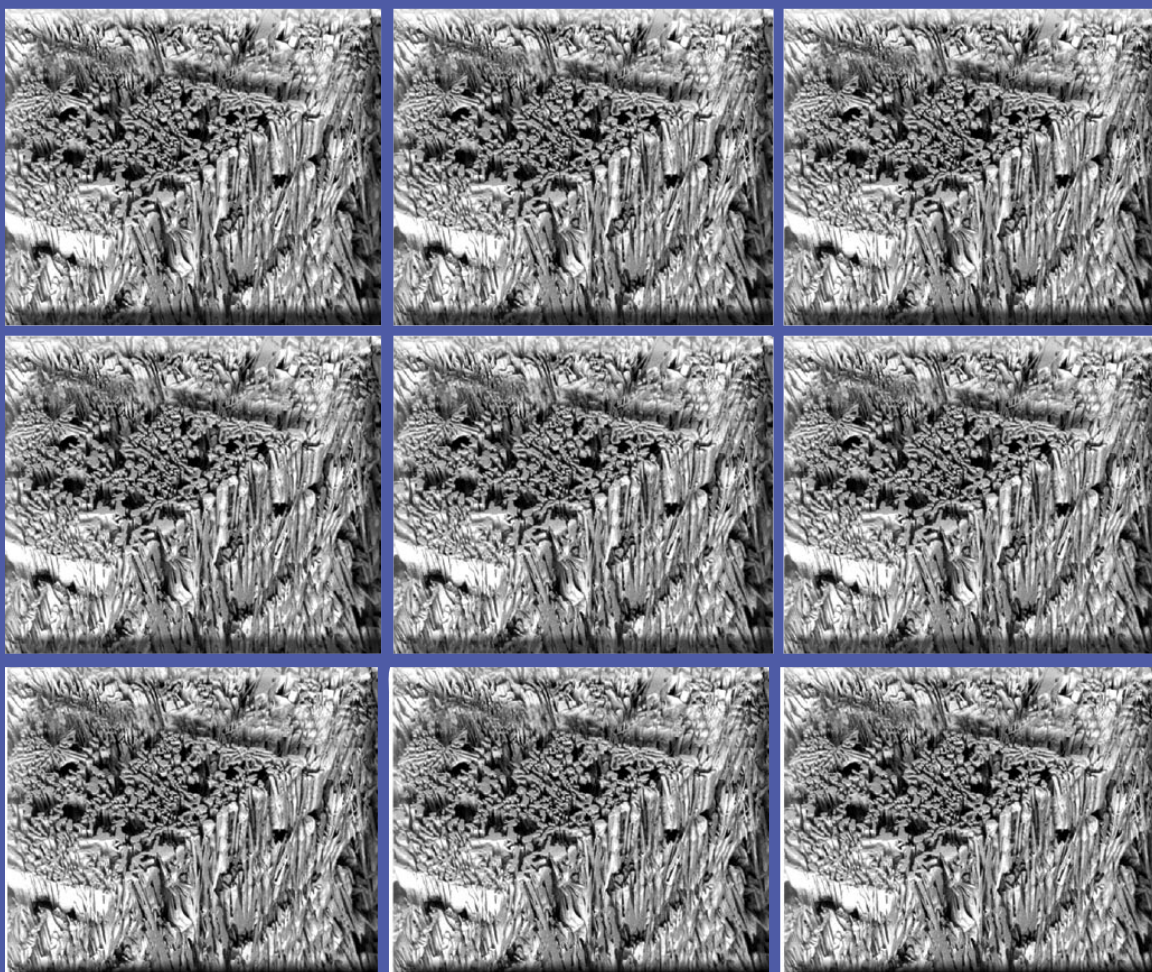
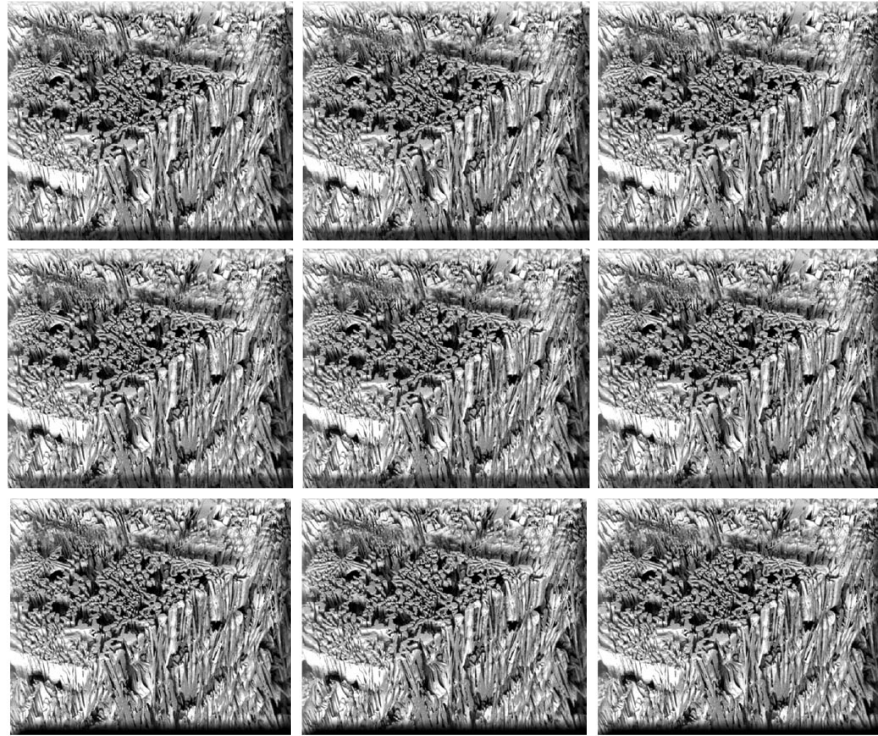




# Abrasion-Resistant Cast Iron Handbook

George Laird  
Richard Gundlach  
Klaus RÖhrig



# **Abrasion-Resistant Cast Iron Handbook**

George Laird  
Richard Gundlach  
Klaus Röhrig

**Edited by  
Karen E. Frink**





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This edition of “Abrasion-Resistant Cast Iron Handbook” has had minimal updates to the front matter. The rest of the book is true to the original print version.

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# Foreword

## Historical Review of Abrasion Resistant Cast Irons

*A special contribution by John Dodd*

This review was conceived as a means of bringing together technical data on the production and properties of alloyed white irons, which has been derived from many years of experience, a vast amount of research and many in-plant investigations by more than a generation of metallurgists. Much of this information has been considered as proprietary by foundrymen who have pioneered improvements in the technology and, obviously, wished to preserve their marketing advantage.

The AFS Special Irons Committee 5-D, which has sponsored this publication, played a major role in the collection and dissemination of sound technical information on the highly alloyed irons since its inception in 1976. The committee also served as a forum for metallurgists and others with a particular interest in this area of abrasion-resistant castings, as well as provided the opportunity to socialize with kindred spirits. Norman Lillybeck (now retired, then with John Deere) deserves recognition as the father of the Special Irons Committee. He felt that the high-alloy irons were not getting the attention they deserved by the AFS Gray Iron Research Committee 5-H and organized a long-delayed vote to create the Special Irons Committee.

Abrasion-resistant castings are tested more severely than many other castings, since much of the casting is worn away in service. Differences in performance are easily noted, and one supplier can be compared with another. Premature failure by fracture or excessive wear can readily be assessed and related to the initial cost. Over the years, three factors have been the driving force for continued technical innovation: (1) improving resistance to fracture; (2) increasing resistance to abrasive wear; and (3) achieving these goals at an acceptable cost.

The high-alloy white irons currently being produced by capable foundries are now working under impact conditions that would have been unthinkable some 30 years ago. Reduced wear rates have allowed substantial reductions in maintenance costs in crushing and grinding operations, which have more than paid for any of the extra alloy costs.

There are many pitfalls for foundries just beginning the production of these vastly improved high-alloy white irons ... also for long established foundries in which the

loss of experienced personnel has downgraded hard-earned expertise.

The following summary of the salient features in the development of the alloy irons is perhaps the easiest way to understand these irons and the potential pitfalls, which are seldom emphasized in the published literature. This approach also shows how the alloys were developed in a continuing process, over the years. Even long-established alloy irons, like Ni-Hard, can be improved, as the late Gordon Cox demonstrated some 50 years after its introduction. Improvements in properties are likely to continue; we can expect modifications to the carbide structure to improve toughness and the introduction of even harder carbides to increase abrasion resistance.

The role of major alloy suppliers in the development of alloy irons cannot be overemphasized. The Electro Metallurgical Company in Niagara Falls, New York supported the development of the high-chromium irons for some 40 years from 1918, and the Climax Molybdenum Company (headquartered in Greenwich, Connecticut) continued this support for a further 25 years from 1960. The International Nickel Company (or INCO) in New York provided strong marketing and development support, worldwide, for the Ni-Hard irons from 1928 for about 50 years. Eventually, the need to compete against new low-cost sources of alloys did not allow a margin on the price of alloys for the continued support of the foundry industry by the major alloy suppliers.

## Development of Abrasion-Resistant Alloy Irons

Like many other technologies, the development of the abrasion-resistant white irons has occurred in clearly defined phases. The aims of higher abrasion resistance and better toughness at an acceptable cost have been the guidelines for development of white irons.

### *Low-Alloy White Irons*

Before the invention of the high-alloy white irons, low-alloy chilled or pearlitic white iron was used for a vast number of applications—from plough shares to rolls in rolling mills—to resist wear and abrasion. There seems to have been no connection between this huge low-alloy iron industry usage and the development of the high-alloy irons. The Electro Metallurgical Company, in whose research laboratories the remarkable properties of the first high-alloy chromium irons were discovered, was, however, a supplier to the low-alloy iron industry. It also used many low-alloy iron components in the crushing

of ferroalloys and recognized the value of its high-chromium alloy iron discovery.

### ***High-Chromium Alloy White Irons***

The high-chromium white irons were discovered in the course of research around 1915. The discovery was made by Frederick Becket, an English metallurgist in the Niagara Falls Research Laboratories of the Electro Metallurgical Company. His objective was to develop new tool steels, based on chromium. This work was intended to offset possible shortages of critical alloys in the manufacture of tool steels during the first-world war, and was not aimed at improving the low-alloy white irons. Becket would seem to have invented the very first highly alloyed iron and, in 1916, described its characteristics and applications in U.S. Patent 1,245,552. High-carbon high-chromium alloys produced in this investigation were found to be less brittle than the unalloyed white irons and were extremely hard; they also had excellent heat and corrosion resistance.

The development and promotion of these 25–30% chromium irons were a major breakthrough in cast iron technology. It seems likely that this work stimulated the research program mounted by INCO a few years later, to find applications for nickel in abrasion-resistant cast irons.

### ***Ni-Hard AR Cast Iron***

In the mid 1920s, research initiated by INCO to increase the market for nickel led to the second important new abrasion-resistant white cast iron: Ni-Hard. Paul Merica, James Vanick and Thomas Wickenden of INCO discovered that a combination of nickel and chromium added to a cast iron would produce a structure of martensite and austenite with a high resistance to abrasion and wear in the as-cast condition. This alloy was then promoted very strongly, worldwide, for more than half a century under the “INCO” trademark of Ni-Hard. Unlike the chromium irons that required electric melting, Ni-Hard could be produced in the cupola furnace. This class of iron had relatively good castability and became, for many years, the most widely produced abrasion-resistant cast iron in the world.

### ***Further Development of High-Chromium Irons***

From 1920 onward, the Electro Metallurgical Co. (now a unit of Union Carbide and Carbon Corp. in New York City) used a range of 15–35% chromium, 1.5–3.0% carbon alloys in their own works, to reduce maintenance costs in crushing and grinding operations and in high-temperature applications. In addition, they promoted these new alloys, worldwide, to increase their market for ferrochromium.

The high-chromium irons could not be melted easily in the cupola furnace, due to the high affinity of chromium for carbon. Thus, their production was largely restricted to foundries with electric melting furnaces, and this undoubtedly hindered their market penetration. As electric melting became more common, large tonnages of the high-chromium alloys

were produced and used in the as-cast condition, i.e., without any heat treatment. They were found to be less brittle than the Ni-Hard irons and were often more resistant to abrasion. The higher-chromium irons also proved to be particularly useful in pumps, where a combination of corrosion resistance and abrasion resistance is necessary.

Large tonnages of the high-chromium irons were produced around the world, particularly in the form of small- to medium-sized castings (although quite heavy steel mill rolls were made in the 1930s). Thick-section high-chromium iron castings, however, did not always seem to have the uniform high abrasion resistance for which the smaller castings were noted. Ni-Hard castings, on the other hand, when made to the correct composition, always had an excellent predictable life in both light and heavy castings.

### ***Inconsistent Performance of Early High-Chromium Alloy Irons***

Careful records were kept by the Electro Metallurgical Company of the many trials and long-term use of the chromium irons in their production plants. Over the next 20 years, numerous investigations could not determine why irons of the same or very similar chemical composition could differ quite markedly in service life. Small and light castings, however, always had excellent abrasion resistance. Over a decade later in 1955, Bradley & Foster, Ltd., a major foundry in England, was forced to accept the fact that, even with excellent control of chemistry within the accepted tolerance limits, their customers' complaints of variable performance were indeed justified. This led to a systematic investigation by them in 1962 of the effect of various alloy additions (Ni, Mn, Mo, Si, Al and Ti) to a wide range of chromium and carbon compositions.

Small specimens from 45 compositions selected were melted, simultaneously, in a tool steel heat-treatment furnace and then cooled to room temperature at various rates. These rates were chosen to simulate the commercial cooling rates of thin- and thick-section castings. The microstructure, hardness and magnetic properties were examined to establish the effect of the alloy additions and the cooling rate. From this investigation, composition limits were selected to ensure that both thin- and thick-section castings could be cast, free of pearlite. Pearlite had been found earlier in some heavy castings and was considered to be the principal factor in reducing service life. This investigation was necessary, because no published information could be found, at that time, as to the effect of alloys on the structure of slowly-cooled high-chromium high-carbon alloys, in either the as-cast or heat-treated condition.

The use of a hollow core was developed to obtain samples from the center of heavy castings, to ensure (by metallography) that they were free of ferrite and pearlite. The structure of all small non-heat-treated castings was checked routinely with a portable magnetic measuring device to ensure that they were austenitic. These investigations and procedures ended the pattern of inconsistent abrasion resistance in castings in

the as-cast condition from Bradley & Foster. The investigations and procedures also led to the use of suitable alloy additions to suppress the undesirable formation of pearlite in heat-treated castings.

Many foundries, at that time, melted both Ni-Hard and high-chromium irons. The production of pearlite-free high-chromium irons, in thick sections, could occur by chance, if enough nickel was picked up from a preceding Ni-Hard melt. As little as 0.3% nickel could ensure a fully austenitic matrix structure in castings containing some 28% chromium, in sections up to 2 in. (5.1 cm) thick. This powerful effect of nickel was not known at that time. Thus, pearlite-free heavy sections were being produced successfully, only if the chromium iron melt followed a melt of Ni-Hard.

In the late 1950s, variable abrasion resistance of heavy castings produced with the Climax 15% chromium, 3% molybdenum iron, were observed by T.E. Norman at the Climax mine in Colorado. This led to the development of the Climax 20% Cr, 2% Mo, 0.5% Ni alloy, to ensure freedom from pearlite, and to ensure optimum abrasion resistance. This alloy was soon modified to contain 1% copper instead of nickel, to avoid a conflict with Abex Foundry's Paraboly 20% Cr, 2% Mo, 0.5% Ni alloy iron patent, published later.

This patented range of composition was developed by H.S. Avery of Abex, who used a wet-sand-abrasion testing machine and wear test results to overcome and understand the variable abrasion resistance of heavy high-chromium iron castings produced there. Avery was able to demonstrate the influence of alloys, such as nickel and molybdenum, on the wear resistance of irons with chromium content of between 12 and 26% and carbon content from 2.5 to 3.4%. Both overalloying and underalloying reduced abrasion resistance of heat-treated samples. Both the Climax 20% Cr, 2% Mo, 1% Ni or Cu alloys and the Abex-Paraboly suppressed the formation of ferrite and pearlite in large slowly-cooled castings and gave dramatically improved abrasion resistance. In hindsight, the change from nickel to copper, by Norman, had the unexpected benefits of a higher hardness in the heat-treated alloy, a cheaper alloy and less problems with retained austenite.

Later investigations, published by R.B. Gundlach, then with Climax Molybdenum, using a specially developed pin abrasive wear test, allowed other foundries to correct their techniques of producing as-cast high-chromium irons to avoid the formation of soft pearlitic or ferritic structures.

#### ***Retained Austenite:***

##### ***A Major Factor in Premature Failure***

In many applications of both Ni-Hard and high-chromium iron castings, chipping or spalling around bolt holes and at exposed edges of mill liners were observed in applications where high-hardness heat-treated grinding balls impacted on either fully austenitic or largely austenitic castings. The sever-

ity of this spalling was always greater with Ni-Hard than with chromium iron, in the same application.

Still in the 1960s, R.H.T. Dixon, an engineer with Mond Nickel Co. in Birmingham England (a Div. of INCO), investigated this spalling problem, since it had resulted in many catastrophic failures of Ni-Hards. The problem was explored with the aid of a drop-ball impact machine. Increasing the hardness, by reducing the austenite content of Ni-Hard grinding balls by heat treatment, increased resistance to spalling in this test. Unfortunately, the improved impact resistance obtainable with alloys of the Ni-Hard type by heat treatment was not sufficient to ensure freedom from spalling in all types of cement mills. As a result, as-cast, largely austenitic, high-chromium irons, which were much more resistant to spalling in the as-cast condition, rapidly replaced the heat-treated Ni-Hard used for balls and liners in many applications.

For some two or three years, austenitic high-chromium iron liners and balls worked well together. Then, the introduction of extremely hard (in excess of 800 HV) martensitic high-chromium iron grinding balls resulted in a whole new spate of premature failures due to spalling and breaking. Based on the earlier experience with Ni-Hard failures, heat-treatment was seen as a way to reduce the austenite content of the high-chromium irons, to improve their resistance to spalling. However, heat treatments that appeared to produce a martensitic matrix did not eliminate spalling and, furthermore, could make it worse.

Based on magnetic and hardness measurements, incomplete transformation to martensite was thought to be the problem. Attempts to further reduce the retained austenite, by tempering at a high temperature, only resulted in a large reduction in hardness.

A major contribution to the metallurgy of the high-chromium alloys was made by the Belgium Foundry Societe Anonyme Foundries Magotteaux. In a 1971 patent, they disclosed a method to optimize both the wear resistance and toughness of the high-chromium alloys. This foundry was the first in the world to measure retained austenite, by x-ray diffraction, as a quality control in the production of high-chromium irons. The castings produced by the technique described in their patent were martensitic, extremely hard, free of pearlite and had a low retained austenite content for maximum resistance to severe repeated impacts.

The patent specification clearly states that, if the retained austenite is too high (i.e., above 3%), heavy, repeated impacts will generate high residual stresses, which can cause premature destruction of the castings by breaking, chipping or spalling. Certainly, the "Maxicrom" high-chromium iron, developed by Victor Joiret at Magotteaux, containing some 17% Cr and 0.8% Mo, with a hardness in excess of 60 HRC, set completely new standards for resistance to both abrasion and spalling, when subjected to severe, repeated impacts, as are grinding balls and liners in milling applications.

Contrary to most other hard materials, the high-chromium irons are most resistant to spalling and premature failure at the highest levels of hardness. Unfortunately, very little of the European research was published, and attempts in the U.S.A. to utilize the high-chromium irons for the extremely large castings needed in the mining industry and for the manufacture of steel mill rolls ran into formidable problems in the late 1960s.

The published technology necessary to overcome these problems was based, to a large extent, on the high-chromium molybdenum iron research undertaken by the Climax Molybdenum Company, and a continuous campaign for the development of production techniques with cooperating foundries. This was a situation exactly analogous to the Ni-Hard research and development program mounted by the International Nickel Company, decades earlier.

### ***Control of Matrix Structure***

Once it had been established in the late 1930s that the abrasion and spalling resistance of the high-chromium irons depended on both the very hard chromium carbides in this type of iron and the matrix structure in which the carbides are imbedded, considerable research efforts at Union Carbide's lab in Niagara Falls were concentrated on control of the matrix structure. Early studies there by P. Burgess found that the properties of high-chromium irons were not significantly changed by conventional heat-treatment processes. As a result of this study, only as-cast austenitic irons were promoted, for many years. Burgess used techniques that were usually used for hardening steels. These techniques were later found to be unsuitable for rapidly cooled austenitic high-chromium alloys. To control the matrix structure, high-chromium alloys must be destabilized by soaking at temperatures in excess of 954C (1750F) for at least three hours, before cooling, to precipitate excess carbon from the austenite and obtain martensitic transformation on cooling.

More sophisticated studies were made, by F. Maratray and R. Usseglio-Nanot of the Climax Molybdenum Co. in France, on the effect of alloys on the transformation of austenite after destabilization. They investigated the effect of chromium and molybdenum in a series of 42 alloys. Later, W.W. Cias extended the studies to include the effects of nickel, manganese and copper. Their results were published by Climax in the form of continuous cooling transformation (CCT) diagrams. These diagrams were of great value in estimating the alloy content necessary to obtain the desired heat-treated structure, free of pearlite, at various cooling rates, appropriate to either individual castings or furnace loads of many tons, without the formation of high-cooling stresses.

Later, a mathematical model was derived by J. Dodd and J.L. Parks at Climax by regression analysis from this CCT data, which allowed a more accurate calculation of the necessary alloy content to suppress pearlite. This has proved invaluable in developing less drastic air-cooling rates to obtain high hardness with low residual stresses.

### ***Further Development of the Ni-Hard Irons***

Research initiated in the mid 1920s by the International Nickel Company led to the discovery of a combination of nickel and chromium additions to a cast iron that would produce a structure of martensite and austenite with a high resistance to abrasion and wear in the as-cast condition.

Unlike the early high-chromium irons, Ni-Hard could be cast in almost any thickness and, with the recommended alloy content, it would be free of pearlite and have excellent abrasion resistance. It is not surprising that for many years Ni-Hard was the first choice for most abrasion resistant applications. The spalling problems described earlier led to a wider use of heat treatment to increase the resistance to spalling, and this was found to also improve the resistance to abrasion.

### ***Ni-Hard 4***

In the late 1940s, research by INCO, to develop new and improved abrasion-resistant irons containing nickel, led to a patent and field tests of a cast iron designed to have a eutectic structure. The improved properties were not considered good enough, at that time, to justify the higher cost, and no major marketing effort was mounted.

Research and field trials by Bradley & Foster in England in the late 1950s, to overcome the spalling failures being experienced with austenitic high-chromium iron castings and the established grades of Ni-Hard, led to the commercial production of a new alloy iron within the very broad scope of the eutectic Ni-Hard patent. This new alloy, selected as a result of those research and field trials, proved to have much better spalling resistance under conditions of repeated severe impacts than did the original Ni-Hard grades 1 and 2 and had excellent abrasion resistance, even when cast in very thick sections.

This led to a renewed interest by INCO in this eutectic alloy iron. A brochure was published by them in England to promote "eutectic Ni-Hard," showing pump castings produced as a result of the development work by the Bradley & Foster, Ltd. Research Department.

J. Rickhard of the Mond Nickel Co. in Birmingham, England, an affiliated company of INCO, began a new research program to optimize the composition and properties of this type of high-alloy iron and, in 1963, INCO designated the "eutectic Ni-Hard" composition (essentially the same as that developed by Bradley & Foster) Ni-Hard 4. The alloy was then promoted, worldwide, as a means of combating the nickel-free high-chromium irons that had replaced the older Ni-Hard irons in many high-impact applications.

Renewed trials of this alloy by INCO in the U.S.A. established that it had good corrosion resistance and excellent abrasion resistance and was considered to be particularly suitable for the production of large slurry pumps. Georgia Iron Works in Augusta, Georgia produced the alloy in large tonnages for this purpose and developed proprietary hardening techniques for increasing abrasion resistance by both cryogenic and heat treatment processes.

### ***Premium Ni-Hard***

Continuous experimental work by Cox (at that time, with Sheepbridge Foundry in England), to support the use of nickel in abrasion-resistant irons, led to the development of higher hardness Ni-Hard irons, with a lower austenite content in the matrix. This silicon-modified Ni-Hard has been marketed under the trade name of Premium Ni-Hard. (Combustion Engineering also marketed a similar product called Crown 700.) The high hardness level is achieved without a high-temperature heat treatment by the use of a higher silicon content.

The carbides in this modified iron are of the same type as in Ni-Hard 1 and 2 and are softer than the carbides in the higher-chromium irons. The alloy has good castability and has proved useful in the production of relatively thin liner plates used in some types of coal pulverizers. Larger castings, such as the crusher rolls for the same type of application, have encountered spalling problems due to the still relatively high levels of retained austenite.

Dr. Gordon Cox deserves a special mention in connection with the development of the abrasion-resistant alloy irons. His excellent papers on the nickel-chromium irons reflect his thorough understanding of both alloy systems and show that there is still room for innovative thinking.

### **Closing Comment**

In looking back on my career in the development of the high-alloy irons, I have to thank Dr. J.E. Hurst, one of the pioneers of cast iron metallurgy and a former managing director of Bradley and Foster, for giving me the task of developing a less brittle alloy iron than Ni-Hard.

Written by:

John Dodd

Dodd & Associates, Arvada, Colorado

*Chairman of the AFS Special Irons Committee (5-D),  
from its inauguration in 1976 until 1988.*

# Preface

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This handbook project started out as a simple dream of compiling current literature and organizing this data into a readable format. Along the way, with the help of my co-authors Rick Gundlach and Klaus Röhrig, the handbook has grown and matured into much more. It has been a fascinating journey and I owe a debt of gratitude to my two friends, Mr. Jack Rice and Mr. Raymond Schleiden, for helping me complete this journey. The AFS 5-D Special Irons Committee also has earned a special mention for their constant support and kind reminders to finish, finish, and finish. Without the AFS 5-D Committee this handbook would never have been completed. Every member, past, present, and future deserves a special thanks from every reader.

I view the content of this handbook as a work in progress. We have tried at every bend of the road to ensure that all facts are relevant and practical. Every chapter was read by members of the Foundry Steering Committee for technical veracity. Even with this said, we hope that you will report omissions

and corrections to the authors. With your help, we can make the second edition of this handbook even better.

The American Foundry Society deserves a special recognition for assigning Ms. Karen Frink to work with me on the editing of this handbook. She was outstanding. She thoroughly read every word, every sentence, and every paragraph and made numerous constructive contributions. It was a sincere pleasure working with her.

Authors rarely make a living from writing technical handbook treatises. As such, I had to sneak in snippets of time to write, organize, create and ponder within my schedule. My wife, Ann, was and is wonderful for supporting me in finding time to write. As is most fitting, she is lovingly acknowledged.

Lastly, I thank the American Foundry Society for finding room to publish this handbook and supporting its effort through the AFS 5-D Committee and partial funding for the authors.

George Laird

# About the Authors

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**George Laird**

George Laird was born in 1957, the last of six sons. He received his bachelor, master and doctorate degrees in mechanical engineering from Oregon State University in Corvallis, Oregon. His doctorate was granted in 1993.

A highlight of his professional career was working at the former U.S. Bureau of Mines in Albany, Oregon, where he was allowed the freedom to explore the complex metallurgy and structure of abrasion resistant irons. Along this professional path he was graced with several wonderful tutors: Dr. Robert Blickensderfer, Dr. Norman MacMillan and Dr. Jeffrey Hawk.

Subsequent to this career as a scientist, Dr. Laird started up his own consulting company, Predictive Engineering, Inc., in 1996. His company specializes in mechanical and thermal analysis of engineered systems using computer-aided modeling tools.

George and his wife live in Corvallis, Oregon.



**Richard B. Gundlach**

Richard B. Gundlach was born in 1945 in Buffalo, New York. He holds BS and MS degrees in Metallurgical Engineering from the University of Michigan in Ann Arbor, Michigan.

Rick is president and a senior metallurgical engineer at Climax Research Services (CRS), a materials consulting and testing laboratory located in Wixom, Michigan.

Prior to cofounding CRS in 1987, Rick was a metallurgical engineer and research supervisor for 18 years at Climax Molybdenum Company (AMAX) in Ann Arbor, Michigan.

He has worked in the field of casting metallurgy and has published and consulted in the areas of cast aluminum; gray, ductile and white irons; austempered ductile irons and steels; thermal fatigue; and abrasion resistant materials. He has authored and coauthored more than 40 technical papers.

Rick, his wife and family live in Farmington, Michigan.



**Klaus K. Röhrig**

Klaus K. Röhrig was born in 1935 in Berlin. He graduated in 1961 from Technical University of Berlin and holds a Doctor of Engineering degree in metallurgy.

After four years as a research assistant at the Foundry Institute of Berlin University, he became a research manager of the Cast Iron Group of Rheinstahl (now Thyssen Krupp). In 1969, he joined Climax Molybdenum Company as market development manager, Cast Iron and Steel, Europe.

From 1989-2000, he was the technical editor of the German foundry magazine, *Geißerei-Praxis*, technical consultant to the foundry industry and consultant to the Nickel Development Institute (NiDI).

Klaus and his wife live in Erkrath, Germany (near Düsseldorf).

# Acknowledgments

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## **AFS Special Irons Committee 5-D**

The American Foundry Society 5-D Committee has been the driving force behind the completion of this handbook. If it were not for this committee, this handbook would never have been published. Special and individual thanks are extended to every current and past member of this committee.

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## **Contributors**

The following companies provided guidance and partial funding for this handbook. The contacts given for each company or organization were lead reviewers within that entity for each chapter of this book. Their encouragement and constructive comments were sincerely appreciated by the authors, during the development of this handbook.

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## **International Molybdenum Association**

The International Molybdenum Association (IMOA) was founded in 1989 and has become the focal point of promotional, statistical and technical activities for the worldwide Mo industry. Membership is broad based and includes producers, consumers, converters, traders and assayers. IMOA's secretariat is based in London.

IMOA's main activities currently include:

- promoting Mo as a material with superior properties and performance in a wide variety of metallurgical, chemical and other product applications;
- monitoring Mo in relation to health, safety and environmental issues. With the increasing amount of legislation in many countries relating to the use and disposal of metals and metal-bearing materials, IMOA provides a centralized service, including research studies on those issues that may affect the Mo industry;
- collecting the industry's most comprehensive historical statistics on world supply and demand of Mo products, which are distributed to all IMOA members on a regular basis;
- organizing meetings and promotional conferences beneficial to the Mo industry;
- preparing worldwide industry guidelines to improve consistency and quality in sampling and assaying procedures for Mo concentrates.

### ***List of IMOA Publications***

- Molybdenum
- Practical Guidelines for Fabrication of Duplex Stainless Steels
- The Evolution of High Performance Stainless Steels
- Applications of Mo Metal & Its Alloys
- Molybdenum in Relation to Health, Safety & the Environment
- Applications of Molybdenum in Environmental & Human Health Protection
- Procedures for the Weighing & Sampling of Ferromolybdenum
- Procedures for the Weighing & Sampling of Molybdenite Concentrates
- Procedures for the Weighing & Sampling of Technical Grade Molybdenum Oxide
- Directory of Members (including products and / or activities)

They can be reached at:

E-Mail: [info@imoa.com](mailto:info@imoa.com)

Web Site: [www.imoa.info](http://www.imoa.info)

### ***Use of Mo in Cast Iron Production***

Molybdenum increases the strength and hardness of cast irons by depressing the pearlite transformation temperature. It also increases elevated temperature strength and creep resistance. High-chromium irons containing 2–3% Mo exhibit significantly greater impact toughness than do Mo-free grades and are ideal for severe abrasive conditions like those encountered in mining, milling, crushing, etc. These cast irons have acceptable properties, as cast. This eliminates the need for a costly heat treatment and makes them a cost effective alternative to other grinding materials. Reduced levels of austenite formers, such as nickel and manganese, also minimize the retention of low-temperature austenite...a potential cause of premature failures.

There has been growing interest in the use of high silicon-molybdenum ductile irons with up to 4% Si and 1% Mo. Their good strength up to 600C makes them a viable and cost-effective replacement for more highly alloyed irons and steels in elevated temperature applications, such as turbocharger housings, engine exhaust manifolds and furnace components. The austempered nodular irons develop a unique microstructure, capable of strengths in excess of 1000 MPa (145 ksi) with good impact toughness. Their exceptional properties are ideal for critical applications, such as the large gears and crankshafts required for power generation, ship propulsion and large mining equipment.

### **Nickel Development Institute (NiDI)**

The Nickel Development Institute (NiDI) was an international, nonprofit organization that provide accurate, complete and up-to-date information on nickel, nickel alloys and nickel-containing stainless steels. This information was used to solve materials engineering challenges, including procurement, fabrication, troubleshooting and life-cycle management.

NiDI was incorporated into the Nickel Institute ([www.nickelinstitute.org](http://www.nickelinstitute.org)), and no longer exists. We suggest contacting the Nickel Institute for more information.

# Chapter 1

## Introduction to AR Cast Irons

---

Abrasion-resistant (AR) cast irons have microstructures and properties every bit as complex as the most exotic, advanced engineered alloy. These materials have been in constant use for centuries and hundreds of articles have been written about their metallurgical characteristics and wear behavior. The literature paints a fascinating picture of these irons in that seemingly small changes in melt conditions, pouring *superheats* (see Glossary), alloy chemistry, cooling rate, or heat treatment can lead to vastly different wear properties. Whether it is a foundryman making AR alloys or an end-user trying to wade through alloy specifications, the underlying science of why these alloys behave the way they do is quite intriguing.

It is difficult to answer all the questions about making and using AR cast irons that contain hard, wear-resistant carbides. Nonetheless, this effort offers a look at the AR cast iron literature and laboratory and field observations. The focus of this work is to create a usable handbook on how to effectively understand the science and technology of AR cast irons.

In this handbook, *AR cast irons* refer to a class of ferrous-based materials that contain hard, wear-resistant carbides and are processed via typical foundry routes. These materials are characterized by their substantial carbide volume fraction (CVF), having from 900 to 3000 diamond pyramid hardness (DPH). The majority of these carbides are formed during solidification, i.e., they precipitate from the liquid during a *eutectic* reaction. (See Glossary for *eutectic*.)

It is these hard carbides that put the wear resistance into AR cast irons. This is both good and bad. Abrasion-resistant irons are well known for their superior resistance to all types of abrasive wear; however, it is these same hard carbides that make these irons prone to low toughness and catastrophic failures upon impact. Consequently, abrasion resistance and toughness tend to be mutually exclusive properties in AR irons.

Abrasion resistance is not a distinct property of a material, like hardness or tensile strength, but depends on the microstructure of the material and the abrasive environment. This coupling of microstructure and environment is rather complex, and ever-so-subtle changes may result in significant changes in the service life of the casting or in the relative performance of different materials. For example, in gouging, cutting or scratching types of abrasion, a penetration by the abrasive is necessary, but not every abrasive contact results in particles being detached from the surface. Elastic and plastic deformation, work-hardening phenomena and the toughness and fatigue resistance of the material play an important role.

There are many good examples concerning the difficulty of correlating abrasion resistance to material properties. In many cases, a common fallback position is that a harder material will be more abrasion resistant. This is correct only to some extent. Rubber or elastomers show good abrasion resistance in certain applications and, in many abrasive applications, soft manganese steel (i.e., containing no eutectic carbides) gives superior performance under certain conditions.

The point of the foregoing discussion is to describe the countless factors that should be considered in the selection of an AR cast iron. A good AR iron will not only have excellent wear performance but also sufficient toughness to avoid catastrophic failure. Microstructure description and how this microstructure determines the wear performance of the alloy will be discussed in depth. This fundamental knowledge should allow one to accurately understand casting and foundry problems relating to these irons and thereby extrapolate the use of new irons to more challenging applications.

### Why Abrasion-Resistant Cast Irons Are Known As “White” Irons

The history of carbide-containing or white iron goes back many centuries. Archeological evidence indicates that the Chinese were using white cast iron for agricultural implements and in various weaponry in the form of war axes and blades as far back as 400 B.C. This metallurgical invention, as with a number of others, was lost through the shifting sands of time until the early eighteenth century.

In R. Elliott’s book, *Eutectic Solidification Processing* (Butterworths, U.K., 1983), the accidental discovery of white iron is attributed to Robert Ransome, an agricultural manufacturer in Ipswich, England. It happened when one of his molds split during casting, spilling hot metal onto the cold foundry floor. While he was cleaning up the spilled iron for remelt, he noticed that the iron appeared “white.” After a little experimentation, he found that the iron was much harder than gray iron (which, as you may have guessed, appears gray, when fractured). Seeing and seizing an opportunity, Ransome constructed plowshare molds, which had an iron backing plate. After casting, the resulting part consisted of a hard, abrasion-resistant layer (white iron) backed up by a stronger, more ductile layer for mechanical ruggedness. His abrasion-resistant plows quickly became the marketplace standard and, in a way, Ransome’s foundry kicked off the modern-day business of producing AR cast irons.

## Classification of AR Cast Irons

Abrasion-resistant cast irons can be classified into five broad groups, based on their alloying and microstructure:

1. Pearlitic ( $\text{FeC}$ )
2. Ni-Hard or Ni-Cr ( $\text{M}_3\text{C}$ )
3. Ni-Hard 4 ( $\text{M}_7\text{C}_3$ )
4. High-Cr ( $\text{M}_7\text{C}_3$ )
5. Specialty ( $\text{M}_x\text{C}$ )

This simple classification will be the foundation of our approach to understanding these irons. Later chapters will expound on this classification and provide a broad working knowledge about the microstructure of these irons and how they perform in a variety of foundry and tribological environments.

### Pearlitic or Chilled AR Cast Irons

Pearlitic white cast iron is the oldest type of cast iron, produced especially for resistance against abrasion. It has been in production for centuries. This family of irons has many names, and foundrymen refer to these irons as *indefinite chill iron*, *mottled white iron*, *chill iron*, or simply *white iron*. In all cases, chilling the iron against a cold surface forms the abrasion-resistant carbide network. The formation of carbides as opposed to graphite is brought about by the ambivalent nature of the iron-carbon ( $\text{Fe-C}$ ) system, where the carbon can be precipitated either as graphite (stable or equilibrium phase) or as iron carbide (metastable or nonequilibrium phase). The former results in the well-known family of gray, nodular, compacted graphite or malleable irons and the latter in low-alloy, AR irons or white irons.

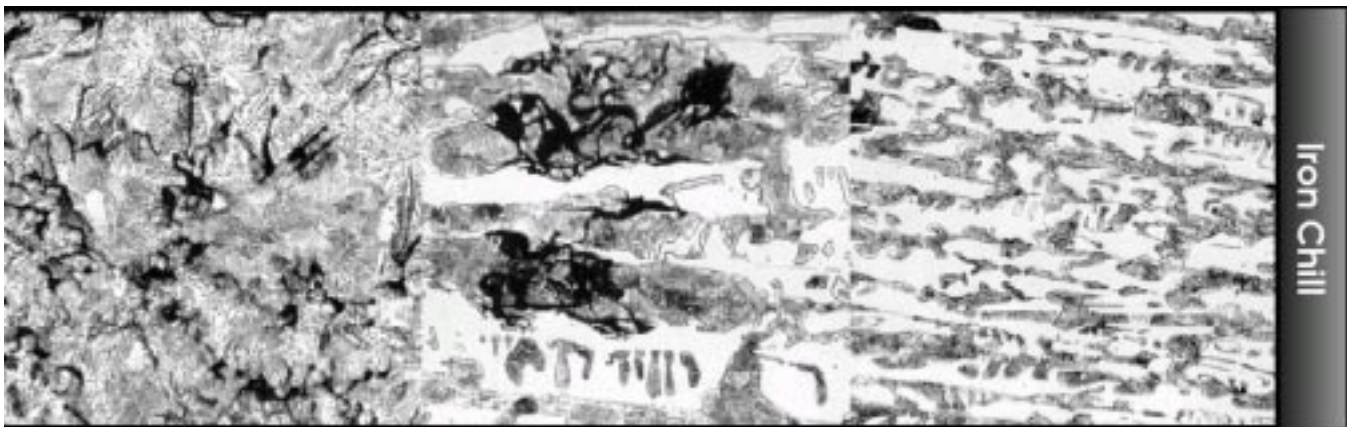
The metastable reaction in the  $\text{Fe-C}$  system occurs due to rapid cooling and typically requires that the liquid iron be solidified against a cold, conducting surface or “chilled.” As the iron rapidly cools, it undergoes a eutectic solidification reaction, forming austenite and iron carbide. The iron carbide is commonly known as *cementite* ( $\text{Fe}_3\text{C}$ ) (see Glossary) and is formed by three iron atoms and one carbon atom. Another name for this eutectic mixture of austenite and cementite is *ledeburite*.

As the solidification front moves away from the chill surface, the solidification rate slows and the iron tends to solidify via the more stable, equilibrium route. Away from the surface, the iron will form a eutectic mixture of austenite and graphite. This transition from metastable to stable is not abrupt but gradual. Consequently, the microstructure of a thick-sectioned, chilled iron will have an abrasion-resistant carbide surface, a transition zone of carbide and graphite and a fully graphitic structure. This microstructural transition is shown in Fig. 1.1

There are several metallurgical procedures that can be employed to vary the thickness of the abrasion-resistant surface layer, besides the use of a bigger or smaller chill. For example, if a thinner carbide shell is desired, the silicon content of the iron can be increased, thereby favoring the formation of graphite over that of carbide. A thicker carbide layer can be produced by adding small amounts of chromium or manganese or by adding elements that inhibit graphite nucleation or graphite growth (i.e. sulfur, bismuth or tellurium).

This type of mixed microstructure—abrasion-resistant on the surface and graphitic underneath—is regarded as beneficial in many cases, because the gray iron (graphite eutectic) provides a backing to the hard, white surface (carbide eutectic). The more ductile backing layer of gray iron enhances the fracture resistance of the component and makes it easy to machine on one side. One disadvantage in the use of these irons is that the abrasion-resistant layer is of finite thickness and can be worn through, exposing the soft, nonabrasion-resistant sub-layer.

After solidification, due to the iron’s low alloy content, the austenite will almost always transform to *pearlite* (see Glossary). Only in rare cases—when the section size is very thin and the cooling rate extremely quick—will the austenite be retained at room temperature. Since the final microstructure is typically a combination of pearlite (which in itself is composed of two distinct phases, i.e., ferrite and cementite ( $\text{Fe}_3\text{C}$ )) and eutectic iron carbide, this alloy is commonly known as pearlitic white cast iron. Pearlitic white cast irons are only standardized in the U.K. and in the forthcoming EU-Standard



**Fig. 1.1.** Pearlitic AR cast iron. The carbide structure forms only near the chill. Further away from the chill, the iron reverts back to its equilibrium graphite structure. 100X.

(see Appendix E for details). Their composition range is very wide and is given in Table 1.1.

The hardness of pearlitic white cast irons is usually between 350 and 500 Brinell hardness (HB) (see Appendix D for a complete definition of hardness techniques and conversion tables). The hardness can be increased by forming more carbide or by increasing the hardness of the pearlite by refining its laminar structure of ferrite and carbide. However, these irons are not amiable to heat treating above 600C, since the metastable carbide network will easily transform into graphite, rendering the iron soft.

The utility of pearlitic AR cast irons is that by a judicious combination of chemical composition, inexpensive raw materials, chill and mold wash, superheat, pouring temperature, inoculation and section size of the casting, well-defined layers of AR cast iron can be engineered over a ductile substrate. The metallurgical principles are well understood, but their practical production requires a substantial amount of experience and care; or in other words, it is often still regarded as more of an art than engineering.

### Ni-Hard or Ni-Cr AR Cast Irons

Industrially, this class of irons, best known as Ni-Hard, comes from the International Nickel Company (INCO) and refers to a group of four iron compositions (Ni-Hards 1, 2, 3 and 4) that use nickel as the primary alloying element for hardenability. Another important distinction within this group of irons is the use of Cr to drive the formation of metal carbides over that of

graphite. As shown in Table 1.2, it is recommended that the Ni-Hard irons contain at least 1.4% Cr with more typical levels around 2% Cr. The presence of Cr stabilizes the eutectic reaction and favors the formation of carbide structure independent of cooling rate. Hence, AR structures containing eutectic mixtures of austenite and carbide can be obtained in thin and thick section sizes independent of the use of chills. It is still possible to obtain traces of graphite in thicker sections or when higher levels of C and Si are employed. Barring these circumstances, the dominant microstructure of Ni-Hard is one composed of a ferrous matrix surrounded by hard metal carbides.

When Cr (or, to some extent, Mn and Mo) is used to stabilize the carbide phase, the metal carbide is known as  $M_3C$  to reflect the mixed nature of this phase. The  $M_3C$  carbide is visually indistinguishable from the  $Fe_3C$  carbide, formed in pearlitic AR cast irons, and occurs through the same eutectic reaction. Based on Ni-Hard's strong use of both Ni and Cr to obtain their desired wear properties, these materials are also sometimes known as Ni-Cr irons.

Of all the AR irons, Ni-Hard is produced in the largest tonnage for a variety of mineral processing industries. Ni-Hard's low cost is due to its low alloy content, its ability to be cast into a variety of shapes and its high hardness in the as-cast condition. Its high hardness is what clearly separates it from pearlitic AR cast irons. Ni-Hard's high hardness results from the formation of martensite versus pearlite in the as-cast condition. This metallurgical shift is the result of Ni-Hard's high Ni content. The presence of 3–5% Ni allows the virgin

**Table 1.1**  
**Chemical Composition and Hardness of Pearlitic AR Cast Irons**

| Alloying Element | %C      | %Si     | %Cr     | %Ni     | %Cu     | %Mn     | %Mo     | %V      | %P      | %S       | Brinell Hardness, HB |
|------------------|---------|---------|---------|---------|---------|---------|---------|---------|---------|----------|----------------------|
| U.K. Std.        | 2.4–3.9 | 0.4–1.5 | 0.0–2.0 | 0.0–2.0 | 0.0–1.5 | 0.2–1.0 | 0.0–0.5 | 0.0–0.3 | 1.0 max | 0.15 max | 350–500              |

**Table 1.2**  
**Chemical Composition and Hardness of Ni-Hard 1 and Ni-Hard 2 AR Cast Irons**

| Alloying Element          | %C      | %Si     | %Cr     | %Ni <sup>(1)</sup> | %Mn     | %Mo     | Brinell Hardness, HB   |
|---------------------------|---------|---------|---------|--------------------|---------|---------|------------------------|
| Ni-Hard 1<br>ASTM A532-1A | 3.0–3.6 | 0.8 max | 1.4–4.0 | 3.3–5.0            | 2.0 max | 1.0 max | 550 (typical)          |
| Ni-Hard 2<br>ASTM A532 1B | 2.9     | 0.8 max | 1.4–4.0 | 3.3–5.0            | 2.0 max | 1.0 max | 550 (typical)          |
| Ni-Hard 1<br>NiDI         | 3.0–3.6 | 0.3–0.5 | 1.5–2.6 | 3.3–4.8            | 0.3–0.7 | 0.4 max | 500–700 <sup>(2)</sup> |
| Ni-Hard 2<br>NiDI         | 2.9 max | 0.3–0.5 | 1.4–2.4 | 3.3–5.0            | 0.3–0.7 | 0.4 max | 500–700 <sup>(2)</sup> |

(1) It is possible to substitute copper (up to about 1.5%) for nickel; however, Cu is not as powerful as Ni in producing a hard structure.

(2) Hardness varies depending on heat treatment.

ASTM = American Society for Testing and Materials recommendations

NiDI = Nickel Development Institute recommendations

austenite to reach the martensite start temperature unhindered by the formation of pearlite. No transformation reaction is ever perfect and the as-cast Ni-Hard microstructure will contain a mixture of austenite and martensite. If the casting is of variable thickness, then thicker sections of the casting may contain traces of pearlite. From this discussion, it is obvious that it is quite difficult to make predictions about a casting's wear performance, based on initial chemistry, without knowing its dimensions or thermal history.

Table 1.2 presents a brief idea of the general compositions for Ni-Hard, based on the U.S. ASTM specification for Ni-Hard. International specifications for Ni-Hard can be found in Appendix E. Although compositional information is given to the exclusion of other mechanical property data, this exclusion does not imply that it is unimportant. Because chemical information can be readily determined, it is often the favorite tool of many foundrymen who concentrate on alloy composition at the expense of mechanical property data. The mechanical properties of an iron (e.g., hardness, fracture toughness, abrasion resistance, etc.) are as much determined by pouring superheat or heat treatment as they are alloy composition. This point will be demonstrated throughout the course of the book. With this caveat in mind, these initial iron compositions are given solely to guide the reader as to how these irons are classified within the foundry world and not as to their wear resistance or toughness.

Although this branch or group of irons is commonly termed Ni-Hard, a variety of compositions are included under the Ni-Hard name and they have a wide range of mechanical and wear properties. There is no such thing as a typical Ni-Hard. This inevitably leads to problems when foundrymen and end-users specify the iron by its common trade name and ask for a certain level of hardness in the finished casting. The problem with this approach is that it works often enough to create an illusion of an effective method for specifying AR iron castings.

Although hardness is readily understood to reflect the ability of the material to resist deformation and hence provide a measure of its strength and wear characteristics, hardness in Ni-Hard and in all AR irons is really an agglomeration of many separate microstructural components. They all work together to resist mechanical deformation. For example, Fig. 1.2 shows a polished plane of a Ni-Hard casting after being slightly etched with a strong acidic chemical. The chemical etching attacks some *constituents* (see Glossary) more strongly than others and creates a rough or optically contrasting surface. In the case of Ni-Hard (and other ferrous-based alloys), martensite is readily attacked by the etch (dissolves away and creates an optical feature) and appears dark in the picture. The other phase present in the matrix is austenite—an iron phase that is quasi-stable at room temperature—which occurs when alloyed with C or Ni or Cr, etc. A complete definition of the various phases that can form in AR cast irons will be given in Chapter 3 on Heat Treatment.

As shown in Fig. 1.2, several other phases or constituents are also present. If we consider AR irons to be composites, then it is natural to think of the carbide phase as the reinforcing

fiber (or metallurgists might say reinforcing constituent) and the more prevalent ferrous phase as the matrix of the composite. To demonstrate this concept, the ferrous matrix can be etched away to reveal the  $M_3C$  carbide phase. Viewed by *scanning electron microscopy* (SEM) (see Glossary), Fig. 1.3 shows a photomicrograph of the Ni-Hard reinforcing fiber.

When a hardness measurement of a Ni-Hard alloy is taken, it represents the combined hardnesses of the matrix and the carbide phase. One technique to increase the hardness of a Ni-Hard casting is to increase its C and Cr contents. As these alloying elements are increased, the iron forms more carbide and the casting increases in hardness. The downside to this alloy modification is that the casting becomes brittle as the carbide phase increases to a point that it dominates the microstructure of the casting. The end result is that the customer has the specified hardness but is left with a dangerously fragile casting that could prematurely or catastrophically fail. Consequently, when quite broad chemistries (i.e., the generic Ni-Hard specification) and only one level of hardness are specified, a wide range of microstructures having vastly different wear characteristics can be created to satisfy a certain hardness requirement.

The example specification in the previous paragraph illustrates the interplay between alloy chemistry, microstructure and mechanical properties. To create the correct alloy for the application, one should be aware of these interactions and play to their strengths. Throughout the subsequent chapters, an effort will be made to link iron chemistry to solidification behavior, to hardenability, to microstructure. With this understanding, the iron's microstructure can be qualitatively linked to the required hardness or required mechanical properties or required wear requirements of the application.



**Fig. 1.2. Ni-Hard AR iron showing the eutectic carbides and matrix phases on a 2D plane that was cut through the iron. 500X.**

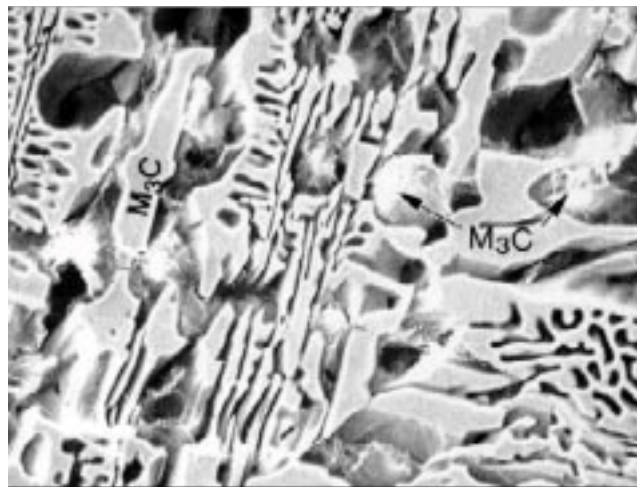
## Ni-Hard 4 AR Cast Irons

With the early success of Ni-Hard, metallurgists were encouraged to develop other variants to this alloy family. One of the drawbacks of the Ni-Hard family was their low toughness under impact conditions. In the early days of alloy development, it wasn't entirely clear why Ni-Hard had such poor toughness. Without going into too much detail, the continuity of the carbide network was found to be critically important in controlling the impact toughness of the alloy. Although the metallurgists of that day were unaware of this microstructure-to-toughness linkage, their new Ni-Hard alloy (Ni-Hard 4) contained a much more fracture-friendly metal carbide, known chemically as  $M_7C_3$ . Consequently, Ni-Hard 4 demonstrated improved toughness and abrasion resistance when compared against the original Ni-Hard.

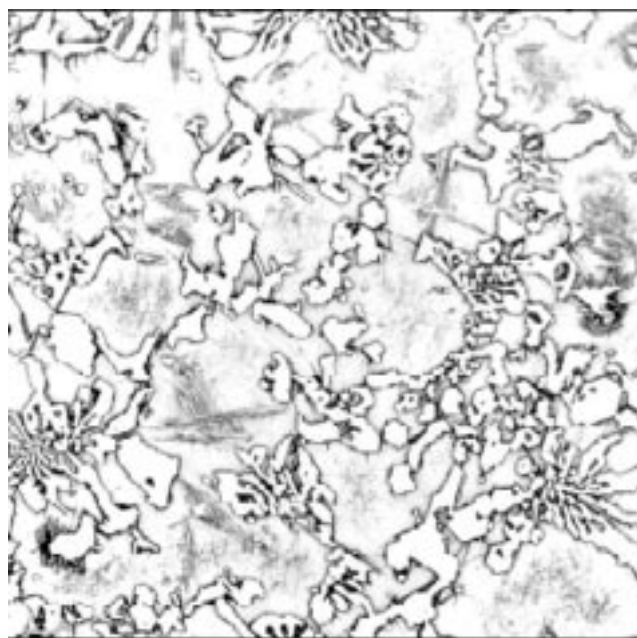
Table 1.3 gives the general specification for Ni-Hard 4. Depending on heat treatment, the hardness can vary significantly. As would be expected, if the toughness and wear resistance show a change over Ni-Hard, then the internal structure or microstructure of the iron must also show a change. Figure 1.4 is a photomicrograph that shows a polished plane from a Ni-Hard 4 casting, where part of the matrix has transformed into martensite (dark, slender and sharp looking features, otherwise known as “acicular” features) as it cooled to room temperature. If one concentrates on only the ferrous constituent, the Ni-Hard 1 or 2 matrix is nearly identical to the Ni-Hard 4 matrix. If we dissolve away the ferrous matrix, Fig. 1.5 reveals the pivotal component to Ni-Hard 4's success—its harder and more discrete  $M_7C_3$  eutectic carbide phase.

Increasing the Cr content of the alloy to a minimum of 7% and the Si content to around 1.8% forms this improved  $M_7C_3$  carbide. Not only is the carbide more discrete, which raises the toughness of the iron, the carbide is also significantly harder than the  $M_3C$  carbide. The downside to Ni-Hard 4 is its higher cost when compared to Ni-Hard.

Whether or not to use Ni-Hard 4 as an AR iron is typically based on a number of factors. For example, it is wonderfully fluid and casts exceedingly well due to its near-eutectic composition and high Si content. Moreover, with careful alloy control, it is possible to obtain as-cast hardnesses in excess of 600 HB and, through the use of various heat treatments, hardnesses of over 700 HB.



**Fig. 1.3.** After chemical attack, the true 3D image of the carbide structure can be seen by using the high depth-of-field capabilities of a scanning electron microscope (SEM). Notice that the carbides are nearly continuous throughout the iron. 1000X.



**Fig. 1.4.** A metallographic section from a Ni-Hard 4 casting. The polished plane shows the characteristic eutectic carbide shapes found in high-alloy AR irons. 500X.

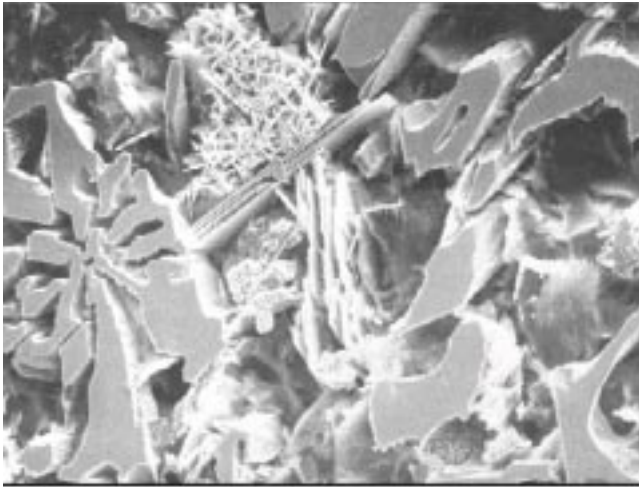
**Table 1.3**  
**Chemical Composition and Hardness of NiHard-4 AR Cast Iron**

| Alloying Element          | %C      | %Si     | %Cr      | %Ni     | %Mn     | %Mo         | Brinell Hardness, HB   |
|---------------------------|---------|---------|----------|---------|---------|-------------|------------------------|
| Ni-Hard 4<br>ASTM A532-ID | 2.5–3.6 | 1.0–2.2 | 7.0–11.0 | 5.0–7.0 | 1.3 max | 1.0 max     | 550 (typical)          |
| Ni-Hard 4<br>NiDI         | 2.6–3.2 | 1.8–2.0 | 8.0–9.0  | 4.5–6.5 | 0.4–0.6 | 0.0–0.4 max | 550–700 <sup>(1)</sup> |

(1) Hardness varies depending on heat treatment.

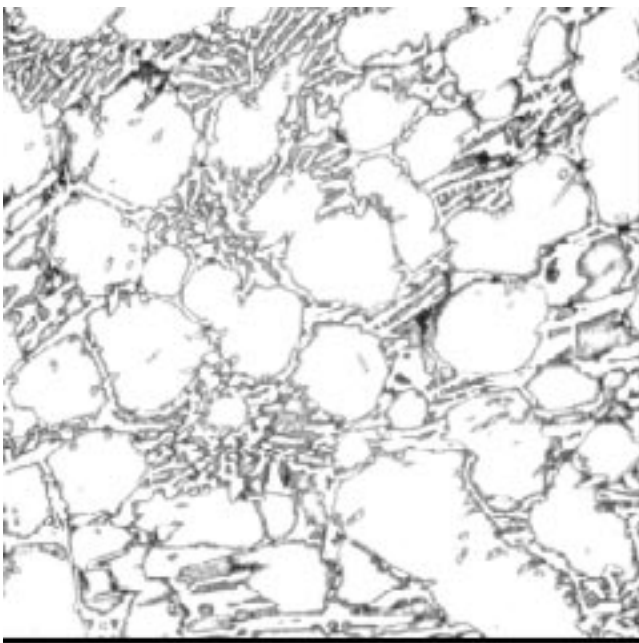
ASTM = American Society for Testing and Materials recommendations

NiDI = Nickel Development Institute recommendations



**Fig. 1.5.** The 3D carbide network of a Ni-Hard 4 iron as seen through SEM. The  $M_7C_3$  carbides are much less continuous than the lower-Cr  $M_3C$  carbides shown in Fig. 1.3. 1200X.

There is an abundance of interesting things occurring within this iron family. For example, if the Si content is not carefully controlled, it is possible to form duplex carbides that have a hard  $M_7C_3$  core surrounded by a soft  $M_3C$  shell. Besides this strange feature, both Ni-Hard 4 and Ni-Hard respond extremely well to cryogenic treatment producing increased hardnesses in the as-cast component of between 80 and 120 HB. All of these phenomena can be explained through direct application of metallurgical principles.



**Fig. 1.6.** A metallographic section from a high-Cr casting. Little contrast is shown between the austenite matrix and the  $M_7C_3$  eutectic carbides. Even when severely attacked with a 2% nital etchant, the Cr-rich austenite matrix survives quite well. As one may suspect, high-Cr irons do quite well in some corrosive environments. 500X.

## High-Cr AR Cast Irons

When toughness, abrasion and, in a few cases, corrosion resistance are critical, the high-Cr family of AR irons is hard to surpass. High-Cr irons have been around since the early 1900s and, historically, were the first true, alloyed AR irons in commercial production. Their mechanical and wear properties are derived from a well-controlled (technologically well-understood) matrix microstructure and their fine distribution of  $M_7C_3$  metal carbides. With proper alloying, extremely thick sections of high-Cr irons can be made with fairly uniform hardness from surface to core. Typically, this high hardenability arises from a combination of Cr, Mo, Ni and Cu alloying elements.

The number of alloy combinations is almost limitless and many foundries enhance their business by promoting one variant or another of the same alloy. As a start, Table 1.4 gives the U.S. ASTM classifications and their possible hardnesses. (Other classifications for high-Cr irons are given in Appendix E.)

Not only are there more variants of high-Cr irons as compared to the Ni-Hard irons, their mechanical and wear properties also span a broader range. It is impossible to define a typical high-Cr white cast iron. One characteristic of the high-Cr family is that it is a bit different from the Ni-Cr irons in its as-cast structure. After pouring, solidifying and cooling to room temperature, the high-Cr irons are supersaturated with C and Cr. In many high-Cr iron compositions, this high alloy content stabilizes the austenitic matrix all the way down to room temperature. Nonetheless, the compositional limits of the high-Cr family are so broad, that the as-cast microstructure can easily be ferritic, pearlitic, bainitic, ausferritic, martensitic, austenitic or usually a mixture. This incredible ferrous matrix flexibility is one of the reasons that high-Cr irons are used in a wide range of applications.

Figure 1.6 shows a polished plane cut from a thick-sectioned high-Cr iron in the as-cast condition. The photomicrograph is quite bland when compared to the more contrasting images presented by the Ni-Hard family. This uniformity of contrast arises from the relatively inert nature of the austenite matrix within the high-Cr iron. The high-Cr content of the austenite renders it highly resistant to attack by the 2% nital (see Appendix F for details on etchants) and, since other decomposition (pearlite or bainite) or transformation (martensite) products are present, the microstructure appears rather bland.

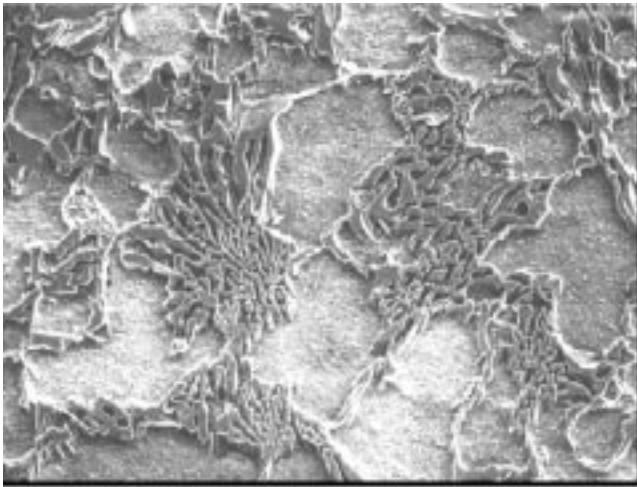
If the matrix is partially removed by etchant attack, Fig. 1.7 shows that the carbide network looks a lot like the Ni-Hard 4 carbides (as it should). Nonetheless, there are many subtle and wonderfully complex differences between the high-Cr and the Ni-Hard 4 irons that will be explained in subsequent chapters.

There are many reasons for and against casting high-Cr irons. The most persuasive argument for them is that high-Cr irons represent a multivariate system. This is an advantage because it offers the possibility to tailor them to specific applications and purposes; however, it involves high risks if

**Table 1.4**  
**Chemical Composition and Hardness of High-Cr AR Cast Irons**

| Alloying Element          | %C      | %Cr | %Mo     | %Mn     | %Si     | %Ni      | Brinell Hardness, HB <sup>(1)</sup> |
|---------------------------|---------|-----|---------|---------|---------|----------|-------------------------------------|
| High-Cr<br>ASTM A532-IIA  | 2.4–2.8 | 12  | 0.5–1.0 | 0.5–1.5 | 1.0 max | 0.5 max  | 400–700 (typical)                   |
| High-Cr<br>ASTM A532-IIB  | 2.4–2.8 | 15  | 1.0–3.0 | 0.5–1.5 | 1.0 max | 0.5 max  | 400–700 (typical)                   |
| High-Cr<br>ASTM A532-IIC  | 2.8–3.6 | 15  | 2.3–3.5 | 0.5–1.5 | 1.0 max | 0.5 max  | 400–700 (typical)                   |
| High-Cr<br>ASTM A532-IID  | 2.0–2.6 | 20  | 1.5 max | 0.5–1.5 | 1.0 max | 1.5 max  | 400–700 (typical)                   |
| High-Cr<br>ASTM A532-IIE  | 2.6–3.2 | 20  | 1.0–2.0 | 0.5–1.5 | 1.0 max | 1.5 max  | 400–700 (typical)                   |
| High-Cr<br>ASTM A532-IIIA | 2.3–3.0 | 25  | 1.5 max | 0.5–1.5 | 1.0 max | 1.5 max. | 400–700 (typical)                   |

(1) Hardness varies depending on heat treatment.



**Fig. 1.7. 3D carbide network in a high-Cr AR iron. Strong similarity can be seen between the carbide network shown in this figure and that shown in Fig. 1.5. This is as it should be, since both irons go through the same type of eutectic reaction. 320X.**

the decision is wrong or the process control is insufficient. For example, composition, casting geometry, melting practices, cooling conditions in the mold and heat treatment must be properly balanced and extremely well controlled for each application to produce crack-free marketable castings. These issues are often more than enough to dissuade the financially timid. Moreover, it is difficult to troubleshoot alloy and casting problems in these irons due to the paucity of information available to the practicing foundryman. In many situations, if the information is not within reach on the desktop, it is out of reach for most foundrymen. It is hoped that this book will provide the desktop reference that will allow practicing foundrymen to understand the mechanisms behind the formation of austenite or martensite in their chosen iron and to understand why these irons are so susceptible to cracking while others are cast unscathed.

### Specialty AR Cast Irons

If standards existed for these specialty irons, they would no longer be special. Their high-alloy content (Table 1.5) and their unusual foundry characteristics distinguish these irons.

#### **Ferritic High-Cr Specialty Irons**

The first example is the ferritic high-Cr family of irons, which contain around 2% C and 30% Cr. These irons are often used in applications where high *corrosion* (see Glossary) resistance is mandatory and good abrasion resistance is required. Their microstructure consists of a stainless steel-like matrix, surrounded by hard  $M_7C_3$  carbides. A key factor to their success is an excessively high Cr content within the ferrous matrix. Their high Cr content reduces corrosive attack from oxidizing acids (e.g.,  $HNO_3$ ) but not particularly from reducing acids (e.g., HCl).

#### **Vanadium-Based (A7) Specialty Irons**

The down side to the V-based irons (also known as A7) are their poor processing characteristics. The reasons behind these characteristics are straightforward if one understands how and why irons harden (in Chapter 3) and how feeding difficulties are tied to solidification characteristics (in Chapter 2).

Other specialty irons are based on vanadium to create hard, abrasion-resistant  $M_xC$  eutectic carbides within their ferrous matrices. This class of irons is a subset of tool steels. If one follows the nomenclature laid down by our forefathers' metallurgists, any ferrous material containing less than 2% C must be an *steel* and not an *iron*. However, a more useful definition is to denote irons as ferrous-based compositions that undergo a eutectic reaction during solidification. Hence, perhaps a new name for these materials might be "tool irons."

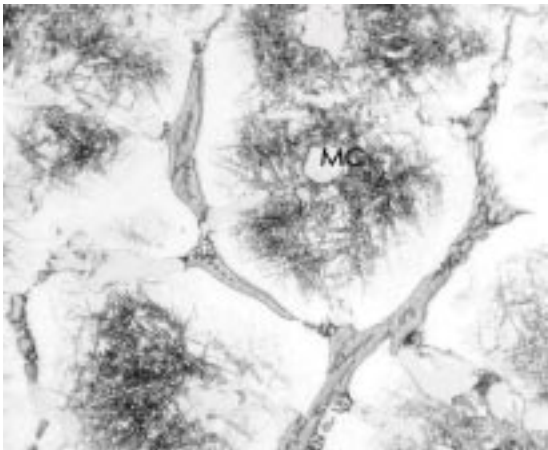
The V-alloyed cast irons are extremely tough and extremely abrasion resistant (in certain applications). An example of this type of iron is shown in Fig. 1.8. The dominant

carbide is V-based and is often noted as MC in the literature. High cost has restricted their usage to applications where order-of-magnitude improvements in wear life can be realized. One such application is in hot strip mill rolls.

### Niobium-Based Specialty Irons

Some metallurgists have added niobium (Nb) to the iron to create very hard niobium carbides (NbC) within the ferrous matrix. This expensive addition produces irons that can be extremely wear-resistant but have some unusual foundry problems. For example, the addition of Nb changes the fluidity of the iron for the worse. That is, as the iron solidifies, niobium carbides will form almost instantly within the melt. These carbides will then impede the flow of metal through the mold and may have unintended consequences, i.e., formation defects in the casting where none existed before.

The addition of Nb can also rob of the iron of hardenability. For example, there is only so much C to go around, and if the addition of Nb tips the scale toward impoverishing the ferrous matrix of C, then the iron will never obtain the hardness that the customer demands. As with any new alloy venture, these shortcomings can be handled with careful thought.



**Fig. 1.8.** Optical micrograph of a polished and etched surface of a castable tool steel alloy. The microstructure is composed of both MC and  $M_7C_3$  carbides, surrounded by a mixed matrix microstructure of austenite and martensite. The blocky MC carbides can be seen in the center of the austenite dendrites, while the blade-like  $M_7C_3$  carbides can be seen at the edges of the dendrites. 1000X, etched with Murakami's and 2% nital.

### D2 and 440C Specialty Irons (Tool Steels)

Another interesting class of AR materials is that represented by the alloys designated commercially as D2 and 440C. Their compositions are given in Table 1.5. Although these materials are not actually irons, they form eutectic  $M_7C_3$  carbides during solidification. Based on this eutectic reaction, both D2 and 440C are classified as AR cast irons. They represent important, castable, AR cast irons that any foundry capable of pouring Ni-Hard and high-Cr could comfortably cast.

The main characteristics of these low-C, abrasion-resistant alloys are the microstructures of fibrous carbides that are widely separated by ferrite. Figure 1.9 shows an example of a heat-treated 440C with a ferritic/austenitic matrix. The iron's high alloy content has overstabilized the matrix, resulting in a mixed structure containing both austenite, ferrite and martensite phases. Since the carbides are quite far apart (when compared to the high-Cr or Ni-Hard irons), the toughness of these alloys is often superior to other AR cast irons. However, wear resistance to severe abrasion is compromised. Nonetheless, if toughness is critical (D2 alloy) or if corrosion resistance *and* toughness are critical (440C), then one of these irons might be the answer.



**Fig. 1.9.** Optical micrograph of a polished and etched surface of a heat-treated 1.1% C, 17% Cr (440C) AR cast iron. The blade-like  $M_7C_3$  eutectic carbides are surrounded by a thin layer of martensite. The very small dots or dark tracings within the ferrite/austenite dendrites are  $M_{23}C_7$  secondary carbides that formed during heat treatment. The light areas are large austenite dendrites. 500X, etched with 2% nital.

**Table 1.5**  
**Chemical Composition and Hardness of Specialty AR Cast Irons**

| Alloying Element          | %C      | %Cr   | %V   | %Nb | %Mo     | %Mn     | %Si     | %Ni     | Brinell Hardness, HB <sup>(1)</sup> |
|---------------------------|---------|-------|------|-----|---------|---------|---------|---------|-------------------------------------|
| Ferritic High-Cr          | 1.5–2.4 | 30–40 | 0.0  | 0.0 | 0.      | 0.0     | 0.0     | 0.0     | 300–450                             |
| V-Based A7 <sup>(2)</sup> | 2.3     | 5.25  | 4.75 | 0.0 | 1.1 max | 0.5     | 0.5     | 0.0     | 500–750                             |
| Nb-Based                  | 2.9     | 16.7  | 0.0  | 2.0 | 0.8     | 0.8     | 0.0     | 0.7     | 440–630                             |
| D2 Tool Steel             | 1.4     | 12.0  | 0.2  | 0.0 | 0.8     | 0.0     | 0.0     | 0.0     | 600–700                             |
| 440C Tool Steel           | 1.1     | 18.0  | 0.0  | 0.0 | 1.0–2.0 | 0.5–1.5 | 1.0 max | 1.5 max | 350–500                             |

(1) Hardness varies depending on heat treatment.

(2) These irons might contain anywhere from 0 to 7% tungsten (W) to boost the formation of hard tungsten carbides (WC).

# Chapter 2

## Solidification of AR Cast Irons

Understanding the solidification behavior of iron can tell a lot about how well it will flow into the mold, its tendency to form porosity, its toughness and its abrasion resistance. The biggest challenge in applying solidification theory to AR irons is trying to grasp the big picture. There exists a wealth of scientific information on solidification, and it is the goal of this chapter to boil this mass of information and esoteric theory down into usable technological information. To this end, this chapter is dedicated only to the *liquid-to-solid transformation* in AR cast irons. In other words, once the iron is solidified, we are done talking about it. What happens further on, as it cools in the mold, is covered in Chapter 3, Heat Treatment.

This chapter should provide the tools needed to gain immediate insight into the microstructure of AR cast irons, without having to resort to a lengthy or costly metallurgical investigation. For example, with basic chemistry information, it is possible to calculate the carbide volume fraction (CVF) and the initial matrix chemistry. With such information, wear properties can be estimated, optimum heat treatments designed and lean alloying strategies pursued. In short, understanding solidification on a practical level can be both interesting and financially beneficial.

### Binary Fe-C Systems (Fe<sub>3</sub>C and M<sub>3</sub>C AR Cast Irons)

Pearlite, chill-cast and Ni-Hards 1 and 2 fall into the Fe<sub>3</sub>C and M<sub>3</sub>C eutectic carbide category.

The reason regular gray iron will “chill” to white cast iron is due to the solidification rate. If the iron is cooled quickly, it races to a subnormal solidification temperature and becomes what is termed, *undercooled*. In this undercooled state, the iron is predisposed or “anxious” to form carbide instead of graphite. Figure 2.1 shows the *phase diagram* representation of this behavior. (See Glossary for *Phase Diagram*.) The *graphite* eutectic reaction occurs at a higher temperature than does the *carbide* eutectic reaction. The trick is to cool the iron fast enough, so that it races past the graphite eutectic and hits the carbide eutectic. In reality, it is a battle between forming graphite or carbide.

Graphite nucleates easier than does carbide (think of this stage as climbing a hill, the graphite hill is lower than the carbide hill), but once started, the carbide phase grows much faster than does the graphite phase (the down slope of the carbide hill is steeper than that of the graphite hill). This

behavior can lead to castings that are carbide-rich on the surface (white) and graphitic within the core of the casting (gray). When hot metal hits the cold mold, the cooling rate is extreme. As the iron fills the mold and the hot liquid is no longer in contact with the cold mold, the cooling rate becomes much slower. This slower cooling rate provides a ready-made opportunity for graphite to form.

Because it is a metallurgical process, this white-to-gray transition can go astray. For example, in some castings, one can find graphite on the surface and carbides in the core. This sort of behavior is termed *inverse grayness* or *inverse chill*. Inverse chill occurs when conditions for graphite nucleation are extremely favorable (e.g., high Si levels in the melt), while the solidification rate is favorable toward carbide growth. Initially, both graphite and carbide will nucleate near the surface. Because the carbide reaction is more vigorous than the graphitic reaction, the iron will contain a mixed structure of graphite and carbides near the surface and a core structure dominated by carbides.

Controlling the carbide- versus graphite-forming reaction, based strictly on solidification rate, is difficult but technologically possible. If carbides are desired, a more consistent method is to reduce the silicon content to the lowest feasible level. For example, as Fig. 2.2a shows, as the Si level is increased, the graphite eutectic temperature similarly increases. This partially explains why gray irons are typically alloyed with 1.5–2.5% Si. The higher the Si, the easier it is to form graphite and not carbide. However, if we are interested in producing a carbide-containing low-alloy iron, then the Si content should be as low as is reasonably possible, say 0.5% Si.

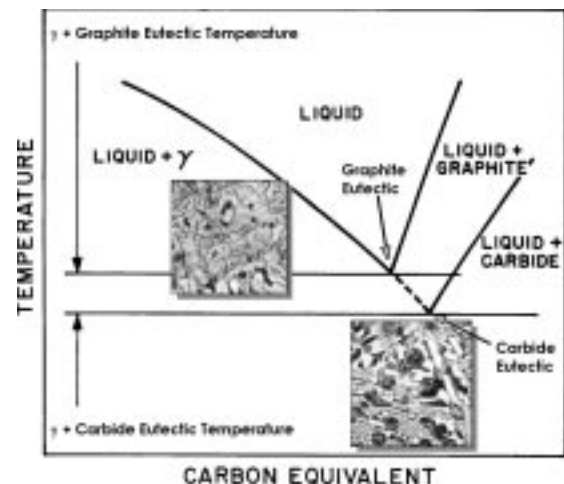


Fig. 2.1. Phase diagram example illustrating temperature interval between graphite and carbide eutectic temperatures.

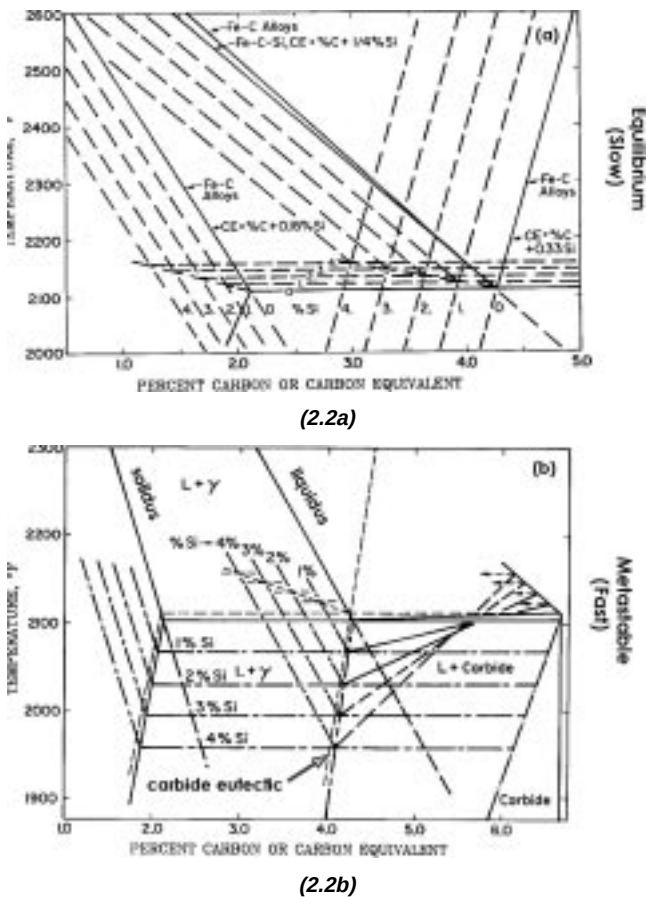
## Carbon Equivalency

Since we are discussing the use of Si in irons, the subject of carbon equivalency should be broached. Most foundry technologists are familiar with the use of a *carbon equivalent* (CE) (see Glossary) to determine whether a graphitic iron is *hypoeutectic* or *hypereutectic*. (See Glossary.) A common carbon equivalent formula is:  $CE = \%C + 1/3\% Si + 1/4\% P$ .

In low-phosphorus irons (say  $<0.1\%$ ), this formula is reduced to just:  $CE = \%C + 1/3\% Si$ .

Figure 2.2a shows how the addition of Si shifts the graphite eutectic composition to the left. Consequently, we can show that Si is a fairly good substitute for C in graphitic irons.

One may wonder if a similar formula exists for irons that solidify without the formation of graphite. Following the same approach used for graphitic irons, Fig. 2.2b provides some data on the effect of Si on the carbide eutectic composition. Using this data, the CE can be calculated to be around:  $CE = \%C + 1/10\% Si$ . Other work by Moore (1972) has provided the following equation for low-alloy white irons as:  $CE = \%C + 1/9\% Si + 1/3.5\% P$ . In other words, the effect of Si on the iron's effective C content is much lower or practically inconsequential.



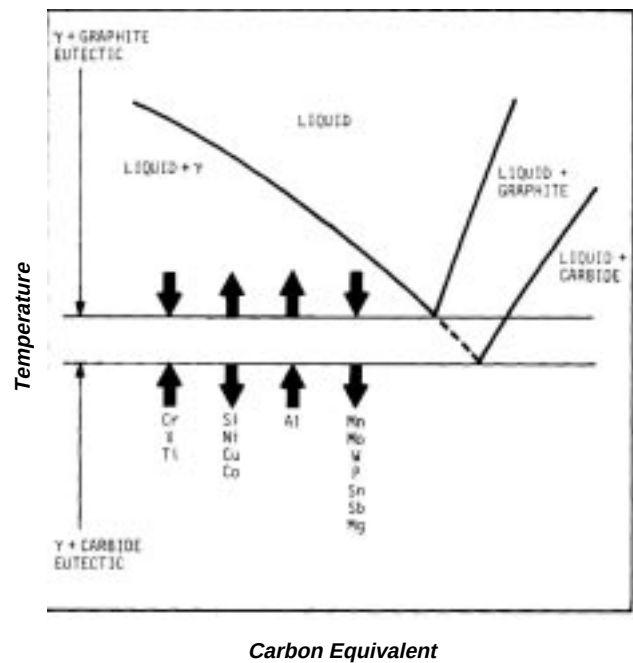
**Fig. 2.2.** Equilibrium (a) and metastable (b) phase diagrams for Fe-C irons containing 0, 1, 2, 3 and 4% silicon. As the Si level is increased, the graphite eutectic temperature is increased and the carbide eutectic temperature is decreased. This relationship helps explain why the addition of Si favors the formation of graphite over that of carbide. (After Heine, 1986)

## Effect of Low-Cr Additions

Another way to look at the issue of carbide versus graphite formation is to focus on the influence of the various alloying elements on the graphite and carbide eutectic temperatures. Figure 2.3 provides a qualitative view of how alloying elements can promote the formation of graphite or carbide. Ideally, when white iron is desired, one would want to move the graphite eutectic temperature down and the carbide eutectic up. Of all the alloying elements, chromium, vanadium and titanium do the best job. Other elements have more of a mixed effect. For example, aluminum pushes both the graphite and carbide eutectic temperatures up. Hence, graphite formation is favored because the cooling liquid will hit the graphite eutectic first.

In many industries, a duplex (bimetallic) casting consisting of a hard AR surface with a tougher gray iron core is often desired. Several alloying approaches are available to control the thickness of the AR surface layer. Traditionally, sulfur (S) has been used in conjunction with tellurium (Te) or bismuth (Bi) to drive the alloy toward carbide formation. The trouble with adding these elements is their troublesome history. In brief, S can easily segregate and form sulfides along the grain boundaries, causing hot shortness; and both Te and Bi pose health hazards in the foundry.

A more acceptable way to drive the alloy toward carbide formation is the use of chromium. A good example of Cr's effectiveness is illustrated by its ability to increase the chill depth. Figure 2.4 provides data on the chilling tendency of common alloying elements in irons containing between 3.0 and 3.3% C.



**Fig. 2.3.** Influence of alloying elements on graphite and carbide eutectic temperatures. Notice how Cr, V and Ti move the carbide eutectic up and the graphite eutectic down. Consequently, the addition of Cr, V or Ti is the perfect way to remove graphite and promote carbide formation. (After Janowak and Gundlach, 1982)

Manganese is quite interesting, since its effect is only pronounced at levels in excess of 1%, while at lower levels, it has very little effect. It follows that, nothing really beats Cr for chilling effectiveness at low cost.

The reason Cr is such a strong carbide former is due to its ability to stabilize the carbide structure. Chemists would say that the carbide energy of formation (see Glossary for *Gibbs Free Energy of Formation*) increases with increasing Cr content. Thermodynamically, the iron wants to form carbides over graphite, regardless of the cooling rate.

At low Cr levels (say below 2%) a lot of this is fairly qualitative. One is always taking odds on what will happen and when it will happen. For example, it is possible to form graphite in some Ni-Hard compositions by increasing the Si and/or C contents. Higher Si and C tend to hinder the chill-inducing ability of the Cr, and a mottled microstructure of graphite flakes and carbides can result. With regard to duplex (bimetallic) castings, the judicious control of Cr, C and Si is all that is typically necessary to specify the depth of the abrasion-resistant surface layer.

## Binary Phase Diagram

In order to use solidification as a tool, a basic understanding of phase diagrams is needed. Although they are almost always approached with some fear and loathing, phase diagrams are extremely powerful and useful. A key point to keep in mind is that they are maps and, as with any map, trying to understand the whole map at once can be overwhelming.

Before attacking the more complex ternary diagrams, let's start with the most basic binary diagram of metallurgy: the phase diagram of the binary Fe-C system, as shown in Fig. 2.5. Figure 2.5 lays out the road map for all commercial pearlitic and Ni-Hard 1 and 2 AR cast irons.

Irons are often defined as containing more than 2% carbon. A more accurate approach is to classify irons as either hypoeutectic, eutectic or hypereutectic, with an emphasis on *eutectic*. This means that an iron is a ferrous material that solidifies with

a eutectic reaction. This definition is more robust and frees us from the compositional constraint that an iron must have more than 2% C. The rationale for this more robust definition will become apparent when we start to deal with alloyed irons.

Many people may ask: What exactly does eutectic solidification mean? With a little help from Fig. 2.6, the solidifica-

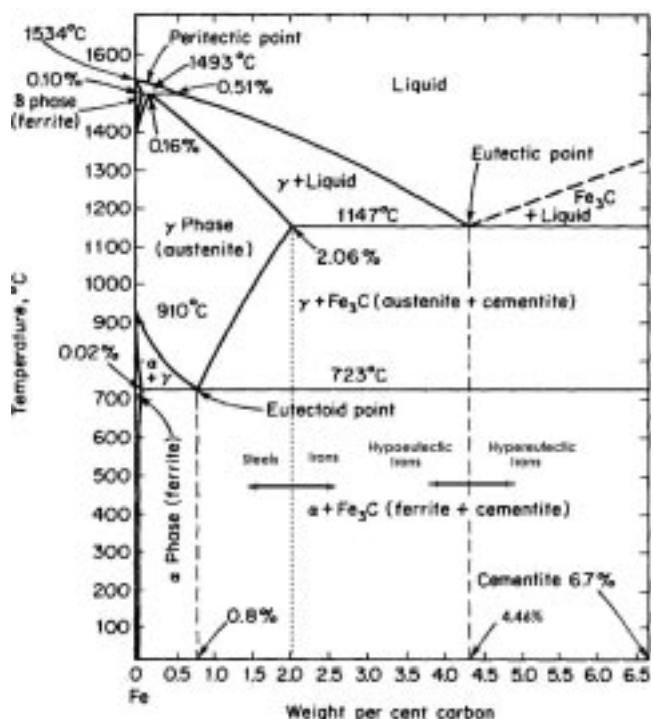


Fig. 2.5. Metastable Fe-C phase diagram. The classic definitive areas of steels and irons are superimposed on the diagram.

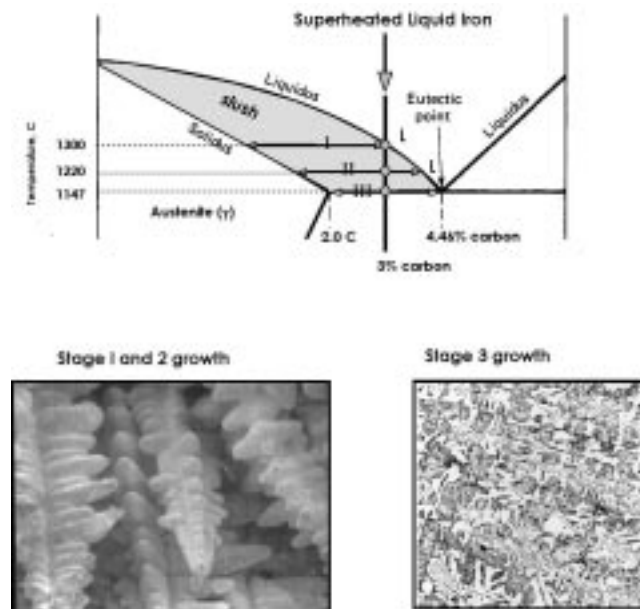


Fig. 2.6. Solidification stages of an Fe<sub>3</sub>C or M<sub>3</sub>C AR iron. Upper schematic shows three stages of eutectic solidification. Lower left shows dendritic structure common to all hypoeutectic irons and steels. (Deep-etched SEM is courtesy of E. Thomas, Atlas Foundry) Lower right shows hypoeutectic structure.

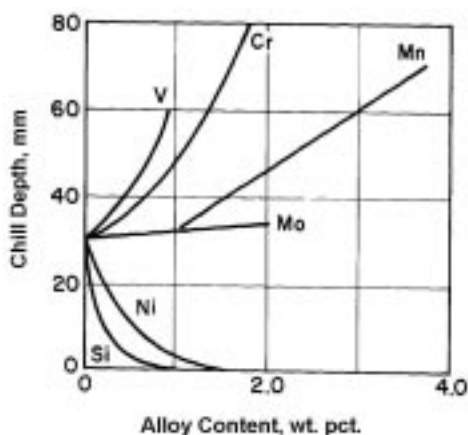


Fig. 2.4. Effects of alloying elements on chill depth. Iron was cast against a chill and obtained a chill depth of slightly more than 30 mm in the unalloyed condition. (After Janowak and Gundlach, 1982)

tion process in an  $\text{Fe}_3\text{C}$  or  $\text{M}_3\text{C}$  AR iron that contains 3% C can be described. As our solidification knowledge improves, we will also cover the combined effects of Cr, Si, Ni, Mn, Mo, etc.; however, for the moment, let us keep it simple and work within the constraints of the binary Fe-C system.

As shown in the top part of Fig. 2.6, the 3% C iron would be classed as hypoeutectic since it is solidifying to the left of the eutectic. The first phase to solidify is austenite and it has a face-centered-cubic (FCC) crystalline structure. Austenite is also referred to as gamma iron or by the Greek letter as “ $\gamma$ ” or “ $\gamma\text{Fe}$ .” The morphology of the austenite phase is typically described as dendritic or having the shape of a Christmas tree. An image of a cast steel dendrite is shown in the lower left of Fig. 2.6 and is representative of austenite dendrites growing in hypoeutectic irons.

As these dendrites grow, the various alloying elements will tend to partition between the dendrite and the remaining liquid. In the binary Fe-C alloy, the only partitioning element is C. The first dendrites to form have a C content of around 1.4% (Stage I at 1300C/2372F). As the iron continues to cool, the phase diagram indicates that the austenite will absorb more and more C atoms. Consequently, at Stage II at 1220C (2228F), the newly formed austenite dendrites will have a C content of around 1.75%. The excess C is simply rejected into the liquid. Hence, as the austenite dendrites continue to grow, the remaining liquid becomes richer and richer in carbon.

At the eutectic, the austenite will have a C content of around 2% and the remaining liquid will have a C content of around 4.4%. This remaining liquid then undergoes a eutectic reaction, forming both austenite and carbide phases. These crystalline phases grow in a coupled manner, i.e., the austenite and carbide grow at approximately the same rates. At the end of solidification, the microstructure consists of primary austenite dendrites and a mixture of eutectic austenite and carbides. As the iron cools to room temperature, some of this austenite will decompose to pearlite or transform to martensite. A typical microstructure of a 3% C pearlitic AR cast iron is shown in the lower right of Fig. 2.6.

If the solidification rate is high (i.e., nonequilibrium), the austenite dendrites will have a nonuniform distribution of carbon. This type of nonuniform distribution of alloying elements is referred to as solidification coring. For example, at the center of the dendrite, we would find a C content of around 1.4%, while near the periphery of the dendrite, the C content would approach 2%.

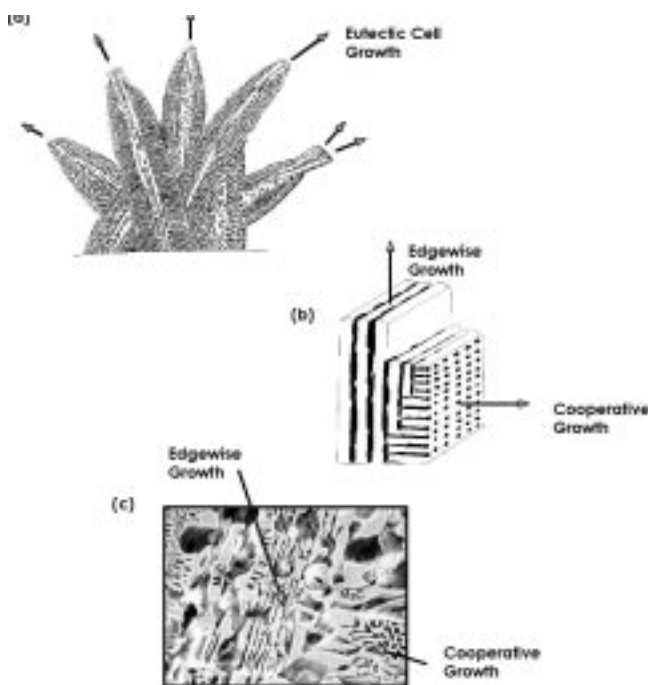
In the real world, castings are centimeters in thickness, and solidification rates are quite low. At these rates, interstitial C atoms quickly diffuse through the austenite crystalline lattice and eliminate any coring effect. In alloyed irons, substitutional elements of Cr, Mo, Ni, Mn, Si, and others are much slower to diffuse within the austenite and some coring can typically be found within the dendrite.

## Eutectic $\text{Fe}_3\text{C}/\text{M}_3\text{C}$ Growth Morphology

Eutectic reactions are the heart and soul of AR irons. For example, all the irons listed in the ASTM A532 specification (see Appendix E, Table E7) undergo eutectic reactions. How these eutectic carbides form, their size and their orientation all affect the toughness and wear resistance of the iron. Eutectic reactions are not complicated. For example, as the iron approaches the eutectic valley, we have tendrils of solid austenite surrounded by high-carbon liquid. When this high-carbon liquid hits the eutectic, carbides and austenite cooperatively spring forth.

How they spring forth as coherent mixtures of austenite and carbide is quite complicated and the subject of ongoing research. For the moment, let us focus on just the carbide phase. In pearlitic and the Ni-Hard 1 and 2 AR grades, the carbides have a *stoichiometry* (see Glossary) of three metal atoms to one C atom. In pure Fe-C alloys (e.g., pearlitic AR irons), the carbide can be denoted stoichiometrically as  $\text{Fe}_3\text{C}$ . Since many of these irons have some fraction of Cr, Mn or Mo (e.g., Ni-Hards 1 and 2)—and all of these alloying elements do partially substitute for the iron atom within the carbide crystalline structure—it is common to denote the carbide as  $\text{M}_3\text{C}$ . The  $\text{M}_3\text{C}$  carbides also go by the name of *cementite*. The eutectic mixture of austenite and cementite is called *ledeburite*.

Figure 2.7 shows a couple of schematics of the eutectic growth process in these AR irons. On a global scale, each *eutectic colony* (see Glossary) represents a separate growth event and is somewhat independent of its neighbors. Within



**Fig. 2.7. Solidification process in a low-alloy AR cast iron: (a) macroscopic view of eutectic colony growth (After Rickard and Hughes, 1961); (b) microscopic view of individual eutectic colonies (After Hillert and Rao, 1968); and (c) deep etched SEM micrograph showing edgewise and cooperative growth carbide structures.**

each eutectic colony, another separate growth process is occurring.

In pearlitic and Ni-Hard 1 and 2 irons forming eutectic mixtures of austenite and  $\text{Fe}_3\text{C}/\text{M}_3\text{C}$ , each eutectic colony hesitates between *edgewise* and *cooperative* growth (Fig. 2.7b). Edgewise growth occurs when the carbide phase is leading. When examined, it looks like a row of stacked plates. Cooperative growth happens when both phases are leading, and, in this case, it looks like Swiss cheese. Figure 2.7c shows an example of both edgewise and cooperative growth modes.

Theoretically, cooperative growth should dominate, but experimental evidence indicates that both edgewise and cooperative growth modes occur at about the same rates. Since both modes are operative, we are presented with a rather confusing microstructure.

This subject of eutectic solidification does have practical implications. From the previous text, it becomes clear that the eutectic forms a continuous network in between the austenite dendrites or grains; depending on the type of eutectic growth, the carbide fraction within the eutectic will be more or less continuous. If the carbides are continuous, then cracks can easily develop within the carbide network, leading to catastrophic fracture of the casting. In other words, the more continuous the carbide network, the lower the toughness of the iron. This microstructural feature also partially explains why  $\text{Fe}_3\text{C}$  and  $\text{M}_3\text{C}$  AR irons are typically hypoeutectic in order to have some degree of toughness. Much more will be said about the importance of this carbide network in Chapter 5, Morphology, etc.

### $\text{Fe}_3\text{C}/\text{M}_3\text{C}$ Carbide Formation

AR irons have superior wear resistance for one dominant reason: the presence of hard, wear-resistant carbides. As a good first approximation, you can use the phase diagram shown in Fig. 2.5 to predict the amount of carbides formed in pearlitic and Ni-Hard 1 and 2 AR cast irons.

For the most part, the amount of carbides formed in these irons is a direct function of the carbon. If we use our 3% C iron as an example and assume that we are casting a thick section (slow cooling and, hence, limited C coring), then we can assume that the iron's microstructure will consist of austenite having a C content of around 2% and  $\text{Fe}_3\text{C}/\text{M}_3\text{C}$  carbides having a C content of around 6.7%. This wt.% is simply based on the weight of three Fe atoms to one C atom and works out to 6.7 wt%. At this point, we can use a little math to construct a formula for calculating the wt% of carbides in a 3% C low-alloy AR iron:

$$\begin{aligned} \text{Eutectic}_{\text{Carbide}}\% &= \\ &= \left( \frac{(\text{Carbon}_{\text{Iron}}\% - \text{Carbon}_{\text{Austenite}}\%)}{(\text{Carbon}_{\text{Carbide}}\% - \text{Carbon}_{\text{Austenite}}\%)} \right) \cdot 100 \\ \text{Eutectic}_{\text{Carbide}}\% &= \left( \frac{3.0 - 2.0}{6.7 - 2.0} \right) \cdot 100 \\ \text{Eutectic}_{\text{Carbide}}\% &= 21 \end{aligned} \quad (2.1)$$

This technique is called the “lever rule,” and details on it can be found in most metallurgical textbooks. The term “lever” is logical, since you are taking a weighted balance between the original melt's C content on the one hand and that of the two individual austenite and carbide phases on the other hand.

It should be mentioned that, since the density of the carbide and iron phases are somewhat similar, the weight fraction is analogous to the volume fraction of carbide.

As the C content of the melt is increased, the carbide weight fraction increases quickly. For example, at the eutectic C content of 4.2%, the weight fraction of carbide is around 45%. This is a lot of carbides and, as one may correctly guess, the iron would be quite brittle but very resistant to straight abrasive wear. Much more will be said about the relationship between microstructure and wear properties in later chapters.

## Ternary Fe-Cr-C Systems ( $\text{M}_7\text{C}_3$ AR Cast Irons)

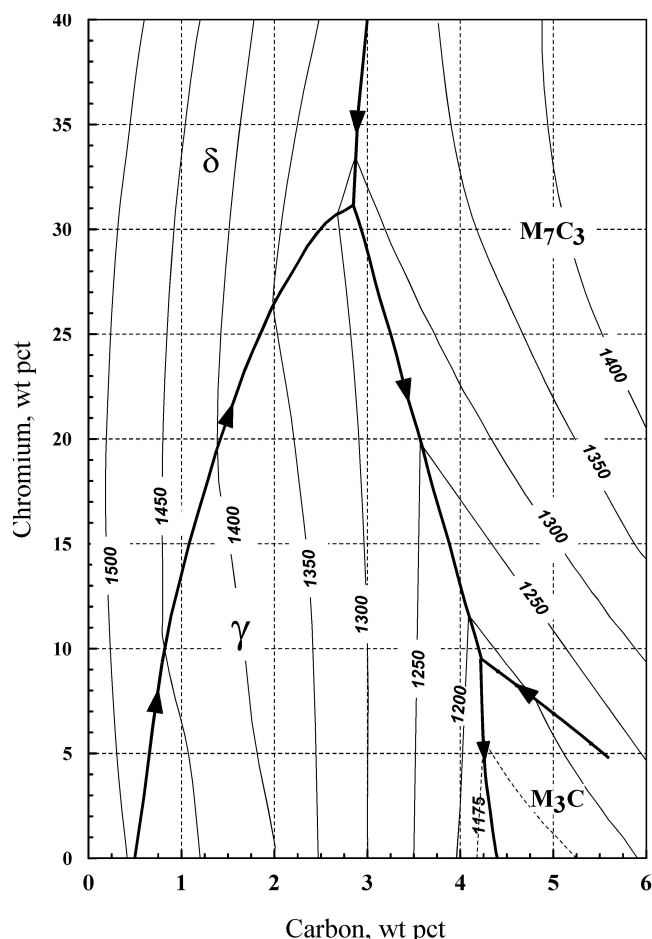
Ni-Hard 4 and high-Cr AR cast irons fall into the  $\text{M}_7\text{C}_3$  eutectic category. All of these irons solidify via a eutectic reaction that forms austenite and  $\text{M}_7\text{C}_3$  carbides. For the most part, the ternary Fe-Cr-C phase diagram can describe their solidification behavior.

Understanding the solidification of the ternary alloy system (Fe-Cr-C) is not trivial. In the low-alloy irons, we can safely work within the binary Fe-C phase diagram and make very good predictions about solidification temperatures and microstructures. As we add more and more Cr to the iron, it becomes more and more complex. Nevertheless, thanks to some very dedicated and smart scientists, we now have the tools to understand how high-Cr AR irons solidify. The end result of their work was the construction of the liquidus surface as shown in Fig. 2.8.

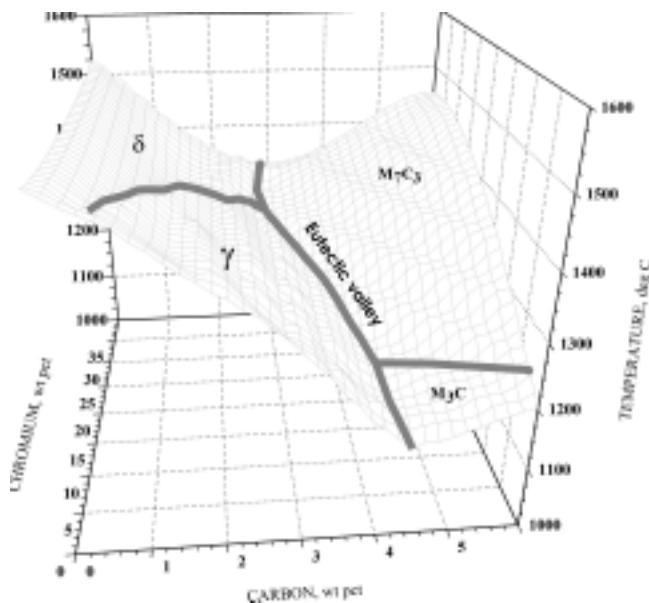
### Ternary Phase Diagram

The ternary phase diagram in Fig. 2.8 was constructed by Thorpe and Chicco (1985) and is based on hundreds of experimental measurements and, in part, on prior work published on Fe-Cr-C phase diagrams, such as the well-acknowledged work by Jackson (1970). The full name of this phase diagram is: Fe-Cr-C *metastable* liquidus surface. It governs the solidification behavior of 95% of today's commercial AR irons.

Figure 2.8 is a 2D representation of a 3D surface. Think of the liquidus lines as contours on a topological map. With a little work, these 2D lines can be transformed into a 3D surface, as shown in Fig. 2.9. The boundaries between major phases are shown as very thick lines superimposed on the liquidus surface. One of the major differences between binary and ternary diagrams is that 2D lines constitute the boundary between liquid and solid in binary diagrams (Fe-C); whereas, in the ternary diagram (Fe-Cr-C), a 3D liquidus surface forms



**Fig. 2.8.** 2D view of the Fe-Cr-C metastable liquidus surface. The contour lines represent constant liquidus temperatures (i.e., the temperature at which the first liquid starts to solidify). For example, a 2.5% C, 0% Cr iron will have the same liquidus temperature as that of a 2% C, 2.5% Cr iron. (After Thorpe and Chicco, 1985)



**Fig. 2.9.** 3D view of Fig. 2.8's Fe-Cr-C metastable liquidus surface. As the eutectic valley is approached, the slope of liquidus surface become steeper and steeper.

(Ternary Systems, continued)

the boundary between liquid and solid. Another way to think of this difference is that, in the binary Fe-C system, the liquidus temperature is determined by the alloy's C content, while in the ternary Fe-Cr-C system, the liquidus temperature is determined by both the C and Cr content. This concept of 3D solidification surfaces versus 2D solidification lines is one of the hardest concepts to grasp in understanding how the Fe-Cr-C system functions.

To thoroughly introduce the Fe-Cr-C system, we should note that, at 0.0% Cr, Fig. 2.8 presents just a section of the Fe-C phase diagram, with temperature plotted as contours instead of along an axis. Similarly, at 0% C, Fig. 2.8 presents a section of the Fe-Cr diagram. The rest of the diagram had to be determined by experimental tests on hundreds of Fe-Cr-C alloys, to determine when they started to freeze and what phases formed on freezing. This data was then used to construct the phase boundaries and its corresponding liquidus surface shown in Fig. 2.8.

There are four distinct liquidus surfaces shown in Fig. 2.8. These surfaces correspond to the phases that form on freezing:  $\gamma$  (austenite),  $\delta$  (ferrite),  $M_3C$  and  $M_7C_3$ . We already are familiar with the austenite and  $M_3C$  phases leaving only two new phases,  $\delta$  (ferrite) and  $M_7C_3$  (high-Cr carbide) to be discussed.

The surface marked as  $\delta$  denotes a compositional phase region in which alloys, falling on it, will freeze as a stable, Cr-rich ferrite, given as *delta ferrite* or  $\delta Fe$  (see Glossary). Subsequent to this initial ferrite formation, a variety of different solidification reactions can occur. For example, ferritic AR irons are cast using compositions of around 2% C and 30% Cr. Looking at Fig. 2.8 and locating the intersection of the vertical line at 2% C and the horizontal line at 30% Cr, the first phase to form is  $\delta$ . It should be mentioned that this type of iron is well known for its excellent corrosion and abrasion resistance and will be covered in greater detail in Chapter 9, Corrosive/Abrasive Wear Performance.

If the iron composition falls onto the  $\gamma$  liquidus surface, austenite dendrites will form and the iron will be classified as hypoeutectic. As the iron cools and continues to solidify, forming more and more *proeutectic austenite* (see Glossary), the composition of the remaining liquid falls toward the eutectic valley shown in Fig. 2.9. Once the eutectic valley is reached, the remaining liquid solidifies as a eutectic mixture of austenite and  $M_7C_3$ .

## Eutectic Reactions and Carbide Formation

The  $M_7C_3$  carbide is quite different from the  $Fe_3C/M_3C$  carbide found in the pearlitic and Ni-Hard 1 and 2 AR irons. For starters, the  $M_7C_3$  carbide (seven metal atoms and three carbon atoms) requires, for its formation, a minimum of 25% Cr within the base-iron melt and, once formed, has a typical Cr content of between 45 and 60%. The crystalline structures of the two carbides also differ. The  $M_7C_3$  carbide has a hexagonal crystalline structure, while the  $M_3C$  carbide is orthogon.

Due, in part, to its different crystalline structure, the  $M_7C_3$  carbide is much harder than the  $M_3C$  carbide; approximately 1400 HV versus approximately 900 HV.

Finally, the morphology of the  $M_7C_3$  carbide is different from that of the  $M_3C$  carbide. The  $M_7C_3$  carbide is often described as rod or blade-like, while the  $M_3C$  carbide forms a more continuous plate-like structure.

Despite all the differences between the two carbides, they still form via identical paths; either as eutectic or hypereutectic carbides. Their morphology and metallurgical properties are discussed at length in Chapter 5, Morphology, Microstructures and Properties.

If the iron's composition falls onto the  $M_7C_3$  field shown in Figs. 2.8 and 2.9, it is classified as hypereutectic, and the first phase to form is the  $M_7C_3$  carbide. Similarly, if the composition falls within the  $M_3C$  field shown in Figs. 2.8 or 2.9, the hypereutectic iron will contain primary  $M_3C$  carbides. In both cases, the iron still ends up solidifying along the eutectic valley shown in these figures.

In the binary diagram (Fig. 2.8), Fe-C, the eutectic reaction happens at a fixed temperature, and the amount of eutectic formed is fixed by the C content. In the ternary Fe-Cr-C system (Fig. 2.9), this no longer holds true. Indeed, as Fig. 2.8 shows, the eutectic reaction is governed by both the C and Cr contents of the iron, and it does not happen at a fixed temperature. Figure 2.9 shows a sloping valley, with eutectic reactions occurring at higher and higher temperatures as the Cr content is increased. This has important foundry consequences. For example, the nearly eutectic composition of Ni-Hard 1 poured at 1315C (2400F) has (based on Fig. 2.8) a eutectic temperature of 1150C (2100F), and the pouring superheat is 165C (300F).

Switching to a nearly eutectic 3.2% C, 25% Cr iron, Fig. 2.8 indicates that this iron will have its eutectic reaction at 1275C (2325F). If this iron is poured at the same temperature as the Ni-Hard, the superheat is only 50C (90F).

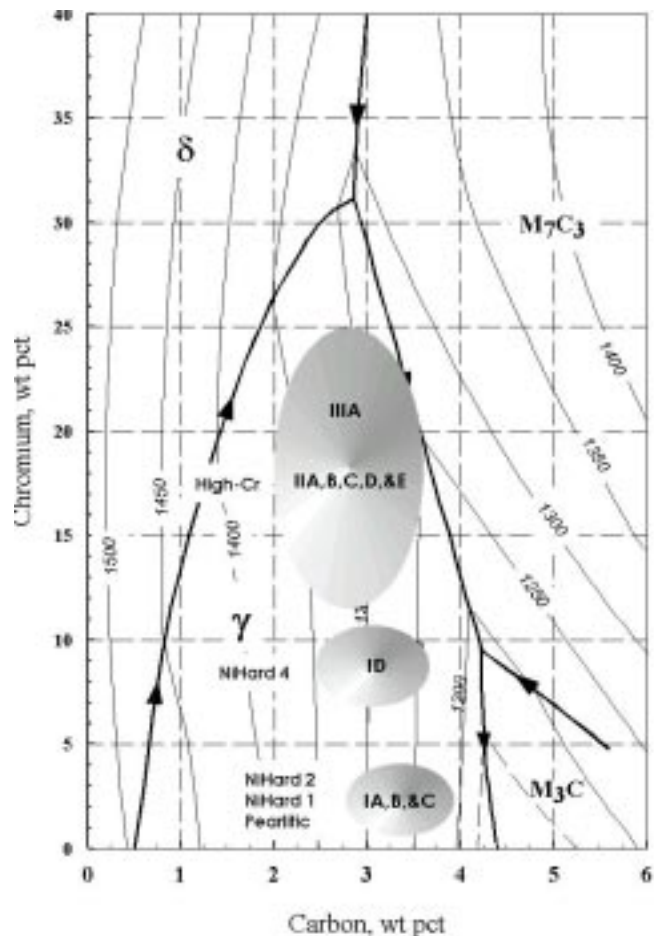
The majority of AR cast irons solidify onto the austenite liquidus surface. To clarify this point, Fig. 2.10 shows the irons that comprise the ASTM A532 specifications (see Appendix E, Standards, Table E7, for this and other iron specifications) superimposed onto the metastable liquidus surface. It is clear from Fig. 2.10 that all of the A532 irons are hypoeutectic. Thus, the first phase to solidify will be austenite followed by a eutectic reaction. The only major solidification difference between the alloy groups shown in Fig. 2.10, is that the low-alloy AR irons (shown as IA, B and C at the bottom of Fig. 2.10) form eutectic mixtures of austenite and  $M_3C$  carbides, while all the other irons (ID, IIA, B, C, D and E and IIIA) form eutectic mixtures of austenite and  $M_7C_3$  carbides.

The reader may wonder why the lower Cr irons (designated as ID) form  $M_7C_3$  carbides and not  $M_3C$  carbides. In the binary Fe-C system, we were dealing with a fixed 2D system. As solidification progresses in the Fe-C system, the composition of the remaining liquid is pushed horizontally toward increas-

ing C content, until reaching a fixed eutectic point. At the eutectic, the reaction  $L \Rightarrow \gamma + M_3C$  then proceeds. This reaction is fixed at a single composition (4.42% C) and a single temperature (1147C/2097F).

In the ternary Fe-Cr-C system, hypoeutectic irons reject both C and Cr during solidification. As the austenite dendrites grow during the solidification process, the remaining liquid becomes enriched in C and Cr, and its representative point is pushed diagonally toward the eutectic valley. Once the liquid reaches the eutectic valley, it will solidify over a range of temperatures. In ternary systems, the eutectic reaction does not occur at a fixed temperature, nor do the two solidifying phases ( $\gamma + M_7C_3$ ) have a fixed composition. That is, as the liquid reaches the eutectic, it starts to form eutectic colonies. As the alloy continues to solidify, the remaining eutectic liquid moves down the eutectic valley, decreasing in temperature.

The reason that the solidifying austenite dendrites reject both C and Cr has to do with complex interrelationships between the substitutional Cr atoms and the interstitial C atoms within the FCC crystalline structure of the austenite dendrite. Simply put, as the Cr content goes up, the C content of the austenite dendrites goes down.



**Fig. 2.10. Majority of AR cast irons have compositions that fall within the shaded regions lying on the austenite liquidus surface.**

Figure 2.11 shows vertical sections through the ternary Fe-Cr-C phase diagram at constant Cr contents increasing from 0 to 20% Cr. (At 0% Cr we are looking at the austenite field from the binary Fe-C phase diagram in Fig. 2.9.) For clarity, this representation has been limited to the austenite field of the solidus surface. The size of this field shrinks with increasing Cr and disappears at 20%. Consequently, Cr is considered to promote the formation of ferrite, since the austenite field shrinks while the ferrite field expands.

As the alloy's Cr content increases, the solidus line moves to the left and the C content, above which a eutectic reaction takes place, decreases more and more, while the temperature of the eutectic reaction increases continuously. The solidus line moving to the left signifies that, when the Cr content of the alloy increases, the C content of the austenite dendrite decreases over the full range of proeutectic solidification.

The slight initial decrease of the eutectic temperature shown in Fig. 2.11 reflects the use of the graphite eutectic (equilibrium) temperature at 0% Cr, which is higher than the carbide eutectic (metastable) temperature at 0% Cr. Otherwise, ignoring the 0% Cr eutectic temperature, the eutectic temperature rises as the Cr content is increased. Hence, as one moves up the eutectic valley, the eutectic austenite will have more and more Cr and less and less carbon.

For hypereutectic irons, carbides are the first phase to form upon solidification. As these Cr- and C-rich carbides form, the remaining liquid is rapidly depleted of C and Cr. For example, as  $M_3C$  and  $M_7C_3$  carbides form, they require approximately 6.8 and 8.9% C, respectively. With respect to Cr, the  $M_3C$  carbide has a maximum Cr solubility of about 15% Cr, while the  $M_7C_3$  carbide is theoretically unlimited; however, in most high-Cr irons, the  $M_7C_3$  carbide will contain no more than about 60% Cr.

Considering the very high carbon and chromium contents of the carbides (called *primary carbides* since they are formed prior to eutectic solidification), it is entirely understandable that the remaining liquid is quickly depleted of C and Cr during their formation. Consequently, as a hypereutectic iron solidifies, the remaining liquid is pushed toward lower C and Cr contents (left and down on the liquidus surface shown in Fig. 2.8) as it crosses the decreasing temperature contours toward the eutectic valley. (See Glossary for *Primary Carbides*.)

It is possible to show this C and Cr segregation behavior in both hypoeutectic and hypereutectic irons by taking careful and precise temperature measurements as the iron solidifies. By superimposing the iron's initial chemistry and its solidification temperatures onto Fig. 2.8, a direct view of this C and Cr segregation behavior is given in Fig. 2.12. As shown in Fig. 2.12, the proeutectic austenitic and ferritic phases of hypoeutectic irons, all reject C and Cr into the liquid (the sequence of arrows moves to the right and up) as they approach the eutectic valley. On the contrary, if the iron is hypereutectic, the liquid

is depleted of C and Cr as it approaches the bottom of the eutectic valley.

Once the iron hits the eutectic line, whether it is hypoeutectic or hypereutectic, it undergoes eutectic solidification over a range of temperatures. As mentioned before, this is one of the major differences between the binary Fe-C and ternary Fe-Cr-C systems. In the ternary Fe-Cr-C system, as the liquid hits the eutectic line and starts to freeze, the composition is not fixed as it cools, and changes as it rolls down the eutectic

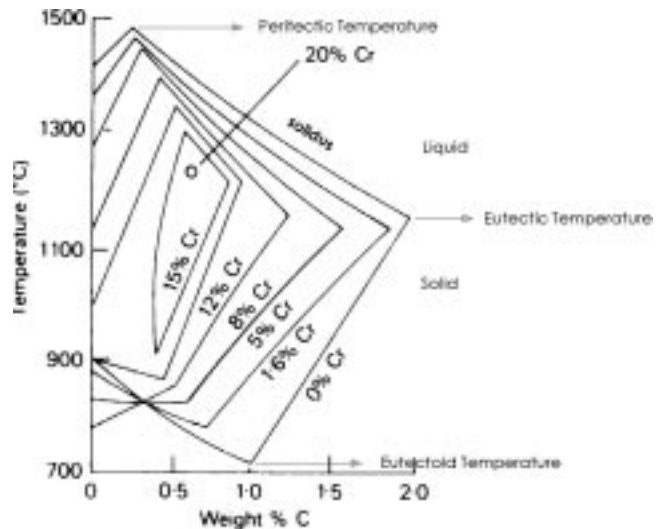


Fig. 2.11. Effect of Cr in shrinking austenite field. At 0.0% Cr, we have the same austenite field as shown in Fig. 2.2. (After Tabrett et al., 1996)

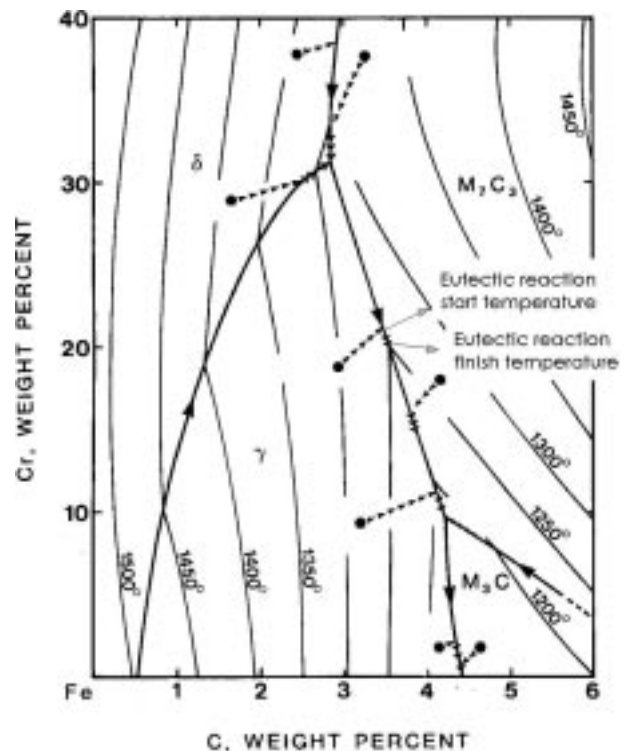


Fig. 2.12. Evolution of composition of remaining liquid during solidification for different Fe-Cr-C alloys superimposed on the metastable Fe-Cr-C liquidus surface. (After Thorpe and Chicco, 1985)

valley. In reporting eutectic solidification temperatures for AR irons, the purists will report both the start and finish of eutectic solidification; however, just the eutectic start temperature will often suffice.

Another way to view the solidification behavior shown in Fig. 2.12, is to construct a vertical 2D slice through the metastable Fe-Cr-C liquidus surface at a fixed Cr content. The slice shown in Fig. 2.13a is taken at 15% Cr, as we attempt to capture the solidification behavior of a 3% C, 15% Cr iron. Since we are not plotting the variation in Cr content during solidification, the compositional information is approximate.

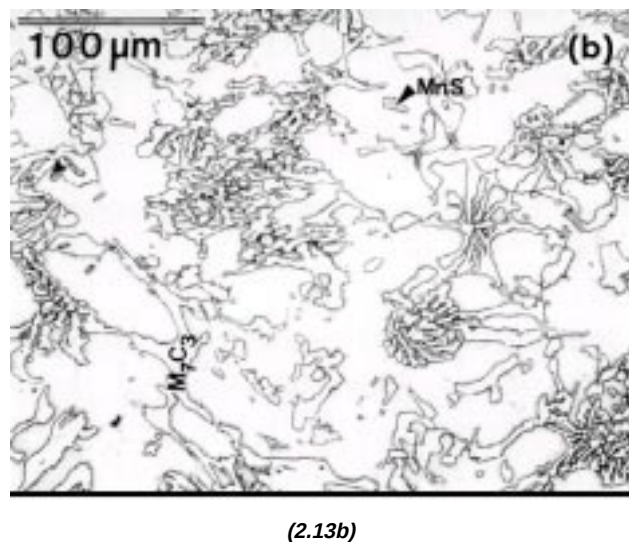
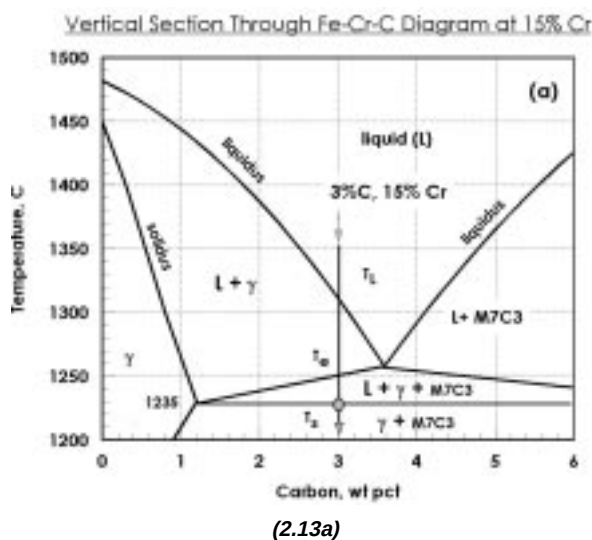
Figure 2.13a is quite a bit different from a similar vertical section at 0% Cr given in Fig. 2.5 representing the binary Fe-C system. The biggest difference is the existence of a three phase region of liquid (L), austenite ( $\gamma$ ) and carbides. Nonetheless, the iron still solidifies following a predictable sequence; that is, growing proeutectic austenite dendrites create a slushy mixture of dendrites and liquid, followed by a eutectic reaction of austenite and carbides. This eutectic mixture then has a eutectic start temperature, marked as  $T_e$  and a eutectic finish temperature, marked as  $T_s$ . At the end of solidification, one obtains a microstructure composed of austenite and  $M_7C_3$  carbides. An example of a 3% C, 15% Cr microstructure is given in Fig. 2.13b.

Figure 2.13a also shows that the eutectic austenite end composition is at 1.2% C. Based on this vertical section, an alloy containing 1.2% C and 15% Cr will form  $M_7C_3$  carbides through a eutectic reaction. Since the alloy is solidifying with a eutectic reaction, the alloy is correctly classified as a hypoeutectic iron. This happens because the Cr decreases the size of the austenite field, as described earlier. By using the data provided in Fig. 2.11, we can construct a line defining the

C and Cr limits for the eutectic austenite solidus field. Figure 2.14 shows this line superimposed upon the Fe-Cr-C liquidus surface. Alloy compositions that fall to the right of this line are classified as irons, since they undergo a eutectic solidification. For example, 440C (known commercially as a corrosion-resistant steel) containing 1.1% C and 17% Cr, is actually an iron because it solidifies with a eutectic reaction. (This is also true of D2 tool steel.)

Quantitative prediction of phase composition and phase volume fraction is much more difficult in ternary systems than in binary systems. Whereas in the binary system, the lever rule is used at the point of eutectic solidification, this point no longer exists in the ternary Fe-Cr-C system. For example, Fig. 2.14 presents the solidification temperatures for the 3% C, 15% Cr iron superimposed upon the liquidus surface. Once the eutectic valley is reached, the eutectic solidification temperature varies from  $T_e = 1257^\circ\text{C}$  ( $2300^\circ\text{F}$ ) to  $T_s = 1235^\circ\text{C}$  ( $2253^\circ\text{F}$ ). Although it is possible to make predictions about the various volume fractions of L,  $\gamma$  and  $M_7C_3$  while the iron is solidifying, it requires the construction of a tie triangle, with each corner of the triangle anchored by one of these phases. Since this procedure is rather complex, the reader is referred to other sources for more complete information (West, 1965).

At the end of eutectic solidification, only two phases exist for the irons that fall within the austenite liquidus surface: austenite and carbide. If the temperature at the end of eutectic solidification is known, a tie line can be constructed through the final eutectic solidification temperature and the iron's initial composition. The ends of this tie line will then intersect the phase boundaries of the eutectic austenite (constructed using the data presented in Fig. 2.11) and eutectic  $M_7C_3$  carbide. Such a construction is shown in Fig. 2.15. As shown, the final composition of the eutectic austenite is around 8.5%



**Fig. 2.13.** As an approximation, a 2D vertical slice (a) can be taken through the 3D Fe-Cr-C phase diagram in Fig. 2.9. Although this type of section does not provide true compositional information, it does provide us with the solidification temperatures. As the alloy crosses the metastable surface, it starts to freeze ( $T_L$ ). As it continues to solidify, C and Cr are rejected into the liquid, and the iron becomes a mixture of austenite dendrites and C- and Cr-enriched liquid. The start of the eutectic solidification ( $T_e$ ) occurs around  $1260^\circ\text{C}$ , and finishes ( $T_s$ ) at  $1230^\circ\text{C}$ . The microstructure of the fully solidified, hypoeutectic iron is shown in Fig. 2.13b.

Cr and 1.2% C. This agrees extremely well with microanalysis results taken at the center of proeutectic austenite dendrites in an as-cast 3% C, 15% Cr iron that yielded numbers of 9.5% Cr and 1.1% C. This is reasonable.

Correlating microanalysis results to phase diagram predictions is difficult, due to this C and Cr coring during solidification. For example, the same microanalysis procedure measured a Cr content of 45% in the  $M_7C_3$  carbide. Because the microanalysis technique is focusing on a spot at the center of the carbide, it is measuring the Cr content of the carbide formed at the very beginning of eutectic solidification and, therefore, at the highest Cr content.

There are two main problems with trying to arrive at composition and volume fraction information using the ternary Fe-Cr-C diagram. First, the tie line constructed in Fig. 2.15 is very sensitive to small changes in the eutectic finish temperature. For example, if the eutectic finish temperature is higher, the Cr content of the austenite is predicted to be higher. Second, few foundries have equipment to accurately measure solidification temperatures in their irons. This limitation alone hinders the use of phase diagram tie lines to predict compositional information.

At this stage, you may be wondering why we are going to such lengths describing tie lines and other solidification esoteric issues. The purpose of this discussion is to gain a strong understanding of what one is actually making when pouring an AR cast iron. With this fundamental knowledge, it will be easier to understand the mechanical properties of the as-cast iron and, more importantly, how the iron behaves as it cools and as it is heat treated. This information will be used throughout future chapters

Although we jumped straight to eutectic solidification for the 3% C, 15% Cr iron, it is worthwhile to back up and work through the solidification process in the ternary system, from start to finish. As with the binary Fe-C system, it is possible to use the ternary Fe-Cr-C phase diagram to gain some idea of the segregation or coring behavior in proeutectic austenite (austenite that has formed prior to eutectic solidification).

Figure 2.16 uses a 3D combination of the austenite solidus surface (constructed from the austenite solidus lines in Fig. 2.11) and the liquidus surface for the Fe-Cr-C system given in Fig. 2.9. As the iron crosses the liquidus surface at 1300C (2372F), very small crystals of austenite nucleate and grow. As these austenite crystals grow, they reject both C and Cr into

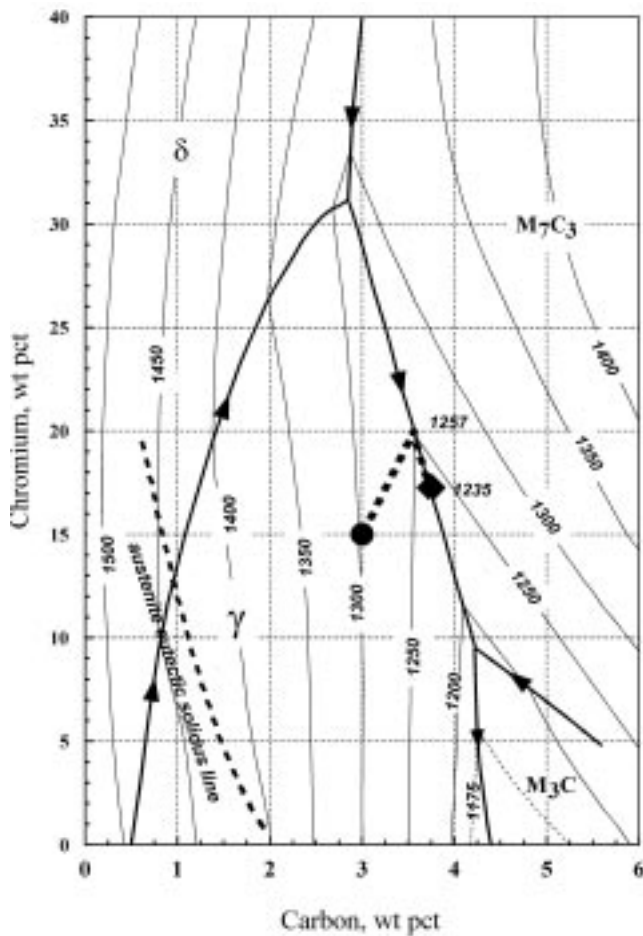


Fig. 2.14. Solidification path (thick dashed line) for a 3% C, 15% Cr iron. The boundary line between steels and irons is also shown on the left hand side of the figure.

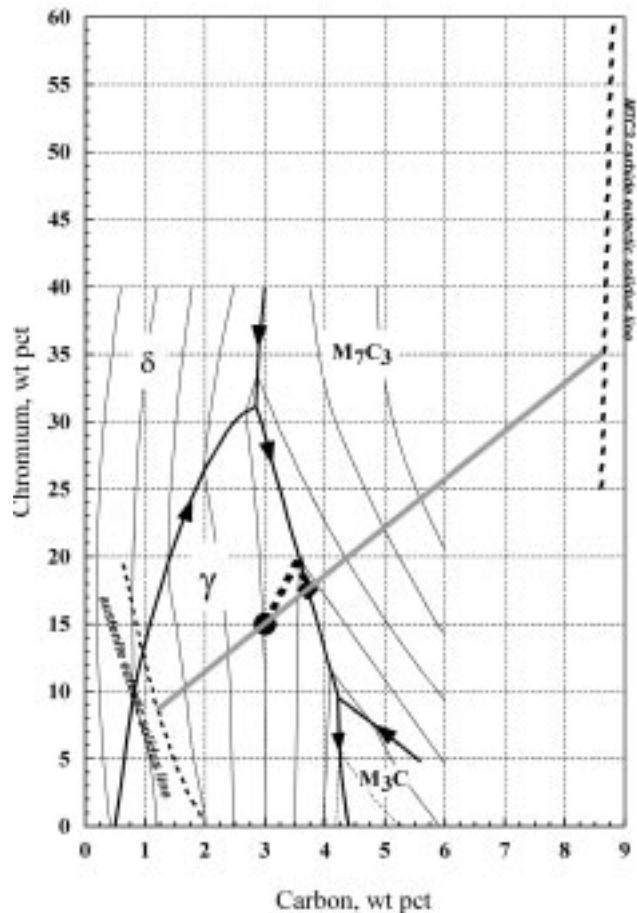


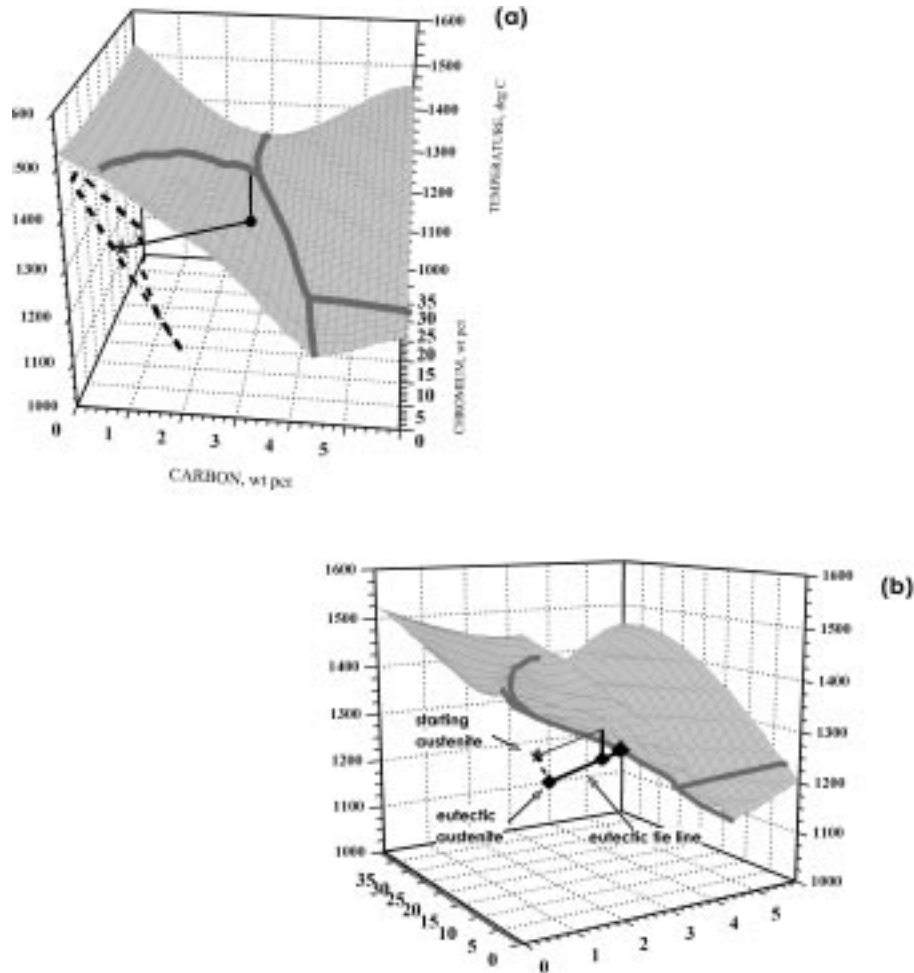
Fig. 2.15. Eutectic tie line allows a qualitative prediction of the composition and volume fraction of the austenite and carbide phases in the 3% C, 15% Cr iron. However, for this tie line to be constructed, the composition and ending eutectic solidification temperature must be known in advance.

the liquid. Subsequent austenite growth takes place within this enriched liquid bath and, consequently, the newly formed austenite is slightly richer in both C and Cr. This behavior is difficult to visualize, since we are more comfortable using straight lines between the different phase boundaries. One way to think about this behavior is that, as the liquid composition slides down the liquidus surface increasing in C and Cr, the composition of the growing austenite dendrites similarly slides down the austenite solidus surface and, likewise, increases in C and Cr. At the very end of proeutectic solidification, the liquid hits the eutectic line and starts eutectic solidification. The eutectic austenite solidus line (the bottom line of the austenite solidus field shown in Fig. 2.16a) is not reached until the very end of eutectic solidification. This reaction is shown in 3D by the tie line in Fig. 2.16b and shown in 2D in Fig. 2.15.

This explanation is meant to be qualitative. Quantitative phase composition from the ternary Fe-Cr-C diagram is very difficult, and requires hard-to-obtain high-quality experimen-

tal data. Nevertheless, the important theme to this discussion is that the solidification of AR cast irons follows predictable patterns. Although we may not be able to determine the exact pattern or, in other words, the exact phase compositions of the austenite, ferrite or carbide phases during solidification, we can make extremely good estimates of the relative C and Cr contents of these phases. This compositional information allows a much more effective approach to the development of new alloys and subsequent heat-treatment strategies.

The C and Cr coring described in the solidification of proeutectic austenite is a very real phenomenon in high-Cr AR irons. Although much of the C coring disappears as the casting slowly cools to room temperature, Cr coring will still exist within the austenite dendrite in all but the thickest of castings. This Cr inhomogeneity gradient yields a certain unpredictability to the mechanical properties and heat-treatment characteristics of thin-sectioned castings. This subject will be covered in more detail in Chapter 3, Heat Treatment of AR Cast Irons.



**Fig. 2.16. Qualitative approach to understanding the solidification of a 3% C, 15% Cr AR cast iron. As the iron crosses the liquidus surface (a), the melt separates into a C- and Cr-enriched liquid and an austenite component at significantly lower C and Cr. As the iron continues to solidify (b), the eutectic valley is reached, and eutectic solidification commences. Only at the very end of eutectic solidification is the austenite eutectic solidus line reached.**

Before ending the discussion of the Fe-Cr-C ternary diagram, we should mention the work done by Arnold and Sare (1987) using commercial AR irons from a number of foundries. They found that the eutectic line representing the bottom of the eutectic valley is, in fact, not a line but a surface, the width of which increases with increasing Cr content, as

illustrated in Fig. 2.17. It means that, in practice, for any given Cr content, a range of C contents will lead to a “fully” eutectic microstructure. It thus becomes easier and easier to hit a eutectic composition with increasing Cr content, to obtain a eutectic structure, despite variations in C content. More importantly, all will have the same microstructure and, likewise, similar mechanical and AR properties.

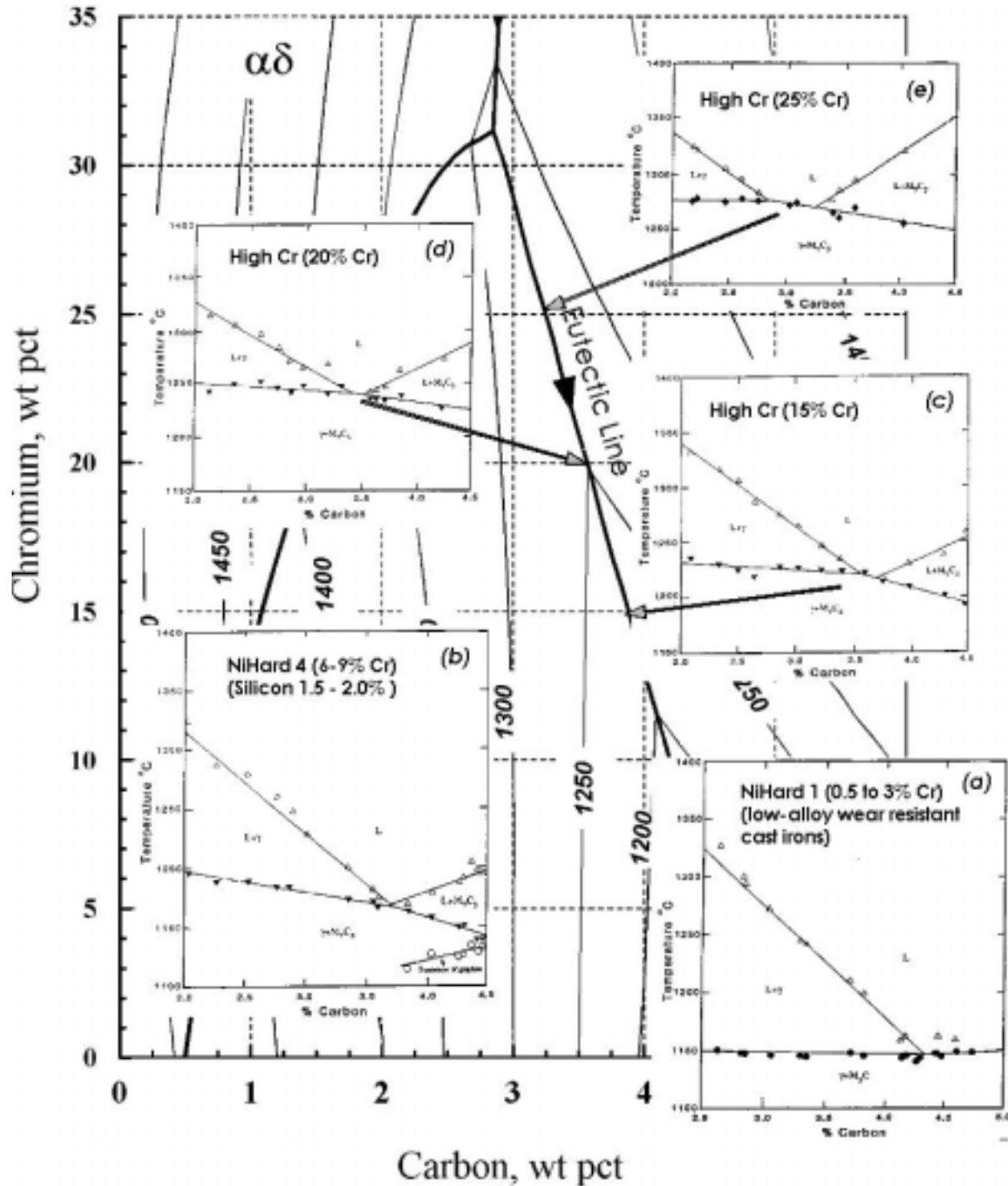


Fig. 2.17. 2D vertical slices at constant Cr content through the 3D metastable liquidus surface. Ni-Hard 4 (b) is the only AR iron that has its solidification dynamics affected by its Si content. As the Cr content is increased (c) through (e), the width of the eutectic “zone” is increased. (After Arnold and Sare, 1987)

### **M<sub>7</sub>C<sub>3</sub> Eutectic Growth Morphology**

As might be expected, the high-Cr eutectic reaction ( $L \Rightarrow \gamma + M_7C_3$ ) has its own unique morphology; eutectic colonies show up as cylinders in the schematic representation (Fig. 2.18a). A eutectic colony of M<sub>7</sub>C<sub>3</sub> carbide and austenite has one dominant growth direction. Hence, during eutectic-colony growth, we get nicely defined cylinders instead of the more random structure shown by the ledeburitic eutectic reaction ( $\gamma + M_3C$ ). For more information on this subject, refer to References/Bibliography at the end of this chapter.

It gets confusing, however, when we view a 2D plane section through the M<sub>7</sub>C<sub>3</sub> eutectic mixture. Figure 2.18b shows an interesting, deep-etched SEM picture of a carefully polished block. Clusters of M<sub>7</sub>C<sub>3</sub> eutectic colonies can be seen running up against other clusters. This photomicrograph beautifully illustrates the random orientation of the eutectic colonies.

In most cases, some effort is taken to arrange the viewing angle such that the viewer only sees one M<sub>7</sub>C<sub>3</sub> eutectic colony at a time. This type of photomicrograph is shown in Fig. 2.18c. In this picture, the eutectic has grown cooperatively in a single direction, and our section plane is approximately perpendicular to the colony's primary growth direction.

### **M<sub>7</sub>C<sub>3</sub> Carbide Formation**

Although we could use the tie line constructed in Fig. 15 and estimate the carbide volume fraction (CVF), it is simpler to use formulae developed by Maratray and Usseglio-Nanot (1971) for calculating CVF:

$$\begin{aligned} \text{CVF}\% &= 12.33 \times \%C + 0.55 \times \%Cr \\ &- 15.2 \text{ accuracy} = \pm 2.1\% \text{ CVF} \end{aligned} \quad (2.2)$$

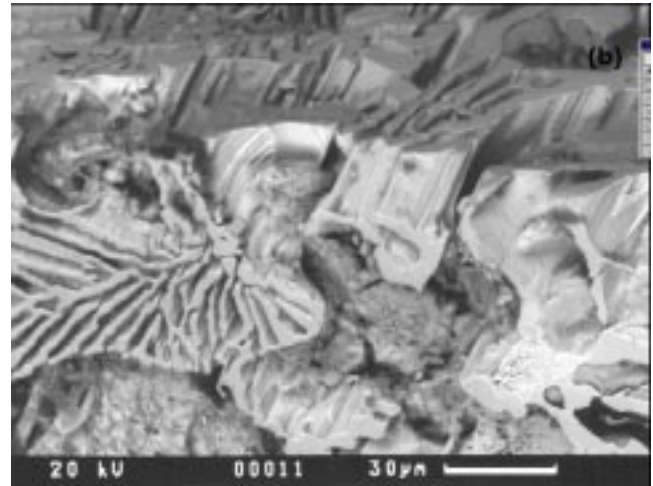
It was experimentally determined and results were provided that are reasonably close to the predictions made by using the phase diagram. For example, for the 3% C, 15% Cr iron covered in earlier sections, the CVF estimate using Maratray's formula, and by using the tie line shown in Fig. 2.15, is:

$$\begin{aligned} \text{CVF}\% &= 12.33 \times 3 + 0.55 \times 15 \\ &- 15.2 \text{ CVF}\% = 30 \\ \text{Tie Line CVF}\% &= 26 \end{aligned} \quad (2.3)$$

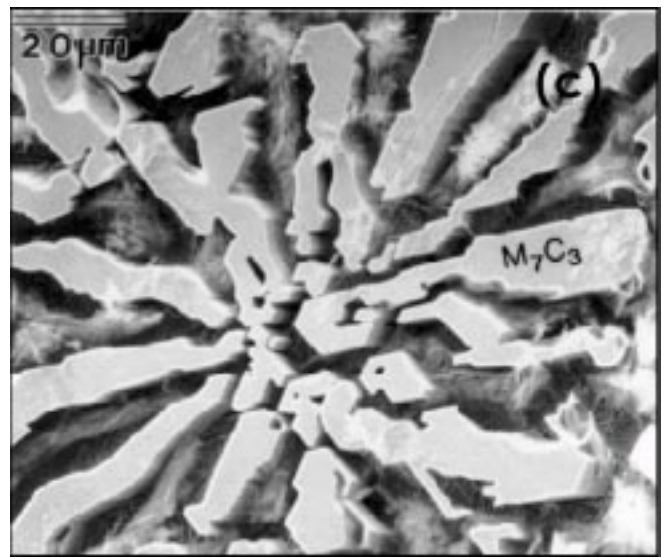
Maratray's formula seems to give results a bit on the high side, as compared to the phase diagram method. Generally, a 2–3% deviation will be found, due to the fact that Maratray's formula includes both eutectic and secondary carbides (secondary carbides are those formed by way of solid-state precipitation from the austenite during heat treatment or cooling to room temperature). The role of secondary carbides, where they come from and what they do, will be covered in Chapters 3 and 4, Heat Treatment and Alloying for Hardenability.



(2.18a)



(2.18b)



(2.18c)

**Fig. 2.18. Solidification characteristics of high-Cr AR cast irons. At first, (a) eutectic colonies grow as tapered cylinders; (b) their random growth orientation is more clearly shown by this deep etched SEM micrograph. (After Tabrett, 1996). Nonetheless, (c) each individual eutectic colony has a specific growth direction and individual carbides can often be described as rods or laths.**

## Determining Solidification Temperatures

There is a simple way to determine solidification temperatures and predict the amount of superheat needed (see Appendix G).

Figure 2.14 shows how proeutectic austenite solidification and eutectic start and finish temperatures can be determined. A comparison to the applied pouring temperature yields the superheat. In the case of the Ni-Hards (1, 2 and 4), these temperatures are somewhat lower, due to the influence of nickel.

With these estimates, one can make a pretty good prediction as to the actual superheat to use for that iron. Doing so is really quite useful in case there is a plan to change casting designs or switch between different irons for the same casting. Predicting the superheat acts as a guide to knowing the iron's relative fluidity in a particular mold, or knowing whether too much or too little superheat is being applied to the current casting process.

Authors have presented equations for calculating the eutectic temperatures (Tong et al., 1989); however, these formulas are only appropriate for hypoeutectic irons having Cr content between 10 and 20% (e.g., high-Cr irons). Considering these drawbacks, there is nothing like using the real solidification data given by the metastable Fe-Cr-C liquidus surface (see Fig. 2.8 or Fig. 2.9).

For example, estimating the solidification start or liquidus temperature, based on the iron's C and Cr content, is merely

a question of plotting its composition on the metastable Fe-Cr-C liquidus surface. Finding the exact eutectic start temperature, based on the phase diagram, is quite difficult. A simple and reasonably accurate approach consists of connecting the alloy's melt chemistry to the phase diagram's origin, and then extending this line to the eutectic line.

An example of this technique is shown in Fig. 2.19. For comparison, Tong's equations predict:

$$T_{\text{Liquidus}} = 1554.5 - 89.2 \times \text{C\%} - 0.77 \times \text{Cr\%}$$

$$T_{\text{Liquidus}} = 1554.5 - 89.2 \times 2.5\% - 0.77 \times 20\%$$

$$T_{\text{Liquidus}} = 1316 \quad (2.4)$$

$$T_{\text{Eutectic}} = 1201.7 - 20.3 \times \text{C\%} + 5.97 \times \text{Cr\%}$$

$$T_{\text{Eutectic}} = 1270 \quad (2.5)$$

Although the start of austenite solidification differs by about 15°C (27°F), the prediction for the start of eutectic solidification is correct. As one can see, the approximate accuracy of our predictions is probably around 15C (27F) for the start of austenite solidification and perhaps around 5C (9F) for the start of eutectic solidification.

To give this section more utility, Table 2.1 presents the start of solidification (liquidus) temperatures and the start of eutectic solidification (solidus) temperatures for a broad range of AR cast irons. As always, when in doubt, the Fe-Cr-C phase diagram should be consulted and a simple line extending from the origin through the alloy's composition should be used to find the solidification temperatures.

**Table 2.1**  
**Solidification Temperatures for ASTM A352 Designated AR Cast Irons**

| Material Class | Type | Designation <sup>(2)</sup> | %Cr                                                         | %C            | Austenite Start <sup>(1)</sup> (Liquidus) °C | Solidus Temp <sup>(1)</sup> °C | Freezing Range °C |
|----------------|------|----------------------------|-------------------------------------------------------------|---------------|----------------------------------------------|--------------------------------|-------------------|
| I              | A    | Ni-Cr-HC                   | 2                                                           | 3.0 + 1% Si   | 1275                                         | 1150                           | 125               |
| I              | A    | Ni-Cr-HC                   | 2                                                           | 3.6 + 1% Si   | 1225                                         | 1150                           | 75                |
| I              | B    | Ni-Cr-LC                   | 2                                                           | 2.5 + 1% Si   | 1325                                         | 1150                           | 175               |
| I              | C    | Ni-Cr-GB                   | Similar to the above irons (carbon is the dominant player). |               |                                              |                                |                   |
| I              | D    | Ni-Hi Cr                   | 8                                                           | 2.5 + 1.5% Si | 1325                                         | 1200                           | 125               |
| I              | D    | Ni-Hi Cr                   | 8                                                           | 3.6 + 1.5% Si | 1225                                         | 1200                           | 25                |
| II             | A    | 12% Cr                     | 12                                                          | 2.6           | 1340                                         | 1245                           | 95                |
| II             | B    | 15% Cr-Mo-LC               | 15                                                          | 2.6           | 1330                                         | 1255                           | 85                |
| II             | C    | 15% Cr-Mo-HC               | 15                                                          | 3.0           | 1300                                         | 1250                           | 50                |
| II             | C    | Cr-Mo-HC                   | 15                                                          | 3.6           | 1250                                         | 1240                           | 10                |
| II             | D    | 20% Cr-Mo-LC               | 20                                                          | 2.3           | 1350                                         | 1275                           | 75                |
| II             | E    | 20% Cr-Mo-HC               | 20                                                          | 3.0           | 1290                                         | 1260                           | 30                |
| III            | A    | 25% Cr                     | 25                                                          | 2.3           | 1340                                         | 1285                           | 55                |
| III            | A    | 25% Cr                     | 25                                                          | 3.0           | 1290                                         | 1270                           | 20                |

(1) Solidification temperatures are accurate to ±10°C.

(2) Designation: HC = High C; LC = Low C; GB = Great Britain.

## Effect of Fluidity on Solidification

As an iron solidifies, its viscosity changes dramatically from start to finish in a continuous fashion. At the solidus it is a very pliable solid having no elastic memory (only until the iron cools down to around 600C (1100F) do strains start locking in and do residual stresses form upon cooling to room temperature). How well an iron will flow (or its fluidity) is often confounded with the applied superheat. Indeed, a higher superheat will increase the fluidity of the metal. (See Appendix H.)

Because the property “fluidity” is of great interest for many foundry applications, its relationship to solidification will be covered in some detail. For example, it is well known that AR cast irons have superior fluidity when compared to cast steels, but exactly *why* is not well covered in the literature.

There is no clear definition of metallurgical fluidity. For our purposes, we can think of fluidity as how far a liquid metal will flow in a constricted channel. A high fluidity then implies that the liquid metal will flow exceedingly well through a constricted channel. One tool used in assessing this length measurement is the fluidity spiral. An example of a spiral mold is shown in Fig. 2.20. To facilitate the fluidity measurement, scribe marks are cut into the spiral’s pattern at regular intervals. After casting, the marks are counted and the fluidity length is computed.

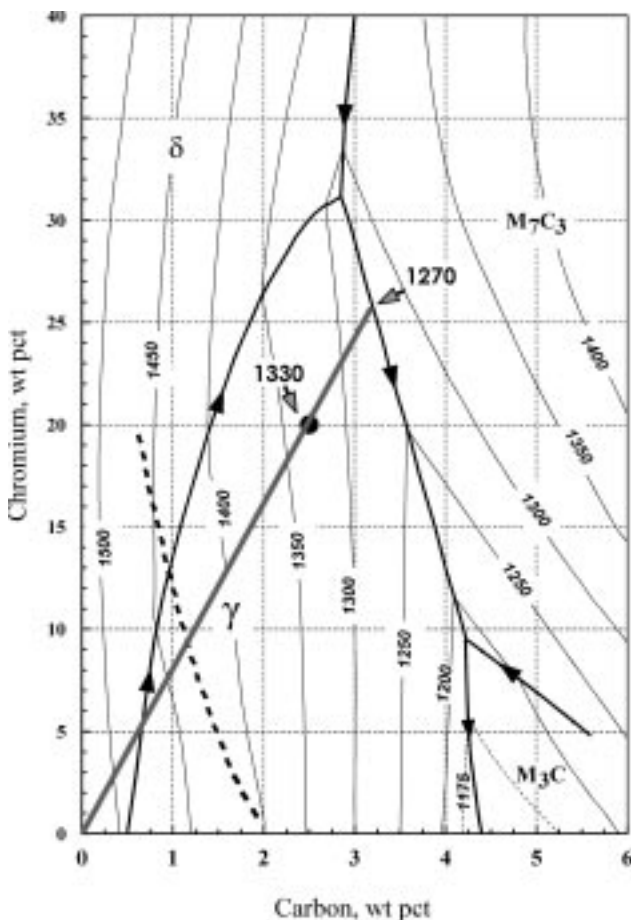


Fig. 2.19. Predicting solidification temperatures using the Fe-Cr-C liquidus surface.

Many factors affect the fluidity of a liquid metal. A short list includes the following:

- Alloy composition
- Superheat
- Melt cleanliness
- Sand-molding process

A brief discussion will cover the first three factors and, at this time, we can only speculate as to the effect of the molding processes.

## Alloy Composition

Solidification studies on fluidity have classed metal alloys into two broad groups; alloys that freeze over a very narrow (short) temperature range (pure metals and eutectics) and alloys that have broader (long) freezing ranges (hypoeutectic and hypereutectic). Figure 2.21 shows schematically how short- and long-freezing range alloys are believed to solidify in the fluidity spiral (or for that matter in any thin section of a casting).

If the alloy freezes within a very narrow temperature range, then the metal will tend to flow until it is choked off at some point, upstream of the advancing solidification front. This is a bit difficult to visualize and actually never happens as neat and clean as shown in Fig. 2.21a. As the metal flows through the spiral, it starts to solidify along the mold walls. However, since the alloy has a very short freezing range, this layer is kept very thin by the passage of hotter liquid as the front progresses through the spiral. Eventually, the metal cools enough for a localized buildup to occur behind the advancing solidification front (which, at this stage, is not advancing very fast). The

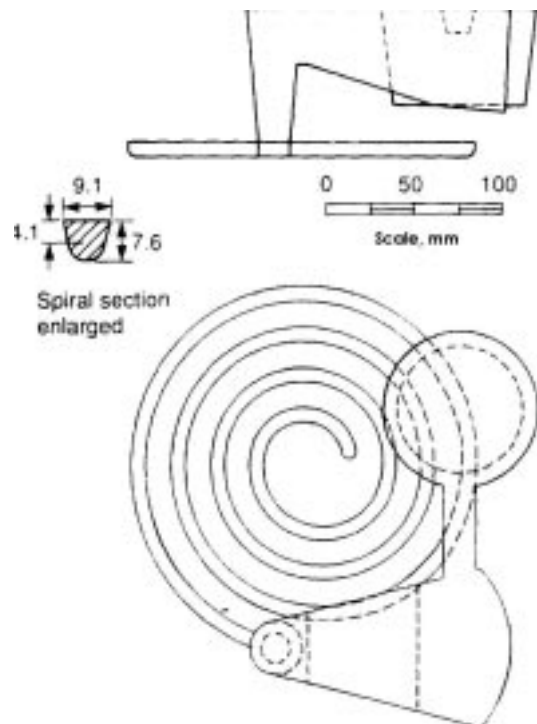
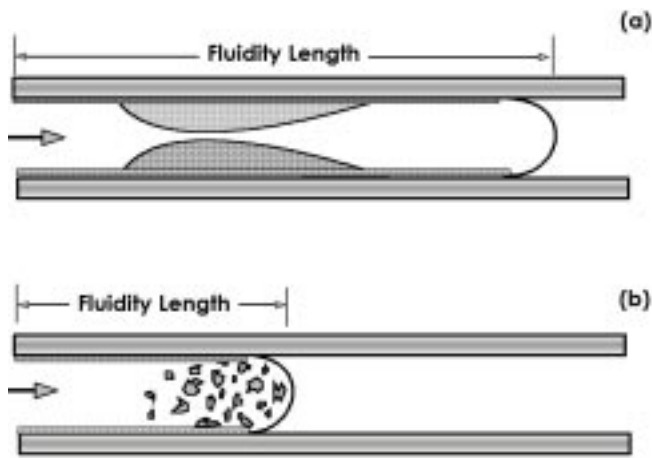


Fig. 2.20. Fluidity spiral for cast alloys. (After Campbell, 1991)



**Fig. 2.21. Idealized solidification patterns for (a) short-freezing range metals (relatively pure metals and eutectics) and (b) long-freezing range metals. In short-freezing range metals, the liquid metal flows unimpeded by early growing dendrites until the “exact” freezing point is hit. In contrast, long-freezing range metals start solidification early by forming many dendrites. These dendrites then break off the walls and tumble along the solidification front. Eventually, the metal will become sluggish and the front will stop advancing while much of the iron is still liquid.**

buildup is always a distance away from the advancing front, because new, hotter metal is continuously being supplied out to the front. However, this new, hot metal must first travel all the way from the pouring basin, through the spiral and then, finally, to the advancing front. As the flow of metal is choked off, a “shrink pipe” will start to form between the choke point and the now-slowly freezing solidification front.

For long-freezing range alloys (Fig. 2. 21b), the solidification behavior seems more logical. Long-freezing range implies that solidification occurs over a broad temperature interval. In this case, the formation of dendrites starts quite early in the solidification process. As the iron continues to freeze, the liquid becomes a viscous mixture of dendrites and liquid. In many cases, when the metal becomes about 50% solid, all fluid flow is pretty much choked off. That is, the dendrites act as a dense filter, hindering the flow of the liquid metal.

In solidification modeling, the fraction of solid that effectively stops the flow of molten metal is termed the critical fraction solid (CFS). For example, when the CFS is reached, the riser will no longer feed the casting; hence, long-freezing range alloys tend to choke off early. As a result, these alloys tend to cause greater amounts of porosity in castings than do other alloys.

To further demonstrate the concept of short- versus long-freezing range irons, Fig. 2.22 presents *differential thermal analysis (DTA) thermograms* (see Glossary) of two long-freezing range and two short-freezing range AR cast irons. For Ni-Hard irons (a) and (b), austenite dendrites start to form at about 75°C (135°F) above the eutectic solidification temperature. For these irons, the early formation of dendrites increases the viscosity of the iron and limits its fluidity well before the

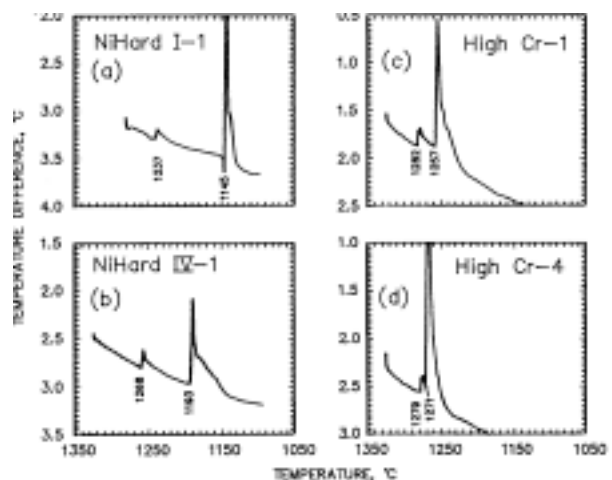
iron completely solidifies. In contrast, high-Cr irons (c) and (d) have short freezing ranges with the formation of austenite dendrites or carbides occurring almost immediately above the eutectic solidification temperature. Irons (c) and (d) tend to be much more fluid than irons (a) and (b), since the flow is unimpeded by the early formation of austenite dendrites.

The conclusion is clear, it is hard to beat the fluidity of a short-freezing range eutectic alloy. Whether the iron is short or long freezing is just a matter of the iron’s composition. In graphitic irons, silicon can quite effectively increase the fluidity of the iron by changing the iron’s composition. That is, the addition of Si can make a long-freezing range hypoeutectic iron more of a short-freezing range eutectic iron. For example, the addition of 2% Si to a 3.5% C graphitic iron will make the iron nearly eutectic and yield a significant increase in the iron’s fluidity.

In the case of AR irons, Si has practically no effect on the iron’s carbon equivalent ( $CE_{white} = \%C + 1/10\%Si$ ), and does very little in moving the fluidity of a long-freezing range hypoeutectic AR alloy toward that of a short-freezing range eutectic one. Consequently, we are left with a question as to the real effect of Si on the fluidity of AR cast irons.

In low-alloy AR irons, Si depresses the start of dendritic and eutectic solidification (see Fig. 2.2b for details of this effect). By keeping the pouring temperature constant, an increase in the alloy’s Si content effectively increases the alloy’s superheat and, thus, its fluidity.

In high-Cr AR irons, the use of Si to increase the alloy’s fluidity is kind of controversial. Chromium will tend to negate the effect of Si on the dendritic and eutectic solidification temperatures, and, as already discussed, Si has only a minor effect on the high-Cr alloy’s carbon equivalence. In practice, Si in the range 0.5–0.8% is believed to increase the fluidity of high-Cr irons by forming a “slippery” SiO<sub>2</sub> film on the surface of the liquid. How much of a fluidity boost this surface film



**Fig. 2.22. Differential thermal analysis (DTA) results for four AR cast irons. Irons (a) and (b) are considered long freezing and irons (c) and (d) are considered shorter freezing irons. Their compositions are: (a) 3.6% C, 3.2% Cr, 4.1% Ni and 1.0% Si; (b) 2.9% C, 8.4% Cr, 5.8% Ni and 1.8% Si; (c) 3.1% C and 15.3% Cr; and (d) 3.2% C and 28.8% Cr.**

provides is unknown, at this time, and is the subject of ongoing research, as reported in Appendix H.

The effect of superheat on fluidity is especially interesting to most foundrymen. For example, if two irons, having different compositions (say a 2.8% C, 25 % Cr iron and a 3.5% C, 15% Cr iron) are poured at the same temperature, they will have a 50°C (90°F) difference in superheat. This constitutes the source of many arguments about fluidity; that is, not clearly separating the effects of composition versus that of superheat.

This problem rears its ugly head whenever a new alloy is developed or whenever alloys are switched to suit different applications. When the application has thin sections or is borderline on the size of its riser (or more correctly its feeder), cold shuts or porosity, respectively, might develop when switching to the lower-C, higher-Cr iron (i.e., longer-freezing range alloy). It might just be assumed that the lower-C, higher-Cr iron has less fluidity. As we now know, this is not the case. Both irons have about the same fluidity. When we compare their positions on the Fe-Cr-C liquidus surface shown in Fig. 2.8, we see that they solidify with about the same amount of eutectic. However, if we wanted to pour both irons with the same superheat, the pouring temperature of the 2.8% C, 25% Cr alloy would need to be increased by 50°C (90°F) over that of the 3.5% C, 15% Cr alloy.

To summarize this discussion on the effect of alloy composition on fluidity, Fig. 2.23 indicates how composition drives the fluidity behavior of irons. Although this data was gathered for graphitic irons, the same behavior should hold for AR cast irons. Basically, at low and high C content, irons have a short freezing range, while at mid-range levels, irons have a long freezing range.

## Superheat

Figure 2.23 also shows the effect of increasing superheat (see Appendix G). As the superheat is boosted, the maximum length the iron can travel within the spiral is increased. This is what we would expect; higher superheat should allow the metal to travel further within the mold. It might be reasoned that the logical thing to do would be to cast all our irons at high superheat and just forget about a specific fluidity for each iron.

Unfortunately, there are negative aspects to using high pouring temperatures. Besides the increased cost (due to energy and refractory use), high pouring temperatures can degrade the surface finish of the casting by fusing the metal onto the sand. It also has limited effectiveness, since the mold will simply absorb much of this added superheat. That is, as you increase the temperature of the metal, heat transfer from the metal to the mold is also increased, negating some of the benefits of higher superheat. The higher heat transfer rate at higher melting temperatures explains why low-carbon cast steels have lower “practical” fluidity than near-eutectic irons having lower melting temperatures. They both have short freezing ranges, but heat is extracted at a higher rate for the steel, which has to be cast at much higher temperatures.

Hence, if one carefully separates the effects of superheat and chemical composition, it is possible to switch alloys with very little worry about forming cold shuts or introducing porosity in what was a sound casting. It is all just a matter of careful verification of the alloy’s true solidification temperature (i.e., the true superheat) and a good understanding of whether the alloy is long freezing (hypoeutectic or hypereutectic) or short freezing (eutectic).

## Melt Cleanliness

An interesting side note about fluidity is that it is highly susceptible to melt cleanliness. A dirty melt having lots of oxides and inclusions also has lots of nucleation sites to initiate solidification. Given two melts having the same composition, the melt having a greater number of suspended oxides and other inert particles will have lower fluidity due to earlier nucleation and growth. Conversely, Powell (1990) has shown that extremely clean melts will not nucleate and solidify until they reach temperatures well below that commonly thought of as the equilibrium freezing temperature. When these ultra pure melts do start to nucleate and grow, they are so undercooled that solidification is extremely rapid and the microstructure is extremely fine (e.g., similar to that seen near a chilled surface).

This knowledge is only now being used in the gray and AR cast iron foundries by the introduction of inert gas melting practices and by the effort to use cleaner charge materials. With a cleaner melt, the iron will solidify more uniformly and more aggressively, resulting in much greater fluidity over that of a dirty melt.

In summary, it is entirely possible to have excellent control over the irons’ fluidity by matching pouring superheat with the iron’s composition. For starters, Table 2.1 and Fig. 2.8 provide the necessary data to start your calculations. If more on this topic is of interest, the reader is encouraged to look at Campbell (1991).

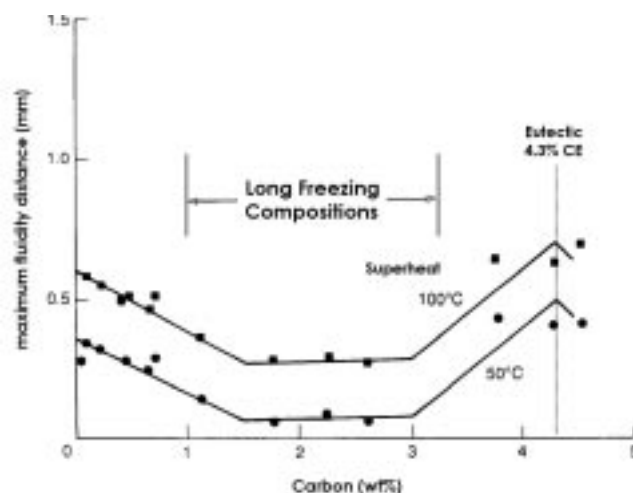


Fig. 2.23. Fluidity as a function of superheat and composition for Fe-C irons. (After Campbell, 1991)

## Structures Formed During Solidification

We now know how AR cast irons solidify, the various temperatures at which they solidify and how they flow during solidification. We also know a little bit about their microstructure (e.g., how much carbide will form on solidification). Our next step is to understand the structures that form during solidification. As a first step, we will take a look at the cast macrostructure of several high-Cr irons shown in Fig. 2.24. (Later on, we will discuss *microstructure*.)

### Macrostructure

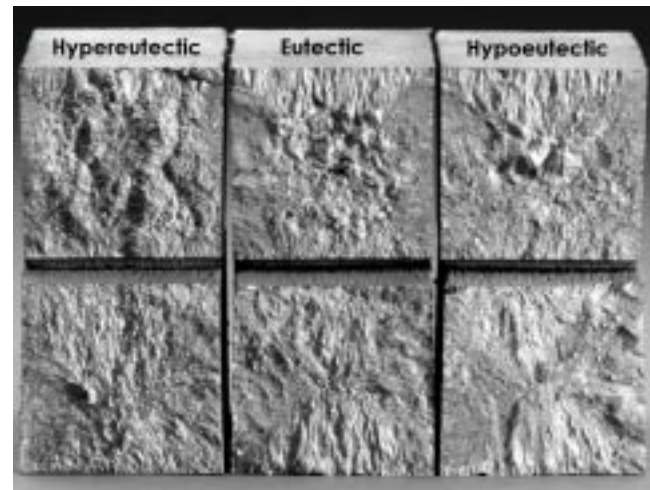
When talking about cast macrostructure, we are referring to the overall cast structure that becomes visible when the iron is fractured or sectioned. In general terms, a cast macrostructure will be *columnar and/or equiaxed* (see Glossary) in appearance. In Fig. 2.24, the eutectic iron poured using a low superheat temperature (upper middle) has columnar grains growing from the surface, with a transition to equiaxed grains near the center of the casting. At higher superheat, the same eutectic iron (lower middle) has a macrostructure that is dominated by columnar grains. This effect was found to be fairly uniform for both hypoeutectic and hypereutectic irons.

Besides superheat, alloy composition influences the growth of columnar and/or equiaxed grains. As the iron becomes more hypereutectic, the growth of equiaxed grains is favored over that of columnar grains. This can be explained by the way carbides and austenite grow during solidification. Large amounts of Cr are required to form  $M_7C_3$  and, hence, these carbides grow very slowly when compared to the growth rate of austenite. Consequently, in hypereutectic irons (where growth is dominated by carbides), growth from the mold walls is not as rapid as in hypoeutectic irons (where growth is dominated by austenite dendrites). Therefore, equiaxed grains have a better chance to nucleate and grow in hypereutectic irons. As might be expected, when the superheat is high, you will get columnar grains, whether you like it or not, since the temperature gradient between the mold and the liquid metal will be quite steep. If this subject fascinates the reader they should look at Ohno (1987) and the other references at the end of this chapter.

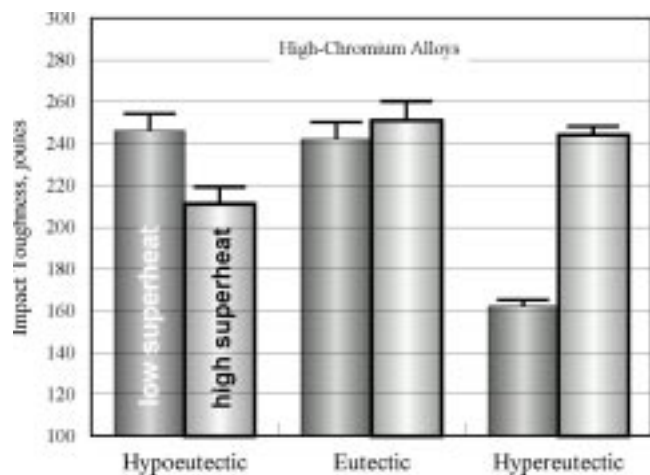
Typically, the cast macrostructure is not considered a very important issue. When specifying a pouring temperature, consideration is mainly given to figuring out the correct temperature for properly filling the mold, obtaining a good surface finish and conserving refractory and energy. In the past, when many foundries were happy just to produce a sound product, this may have been completely adequate. However, controlling cast macrostructure can yield important benefits for the foundry willing to give more thought to their pouring superheats.

Figure 2.25 provides data as to the effect of cast macrostructure versus superheat for hypoeutectic, eutectic and hy-

pereutectic high-Cr AR cast irons (the same ones shown in Fig. 2.24, except in reverse order). For hypoeutectic irons, we follow the predictable pattern of where a fine equiaxed structure gives us a 10–15% boost in impact toughness. For eutectic irons, it is anybody's guess as to the effectiveness of controlling superheat. Most interesting is the fact that, for hypereutectic irons, the columnar structure, produced by high superheat, yields a 30–35% improvement in toughness. The reason this occurs will be covered in Chapter 5, Morphology, etc. In the meantime, the gist of the message is that controlling superheat and, therefore, cast macrostructure is an important option for the foundryman.



**Fig. 2.24.** Examples of cast macrostructure in high-Cr AR cast irons. The top row of bars were cast at low superheat (50C) while the bottom row of bars were cast at high superheat (150C). Each bar is approximately 5x5x18 cm and was fractured in a drop-weight impact test. The nominal compositions of the bars are: Hypereutectic–3.7% C, 26% Cr; Eutectic–3.1% C, 26% Cr; and Hypoeutectic–2.8% C, 26% Cr.



**Fig. 2.25.** Single-blow impact data on 5x5x18-cm cast bars showing the effect of casting superheat on impact toughness. (After Laird and Dogan, 1996)

## Microstructure

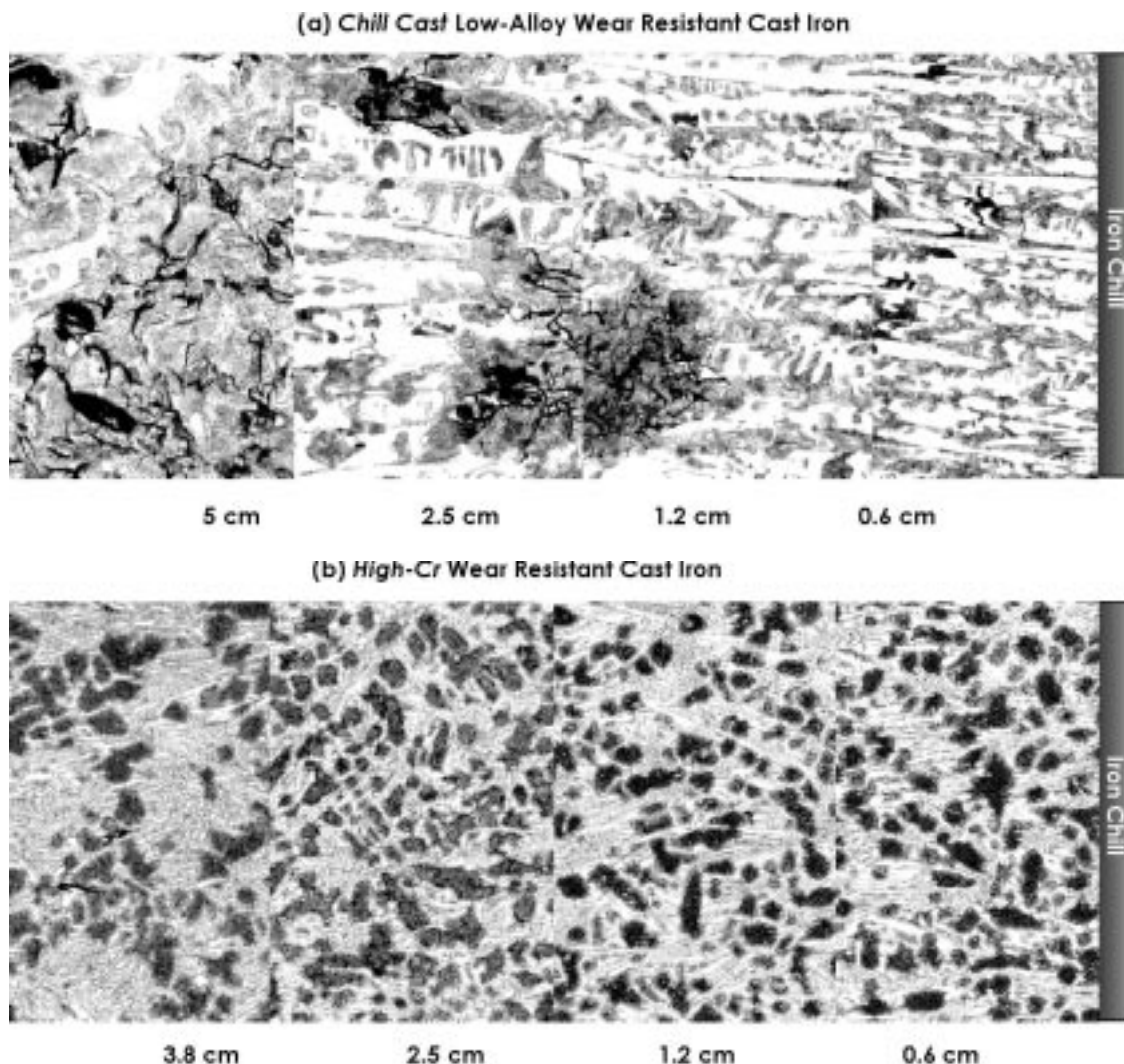
The rate of solidification is well known to have distinct effects on the cast iron's microstructure. If the iron is cooled quickly (which generates a high solidification rate), we see a refinement of the microstructure, whether it is smaller austenite dendrites or smaller carbides (primary or eutectic, it does not matter). This refinement has been documented by a number of people (see Matsubara et al. and Ogi et al., 1981, Laird and Dogan, 1996 or Powell, 1990) and results from a preference for nucleation over growth when the freezing rate increases.

As was mentioned in the first chapter, a high freezing rate did bring the discovery of "white cast irons." Foundries are still banking on this discovery when they use chills to create carbides at specific locations. Figure 2.26a shows the chill cast microstructure of a Ni-Hard-type AR iron. By carefully controlling the C, Si, Ni and Cr contents, the depth of chill can be metallurgically engineered to meet the conflicting demands of toughness and wear resistance. Although this sounds quite simple, definitive data do not exist between C, Si, Ni and

Cr to allow accurate predictions of chill depth as a function of chemical composition. Moreover, as would be expected, other elements influence the depth of chill. The reader is referred to work by Boyes and Fuller (1964) as a good starting point for their studies on controlling chill depth in low-alloy AR cast irons. For further reference Table 2.2 is provided.

Figure 2.26a also indicates that the  $M_3C$  carbides are quite long but isolated or discrete, as compared to slower cooled, sand cast Ni-Hard (see Fig. 1.1). The isolated, parallel nature of the carbides shown in Fig. 2.26a is not the result of alloying but of the much higher solidification rate produced by the iron chill and, it should not be overlooked, a low superheat temperature. This is quite advantageous, since it is much harder for a crack to propagate across dendrites than along a near-continuous path of brittle carbides, represented by the sand cast iron shown in Fig. 1.1.

Figure 2.26b shows a similar microstructural refinement in a high-Cr AR cast iron. The microstructure immediately



**Fig. 2.26.** Examples of chill cast microstructures showing the effect of the solidification rate on the cast microstructure. All photomicrographs were taken on a polished plane, perpendicular to the chill surface, at 100X. (Irons courtesy of United Foundries, Inc.)

adjacent to the chill face is composed of fine, rod- and blade-shaped  $M_7C_3$  carbides and small proeutectic austenite dendrites. Farther away from the chill surface, the austenite dendrites become larger and the carbide morphology becomes more blade-like. This transition, from a microstructure dominated by a fine dendritic structure to that of a much coarser structure as a function of solidification rate, is found in the literature, but perhaps is not well known.

Another interesting relationship between cast microstructure and solidification rate is the carbide morphology found in the eutectic cell. A good example of this morphology is shown in Fig. 2.18c, where near the center of the colony we have a few rod-like carbides with much larger, blade-like carbides radiating outward. This carbide morphology was explained by Dogan and Laird (1997).

**Table 2.2**  
**Relative Effects of Alloying Elements on**  
**Chilling Tendency of Hypoeutectic AR Cast Irons**

| Addition of<br>0.1% Alloy to<br>Hypoeutectic<br>Base Iron           | Change in % <sup>(1)</sup> |                               | Applicable<br>Alloying<br>Range<br>(Wt%) | Relative<br>Influence<br>Factor Based<br>on Si = +1.0 <sup>(2)</sup> |
|---------------------------------------------------------------------|----------------------------|-------------------------------|------------------------------------------|----------------------------------------------------------------------|
|                                                                     | 100%<br>Carbide<br>Zone    | Carbide +<br>Graphite<br>Zone |                                          |                                                                      |
| Silicon                                                             | -18                        | -18                           | 0.4–1.7                                  | +1.0                                                                 |
| Carbon                                                              | -6                         | -7.5                          | 2.4–3.7 CE                               | n.a.                                                                 |
| Phosphorus                                                          | -7                         | -4                            | 0.0–1.2                                  | n.a.                                                                 |
| Manganese <sup>(3)</sup><br>(in excess of<br>sulfur<br>requirement) | +6                         | +6                            | 0.0–3.0                                  | -0.25                                                                |
| Sulfur <sup>(3)</sup><br>(see above)                                | +28                        | +28                           | 0.008–0.33                               | n.a.                                                                 |
| Nickel                                                              | -4.5                       | -4.5                          | 0.0–5.3                                  | +0.3–0.4                                                             |
| Copper                                                              | -6.3                       | n.a.                          | 0.0–4.3                                  | +0.2–0.35                                                            |
| Aluminum                                                            | -55                        | -55                           | 0.0–0.18                                 | n.a.                                                                 |
| Titanium                                                            | -8.5                       | n.a.                          | 0.0–0.85                                 | n.a.                                                                 |
| Boron                                                               | +60                        | n.a.                          | 0.0–0.20                                 | n.a.                                                                 |
| Chromium                                                            | +16                        | +16                           | 0.0–3.0                                  | -1.0                                                                 |
| Molybdenum                                                          | +8                         | +8                            | 0.0–2.0                                  | -0.35                                                                |
| Vanadium                                                            | +20                        | +20                           | 0.0–0.7                                  | -2.0                                                                 |
| Tin                                                                 | +12.5                      | n.a.                          | n.a.                                     | n.a.                                                                 |

(1) Bars, 1.125-in. thick, 5-in. wide and 9-in. high, were cast against a large chill. The change in chill depth was measured for each alloy addition; please note alloy ranges that were used in the experimental investigation. A positive sign (+) indicates that the depth of the carbide formation increases with the alloying addition while a negative sign (–) indicates that the chill depth decreases. The data came from Boyes and Fuller (1964).

(2) Positive (+) means that it acts in the same way as Si, while a negative sign (–) means it acts in an opposite manner to Si. The source for this data can be found in Boyes and Fuller (1964).

(3) Manganese and sulfur are linked because they form MnS.

n.a. not available.

As the eutectic cell grows (it represents one nucleation event and hence grows cooperatively), the solidification rate varies within the eutectic cell. As the carbides within the core of the eutectic cell release heat, the solidification rate slows around these initial rod-like carbides. The lower solidification rate then favors the growth of blade-like carbides surrounding the rod-like carbides at the center of the eutectic colony.

Sometimes, there is a “hole” found within the rod-like carbides. This hole is real and is filled with austenite, albeit, austenite that is highly depleted in Cr and C. Why this hole exists was explained by Pearce (1984) in his very thoughtful and practical paper on AR cast irons. Pearce makes a good argument—using transmission electron microscopy (TEM) of the carbide microstructure—that the hexagonal structure (what we are calling rod-like) initially freezes as a hexagonal shell, which then grows inward, forming a shrinkage cavity. Although this mechanism is difficult to visualize, and fails to explain the lack of holes within blade-like carbides, it is the only scientific mechanism that has been offered, to this date.

## Alloy Segregation

The groundwork has been laid as to what happens during the solidification of AR cast irons. It is known that, during the solidification event, molten liquid freezes into two solid forms: a metallic phase (austenite or ferrite) and a carbide phase ( $M_3C$  or  $M_7C_3$ ). Using the phase diagrams, one can also make reasonable estimates as to the C and Cr contents of the ferrous and carbide phases and their relative volume fractions. Also, there should be a good understanding of the cast solidification macrostructure and microstructure.

In this chapter, the emphasis has been on the C and Cr contents of the iron. For example, whether the iron forms  $M_3C$  or  $M_7C_3$  carbides is mainly controlled by the Cr content. Likewise, whether the iron becomes *hypoeutectic* or *hypereutectic* is mainly controlled by *both* the C and Cr contents. This section will expand on this basis and bring into focus the important effects of Si and Ni on pearlitic, Ni-Hard 1, Ni-Hard 2 and Ni-Hard 4 cast irons, and the general effects of Si, Ni, Mo, Mn and Cu on the solidification behavior of AR cast irons.

When reviewing specifications for AR cast irons, these other alloying elements often have distinct ranges. For the most part, these elements are only added for *hardenability* (see Chapter 4, Alloying for Hardenability). For example, when casting a 3% C, 15% Cr iron, the ASTM A532 specification (see Appendix E, Standards, Table E7) calls for an Mo level of between 2.3 and 3.5%. If the casting is thin (under 25 mm) and heat treated under the right conditions, it is entirely possible to obtain a hard AR iron using 1.0% Mo. Although the iron would not meet “specification,” it would be essentially identical to other thicker castings containing between 2.3 and 3.5% Mo. A cautionary note, however: the trouble with lean alloying (e.g., using low levels of Ni and Mo) is that you do not realize that you have a problem until the castings

have been heat treated and are ready to be shipped to the customer. (More information on lean alloying can be found in Chapter 4, Alloying for Hardenability.)

An important concern when adding alloying elements to AR cast irons is where the alloying elements will end up: in the carbide or in the metallic matrix. Once an alloying element is absorbed by the carbide, it no longer has any effect on the iron's hardenability. It means that the hardenability of the metallic matrix is determined by the alloying elements in the matrix, and not by those parts of the alloying elements locked within the carbides.

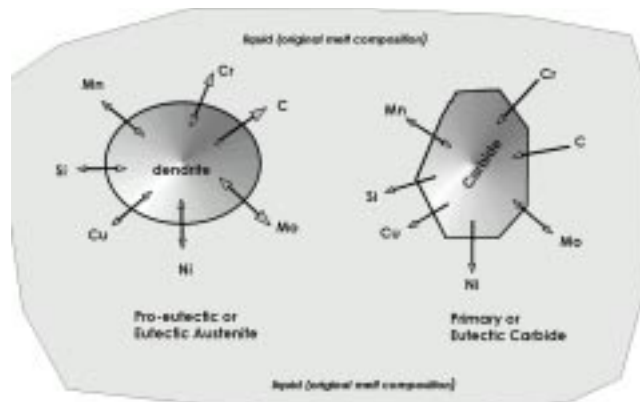
As a start, Fig. 2.27 provides a schematic of how the various hardenability elements (which naturally include C and Cr) segregate or partition during solidification. It should also be kept in mind that the partition behavior is also a matter of solidification rate and, once the iron is solid, the time available for diffusion. It is a complex subject.

During austenite formation (proeutectic or eutectic), the growing austenite dendrite will tend to accept or reject certain alloying elements. As we know from previous text on the solidification behavior of Fe-Cr-C irons, the growing austenite dendrite will readily reject or segregate C and Cr into the liquid. The dendrite will also reject Mo into the liquid, and, for that matter, other strong carbide-forming elements, like vanadium (V), tungsten (W) and niobium (Nb). The austenite dendrite will also reject chill-inducing elements, such as P and S (and other trace elements, such as Bi, Te and Mg).

Manganese, since it is a weak carbide former, is not strongly rejected by the growing austenite dendrite. At the 1% melt level (an Mn content that is often found in AR cast iron compositions), it is almost evenly divided between the dendrites and the carbides. We find the same behavior for the hardenability elements of Si, Cu and Ni. During solidification of the dendrites, there is little rejection of these elements.

The exact opposite takes place during the formation of the carbide phase (be it the low-Cr  $M_3C$  or the high-Cr  $M_7C_3$  carbide). Silicon, Cu and Ni are completely rejected into the liquid. In other words, Si is insoluble in the carbide phase, while the solubility of Cu and Ni is very, very limited. For all practical purposes, Si, Cu and Ni are just taken to be insoluble in the carbide phase.

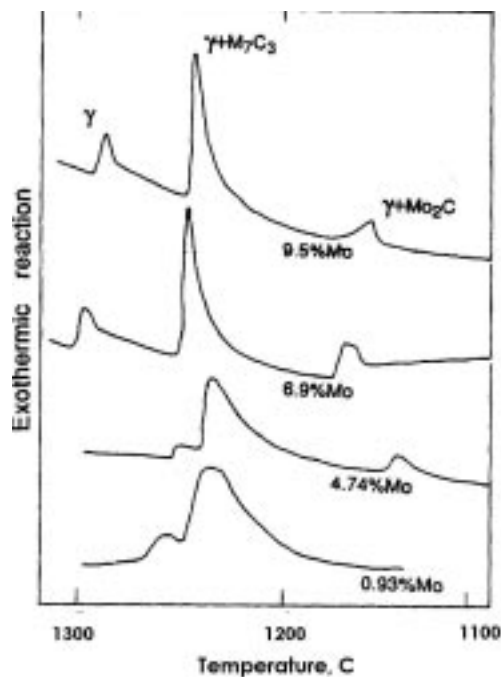
As can be expected, C, Cr and, to some extent, Mo and V will be directly absorbed by the growing carbide. The reason we say "to some extent Mo," is that Mo also tends to form its own eutectic carbides, given half a chance. At Mo levels over 2%, one can find  $Mo_2C$  carbides dotted along the austenite and eutectic grain boundaries. At even higher levels, say from 3 to 5% Mo, the presence of Mo carbides is quite apparent along those boundaries. The reason  $Mo_2C$  carbides are located there is that they solidify during a separate, lower-temperature eutectic reaction. Figure 2.28 shows a few DTA thermograms of a 3% C, 20% Cr AR cast iron with different Mo contents. The upper peaks shown in these DTA thermograms represent the eutectic reaction  $L \Rightarrow \gamma + Mo_2C$ .



**Fig. 2.27. Schematic of approximate segregation behavior of hardenability elements during solidification.**

Manganese is found in both the  $M_3C$  and  $M_7C_3$  carbides; much as it is found in the austenite dendrite. The Mn content of the melt is the same one found in both the dendrites and the carbides.

Table 2.3 provides experimental data concerning the partition behavior of alloying elements during solidification. The numbers given in Table 2.3 represent the partition ratio of a particular element between the carbide and the matrix (sometimes, the partition ratio is also termed the partition coefficient). The microchemistries provided in this table are from Laird (1991) and were obtained using an *electron microprobe* (see Glossary).



**Fig. 2.28. DTA thermograms of a 3% C, 20% Cr AR cast iron containing varying amounts of Mo. They readily demonstrate that the formation of  $Mo_2C$  occurs at the very end of the solidification process. As can be imagined, the only remaining place for the  $Mo_2C$  to form is at the grain boundaries of the eutectic cells, where it can lower the fracture toughness of the iron. (Courtesy of M. Ikeda)**

(Alloy Segregation, continued)

Besides the partition information presented in Table 2.3, it is useful to spend a few minutes on the chemical information. For each iron, the original melt composition is given, followed by the microanalysis information for the carbide and matrix phases. These numbers provide a definitive measure of where the alloying elements (including C and Cr) end up at the end of solidification.

One of the interesting things to notice in Table 2.3 is that, as the Cr content of the iron increases, the Cr content in the matrix similarly increases (but perhaps not as fast as one might expect). The amount of Cr contained within the matrix is not only a simple function of the melt's original melt composition but also of the melt's C content.

Based on knowledge of the phase diagram, one can understand that, as the C content increases, the amount of carbides

will also increase. Any formation of carbides will consume a large amount of Cr and, hence, lower the amount of Cr that will be available to the matrix. This is evident in Table 2.3 when examining the first three high-Cr alloys. Moreover, this relationship is borne out by the phase diagram if one plots the iron's composition on Fig. 2.14 and draws a tie line through the origin. If the C content is increased while holding the Cr content constant, a quick investigation of Fig. 2.14 will show that the Cr content of the eutectic austenite decreases.

It should not be overlooked that Cr is a very powerful "hardenability" element and is the dominant reason that high-Cr AR irons are often partially austenitic in the as-cast state. Also, when the Cr content of the matrix is over 11%, the iron will have significant corrosion resistance (see Chapter 9). Thus, it is more than just of academic interest to understand the partitioning behavior of Cr within the iron if good corrosion resistance is of concern.

**Table 2.3**  
**Alloy Segregation (Partition) During Solidification of AR Cast Irons**

|                                                                                                                                                                            | Composition (Wt%) |                    |      |     |     |     |      |      | Total | Stoich. <sup>(1)</sup>           | Partition Ratios = Carbide% / Matrix% |      |      |     |      |      |      |
|----------------------------------------------------------------------------------------------------------------------------------------------------------------------------|-------------------|--------------------|------|-----|-----|-----|------|------|-------|----------------------------------|---------------------------------------|------|------|-----|------|------|------|
|                                                                                                                                                                            | Cr                | C                  | Fe   | Ni  | Si  | Mn  | Mo   | Cu   |       |                                  | Cr                                    | C    | Ni   | Si  | Mn   | Mo   | Cu   |
| <b>Ni-Hard AR Cast Iron or ASTM IA:</b>                                                                                                                                    |                   |                    |      |     |     |     |      |      |       |                                  |                                       |      |      |     |      |      |      |
| Melt                                                                                                                                                                       | 3.2               | 3.6                | n.a. | 4.1 | 1.0 | 0.5 | 0.0  | 0.0  |       |                                  |                                       |      |      |     |      |      |      |
| Carbide                                                                                                                                                                    | 5.8               | 6.1                | 86.6 | 1.4 | 0.0 | 0.4 | 0.0  | 0.0  | 100.5 | M <sub>77</sub> C <sub>23</sub>  | 4.8                                   | 5.5  | 0.23 | 0.0 | 1.3  | n.a. | n.a. |
| Matrix                                                                                                                                                                     | 1.2               | 1.1 <sup>(2)</sup> | 89.8 | 6.1 | 2.0 | 0.3 | 0.0  | 0.0  | 100.7 | (M <sub>3</sub> C)               |                                       |      |      |     |      |      |      |
| <b>Ni-Hard 4 AR Cast Iron or ASTM ID:</b>                                                                                                                                  |                   |                    |      |     |     |     |      |      |       |                                  |                                       |      |      |     |      |      |      |
| Melt                                                                                                                                                                       | 8.4               | 2.9                | n.a. | 5.8 | 1.8 | 1.0 | 0.0  | 0.0  |       |                                  |                                       |      |      |     |      |      |      |
| Carbide                                                                                                                                                                    | 30.2              | 8.8                | 60.2 | 0.7 | 0.0 | 1.3 | n.a. | n.a. | 101.3 | M <sub>70</sub> C <sub>30</sub>  | 8.9                                   | 6.8  | 0.11 | 0.0 | 1.6  | n.a. | n.a. |
| Matrix                                                                                                                                                                     | 3.4               | 1.3 <sup>(2)</sup> | 85.2 | 6.9 | 2.5 | 0.9 | n.a. | n.a. | 100.3 | (M <sub>7</sub> C <sub>3</sub> ) |                                       |      |      |     |      |      |      |
| <b>Ni-Hard 4 AR Cast Iron or ASTM ID:</b>                                                                                                                                  |                   |                    |      |     |     |     |      |      |       |                                  |                                       |      |      |     |      |      |      |
| Melt                                                                                                                                                                       | 8.0               | 3.4                | n.a. | 5.8 | 2.2 | 1.1 | 0.0  | 0.0  |       |                                  |                                       |      |      |     |      |      |      |
| Carbide                                                                                                                                                                    | 24.5              | 8.4                | 65.5 | 0.7 | 0.0 | 1.3 | n.a. | n.a. | 100.6 | M <sub>71</sub> C <sub>29</sub>  | 9.4                                   | 8.4  | 0.09 | 0.0 | 1.6  | n.a. | n.a. |
| Matrix                                                                                                                                                                     | 2.6               | 1.0 <sup>(2)</sup> | 86.5 | 7.4 | 3.2 | 0.8 | n.a. | n.a. | 101.6 | (M <sub>7</sub> C <sub>3</sub> ) |                                       |      |      |     |      |      |      |
| <b>High-Cr AR Cast Iron or ASTM IIB:</b>                                                                                                                                   |                   |                    |      |     |     |     |      |      |       |                                  |                                       |      |      |     |      |      |      |
| Melt                                                                                                                                                                       | 18.4              | 2.2                | n.a. | 0.4 | 0.3 | 0.5 | 0.0  | 0.0  |       |                                  |                                       |      |      |     |      |      |      |
| Carbide                                                                                                                                                                    | 50.2              | 8.7                | 41.0 | 0.0 | 0.0 | 0.7 | 0.1  | 0.0  | 100.4 | M <sub>70</sub> C <sub>30</sub>  | 4.3                                   | 8.7  | 0.0  | 0.0 | 1.2  | n.a. | 0.0  |
| Matrix                                                                                                                                                                     | 11.5              | 1.0 <sup>(2)</sup> | 85.4 | 0.6 | 0.7 | 0.6 | 0.0  | 0.1  | 100.2 | (M <sub>7</sub> C <sub>3</sub> ) |                                       |      |      |     |      |      |      |
| <b>High-Cr AR Cast Iron or ASTM IIC:</b>                                                                                                                                   |                   |                    |      |     |     |     |      |      |       |                                  |                                       |      |      |     |      |      |      |
| Melt                                                                                                                                                                       | 15.3              | 3.1                | n.a. | 0.4 | 0.4 | 0.6 | 0.9  | 0.8  |       |                                  |                                       |      |      |     |      |      |      |
| Carbide                                                                                                                                                                    | 45.5              | 8.4                | 44.2 | 0.0 | 0.0 | 0.6 | 1.6  | 0.0  | 100.4 | M <sub>71</sub> C <sub>29</sub>  | 4.8                                   | 7.6  | 0.0  | 0.0 | 1.0  | 3.2  | 0.0  |
| Matrix                                                                                                                                                                     | 9.5               | 1.1 <sup>(2)</sup> | 86.3 | 0.5 | 0.5 | 0.6 | 0.5  | 1.3  | 100.2 | (M <sub>7</sub> C <sub>3</sub> ) |                                       |      |      |     |      |      |      |
| <b>High-Cr AR Cast Iron or ASTM IIE:</b>                                                                                                                                   |                   |                    |      |     |     |     |      |      |       |                                  |                                       |      |      |     |      |      |      |
| Melt                                                                                                                                                                       | 20.2              | 3.0                | n.a. | 0.4 | 0.5 | 0.7 | 1.9  | 0.8  |       |                                  |                                       |      |      |     |      |      |      |
| Carbide                                                                                                                                                                    | 53.0              | 8.6                | 36.1 | 0.0 | 0.0 | 0.7 | 1.8  | 0.0  | 100.2 | M <sub>70</sub> C <sub>30</sub>  | 4.5                                   | 8.6  | 0.0  | 0.0 | 1.0  | 2.2  | 0.0  |
| Matrix                                                                                                                                                                     | 11.7              | 1.0 <sup>(2)</sup> | 82.4 | 0.6 | 0.8 | 0.7 | 0.8  | 1.5  | 99.6  | (M <sub>7</sub> C <sub>3</sub> ) |                                       |      |      |     |      |      |      |
| <b>High-Cr AR Cast Iron or ASTM IIIA:</b>                                                                                                                                  |                   |                    |      |     |     |     |      |      |       |                                  |                                       |      |      |     |      |      |      |
| Melt                                                                                                                                                                       | 28.8              | 3.2                | n.a. | 0.3 | 0.2 | 0.5 | 0.5  | 0.1  |       |                                  |                                       |      |      |     |      |      |      |
| Carbide                                                                                                                                                                    | 60.9              | 8.6                | 31.3 | 0.0 | 0.0 | 0.4 | 0.5  | 0.0  | 101.5 | M <sub>71</sub> C <sub>29</sub>  | 4.6                                   | 14.3 | 0.0  | 0.0 | 0.80 | n.a. | 0.0  |
| Matrix                                                                                                                                                                     | 13.3              | 0.6 <sup>(2)</sup> | 85.0 | 0.4 | 0.6 | 0.5 | 0.3  | 0.1  | 101.0 | (M <sub>7</sub> C <sub>3</sub> ) |                                       |      |      |     |      |      |      |
| <i>(1) Stoichiometry of the carbide (i.e., M<sub>7</sub>C<sub>3</sub> = M<sub>70</sub>C<sub>30</sub>) is based on exact microprobe data and ( ) shows rounded numbers.</i> |                   |                    |      |     |     |     |      |      |       |                                  |                                       |      |      |     |      |      |      |
| <i>(2) Matrix C concentrations were referenced against a low-C standard (NBS 160A, 19% Cr, 14% Ni, 3% Mo, 0.06% C steel).</i>                                              |                   |                    |      |     |     |     |      |      |       |                                  |                                       |      |      |     |      |      |      |
| <i>n.a. not applicable or not analyzed.</i>                                                                                                                                |                   |                    |      |     |     |     |      |      |       |                                  |                                       |      |      |     |      |      |      |

Table 2.3 confirms the phase diagram compositional predictions for a 3% C, 15% Cr iron, showing a matrix C content of 1.1% (which has an error of  $\pm 0.2\%$ ). For other irons, the measured C content within the matrix phase is in accord with the phase diagram predictions concerning eutectic austenite (i.e., as the matrix Cr content increases the C content decreases).

The data of Table 2.3 further confirm that Si and Cu are completely rejected from the carbide phase. Nickel is also largely rejected by the carbide phase. This is useful information for the foundryman thinking about hardenability, since any addition of Si, Cu and Ni will be magnified by the amount of carbides that are formed during solidification. For example, in case we are interested in adding 1% Ni to a 3% C, 20% Cr AR cast iron, the actual matrix content can be calculated as follows:

$$\begin{aligned} \text{Ni}_{\text{Matrix}} &= \text{Ni}_{\text{Melt}} / (1 - \text{CVF} / 100) \\ \text{Ni}_{\text{Matrix}} &= 1\% / (1 - 33 / 100) \\ \text{Ni}_{\text{Matrix}} &= 1.5\% \end{aligned} \quad (2.6)$$

The same approach can be used to calculate the actual amount of Si and Cu that is present in the matrix.

Silicon plays a diverse role in many AR cast iron's. For example, many high-Cr irons have a nominal Si level of 0.6%. As an alloying element, Si imparts little hardenability; contrarily, it does promote the formation of pearlite, and thus lowers the hardenability of the iron. Consequently, the main reason for adding Si to a high-Cr iron is to control the melt oxidation and increase fluidity. Silicon will form a thin tenacious layer of  $\text{SiO}_2$  on the surface of the melt. This thin layer will hinder the formation of  $\text{Cr}_2\text{O}_3$ , thereby providing some measure of protection against further oxidation. Silicon is also reported to make furnace slag less viscous and easier to handle on removal, prior to tapping.

As an alloying element, Si separates fully to the matrix, so even limited additions of Si can have unwanted consequences. For example, if a 3% C, 25% Cr iron is alloyed with 0.6% Si, the real matrix content is actually around 0.9%, assuming an approximate carbide volume fraction (CVF) of 35%. This multiplier effect explains why Si can have dangerous consequences.

Unfortunately (or fortunately), the other hardenability elements do not behave in the same way as Si, Cu or Ni. As shown in Table 2.3 and illustrated in Fig. 2.27, Mn is equally distributed between the carbide and matrix and poses no problems for our hardenability calculations. However, one of the more frequently used hardenability elements, Mo, has a more complicated behavior since it partitions or segregates between the austenite dendrite and the  $\text{M}_7\text{C}_3$  carbide phase and may form its own carbide phase ( $\text{Mo}_2\text{C}$ ). It has been estimated that only about one third of the original melt addition of Mo ends up in the matrix, where it affects the hardenability of the iron. Hence, if you add 2% into the melt, you will be lucky to obtain the same hardenability effect as that of 0.8% Mo in a 1% C steel.

In general, one should think of a AR cast iron as having two components: a heat treatable matrix and a network of carbides (whether  $\text{M}_3\text{C}$ ,  $\text{M}_7\text{C}_3$ ,  $\text{Mo}_2\text{C}$  or whatever). If one focuses on the matrix compositions given in Table 2.3, it becomes obvious that we are basically heat treating a special class of high-C tool steel. For example, a common tool steel is D2, which contains approximately 1.4% C, 12% Cr, 0.8% Mo, 0.3% Si and perhaps 0.25% V for grain size control. (Alternatively, the H11 composition of 0.4% C, 5% Cr, 1% Mo and 0.9% Si is also a good fit for many AR cast iron matrix compositions.) The D2 composition is fairly close to the matrix composition of alloy ASTM IIE shown in Table 2.3.

Foundrymen interested in heat treating a 3% C, 20% Cr iron will learn a lot from the literature on heat treating D2 (and, likewise, many of the other tool steel alloys, e.g., H11) (Roberts et al., 1980). This is the real beauty of understanding the working of solidification and segregation in AR irons. It allows us to dive in, eliminate the extraneous and develop fundamental mechanisms of AR irons' behavior.

## Alloy-Specific Notes on Solidification

### Pearlitic and Ni-Hard 1 and 2 AR Cast Irons

In much of the discussion on segregation, the emphasis has been placed on high-Cr compositions with an assumption of fairly slow cooling rates. This is at direct odds with the conditions pertaining to chill-cast, pearlitic AR iron grades, and in many applications where Ni-Hard 1 and 2 are prevalent. Although the same principles apply (e.g., Si is still completely rejected from the growing carbide phase), the intensity of these reactions increases.

Take, for example, the behavior of a eutectic pearlitic iron having a CVF of 35%. At this level, the Si content of the dendrite is approximately 1.5 times the original melt's content. At high solidification rates, when diffusion is minimized, dendritic Si content can be even greater than predicted by straight partitioning between carbide and austenite phases.

The casting of steel mill rolls presents an example of a pearlitic or lean Ni-Hard 1 alloy that experiences a high solidification rate. In this casting process, the alloy is cast against a very thick steel cylinder (e.g., about 20 cm (8 in.) thick) to create a 5–10 cm (2–4 in.) thick shell of AR iron. This shell is then backed up with a gray or nodule iron to form the body and necks of the roll. Typical steel mill rolls are shown in Chapter 10, Applications.

Another name for pearlitic or lean Ni-Hard 1 alloys that have been chemically tailored for steel mill roll applications is "indefinite chill." Figure 2.29 shows a micrograph of the indefinite chill microstructure. The elongated graphite zones between the eutectic carbide structure is the main microstructure feature of indefinite chill irons. This structure is brought about by the carbide eutectic grains growing parallel to the heat flow. As the carbide structure grows, Cr is absorbed and



**Fig. 2.29.** Optical micrograph of a polished section, perpendicular to the chill face. Dark stripes that run perpendicular to the chill face are graphite stringers. General composition is 3.9% C, 0.7% Si and 1.8% Cr. 80X.

silicon is rejected into the remaining liquid. This interlayer of liquid is then driven toward the graphite eutectic leading to the thin graphite stringers shown in Fig. 2.29. The amount of graphite formed can be manipulated to some extent by varying the ratio of graphitizers (Si, Ni, Al, etc.) to carbide stabilizers (Cr, Mn, Mo, V, etc.) added to the melt depending on the required chill depth. The indefinite chill structure is popular with roll manufacturers because of its unique combination of resistance to wear and thermal shock.

### Ni-Hard 4 AR Cast Irons

Silicon is the one common alloying element that has a real impact on the solidification behavior of pearlitic and Ni-Hard 1, 2 and 4 AR cast irons. Its use is most striking in Ni-Hard 4 (ASTM ID). Although the ASTM specification requires an Si level of between 1.0 and 2.2%, this specification range is too wide to always result in the production of an iron having the same microstructure. Indeed, at the lower end of the specification (<1.4% Si), a duplex carbide is formed, having a hard  $M_7C_3$  core surrounded by a softer  $M_3C$  shell. At the higher end of the specification (>1.5% Si), the carbide will be entirely  $M_7C_3$ .

Figures 2.30a and 2.30b show a few results of this interesting behavior. As the Si content is increased, the formation of  $M_3C$  is suppressed. Adequate suppression occurs at a threshold value of 1.5% Si, while complete suppression is obtained at even higher Si levels.

To explain this behavior, we need to examine the phase diagram for a low-Si Ni-Hard 4 alloy (0.3% Si). As the Ni-Hard 4 starts to freeze, proeutectic austenite dendrites will start to form in the hypoeutectic iron. As the remaining liquid becomes enriched in C and Cr, the iron moves up and to the right until it hits the eutectic line. At this stage, the eutectic reaction of  $L \Rightarrow \gamma Fe + M_7C_3$  is initiated. As the remaining liquid continues to freeze, it will slide down the eutectic line until it hits the peritectic point at 4.2% C and 9.7% Cr. Once at the peritectic point, shells of  $M_3C$  will start to form around

the existing cores of  $M_7C_3$ . If enough liquid remains after the peritectic reaction, it then continues to solidify as ledeburite ( $L \Rightarrow \gamma Fe + M_3C$ ). (see Glossary for Peritectic Reaction.)

If the Si content is above 1.5%, the peritectic reaction is hindered. At even higher levels, say, around 1.7% Si, the peritectic reaction is effectively suppressed. Proof of this mechanism is given by the DTA thermograms shown in Fig. 2.30c for low- and high-Si Ni-Hard 4 irons. As the Si content is increased, the  $M_3C$  formation (i.e., the peritectic reaction) is suppressed and the  $M_3C$  peak disappears.

This behavior is explained by Si's effect on increasing the activity coefficient of C within the melt. As eutectic solidification commences, Si is completely rejected from the growing  $M_7C_3$  carbide. This creates a high-Si halo around the growing carbide. Consequently, this high-Si halo increases the localized C activity and shifts the liquid composition to the right (i.e., higher C). This compositional shifting to a pseudo higher C content drives the growth of existing  $M_7C_3$  carbides until all the remaining liquid has solidified.

To close on this subject of Ni-Hard 4, NiDI's 1996 publication (see references) recommends that Ni-Hard 4 should have between 1.5 and 2.0% Si with an optimum Si content at 1.9%. If the reader is interested, they should also study the work of Laird et al. (1991) and Thorpe and Chicco (1981).

### High-Chromium AR Cast Irons

The saga on peritectic reactions is not yet over. In a few experiments with 15% Cr irons, it was found that the hypoeutectic (3.5% C) and eutectic (3.7% C) grades all gave uniform microstructures of austenite and  $M_7C_3$ , while the hypereutectic (4.3% C) grade gave a mixed microstructure of austenite,  $M_7C_3$  and  $M_3C$  (or duplex carbides having a core of  $M_7C_3$  surrounded by a shell of  $M_3C$ ) (see Laird and Dogan, 1996). Figure 2.31 shows an SEM photomicrograph of this duplex microstructure.

The formation of  $M_3C$  carbides in just the hypereutectic grade is due only to the initial formation of primary  $M_7C_3$  carbides, depleting the liquid of C and Cr, which drives the remaining liquid straight toward the peritectic point (this can be visualized by drawing a straight line on Fig. 2.8 from the origin to this hypereutectic composition). In the hypoeutectic and eutectic grades, the liquid is enriched in C and Cr during solidification and the remaining liquid is driven away from the peritectic point. In short, when the Cr level is not above 15% Cr, be careful in dealing with hypereutectic grades, or the results may not be what was expected.

### Specialty AR Cast Irons

#### D2 and 440C AR Cast Irons

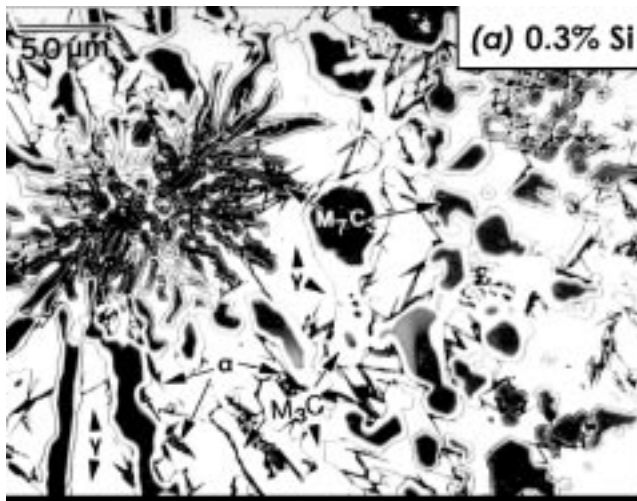
Before covering the other specialty alloy systems, we should finish the discussion from Chapter 1 about the Fe-Cr-C system covering the special alloys, known commercially as D2 and 440C (see Table 1.5 for details). Two large black dots, representing D2 and 440C, are superimposed on the Fe-Cr-C

liquidus surface shown in Fig. 2.32. Both of them fall to the right of the austenite eutectic solidus (dashed) line and are thus classified as irons (since eutectic carbides are formed). A further look at Fig. 2.32 indicates that a number of low-C, high-Cr irons can be devised by keeping to the right of the austenite eutectic solidus line. This is a handy piece of information for the foundryman interested in developing tough, AR or abrasion/corrosion-resistant irons.

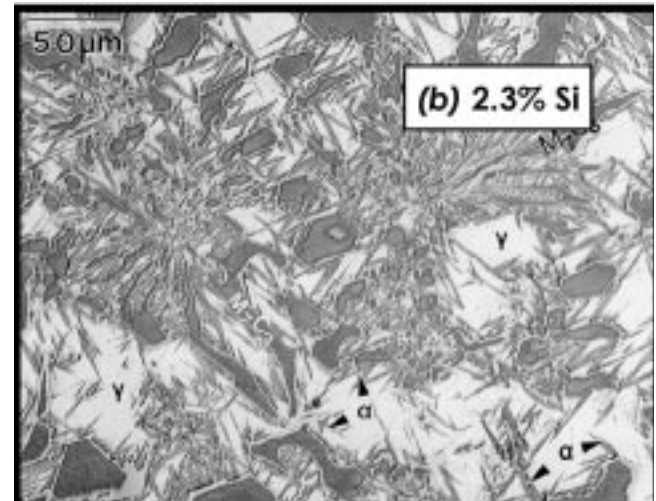
As can be interpreted on this phase diagram (Fig. 2.32), both of these irons form  $M_7C_3$  eutectic carbides, albeit through slightly different solidification routes. The D2 iron follows

the standard solidification route for a hypoeutectic iron, while the 440C iron starts with the formation of  $\delta$  (delta ferrite) and then follows the peritectic line ( $L + \delta \Rightarrow \gamma$ ) until the remaining liquid hits the peritectic point ( $L + \delta \Rightarrow \gamma + M_7C_3$ ) and finally finishes within the eutectic valley ( $L \Rightarrow \gamma + M_7C_3$ ).

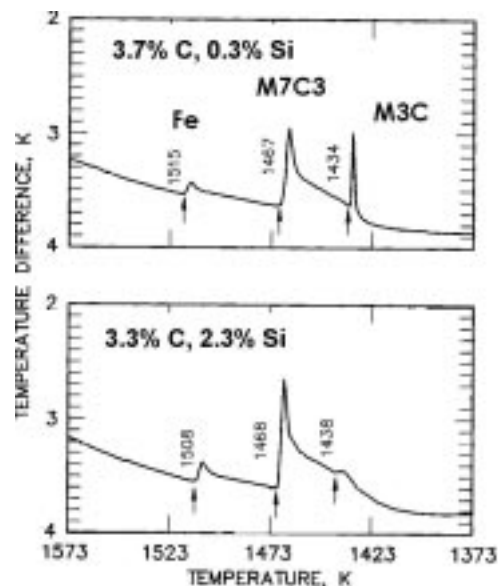
Although one may think that the 440C iron would be less fluid than the D2 iron, they are actually quite similar in fluidity. They both have extremely long solidification ranges and could be expected to suffer from piping and hot tearing problems.



(2.30a)



(2.30b)



(2.30c)

**Fig. 2.30.** Silicon has a strong tendency to suppress the formation of  $M_3C$  during solidification. This is shown quite sharply in Ni-Hard 4 irons containing a range of Si contents. The two photomicrographs show polished sections that have been etched with 2% nital and then stained with Murakami's. The Murakami's will stain the  $M_7C_3$  carbide and not the  $M_3C$  carbide. The  $M_7C_3$  carbide will appear dark and the  $M_3C$  carbide will appear light. As shown in (a), if the Si content is 0.3%, the iron will form soft  $M_3C$  carbides around the much harder  $M_7C_3$  carbides. At higher Si content, (b) shows that it is possible to completely suppress the formation of  $M_3C$  carbides. The DTA thermograms (c) provide another method to confirm these results.



Fig. 2.31. SEM (in backscatter mode) of duplex microstructure. Dense phases (matrix and  $M_3C$  carbides) reflect more electrons toward the detector and appear lighter, while less dense phases (e.g.,  $M_7C_3$ ) reflect fewer electrons and appear darker. The formation of the peritectic shell of  $M_3C$  carbide is shown around most all of the  $M_7C_3$  carbides.

These irons, however, serve slightly different applications. Both have good toughness (i.e., compared to a regular AR cast iron) because they have very low CVFs (around 10–12% versus 20–35% for high-Cr AR cast irons). They also respond similarly to abrasive wear when heat treated to the same

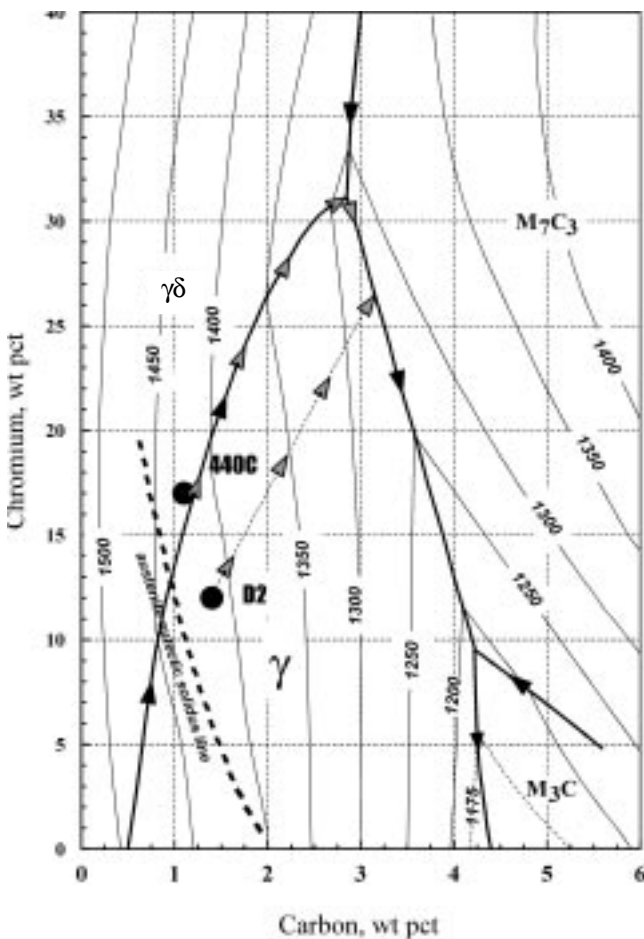


Fig. 2.32. Solidification paths for low-C AR cast irons, known commercially as D2 and 440C.

hardness. The similarity ends here, however. From the phase diagram, it is seen that the Cr content of 440C austenite dendrites is around 15%, while that for the D2 dendrites is around 10%. It follows that the corrosion resistance of the 440C is superior to that of D2. This shows, once more, how phase diagrams can be powerful tools for the prediction of abrasion and corrosion resistance of existing irons and for the development of new irons.

### Ferritic High-Cr (Fe-Cr-C-X) AR Cast Irons

The Fe-Cr-C-X irons are characterized by the use of alloying additions of one or more of the following elements in place of the X: V, Nb, W, Mo and Ti. These should be in sufficient quantities to form discrete carbides having the chemical notation close to one metal (M) atom for one carbon (C) atom or  $MC$ .

**Vanadium Addition**—One member of this alloy family is commonly known as A7 (AISI classification) and is denoted as a special wear-resistant cold work die steel by Roberts and Cary (1980). Its base composition is 2.3% C, 5% Cr and 5% V with Si, Mn and Mo added for hardenability. A complete specification is given in Table 1.5. The appropriate Fe-Cr-C-V liquidus surface for this iron is shown in Fig. 2.33. It was constructed by holding the Cr content constant at 5% while varying the C and V contents. Hence, this surface is only valid for Cr contents close to 5%.

Following the definition of what constitutes an iron, the A7 alloy should be classified as a hypoeutectic iron. The first phase to form is  $\gamma$ , followed by the V eutectic reaction of VC carbides and austenite ( $L \Rightarrow \gamma + VC$ ). The iron continues to solidify along this eutectic valley until the remaining liquid hits the peritectic point and starts to form  $M_7C_3$  carbides ( $L + VC \Rightarrow \gamma + M_7C_3$ ). At the very end, the remaining liquid slides

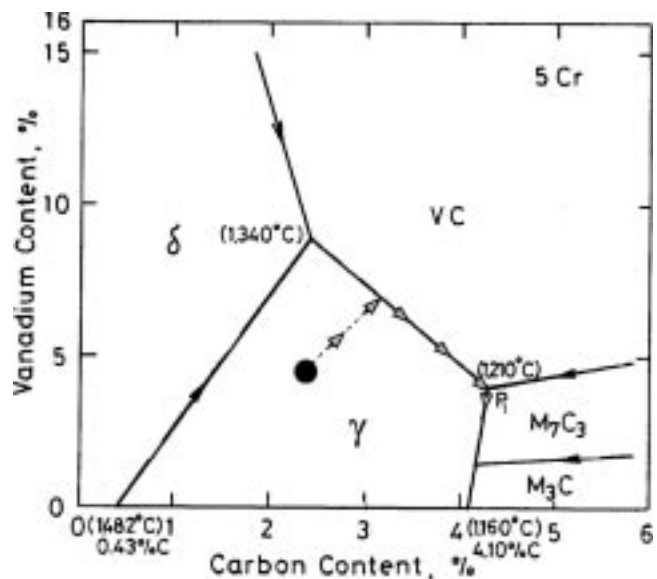


Fig. 2.33. Metastable liquidus surface for the Fe-Cr-V-C system with the solidification path for A7 (2.3% C, 5% Cr, 4.75% V iron) superimposed on the liquidus surface. The black dot marks the start of the A7 solidification sequences. (After Sawamoto et al., 1986)

down the Cr carbide eutectic line and forms a small amount of eutectic  $M_7C_3$  ( $L \Rightarrow \gamma + M_7C_3$ ).

After solidification, the microstructure of the A7 iron appears as shown in Fig. 1.8, presenting large dendrites interspersed with eutectic VC carbides and  $M_7C_3$  carbides at the grain boundaries. The Cr added to this iron brings the benefit of additional CVF and increased hardenability. Because of their high V content, such irons are expensive but can provide superior toughness and abrasion resistance in some applications. The discrete VC carbides allow toughness, and their extreme hardness (2400 DPH) provides superior abrasion resistance under some conditions.

**Niobium Addition**—The effect of adding Nb to the Fe-Cr-C systems is similar to that of V, but much stronger. Due to the very strong affinity of Nb for C, in this case, the solidification begins with the formation of NbC carbides and then follows the standard solidification pattern for an Fe-Cr-C alloy.

Figure 2.34 shows the microstructure of a high-Cr Nb and V alloyed AR cast iron. Although only 0.5% Nb is added, it readily forms NbC. For comparison, an addition of 2% V dissolves mostly into the  $M_7C_3$  carbides leaving trace amounts in the matrix. This confirms Nb's much stronger carbide-forming characteristics.

An important characteristic shown in the micrograph given of Fig. 2.34 is the clustering of the NbC carbides at the center of the austenite dendrites. The situation of these carbides at the center of the dendrite indicates that they solidify first; i.e., prior to the formation of the austenite. The presence of hard Nb carbides at the center of the softer dendrites can provide increased wear resistance. This is true whether the iron is a low-alloy or high-alloy AR iron, and is of interest to those wanting to strengthen the dendritic structure of the iron.

**Tungsten, Molybdenum and/or Titanium Addition**—Other exotic additions to special AR cast irons are W, Mo and Ti. All are strong carbide formers; however, the effects of W and Mo are similar to that of V, while Ti behaves much more like Nb. Considering the lack of known commercial applications for

cast alloys with high-alloy additions of W and Mo, the interested reader is referred to Roberts and Cary (1980) for more information. Nonetheless, we need to mention that in specialty tool steels, W and Mo combine to form a very hard carbide with a stoichiometry referred as  $M_6C$ , where the M consists of Fe, W and Mo atoms and has a hardness of around 2400 DPH.

Irons alloyed with Ti enjoy more usage. In cast iron, Ti forms TiC carbides having hardnesses of around 3000 DPH. Titanium, of course, will also consume a good fraction of any available N within the melt, forming titanium nitrides (TiN). However, since the C presence is orders-of-magnitude greater, TiC will readily form.

Titanium is very reactive and will almost immediately form small TiC carbides within the melt (with a melting temperature of 1700C/3092F). One concern is their lower density:  $TiC = 4.5 \text{ g/cm}^3$ , as compared to the cast iron's  $\sim 7.5 \text{ g/cm}^3$ . In thick castings, TiC carbides can float to the top. As a caveat, all metallurgical processes have their limits, yet TiC CVFs of around 6–8% have been successfully introduced into Ni-Hard compositions. Be warned, however, that the presence of solid particles in the melt will have a drastic affect on the iron's fluidity.

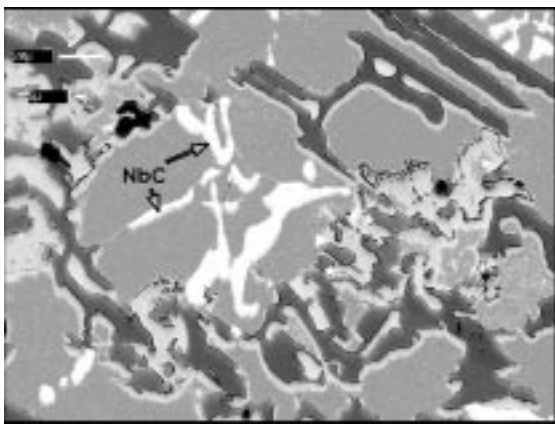
## Effects of Tramp Elements on Solidification

Often, there are iron specifications with rigid limits for phosphorus (P) and sulfur (S) (see Appendix E, Standards, Table E7). For example, the ASTM specifications call for P and S limits of 0.30% and 0.15%, respectively, for Type IA, B and C; and P and S limits of 0.10% and 0.06% for all others. In other countries, similar restrictions on P and S are applied. The reasons for these limits are based on experience rather than theory.

### Phosphorus

In the early days of iron manufacturing, P was often added to increase the fluidity of the iron. In actuality, P forms a low melting eutectic ( $L \Rightarrow \gamma\text{Fe} + \text{Fe}_3\text{P}$ ). As the iron solidifies, P segregates into the liquid and, after everything else has solidified, forms a brittle phosphide eutectic. By extending the time that the iron remains fluid, its "fluidity" is then increased, albeit with some disastrous mechanical property consequences. Figure 2.35 shows the morphology of the phosphide eutectic. Its porous morphology is easily fractured and its presence along the grain boundaries leads to easy and fast intergranular fracture (i.e., the fracture path runs along the grain boundaries).

A little-known reason for adding P to AR cast irons is the elimination of the appearance of hot tears at the surface of the castings. However, the hot tear still exists, it is just filled in or "healed" with the phosphide eutectic. Visually, it appears as if the hot tearing has been prevented but, mechanically, the iron has become even weaker.



**Fig. 2.34.** SEM (backscatter mode) of a 2.3% C, 28% Cr, 0.5% Nb, 2% V iron. The very dense NbC carbides appear white under backscatter mode. The other surrounding carbides are of the  $M_7C_3$  type. No VC carbides were found. Note: image analysis was used to highlight the NbC carbides.

In AR cast irons, P is typically not engineered into the iron. It is considered a *tramp element*, thus an undesirable element. This leads one to wonder at what level P is deleterious. Phosphorus strongly segregates into the liquid during solidification and, as a result, phosphide eutectic has been found at levels as low as 0.02% P. It is, thus, entirely likely that we will find the brittle phosphide eutectic in irons having 0.10% P and, to a greater extent, in irons containing 0.30% P.

However, irons are not steels. Abrasion-resistant irons have their own preexisting network of brittle carbides and the addition of trace amounts of another brittle phase has little measurable effect on its mechanical and wear properties. Since the ASTM types IA, B and C already contain plate-like  $M_3C$  carbides, their tolerance for additional plate-like phosphide carbides (max. 0.30% P) is considerably greater than that of the higher-Cr AR irons containing rod- and blade-like  $M_7C_3$  carbides (max 0.10% P).

From this discussion, foundrymen should keep in mind that there is no level of phosphorus, that, once exceeded, will “destroy” their iron. That is, many applications could reasonably tolerate phosphorus levels of 1% or more, while other applications will require that the P level be no more than 0.1%. It is all a matter of understanding the iron and the application.

## Sulfur

Sulfur is another restricted or *tramp element* in AR irons. Sulfur, like P, strongly segregates into the liquid during solidification. For carbide-containing irons, S is no big deal. Sulfur is a big concern for gray iron foundries, where graphite morphology (flake or spheroidal) is affected by the S content. Sulfur is also a problem for steel foundries, for a number of reasons. It has a strong affinity for Mn and will readily form MnS during solidification. This is part of the problem.

Because S strongly segregates into the liquid, these MnS inclusions will naturally form along the grain boundaries and act as initiation sites for intergranular fracture. Consequently, when Mn content is low and S content high, embrittlement will occur when hot working the steel at around 1000C (1832F) (it is called hot shortness). Under these conditions, the intergranular sulfide eutectic (just like the phosphide eutectic, earlier) will melt and cause hot shortness or cracking of the steel as it is deformed.

For irons, these issues do not strike a strong note. A few MnS inclusions are never a big deal in a material having 10–40% CVF and, furthermore, cast irons are rarely hot worked. Realistically, AR cast irons could easily handle higher levels of S (above 0.15%), with little degradation in properties. Thus, while it is best not to purposely add S to the iron, would we pig an AR iron having 0.15% S? Most likely not. On the other hand, would we purposely try to make a AR cast iron having 0.15% S? Most likely not, in view of sulfur’s known tendency to induce hot tearing and reduce toughness.

## Inoculation for Better Solidification

Inoculation is a sensitive subject to many scientists around the world. Definitive and conclusive research showing the formation of finer carbides through inoculation are about nonexistent. At best, the evidence that the eutectic carbide size can be controlled by inoculation is very limited. This has not hindered some researchers to claim that they have been able to form spherical carbides via the addition of special inoculants or alloying compositions. In all cases, these claims are not fully substantiated. That is, 2D views showing smaller carbides do not imply that the carbide’s 3D shape has evolved from rod-like to sphere-like. Until further proof is delivered, this discussion will center on the limited research available on the refinement of the eutectic microstructure through increased nucleation.

Now is a good time to emphasize a few points about nucleation. Nucleation is not uniform and not perfect. As the iron approaches the metastable liquidus surface, small solid particles will act as nucleation sites. Initially, only a few sites will be attractive enough to initiate growth. As the iron cools further, a greater driving force develops to transform the liquid into solid. At this stage, additional solid particles will become attractive to the liquid metal for nucleation and growth.

As the temperature drops, dendrites start growing below the metastable surface. However, dendrites that nucleated early in the solidification process grow much larger than the dendrites that started to form in the later stages of solidification. Consequently, the iron will have very large eutectic carbides surrounded by a smattering of smaller eutectic carbides.

Therefore, the aim of inoculation is to introduce solidified particles into the melt, in order to provide an overwhelming number of attractive nucleation sites, such that growth occurs throughout the liquid at the same time. Competition between

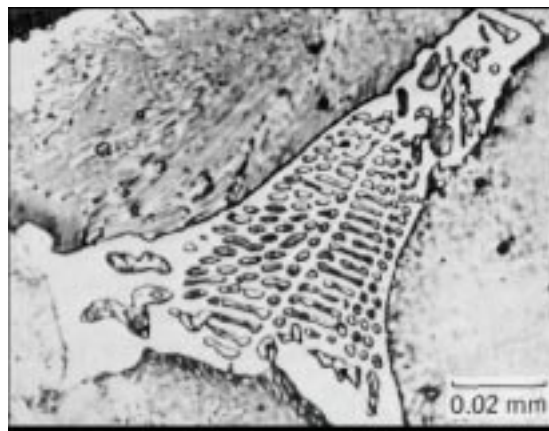


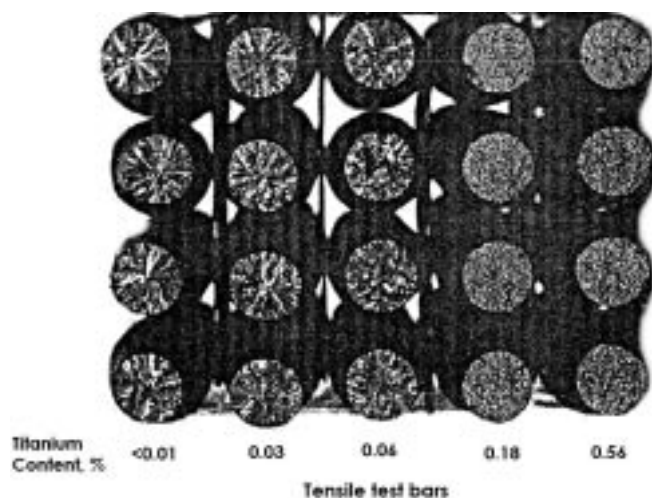
Fig. 2.35. Phosphide eutectic at the grain boundary within a gray iron. (Courtesy BCIRA)

the growing eutectic grains hinders the formation of large grains and thus promotes the formation of smaller, more uniform-sized eutectic grains (which results in finer eutectic carbides). The goal of inoculation is to create a finer-grained iron with improved mechanical and wear properties. This goal has rarely been achieved. Additionally, as demonstrated in the section on Effect of Fluidity on Solidification, a more uniformly solidifying melt will also have increased fluidity.

One bright piece of work on inoculation was done by Rickard (1960) on a ferritic high-Cr iron (1.5% C, 30% Cr). Figure 2.36 shows some results from Rickard's work. The evidence is quite conclusive that Ti does promote a finer grained microstructure in this hypoeutectic iron. Since the iron is hypoeutectic and falls within the  $\delta$  field on the metastable Fe-Cr-C liquidus surface (Fig. 2.8), the first phase to solidify is ferrite. However, when Ti is added, the first phase to solidify is actually TiC, followed by the formation of ferrite.

In drawing conclusions from Rickard's work, we could say that TiC acts as a nucleant to the ferrite phase. As the hypoeutectic iron cools and approaches the liquidus surface, solid TiC particles provide numerous sites for the ferrite to nucleate and grow. Growth then occurs at many sites and in all directions, resulting in many small grains.

Besides the qualitative data on grain refinement (Fig. 2.36), Rickard also provided tensile test data indicating that grain refinement increased the iron's tensile strength from 45 MPa (65 kpsi) to 54 MPa (78 kpsi). This is logical, since it has long been known that grain refinement in steel will result in increased tensile strengths. How much refinement of microstructure relates to abrasion and impact properties is unknown at this time.



**Fig. 2.36. Fractured tensile test bars after inoculation with varying amounts of titanium. As the level of titanium is increased, the grain size of the hypoeutectic 1.5% C, 30% Cr AR iron decreases.**

In more recent years, other researchers have looked at Ti inoculation and have found similar results for hypoeutectic irons. For example, Basak et al. (1982) looked at a single addition of 0.35% Ti in a 2.5% C, 4.5% Mn, 11% Cr iron. In this case, the hypoeutectic iron falls within the  $\gamma$  (austenite) field of Fig. 2.8, and austenite is the metallic phase that is nucleated by the solid TiC particles. Microstructural refinement was noted, but it is difficult to draw many conclusions from an experimental program that consists of two irons, one uninoculated and one inoculated.

Other scientists (e.g., Liang and Su, 1991) have played around with various additions of mischmetal (a combination of rare earth elements with an emphasis on cerium) as inoculants. Some modification was shown for a low-Cr near-eutectic iron (4.2% C, 3.7% Cr –  $L \Rightarrow \gamma\text{Fe} + \text{M}_3\text{C}$ ). However little change was noted in a high-Cr hypoeutectic iron (3% C, 18% Cr –  $L \Rightarrow \gamma\text{Fe} + \text{M}_7\text{C}_3$ ).

Until now, inoculation research has centered on hypoeutectic irons in which either ferrite or austenite nucleation is enhanced by the presence of TiC particles. It is interesting to speculate as to whether or not eutectic or hypereutectic alloys could be refined in a similar manner. However, as of this writing, it is unsure as to how much benefit can be derived from the grain refinement in AR cast irons.

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# Chapter 3

## Heat Treatment of AR Cast Irons

Cast iron is usually defined as an alloy of iron and carbon with a C content above 2 weight percent. Other alloying elements can total about 6 wt.% in low-alloy irons and much higher in more highly alloyed irons such as Ni-Hard 4 and high-Cr white irons. All AR irons contain structures consisting of eutectic carbides and one or two ferrous metallic phases. White cast iron can exhibit a wide variety of properties, depending on composition, phases and microconstituents present, which can vary depending on the heat treatment. Heat treatment is performed to make castings very hard for maximum abrasion resistance, or soft to make them machinable. In subsequent sections of this chapter, various types of heat treatment are described for Ni-Cr and high-Cr white irons. In this section, an outline of the phase transformations associated with heat treating of cast irons will be given.

### Phase Transformation

In cast irons, metallic iron can exist in two allotropic or crystalline forms: ferrite and austenite. In ferrite, the iron atoms are arranged in a body-centered-cubic (BCC) crystalline lattice. In austenite, the iron atoms are arranged in a face-centered-cubic (FCC) crystalline lattice. Furthermore, ferrite has a very low solubility for C (<0.03%), whereas austenite has a high solubility for C (>1%). Ferritic iron is the stable phase at room temperature. Austenitic iron is generally stable at elevated temperatures, unless it has been made stable at room temperature by elevated levels of C and/or alloying elements such as Ni, Cr and Mn.

It is this characteristic of metallic iron that makes it possible to heat treat castings to a high hardness. That is, the iron can be heated to an elevated temperature to form an FCC lattice, which dissolves large amounts of C. Then it can be quenched (rapidly cooled) to room temperature where the BCC lattice forms and carbon atoms are rejected and form iron carbide. Depending on the transformation temperature, the distribution of iron carbide is more or less finely distributed and the hardness and strength of the metallic matrix is varied.

### Equilibrium Diagrams

The basis for understanding the heat treatment of cast irons is the iron-carbon phase diagram (Fig. 3.1). Because it is thoroughly explained in elementary textbooks, it will be treated very briefly here. Figure 3.1 actually shows two diagrams; the stable iron-graphite diagram (dashed lines) and the metastable iron-iron carbide diagram. The stable condition is almost always deliberately avoided in AR cast irons; therefore, the

metastable diagram is of more interest. These diagrams show which phases are to be expected at equilibrium (or metastable equilibrium) for any combination of C concentration and temperature. Table 3.1 provides a summary of important metallurgical phases and microconstituents.

At the low-carbon end (left side of the diagram in Fig. 3.1) ferrite (*alpha*-iron/ $\alpha$ -Fe) is found (see Glossary for Delta and Alpha Ferrite), which can, at most, dissolve 0.028 wt.% C. At higher temperatures, austenite (gamma-iron/ $\gamma$ -Fe) is found, which can dissolve up to 2.1 wt.% C. At the C-rich side, we find cementite ( $\text{Fe}_3\text{C}$ ). Of less interest, except for highly alloyed irons, is the delta-ferrite ( $\delta$ -Fe) existing at the highest temperatures. Between the single-phase fields are found regions with mixtures of two phases, such as ferrite + cementite, austenite + cementite and ferrite + austenite. In heat treating cast irons, the liquid phase is always avoided. Some important boundaries at single-phase fields have been given special names that facilitate the discussion. These include:

- $A_1$ , the so-called eutectoid temperature, which is the minimum equilibrium temperature for austenite.
- $A_3$ , the low-temperature boundary of the austenite region at low C content, that is, the  $\gamma/\gamma + \alpha$  boundary.
- $A_{cm}$ , the low temperature boundary of the austenite region at high C contents, that is, the  $\gamma/\gamma + \text{cementite}$  ( $\text{Fe}_3\text{C}$ ) boundary.

Relevant definitions of terms associated with phase transformations of cast irons can be found in Table 3.1. The C content at which the minimum austenite temperature is attained is called the eutectoid C content (nominally 0.8% C). The ferrite-cementite phase mixture of this composition, formed during cooling, has a characteristic appearance called pearlite, and can be treated as a microstructural entity or microconstituent. It is an aggregate of alternating ferrite and cementite lamellae.

If alloying elements are added to the iron-carbon alloy cast iron, the positions of the  $A_1$ ,  $A_3$  and  $A_{cm}$  boundaries and the eutectoid composition are changed. The diagrams in Figs. 3.2a to 3.2c illustrate the effects of Cr, Mn and Si on the austenite (gamma) phase field. The influence of Ni is reportedly similar to that of Mn. Diagrams introduced by Bain (1939) (Fig. 3.3) show the variation of  $A_1$  temperature and the eutectoid C content with increasing amounts of a selected number of alloying elements. It suffices here to mention that (1) all important alloying elements decrease the eutectoid C content; (2) of the austenite-stabilizing elements, Mn and Ni decrease the  $A_1$  temperature; and (3) the ferrite-stabilizing elements (Cr, Si, Mo and W) increase the  $A_1$  temperature.

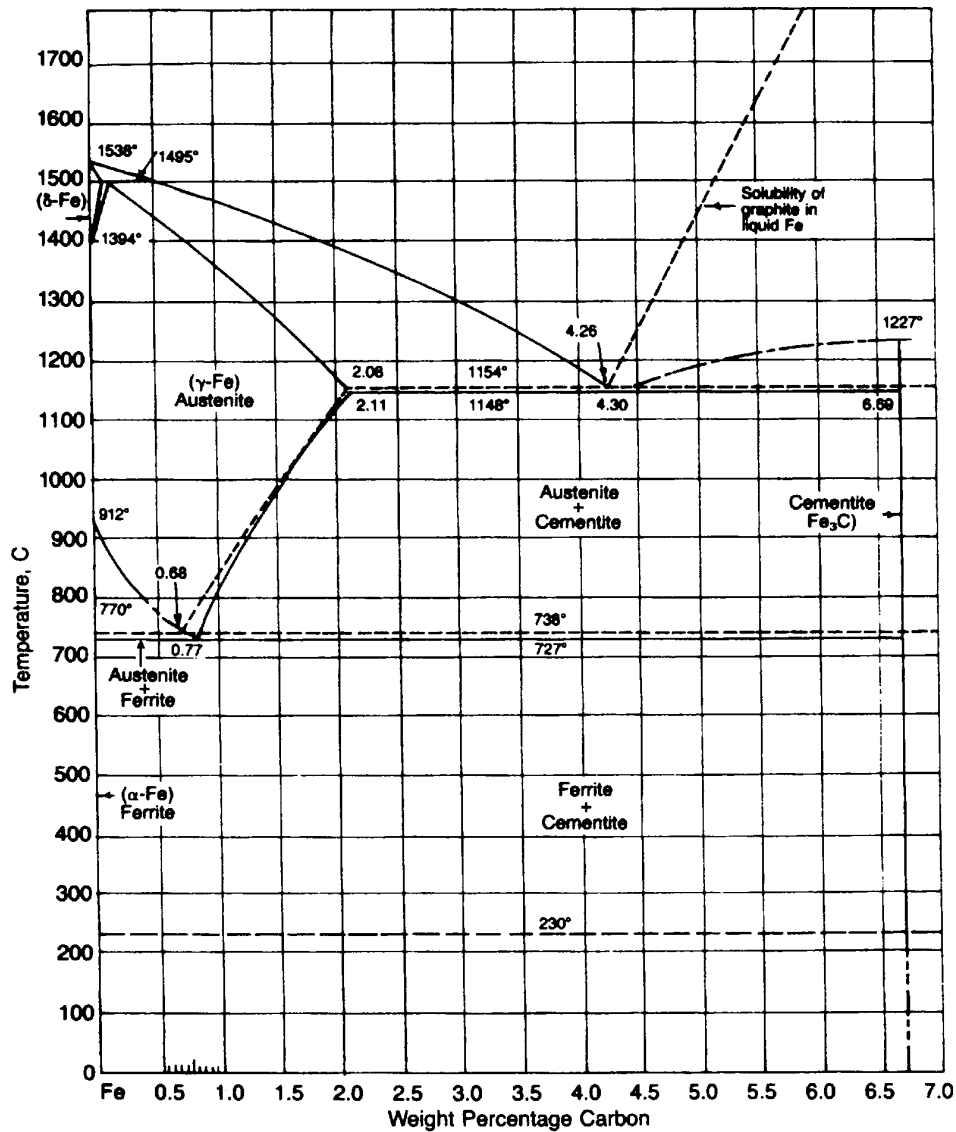


Fig. 3.1. Fe-C phase diagram. (Metals Handbook, ASM, 1973)

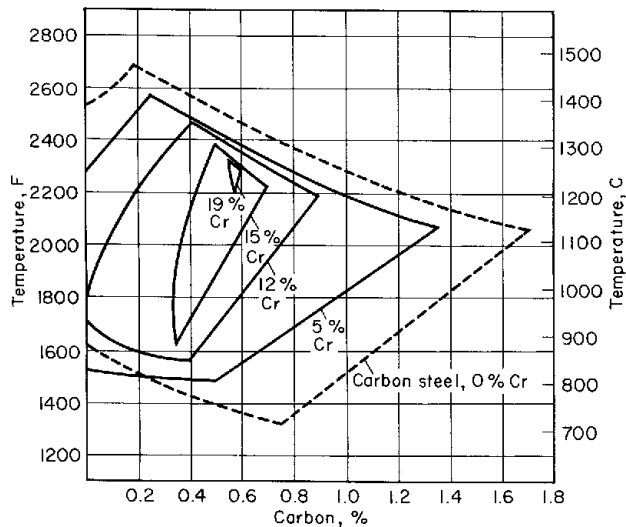


Fig. 3.2a. Influence of Cr on the austenite phase field in Fe-Cr-C system.

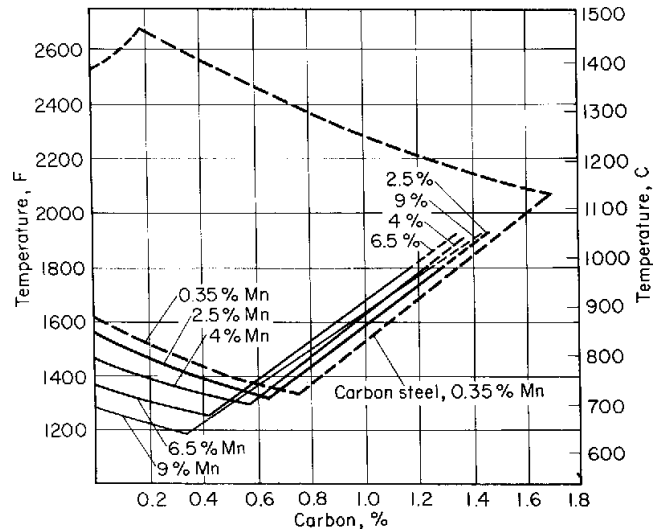
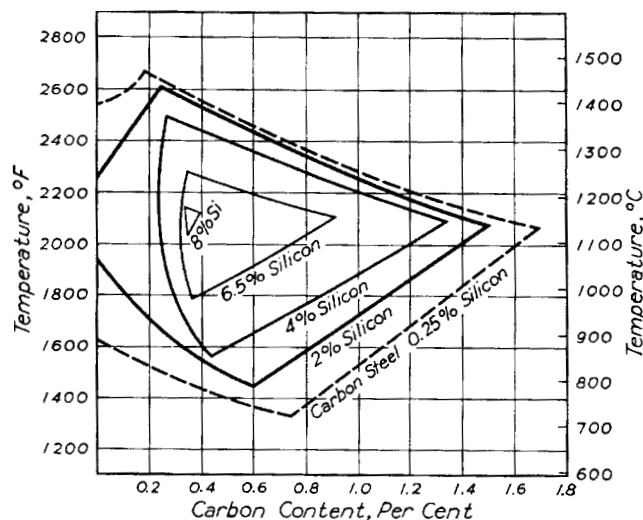


Fig. 3.2b. Influence of Mn on the austenite phase field in Fe-Mn-C system.

**Table 3.1**  
**Summary of Important Phases and Microconstituents in AR Cast Irons**

| Phase<br>(Microconstituent)                           | Crystal Structure<br>of Phases                                            | Characteristics                                                                                                                                                                                                                                                                                                                   |
|-------------------------------------------------------|---------------------------------------------------------------------------|-----------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------|
| $\alpha$ -Ferrite ( $\alpha$ -iron)<br>(alpha-iron)   | BCC (body-centered-cubic)                                                 | Relatively soft low-temperature phase;<br>stable equilibrium phase.                                                                                                                                                                                                                                                               |
| $\delta$ -Ferrite ( $\delta$ -iron)<br>(delta-iron)   | BCC                                                                       | Isomorphous with $\alpha$ -iron;<br>high-temperature phase;<br>stable equilibrium phase.                                                                                                                                                                                                                                          |
| $\gamma$ -Austenite ( $\gamma$ -iron)<br>(gamma-iron) | FCC (face-centered-cubic)                                                 | Relatively soft medium-temperature phase;<br>stable equilibrium phase.                                                                                                                                                                                                                                                            |
| Cementite ( $\text{Fe}_3\text{C}$ )                   | Complex-orthorhombic                                                      | Hard metastable phase.                                                                                                                                                                                                                                                                                                            |
| Graphite                                              | Hexagonal                                                                 | Stable equilibrium phase                                                                                                                                                                                                                                                                                                          |
| Pearlite                                              | Mixed/complex-orthorhombic                                                | Metastable microconstituent;<br>lamellar mixture of ferrite and cementite                                                                                                                                                                                                                                                         |
| Martensite                                            | BCT (body-centered-tetragonal) (super-saturated solution of C in ferrite) | Hard metastable phase; lath morphology when $<0.6$ wt% C; plate morphology when $>1.0$ wt% C and mixture of those in between.                                                                                                                                                                                                     |
| Bainite                                               | Mixed                                                                     | Hard metastable microconstituent; non-lamellar mixture of ferrite and cementite on an extremely fine scale; upper bainite formed at higher temperatures has a feathery appearance; lower bainite formed at lower temperatures has an acicular appearance. Hardness of bainite increases with decreasing temperature of formation. |

These classifications relate directly to the synergisms in quench-hardening, as described later in Chapter 4, Alloying for Hardenability. Modern thermodynamic calculations (i.e., Calphad or ThermoCalc applications) allow accurate determinations of these shifts that affect the driving force for phase transformation. These methods also permit calculation of complete ternary and higher-order phase diagrams, including alloy carbides.



**Fig. 3.2c.** Influence of Si on the austenite phase field in Fe-Si-C system.

## Transformation Diagrams

The kinetics of phase transformation are as important as the equilibrium phase relationships described in the previous section for the heat treatment of AR cast irons. The metastable constituent, martensite, which is of utmost importance to the properties of AR cast irons, can generally form by comparatively rapid cooling to ambient temperature, that is, when transformation to pearlite and bainite are avoided. Both pearlite and bainite are products of the eutectoid decomposition of austenite into a mixture of ferrite and cementite. Eutectoid decomposition of austenite relies on the diffusion of carbon and iron atoms into the growing ferrite and carbide phases. Alloying elements slow diffusion and delay transformation. Consequently, alloying additions are made to increase the “hardenability” of an iron-carbon alloy.

Martensite, the hardest metallic constituent, forms from austenite by a shear transformation during quenching. It’s hardness increases monotonically with C content up to about 0.7 wt.%. If martensite is subsequently heated to a moderately elevated temperature, it decomposes to tempered martensite, which has a more stable distribution of ferrite and carbide. The re-heating process is known as tempering or drawing.

Transformation can be illustrated with two different types of commonly-used transformation diagrams. These include isothermal transformation diagrams and continuous-cooling transformation diagrams.

Isothermal transformation (IT) diagrams, also referred to as time-temperature-transformation (TTT) diagrams, describe the decomposition of austenite at constant temperatures occurring below the  $A_1$  temperature. An example of an IT diagram for Ni-Hard 1 is shown in Fig. 3.4a. Here, it can be seen that at 550C, pearlite transformation is delayed 10 hours.

Continuous cooling transformation (CCT) diagrams describe the decomposition of austenite on continuous cooling from the austenite phase field to room temperature. An ex-

ample of a CCT diagram for Ni-Hard 1 is shown in Fig. 3.4b. Here, it can be seen that pearlite transformation is delayed about 4 hours. Continued discussion of the kinetics of phase transformation, the use of transformation diagrams and the selection of alloying elements for hardenability will be presented in Chapter 4, Alloying for Hardenability.

## Nickel-Chromium (Ni-Hard 1 and 2) White Irons

The optimum matrix microstructures for Ni-Hard 1 and Ni-Hard 2 (ASTM A 532 Grades IA–IC) are most often achieved in the as-cast condition. Oftentimes, the only heat treatment required is tempering (or double tempering) to improve resistance to impact loading. The high alloy content affords adequate hardenability to avoid pearlite or bainite in all but the thickest cast sections (see Chapter 4, Alloying for Hardenability, for further discussion on this topic).

### As-Cast Microstructure

The nickel content must be adjusted in the range of 3.3 to 4.5% to provide the “as-cast” hardenability of Ni-Cr alloys. The diagram in Fig. 3.5 illustrates the influence of Ni on hardness in various cast sections. With too little Ni, the hardness falls off in larger cast sections due to partial transformation to pearlite. With too much Ni, the hardness falls off in lighter cast sections due to undue suppression of the martensite transformation temperature range and an accompanying increase in retained austenite. Furthermore, with increasing Ni content and section size, increased levels of Cr are required to inhibit graphitization. Table 3.2 lists the recommended Ni and Cr contents for varying cast section sizes and molding mediums.

Generally, the high alloy content of the Ni-Cr white irons causes the as-cast structures to have relatively high retained

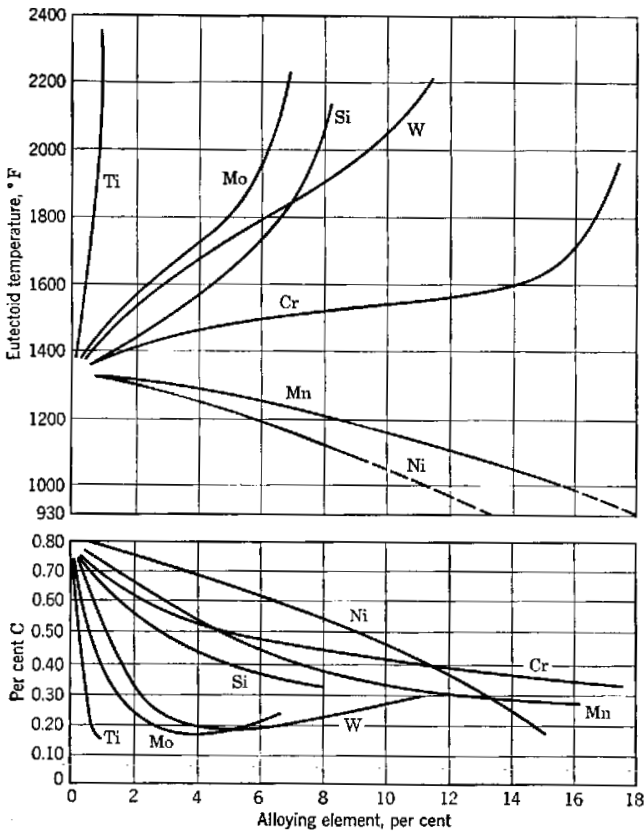


Fig. 3.3. Influence of various alloying elements on the critical  $A_1$  temperature and eutectoid C content in Fe-C alloys.

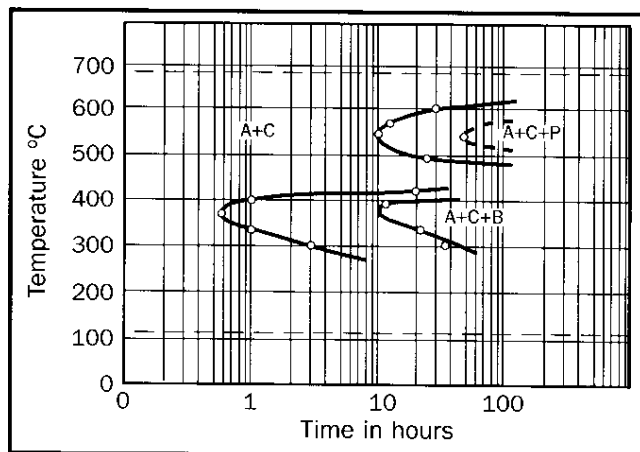


Fig. 3.4a. Isothermal transformation (IT) diagram for Ni-Hard Type 1: A = austenite; B = bainite; C = carbide; P = pearlite.

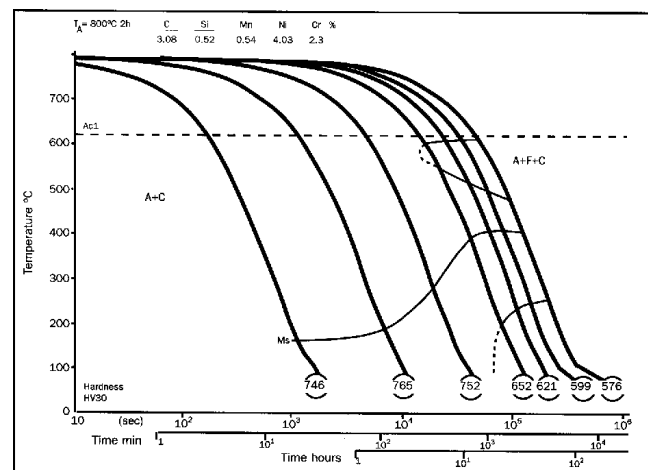


Fig. 3.4b. Continuous-cooling transformation (CCT) diagram for Ni-Hard Type 1: A = austenite; B = bainite; C = carbide; P = pearlite; F = ferrite;  $M_s$  = martensite start.

austenite contents. The elements Ni and Mn tend to lower the martensite start temperature and increase the level of retained austenite in the as-cast structure. On the other hand, Si tends to raise the martensite start temperature and increase the amount of martensite. As a result, the as-cast hardness of the Types IA–IC Ni–Cr white irons can range from 450 to over 700 HB, depending on composition and section size. While high hardness and low retained austenite are most often desired, some have suggested that excessive hardness can lead to cracking and/or breakage due to low toughness. In those instances, the composition can be adjusted to produce higher retained austenite content (and lower hardness), but tempering may be the best approach.

Regarding an increase in retained austenite, Mn is actually more influential than Ni. While both elements depress the  $M_s$  temperature, Ni raises C activity, lowers the solubility of C in austenite and, therefore, has less influence in lowering the  $M_s$  temperature.

Silicon has an equally large influence on retained austenite. It significantly increases C activity and has a very weak effect in depressing the  $M_s$  temperature. Consequently, Si strongly reduces the C solubility in austenite and thus raises the  $M_s$  temperature and, consequently, the as-cast hardness.

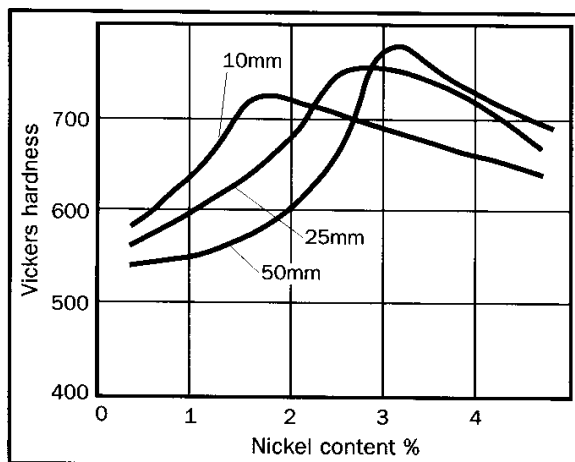


Fig. 3.5. Hardness of as-cast Ni-Hard 1 in various cast sections as a function of Ni content.

Table 3.2  
Guidelines for Ni and Cr Contents of Ni-Hard 1 and Ni-Hard 2  
as a Function of Section Size

| Section Size (mm) | Ni-Hard 1    |               |     |     | Ni-Hard 2    |               |     |     |
|-------------------|--------------|---------------|-----|-----|--------------|---------------|-----|-----|
|                   | Sandcast %Ni | Chillcast %Cr | %Ni | %Cr | Sandcast %Ni | Chillcast %Cr | %Ni | %Cr |
| <12               | 3.8          | 1.6           | 3.3 | 1.5 | 4.0          | 1.5           | 3.5 | 1.4 |
| 12–25             | 4.0          | 1.8           | 3.6 | 1.7 | 4.2          | 1.7           | 3.8 | 1.5 |
| 25–50             | 4.2          | 2.0           | 3.9 | 1.9 | 4.4          | 1.8           | 4.1 | 1.6 |
| 50–75             | 4.4          | 2.2           | 4.2 | 2.1 | 4.6          | 2.0           | 4.4 | 1.8 |
| 75–100            | 4.6          | 2.4           | 4.5 | 2.3 | 4.8          | 2.2           | 4.7 | 2.0 |
| >100              | 4.8          | 2.6           | 4.8 | 2.5 | 5.0          | 2.4           | 5.0 | 2.2 |

## Recommended Heat Treatments

Ni-Hard was originally intended to be put into service in the as-cast condition. Nowadays, however, almost all castings are heat treated. There are three basic heat treatments that are applied to Ni–Cr/Ni-Hard iron castings.

1. **Tempering**—Tempering in the range 225–275°C (437–527°F) (single-stage tempering). Tempering at low temperatures makes martensite less brittle and improves toughness. Some improvement in resistance to repeated impact fatigue has also been ascribed to single-stage tempering.
2. **Double Tempering**—Double tempering, first at 450°C (842°F) followed by air-cooling and tempering at 275°C (527°F). This two-stage tempering reportedly reduces hardness somewhat, but substantially increases resistance to repeated impact fatigue.
3. **Quenching and Tempering**—Full heat treatment above the critical temperature in the range 750–850°C (1382–1562°F), followed by slow cooling and low-temperature tempering, can maximize hardness and abrasion resistance.

Most castings of Ni-Hard 1 and 2 are given single-stage or two-stage heat treatments, while most Ni-Hard 4 is given full quench-and-temper heat treatment. For the more complex castings, where heat treatment might produce too much added stress, refrigeration treatments to as low as –196°C (–320°F) (liquid nitrogen) are applied to further transform existing austenite to martensite, thus increasing the hardness.

In addition to tempering, the Ni–Cr (Ni-Hard) castings can be given additional treatments to alter their microstructure and improve certain properties. These treatments will be discussed briefly in the next few paragraphs as they relate to the Ni–Cr (Ni-Hard 1 and Ni-Hard 2) castings.

A more comprehensive explanation of subcritical heat treatment, refrigeration treatment and stress relief will be given in the section on high-Cr and Cr–Mo alloys.

## High-Temperature Heat Treatment

Hardening of Class I Type D, Ni-Hard 4 iron castings is performed by supercritical heat treatment when as-cast hardness is insufficient. An austenitizing heat treatment usually comprises heating at temperatures between 750 and 790°C (1382 and 1454°F) with a soak time of 8 hours. Air or furnace cooling, not over 30°C/hr (50°F/hr), is conducted followed by a tempering/stress-relief heat treatment. Currently, refrigeration heat treatment, followed by tempering, is the more commonly practiced method for increasing the hardness (reducing retained austenite).

## Refrigeration Treatment

To achieve a hardness of 550 HB, it is necessary that the as-cast austenite-martensite microstructure have at least 60% martensite present. When martensite content is increased to 80–90%, however, hardness values exceed 650 HB. To reduce the amount of retained austenite (form more martensite) deep-freeze treatments are commonly applied. Refrigeration

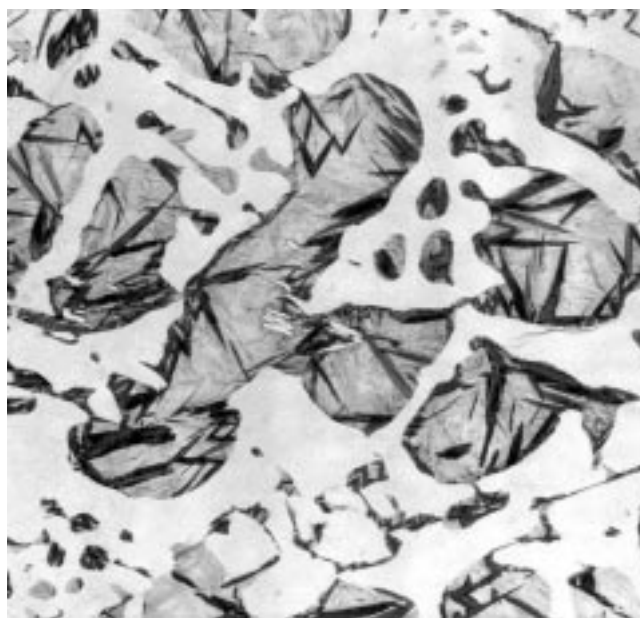
to temperatures between  $-70$  and  $-180^{\circ}\text{C}$  ( $-94$  and  $-292^{\circ}\text{F}$ ) for 1/2 to 1 hour will usually raise the hardness level by 100 HB units. A subsequent tempering heat treatment usually follows, and is recommended. The typical before-and-after refrigerated Ni-Cr iron microstructures are shown in Figs. 3.6a and 3.6b.

### Tempering Treatment

All Ni-Cr white iron castings are given a tempering heat treatment because, properly made, they have a martensitic matrix structure, as cast. Tempering is performed between  $205$  and  $260^{\circ}\text{C}$  ( $400$  and  $500^{\circ}\text{F}$ ) for at least 4 hours. This tempers the martensite, relieves some of the transformation stresses



(3.6a) before refrigeration



(3.6b) after refrigeration

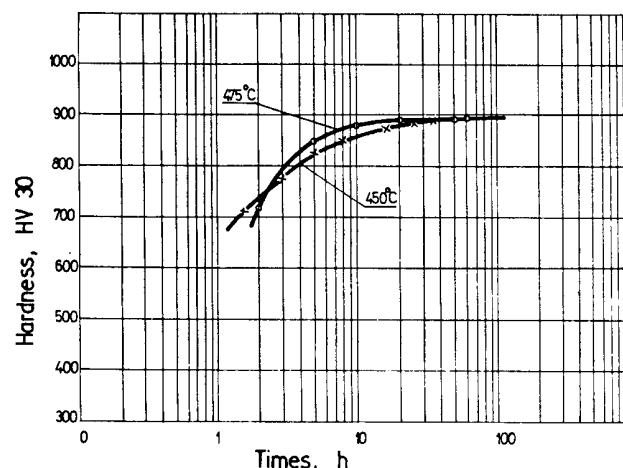
Fig. 3.6. Microstructure Ni-Hard 1 alloy, before and after refrigeration.

and increases the strength and impact toughness by 50–80%. Some additional martensite may form on cooling from the tempering temperature. This heat treatment does not reduce hardness or abrasion resistance.

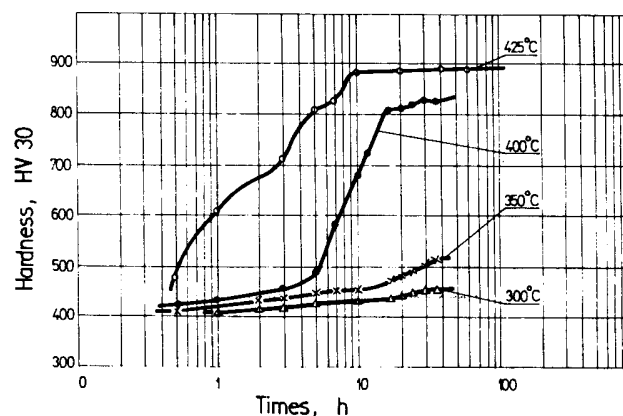
### Subcritical Heat Treatment

Destabilization of austenite in Ni-Cr, Ni-Hard 1 and 2 alloys (ASTM A 532 Class IA and IB) at subcritical temperatures was suggested by Rote et al. (1951) and confirmed by Tomovic et al. (1988). Subcritical heat treatment is performed in the bay between the pearlite and bainite transformation regions. When held at temperatures between  $425$  and  $475^{\circ}\text{C}$  ( $797$  and  $887^{\circ}\text{F}$ ), a predominantly austenitic matrix structure destabilizes, by carbide precipitation, such that the depletion of dissolved C in the austenite raises the  $M_s$  temperature and, upon cooling, it transforms to a more fully martensitic microstructure of quite high hardness. The amount of time at temperature to reach maximum hardness (destabilization) was determined for one alloy to be from 10 to 20 hours (see Figs. 3.7a and 3.7b).

Subcritical heat treatment of Ni-Hard 1 and 2 alloys is not commonplace. It is reportedly used in the processing of as-cast, permanent-mold, grinding balls, where high levels of austenite are retained as a result of rapid (air) cooling from high temperatures.



(3.7a) shows 450 and 475°C



(3.7b) shows 300, 350, 400 and 425°C

Fig. 3.7. Change in hardness in Ni-Hard 1 alloy with time at temperature subsequent to quenching from  $1100^{\circ}\text{C}$ .

## High-Chromium and Chromium-Molybdenum White Irons

### As-Cast Microstructure

To a large extent, the metal matrix in high-Cr irons has characteristics of tool steel. It can be made less brittle to increase tolerance to impact loading, or hardened to reduce wear, though generally at the expense of toughness. Matrix structures can be *pearlite*, *austenite* or *martensite*, or some combination of these.

As with steels, maximum abrasion resistance is achieved with a fully martensitic matrix structure. As-cast austenitic irons, though relatively soft (500–600 HB), can give satisfactory performance in many abrasive wear applications due to pronounced work hardening in service. Pearlitic matrix structures have modest abrasion resistance and low toughness and, therefore, are not generally desirable. Pearlite will always occur when alloying is insufficient to inhibit its formation. Lack of hardenability is a major cause of premature failure.

In intermediate Cr irons (ASTM A 532 Class II), most of the Cr is tied up in the Cr carbides and, therefore, other alloying elements are almost always needed to provide adequate hardenability. In higher Cr irons (ASTM A 532 Class III), more Cr remains dissolved in the metallic matrix and less alloying is required for hardenability.

Resistance to crack propagation is greatest in irons having a fully austenitic matrix. Under conditions of repeated impact, however, martensitic irons, which have been tempered to minimize retained austenite, have the maximum resistance to spalling (surface cracking or flaking).

### As-Cast Austenitic White Irons

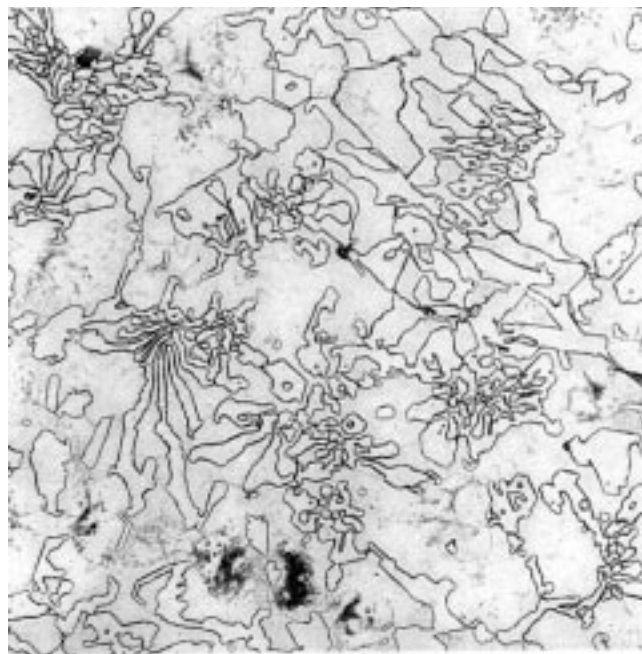
Solidification in the hypoeutectic alloys occurs by the formation of austenite dendrites followed by the eutectic formation of austenite and  $M_7C_3$  Cr carbides. Under equilibrium conditions, additional Cr carbide will precipitate from the austenite matrix on cooling from the eutectic to the critical temperature of about 760C (1400F); transformation to ferrite and carbide occurs on subsequent cooling. However, when cooling occurs under nonequilibrium conditions (such as encountered in most commercial castings), precipitation at Cr carbide is suppressed and the austenite becomes supersaturated in carbon and chromium.

Due to elevated C and Cr contents, a metastable austenitic structure normally develops, provided pearlitic and bainitic transformations have been inhibited (Fig. 3.8). With sufficient alloying with Mo, Mn, Ni and Cu, pearlitic transformation can be avoided in any section thickness.

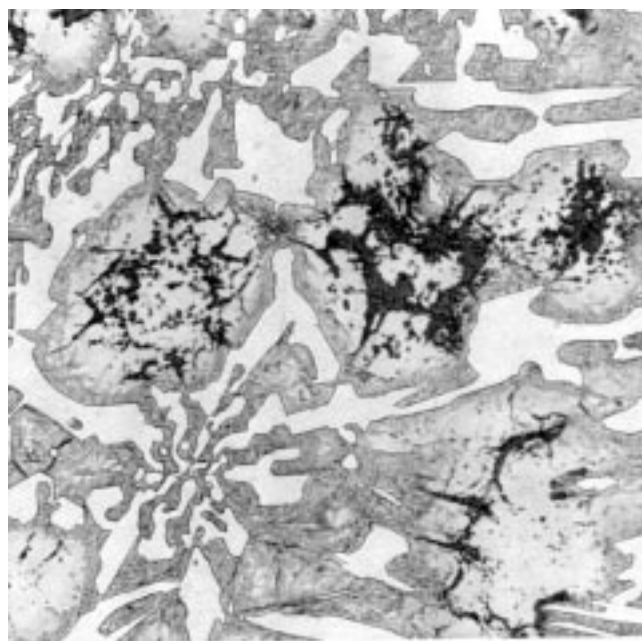
### As-Cast Martensitic White Irons

Partially martensitic structures can be obtained, as cast, in heavy-section castings that are allowed to cool slowly in the mold. With slow cooling rates, some austenite destabilization occurs and partial transformation to martensite will result. But in these castings, martensite is mixed with large amounts of

retained austenite (Fig. 3.9); therefore, hardness levels are lower than can be achieved in heat-treated martensitic castings. These castings must contain sufficient alloy to suppress pearlite on cooling. Some higher Si compositions have been developed to assist further martensite formation by refrigeration treatments. Subcritical annealing has been employed to reduce austenite content and, at the same time, increase hardness and toughness of partially martensitic-austenitic castings, such as rolling mill rolls.



**Fig. 3.8.** As-cast microstructure of high-Cr iron with an austenitic matrix.



**Fig. 3.9.** As-cast austenitic-martensitic microstructure in high-Cr white iron.

### As-Cast Pearlitic White Irons

For castings that will be heat treated, the desired mold-cooled structure is often pearlitic. This softer structure will facilitate removal of gates and risers and minimize the transformational and thermal stresses that cause cracking or spalling. The pearlitic structure will also shorten the response to heat treatment. To obtain maximum hardness and abrasion resistance, martensitic matrix structures must be produced by full heat treatment. Therefore, the casting must contain sufficient alloy to avoid pearlite formation on cooling from the heat treatment temperature.

Careful design of alloy composition will assure that a substantially pearlitic (softer) structure develops in the casting after mold-cooling, but will provide enough hardenability to prevent pearlite formation during subsequent heat treatment. Heavy-section castings made pearlitic by such a “lean-alloy” approach, can oftentimes be removed from the mold without danger of cracking, once the castings have reached black heat.

### Heat-Treated Martensitic White Irons

Maximum hardness, abrasion resistance and resistance to spalling are generally achieved in fully martensitic structures. As-cast martensitic structures generally contain plate martensite and significant quantities of retained austenite, and do not display the maximum hardness and toughness attainable in high-Cr iron castings. A uniform distribution of lath marten-

site gives the best properties and is only obtained by full heat treatment. (See Chapter 6, Quality Control, Inspection and Specifying, for examples of plate and lath martensite structures) In many applications, only a minimal amount (10% max) of retained austenite is tolerable and, yet, the high-Cr iron alloys are quite prone to having significant quantities (>30%) of retained austenite after heat treatment. For many applications, further reduction of austenite is required.

### Destabilization of Austenite

As cast, the metallic matrix of a high-Cr iron is largely austenitic and will contain an excess of dissolved C; the austenite matrix is stabilized by its high C and Cr contents. The principal role of heat treatment is to destabilize the austenite by reducing the C content via precipitation of secondary carbides. Carbide precipitation depletes the austenite of excess C and raises the martensite transformation temperature range to above room temperature. The heat treatment of commercial high-Cr white iron castings is commonly performed in the temperature range 930 to 1060C (1700–1940F).

When high-Cr irons are heated to the austenitizing temperature, they must be held at that temperature for a significant length of time to allow for the precipitation of carbides ( $M_7C_3$  Cr-carbides). The time required to reach equilibrium varies with austenitizing temperature. The carbide precipitation reaction has been investigated by Maratray (1970) for many austenitic high-Cr iron alloys and a typical transformation diagram is shown in Fig. 3.10. The transformation behavior, like other diffusion-controlled transformations, displays a “C” curve with the nose of the curve occurring around 1000C (1832F). At higher or lower temperatures, somewhat longer times are required to reach equilibrium (full austenite destabilization).

Carbide precipitation occurs in a reasonable length of time at temperatures above 930C (1706F). For instance, at 1010C (1850F) a holding time of two hours is adequate, whereas at 927C (1700F) a holding time of at least eight hours is required. At lower temperatures, the precipitation and rate of growth of the Cr-carbides are very sluggish due to the rather low rate of diffusion of Cr to the growing carbides.

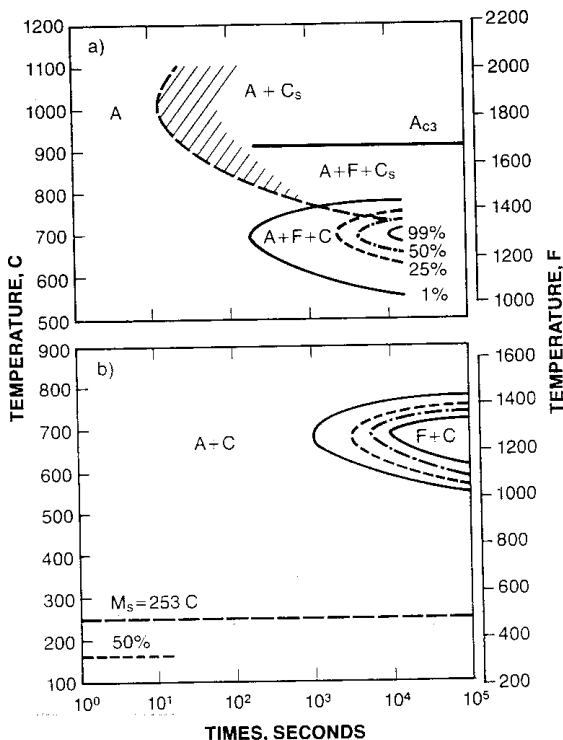


Fig. 3.10. Isothermal transformation diagrams for high-Cr white iron containing 2.5%C–20%Cr–1%Mo; a) before and b) after austenite destabilization: A = austenite; C = carbide; P = pearlite; F = ferrite; M<sub>s</sub> = martensite start; C<sub>s</sub> = secondary carbide.

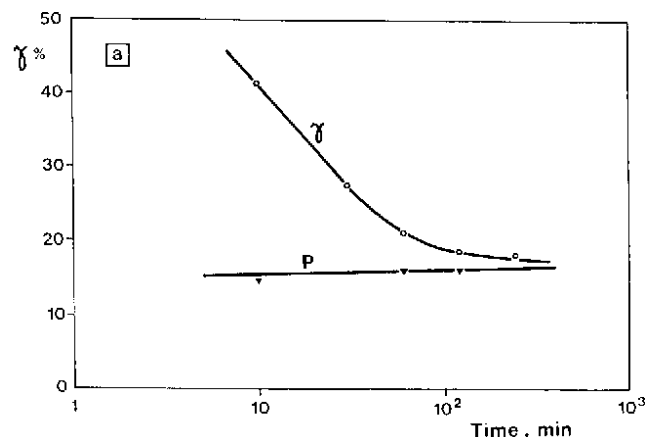


Fig. 3.11. Effect of solution treatment time and initial matrix structure on the amount of retained austenite upon cooling to room temperature.

### Austenitizing Time

The carbide precipitation reaction at the austenitizing (destabilization) temperature is necessary for a successful heat treatment. For proper control of hardness, the rejection of carbon, by carbide precipitation, must reach equilibrium. Otherwise, each set of castings processed through the heat-treatment cycle will have less than optimum hardness and wear resistance. Production castings are generally in one of two conditions prior to heat treatment; (1) as-cast austenitic or austenitic-martensitic or (2) largely pearlitic. Maratray (1982) has shown that castings that are initially pearlitic reach an equilibrium carbon content very rapidly, relative to austenitic structures. This is shown in Fig. 3.11 where the retained austenite content on cooling after cooling remains almost constant for the initially *pearlitic* matrix structure, whereas the *austenitic* sample required over 100 minutes to reach equilibrium.

Maratray (1982) reported that, in commercial practice, “heavy-section pearlitic castings (still) require over one hour at the solution temperature, probably due to the large amount of C precipitated as secondary carbides during the initial cooling of the castings. Castings that are largely austenitic require a holding time of at least two hours to attain equilibrium C content and even longer times are required if the casting is heavily segregated or has been heat treated at extremely high homogenization temperatures prior to solution heat treatment.”

### Austenitizing Temperature

Alloy composition can readily be designed to develop hard martensitic microstructures on air cooling. However, the as-quenched hardness is strongly affected by the hardening (austenitizing) temperature used and is directly related to the iron composition. Therefore, it must be optimized for each alloy. The amount of carbon dissolved in the austenite is controlled by the austenitizing temperature—carbon solubility increases with austenitizing temperature.

Although higher C content produces martensite having higher hardness, a high carbon content, along with the high alloy content of these irons, is sufficient to depress the martensite transformation range to below room temperature. Consequently, hardness will increase with austenitizing temperature up to a point; but with further increases in austenitizing temperature, hardness falls off due to the depression of the martensite temperature range and the accompanying increase in retained austenite content. The graph in Fig. 3.12 illustrates the effect of austenitizing temperature on martensite hardness, austenite content and their effects on overall hardness. Maximum hardness is generally achieved when retained austenite is in the vicinity of 20% (Fig. 3.12).

Because of this sensitivity to austenitizing temperature, furnaces that produce accurate and uniform temperatures are most desirable. The successful heat treatment produces austenite destabilization by precipitation of fine, secondary  $M_7C_3$  carbides, which are visible in the hardened martensitic microstructure (see Fig. 3.13).

Other influences on hardening temperatures are Cr content, Cr-C interaction, alloying elements and cooling rate.

**Influence of Cr Content**—Chromium content has a significant influence on the optimum hardening temperature. The bulk Cr content of the alloy has a direct effect on the amount of Cr dissolved in the metallic phase. In general, the 25–28% Cr alloys (Class III irons) have significantly more Cr dissolved

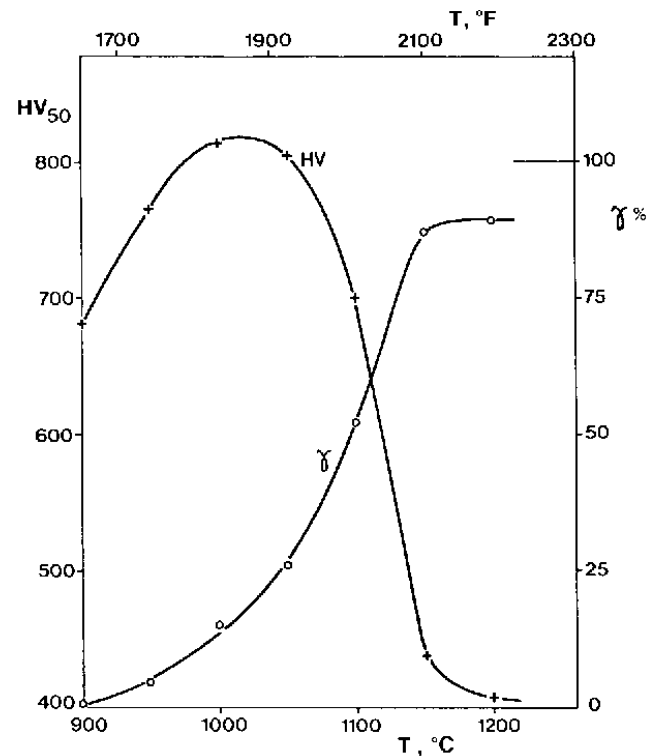


Fig. 3.12. Typical relationship between solution temperature, retained austenite content and hardness after quenching.

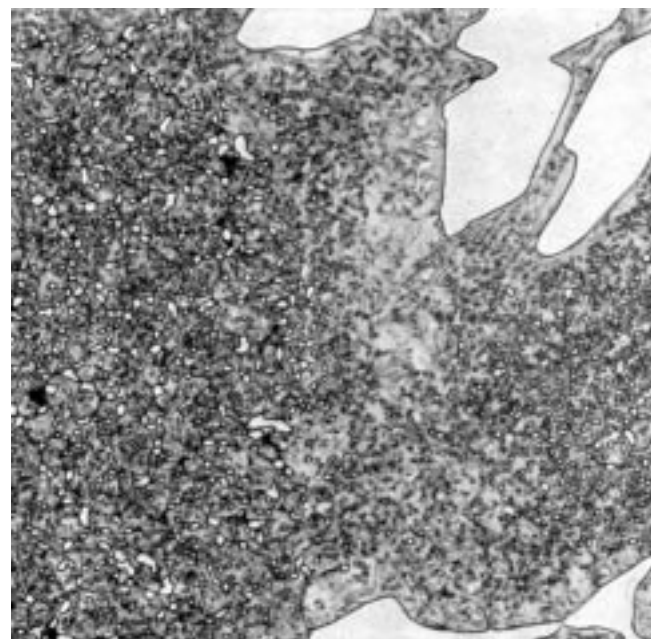


Fig. 3.13. Microstructure of heat treated (destabilized) martensitic high-Cr iron, illustrating fine, secondary  $M_7C_3$  carbides in a martensite matrix.

in the austenite phase than do the 15% Cr alloys (Class II irons). The differences in Cr content, have an effect on the solubility of C in austenite. This effect is illustrated in the isothermal section of the Fe-Cr-C phase diagram (Fig. 3.14).

Figure 3.14 shows three isothermal sections superimposed on one another and they cover the typical range of austenitizing temperatures employed in heat treating high-Cr white irons. The feature of interest is the boundary line separating the austenite phase field from the austenite +  $M_7C_3$  phase field. This boundary describes the solubility of C in austenite at equilibrium. It is seen that for a given temperature, the solubility of C in austenite decreases with increasing Cr content. Likewise, for a given Cr content, the solubility of C increases with temperature.

Consequently, as Cr content increases, a higher austenitizing temperature is required to achieve the same C content in the austenite. Therefore, as Cr content increases, the optimum hardening temperature increases (Fig. 3.15).

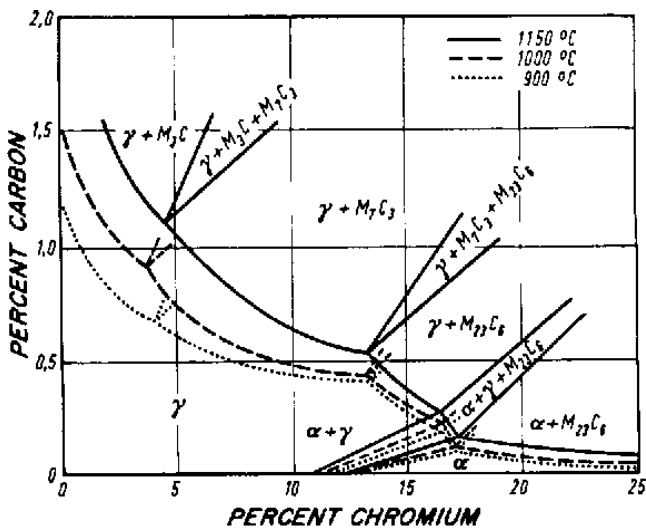


Fig. 3.14. Isothermal sections of the Fe-Cr-C system.

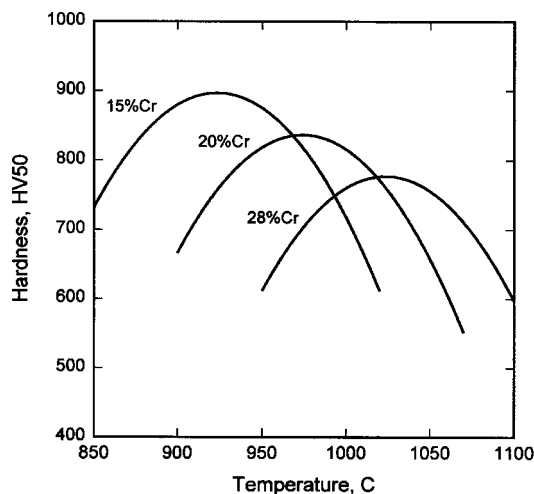


Fig. 3.15. Influence of Cr on the optimum hardening temperature in high-Cr white irons.

Chromium also has a significant influence on the maximum hardness of the hardened alloy because Cr affects the martensite start ( $M_s$ ) temperature. With increasing Cr, the  $M_s$  temperature is depressed and austenite becomes more stable. Consequently, at maximum hardness, the optimum balance between martensite and austenite occurs at a lower dissolved C content and the maximum hardness is reduced. This effect is also shown schematically in Fig. 3.15.

Class II irons containing 12–20% Cr are austenitized in the temperature range 950–1010C (1742–1850F). Class III irons containing 23–28% Cr are austenitized in the temperature range 1010–1090C (1850–1994F).

**Influence of Cr-C Interaction**—Carbon strongly influences the level of Cr dissolved in the metallic phase. As the C content is raised, the amount of Cr consumed by the formation of eutectic carbides increases and the amount of Cr remaining in solution in the metallic phase is reduced. The use of Cr:C ratio has been used as a tool to simplify our understanding of the effects of Cr and C in the alloy. In general, as the ratio of %Cr:%C in the alloy increases, Cr in the metallic phase increases and, likewise, the optimum hardening temperature increases.

**Influence of Alloying Elements**—Like Cr, the alloying elements that are added to high-Cr white irons for hardenability also influence the optimum hardening temperature. Most alloying elements, including Mo, Mn, Ni and Cu, lower the  $M_s$  temperature and tend to stabilize austenite. They also have an effect on C activity and, thereby, influence the C solubility in austenite. Consequently, they are expected to modify the optimum austenitizing temperature.

Maratray (1982) evaluated the influence of common alloying elements on retained austenite and showed that they are equally as influential as Cr content. He showed that Mn shifts the optimum hardening temperature to lower temperatures. He determined that the temperature to produce 50% austenite + 50% martensite on quenching to room temperature is 65°C lower with an addition of 3% Mn to a 1.6% C-16% Cr alloy (Fig. 3.16).

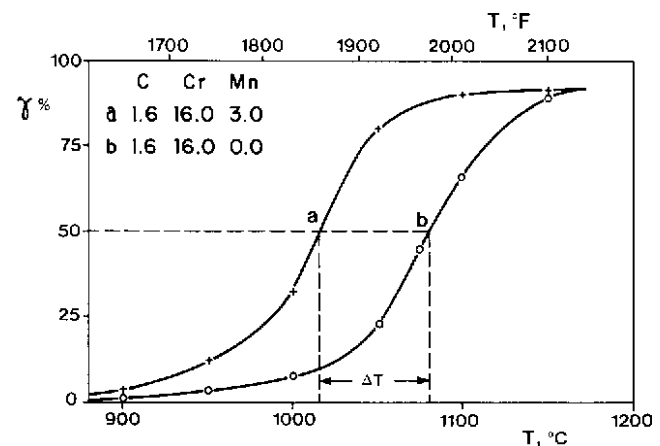


Fig. 3.16. Reduction in solution temperature ( $\Delta T$ ) to obtain 50% retained austenite on quenching, caused by an increment of 3% Mn in a 1.6C-16%Cr alloy.

Maratray (1982) reported on the shift in austenitizing temperature to produce an as-quenched microstructure of 20% austenite + 80% martensite for the alloying elements Mn, Ni, Mo and Si. The effects are shown in Table 3.3, which indicates the shift in austenitizing temperature for each 1% addition of the element. Manganese and nickel were shown to lower the optimum austenitizing temperature, while Mo and silicon raised the optimum austenitizing temperature. Silicon had, by far, the largest influence on austenitizing temperature. Silicon has a minimal effect on  $M_s$  temperature, but a major effect on reducing C solubility. This is why it is so effective in increasing the hardness (martensite content) of as-cast martensitic grades of high-alloy white irons.

**Influence of Cooling Rate**—Upon cooling from the austenitizing temperature to 700°C (1292°F), additional carbide precipitation can occur, particularly in massive slow-cooling sections. Carbide precipitation during cooling from the austenitizing temperature causes further depletion of C in the austenite matrix, a rise in the  $M_s$  temperature and a reduction in hardness of the as-cooled martensitic structure. The influence of cooling rate on reduction of dissolved C content is demonstrated by the reduction of retained austenite observed in samples quenched from various austenitizing temperatures and at various rates ranging from that of 10-mm to 500-mm diameter bars (Fig. 3.17). Note that at slower cooling rates (heavy sections), the influence of austenitizing temperature is less significant. The influence of cooling rate and quenching temperature in typical Ni-Hard 4 and high-Cr white iron alloys is illustrated in Figs. 3.18a and 3.18b, respectively.

The results in Fig. 3.18 show that austenite stability (retained austenite) is reduced with decreasing cooling rate above 700°C (1292°F); however, there are circumstances where

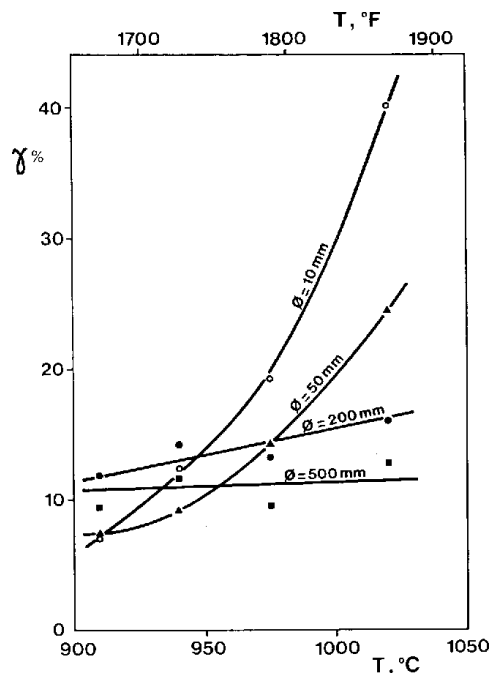


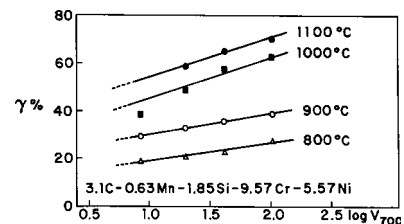
Fig. 3.17. Effect of cooling rate, indicated by section thickness and the solution treatment temperature on the retained austenite after cooling to ambient temperature of Alloy 25, containing 2.6C, 16Cr, 1.6Mo, 1.2Mn and 1.2Ni.

**Table 3.3**  
**Influence of Alloying Elements on Austenitizing Temperature to Produce 20% and 50% Austenite in an Austenite + Martensite Structure**

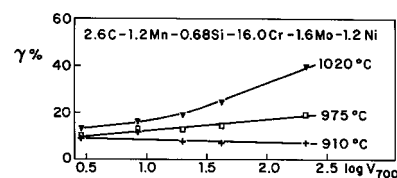
| Element (M) | Increase % Element (M) | T (°C) | 50% $\gamma$ (20% $\gamma$ ) $\Delta T / \Delta \%M$ |
|-------------|------------------------|--------|------------------------------------------------------|
| Mn          | 1.94                   | -64    | -33 (-35)                                            |
| Ni          | 1.80                   | -38    | -21 (-22)                                            |
| Cr          | 14.25                  | 22     | 1.6 (1.2)                                            |
| Cr          | 13.30                  | 16     | 1.2 (2.5)                                            |
| Mo          | 1.85                   | 24     | 13 (43)                                              |
| Si          | 1.23                   | 42     | 34 (79)                                              |

*M = element specified*  
*Values in parentheses represent the influence of alloying elements when obtaining 20% austenite + 80% martensite.*

austenite will become stabilized with low cooling rates through the martensitic transformation region. Maratray (1982) has shown that, with particularly low cooling rates, as-quenched hardness decreases due to increased retained austenite (Fig. 3.19). Many years ago, the phenomenon of austenite stability in Cr-alloyed steels was demonstrated by Lyman and Troiano (1946). In interrupted quenching experiments, they showed that austenite stabilization occurs with extended holding just above or below the  $M_s$  temperature in steels containing 3–9% Cr. Unpublished research on high-Cr white irons by the author has confirmed that austenite stability occurs with slow cooling below about 430°C (800°F).



(3.18a) Ni-Hard 4 white iron



(3.18b) 16% Cr white iron

Fig. 3.18. Influence of cooling rate on retained austenite content for (a) Ni-Hard 4, and (b) 16% Cr white iron, when cooled from various austenitizing temperatures. (After Maratray & Poulalion, 1982)

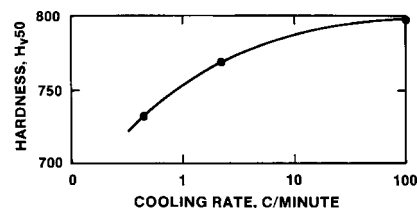


Fig. 3.19. Effect of cooling rate through the martensite transformation temperature region on room-temperature hardness for a high-Cr iron containing 1.7% C-16% Cr-3% Mo and 1% Mn and having an  $M_s$  temperature of 235°C (455°F).

## Recommended Heat Treatments

### Subcritical Heat Treatment (Tempering)

Castings can be put in service in the as-hardened condition without further tempering or subcritical treatments; however, the microstructure, after hardening, usually contains significant levels of retained austenite, which experience has shown to be a source of failure from spalling. Low temperature tempering at around 205C (400F) is much too low for transforming retained austenite; therefore, much higher temperatures in the range 450–600C (842–1112F) have been adopted. These treatments, referred to as subcritical treatments, provide an opportunity to eliminate retained austenite. In some instances, a significant increase in hardness is also observed.

Subcritical heat treatment (tempering) is sometimes performed, particularly in large heat-treated martensitic castings, to reduce retained austenite content and increase resistance to spalling. The tempering parameters necessary to eliminate retained austenite are very sensitive to time and temperature, and vary depending on the casting's composition and prior thermal history. Excess time or temperature results in softening and a drastic reduction in abrasion resistance. Insufficient tempering results in incomplete elimination of austenite. The amount of retained austenite present cannot be determined metallographically; those experienced with this heat-treatment practice have developed techniques, using specialized magnetic instruments, to determine the level of retained austenite after tempering.

Studies have shown that chemical composition, subcritical treatment temperature and time and variations in the prior hardening treatment influence the response to subcritical treatment, as the work by Maratray (1977) demonstrates. Also, subcritical treatments can increase hardness and reduce retained austenite in as-cast high-Cr irons produced in thick sections. About the same temperatures used for the heat-treated irons are utilized in the subcritical treatment of as-cast irons; the available data indicate that much longer holding times are required.

Maratray and Poulalion (1977) have proposed that there are basically three types of reactions that may take place on subcritical heat treatment of austenite. The reaction that always occurs first is the precipitation of carbide from the austenite. When adequately controlled, this carbide precipitation results in austenite destabilization and subsequent transformation to martensite on cooling to room temperature. The other two reactions proposed are: eutectoid decomposition of austenite to ferrite and an alloy carbide; or transformation to pearlite. Experience with a variety of alloys indicates that these reactions occur in the temperature range 480–650C (896–1202F).

Thermodynamic data on the heats of formation of the various Cr carbides that can form indicate that  $M_{23}C_6$  carbide is much more stable than  $M_7C_3$  carbide in this temperature range. Biss (1981) has determined that subcritical heat treat-

ment of an as-cast alloy at 625C (1160F) produced a structure of ferrite and fibrous carbides of the  $M_{23}C_6$  type, as shown in Fig. 3.20. He also found fresh martensite but no evidence of carbide precipitation within the remaining austenite.

It appears that the retained austenite in high-Cr irons can be divided into two types: austenite that will decompose at temperatures around 500C (932F) or austenite that must be heated to approximately 600C (1112F) before any significant decomposition occurs. Alloys subcritically heat treated at around 600C (1112F) will transform to pearlite, whereas those treated around 500C (932F) will not. The austenite that responds to heat treatment at around 500C (932F) has a high Cr:C ratio and a relatively low C content which, it would seem, favors the precipitation of the higher order  $M_{23}C_6$  carbide. This precipitation results in austenite destabilization and subsequently results in martensite formation on cooling.

Austenites of higher Cr content and lower C content favor formation of higher-order carbides. When the austenite composition does not favor formation of these higher-order carbides, longer time at temperature does not seem to have any effect, as has been observed in as-cast irons, irons having high retained austenite, high dissolved C content, and low  $M_s$  temperatures. As-cast 2.9% C-17.5% Cr alloys having been variously alloyed and containing well over 50% austenite were tempered 20 hours at 540C (1004F) with little or no visible change in structure and hardness. Yet, at 625C (1157F) they readily transformed to a mixed structure of pearlite plus fibrous carbides in ferrite. This structure is characteristic of the microstructures desired when producing hard pearlitic rolls, such as are used in hot-rolling applications where structural stability at elevated temperatures is needed.



**Fig. 3.20.** TEM micrograph illustrating fibrous carbides in ferrite in a high-Cr white iron containing 2.9% C-17% Cr-1.4% Mo-1.1% Ni, which was as-cast + subcritically heat treated for two hours at 625C (1160F). (After Biss, 1981)

When the composition and retained austenite levels are such that the alloy responds to heat treatment in the 500C (932F) temperature range, the austenite can be destabilized. With sufficient carbide precipitation during subcritical heating, the  $M_s$  temperature of the remaining austenite will be raised sufficiently and the austenite will subsequently convert to martensite on cooling to room temperature. In general, as the amount of retained austenite increases (reflecting a higher dissolved C content in the austenite), higher tempering temperatures are required to ensure its transformation, either during tempering or on subsequent cooling. Tempering on the low side of this temperature range minimizes tempering of the primary martensite and thereby produces a higher final hardness, as shown in Fig. 3.21. Note that small changes in time or temperature can drastically alter the heat treatment response. Furthermore, subcritical heat treatments do not usually produce an increase in hardness.

It appears that the volume of retained austenite in the iron casting is the best practical indication of the transformation characteristics of the remaining austenite. This is illustrated in Fig. 3.22, for an iron alloy that has been austenitized at three temperatures to produce 7, 60 and 100% austenite on quenching. In conditions with substantial retained austenite content, a significant increase in hardness occurred, whereas no significant change in hardness occurred in the alloy with only 7% austenite—other than martensite tempering. When it is possible to determine the volume of retained austenite, it is useful for predicting the appropriate subcritical heat treatment parameters for the casting (Fig. 3.23). In typical iron castings, a high level of retained austenite indicates that a high tempering temperature (near 600C or 1112F) is required for transformation of the austenite on tempering, but the austenite will transform largely at temperature, rather than on subsequent cooling. At lower levels of austenite, lower temperatures should be employed, and austenite transformation will occur on cooling.

### Refrigeration Treatment

To achieve high hardness and wear resistance in high-Cr irons without heat treatment, some have used refrigeration treatments. With as-cast irons having high retained austenite levels, it is possible to convert austenite to martensite with refrigeration. There are patented alloys specifically designed for this approach. Naturally, the alloys must have microstructures essentially free of pearlite in the as-cast condition to achieve maximum benefit from refrigeration treatments. Subsequent tempering is recommended to make the martensite less brittle and to increase toughness.

When large amounts of retained austenite are present (reflecting a very low  $M_s$  temperature), the alloys are more resistant to further martensitic transformation upon refrigeration and less hardness improvement can be expected. Maratray (1982) has shown that response to refrigeration is related to the stability of the austenite phase. This effect is illustrated in Fig. 3.24, where a series of samples were prepared having a range of  $M_s$  temperatures. As the  $M_s$  temperature of the alloy decreased, there was a disproportionate increase in the

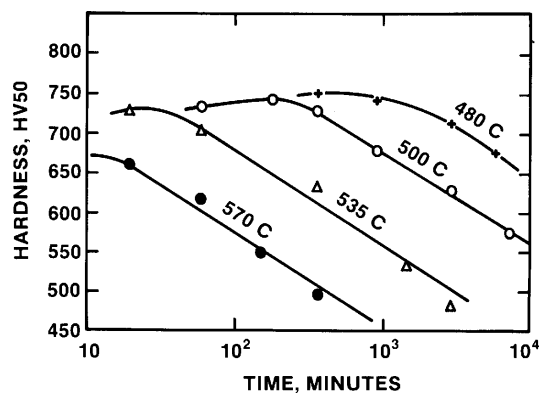


Fig. 3.21. Influence of subcritical tempering temperature on the maximum final hardness of an iron containing 1.6%C-16%Cr-3%Mo, which had been austenitized and quenched from 1000C (1830F) and had a hardness of around 840 HV prior to tempering.

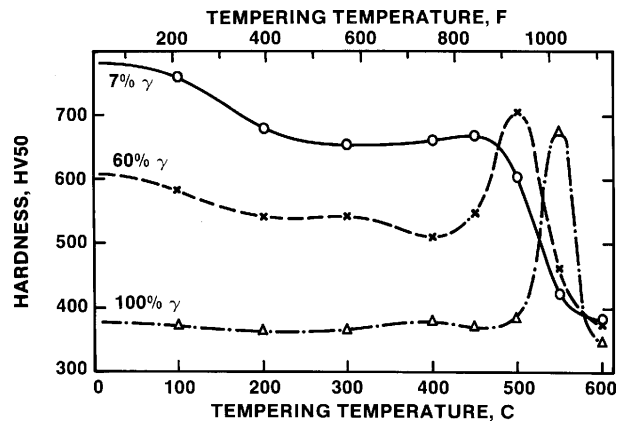


Fig. 3.22. Relationship between tempering temperature and retained austenite content in a high-Cr iron containing 1.3%C-17%Cr-0.9%Mn.

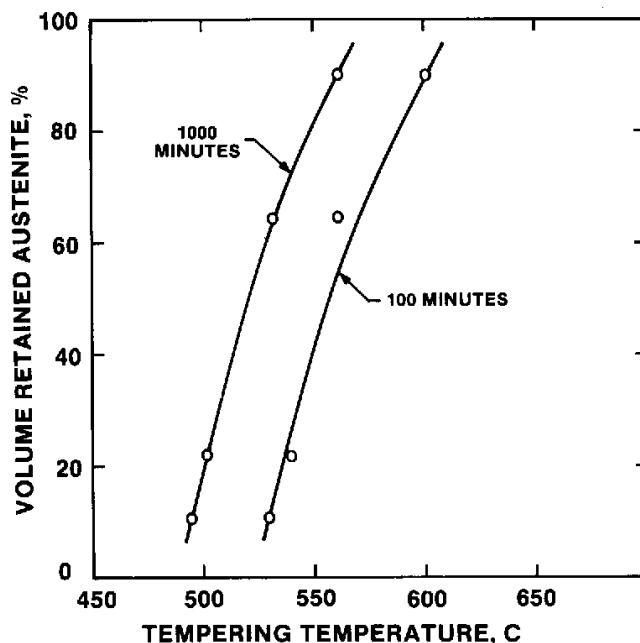


Fig. 3.23. Relationship between initial retained austenite content and the tempering time and temperature required to obtain maximum hardness for an iron containing 1.6%C-16%Cr-3%Mo.

amount of austenite remaining after refrigeration. The data also indicated that refrigeration was only effective at temperatures down to  $-100^{\circ}\text{C}$  ( $-148^{\circ}\text{F}$ ). Below this temperature, no significant additional transformation is likely.

Refrigeration is commonly performed using liquid nitrogen. In the heat treatment of any white cast iron, care must be taken to avoid cracking by thermal shock. Cracking is less likely to occur during refrigeration when castings are *sprinkled* with liquid nitrogen, rather than *immersed*. Most practices leak liquid nitrogen into the chamber holding the castings, and cooling is generally accomplished by conduction of heat to the cold nitrogen gas. The time at temperature is not very critical, other than to assure the castings reach the desired control temperature. Soak times at temperature are typically one-half hour.

### Annealing Treatment

Castings can be annealed to make them more machinable, either by subcritical annealing or a full anneal. Subcritical annealing is accomplished by pearlitizing, via soaking in the narrow range between  $690$  and  $705^{\circ}\text{C}$  ( $1274$  and  $1301^{\circ}\text{F}$ ) for an extended time (depending on the alloy). Subcritical annealing will produce hardness in the range  $400$  to  $450$  HB. Alloys with extreme hardenability will not respond well to subcritical annealing, taking many hours or even days.

Improved machinability can be achieved with full annealing, whereby castings are heated in the range  $955$  to  $1010^{\circ}\text{C}$

( $1750$  to  $1850^{\circ}\text{F}$ ) followed by slow cooling to  $760^{\circ}\text{C}$  ( $1400^{\circ}\text{F}$ ), or just below the  $A_1$  temperature and holding at this temperature for  $10$ – $50$  hours, depending on composition, to ensure complete transformation to ferrite and carbides.

Softening can also be achieved by combining subcritical and supercritical heat treatment. That is, by subcritical heat treatment below the  $A_1$  temperature to pearlitize the structure, followed by heating just above the  $A_1$  temperature, then slow cooling through the  $A_1$  temperature to reform pearlite.

Annealing does not affect the shape of the eutectic carbides or the potential for subsequent hardening. The guidelines used for hardening as-cast castings also apply to annealed castings.

### Stress-Relief Treatment

The practical meaning of toughness for an AR white iron is survival without breakage in a particular application. The conditions resulting in fracture can be from single or repeated impact, or from static or dynamic bending stresses. Here, the inherent low ductility of the iron is often overridden by the presence of internal or residual stresses due to shape and heat treatment of the casting. So, the integrity of the casting and the care and control used in heat treatment are decisive factors in service performance. High residual stresses generated by either steep temperature gradients in cooling, or by differential transformation in various sections of the casting while cooling through the martensitic transformation range can detract from the apparent ductility and strength of the casting in service.

Very little information is available on the amount of stress relief that occurs with tempering. The predominant stresses present in heat-treated castings develop as a result of the volume change accompanying the austenite-to-martensite transformation. Low-temperature tempering, in the range of  $205$  to  $235^{\circ}\text{C}$  ( $401$  to  $455^{\circ}\text{F}$ ), is particularly desirable because a substantial improvement ( $20\%$ ) in fracture toughness occurs when tempering the martensite phase. Tempering at temperatures sufficient to significantly relieve stresses, i.e., above

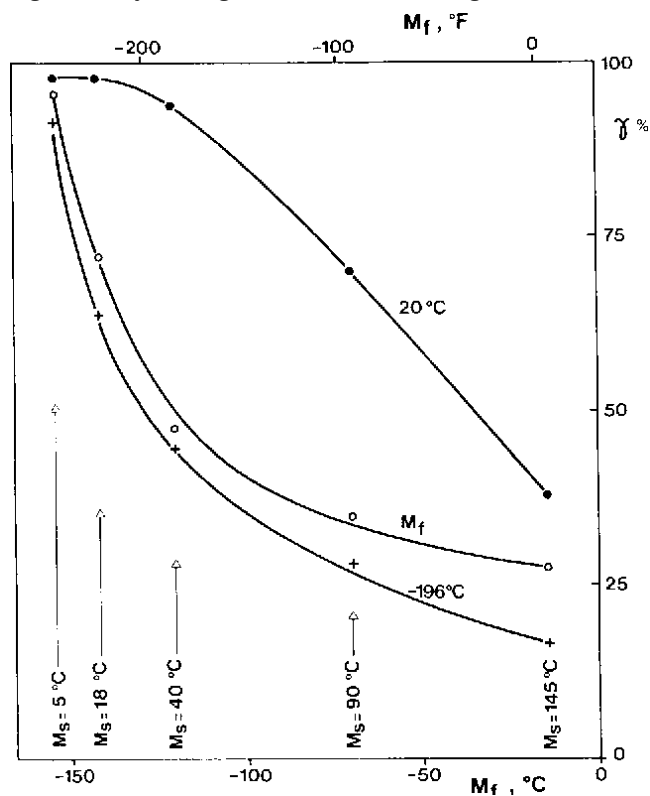


Fig. 3.24. Relationship between initial C content of austenite, indicated by  $M_s$  temperature, and retained austenite content after cooling from the solution treatment to ambient temperature, apparent  $M_f$  temperature, and  $-196^{\circ}\text{C}$  for a Ni-Hard 4 alloy.

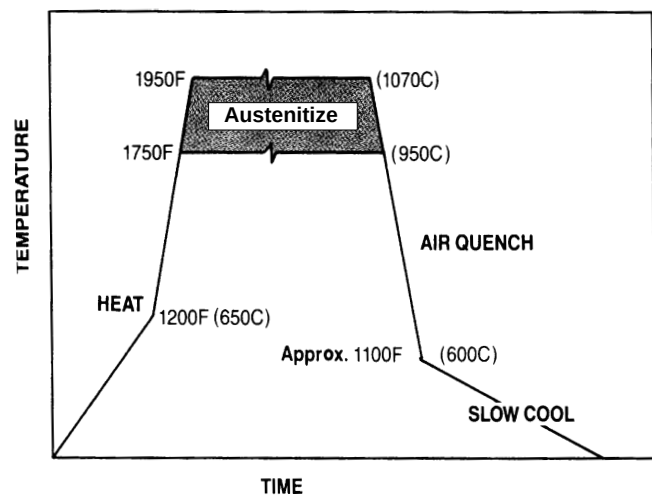


Fig. 3.25. Heat-treatment schedule for supercritical heat treatment of Ni-Hard 4 and the high-Cr white cast irons.

565C (1050F), can substantially reduce abrasion resistance. Therefore, it is much more desirable to minimize the development of cooling stresses and transformational stresses via slow cooling from 482C (900F) to ambient temperature. This is accomplished by slow, controlled cooling to minimize temperature gradients and differential transformation in the casting. Often it is recommended that castings be left in the mold to below 60C (140F) to avoid high residual stresses.

## Practical Guidelines for Heat Treatment

Toughness and abrasion resistance are improved by heat treatment to a martensitic microstructure. Figure 3.25, which illustrates the process, emphasizes the importance of slow heating to 650C (1202F) to avoid cracking. For complex shapes, a maximum rate of 30°C/h (50°F/hr) is recommended. Simple shapes and fully pearlitic castings can be heated at faster rates. The heating rate can be accelerated above red heat.

**Austenitization**—There is an optimum austenitizing temperature to achieve maximum hardness (Fig. 3.26), which varies for each composition. The austenitizing temperature determines the amount of C that remains in solution in the austenite matrix. Too high a temperature increases the stability of the austenite—the higher retained austenite content reduces hardness. Low temperatures result in low-C martensite, reducing both hardness and abrasion resistance. Because of this sensitivity to temperature, furnaces that can produce accurate and uniform temperatures are most desirable. The successful heat treatment produces austenite destabilization by precipitation of fine secondary  $M_7C_3$  carbides within the austenite matrix, as illustrated in Fig. 3.13.

Class II irons containing 12–20% Cr are austenitized in the temperature range 950–1010C (1742–1850F). Class III irons containing 23–28% Cr are austenitized in the temperature range 1010–1090C (1850–1994F).

Castings should be held at temperature long enough to accomplish equilibrium dissolution of C and Cr carbides to ensure proper hardening response. A minimum of four hours at temperature is necessary. For heavy sections, the rule of one hour per inch of section thickness, when the castings reach temperature, is usually adequate. For castings that are fully pearlitic prior to heat treatment, the holding time at temperature can be reduced.

**Quenching**—Air quenching (vigorous fan cooling) the castings from the austenitizing temperature to below the pearlite temperature range, i.e., to between 550 and 600C (1022 and 1112F), is highly recommended. The subsequent cooling rate should be substantially reduced to minimize stresses; still-air or even furnace cooling to ambient temperature is common. Complex and heavy-section castings are often placed back into the furnace, which is at 550 to 600C (1022 to 1112F), and allowed sufficient time to reach uniform temperature within the casting. After the temperature is equalized, the castings are either furnace or still-air cooled to ambient temperature.

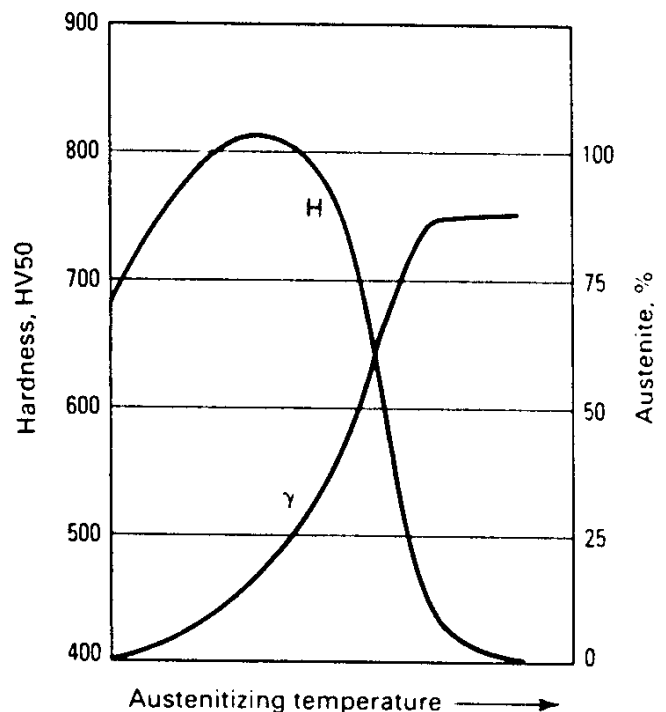


Fig. 3.26. Relationship between austenitizing temperature and hardness, illustrating the optimum hardening temperature for the alloy: H = hardness;  $\gamma$  = austenite..

**Tempering**—Castings can be put into service in the hardened (as-cooled) condition without further tempering or subcritical heat treatments; however, tempering in the range of 200 to 230C (392 to 446F) for two to four hours is recommended to restore some toughness in the martensitic matrix and to further relieve residual stresses. The microstructure after hardening always contains retained austenite in the range of 10 to 30%. Some retained austenite will be transformed following tempering at low temperatures, but if spalling is a problem, higher temperature tempering should be used to further reduce austenite content.

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# Chapter 4

## Alloying for Hardenability

### Hardenability Concepts

The metallic matrix of the AR cast irons consists of various combinations of the microconstituents *austenite*, *pearlite*, *bainite* and *martensite*. Martensite is by far the hardest of these microconstituents. Pearlite, bainite and martensite all form from austenite as it transforms upon cooling below the  $A_1$  (critical) temperature. Each microconstituent forms at a different temperature: pearlite forms just below the  $A_1$  temperature; bainite forms at intermediate temperatures; and martensite at the lowest temperatures. Martensite forms only if the austenite has not already transformed to pearlite or bainite.

Maximum hardness is attained when the matrix microstructure is cooled rapidly enough to develop a fully martensitic microstructure. The process of obtaining martensite by rapid cooling is described as the *hardening process*. Hardenability is defined as the ability to harden by rapid cooling. The term hardenability has also been defined by Siebert, Doane and Breen (1977) as “the capacity to transform partially or completely from austenite to some percentage of martensite at a given depth when cooled under some given conditions.”

Hardenability is a very important concept in the heat treatment of ferrous alloys and is a key factor in the selection of the alloy composition of AR cast irons.

### Phase Transformation

An understanding of the transformation behavior of austenite is critical to understanding the hardening process. The austenite matrix is stable at temperatures down to the  $A_1$  temperature. With slow cooling below the critical ( $A_1$ ) temperature, austenite will transform to pearlite or bainite (microconstituents consisting of an aggregate of ferrite plus carbide). Pearlite forms just below the critical temperature. Bainite forms at intermediate temperatures between 204 and 427°C (400 and 800°F). Martensite generally forms below 204°C (400°F).

When decomposition of austenite to pearlite (or bainite) takes place, both the ferrite phase and carbide phase form simultaneously from the austenite grains. The transformation products form in alternate layers of ferrite and carbide and the rate of growth of these two phases is dependent on the diffusion of carbon and the rearrangement of metallic atoms to form the new phases. When viewed in a microscope, alternate layers of cementite ( $\text{Fe}_3\text{C}$ ) and ferrite can be seen. The formation of pearlite (or bainite) from austenite is a

nucleation and growth process. With all nucleation processes, some time is required before nucleation occurs. Growth of the ferrite and carbide phases is controlled by diffusion and, thus, is also a time-dependent reaction.

Upon cooling to temperatures below 204°C (400°F), austenite will transform to martensite. Martensite is often referred to as another phase having an acicular structure, which means needle-like. It is important to recognize that the martensite reaction is not a time-dependent reaction, like pearlite formation, but occurs by a different transformation mechanism than pearlite or bainite. Martensite formation is not dependent on diffusion but, rather, begins to form when a definite temperature (martensite “start”,  $M_s$ ) is reached. It continues to form only with further cooling. It is completely transformed at the martensite “finish” ( $M_f$ ) temperature. Martensite forms over a temperature range of about 150°C (300°F). If the cooling is interrupted within this temperature range, the martensite transformation ceases and some austenite remains.

### Kinetics of Phase Transformation

When the metal is cooled rapidly, the transformation of austenite to pearlite (or bainite) can be incomplete and the remaining austenite will transform to martensite. With sufficiently rapid quenching, the pearlite and bainite reactions can be completely suppressed, and a fully martensitic structure can be obtained. The degree to which cast iron can be made hard is determined by the amount of martensite that forms, either in the as-cast microstructure or after heat treatment.

As alloying elements are added to cast irons, the kinetics of pearlite and bainite phase transformations are usually slowed and the hardenability (the ability to obtain martensite) is increased. That is, alloying elements increase the time for the nucleation and growth of pearlite. This makes it possible to avoid transformation to pearlite (and bainite) and to obtain martensite on quenching at slower cooling rates. Thus, even heavy-section castings can be hardened by air-cooling.

The most common alloying elements used to increase hardenability in abrasion-resistant cast irons are chromium, nickel, manganese, molybdenum and copper. Each alloying element contributes to hardenability differently; some elements are more effective than others. Certain elements, like molybdenum, have synergistic effects with other elements, further augmenting hardenability. Alloying elements that reduce C solubility in austenite, such as the element Si, may actually reduce hardenability.

Hardenability is related to the alloy content and carbon content. Due to the fact that some Cr is consumed in forming eutectic  $M_7C_3$  carbides, the amount of Cr contributing to the hardenability of the alloy is considerably less than the bulk Cr content. Irons of higher Cr:C ratios have greater Cr contents in the metallic matrix and, thus, have greater hardenability. Likewise, in heat treatment, the depletion of C and Cr from austenite destabilization (carbide precipitation) causes a reduction in hardenability. As a result, lower austenitizing temperatures produce lower hardenability.

### **M<sub>s</sub> Temperature and Retained Austenite**

As described previously, martensite formation occurs over a relatively wide temperature range. For some alloys the M<sub>s</sub> temperature is sufficiently close to room temperature that little martensite forms before reaching room temperature. As a result, the casting can contain significant quantities of retained austenite in the as-cast condition, and even in the as-quenched condition after heat treatment. Of course, refrigeration below room temperature will cause further transformation to martensite and further reduction in retained austenite.

In many applications, elevated levels of retained austenite can be detrimental to the performance of the casting. In those applications, the austenitic-martensitic structure of as-cast irons is entirely unacceptable. Furthermore, it is necessary to select the alloy composition and heat treatment that produces the least amount of retained austenite after hardening, i.e., a composition and hardening temperature that produces an M<sub>s</sub> temperature of 150C (300F) or above.

Virtually all alloying elements lower the M<sub>s</sub> temperature, but carbon has the greatest effect. Several different investigators have studied the influence of C and alloying elements on M<sub>s</sub> temperature, and numerous equations describing the effects of each element have been developed. A few of these equations are as follows:

$$M_s (^\circ F) = 930 - 570(\%C) - 60(\%Mn) - 50(\%Cr) - 30(\%Ni) - 20(\%Mo) - 20(\%Si) - 20(\%W) \quad \text{(After Payson \& Savage, 1944)} \quad (4.1)$$

$$M_s (^\circ C) = 539 - 423(\%C) - 30.4(\%Mn) - 12.1(\%Cr) - 17.7(\%Ni) - 7.5(\%Mo) \quad \text{(After Andrews, 1965)} \quad (4.2)$$

Note that, among the alloying elements, Mn, Cr and Ni appear to have the greatest influence, followed by Mo and W.

The influence of Cr is well demonstrated by the fact that fully austenitic microstructures are readily obtained in as-cast high-Cr iron castings. The influence of Ni on promoting retained austenite is equally illustrated by the high retained austenite content in Ni-Hard castings.

Retained austenite is a concern in heat-treated high-Cr iron castings, as well. Many producers are careful to avoid excessive levels of Ni and Mn in castings that are exposed to conditions with repeated impact. While elevated levels of retained austenite are not particularly harmful to abrasion resistance, retained austenite can significantly reduce resistance to spalling.

## **Alloying for As-Cast Hardenability**

Many castings are used in the as-cast condition; others are designed to be subsequently heat treated. With Ni-Hard castings, the desired as-cast microstructure is always martensitic or austenitic-martensitic. With high-Cr and Cr-Mo iron castings, the desired as-cast microstructure is often pearlite.

Castings that will be used in the heat-treated condition, still require that the as-cast microstructure be controlled. As-cast microstructures can vary from being fully pearlitic to being largely austenitic or austenitic-martensitic. With as-cast pearlitic microstructures, the castings generally have lower residual stresses and are less likely to crack. Furthermore, they are more machinable and usually exhibit a better response to heat treatment.

### **Nickel-Chromium (Ni-Hard) Irons**

With Ni-Hard irons, the nickel content must be balanced with section size. For lighter sections, lower Ni is required. Excess Ni produces elevated amounts of retained austenite, less martensite, and overall lower hardness. For heavy cast sections, more Ni is required to avoid pearlite and bainite. As section size increases further, more alloy is required and, eventually, Mo is added to develop maximum hardenability. Recommended Ni-Hard compositions are given in Table 4.1 for castings of varying section size. With higher Si content, lower Ni:Cr ratios are required.

There is a need to balance Ni, Si and Cr to avoid graphite in Ni-Hard castings. Graphite formation is greatly exacerbated by higher C content. To avoid graphite formation, it is important to control the Ni:Cr ratio.

### **High-Chromium White Irons**

A wide range of matrix microstructures can be obtained by varying the alloy content of the high-Cr irons and by applying heat treatment. With suitable alloying, the phase changes that occur in the high-Cr irons, as they are cooled in the mold, are extremely sluggish. With the exception of castings that com-

**Table 4.1**  
**Guidelines for Ni and Cr Contents of Ni-Hard 1 and Ni-Hard 2 as a Function of Section Size**

| Section Size (mm) | Ni-Hard 1    |              |               |               | Ni-Hard 2    |              |               |               |
|-------------------|--------------|--------------|---------------|---------------|--------------|--------------|---------------|---------------|
|                   | Sandcast %Ni | Sandcast %Cr | Chillcast %Ni | Chillcast %Cr | Sandcast %Ni | Sandcast %Cr | Chillcast %Ni | Chillcast %Cr |
| <12               | 3.8          | 1.6          | 3.3           | 1.5           | 4.0          | 1.5          | 3.5           | 1.4           |
| 12–25             | 4.0          | 1.8          | 3.6           | 1.7           | 4.2          | 1.7          | 3.8           | 1.5           |
| 25–50             | 4.2          | 2.0          | 3.9           | 1.9           | 4.4          | 1.8          | 4.1           | 1.6           |
| 50–75             | 4.4          | 2.2          | 4.2           | 2.1           | 4.6          | 2.0          | 4.4           | 1.8           |
| 75–100            | 4.6          | 2.4          | 4.5           | 2.3           | 4.8          | 2.2          | 4.7           | 2.0           |
| >100              | 4.8          | 2.6          | 4.8           | 2.5           | 5.0          | 2.4          | 5.0           | 2.2           |

pletely transform to pearlite on mold cooling, large amounts of retained austenite usually exist in high-Cr iron castings. This austenite is generally supersaturated, in respect to C, and it is the high Cr and C contents in the austenite that cause it to be retained at room temperature. The degree of supersaturation can be minimized by delaying cooling of the casting from the solidus. Slow cooling allows precipitation of carbides and thus a reduction in the stability of the austenite. This is the basis of the controlled cooling and tempering approach used for certain classes of castings.

Experimental data indicate that the hardenability of as-cast high-Cr irons is substantially greater than that of irons that are subsequently heat treated. It has proved difficult to quantify the effects of alloying on as-cast high-Cr irons, but there are data available from the investigations of Maratray (1970) and Gundlach (1974), which suggest that the effects of alloying elements on as-cast irons are similar to, but not quantitatively the same as, those observed for heat-treated irons.

Many complex cast sections, such as slurry pump components, are often used in the as-cast austenitic/martensitic condition to avoid the possibility of cracking and distorting during heat treatment. To avoid forming pearlite during mold cooling, alloying additions are usually required. As the C content is increased, more Cr is consumed to form additional carbide and, therefore, larger alloying additions are required. Table 4.2 presents a guide to appropriate alloying for the prevention of pearlite in the various classes of mold-cooled as-cast irons.

*As-cast Austenitic/Martensitic Microstructures*—Solidification in the hypoeutectic alloys occurs by the formation of austenite dendrites followed by the eutectic formation of austenite and  $M_7C_3$  Cr carbides. Under equilibrium conditions, additional Cr carbide will precipitate from the austenite matrix on cooling from the eutectic to the critical temperature, about 760C (1400F) and transformation to ferrite and carbide occurs on subsequent cooling. However, when cooling under nonequilibrium conditions, such as encountered in most commercial castings, the austenite becomes supersaturated in C and Cr. Due to elevated C and Cr contents, a metastable

austenitic microstructure normally develops, provided pearlitic and bainitic transformations have been inhibited. With sufficient alloying with Mo, Mn, Ni and Cu, pearlitic transformation can be avoided in virtually any cast section.

With lighter cast sections, the mold-cooled structure is generally austenitic. Austenitic/martensitic structures are obtained as-cast in heavy-section castings that cool slowly in the mold. With slow cooling rates, austenite stabilization is incomplete and partial transformation to martensite occurs. But in these castings, martensite is mixed with large amounts of retained austenite and, therefore, hardness levels are lower than can be achieved in heat-treated martensitic castings. These castings must contain sufficient alloy to suppress pearlite on cooling. Some compositions (higher Si alloys) have been developed to assist martensite formation in refrigeration treatments. Subcritical annealing has also been employed to reduce austenite content and, at the same time, increase hardness and toughness.

## Alloying for Hardenability in Heat Treatment

To obtain maximum hardness and abrasion resistance, martensitic matrix structures must be produced by full heat treatment. The casting must contain sufficient alloy to avoid pearlite formation on cooling from the heat treatment temperature.

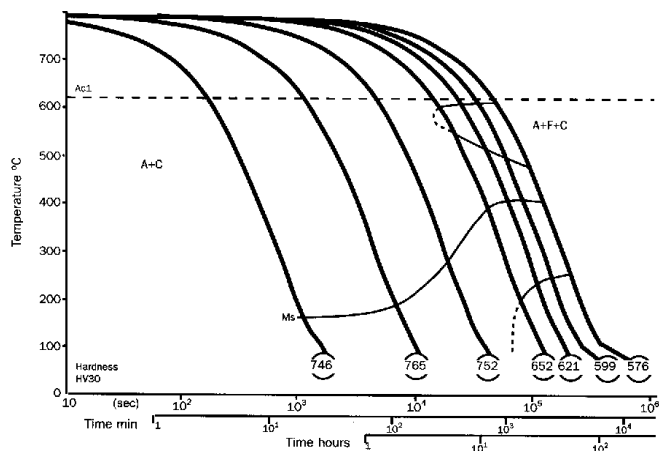
Chromium, nickel, copper, manganese and molybdenum are used to delay transformation to pearlite on heat treatment and Cias (1974) and Maratray (1970) have shown how variations in these elements affect continuous cooling transformation behavior in high-Cr white irons.

Typical continuous cooling transformation (CCT) diagrams illustrating the hardenability in several common alloys including Ni-Hard 1, Ni-Hard 4 and high-Cr iron alloys from Classes II and III are presented in Figs. 4.1a–4.1g. The number encircled at the end of each cooling curve represents the as-quenched hardness (HV).

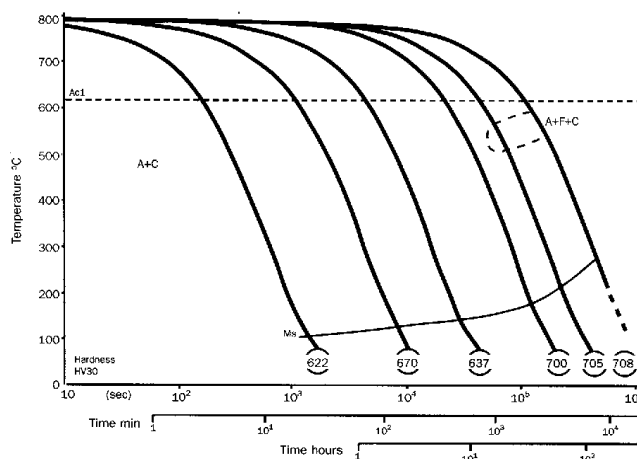
**Table 4.2**  
**Minimum Alloy Content to Avoid Pearlite in**  
**Mold-Cooled Castings of Indicated Effective Section Size**  
**(Plate Thickness or Radius of Rounds)**

| ASTM A 532<br>Class | % Cr <sup>(1)</sup> | % C <sup>(1)</sup> | —Plate Thickness or Radius of Rounds— |                  |                        |
|---------------------|---------------------|--------------------|---------------------------------------|------------------|------------------------|
|                     |                     |                    | 25 mm<br>(1 in.)                      | 50 mm<br>(2 in.) | 100 mm<br>(4 in.)      |
| IIB, IIC            | 14–18               | 2.0                | 1.0 Mo                                | 1.5 Mo           | 1.5 Mo + 1.0 (Ni + Cu) |
|                     |                     | 3.5                | 2.0 Mo                                | 2.5 Mo           | 2.5 Mo + 1.0 (Ni + Cu) |
| IID, IIE            | 18–23               | 2.0                | 0.5 Mo                                | 1.0 Mo           | 1.0 Mo + 1.0 (Ni + Cu) |
|                     |                     | 3.2                | 1.5 Mo                                | 2.0 Mo           | 2.0 Mo + 1.0 (Ni + Cu) |
| IIIA                | 23–28               | 2.0                | —                                     | 0.5 Mo           | 1.0 Mo                 |
|                     |                     | 3.0                | 1.0 Mo                                | 1.5 Mo           | 1.5 Mo + 1.0 (Ni + Cu) |

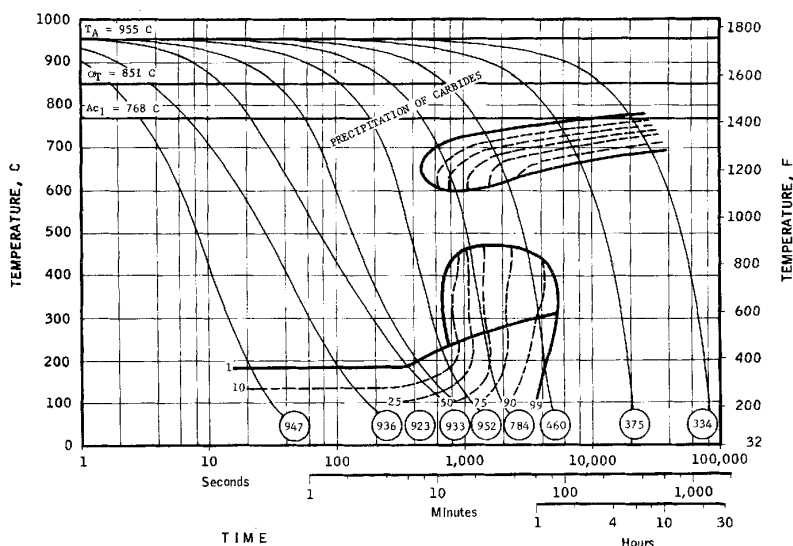
(1) In base irons containing 0.6% Si and 0.8% Mn.



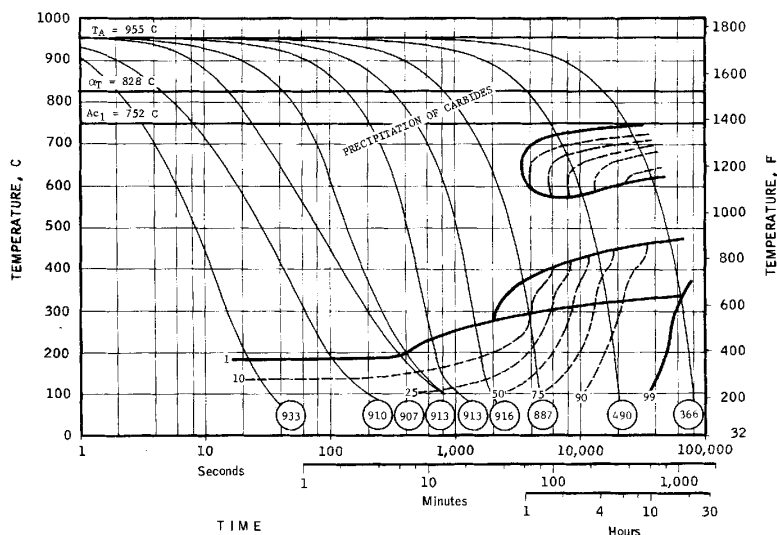
**Fig. 4.1a. CCT diagram for Ni-Hard 1 containing 3.1% C, 0.6% Si, 4.0% Ni, 2.3% Cr and 0.5% Mn.**



**Fig. 4.1b. CCT diagram for Ni-Hard 4 containing 3.1% C, 1.2% Si, 5.4% Ni, 8.3% Cr and 0.6% Mn.**



**Fig. 4.1c. CCT diagram for high-Cr iron containing 2.9% C, 0.6% Si, 0.8% Mn, 17.6% Cr, 0.5% Mo.**



**Fig. 4.1d. CCT diagram for high-Cr iron containing 2.9% C, 0.6% Si, 0.8% Mn, 17.5% Cr, 1.5% Mo.**

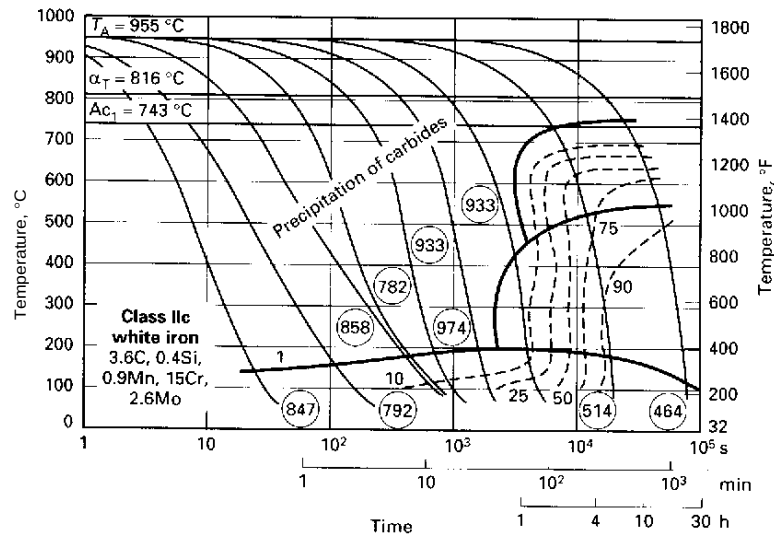


Fig. 4.1e. CCT diagram for high-Cr iron containing 3.6% C, 0.4% Si, 0.9% Mn, 15% Cr, 2.6% Mo.

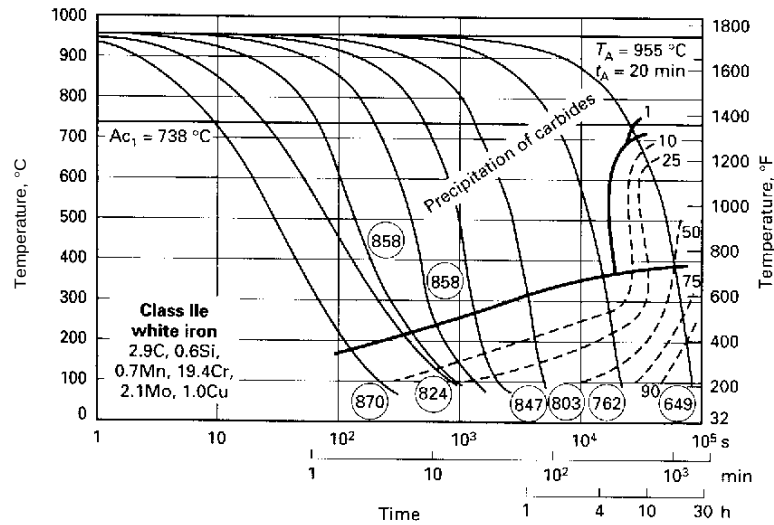


Fig. 4.1f. CCT diagram for high-Cr iron containing 2.9% C, 0.6% Si, 0.7% Mn, 19.4% Cr, 2.1% Mo, 1.0% Cu.

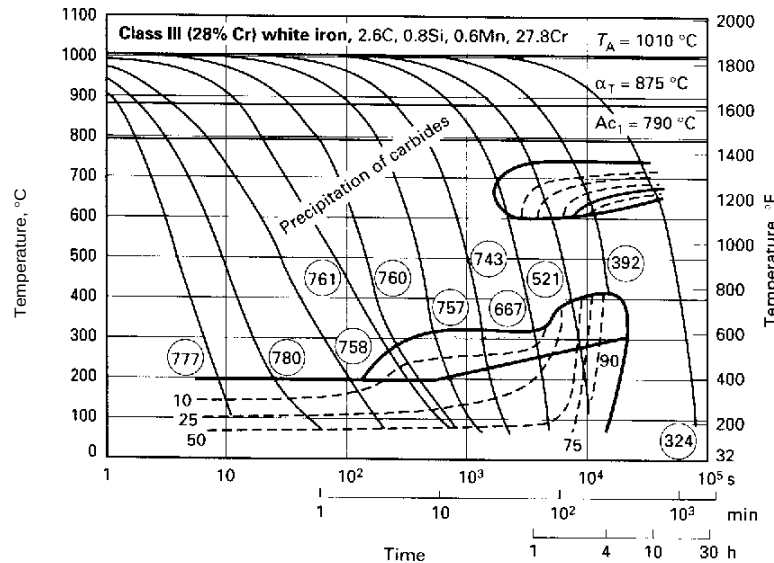


Fig. 4.1g. CCT diagram for high-Cr iron containing 2.6% C, 0.8% Si, 0.6% Mn, 27.8% Cr.

## Selecting Alloy Compositions

Alloying must be sufficient to ensure that a pearlite-free microstructure is obtained during heat treatment. Of necessity, the heat treatment requires an air quench from the austenitizing temperature. Faster cooling rates should not be used, because the casting can develop cracks due to high thermal and/or transformation stresses. Thus, the alloy must have sufficient hardenability to allow air hardening. Over-alloying with Mn, Ni and Cu will lower the  $M_s$  temperature and promote retained austenite, which detracts from resistance to abrasion and spalling.

Dodd, Gundlach and Morton (1982) developed an equation for hardenability of high-Cr irons from CCT data using regression analysis. All irons were austenitized at 955C (1750F) and hardenability is defined as time (in seconds) to reach the pearlite nose. The relationship that incorporated first-order terms for the alloying elements, plus some alloy interaction terms, is:

$$\begin{aligned} \text{Log pearlite time (sec)} = & 1.07 + 0.43(\%Ni) \\ & + 0.44(\%Cu) + 0.25(\%Cr/C) \\ & + 0.59(\%Mo \times \%Mn) + 0.56(\%Mo \times \%Ni) \end{aligned} \quad (4.3)$$

This equation implies that Ni and Cu have strong singular contributions to hardenability, but closer scrutiny of the data indicates that additions of Ni and Cu, either by themselves or as a combination (Ni + Cu), are not as powerful as this equation predicts (Fig. 4.2). Data points for Ni- and Cu-alloyed irons identified in Fig. 4.2 fall well above the curve and indicate that the equation gave an overly optimistic prediction of hardenability for those alloys. It was only when Mo was added (minimum of 0.5%) that these elements showed strong contributions to hardenability. This synergistic effect with Mo is already partially expressed in the “Mo × Ni” interaction term of the equation, but it also occurs with copper.

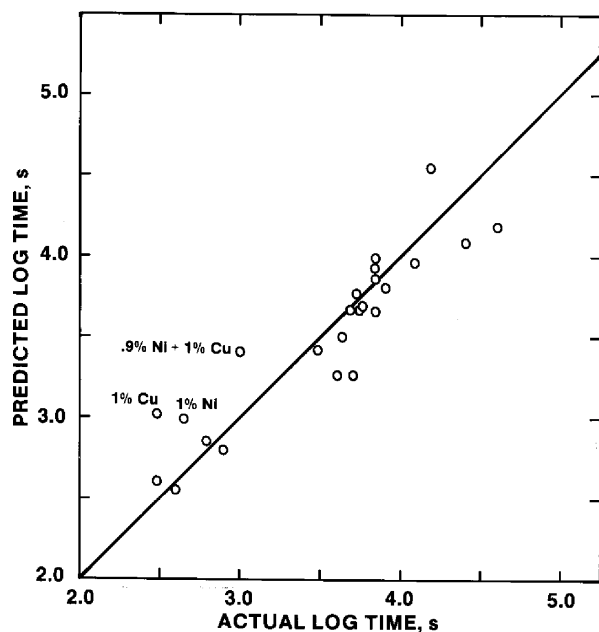


Fig. 4.2. Correlation of time to reach the pearlite nose and the time predicted by regression analysis (Eq. 4.3).

## Correlation of Composition With Section Size

For each composition there is a critical cooling rate required to avoid the pearlite nose of the CCT diagram. Therefore, it is necessary to know the cooling rate of the critical section of a casting that must be hardened. Ideally, this is determined in heat treatment experiments by instrumenting the casting with thermocouples. Alternatively, it can be estimated by observing the time it takes for the casting to reach “black heat” (no color) on quenching. Once the critical cooling rate is determined, one can go to Equation 4.3 to estimate the alloying required to develop adequate hardenability.

Since almost all cast sections must be air-cooled to avoid cracking and high levels of residual stress, it is desirable to know the rates for cooling castings in air. The cooling rates of steel rounds of various diameters cooled in still air from the austenitizing temperature is offered by Siebert et al. (1977) as a guide (Fig. 4.3).

The following equation has been derived from the data of Fig. 4.3. It gives a conservative estimate of the time to cool in air to the pearlite nose for plate sections ranging from 1 inch to 12 inches thick:

$$\text{Log } T = 0.90 \text{ Log } t - 1.0 \quad (4.4)$$

where  $T$  = section thickness, mm

$t$  = time to pearlite nose, seconds

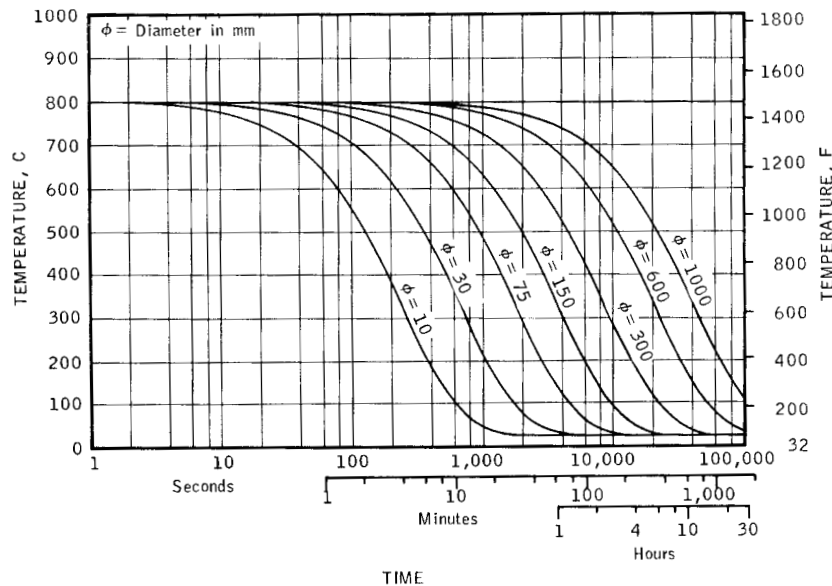
From this relationship a correlation of composition with section size can be estimated by combining Equations 4.3 and 4.4 with the following results:

$$\begin{aligned} \text{Log } T = & 0.06 + 0.39(\%Ni) + 0.40(\%Cu) + 0.23(\%Cr/C) \\ & + 0.53(\%Mo \times \%Mn) + 0.50(\%Mo \times \%Ni) \end{aligned} \quad (4.5)$$

This relationship is given as a guide to alloy selection. It will help the designer estimate the hardenability of high-Cr white iron alloys. It should not be applied rigorously, but rather, each foundry should test the composition in heat treatment trials to verify that sufficient hardenability is achieved.

Hardenability is known to increase with higher austenitizing temperature, i.e., less austenite destabilization. Therefore, the hardenability predicted by this equation is somewhat conservative. Considering the very deleterious effect pearlite has on hardness, a conservative approach is desirable.

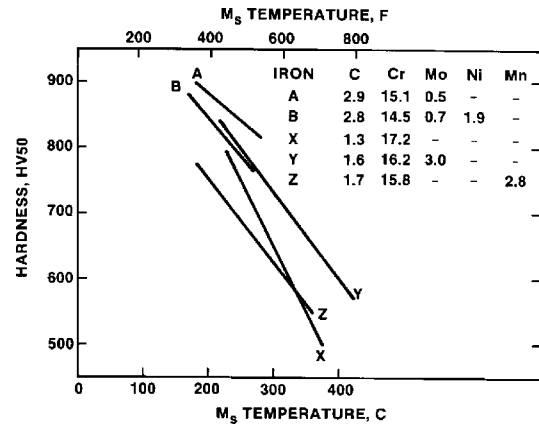
In addition to suppressing or delaying transformation to pearlite, the alloying elements used to obtain hardenability, particularly Ni, Mn and Cu, markedly depress the martensitic transformation temperature range. This must be taken into consideration when designing an alloy, since inappropriate alloying can result in a very low  $M_s$  temperature with large amounts of retained austenite, and/or castings with reduced hardness. Because Ni and Cu promote retained austenite, they are usually restricted to combined levels of 1.2% maximum. Manganese behaves similarly and should be restricted to 1.0% maximum. Figure 4.4 shows the relationship between the maximum obtainable hardness and the  $M_s$  temperature of five heat-treated high-Cr irons. As previously discussed, carbon is



**Fig. 4.3. Air-cooling response of steel products of increasing cross section (center temperature vs. log time).**

more potent than the alloying elements in depressing the  $M_s$  temperature. Thus, an important feature of the metallurgy of the high-Cr-Mo alloy irons is control of the C content of the matrix. This is invariably dramatically higher than consideration of the equilibrium diagrams would suggest, due to the very sluggish diffusion of C and Cr in these high-Cr alloys. This topic has been discussed, in detail, by Maratray (1982).

Table 4.3 is offered as a guide to alloying for air quenching heat-treated castings of various sections. These recommendations for alloying specific ASTM grades are based on studies of continuous-cooling transformation of high-Cr irons. The summary provides specific references to available data. The regression Equations 4.3, 4.4 and 4.5 should be consulted when considering nonstandard grades and thick section castings.



**Fig. 4.4. Relationship between hardness and  $M_s$  temperature for five alloyed high-Cr irons that have been austenitized at various temperatures and quenched to room temperature.**

**Table 4.3**  
**Minimum Alloy Content to Avoid Pearlite in**  
**Heat-Treated (Air-Quenched) Castings of Indicated Effective Section Size**  
**(Plate Thickness or Radius of Rounds)**

| ASTM A 532 Class | % Cr <sup>(1)</sup> | % C <sup>(1)</sup> | Plate Thickness or Radius of Rounds |                        |                                       |
|------------------|---------------------|--------------------|-------------------------------------|------------------------|---------------------------------------|
|                  |                     |                    | 25 mm (1 in.)                       | 50 mm (2 in.)          | 100 mm (4 in.)                        |
| IIB, IIC         | 14-18               | 2.0                | 1.5 Mo                              | 1.5 Mo + 0.5 (Ni + Cu) | 2.0 Mo + 1.0 (Ni + Cu)                |
|                  |                     | 3.5                | 3.0 Mo                              | 2.0 Mo + 1.0 (Ni + Cu) | 2.5 Mo + 1.0 (Ni + Cu) <sup>(2)</sup> |
| IID, IIE         | 18-23               | 2.0                | 1.0 Mo                              | 2.0 Mo                 | 2.0 Mo + 0.5 (Ni + Cu)                |
|                  |                     | 3.2                | 1.5 Mo                              | 2.0 Mo + 0.7 (Ni + Cu) | 2.0 Mo + 1.2 (Ni + Cu) <sup>(2)</sup> |
| IIIA             | 23-28               | 2.0                | 0.5 Mo                              | 1.5 Mo                 | 1.5 Mo + 0.5 (ni + Cu)                |
|                  |                     | 3.0                | 1.5 Mo                              | 1.5 Mo + 0.6 (Ni + Cu) | 1.5 Mo + 1.2 (Ni + Cu) <sup>(2)</sup> |

(1) In base irons containing 0.6% Si and 0.8% Mn.

(2) Ni and Cu promote retained austenite and should be restricted to combined levels of 1.2% maximum; Mn behaves similarly and should be restricted to 1.0% maximum. Source: ASM Handbook, vol. 15, 9th Ed., p. 683.

## Lean Alloy Approach

For castings that will be heat treated, the desired mold-cooled structure is often pearlite. This softer structure will facilitate removal of gates and risers and minimize the transformational and thermal stresses that cause cracking. The pearlitic structure will also shorten the response to heat treatment. Alloys of low hardenability that transform completely to pearlite on mold cooling need only be austenitized for one hour at temperature to obtain martensite on subsequent cooling. Whereas, the precipitation of carbides from supersaturated austenite can take up to four hours at the austenitizing temperature to achieve sufficient destabilization.

Careful design of alloy composition will assure that a substantially pearlitic structure develops in the casting after mold cooling, but will provide enough hardenability to prevent pearlite formation during subsequent heat treatment. Heavy-section castings made pearlitic by such an alloy content can oftentimes be removed from the mold, once the castings have reached black heat (i.e., below 650C).

It should be emphasized that overalloying, as well as underalloying, should be avoided. If Mn, Ni or Cu are added in excessive quantities to prevent pearlite formation in heavy sections, too much austenite will be retained in light sections. As a result, air-cooled hardness will be lower than desired in the thin sections and both wear resistance and resistance to spalling will be reduced.

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# Chapter 5

## Morphology, Microstructures and Properties

Abrasion-resistant cast irons are one of the most complex microcomposite engineering materials available to wear-resistance technologists. Their range of mechanical, thermophysical and wear properties provide a broad range of options in meeting the challenges of a particular environment. Abrasion-resistant irons can be tough or they can be brittle; they can have no corrosion resistance or they can be as “stainless” as 304 *stainless steel*—it all depends on the morphology and microstructure. (See Glossary for *Stainless Steel*.)

The aim of this chapter is to demonstrate how the microconstituents of AR irons combine to form micro-composite materials having wide-ranging properties. It is this combination of distinct constituents and phases that makes AR irons so useful, yet so complex. This chapter is devoted to the study of these individual constituents and phases. From this basis, AR irons are discussed as a combination of two constituents: carbide and matrix. Within each constituent, a variety of phases can be found. A brief list of the various constituents’s hardnesses is provided in Appendix A. For example, the metallic matrix can be austenitic, martensitic, pearlitic or, typically, a combination of austenite and martensite.

For each class of AR cast iron, brief discussions are provided about mechanical and physical properties. General relationships will be discussed to highlight the relationships between microstructure and properties. This discussion is not meant to be exhaustive, but only to highlight general guidelines. If more specific data is needed, Appendix B should be consulted.

### Carbide Constituents

The world of AR irons revolves around the type, volume fraction and morphology of carbides present in the iron. Table 5.1 provides a quick summary of the different types of carbides that can be found in AR cast irons. For the majority of users, there are only two carbides of interest, the  $M_3C$  and the  $M_7C_3$ . For example, all of the irons defined within the ASTM A532 specifications contain either the eutectic  $M_3C$  or  $M_7C_3$  carbide. It is these carbides that will be covered in detail.

### $M_3C$ Carbide Morphology

Table 5.1 provides a starting point for our understanding of this carbide. What really separates the  $M_3C$  carbide from the other carbides is its morphology. Figures 5.1a–5.1c provide a good look at the plate-like morphology of the  $M_3C$  carbide as

the iron’s composition is shifted from hypoeutectic to hyper-eutectic.

The basic morphology of the carbide remains the same whether it is formed during eutectic solidification (hypoeutectic and eutectic) or formed during primary solidification as primary carbide (hypereutectic). The  $M_3C$  carbide morphology is often described as being composed of thin plate-like structures. These plates can appear to be rather solid or to be riddled with holes. In all cases, the carbide network is continuous. The continuity of these thin plate-like structures is one of the major microstructural features in  $M_3C$ -type AR irons.

Figure 5.2 shows a schematic of how the  $M_3C$  carbide is often observed and described in the literature. If the carbide is captured at just the right angle, it may appear as discrete plates; if captured at another angle, it may appear as a solid mass of carbides punctured with holes. In either case, more careful observation by the SEM would show that the carbide is always continuous and is always composed of the basic building blocks shown in Fig. 5.2.

This fact is still questioned every once in awhile. For example, high cooling rates sometimes appear to produce a discontinuous  $M_3C$  carbide network. Figure 5.3a shows an optical polished plane view that appears to show the  $M_3C$  carbide as being discontinuous. Contrary to the polished plane view, the deep-etched micrograph (Fig. 5.3b) shows how 2D pictures can be misleading; i.e., the  $M_3C$  carbide is still continuous.

The  $M_3C$  carbide can also appear as secondary carbides (carbides formed via a solid-state transformation). In its most famous form, it appears as a mixture of ferrite and  $M_3C$ , i.e., pearlite. In Ni-Hard 4 irons and other irons with low-Cr austenite matrices (e.g., Ni-Hard 1, Ni-Hard 2 and, of course, pearlitic irons), large secondary  $M_3C$  carbides will form within the austenite matrix as the casting cools to room temperature. Figures 5.4a and 5.4b show the morphology of secondary  $M_3C$  carbides in a Ni-Hard 4 dendrite. If anything, the characteristic plate-like  $M_3C$  morphology is shown even clearer in these deep-etched SEM micrographs.

### $M_3C$ Mechanical and Physical Properties

The hardness of the  $M_3C$  carbide is nothing extraordinary. Table 5.1 shows that it is the softest of the carbides, having a hardness range of 800–1000 DPH. (Only the phosphide carbide  $Fe_3P$  is softer, having a hardness of around 500 DPH.) What makes the  $M_3C$  carbide extraordinary is its ability to form in massive quantities. For example, it is common to find

pearlitic and Ni-Hard 1 grades of AR irons containing up to 40% carbide volume fraction (CVF). At such high CVF, the  $M_3C$  carbide dominates the iron's abrasion resistance.

The reactivity or resistance to oxidation of the  $M_3C$  carbide was not found in the literature surveyed by the authors. It can be reasoned that its chemical inertness or reactivity is tied to its Gibbs Free Energy of Formation. The property of *Gibbs Free Energy* (see Glossary) is the energy released during the formation of the compound. If the compound releases a large amount of energy, then a similarly large amount of energy is required to decompose the compound. Figure 5.5 provides some data on the Gibbs Free Energy for the  $M_3C$ ,  $M_7C_3$  and  $M_{23}C_6$  carbides as a function of Cr content. The numbers are

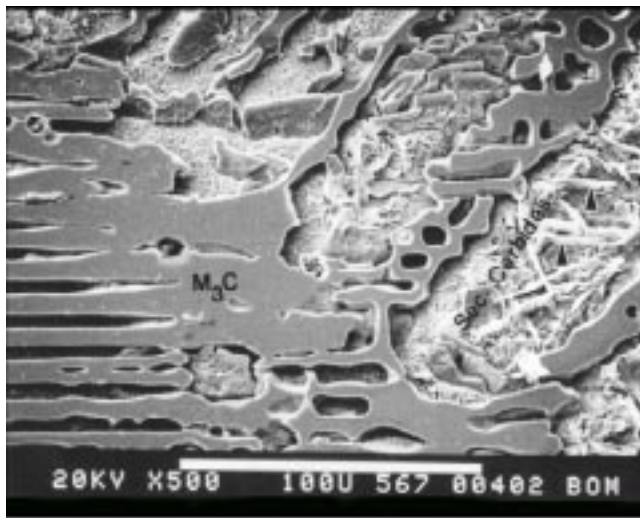
negative, reflecting the amount of energy required to return the compound elements to their original state (e.g.,  $M_3C \rightarrow 3 M$  and 1 C atoms).

Focusing on just the  $M_3C$  carbide, the Gibbs Free Energy of Formation becomes less and less negative as the Cr content decreases. At 0.0% Cr, the carbide becomes unstable as the Gibbs Free Energy approaches zero. This has a direct application in pearlitic irons where the Cr content can be extremely low (<0.5%). In pearlitic irons, depending on the Si and Ni content, the  $M_3C$  carbide (i.e.,  $Fe_3C$  carbide) can be decomposed into graphite at somewhat low heat treatment temperatures, say 600C (1112F). This transformation is at the heart of the *malleable iron* process. (See Glossary for *Malleable Iron*.)

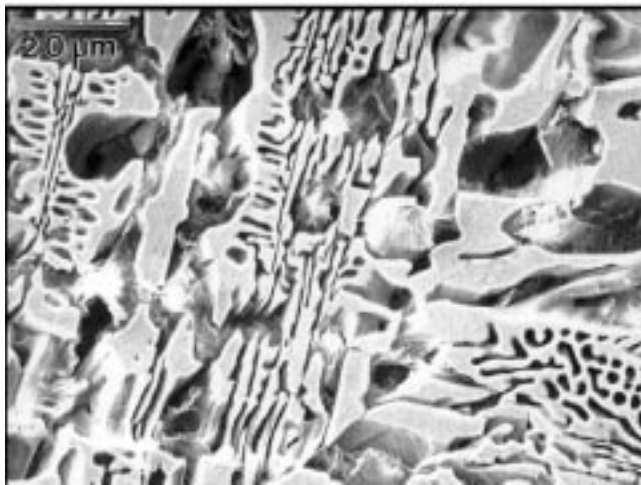
**Table 5.1**  
**Carbides in AR Cast Irons**

| Carbide Type | Chemistry (Wt%)<br>Major Alloying Elements                                                                                                                                                                                                                                                                                                                                                  | %C      | Density (g/cm <sup>3</sup> ) | Crystal-<br>lography               | Mor-<br>phology          | Hardness (DPH) | Gibbs Energy (kJ/mole) |
|--------------|---------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------|---------|------------------------------|------------------------------------|--------------------------|----------------|------------------------|
| $M_3C$       | Majority Fe and, in most Ni-Hards, up to 5% Cr and 1.5% Ni.<br>The $M_3C$ carbide has extensive solubility of Mn and, if 1% Mn is added to the melt, 1% Mn will be present in the carbide. Cr can also dissolve into the $M_3C$ carbide up to 17%. Mo, V and other metal atoms have very limited solubility within the $M_3C$ carbide.                                                      | 6.7–6.9 | 7.4–7.6                      | Orthorhombic                       | Plate-like               | 800–1100       | +1 to –10              |
| $M_7C_3$     | Majority Cr, however the Cr content can vary from 24 to 50%. <sup>(1)</sup> The carbide can also contain several% of Mo (<7%) and V (<30%), depending on the alloy content. Mn has similar solubility as in the $M_3C$ carbide.                                                                                                                                                             | 8.6–8.9 | 6.6–6.9                      | Trigonal (Pseudo-Hexagonal)        | Rod and Blade-like       | 1000–1800      | –10 to –20             |
| $M_{23}C_6$  | Majority Cr with a lower limit of around 60%. Solubility of other elements is assumed to follow that of the $M_7C_3$ and $M_3C$ types, e.g., W<10%, Mo <10% and V <10%.                                                                                                                                                                                                                     | 5.3–5.7 | 7.2–7.7                      | Complex FCC (D8 <sub>4</sub> Type) | Rod-like                 | 1000           | –8 to –15              |
| $Mo_2C$      | In wear irons having several percent Mo, this carbide will commonly appear at the austenite grain boundaries. The presence of other carbide-forming elements, other than Mo, within this carbide has not been confirmed in the literature. At this time, it is assumed to form as a pure carbide, although it is believed to contain some Cr when solidified within a high-Cr AR cast iron. | 5.9     | 9.1                          | Orthorhombic                       | Blade-like               | 1500–1800      | –25 to –30             |
| MC (VC)      | Vanadium carbide is one of the more misunderstood carbides. It will only appear as a eutectic carbide at alloy levels in excess of 4.5%. At lower alloy contents, it will partition into the $M_xC_y$ family of carbides and into the ferrous matrix. V carbides formed in Cr-containing irons can have up to 20% Cr dissolved within the carbide.                                          | 17–20   | 5.6–7.0                      | Complex FCC (B1 Type)              | Rod-like                 | 2000–3000      | –40 to –80             |
| NbC          | Niobium carbide is niobium carbide, i.e., other alloying elements have very limited solubility within its crystal structure. Moreover, Nb has very limited solubility within ferrite and austenite; hence, what you add is what you get.                                                                                                                                                    | 11      | 7.8                          | Complex FCC (B1 Type)              | Coral-like or petal-like | 2000–2500      | –110 to –140           |
| TiC          | Titanium carbide forms as a pure, unmixed carbide.                                                                                                                                                                                                                                                                                                                                          | 20      | 4.9                          | Complex FCC (B1 Type)              | Blocky chunks            | 2000–3100      | –180                   |

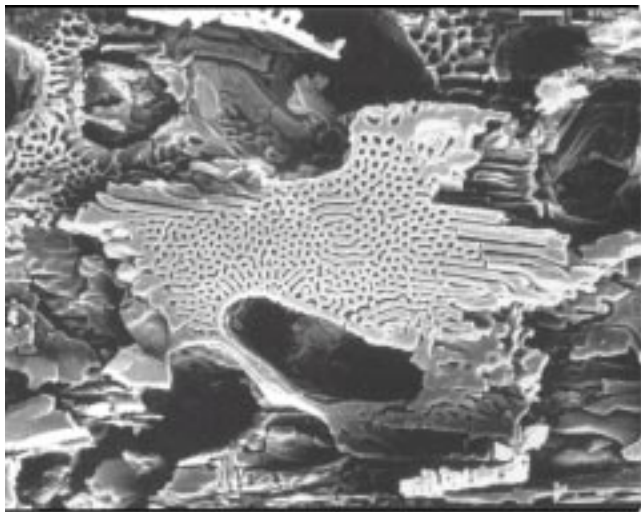
(1) The  $M_7C_3$  carbide can be manufactured to contain  $Cr_7C_3$  but, in typical foundry casting practice, the limit is around 50% Cr.



(5.1a) hypoeutectic



(5.1b) eutectic



(5.1c) hypereutectic

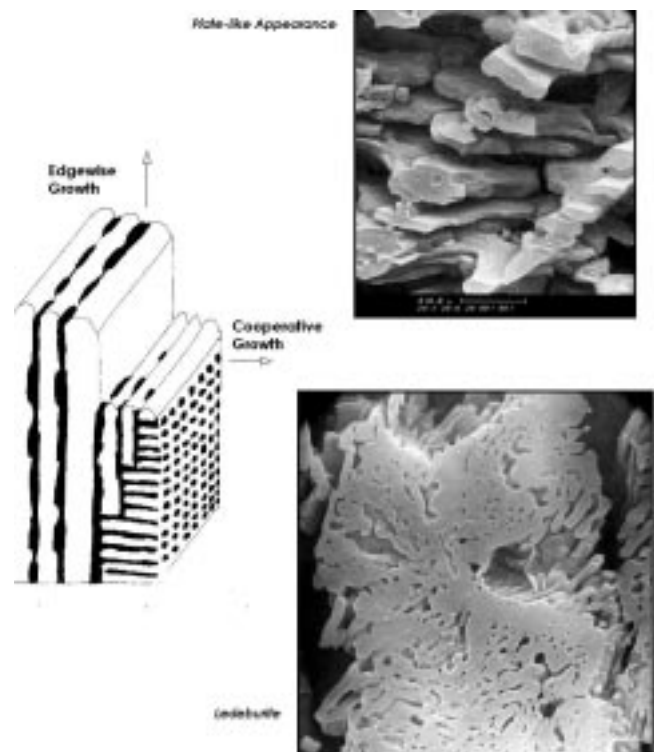
**Fig. 5.1.** Examples of  $M_3C$  carbide morphology as the iron's composition moves from hypo- to hypereutectic. All SEM photomicrographs were taken after deep-etching in aqua regia (see Appendix F for details on etching procedures).

## $M_7C_3$ Carbide Morphology

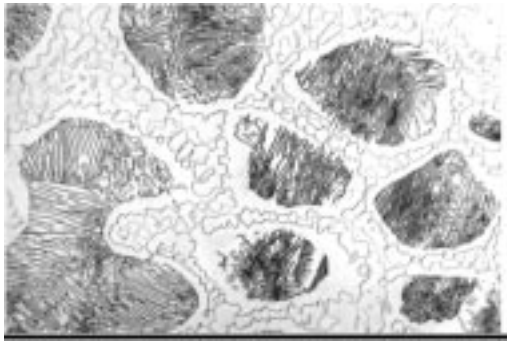
The morphology of the  $M_7C_3$  carbide is truly different from that of the  $M_3C$  carbide. This fundamental difference, to a large extent, is what separates pearlitic and Ni-Hard 1 and 2 from high-Cr irons. Figures 5.6a–5.6c provide a sequence of deep-etched SEM micrographs showing this  $M_7C_3$  carbide morphology in hypoeutectic, eutectic and hypereutectic cast irons.

The  $M_7C_3$  carbide morphology consists of rod- and blade-like structures characterized by a dominant growth direction along the axis of the rod or blade. The growth of these rod- and blade-like structures is confined within a eutectic cell for hypoeutectic and eutectic irons, while in hypereutectic irons the carbides can be extremely thick and long. In the former, the length of any one rod or blade is limited by the size of the eutectic colony. In the latter, the length of a hypereutectic rod is only limited by the growth of adjacent hypereutectic rods within the melt. Consequently, hypereutectic irons can have individual carbide “fibers” that are extremely long (in the range of millimeters).

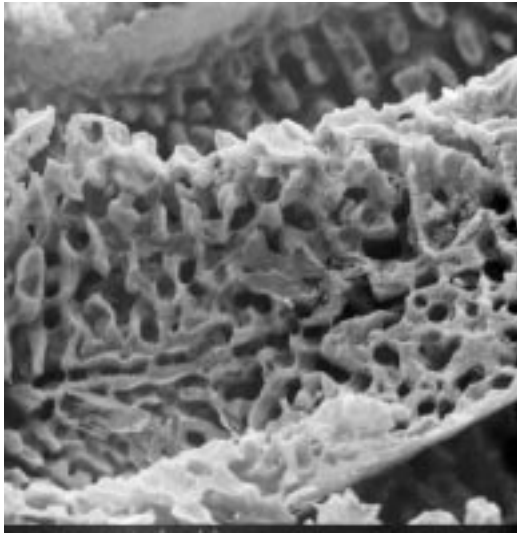
A good example of this growth behavior is shown in Figs. 5.7a and 5.7b. If the iron is sectioned perpendicular to the dominant growth direction, Fig. 5.7a shows that the carbide morphology appears to consist of discrete particles. At right angles to this growth direction, the true growth morphology of the  $M_7C_3$  carbide is displayed in Fig. 5.7b.



**Fig. 5.2.** Depending on the orientation and cooling rate, the  $M_3C$  carbide can appear to have a variety of morphologies. With more careful observations, the morphology is just a variant of the standard plate-like form. (Photomicrographs are courtesy of G.L.F. Powell)



(5.3a) polished plane view of discontinuous  $M_3C$  carbide, within a pearlitic matrix

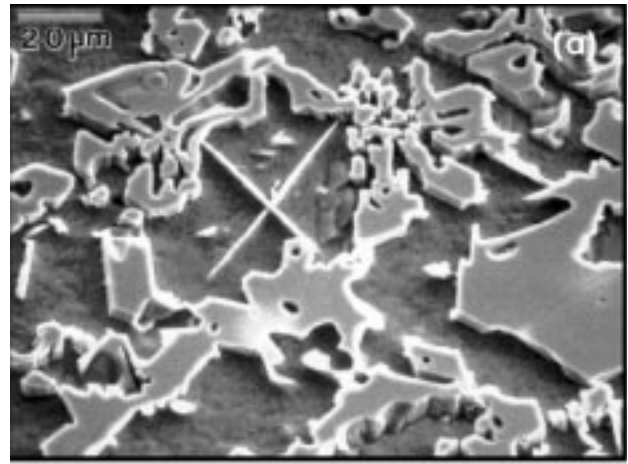


(5.3b) deep-etched view of discontinuous  $M_3C$  carbide, which, of course, is continuous

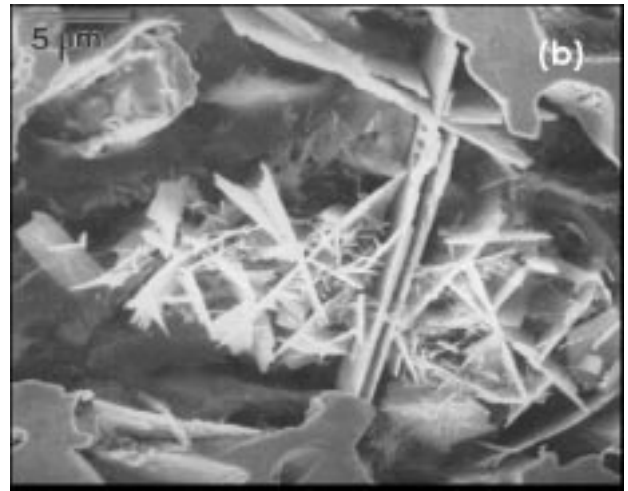
Fig. 5.3. The description of carbide morphology is all in the plane and depth of views that are used. In understanding  $M_3C$  carbide morphology it is always useful to think in 3D. For the  $M_3C$  carbide morphology, try to visualize a plate that is riddled with holes and how it may appear from different view angles and different growth directions. (Photomicrographs are courtesy of G.L.F. Powell)

The iron shown in Fig. 5.7 was sand cast and had a composition of 3% C, 25% Cr. The region shown in the micrographs was obtained from a section that was immediately adjacent to the surface. Since a strong thermal gradient existed near the surface (i.e., acting somewhat as a chill), the carbides dominantly grew perpendicular to this surface. This effect was apparent only within 5 mm of the surface. As would be expected, further away from the surface, the carbide growth morphology became more random.

Another characteristic that is unique about the  $M_7C_3$  morphology is the formation of blade-like carbides between interdendritic spaces. If the iron is strongly hypoeutectic (CVF is less than 20%), the eutectic liquid is squeezed between the solidifying dendrites. As eutectic solidification commences, carbide growth is constrained between the pre-existing dendrites, resulting in the carbide microstructure shown in Fig. 5.6a.



(5.4a) fast cooled



(5.4b) slow cooled

Fig. 5.4. Secondary  $M_3C$  carbides in Ni-Hard 4. The  $M_3C$  carbides still retain their characteristic plate-like morphology. During fast cooling (a), few  $M_3C$  carbides are formed. If the iron is slow cooled (b), many more  $M_3C$  carbides are formed.

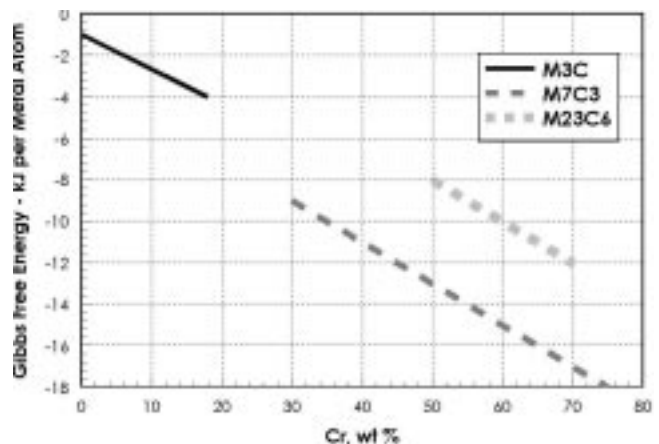
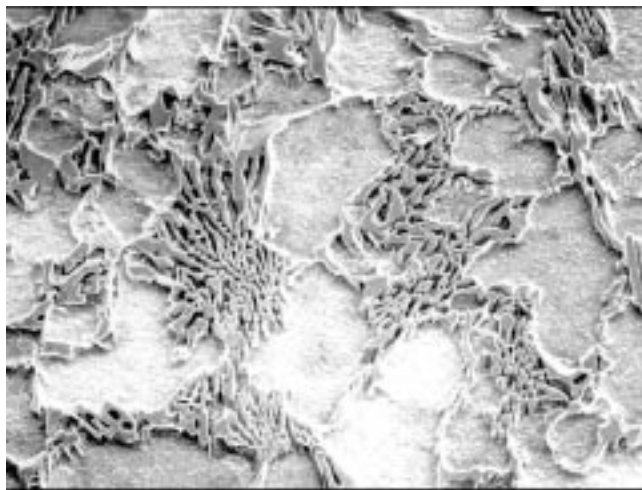
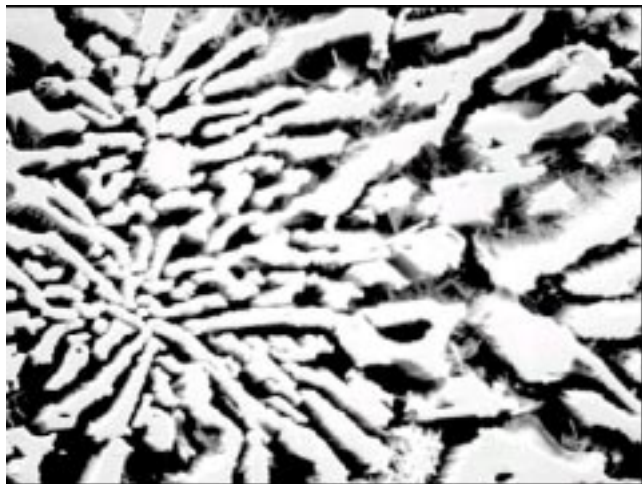


Fig. 5.5. Gibbs Free Energy of Formation for the  $M_3C$ ,  $M_7C_3$  and  $M_{23}C_6$  carbides from austenite. As the Cr content increases, so does the Gibbs Free Energy. Noteworthy is the Gibbs Free Energy for the  $M_3C$  carbide. At very low Cr content ( $Fe_3C$ ), the carbide is thermodynamically unstable (i.e., it takes very little energy to decompose the carbide into graphite and austenite). (After Benz et al., 1974)

Another place that  $M_7C_3$  blade-like structures appear is within eutectic colonies. This aspect was covered in Chapter 2, Solidification, but for completeness, Figs. 5.8a and 5.8b provide a few more pictures of blade-like microstructures that occur in eutectic colonies and between growing dendrites.



(5.6a) hypoeutectic



(5.6b) eutectic



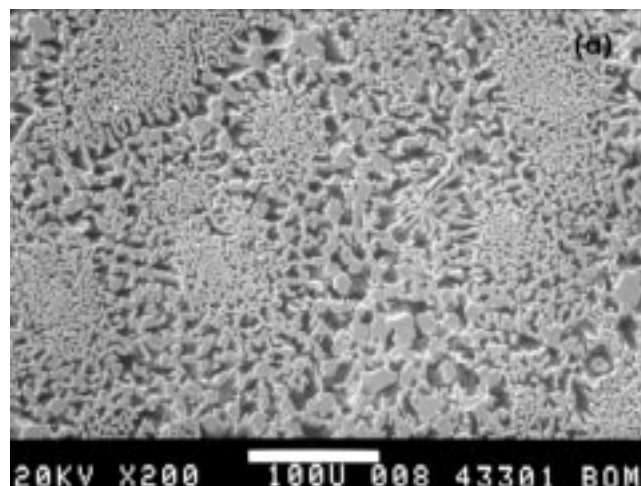
(5.6c) hypereutectic

Fig. 5.6. Examples of the  $M_7C_3$  carbide morphology as the iron moves from hypo- to hypereutectic. All SEM photomicrographs were taken after deep-etching in aqua regia.

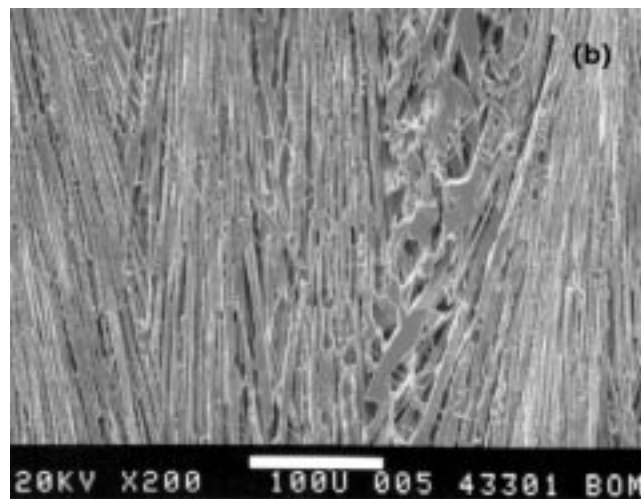
Secondary carbides will also have the same characteristic  $M_7C_3$  morphology. In this case, since there is little competition from solidifying dendrites, the morphology is mostly rod-like. Figure 5.9 shows the morphology displayed by secondary  $M_7C_3$  carbides in a high-Cr white cast iron after heat treatment.

#### **$M_7C_3$ Mechanical and Physical Properties**

The hardness of  $M_7C_3$  can vary from between 1000 to 1800 DPH, with more generally accepted values between 1250 and 1500 DPH. There are many reasons why the hardness of  $M_7C_3$  varies. Depending on what axis the carbide is indented, totally different hardnesses will result. If the carbide is indented across the end of the rod, the hardness is around 1500 DPH, whereas if the carbide is indented along the length of the rod, the hardness is only 1000 DPH. This hardness anisotropy is one reason chill-cast alloys yield slightly higher hardness numbers. That is, near the surface of the iron, the indenter must push against the “harder” side of the carbide.



(5.7a) isolated particles



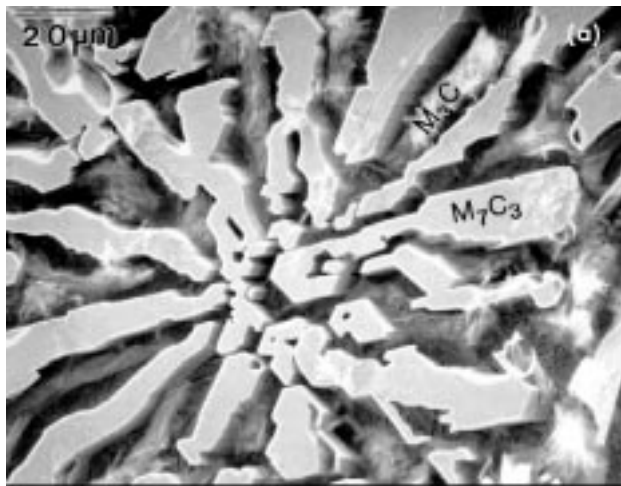
(5.7b) long rods

Fig. 5.7. Dominant growth characteristics of the  $M_7C_3$  carbide can produce long rods if growth is unimpeded. Depending on the polished plane, the carbide morphology can appear as isolated particles (a) or long rods (b). (Courtesy of Dogan and Hawk)

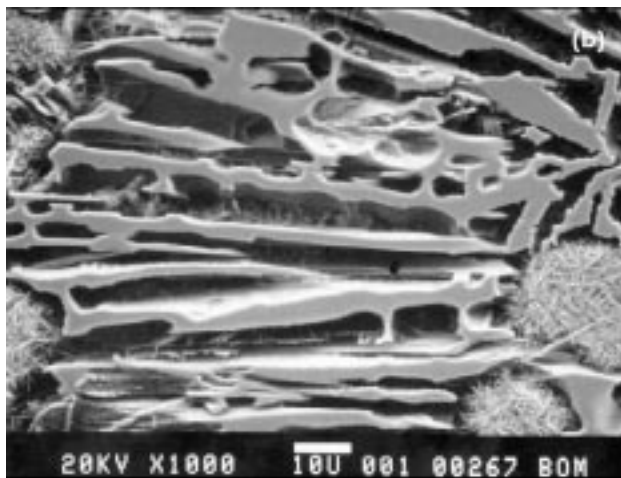
*(Carbide Constituents, continued)*

Other sources on the hardness of the  $M_7C_3$  carbide have shown even greater variability. Measurements made by Kagawa et al. (1992) show great ranges in hardness for the  $M_3C$  and  $M_7C_3$  carbides, depending on the Cr content of the carbides. Their hardness measurements were made on chill-cast alloys. Indentations were made perpendicular to the axis of the  $M_7C_3$  rods or on the “hard” side. Figure 5.10 briefly summarizes part of their work.

Using Fig. 5.11, we can easily estimate the Cr content of the  $M_7C_3$  carbide for many common AR irons. With this number, the hardness of the  $M_7C_3$  carbide can be determined from Fig. 5.10. Very simply, as the Cr content of the iron increases, the hardness of the  $M_3C$  and  $M_7C_3$  carbides increases. Since pearlitic and Ni-Hard 1 and 2 irons contain very little Cr, we can expect to see little variation in the  $M_3C$  carbide hardness. In high-Cr irons, this relationship between Cr content and  $M_7C_3$  carbide hardness provides insight into the abrasion



(5.8a) rod-like structure

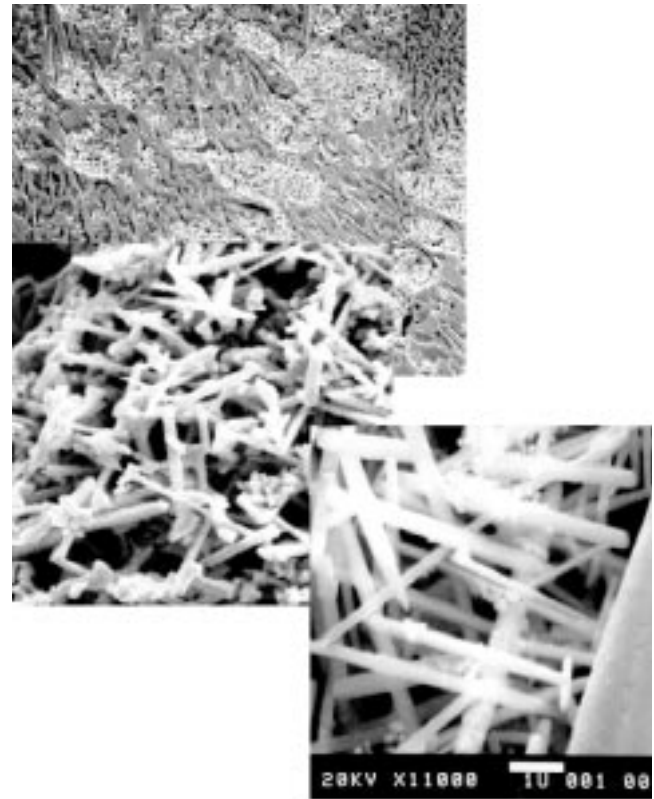


(5.8b) blade-like structure

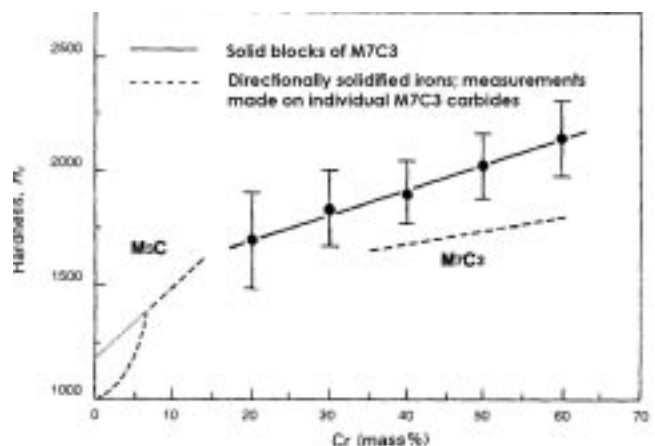
**Fig. 5.8.** The many faceted nature of the  $M_7C_3$  carbide. The basic building block is the rod-like structure, which can be forced to grow as blades when constrained between dendrites. In (a) and (b), the “fur balls” are secondary  $M_3C$  carbides formed during heat treatment. Both micrographs are from the same Ni-Hard 4 iron.

performance of Ni-Hard 4 versus high-Cr abrasion resistance irons. Very simply, higher Cr irons will have harder  $M_7C_3$  carbides.

The chemical inertness of the  $M_7C_3$  carbide is partly tied to its Cr content. As shown in Fig. 5.5, as the Cr content increases, so does the negative Gibbs Free Energy of Formation, and so follows the stability of the  $M_7C_3$  carbide. This is



**Fig. 5.9.** Secondary  $M_7C_3$  carbides in a high-Cr white cast iron. The carbide morphology is basically rod-like with just a few occurrences of rods joining up to form blades.



**Fig. 5.10.** Hardness of the  $M_3C$  and  $M_7C_3$  carbides as a function of Cr content. Measurements were made using a 1-kgf load and represent an average of at least 10 indentations. The large error bars reflect the variability in the hardness measurement. Indentations were made on polished cross sections that were sectioned parallel to the chilled surface. (After Kagawa et al., 1992)

important for users selecting AR irons for extreme corrosion/oxidation environments.

In the literature, pure  $\text{Cr}_7\text{C}_3$  is said to be attacked at 800C (1472F) and will burn at temperatures in excess of 1100C (2012F), if oxygen is present. It can be reasoned that, at concentrations less than pure  $\text{Cr}_7\text{C}_3$ , these temperatures would be lowered.

In oxidizing environments (e.g., the acids  $\text{HNO}_3$  and  $\text{H}_2\text{O}_2$  create oxidizing solutions), the  $\text{M}_7\text{C}_3$  carbide will remain relatively stable due the formation of a thin Cr oxide film. As can be guessed, the  $\text{M}_3\text{C}$  carbide is not nearly as stable since its Cr content is typically only a couple of percent, as compared to the  $\text{M}_7\text{C}_3$ 's Cr content of between 35 to 50%. All in all, the inertness of the  $\text{M}_7\text{C}_3$  carbide is much greater than that of the  $\text{M}_3\text{C}$  carbide.

Finally, in reducing environments (e.g., the acids  $\text{HCl}$  and  $\text{H}_2\text{SO}_4$  create reducing solutions), the  $\text{M}_7\text{C}_3$  carbide is readily dissolved. Chapter 9 on Corrosive/Abrasive Wear Performance will cover the corrosive behavior of AR irons to a much greater extent; however, for the moment, it is good to start thinking about the behavior of these materials as the sum of their individual constituents.

## $\text{M}_{23}\text{C}_6$ Carbide Morphology

This carbide is rarely seen in commercial irons and, to the authors' knowledge, no commercial irons are manufactured that form primary or eutectic  $\text{M}_{23}\text{C}_6$  carbides. It does appear as a secondary carbide in very high-Cr ferritic and austenitic AR irons (e.g., 440C) and is commonly found along the grain boundaries in heat treated low-carbon stainless steels. Its morphology is shown in Figs. 5.12a and 5.12b as a secondary carbide in a heat-treated iron having a base composition of 2.5% C and 29% Cr.

## $\text{M}_{23}\text{C}_6$ Mechanical and Physical Properties

The hardness of  $\text{M}_{23}\text{C}_6$  carbides is lower than that of the  $\text{M}_7\text{C}_3$  carbide but slightly higher than that of the  $\text{M}_3\text{C}$  carbide. It was not possible to find data on the range of hardnesses of the  $\text{M}_{23}\text{C}_6$ , only on the pure  $\text{Cr}_{23}\text{C}_6$  carbide, for which a hardness of 1000 DPH is reported.

Following the Gibbs Free Energy data provided in Fig. 5.5, it would be expected that the chemical stability of the  $\text{M}_{23}\text{C}_6$  carbide would be lower than the  $\text{M}_7\text{C}_3$  carbide. However, due to its higher Cr content, the  $\text{M}_{23}\text{C}_6$  carbide is less reactive in oxidizing environments than the  $\text{M}_7\text{C}_3$  carbide.

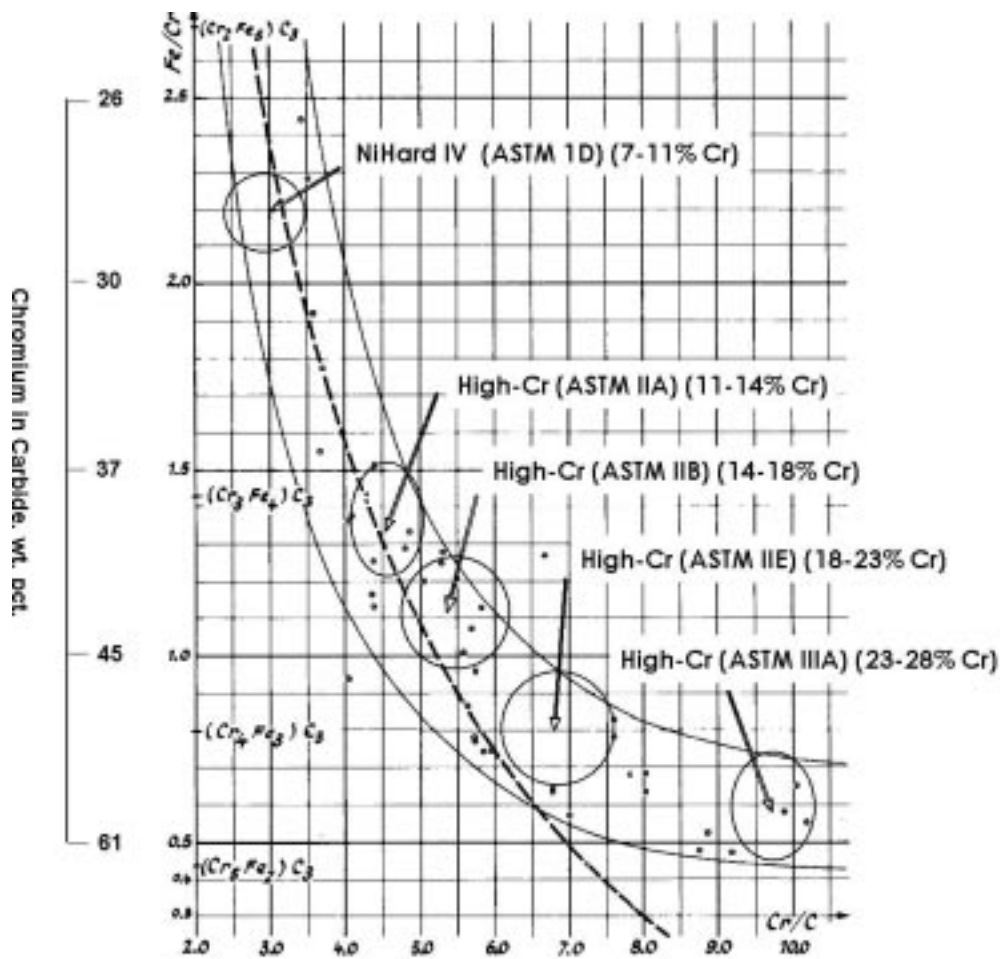
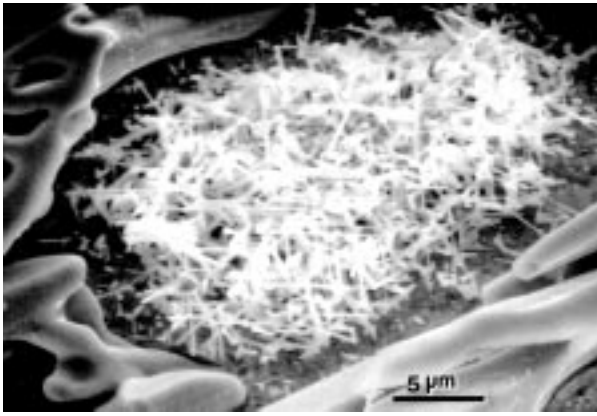
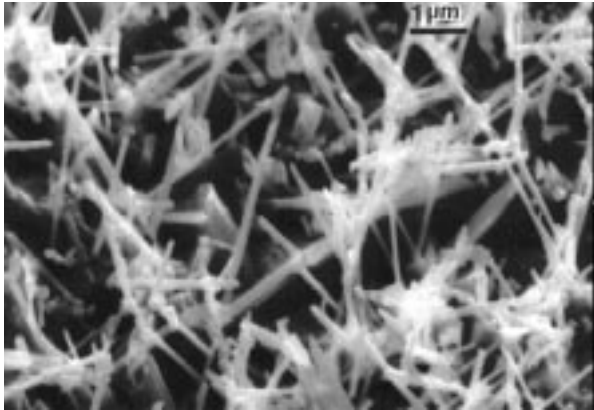


Fig. 5.11. Relationship between Cr:C ratio of the iron and the expected Cr content of the  $\text{M}_7\text{C}_3$  carbide. For example,  $(\text{Cr}_2\text{Fe}_5)\text{C}_3$  has about 24% Cr while the  $(\text{Cr}_5\text{Fe}_2)\text{C}_3$  carbide has about 65% Cr. (After Maratray and Usseglio-Nanot, 1971)



(5.12a)



(5.12b)

**Fig. 5.12. Two different magnifications showing the morphology of the  $M_{23}C_6$  carbide in an austenite dendrite of a 2.5% C, 29% Cr AR cast iron. The morphology is completely rod-like. (Courtesy of Powell and Laird)**

### Mo<sub>2</sub>C Carbide Morphology

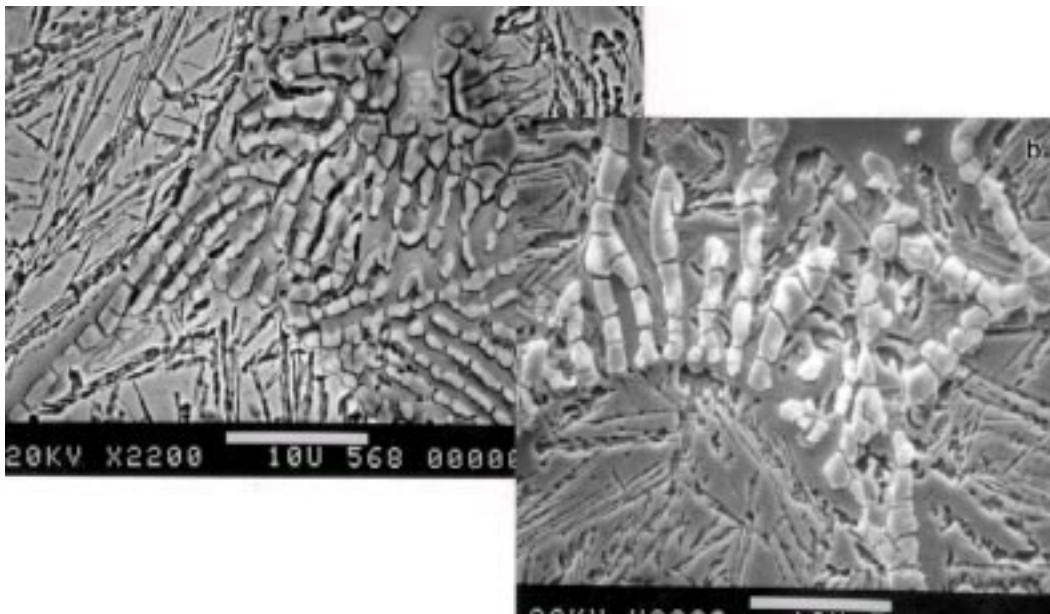
The presence of this carbide is rarely desired. As shown in Fig. 5.13, its fibrous appearance is very similar to that of the soft phosphate carbide ( $Fe_3P$ ) discussed in Chapter 2 on Solidification. The Mo<sub>2</sub>C morphology is characterized by blade-like segments. The Mo carbide forms at the very end of the solidification sequence in AR irons (Fig. 2.28). At this solidification stage, the remaining liquid pool is tightly constrained by the surrounding dendrites. As solidification progresses ( $L \rightarrow Mo_2C + \gamma Fe$ ), the eutectic grows as a mixture of blade-like carbide and austenite. As with other metal carbides, Mo<sub>2</sub>C has a preferential growth direction resulting in the blade-like morphology shown in Fig. 5.13.

The Mo<sub>2</sub>C eutectic carbide is easy to find in AR irons that are alloyed with an Mo content in excess of 3%. Since it forms within the constrained interdendritic spaces, its morphology is often dictated by the other phases that form first. An example of this structure is shown in Fig. 5.14 where the dendrites and M<sub>3</sub>C carbides have all formed prior to the Mo eutectic. At 2% Mo, Mo<sub>2</sub>C carbides can be found along the grain boundaries. At lower levels, Mo dissolves into the austenitic matrix and M<sub>7</sub>C<sub>3</sub> carbide.

Although it might be suspected that the Mo carbide would form as a mixture or as an M<sub>2</sub>C type in Cr-containing AR irons, it has only been confirmed as pure Mo<sub>2</sub>C (Maratray and Usseglio-Nanot, 1971).

### Mo<sub>2</sub>C Mechanical and Physical Properties

The hardness of the Mo<sub>2</sub>C carbide varies from 1500 to 1800 DPH. Since they form as very fine interdendritic carbides and, in most irons, are present in trace quantities (<1% carbide volume fraction), their effect on the iron's hardness and wear properties is negligible.



**Fig. 5.13. Morphology of the Mo<sub>2</sub>C carbide within an austenite dendrite of a 2.3% C, 5% Mo AR cast iron. The morphology is blade-like. Images were created by heavily etching the sample with 2% nital and then viewing in the secondary electron mode of the SEM.**

Molybdenum carbide is rapidly attacked in air at 700–800C (1292–1472F), and can be easily dissolved by concentrated HF, HNO<sub>3</sub> and H<sub>2</sub>SO<sub>4</sub>. Interestingly enough, it is impervious to concentrated HCl. During heat treatment, the carbide remains stable and will not transform. In part, this property is characteristic of its high Gibbs Free Energy of Formation (see Table 5.1).

### MC (VC) Carbide Morphology

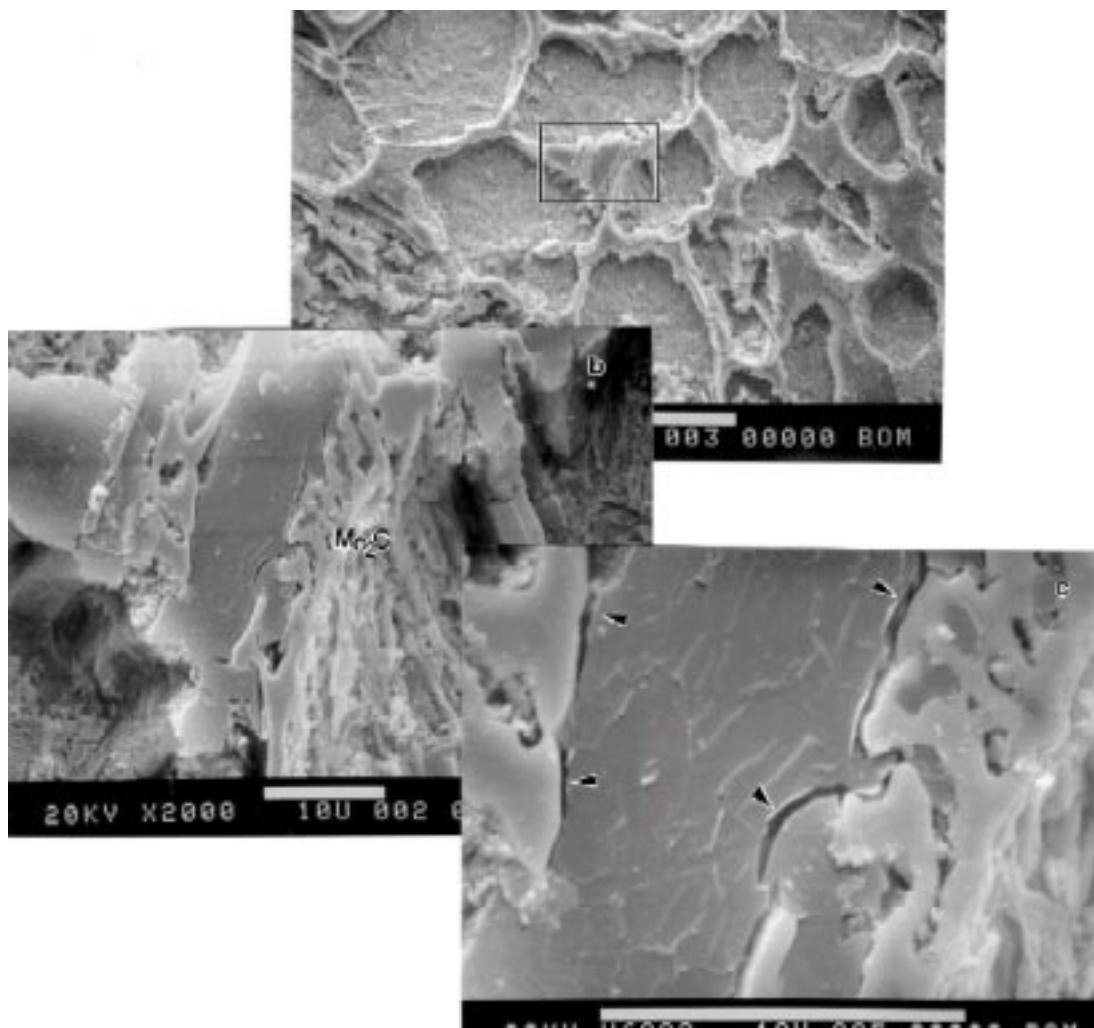
Vanadium carbide only appears as a mixed carbide, much like the M<sub>3</sub>C and M<sub>7</sub>C<sub>3</sub> carbides. In AR irons containing more than a couple percent Cr or Mo, the MC (VC) carbide's stoichiometry will take on the following formula: V<sub>x</sub>Cr<sub>y</sub>Mo<sub>z</sub>C, where  $x > 0.50$  and  $x + y + z = 1.0$ . The MC structure has a great affinity for other metallic alloying elements, but in the form discussed within this section, its dominant alloying element is vanadium. Figure 5.15 quantifies the distribution of alloying elements in the MC carbide for AR irons containing 5 and 15% chromium.

The eutectic MC carbide can provide superior wear resistance and toughness in exotic AR cast irons and in high-C cast tool steels. Although it is quite hard, its morphology is its outstanding characteristic. Figure 5.16 shows examples of the MC morphology. As shown in Figs. 5.16b and 5.16c, the eutectic MC carbide forms as slender, discrete rods that grow from a central nucleation point. Growing as isolated rods, the MC carbide morphology hinders crack propagation. In combination with the right type of metallic matrix (e.g., tempered martensite), a high toughness AR iron can be manufactured.

### MC (VC) Mechanical and Physical Properties

The hardness of this carbide varies from 2000 to 3000 DPH. Researchers have noted that the hardness of this carbide varies with composition and, under the right conditions, can be almost as hard as titanium carbide. As a mixed carbide formed via a eutectic reaction, more likely, its hardness is around 2000–2500 DPH.

As the hardness increases so does the negative Gibbs Free Energy of Formation as given in Table 5.1. Unfortunately, a



**Fig. 5.14.** Deep-etched SEM micrographs of a 2.3% C, 5% Mo AR cast iron. The top SEM shows the overall carbide morphology. The iron is composed of about 10% M<sub>3</sub>C and 2% Mo<sub>2</sub>C carbides. The region enclosed by the box is examined at higher magnifications in the middle and bottom SEMs. The Mo<sub>2</sub>C carbide is fibrous and is poorly bonded to the M<sub>3</sub>C carbide. Cracks between the M<sub>3</sub>C and Mo<sub>2</sub>C carbide are shown in the bottom SEM.

(Carbide Constituents, continued)

high Gibbs Free Energy does not always imply that the carbide will be inert. At heat treatment temperatures in excess of 800C, the vanadium carbide will be aggressively oxidized in the presence of air.

Vanadium based carbides will be attacked by  $\text{HNO}_3$ ,  $\text{H}_2\text{SO}_4$  and  $\text{HClO}_4$  acids but remain relatively unscathed when immersed in HCl.

### NbC Carbide Morphology

This unusual carbide is sometimes found in a few AR cast irons. Its appearance is often surrounded with some mystique. Nevertheless, niobium carbide (NbC) is one of the easiest carbides to understand. First, it only appears in the pure form. Second, Nb has little or no solubility in the other phases found in AR cast irons (i.e.,  $\gamma\text{Fe}$ ,  $\text{M}_7\text{C}_3$  and  $\text{M}_3\text{C}$ ). And, third, since it is insoluble in austenite and has a high negative Gibbs Free

Energy of Formation, it is quickly rejected by growing dendrites, forming NbC quite early in the solidifying sequence.

When Nb is added to hypoeutectic AR cast irons at levels below about 0.6%, it will form as thin, petal-like clusters at the centers of austenite dendrites. At Nb content above 1%, primary NbC will start to form as compact, blocky carbides. Figures 5.17a–5.17c show the morphology of these carbides. In Figs. 5.17b and 5.17c, the austenite dendrite has been dissolved away during the deep-etching procedure leaving the eutectic NbC and primary NbC carbides exposed.

### NbC Mechanical and Physical Properties

The hardness of NbC is sometimes considered greater than that of the MC (VC) carbide. In practice, the two carbides are probably equivalent. If anything, the range of hardness reported for NbC of between 2000 to 2500 DPH places it as softer than the MC carbide.

Niobium carbide is very inert and stable. Most acids will not dissolve NbC (even boiling aqua regia will not attack powdered NbC); however, a mixture of  $\text{HNO}_3$  and HF will cause gentle dissolution. The carbide is also relatively impervious to oxidation, and will only start to oxidize at temperatures in excess of 1100C (2012F).

### TiC Carbide Morphology

Titanium carbide (TiC) is the most inert and the hardest of all the carbides. In AR irons, Ti will sometimes be added as an inoculant to the melt. Once in the high-C melt (i.e., AR iron), the free Ti will immediately form small, cubic particles of TiC. As the rest of the iron starts to solidify, these small particles act as nucleants for austenite and carbide growth. Figure 5.18 provides a glimpse of their morphology within a prior austenite dendrite.

### TiC Mechanical and Physical Properties

The hardness and the Gibbs Free Energy of Formation are both extreme for this carbide. Although a hardness value of 3100 DPH is often quoted, its hardness has been found to vary

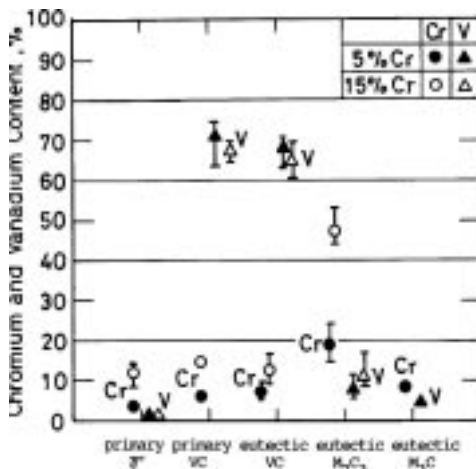


Fig. 5.15. Distribution of Cr and V in  $\gamma$ -austenite and the carbide phases in a V-alloyed AR cast iron. Vanadium will also substitute for Cr and Fe atoms within the  $\text{M}_7\text{C}_3$  and  $\text{M}_3\text{C}$  carbides. Conversely, V has very little solubility in the austenite dendrite. (After Sawamoto, 1986)

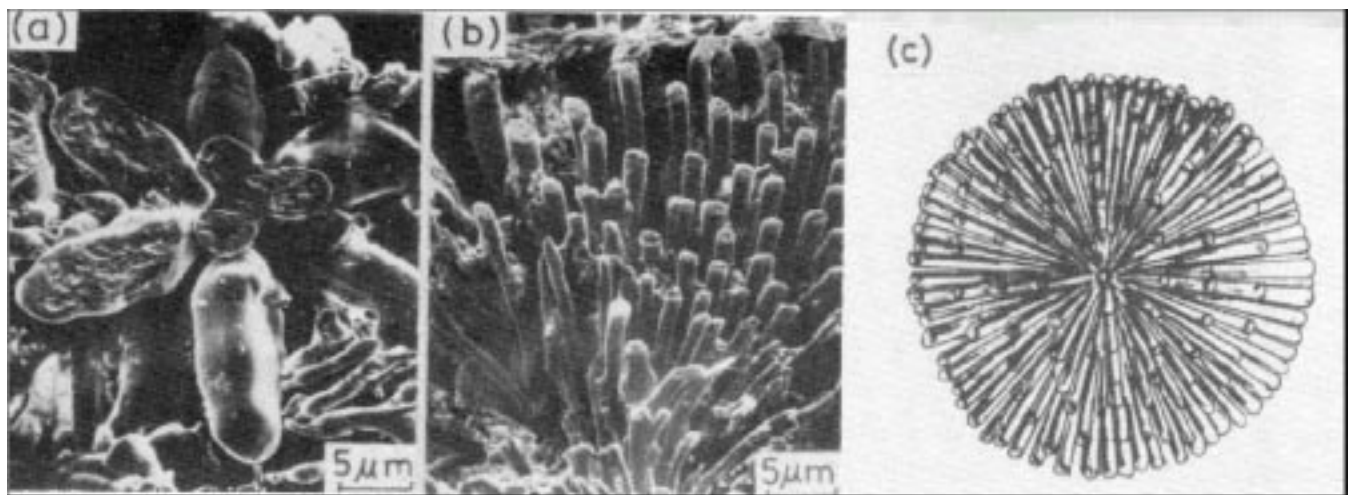
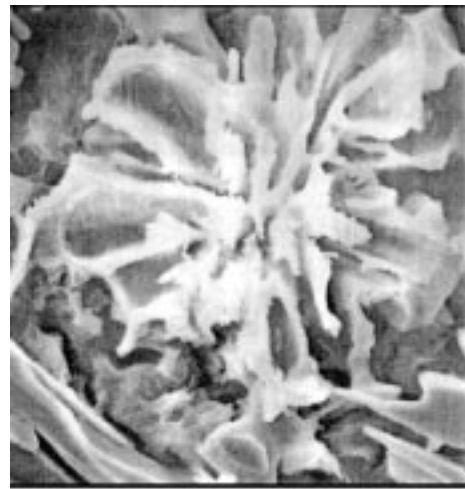


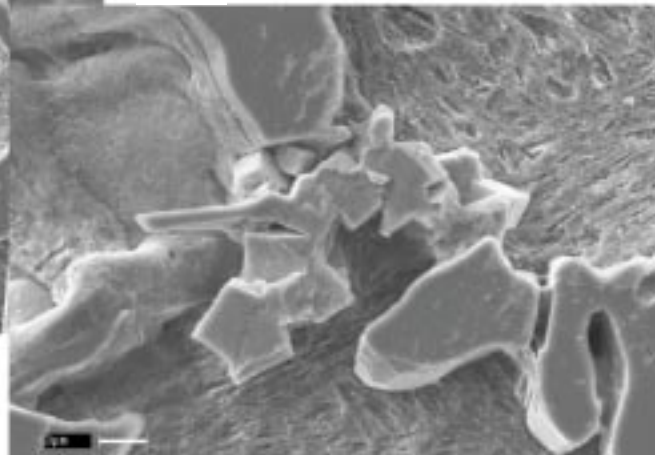
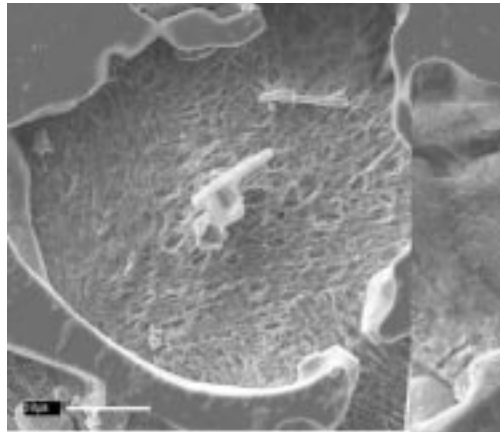
Fig. 5.16. Deep-etched SEM micrographs of primary V carbides (VC), eutectic VC and a schematic of how a VC eutectic colony might appear: (5.16a) primary VC; (5.16b) eutectic VC; (5.16c) schematic. (After Sawamoto, 1986)



(5.17a)

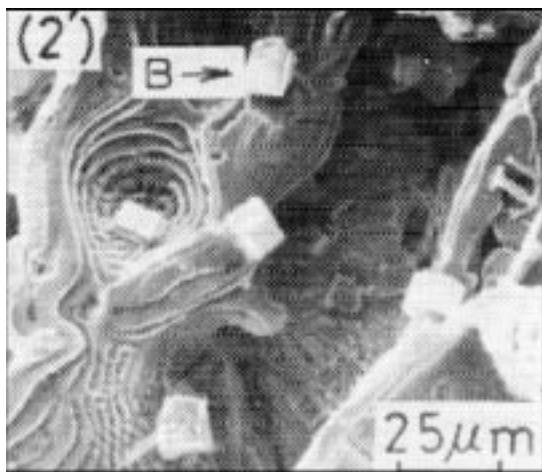


(5.17b)



(5.17c)

**Fig. 5.17. Morphology of the Nb carbide; (a) schematic of a NbC eutectic cluster, (b) deep-etched morphology of the eutectic NbC and (c) primary NbC surrounded by  $M_3C$  carbides. (After Sawamoto, 1986 and Baik and Loper, Jr., 1988)**



**Fig. 5.18. Morphology of TiCs in a hypoeutectic AR cast iron. The symbol B marks a TiC. The other carbide visible in the deep-etched SEM micrograph is  $M_7C_3$ . (After Sawamoto, 1986)**

from as low as 1600 to as high as 3100 DPH. An interesting study, using a 50-g indentation load, showed a linear variation of hardness from 1900 to 2900 DPH, as the composition of the TiC was varied from  $TiC_{0.8}$  to  $TiC_{1.0}$ , respectively. Most likely, the true hardness value of TiC lies somewhere between 2500 and 2900 DPH, making this the hardest of the carbides covered in Table 5.1.

Titanium carbide has the highest Gibbs Free Energy of Formation of the carbides covered in Table 5.1. As would be expected, TiC is very stable and is resistant to most acids. A combination of acids, e.g.,  $HNO_3 + HCl$  or  $HNO_3 + H_2SO_4$ , will cause complete dissolution of TiC after a 24-hour soak. Otherwise, TiC is extremely resistant to most forms of environmental attack.

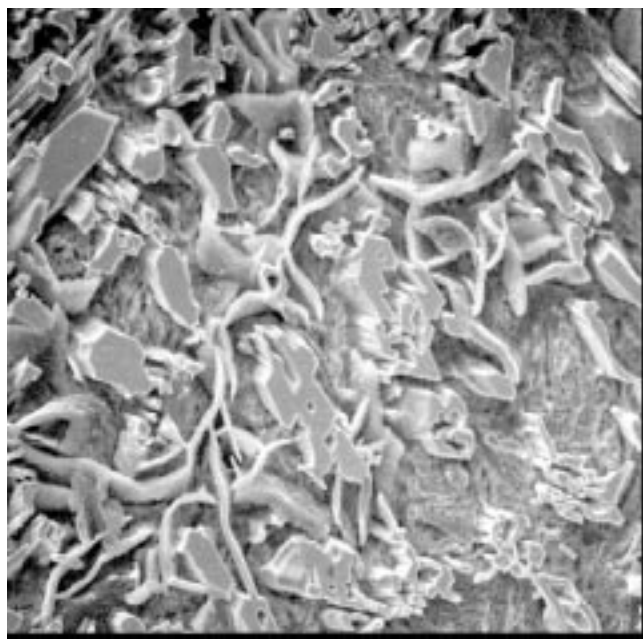
## Graphite Constituents

Although graphite is not often considered a constituent of AR cast irons, it does play a vital role in many applications that require good abrasion resistance and good thermal shock characteristics. In regard to thermal shock, the presence of a few flakes of graphite act as “expansion” joints within the iron, thereby allowing large thermally induced strains to be absorbed within the iron, without cracking the structure.

In most AR cast irons, graphite is not a desirable constituent. Figure 5.19 shows such a case, where graphite was formed in a very high-silicon pseudo-Ni-Hard composition. Additionally, mixed graphitic/carbide wear irons are sometimes made by changing the morphology of the graphite constituent from flake to nodule. In this condition, the iron will have high toughness and moderate wear resistance, albeit, quite low, as compared to nongraphitic AR irons.

Table 5.2 presents the relevant engineering data for graphite. The presence of graphite in the iron’s microstructure is the mechanical equivalent of having holes or voids in your casting. For example, with a hardness of around 20 DPH, the compressive or tensile strength of graphite is practically nil. On a fracture mechanics basis, flakes and nodules act as sharp cracks and voids, respectively. Given a choice, it is far better to have a small void rather than a sharp crack in the iron’s structure (much like the difference between flake and ductile gray iron).

On a positive note, graphite provides an effective means to impart good thermal shock resistance to AR cast irons. This is important to foundries producing rolls for hot-strip rolling mills.



**Fig. 5.19. Graphite flakes in a low-alloy, high-Si AR cast iron. The iron’s composition is 3.1% C, 2% Si, 4% Ni and 2.5% Cr. If Mg is added as a ladle addition, these flakes become nodules.**

## Metallic Matrix Constituents

The metallic matrix in AR cast irons is often a mixture of several different ferrous phases. The primary phases are austenite, martensite and pearlite. Depending on the volume fraction of austenite, martensite, or pearlite, the hardness, toughness and wear properties of the iron can vary significantly.

The emphasis of this section is on this ferrous phase and what one can expect in terms of morphology, crystallography and hardness. Each ferrous phase: austenite, ferrite and martensite, and the two most common nanocomposite phases of pearlite and bainite are discussed. This approach provides some perspective on how the metallic matrix fits into the composite analogy for AR cast irons. Table 5.3 provides a starting point for this discussion.

### Austenite Morphology

Austenite is the first phase to form in all ASTM-specified AR cast irons. An example of its dendritic morphology is shown in Figs. 5.20a and 5.20b. The dendritic arms grow preferentially according to the thermal gradient within the casting and, to some extent, show how the alloying elements (C, Cr, Si, Mo, etc.) segregate during solidification.

Once solid, the austenite dendrites form a crystalline face-centered cubic (FCC) structure. Figures 5.21a–5.21c show how the FCC crystal structure might look on an atomic basis. The atoms can be visualized as fitting into a unit cell having a specific arrangement. This unit cell then forms a repeating arrangement throughout the austenite dendrite.

In between the atoms of the FCC and, likewise, the body-centered cubic (BCC), unit cells lie in interstitial spaces. The size of these interstitial spaces are fixed by the crystal structure, be it FCC or BCC. In FCC austenite, interstitial alloying elements, such as C or N, easily fit into these interstitial spaces. Figure 5.21 shows the location of the interstitial sites most favored by C or N. The octahedral sites for the FCC unit cell can theoretically accommodate atoms that are 0.52 angstroms (Å) or  $0.52 \cdot 10^{-10}$  m in radius. The typical iron atom has a radius of 1.28 Å. Carbon atoms have a radius of 0.77 Å and, therefore, must squeezed into the FCC lattice to obtain the solubility referenced in Table 5.3.

The austenite FCC crystal structure is much more complex than just the simple “sphere and stick” figure shown in Fig. 5.21c. As mentioned in Chapter 3 on Heat Treatment, depending on the temperature, the austenite FCC crystal structure will have an equilibrium C content fixed by how the atoms vibrate back and forth, due to thermal energy (temperature) and other lattice perturbations caused by substitutional alloying elements such as Cr, Mn, Ni, Mo, etc. Considering these facts and the atomic radius of C, It is no surprise that as the C content is increased, the lattice spacing of the austenite FCC crystal must also correspondingly increase. This result is shown in Fig. 5.22, along with the similar affect on the martensitic body-centered-tetragonal (BCT) crystal structure.

Depending on the Cr content of the austenite matrix and the heat-treatment temperature, the C content of the austenite can vary from a maximum of 2.11% down to less than 0.3%. At room temperature, the equilibrium solubility of C in austenite is nearly zero. Nonequilibrium heating and cooling rates can have a dramatic effect on the dissolved C content within the austenite matrix.

Of all the elements, C has the greatest influence in stabilizing austenite at room temperature. Very simply, C has the strongest effect on pushing the martensite start temperature to lower temperatures. If high-C austenite is hindered from decomposing to pearlite as it cools to room temperature, the

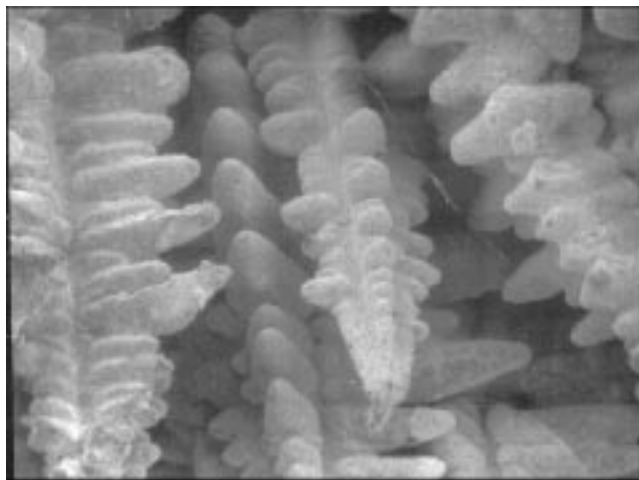
most likely dominant phase will be austenite. This un-decomposed and un-transformed austenite is termed *retained* or *metastable* austenite. An example of the relationship between retained austenite and C is given in Fig. 5.23.

#### Austenite Mechanical and Physical Properties

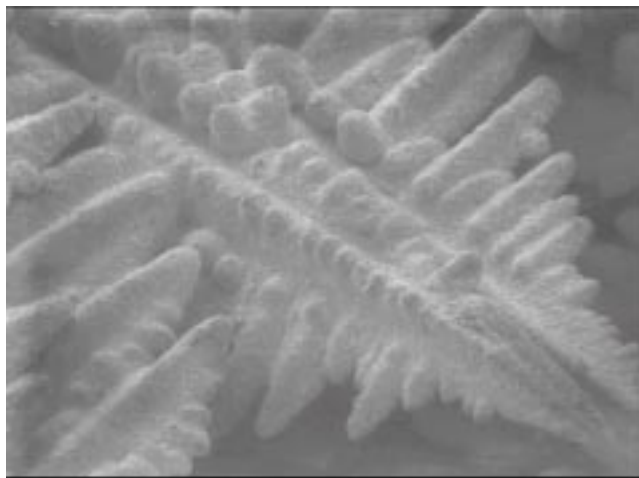
The hardness of retained austenite (all austenite at room temperature in AR cast irons is retained austenite) is controversial. As given in Table 5.3, the hardness varies from 250 to 500 DPH. This variation is due to a number of factors. For example, the hardness of low-alloy austenite has been reported to be between 250 and 350 DPH and high-alloy austenite between 300 and 600 DPH.

**Table 5.2**  
**Graphite in AR Cast Irons**

| Phase    | Chem-istry | Density (g/cm <sup>3</sup> ) | Crystal-lography | Mor-phology     | Hardness (DPH) |
|----------|------------|------------------------------|------------------|-----------------|----------------|
| Graphite | 100% C     | 2.175                        | Hexagonal        | Flake or nodule | 15–40          |

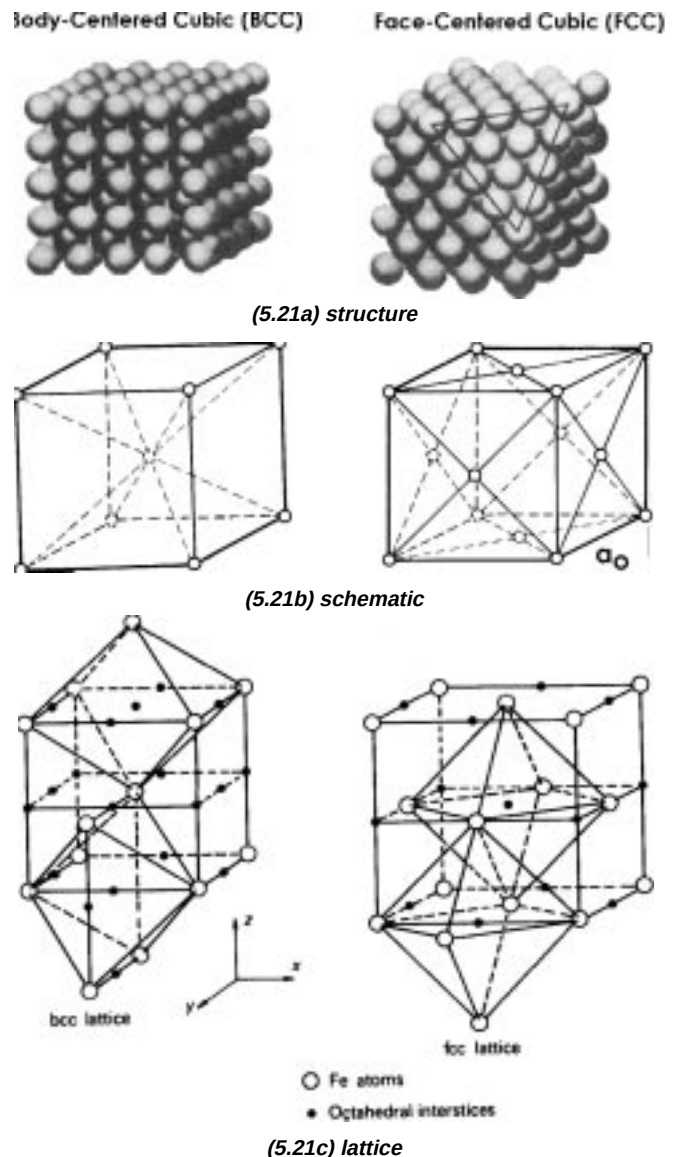


(5.20a)



(5.20b)

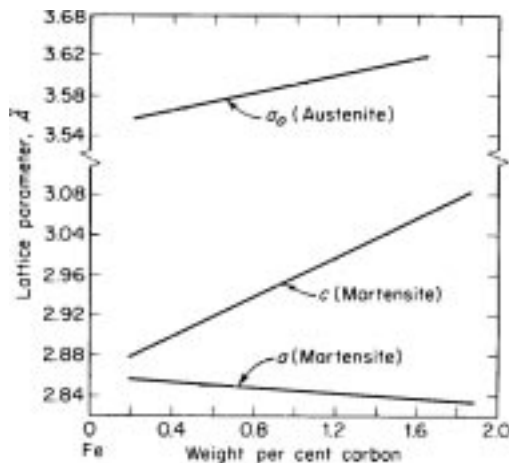
**Fig. 5.20.** Two magnifications of ferrous dendrites in cast steel. (Micrograph courtesy of Ms. Elaine Thomas of Atlas Foundry, Tacoma, WA)



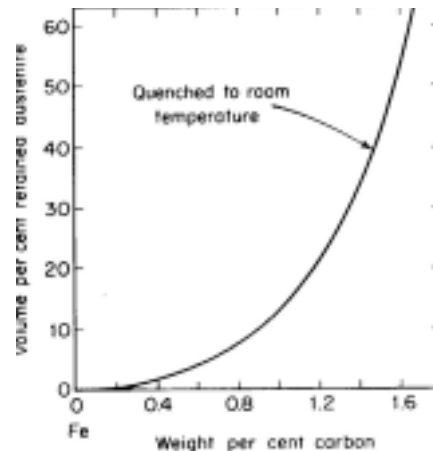
**Fig. 5.21.** Schematics of BCC and FCC crystal structures in iron. The open lattice schematics show the octahedral interstices in BCC and FCC iron. The octahedral interstices in BCC iron are about one-third as big as the FCC interstices. This crystallographic feature explains much about the solubility of C and N in ferrite versus austenite. (After Honeycombe and Bhadeshia, 1995)

**Table 5.3**  
**Ferrous Phases and Nanocomposite Mixtures in AR Cast Irons**

| Ferrous Phase                 | Chemistry (Wt%)<br>Major Alloying Elements                                                            | Carbon Content                                                                                                                                               | Density (g/cm <sup>3</sup> ) | Crystallography                               | Morphology                                                               | Hardness (DPH) |
|-------------------------------|-------------------------------------------------------------------------------------------------------|--------------------------------------------------------------------------------------------------------------------------------------------------------------|------------------------------|-----------------------------------------------|--------------------------------------------------------------------------|----------------|
| Austenite                     | Cr, Ni, Si, etc.<br>all can play a role in the chemical make up of austenite.                         | Max 2.0% 1150C (2100F)<br>Min. 0.8% 723C (1333F)<br>( $\gamma$ -Fe is not an equilibrium phase at room temp.)                                                | 7.84<br>(0.9% C)             | FCC                                           | Dendritic                                                                | 250–500        |
| Ferrite                       | Ferrite will behave similarly to austenite except for its much lower C solubility.                    | Max 0.02% 723C (1333F)<br>Min <0.00005% 20C (68F)                                                                                                            | 7.86                         | BCC                                           | Dendritic                                                                | 215–270        |
| Martensite                    | Follows the behavior of austenite.                                                                    | Max 2% C<br>Min. >0.00005% C                                                                                                                                 | 7.63<br>(0.9% C)             | BCT                                           | Lath-like or Plate-like (lenticular)                                     | 300–900        |
| <b>Nanocomposite Mixtures</b> |                                                                                                       |                                                                                                                                                              |                              |                                               |                                                                          |                |
| Pearlite                      | Just a simple compound of ferrite and M <sub>3</sub> C.                                               | ~0.8% C.<br>Higher C austenites will form discrete M <sub>3</sub> C carbide particles surrounded by pearlite.                                                | 7.78                         | Nanocomposite of ferrite and M <sub>3</sub> C | Lamellar mixture of ferrite and M <sub>3</sub> C                         | 250–450        |
| Bainite                       | Just a simple compound of ferrite and M <sub>3</sub> C that nucleates and grows in a complex fashion. | Formed from max 2.0% C austenite.<br>However, the resulting bainite follows the ferrite max/min limits on C with the remaining C dumped as M <sub>3</sub> C. | 7.7–7.8                      | Nanocomposite of ferrite and M <sub>3</sub> C | Ferrite plates (lenticular) interspersed with M <sub>3</sub> C carbides. | 350–600        |



**Fig. 5.22. Variation of the lattice parameters of austenite (FCC) and martensite (BCT) as a function of C content. (After Roberts, 1953)**



**Fig. 5.23. Retained austenite in steel as a function of C content. (After Roberts, 1953)**

This variation in hardness as a function of alloying content (mainly carbon) is largely due to the unstable or metastable nature of retained austenite at room temperature. When a hardness measurement is made, an indenter is pushed into the austenite crystal causing irreversible plastic strain. The austenite around and underneath the indenter experiences on the order of 10% plastic strain. Depending on the stability of the retained austenite, this mechanical deformation will cause a certain volume fraction of the retained austenite to transform to hard, fresh martensite. The mechanical transformation of austenite to martensite results in the formation of what is

called strain-induced-martensite (SIM). This process is extremely important to the abrasion resistance of many irons and will be covered in greater detail in Chapter 8 on Abrasive and Repetitive-Impact Wear Performance.

It would seem only natural, depending on the stability of the retained austenite and its penchant for the formation of SIM, that room temperature hardness of retained austenite would vary significantly. Most likely, the true hardness or the hardness of just austenite and not SIM, is much softer than 600 DPH or even the maximum hardness given in Table 5.3 of 500

DPH. Guidance for the true hardness of austenite can be found in hardness measurements made on highly stable austenitic Mn alloys, e.g., Hadfield steel, which has a hardness of between 170 to 230 DPH.

The chemical stability of low-Cr austenite is nothing special and behaves identically to that of normal steel. It can be easily attacked with different types of acids, e.g.,  $\text{HNO}_3$  and  $\text{HCl}$  and is easily corroded in environments containing moisture. The exact opposite is true for high-Cr austenite.

The corrosion-resistant properties of austenitic stainless steels or, more accurately, austenitic corrosion-resistant steels are well known. At higher Cr content, the austenitic matrix is nearly identical to that found in these steels and is quite resistant to attack from  $\text{HNO}_3$  (an oxidizing acid) yet can be easily dissolved by  $\text{HCl}$  (a reducing acid). These chemical properties are readily applied in the use of etchants to reveal microstructures (see Appendix F).

With proper alloying, it is not difficult to create a high-Cr austenitic matrix in an AR cast iron that is identical to one formed in a corrosion-resistant steel. If the matrix Cr level is above 11%, then a tight, protective  $\text{Cr}_2\text{O}_3$  oxide film will have a favorable chance to form on the surface of the austenite and provide a barrier to further oxidation. However, if this Cr level is not maintained throughout the austenitic matrix, then selected corrosive attack can occur. This type of selected attack around depleted regions (e.g., carbides) can lead to higher-than-predicted corrosion rates for the iron (see Chapter 9 on Corrosive/Abrasive Wear Performance for greater details).

## Ferrite Morphology

In hypoeutectic irons, the first phase to form is austenite or, when the C is low and the Cr high, ferrite. Whatever the case, the morphology of the dendrite is the same. For ferrite to form during solidification in an AR iron, the iron's composition must fall on the ferrite liquidus surface shown in Fig. 2.8. As shown in the phase diagram, the formation of ferrite generally implies that the iron has a lower C content than that of other AR cast irons.

In understanding the ferrite BCC crystal structure, it is useful to review the data available from the binary Fe-C system (or in other words, at 0.0% Cr). In the Fe-C system, ferrite has a maximum C solubility of 0.18% at 1496C (2725F). As the iron cools, the solubility of C in ferrite decreases until, at room temperature, the maximum C solubility is less than 0.00005%. Figure 5.21 presents a few schematics that show where the carbon atoms fit into the BCC crystal structure of ferrite. More complete information is provided in Table 5.3.

The fundamental reason that ferrite has such low C solubility (and, for that matter, N or any other small atom), is that there is very little room for the atoms to fit into the BCC crystal lattice. Just to put things into perspective, as in the FCC austenitic structure, carbon fits into the BCC structure as an

interstitial. In the ferrite BCC crystal, interstitials fit into the octahedral interstices as shown in Fig. 5.21. These interstitial sites are 1/3 as large as those for the austenite FCC crystal. This huge difference accounts for the very limited solubility of C in ferrite.

At higher temperatures, Cr expands the C solubility in the ferrite BCC structure. In ferrite (and, for that matter, austenite), Cr atoms act as substitutes in the ferrite BCC structure. The native crystal structure of Cr is BCC. It expands the ferrite phase boundary with respect to C (conversely, it shrinks the austenite phase field or loop). Because Cr expands the ferrite loop or acts as a ferrite promoter, it stabilizes the ferrite phase.

Upon cooling to room temperature, it is possible to have a 2% C, 30% Cr AR cast iron having a microstructure consisting of ferrite and  $\text{M}_7\text{C}_3$  carbides. In this case, the ferrite at room temperature is actually at equilibrium. However, it will be supersaturated with C and, if heat treated within the range of 600 to 750C (1112 to 1382F), will decompose to pearlite.

Ferrite can also form in AR irons when large amounts of Si are added to the iron. This mechanism is rather complex and not well documented in the literature. During solidification, Si is completely rejected by the growing carbide. Consequently, a high Si layer will surround the primary or eutectic carbides. If the Cr content is high, then the austenite being formed will then also be low in carbon. Starting with low-C austenite, the localized high-Si regions around the carbides will further lower the dissolved C content in the austenite.

Whether ferrite is formed due to a shift in the solidification or due to solid-state transformation upon cooling to room temperature is unclear. Nonetheless, large amounts of free ferrite (i.e., ferrite that is not locked up as pearlite) were found in an experimental iron having a chemical composition of 2.6% C, 2.6% Si and 20% Cr. In commercial practice, following commercial alloy guidelines, solid-state transformation of austenite to ferrite will not be found.

## Ferrite Mechanical and Physical Properties

In practical terms, ferrite is very soft. Of all the phases or allotropes of iron, ferrite is the softest. Although its hardness can be raised by the addition of alloying elements, e.g., Cr, it is basically soft, with hardness values between 215 and 270 DPH.

Chemically, the only real change of interest occurs when Cr is added in excess of 11%. At these higher levels, the ferrite matrix behaves almost identically to that of the austenitic matrix (i.e., a corrosion-resistant steel).

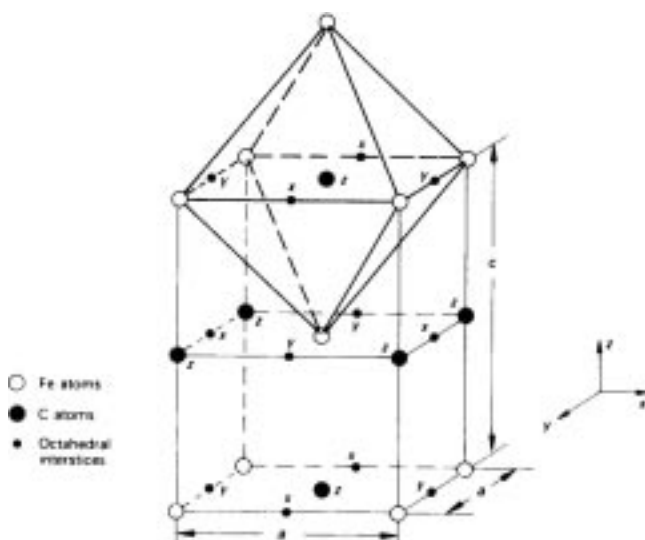
## Martensite Morphology

Abrasion-resistant cast irons can be cast having ferritic or austenitic matrixes, but never martensitic matrixes. This may sound a bit pedantic, but it focuses our attention on the order of events in AR cast irons. If the iron is cast austenitic, the iron has the possibility of transforming to martensite upon cooling to room temperature, as noted in Chapter 3 on Heat Treatment.

Assuming that the cooling rate or hardenability is sufficient, austenite will undergo a sudden lattice shift or crystallographic arrangement. This shift is diffusionless and almost instantaneous. The end result of this shift is that the FCC crystal is transformed into another arrangement, called body-centered tetragonal (BCT). What is most unique about this transformation is that it is diffusionless. The original C content of the austenite is now the C content of the martensite. Figure 5.24 gives an idea of the BCT crystal structure and where the C atoms reside.

The BCT crystal structure shown in Fig. 5.24 is similar to the BCC structure. The high-C content distorts the BCC structure to such an extent that the structure is no longer symmetric. As C is increased, the BCT structure becomes more and more distorted (Fig. 5.25). This has the effect of increasing the hardness of the martensite. Additionally, when FCC austenite transforms to BCT martensite, a volume expansion occurs that is nominally around 4.3%. This volume expansion is shown in Fig. 5.25. It is interesting to note that, as the C content of the austenite increases, the resulting volume expansion, due to martensitic transformation, decreases. Nonetheless, this decrease is fairly minor, compared to the overall magnitude of the transformation.

In most cases, people just assume that a volume expansion of 4% occurs as austenite transforms to martensite. This is a good assumption as shown in Fig. 5.25. It should be noted that this volume expansion assumes 100% transformation of austenite to martensite. However, let's take an AR iron having a 30% CVF with an austenitic matrix having a C content of 0.8%. If 80% of the austenite matrix transforms to martensite (i.e.,  $70\% \cdot 80\% = 56\%$  of the total structure will transform from austenite to martensite), the iron will then expand about 2.4% ( $0.56 \cdot 4.3\% = 2.4\%$ ).



**Fig. 5.24.** The martensite BCT showing the octahedral sites (marked as x, y and z) where C atoms can reside. Sufficient C must exist to fully populate the z-set to create a stable BCT crystal structure. (After Cohen, 1962)

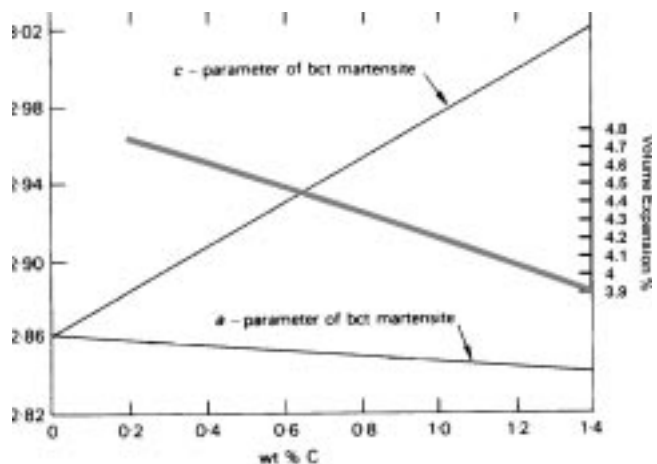
This discussion on the volume expansion during the transformation from austenite to martensite is important to the understanding of heat treating and wear performance of AR cast irons. That is, if there is inconsistent transformation in the iron, extremely high tensile stresses will develop leading to catastrophic failure, either in the foundry or in the field.

The morphology of the martensitic phase is typically described as lath-like or lenticular. The image of a lath as a long and skinny rectangle is easy to visualize. The image of a lenticular plate is much harder to visualize. One way to think of it is to visualize a double-convex lens or a long skinny ellipse having sharp points on both ends. The type of morphology that forms is determined by a number of factors. As a general rule, Ni-Hards 1, 2 and 4 will form lenticular or plate-like martensite, while the high-Cr irons will tend to form lath-like martensite.

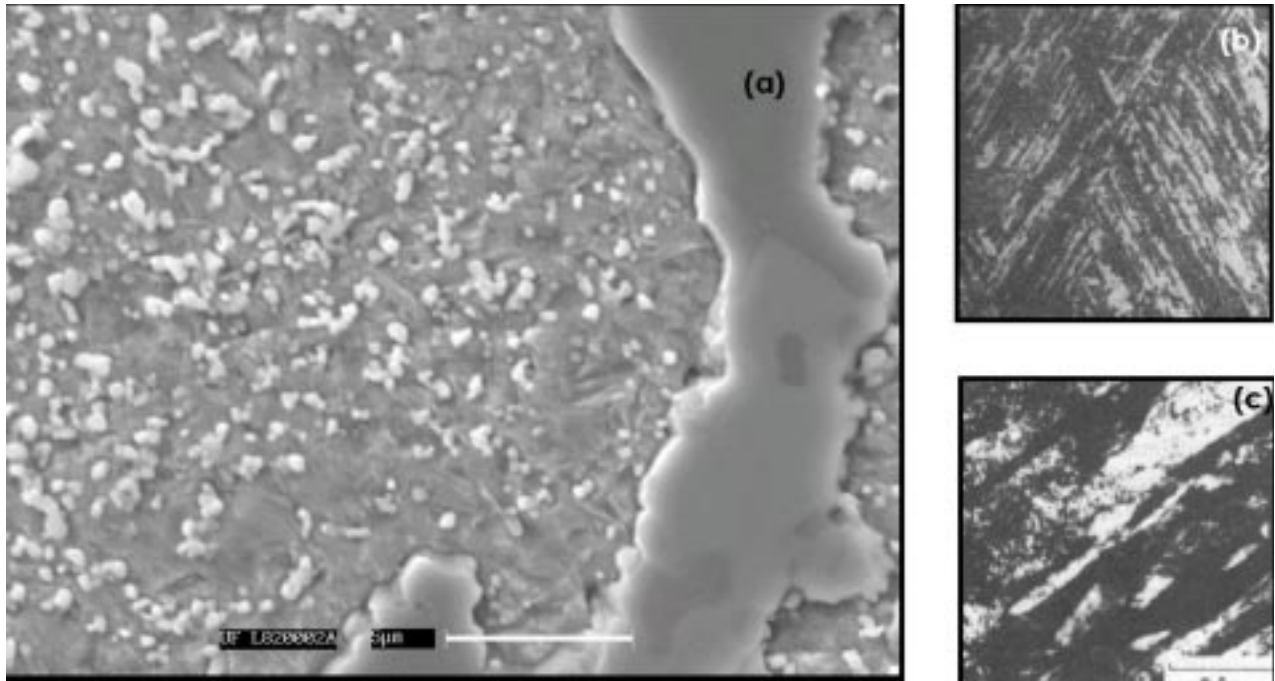
Figure 5.26 shows some examples of lath-like martensite. In Fig. 5.26a, very fine striations can be seen among the secondary carbides of the destabilized iron. These striations provide a weak description of the morphology of lath-like martensite in high-Cr irons. It is easier to use the work performed by Honeycombe and Bhadeshia (1995) on low-C steels showing the more characteristic lath-like structure, as can be seen in Figs. 5.26b and 5.26c.

In all cases, lath-like martensite is arranged in somewhat parallel sheets within a former austenite grain. This sheet-like characteristic gives lath-like martensite high toughness. Between these sheets of martensite are fine layers of retained austenite. These alternating layers of lath-like martensite and retained austenite have been exploited to produce commercial grades of high-strength, high-toughness steels.

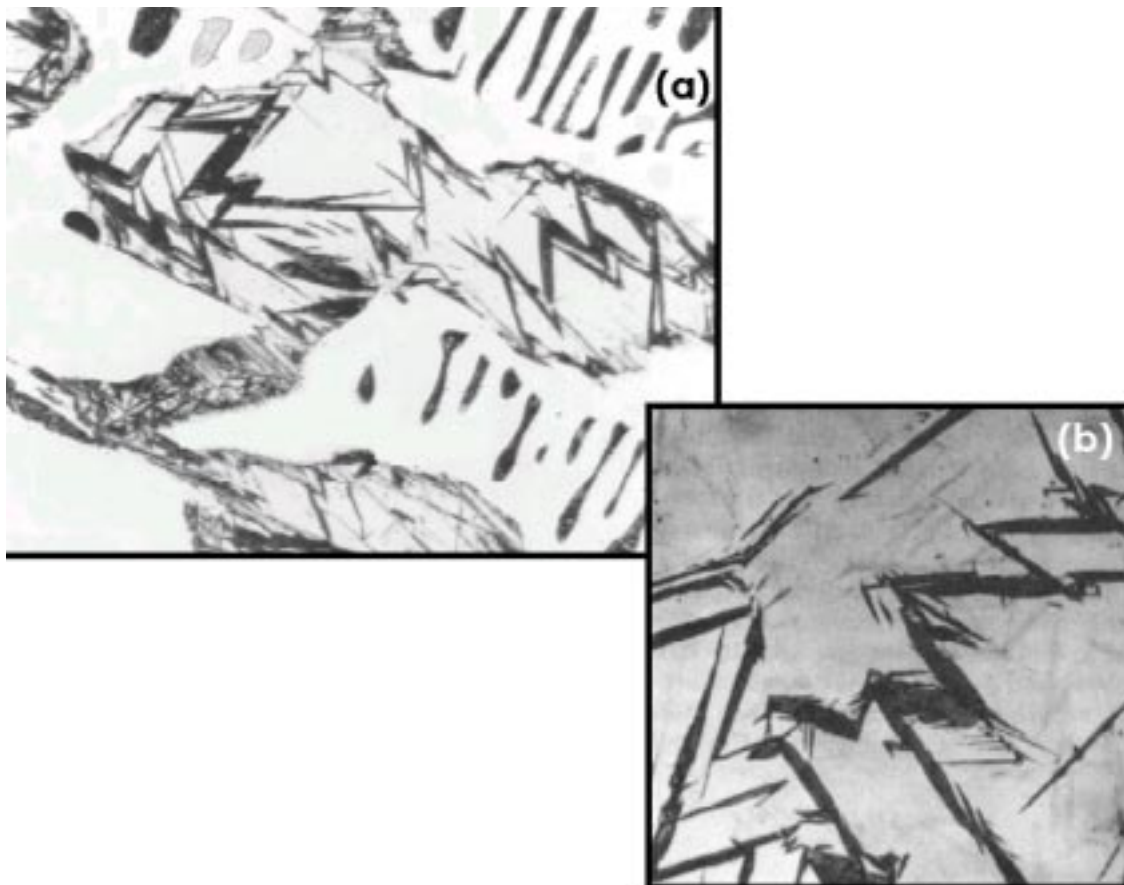
Examples of lenticular martensite are shown in Fig. 5.27. The formation of lenticular martensite is quite energetic and results in one plate nucleating another plate, creating the zigzag pattern shown in Figs. 5.27a and 5.27b. It has been speculated that microcracks are often initiated at the apexes of these lenticular plates, due to their rapid growth and tendency to initiate at right angles from other plates.



**Fig. 5.25.** As the C content of the austenite increases, the volume expansion of the martensite formed from this austenite decreases.



**Fig. 5.26.** Morphology of lath-like martensite in (a, left) high-Cr iron and (b and c, right), in a low-C steel (0.16% C). Discerning the martensitic morphology in the high-Cr iron is nearly impossible due to the presence of secondary carbides. (5.26a is courtesy of K. Collins, Dept. of Energy, Albany Lab.; 5.26b: c after Honeycombe and Bhadeshia, 1995)



**Fig. 5.27.** High-C martensite forms almost explosively. This burst phenomenon results in one martensitic plate nucleating another, along preferred crystallographic planes. The characteristic morphology is that of lenticular plates. The end result is the jagged pattern shown in upper left (a Ni-Hard 1 iron) and in lower right (an iron having a composition of 1.8% C, 3% Mn and 2% Si). (After Honeycombe and Bhadeshia, 1995)

If conditions are right for lenticular martensite, they are also right for retained austenite. As the amount of C increases in the austenite, the martensite start temperature is lowered. As the martensite start temperature is lowered, then the austenite becomes more stable, and more of it is retained upon reaching room temperature. Irons that form lenticular martensite will naturally have a greater volume fraction of retained austenite in the as-cast condition. As an example, lenticular martensite can be seen in any high hardness as-cast Ni-Hard 1, 2 or 4 iron, along with significant volume fractions of retained austenite.

### Martensite Mechanical and Physical Properties

In wear irons, the most important property of martensite is its hardness. The hardness of the martensite is determined by the amount of distortion in the BCT crystal structure. As the C content increases in the BCT lattice, the lattice distortion increases (see Fig. 5.25). As the BCT lattice is distorted or strained, its resistance to dislocation movement increases or, as given in Fig. 5.28, its hardness increases. In the simplest terms, the C content of the martensite determines its hardness.

Although extremely high martensite hardnesses are possible, as shown in Fig. 5.28, reality is somewhat different. As the C content increases, it becomes more and more difficult to transform the austenitic structure to one that is dominantly martensitic. Consequently, the resulting microstructure fails to achieve what you may believe should be its “true” hardness. For example, Fig. 5.29 shows the resulting hardness of an austenitic structure that only partially transforms to martensite upon cooling to room temperature.

Martensite is not the most stable phase in ferrous-based alloys. Due to its highly distorted BCT lattice structure, it has a tendency to decompose to its baser constituents, such as carbide and ferrite, or, if it is a high-C martensite, to a lower C martensite, with the excess C precipitated out as carbide at fairly low heat-treatment temperatures. As with most metastable phases, this process of decomposition can be accelerated by an increase in temperature.

The term for this temperature-accelerated martensitic decomposition process is called *tempering*. As covered in Chapters 3 and 4 on Heat Treatment and Hardenability, the temper process allows the highly strained BCT lattice to rid itself of a few C atoms, forming lower C martensite and epsilon ( $\epsilon$ ) carbide. If continued to equilibrium, the end result of this process is the formation of a mixture of ferrite and carbide.

A side effect of this tempering process is that the martensite lattice will contract or shrink. As C is dumped from the BCT lattice, the lattice contracts, resulting in a shrinkage of the casting from its earlier distorted shape. This process is commonly used by heat-treatment specialists to remove warpage in martensitic parts. Tempering typically softens martensite. Nevertheless, depending on the alloy's C content and tempering temperature, the iron may harden. Figure 5.30 shows a plot of how the hardness of martensite is affected by the tempering process.

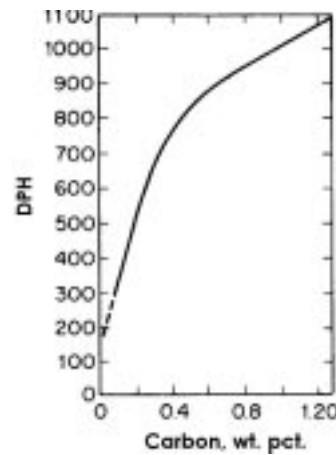


Fig. 5.28. The effect of C on the hardness of martensite. It should be kept in mind that, as the C content increases, it becomes more and more difficult to form martensite. In most irons, the C content is between 0.4 and 1% C, yielding a range of hardnesses from 700 to 940 DPH. (After Bain, 1939)

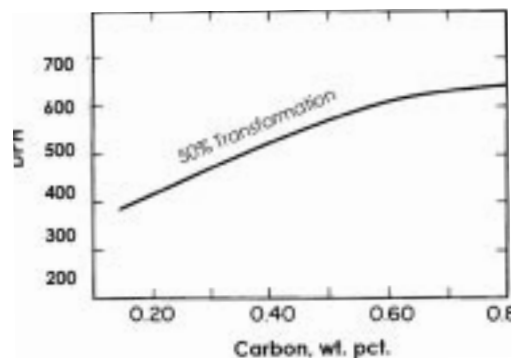


Fig. 5.29. In most irons, complete transformation of austenite to martensite rarely occurs. For example, if only 50% of the austenite transforms to martensite, the resulting hardness of the bulk matrix would be much softer than otherwise expected. (After Hodge and Orehoski, 1945)

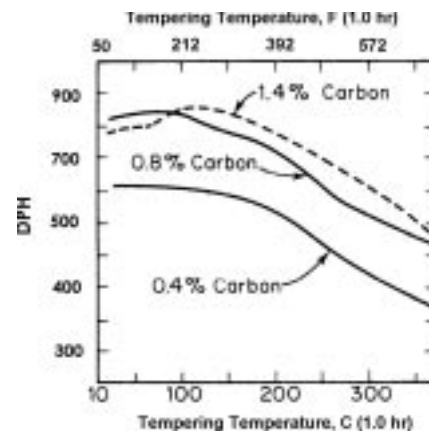


Fig. 5.30. If very little retained austenite is present, tempering will lower the hardness of fresh martensite. At the higher C content (e.g., >0.8% C), tempering will also promote the transformation of retained austenite to fresh, hard martensite, yielding a small increase in hardness at temperatures just above room temperature. At higher temperatures (above 100C/212F) and higher C content (e.g., 1.4% C), an increase in hardness may result due to the formation of epsilon carbide. (After Lement et al., 1954)

As might be expected, the chemical stability of martensite is much less than that of austenite or ferrite, due to its highly strained state. When exposed to acids (for example a dilute solution of  $\text{HNO}_3$ ), the martensitic phase will be readily dissolved in preference to austenite or ferrite. The end result of this process allows the metallographer to clearly confirm whether or not martensite is present within the alloy.

## Nanocomposite Phase Morphology

Pearlite and bainite are often considered as individual phases. In reality, they are actually *nanocomposites* of ferrite and  $\text{M}_3\text{C}$  arranged in different morphologies. (The size of the carbide phase present in pearlite and bainite is on the scale of hundreds of nanometers.) We have already discussed AR cast irons as being composites of carbides and a ferrous matrix, and it is useful to start the discussions on pearlite and bainite from the same viewpoint. That is, we have actually defined all of the major phases and constituents of AR irons (and, for that matter, all other ferrous alloys). The rest of this chapter will focus on the combination of these phases and constituents to form higher-level microstructures (e.g., pearlite or an actual AR iron).

### Pearlite Phase Morphology

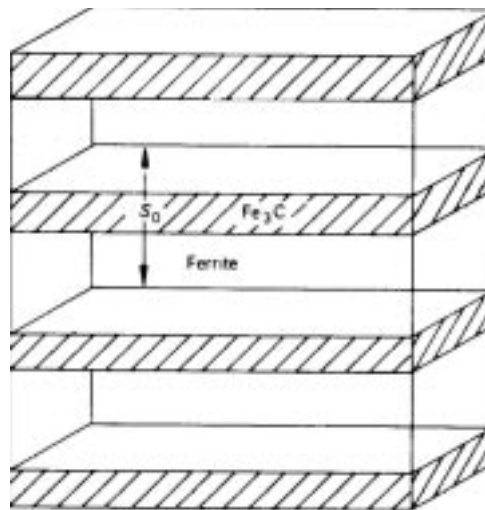
Pearlite has been said to be the most familiar microstructural feature in the whole science of metallography. In 1887, H. C. Sorby was the first to microscopically observe “pearlite” and correctly assume that it was a lamellar mixture of iron and iron carbide. The name pearlite arises from its luminescent optical appearance, much like the inside surface of an oyster shell or mother-of-pearl. Figures 5.31a–5.31c provide some examples of this morphology.

As covered in Chapter 3 on Heat Treatment, pearlite grows by a diffusion-controlled process. Pearlite colonies nucleate and grow from austenite grains. The key to this growth process is the high mobility of carbon. Because ferrite will tolerate almost no carbon, the difference between the austenite’s C content must be “dumped” when forming the  $\text{M}_3\text{C}$  carbide.

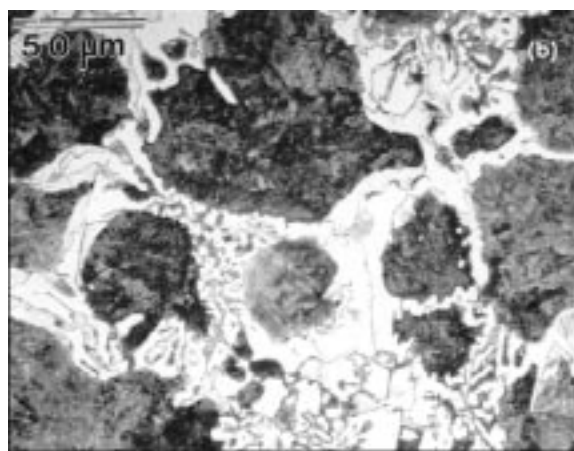
It is relatively easy to calculate the amount of carbide formed within the pearlite microstructure. (For simplicity, we will assume that the density of the original austenite, ferrite and carbide are approximately similar.) At the eutectoid, the C content is 0.8%. Then, using a simple form of the lever rule and logically assuming that the C content of ferrite is zero:

$$\text{Pearlite CVF} = 0.8 / 6.7 = 12\% \quad (5.1)$$

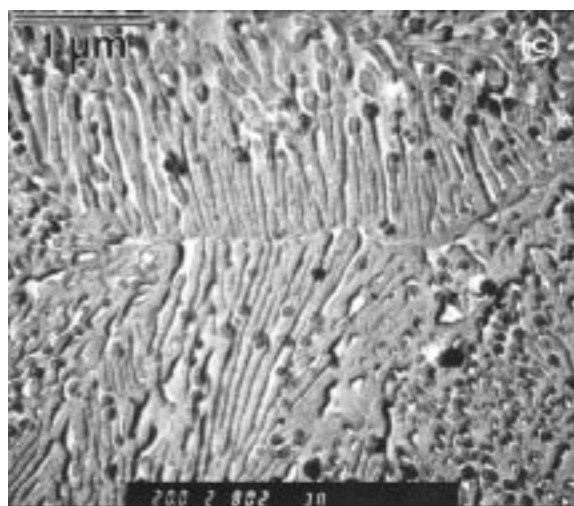
The morphology of the pearlite will always consist of layers of carbide and ferrite. The spacing between these layers is controlled by complex relationships between alloying and temperature, which are well covered in any basic metallurgy textbook. However, for completeness, some rules of thumb are given for controlling the interlamellar spacing in pearlite:



(5.31a)



(5.31b)



(5.31c)

**Fig. 5.31. Pearlite is the most common nanocomposite phase found in steels and irons. Its structure is defined by a lamellar arrangement of cementite or  $\text{Fe}_3\text{C}$  and ferrite. For AR irons, (b) shows the classic pearlitic morphology in a high-Cr iron (3% C, 15% Cr) after heat treating for 12 hr at 650C (1202F). At much higher magnifications (20,000X) (c), the lamellar microstructure can be seen.**

- Because the spacing between the carbide plates is controlled by the diffusion of C (C is almost completely rejected from the ferrite and hence, can only go into the carbide), temperature has a dominant effect on the pearlite morphology.
- In general, coarse pearlite is formed at high temperatures, while finer pearlite is formed at lower temperatures.
- Alloying elements, such as Cr, Ni, Mo and Mn, also have a powerful effect on pearlite refinement by affecting the nucleation and growth of pearlite.

Needless to say, the morphology of pearlite can be tailored by alloying and thermal treatments to achieve a wide range of mechanical properties.

### **Pearlite Mechanical and Physical Properties**

The hardness of pearlite is typically not much of a concern for producers and users of AR cast irons. Table 5.3 shows a range of 250 to 450 DPH for pearlite. Pearlitic hardness is controlled by the interlamellar spacing of the ferrite and carbide layers. Pearlite with extremely fine layers of carbide will give a higher hardness than will pearlite with extremely coarse layers. Commercially produced pearlitic irons will also typically contain some austenite and, maybe, a bit of martensite. This multiplicity of phases makes it very difficult to pinpoint the actual hardness of the pearlite phase.

The chemical stability of pearlite follows the behavior for that of ferrite and  $M_3C$  carbide. When chemically attacked, the ferrite will dissolve, first producing the optical image shown in Fig. 5.31b.

### **Bainite Phase Morphology**

Bainite, named after its discoverer E. C. Bain, is another nanocomposite structure composed of ferrite and carbide. In perspective, pearlite has a high temperature reaction occurring between 500 and 750C (930 and 1380F), while martensite has a very low temperature reaction starting at temperatures around 250C (480F). This leaves a wide range of temperatures, from 250 to 500C (480 to 930F), in which neither of these phases form. This is the bainitic region, where fine aggregates of ferrite plates (or laths) and  $M_3C$  particles form.

Bainite has characteristics of both pearlite and martensite. For example, the growth of the ferrite phase is only partly controlled by C diffusion. Consequently, bainite will grow about an order of magnitude faster than would be expected if it was based strictly on C diffusion. This faster growth rate is attributed to the ferrite's morphology. As shown in Fig. 5.32, a small sub-unit (or ferrite sliver) is nucleated and grows along a preferred crystallographic plane. During its growth, C is only partly dumped into the surrounding austenite. Immediately after its growth, we would find the ferrite lath supersaturated with carbon. Subsequent to further growth, the ferrite lath must dump C into the surrounding austenite or form small carbide particles, which, for our highly alloyed irons, are  $M_3C$ . By growing as a small sliver of supersaturated ferrite

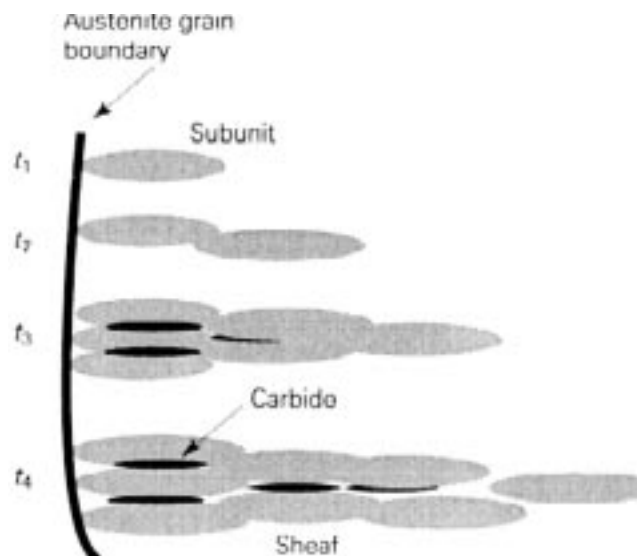
without being totally dependent on C diffusion, bainite has the unique distinction of having a growth rate between that of pearlite (i.e., completely dominated by C diffusion) and martensite (i.e., diffusionless transformation).

When the term bainite is used, one can also specify it as upper or lower. Although both forms can be present in AR cast irons, both forms are still comprised of ferrite and carbide, albeit in different morphological arrangements. Upper bainite, which will form, logically enough, at higher temperatures, is a coarser mixture than that of lower bainite, which is formed at lower temperatures. Since bainitic AR cast irons are not industrially prevalent, our discussion of bainite will be brief.

The transformation of austenite to bainite results in a volume expansion of about 3%. This transformation is quite aggressive and causes large distortions within the parent austenite matrix. Figure 5.33 provides an example of the severity of this transformation. In many respects, it is identical to that of the martensitic transformation, except that it occurs at much higher temperatures. Since it does occur at higher temperatures, some of the transformation strain is relaxed due to dislocation movement.

The morphology of bainite looks quite similar to that of plate or high-C, lenticular martensite. For example, Fig. 5.34 shows an example of a bainitic AR cast iron and, if you didn't know it, you might say that it was martensite and not bainite. The ferrite plates etch similarly to martensite, i.e., both phases are readily dissolved by acids. The result of this metallographic process is to make the ferrite plates appear dark against the relatively unattacked austenite phase

In AR irons, the complete transformation of austenite to bainite is extremely difficult due to their high-alloy contents. As mentioned in Chapter 3 on Heat Treatment, the bainitic start temperature is easily depressed by C, Mn, Ni, Cr and Mo—all elements commonly found in AR cast irons. If



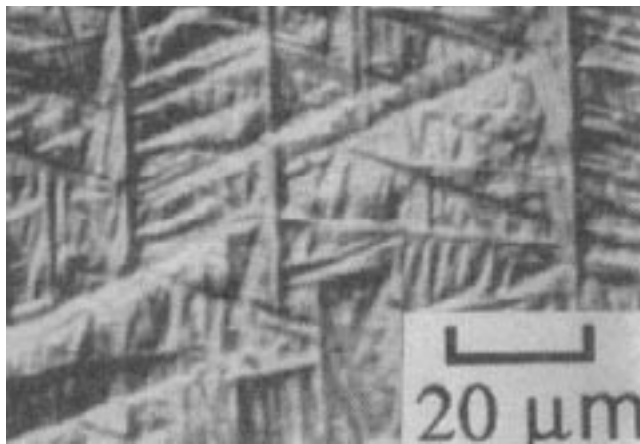
**Fig. 5.32. Schematic illustration of bainite growth (from  $t_1$  to  $t_4$ ) within an austenite grain. Initially, only a small ferrite subunit forms ( $t_1$ ). As growth progresses, carbide is formed and a full bainite sheaf appears ( $t_4$ ). (After Honeycombe and Bhadeshia, 1995)**

bainite is observed in an AR iron, it will represent only a portion of the metallic matrix and, as said by Maratray and Usseglio-Nanot (1971), “it is not possible to get more than 50% bainite, even after long holding times.”

#### **Bainite Mechanical and Physical Properties**

The hardness of bainite is quite varied depending on the transformation temperature and the amount of C in the austenite, prior to transformation. Considering its variability no one hardness number can be quoted for bainite.

The chemical inertness of bainite is similar to that of martensite. It is easily attacked by most acids since, like martensite, it is a highly strained structure. Interestingly enough, bainite is resistant to tempering. Since the bainitic structure is composed of ferrite and carbide, a low temperature temper (below 250C or 482F) has little effect on the microstructure and hardness of bainite. At higher temperatures, the carbides embedded or between the ferrite plates will tend to coarsen and some degradation of hardness will result.



**Fig. 5.33.** Bainite morphology in a medium-C steel. The surface was first polished and then transformed to bainite. The transformation from austenite to bainite results in localized distortion. (After Honeycombe and Bhadeshia, 1995)

## **Metallic Matrix Microstructures**

We have covered the basic building blocks of AR cast irons. The purpose of the following sections is to show how these fundamental constituents and phases can be arranged to form the wide range of in-situ metallic composites otherwise known as AR cast irons. Some comments will be made about hardness, toughness and wear characteristics as they relate to microstructure; however, specific, alloy-dependent mechanical and wear data will be given in Appendix B.

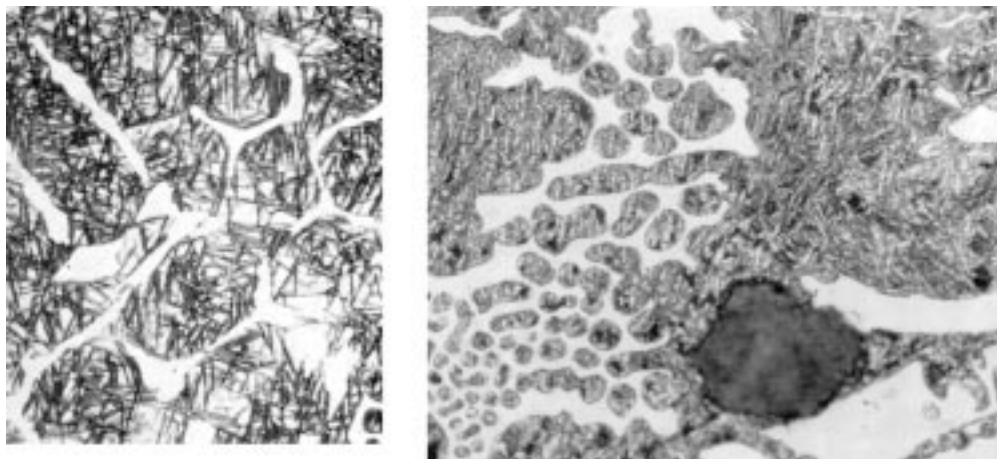
### **M<sub>3</sub>C Eutectic Carbide Category**

M<sub>3</sub>C eutectic carbides characterize this group of irons. Depending on the ferrous matrix phase, the iron is classed as pearlitic or as a Ni-Hard. Pearlitic irons are self-descriptive while the Ni-Hards require some explanation. The metallic matrices of Ni-Hard 1 and 2 are mixtures of austenite and martensite. The proportion of austenite versus martensite is a function of the alloy chemistry and its cooling rate subsequent to solidification (see Chapter 4 on Hardenability). Regardless of the irons' matrix phase, the M<sub>3</sub>C carbide network strongly influences both the microstructure and mechanical properties.

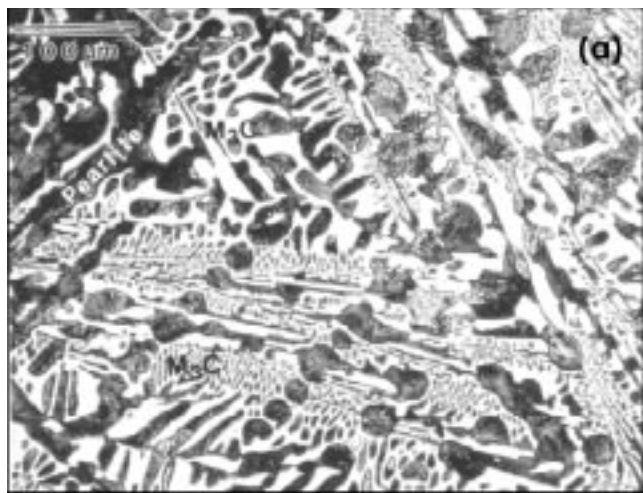
#### **Pearlitic AR Cast Irons**

Figures 5.35a–5.35c show examples of how the constituents of a pearlitic AR cast iron fit together. Although this comment may seem redundant, the secret to understanding AR cast irons is to understand how the carbides and matrix fit together to form the iron as a whole. The carbides should always be thought of as a separate hard phase. Heat treatments merely change the proportion or type of phase present in the metallic matrix.

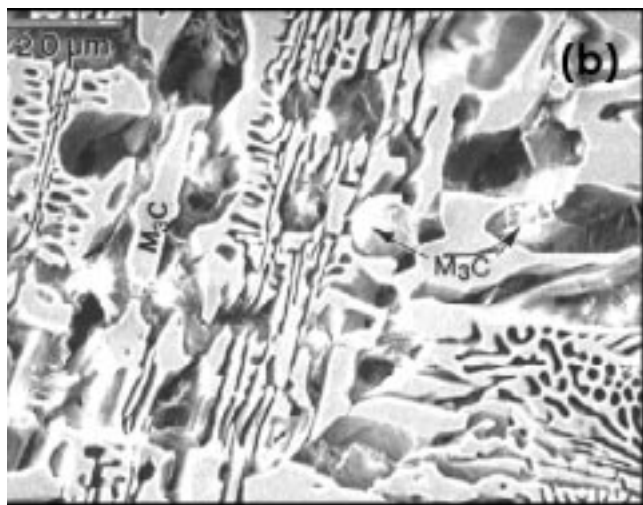
In the case of pearlitic AR irons, heat treatments usually consist of a stress relief and rarely reach into the realm of causing wide-ranging changes in the microstructure of the matrix. One direct way to change the mechanical behavior of this class of irons is to change the carbide volume fraction



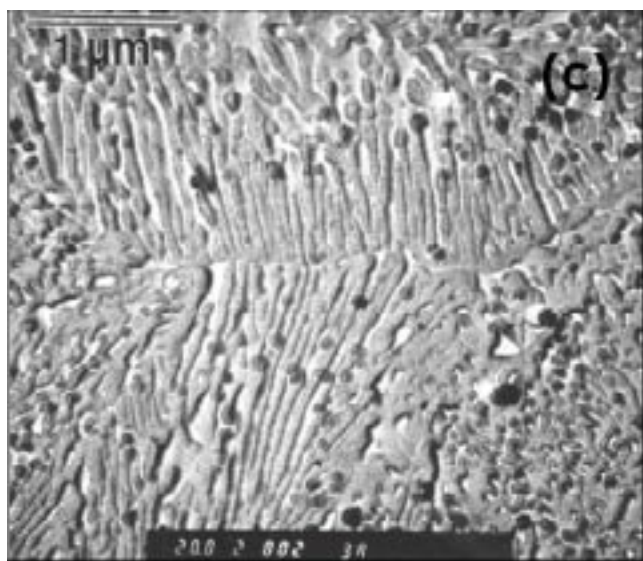
**Fig. 5.34.** Bainitic high-Cr AR cast iron (2.6% C, 12.4% Cr, 1.2% Mo) after duplex heat treatment consisting of a 1000C (1832F) destabilization, followed by a 24-hour hold at 250C (482F); and a low-Cr AR iron containing a mixed microstructure of M<sub>3</sub>C carbide, nodule graphite and ferrous matrix. The metallic matrix in both irons is a mixture of bainite and austenite. The dark, lenticular structures are the easily etched bainitic phase, while the more resistant austenite phase remains unattacked and appears much lighter. (After Maratray and Usseglio-Nanot, 1970, and courtesy of United Foundries)



(5.35a) standard optical 2D view



(5.35b), 3D view of M<sub>3</sub>C carbide network



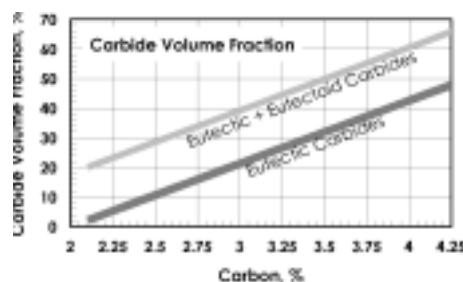
(5.35c) high resolution image of pearlite matrix

Fig. 5.35. Constituents of a pearlitic AR cast iron.

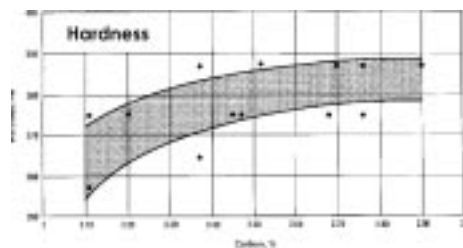
(CVF). Increasing the alloy's CVF is accomplished by increasing its C content. Figures 5.36a–5.36c provide a look at the relationships between C content (CVF) and mechanical properties for pearlitic irons. As the CVF increases, the hardness logically increases and, as might be expected, the toughness of the iron decreases. The decrease in toughness as the CVF increases is due to the brittle nature of the M<sub>3</sub>C carbide. This is a fixed relationship for pearlitic irons.

Pearlite lamellar spacing can also influence the toughness of the iron, to a lesser degree. Control of lamella spacing is a function of cooling rate and alloying as detailed in Chapter 4 on Hardenability. More extensive mechanical property data is provided in Appendix B.

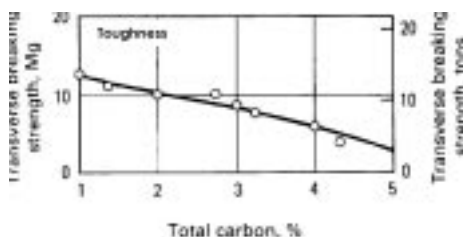
To reaffirm this message about the linkage of microstructure to mechanical properties, Fig. 5.37 shows the one-to-one relationship between C content and hardness in pearlitic AR cast irons. There will always be a hardness variability in the final product due to differences in cooling rates between the irons and, in these very low alloys, the potential to form graphite during solidification (albeit marginal, in most cases). Considering these limitations, the graphical fit shown in Fig. 5.37 is actually quite good.



(5.36a) CVF



(5.36b) hardness



(5.36c) toughness

Fig. 5.36. The mechanical properties of pearlitic AR cast irons are tied directly to the iron's CVF since the matrix varies only slightly as function of alloy content.

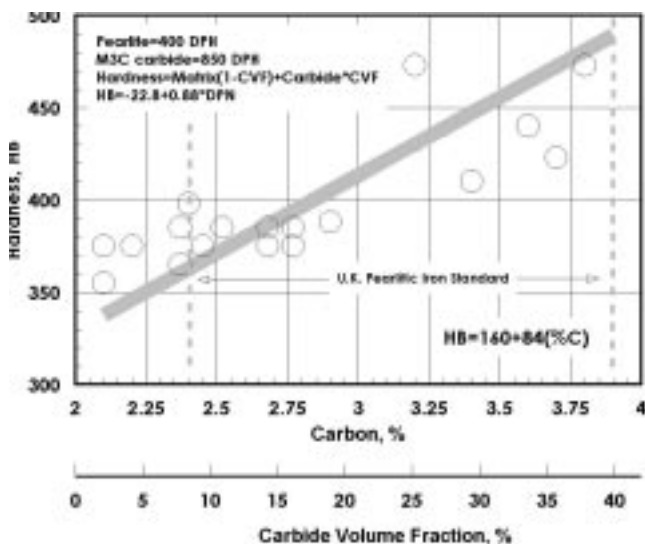
The reason we have some success in predicting the hardness based on the alloy's C content is twofold. The main reason is that the matrix is pearlitic. Second, the alloy must be chill-cast to obtain its carbide structure. As such, the carbide structure increases linearly as the C content is varied. Once the discussion moves to irons that are not chill-cast and have matrices that are mixtures of austenite, martensite, pearlite or bainite (next section), it will be found that simple predictions about mechanical properties are difficult.

#### **Ni-Hard 1 and Ni-Hard 2 AR Cast Irons**

Although Ni-Hard 1 and 2 irons are typically used in the as-cast or as-cast-and-tempered state, their matrix microstructure varies based on subtle changes in alloy chemistry and cooling rate. The eutectic  $M_3C$  carbides remain inert while the metallic matrix changes from austenite to a mixture of austenite, martensite and—if the alloy chemistry is excessively lean (e.g., low on Ni) and/or the cooling rate excessively slow—pearlite. In the Ni-Hard alloys, a number of different matrix microstructures are possible.

Figures 5.38a–5.38c show the microstructure of a Ni-Hard 1 alloy. The microstructure is essentially a mixture of  $M_3C$  carbides, martensite and austenite. The ratio of these three phases controls the mechanical response of the iron. Considering that most foundries strive for this mixed microstructure, efforts will be focused on describing how variations in CVF, martensite and austenite affect the bulk mechanical properties of Ni-Hard 1 and 2.

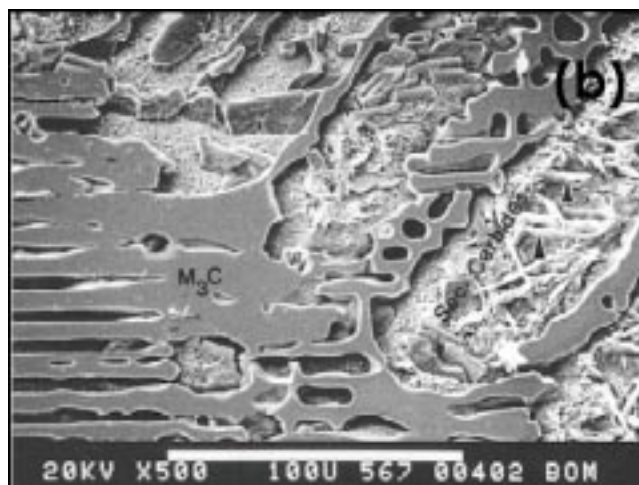
The large martensitic plates within the austenite dendrite are characteristic of the matrix microstructures found in the



**Fig. 5.37.** Hardness in pearlitic AR cast irons is estimated using a simple rule-of-mixtures calculation. The hardness follows the U.K. specification for this iron, which is from 350 to 500 HB for C content that ranges from 2.4 to 3.9%. Hardness values for the pearlite and carbide phase were estimated from Tables 5.1 and 5.3.



**(5.38a)** standard optical 2D view



**(5.38b),** 3D view of  $M_3C$  carbide network



**(5.38c)** high resolution image of martensite/austenite matrix

**Fig. 5.38.** Constituents of a Ni-Hard 1 AR cast iron.

Ni-Hard family. If the origin of the austenite phase was traced, it would be found that its C content is near the maximum solubility limit of 2% (see Chapter 2 on Solidification). As the iron cools, some of this C is dumped from the austenite matrix into proeutectoid  $M_3C$  carbides (known as  $M_3C$  secondary carbides). The quantity of secondary carbides formed in the Ni-Hard alloy is dependent, primarily, on the cooling rate. That is, slower cooling allows greater nucleation and growth of secondary carbides. This results in a lower dissolved C content in the austenite. Typically, depending on section size, shakeout temperature and alloying, the austenite will have a C content somewhere around 0.6% as it approaches the martensite start ( $M_s$ ) temperature.

With its relatively high C content, the  $M_s$  temperature will be depressed forming more plate-like, lenticular martensite than lath-like martensite. Moreover, since we are dealing with high-carbon austenite, the transformation to martensite will be incomplete and significant fractions of retained austenite will remain within the microstructure. In many commercial irons, the microstructure is primarily austenitic. This austenite is very unstable and easily transformed to hard martensite upon impact loading or severe abrasive wear. This can be a mixed blessing, with wonderful results under straight abrasive wear or disastrous results under high impact conditions. The stability of retained austenite and its effect on abrasion and impact is covered in subsequent chapters.

Figure 5.39 provides an overview of how the hardness of Ni-Hard 1 and 2 can vary as a function of the alloy's CVF. Because the CVF is determined by the alloy's C content (see Chapter 2, Solidification), it makes sense to view the hardness of the alloy on the basis of CVF and matrix microstructure. This concept moves us away from viewing hardness or other

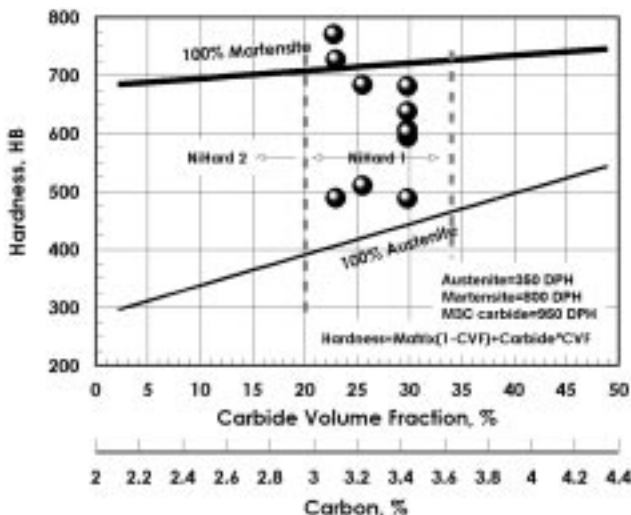


Fig. 5.39. Hardness relationship between CVF and the matrix microstructure in Ni-Hards 1 and 2. The circles are hardness data from Ni-Hards 1 and 2 castings. The hardness boundary lines were constructed using a rule-of-mixtures formula based on individual constituent hardness values estimated from data given in Tables 5.1 and 5.3. The dashed vertical lines show the ASTM Type IA (Ni-Hard 1) carbon limits.

mechanical properties as being determined by the alloy's chemistry.

Boundaries are provided on Fig. 5.39 showing hypothetical limits on hardness for the Ni-Hard 1 and 2 family of irons. As can be seen, the low-alloy Ni-Hard family can take on a broad range of hardnesses by varying the proportion of austenite to martensite at any given CVF. This relationship is directly shown in Fig. 5.40 where the ratio between austenite to martensite is varied as the CVF is held constant.

Besides the hardness variability associated with microstructural phases, hardness variability is also introduced by the carbide morphology. For example, if the hardness measurement is taken on the surface of a small casting, it is possible that the surface will have the characteristics of a chilled iron, while farther away from the surface, the carbides

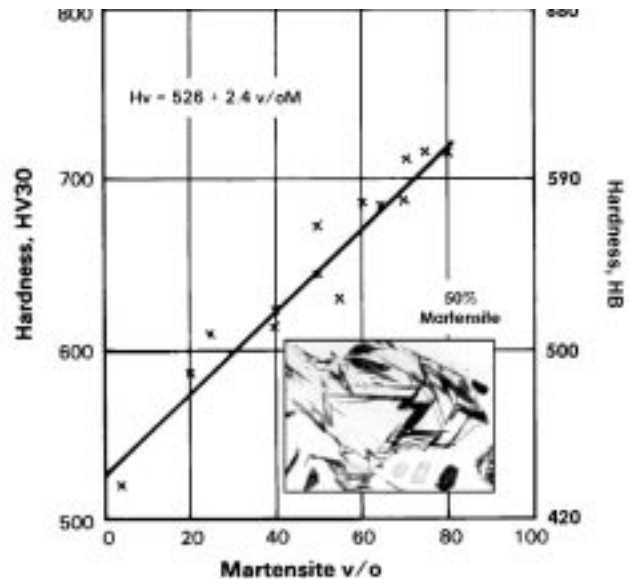


Fig. 5.40. The effect of varying the proportion of austenite versus martensite in a low-C Ni-Hard 2 iron (approximate CVF = 20%). At 0% martensite, the matrix is assumed to be 100% austenite. (After Cox, 1989)

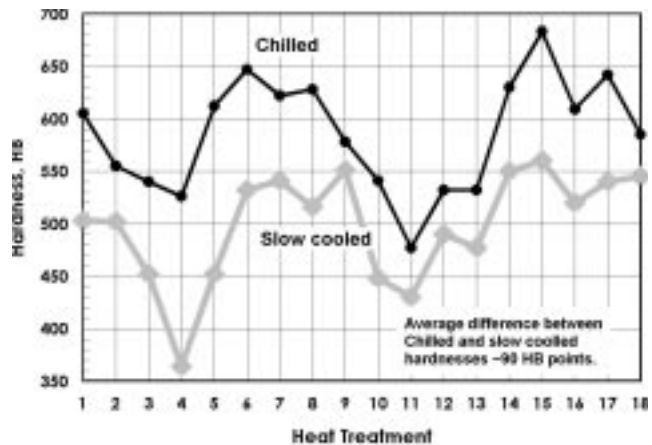


Fig. 5.41. A chilled microstructure will typically yield higher hardnesses than a nonchilled microstructure. Data was gathered from a Ni-Hard 1 keel block iron having a composition of 3.2% C, 1.1% Si, 4.3% Ni and 1.8% Cr that was subjected to 18 different heat treatments.

will be larger and more discrete. This behavior is shown in Fig. 5.41 where hardness data was taken at the tip of a keel block and then taken about 1 inch back from the keel. All hardness data is from similar Ni-Hard 1 compositions (3.2% C, 1.1% Si, 4.3% Ni and 1.8% Cr) that were subjected to 18 different heat treatments. The purpose of this investigation by Rote et al. (1951) was to construct a Time-Temperature-Transformation (TTT) diagram.

Because a chilled microstructure is always present in chilled cast irons, variation in hardness measurements due to carbide size is not always noticed. In thin-section AR cast irons, the surface microstructure takes on the characteristics of a chilled iron with many small carbides. Farther away from the surface, larger more discrete carbides form. Higher hardnesses will typically result from finer microstructures.

For decades, attempts have been made to break up the continuity of the  $M_3C$  carbide network in Ni-Hard 1 and 2 irons and thereby improve the iron's toughness and, perhaps, its abrasion resistance. One technique is to force the iron to

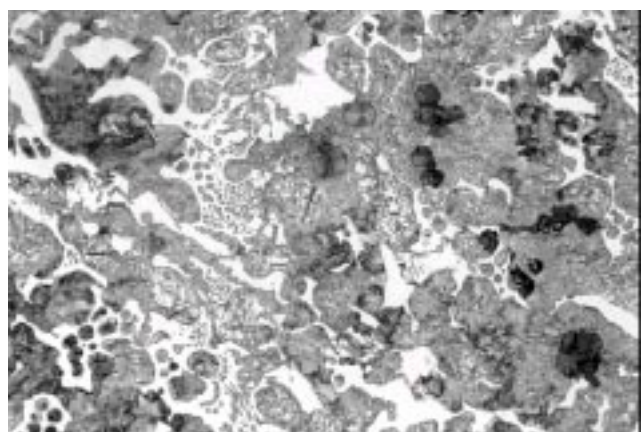
solidify with both carbide and graphite by increasing the iron's Si content.

An example of such an iron is shown in Figs. 5.42a and 5.42b. This iron was inoculated with Mg to produce nodular graphite. Although the  $M_3C$  carbide network is broken up, the mechanical property results are not very spectacular. Hardness is still adequate, in most cases, but the wear resistance does suffer significantly. For example, using the pin-on-drum abrasion test (see Chapter 8, Abrasive and Repetitive-Impact Wear Performance), a high-C Ni-Hard 1 (3.6% C) gave a value of 0.27 mm<sup>3</sup>/m (volume loss of material per meter of abrasive path traversed), while the graphite-modified Ni-Hard gave a value of 0.43 mm<sup>3</sup>/m. In comparison, a 1% C forged steel ball with an estimated hardness of around 750 HB yielded a value of 0.61 mm<sup>3</sup>/m. (Note: lower values indicate better abrasion resistance.) More complete mechanical and wear property data are provided in Appendix B.

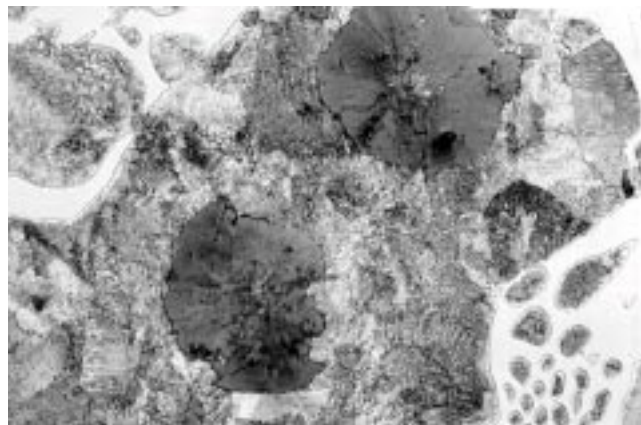
The real value of breaking up the continuity of the carbide network is in increasing the impact toughness of the iron. Without going into too many details, the graphite-modified iron had superior repetitive-impact resistance (spalling resistance) with a value of 0.14 mm<sup>3</sup>/impact (see Chapter 8 Abrasive and Repetitive-Impact Wear Performance) in a ball-on-ball impact test, while the forged steel material had a spalling resistance of 0.23 mm<sup>3</sup>/impact. (Note: lower values indicate better repetitive-impact resistance). Some advantage is gained by modifying the carbide network in Ni-Hard 1 and 2; however, the sacrifice in abrasion resistance is often too high a price to pay for most applications.

Since Ni-Hard 1 and 2 can have a variety of matrix microstructures and CVFs, their strength and toughness vary significantly. Mechanical property data on Ni-Hard 1 and 2 are extremely limited and this lack of data prevents the creation of property maps showing broad trends. Table 5.4 presents a listing of the available data on Ni-Hard 1 to provide some insight into how microstructure affects mechanical properties.

As shown in Table 5.4, it is difficult to make definitive statements about the relationships between CVF and bending strength or fracture toughness. Likewise, it is hard to say what type of matrix microstructure will provide the best strength-to-toughness ratio. This is unfortunate. This state of affairs is somewhat remedied by the much more complete information available for the high-Cr irons, later in the chapter. Much more extensive data is available linking microstructure to mechanical properties for this class of irons. Moreover, many of these correlations are directly applicable to the Ni-Hard 1 and 2 irons. High-Cr irons have mixed matrix microstructures of austenite, martensite, pearlite and bainite much as Ni-Hard 1 and 2 and, as covered, the CVF can similarly vary depending on alloy content. The main difference between the two categories of iron is the carbide type (i.e.,  $M_3C$  versus  $M_7C_3$ ). If allowances are made for carbide morphology, then the data for the high-Cr irons can be safely used to understand the general trends between microstructure and mechanical properties in Ni-Hard 1 and 2.



(5.42a) 200X



(5.42b) 400X

**Fig. 5.42. Mixed carbide and graphite eutectic morphology in a matrix of bainite and martensite. The composition is 3.1% C, 2% Si, 4.4% Ni, 2.6% Cr, with a ladle addition of 0.02% Mg. While the hardness is adequate at 620 HB, the abrasive wear resistance was quite poor due to the iron's lower CVF (13%, whereas, without graphite, it would be 23%) and the presence of very soft graphite that is easily pulled out during abrasive wear.**

**Table 5.4**  
**Microstructural/Mechanical Property Data for Ni-Hard 1**

| Alloy <sup>(1)</sup> | Chemistry                                  | Sample Geometry                                                 | Thermal History                       | Micro-structure                                   | CVF | Hard-ness HB | Bending Strength                   | Impact Energy               | Fracture Toughness (K <sub>IC</sub> )    |
|----------------------|--------------------------------------------|-----------------------------------------------------------------|---------------------------------------|---------------------------------------------------|-----|--------------|------------------------------------|-----------------------------|------------------------------------------|
| A1                   | 3.5% C,<br>0.5% Si,<br>4.1% Ni,<br>1.3% Cr | 22 mm dia. rod;<br>as-cast<br>surface.                          | As-cast                               | Aus. with<br>some Mart.                           | 32  | 460          | 375 Mpa <sup>(2)</sup><br>(55 ksi) | 23<br>Joules <sup>(3)</sup> | —                                        |
| A2                   | 3.5% C,<br>0.5% Si,<br>4.1% Ni,<br>1.3% Cr | 22 mm dia. rod;<br>as-cast<br>surface.                          | Destabilized<br>and air<br>Cooled     | Untempered<br>Mart. with<br>some residual<br>Aus. | 32  | 630          | 600 Mpa <sup>(2)</sup><br>(90 ksi) | 21<br>Joules <sup>(3)</sup> | —                                        |
| A3                   | 3.5% C,<br>0.5% Si,<br>4.1% Ni,<br>1.3% Cr | 22 mm dia. rod;<br>as-cast<br>surface.                          | 500C (932F)<br>hold and air<br>cooled | Bain. and<br>Mart., very<br>low retained          | 32  | 660          | 470 Mpa <sup>(2)</sup><br>(70 ksi) | 18<br>Joules <sup>(3)</sup> | —                                        |
| B1                   | 3.3% C,<br>1% Si,<br>4.5% Ni,<br>1.3% Cr   | Samples were<br>cut from<br>670-mm<br>dia rolls.                | Aus.<br>n.a.                          | Bain./Mart.<br>with some<br>graphite.             | 28  | 550          | n.a.                               | n.a.                        | 22 <sup>(4)</sup><br>Mpa√m<br>20 ksi√in. |
| B2                   | 3.5% C,<br>0.8% Si,<br>5.0% Ni,<br>1.9% Cr | Samples were<br>cut from<br>670-mm<br>dia rolls.                | n.a.                                  | Mart. With<br>some<br>graphite.                   | 32  | 600          | n.a.                               | n.a.                        | 22 <sup>(4)</sup><br>Mpa√m<br>20 ksi√in. |
| C1                   | 3.7% C,<br>0.8% Si,<br>4% Ni,<br>2.1% Cr   | Sand cast as<br>5-mm thick<br>plates. (chill<br>cast structure) | As-cast                               | n.a.                                              | 36  | 610          | n.a.                               | n.a.                        | 26 <sup>(5)</sup><br>Mpa√m<br>24 ksi√in. |
| C2                   | 3.7% C,<br>0.8% Si,<br>4% Ni,<br>2.1% Cr   | Sand cast as<br>5-mm thick<br>plates. (chill<br>cast structure) | 275C (525F)<br>and air<br>cooled      | n.a.                                              | 36  | 580          | n.a.                               | n.a.                        | 26 <sup>(5)</sup><br>Mpa√m<br>24 ksi√in. |
| C3                   | 3.7% C,<br>0.8% Si,<br>4% Ni,<br>2.1% Cr   | Sand cast as<br>5-mm thick<br>plates. (chill<br>cast structure) | 450C (850F)<br>and<br>275C (525F)     | n.a.                                              | 36  | 670          | n.a.                               | n.a.                        | 19 <sup>(5)</sup><br>Mpa√m<br>17 ksi√in. |
| D1                   | 3.2% C,<br>0.6% Si,<br>3.9% Ni,<br>2.2% Cr | n.a.                                                            | As-cast                               | Aus. and<br>Mart.                                 | 26  | 550          | n.a.                               | n.a.                        | 24 <sup>(5)</sup><br>Mpa√m<br>22 ksi√in. |

*Microstructure: Mart. = martensite; Aus. = austenite; Bain. = bainite.*

*(1) A = Data from Parent (1972); B = Eriksson (1973); C = Sare (1979); D = Diesburg and Borik (1974).*

*(2) Bending strength test was performed on 22 mm dia. rods with a support span of 290 mm.*

*(3) Impact test not described.*

*(4) ASTM compact tension specimen with fatigue precrack; a high-quality fracture toughness.*

*(5) Double-torsion bend test; no precrack. As a comparison, low carbon Ni-Cr-Mo steels gave values of around 80 Mpa√m (70 ksi√in.).*

*n.a. not available.*

## M<sub>7</sub>C<sub>3</sub> Eutectic Carbide Category

### Ni-Hard 4 AR Cast Irons

This alloy represents the start of a whole new class of irons. The crucial difference between classes is that the eutectic carbides are of the M<sub>7</sub>C<sub>3</sub> type versus M<sub>3</sub>C for the lower alloy irons. Barring this one difference, the irons are more similar than dissimilar. The matrix microstructure of Ni-Hard 4 is nearly identical to that of the Ni-Hard 1 and 2 irons containing a mixture of martensite and austenite.

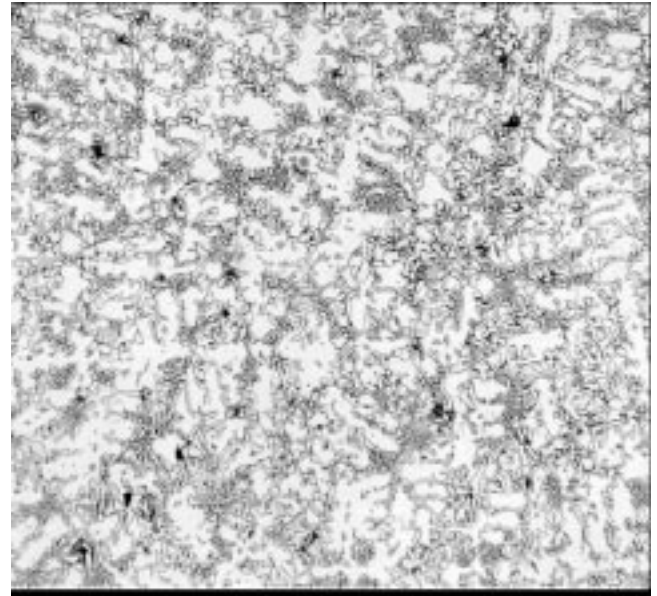
Figure 5.43a shows the general microstructure of a Ni-Hard 4 iron. In this case, the as-cast matrix microstructure is more austenitic than martensitic due to the alloy's higher-than-average Ni and Mn contents. In Fig. 5.43b, plates of martensite can be seen in the spaces between the eutectic carbides and occasionally venturing out into the former austenite dendrite. Correctly alloyed and cooled, Ni-Hard 4 irons will have amounts of retained austenite similar to that found in the Ni-Hard 1 and 2 irons.

The microstructure of as-cast Ni-Hard 4 is composed of an M<sub>7</sub>C<sub>3</sub> eutectic carbide network with a metallic matrix that is a mixture of retained austenite, martensite and just a few M<sub>3</sub>C secondary carbides (e.g., see Fig. 5.4b). If the iron is subjected to a destabilization heat treatment, the matrix microstructure becomes significantly different. After heat treatment, the matrix becomes heavily populated with secondary carbides and results in the austenite C content being lowered from around 1.6% to 0.9% C. On a volume fraction basis, 0.7% C results in the formation of 10% M<sub>3</sub>C carbide within the austenite dendrite. This is a lot of carbide. As an example, Fig. 5.44a shows the same iron as that shown in Fig. 5.43a after heat treatment.

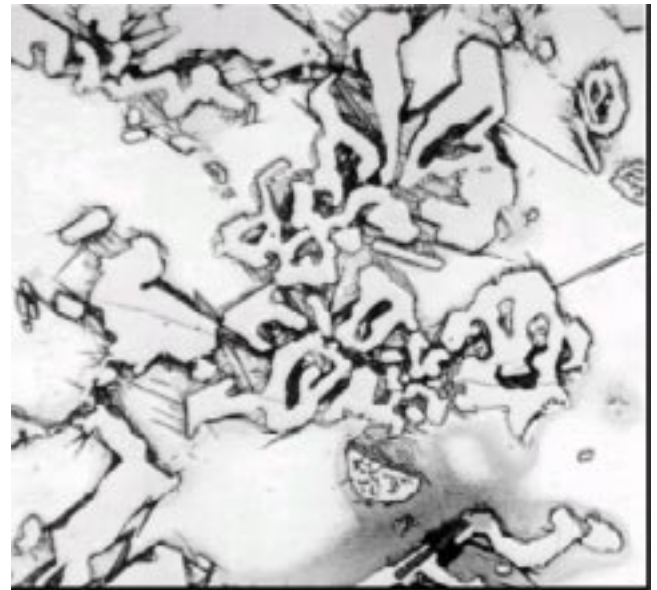
An interesting feature of Fig. 5.44b is that a significant fraction of retained austenite still remains after heat treating (the light regions shown between the M<sub>7</sub>C<sub>3</sub> eutectic carbides). Higher-than-expected retained austenite content in Ni-Hard 4 alloys are due to the alloys' higher Ni content (5–7%). The use of Ni is critical to obtain a hard as-cast microstructure in the Ni-Hard family; however, if the iron is heat treated, as shown in Fig. 5.44, the presence of high levels of Ni within the austenite dendrite will tend to stabilize the remaining austenite (Ni is completely rejected by the carbide phase) and hinder its transformation to martensite by depressing the martensite start temperature. Consequently, it is very difficult to obtain fully martensitic microstructures in Ni-Hard alloys at room temperature. As a result, subzero heat treatments are effective for Ni-Hard 4 alloys and, to some extent, for Ni-Hard 1 and 2 alloys.

The heat-treated iron shown in Fig. 5.44 has a microstructure that is significantly different than that of the as-cast iron shown in Fig. 5.43. Quantitatively, Fig. 5.45 provides two pie charts that directly compare the as-cast and heat treated microstructures. The biggest difference lies in the transformation of austenite to martensite and the increase in M<sub>3</sub>C secondary carbides from 3 to 10%. When combined, these changes result in the as-cast hardness increasing from around 450 HB to over 700 HB in the heat-treated condition.

This increase in hardness presents an interesting interplay between the microconstituents of the iron during heat treatment. Many researchers report hardness as a function of retained austenite or martensite. It is actually more complex than just a simple ratio between austenite and martensite. As shown in Fig. 5.43, when austenite is transformed into martensite, we are actually creating a nanocomposite material of nanometer sized M<sub>3</sub>C carbides that are embedded in a matrix that is dominantly martensitic. The presence of these tiny

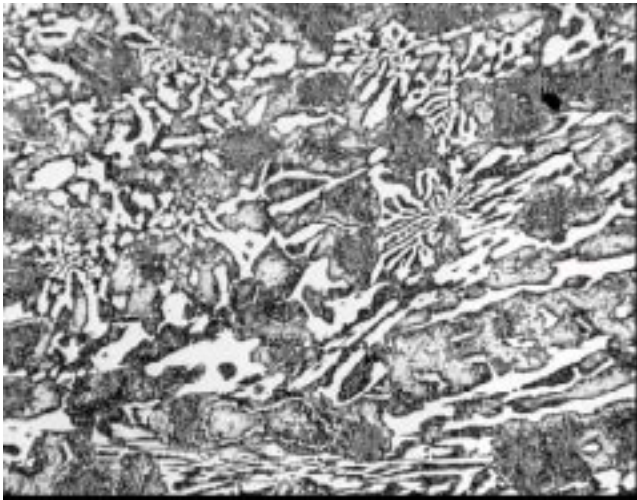


(5.43a) 100X

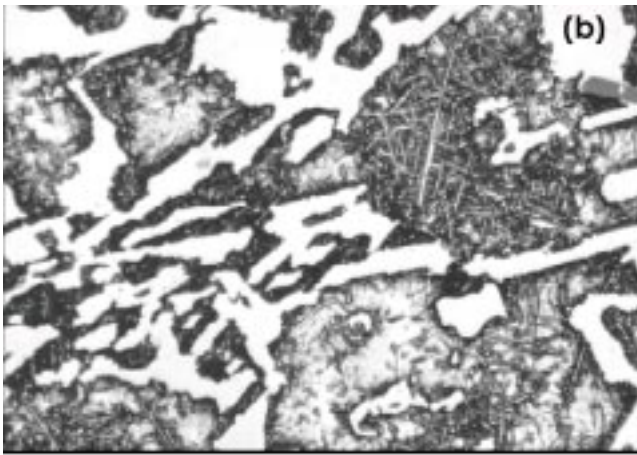


(5.43b) 1000X

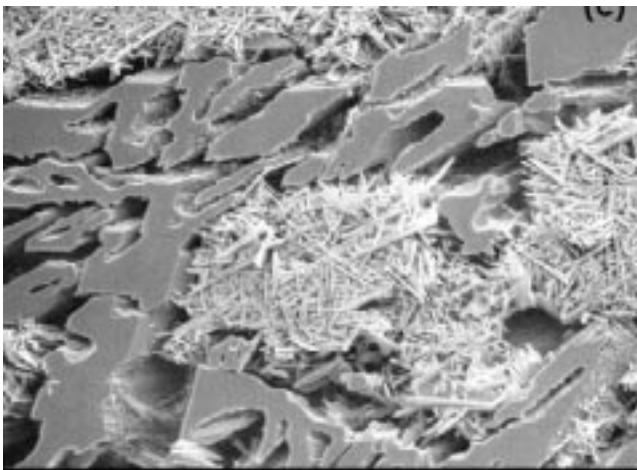
**Fig. 5.43. Microstructure of a Ni-Hard 4 in the as-cast condition. The iron's composition is 2.9% C, 1.8% Si, 5.8% Ni, 1% Mn and 8.4% Cr. The iron was cast as a 75-mm ball in a sand mold and had an as-cast hardness of 450 HB. Typical hardnesses for Ni-Hards are between 550 and 700 HB. The alloy's lower-than-expected hardness is due to its higher-than-normal levels of Ni and Mn (which increases the alloy's retained austenite content). Etched in 2% nital.**



(5.44a) 500X, 2% nital etch



(5.44b) 1000X, 2% nital etch



(5.44c) 1200X, deep-etched in aqua regia and 2% HF

Fig. 5.44. Microstructure of a Ni-Hard 4 composition, similar to that shown in Fig. 5.43 after a destabilization heat treatment at 750C (1380F). The iron has a tremendous quantity of  $M_3C$  secondary carbides after heat treatment. Essentially, we have a microcomposite structure (eutectic carbides and metallic matrix) with the matrix further subdivided into a nanocomposite structure of secondary carbides and matrix.

carbides within the ferrous matrix significantly affects the toughness of the iron. As a rule of thumb, heat-treated Ni-Hard 4 will always be less tough than as-cast Ni-Hard 4 (say about 10% less). Exact toughness numbers for as-cast and heat-treated Ni-Hard 4 can be found in Appendix B.

As in Ni-Hard 1 and 2, the CVF can be increased or decreased by adjusting the C content of the alloy. Figure 5.46 provides an idea of the range of hardness that can be expected in Ni-Hard 4, based on the alloy's C content.

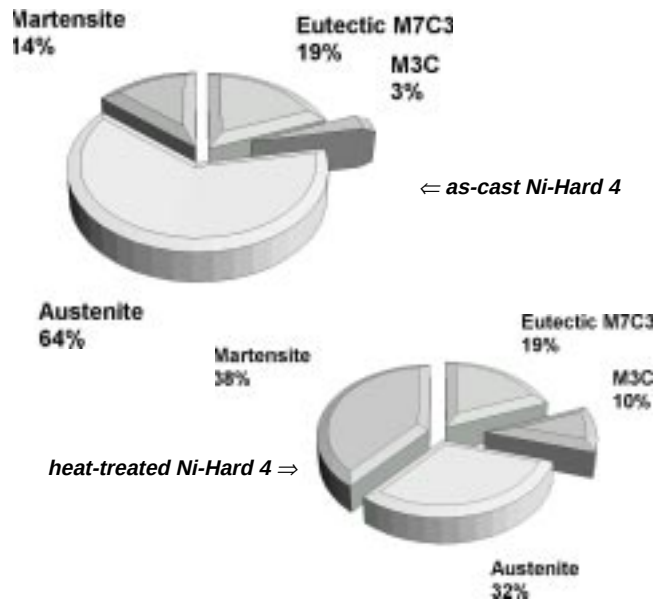


Fig. 5.45. The proportion of microstructural components are dramatically changed by heat treatment. After heat treatment, the volume fraction of martensite increases from 14 to 38% and the amount of  $M_3C$  secondary carbide also increases from around 3 to 10%. These microstructural changes are reflected in an increase in the casting's hardness from 450 to 700 HB.

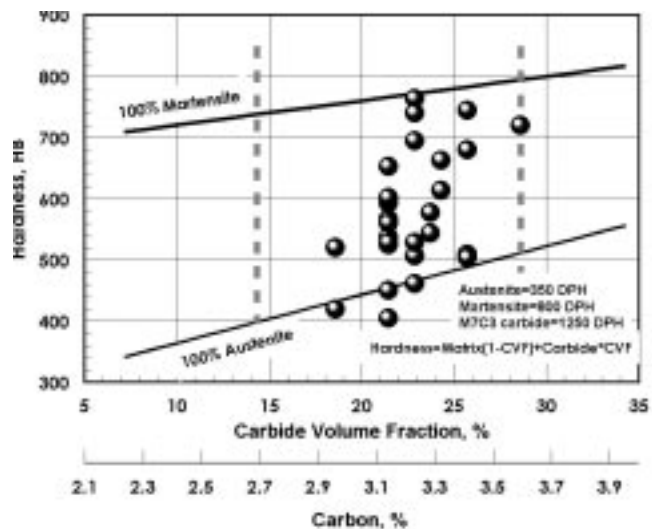


Fig. 5.46. Hardness relationship between CVF and the matrix microstructure in Ni-Hard 4. The circles are hardness data from Ni-Hard 4 castings that were subjected to a variety of heat treatments. The boundary lines were constructed with a rule-of-mixtures formula using constituent hardness values estimated from data given in Tables 5.1 and 5.3. The dashed vertical lines show the ASTM Type ID (Ni-Hard 4) carbon limits.

As the CVF increases, hypoeutectic Ni-Hard 4 irons take on the appearance of hypereutectic irons. Figures 5.47a and 5.47b show a high-C Ni-Hard 4 (3.5% C, 2.1% Si, 6.5% Ni and 8.5% Cr) in the as-cast condition (sand cast as a 75-mm diameter ball). The iron is not typical of commercial high-C Ni-Hard 4 with its higher than normal Si and Ni contents. The result of this extra alloying is the graphite flakes shown in the lower right hand side of Fig. 5.47a.

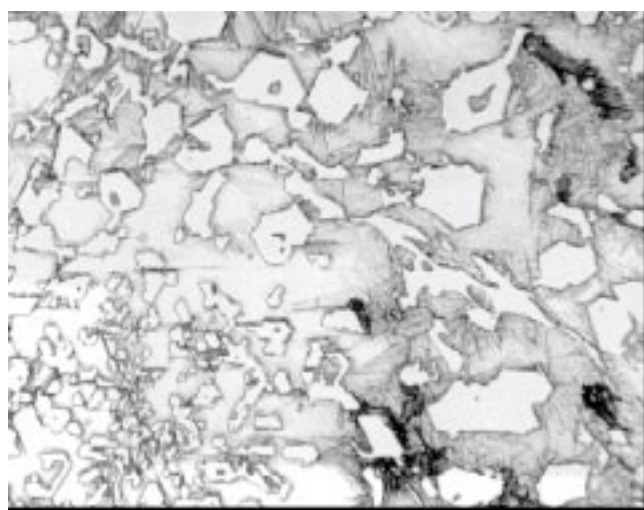
The presence of primary  $M_7C_3$  carbides in a hypoeutectic Ni-Hard 4 is not what one would expect (alloying of 3.6% C and 8.5% Cr places the iron well on top of the austenite liquidus surface shown in Fig. 2.8). During eutectic solidification, the carbides will completely reject Si and Ni into the remaining liquid. This remaining liquid then becomes supercharged with Si and Ni, raising the activity coefficient of carbon. This effect increases the liquid's "thermodynamic C

content" and pushes it across the eutectic boundary and onto the hypereutectic liquidus surface (see Fig. 2.17b). Consequently, primary  $M_7C_3$  carbides will then form at the very end of the solidification reaction.

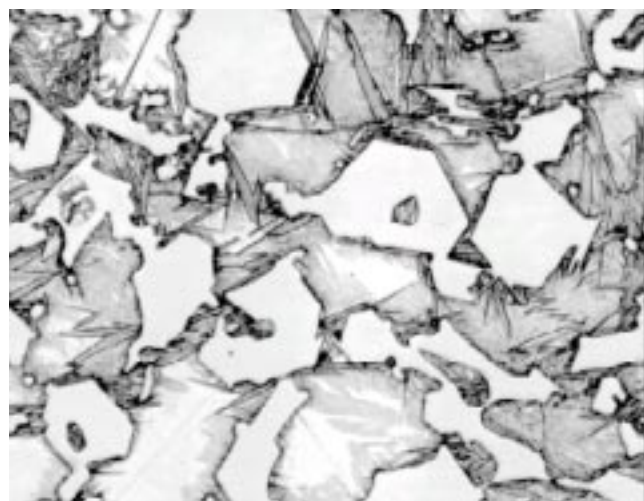
Another interesting solidification reaction in Ni-Hard 4 irons can create duplex carbides containing a core of  $M_7C_3$  and a shell of  $M_3C$ . As covered in Chapter 2 on Solidification, if the Si level is not kept above 1.5%, the eutectic reaction of  $L \rightarrow \gamma Fe + M_7C_3$  will drift down across the peritectic point, forming shells of  $M_3C$  carbides via the peritectic reaction of  $L + M_7C_3 \rightarrow \gamma Fe + M_3C$ . The microstructure formed in a low-Si Ni-Hard 4 is shown in Fig. 5.48. The Vickers hardness indents provide a good idea of the relative hardnesses of the austenite,  $M_3C$  and  $M_7C_3$  phases, from right to left. Needless to say, the presence of the softer  $M_3C$  carbide is not a good thing when what you really want is the harder  $M_7C_3$  carbide.

Chill casting Ni-Hard 4 seems to only moderately increase the surface hardness of the iron. Fischer (1983) chill cast a Ni-Hard 4 using an alloy with the following composition: 3.3% C, 2.1% Si, 5.2% Ni and 7.8% Cr. In his work, the hardness near the chill was around 660 HB, while 50 mm back from the surface the hardness decreased to 610 HB. This decrease of 50 HB appears to be reasonable, based on other work for Ni-Hard 1 alloys.

Defining the toughness of Ni-Hard 4 is problematic. In the as-cast state (cooled fairly quickly from the eutectic temperature), the iron will be dominantly austenitic. The austenitic matrix will be fairly homogenous, containing only a few secondary carbides. Once the iron is heat treated or allowed to transform, partly or totally, to martensite, the matrix will now be a nanocomposite material, containing brittle martensite and even more brittle secondary carbides. Even after tempering, the martensitic matrix will still contain significant amounts of secondary carbides. Once Ni-Hard 4 irons are heat treated, the martensitic matrix will contain about 10%  $M_3C$  secondary carbides. These carbides significantly degrade the toughness of the martensite, no matter how well the iron has been tempered or drawn. Broadly speaking, austenitic irons will



(5.47a) 500X



(5.47b) 1000X

**Fig. 5.47. A hypoeutectic high-C Ni-Hard 4 (3.5% C) has a very unique microstructure containing proeutectic austenite dendrites,  $M_7C_3$  eutectic carbides and, in scattered locations, primary  $M_7C_3$  carbides. If the C, Ni and Si contents are not carefully balanced, the iron will also contain graphite. The sand cast hardness is 470 HB; etched in 2% nital**



**Fig. 5.48. A hypoeutectic low-Si Ni-Hard 4 (~1% Si). Vickers hardness indents can be seen. Etched in 2% nital followed by Murakami's etch, 250X. (Courtesy of Department of Energy)**

always be tougher and have higher fracture toughness values. However, since martensite has a higher yield strength, martensitic irons will have higher tensile strengths.

Table 5.5 provides a sampling of the mechanical properties for Ni-Hard 4. Complete data will be given in Appendix B. In the meantime, the only discernible trend is that austenitic structures provide better fracture toughness, while martensitic structures provide less toughness. It is hard to make any broad conclusions about mechanical properties and their relationship to microstructure, based on the given data. Considering that the Ni-Hard 4 alloy shares the same type of eutectic carbide morphology as high-Cr irons, much of the discussion on relating structure to properties is directly relevant to the Ni-Hard 4 alloys.

### High-Cr AR Cast Irons

Of all the irons covered to this point, high-Cr irons are the most studied. Why this is so is not clear. On a tonnage basis, Ni-Hard 1 leads the pack and, on a historical basis, chill cast AR irons have been around for centuries before the advent of the high-Cr irons. The reason for high-Cr iron's premier position among producers and researchers lies in the wide range of mechanical properties that can be obtained by alloying and heat treatment manipulation (i.e., microstructural manipulation). For many users, high-Cr's higher toughness and excellent abrasion and repetitive-impact resistance are reason enough to choose high-Cr irons. However, obtaining the desired microstructure for such properties is not trivial and perhaps explains why more research has been devoted to this particular family of AR cast irons.

**Table 5.5**  
**Microstructural / Mechanical Property Data for Ni-Hard 4 AR Cast Irons**

| Alloy <sup>(1)</sup> | Chemistry                                  | Sample Geometry                                        | Thermal History              | Micro-structure      | CVF | Hard-ness HB | Bending Strength                    | Impact Energy | Fracture Toughness (K <sub>IC</sub> )    |
|----------------------|--------------------------------------------|--------------------------------------------------------|------------------------------|----------------------|-----|--------------|-------------------------------------|---------------|------------------------------------------|
| A1                   | 3.3% C,<br>2.1% Si,<br>5.2% Ni,<br>7.8% Cr | Surface of chill cast sample.                          | As-cast                      | Aus. with some Mart. | 24  | 660          | 760 Mpa <sup>(2)</sup><br>(110 ksi) | n.a.          | n.a.                                     |
| A2                   | 3.3% C,<br>2.1% Si,<br>5.2% Ni,<br>7.8% Cr | 50-mm from chill cast surface.                         | As-cast                      | Aus. with some Mart. | 24  | 610          | 600 Mpa <sup>(2)</sup><br>(88 ksi)  | n.a.          | n.a.                                     |
| B1                   | 3.4% C,<br>1.8% Si,<br>5.5% Ni,<br>9.6% Cr | Sand cast as 5-mm thick plates. (chill cast structure) | As-cast                      | n.a.                 | 26  | 510          | n.a.                                | n.a.          | 31 <sup>(4)</sup><br>MPa√m<br>28 ksi√in. |
| B2                   | 3.4% C,<br>1.8% Si,<br>5.5% Ni,<br>9.6% Cr | Sand cast as 5-mm thick plates. (chill cast structure) | 800C (1470F) and air cooled  | n.a.                 | 26  | 700          | n.a.                                | n.a.          | 27 <sup>(4)</sup><br>MPa√m<br>25 ksi√in. |
| B3                   | 3.4% C,<br>1.8% Si,<br>5.5% Ni,<br>9.6% Cr | Sand cast as 5-mm thick plates. (chill cast structure) | 800C (1470F) and 275C (525F) | n.a.                 | 26  | 600          | n.a.                                | n.a.          | 26 <sup>(4)</sup><br>MPa√m<br>24 ksi√in. |
| C1                   | 3.4% C,<br>1.6% Si,<br>6.2% Ni,<br>8.6% Cr | n.a.                                                   | 210C (410F)                  | Aus. & Mart.         | 26  | 510          | n.a.                                | n.a.          | 25 <sup>(3)</sup><br>MPa√m<br>23 ksi√in  |

*Microstructure: Mart. = martensite, Aus. = austenite*

(1) A = Data from Fisher (1983); B = Sare (1979); C = Diesburg and Borik (1974).

(2) Bending strength tests were performed on flat bars 127 mm long, 22 mm wide and 10 mm thick. The support span was 82.5 mm. All surfaces were polished.

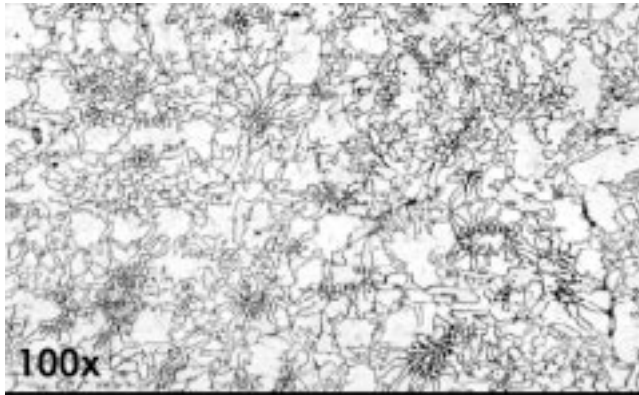
(3) ASTM compact tension specimen with fatigue precrack; a high-quality fracture toughness.

(4) Double-torsion bend test; no precrack. As a comparison, low carbon Ni-Cr-Mo steels gave values around 80 Mpa√m (70 ksi√in.).

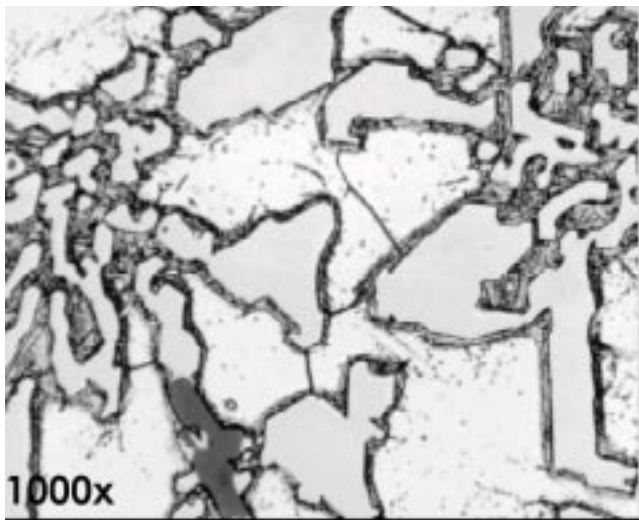
n.a. = not available.

**As-Cast Microstructure**—Figures 5.49a and 5.49b show examples of an as-cast high-Cr iron (3% C, 20% Cr, ~2% Mo and ~1% Cu, a composition also known as Paraboly, developed by Abex Corporation). The as-cast microstructure is a mixture of  $M_7C_3$  eutectic carbides, austenitic and a bit of martensite. The iron's usage of both Mo and Cu provides high hardenability (the pearlite nose is shifted far to the right). In the as-cast state, its matrix is supersaturated in C and Cr to such an extent that the martensite start temperature is depressed to below room temperature for the bulk of the austenitic matrix. Although very thick-section castings may partially transform to pearlite, bainite and/or martensite, the typical as-cast matrix microstructure for a 3% C, 20% Cr high-Cr iron is austenitic.

The majority of high-Cr irons are hypoeutectic and therefore, fall onto the austenite liquidus surface shown in Fig. 2.8. Consequently, they will have a microstructure that is primary composed of austenite dendrites and have CVFs less than 35%. An example of a very hypoeutectic iron is shown in Figs. 5.50a–5.50c. The  $M_7C_3$  eutectic carbides are squeezed be-

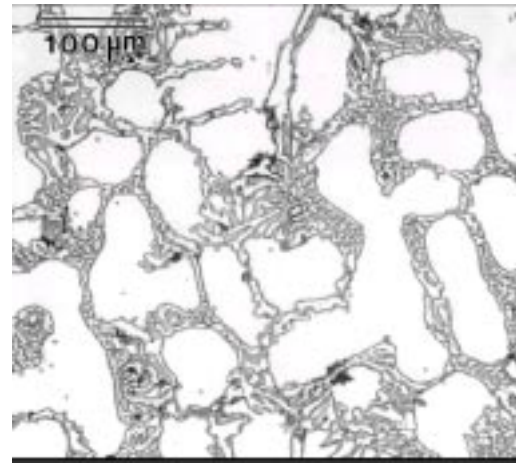


(5.49a) 100X

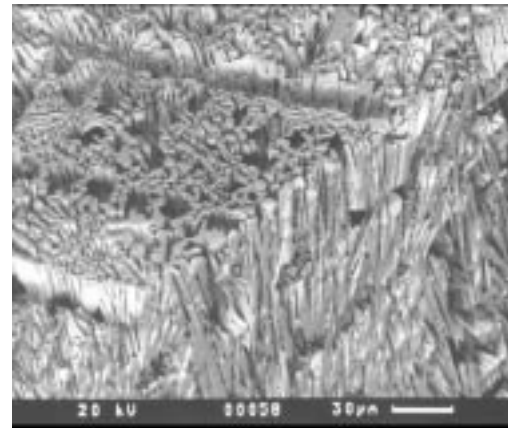


(5.49b) 1000X

**Fig. 5.49.** Hypoeutectic as-cast high-Cr iron (3% C, 20% Cr, 1.9% Mo, 0.7% Mn, 0.5% Si, 0.4% Ni and 0.8% Cu). Sample was cut from a 75-mm dia sandcast ball having a hardness of 480 HB. In (b), a thin layer of martensite can be seen around the eutectic carbides. The fine “dots” shown within the dendrites and along their boundaries are  $Mo_2C$  carbides. The dark shape near the bottom of (b) is an MnS inclusion.



(5.50a) austenite matrix



(5.50b)  $M_7C_3$  eutectic carbide network



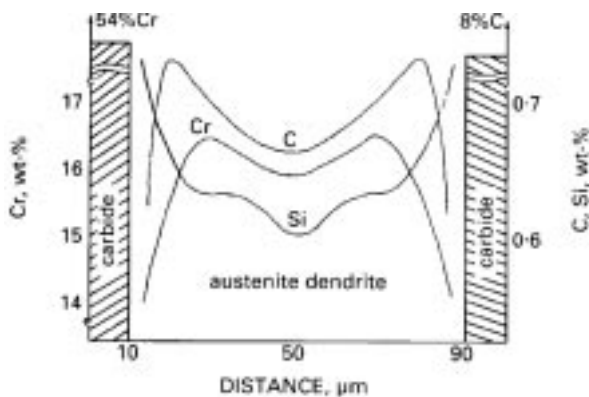
(5.50c) 1000X of (5.50a)

**Fig. 5.50.** As-cast hypoeutectic high-Cr iron (2.2% C, 18.4% Cr) showing (a) its austenitic matrix and (b) its  $M_7C_3$  eutectic carbide network. In (c) at higher magnification, a thin sheath of martensite can be seen around the eutectic carbides. The darker spots are traces of pearlite. ((b) courtesy of Univ. South Australia)

tween the austenite dendrites. If we dissolve the austenite matrix, we can view the carbide backbone of a typical hypoeutectic alloy. This network of hard carbides imbues the iron with exceptional abrasion resistance and contributes to its brittle nature under impact or tensile loading (e.g., rapid cooling or heating during heat treatment). This is the nature of all carbides in all irons.

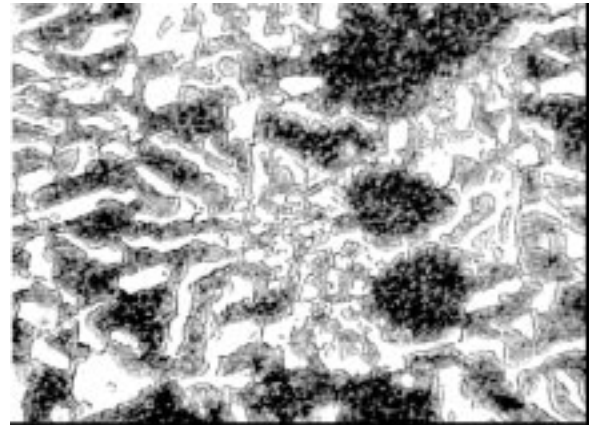
An interesting microstructural feature of as-cast high-Cr irons is that a thin sheath of martensite will often form around the eutectic carbides. This thin layer is shown in Figs. 5.49b and in 5.50c. As the iron cools, the eutectic carbides will coarsen depleting the nearby austenite of C and Cr. This behavior is shown schematically in Fig. 5.51 whereby C and Cr are depleted near the carbide while Si increases since it is completely rejected by the carbide during solidification and solid-state coarsening. Within this thin destabilized region, the martensite start temperature has been raised above room temperature and a thin sheath of martensite will form between and around the  $M_7C_3$  eutectic carbides. If the Si content is a little too high or the Cr content is a little too low, then pearlite can form (see Fig. 5.50c). This coarsening and destabilization process explains why high-Cr irons are never 100% austenitic upon reaching room temperature.

**Heat-Treated Microstructure**—High-Cr irons are often used in applications where high hardness and good repetitive-impact resistance are crucial. To obtain this mix of mechanical properties, the iron is often given a destabilization heat treatment (see Chapter 3, Heat Treatment). The microstructural result of this treatment is shown in Fig. 5.52. As in heat-treated Ni-Hard 4, secondary carbides are formed during the destabilization treatment. Depending on the chemistry of the iron,  $M_7C_3$  or  $M_{23}C_6$  carbides are formed. Since both types of carbides are rod-shaped and brittle, it is somewhat academic as to which carbide is present. The important concept is that the resulting microstructure is viewed as a nanocomposite structure of martensite and secondary carbide. And, considering the nature of the martensitic transformation in high-Cr

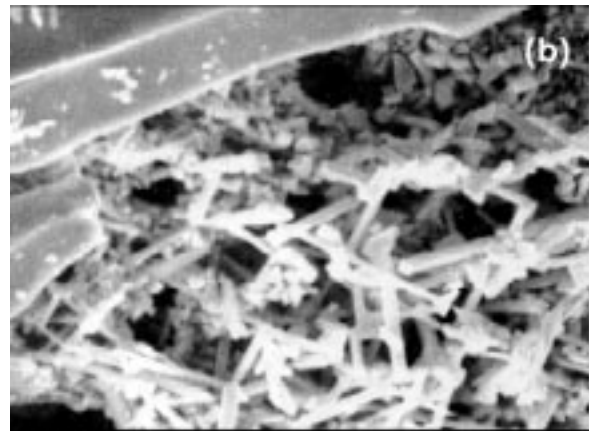


**Fig. 5.51.** As-cast high-Cr chemistry profile in a large austenite dendrite. Near the eutectic carbides, the C and Cr atoms have diffused toward the eutectic carbides, resulting in a slight coarsening during the slow cool from the eutectic temperature. (After Dupin et al., 1982)

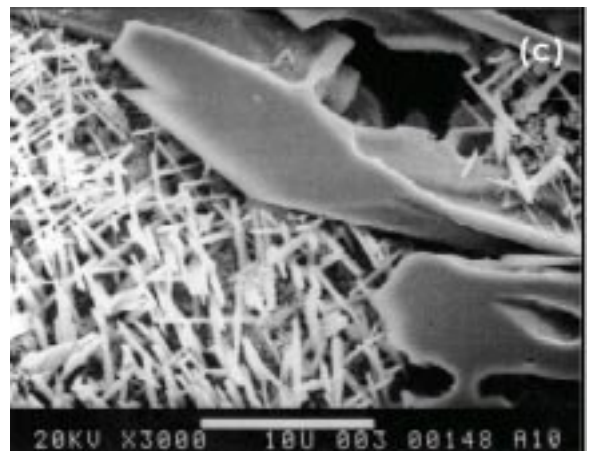
irons, the austenite phase will never fully transform to martensite. As a result, the heat-treated matrix microstructure is always a mixture of martensite, retained austenite and secondary carbides.



(5.52a) 250X

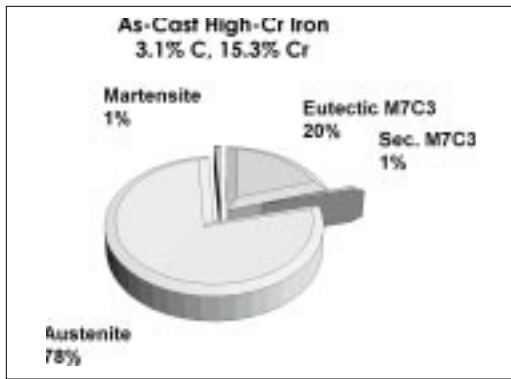


(5.52b) 10,000X

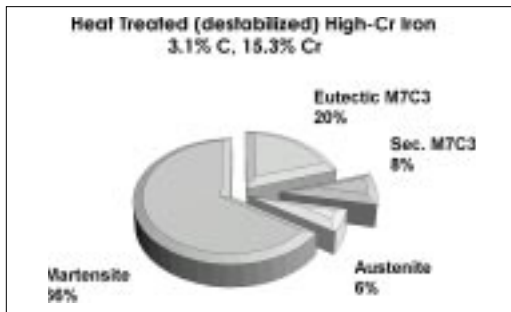


(5.52c) 3000X

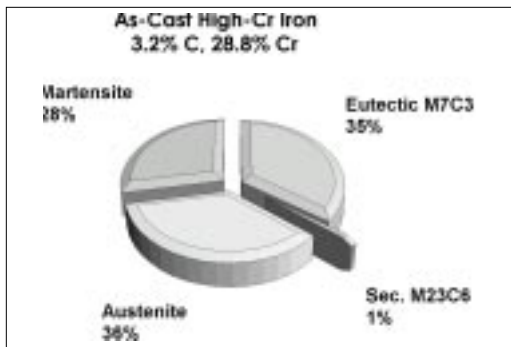
**Fig. 5.52.** Heat-treated, near-eutectic high-Cr iron (3% C, 25% Cr). Low-C lath martensite is formed around the eutectic carbides while higher C lath/plate martensite appears within the prior austenite dendrites (not shown). Depending on the chemistry of the iron, the secondary carbides may be (b)  $M_{23}C_6$  or (c)  $M_7C_3$ . For the iron shown in (a),  $M_{23}C_6$  carbides are formed during the destabilization heat treatment.



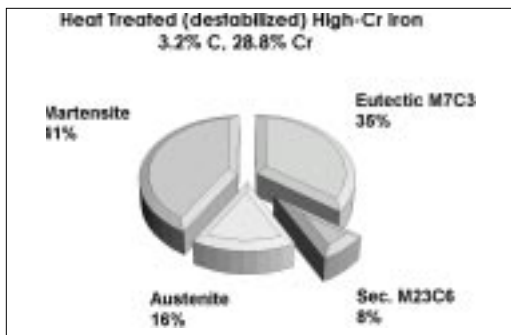
(5.53a) as-cast 15.3% Cr



(5.53b) heat-treated 15.3% Cr



(5.53c) as-cast 28.8 Cr



(5.53d) heat-treated 28.8 Cr

Fig. 5.53. The direct result of the high-temperature heat treatment is the formation of secondary carbides. Only upon cooling does the destabilized austenite transform to martensite. In the 3.1% C, 15.3% Cr iron, the secondary carbides are of the  $M_7C_3$  type, while in the 3.2% C, 28.8% Cr iron, the secondary carbides are  $M_{23}C_6$ . The precipitation of these carbides removes C and, to some degree, Cr from the austenite matrix. Upon cooling to room temperature, the destabilized matrix will mostly transform to martensite. In the higher-alloyed irons, this transformation is not nearly as complete, due to the iron's higher Cr content.

The pie-charts shown in Fig. 5.53 provide a quantitative idea as to how the heat treatment process affects the microstructure in high-Cr AR cast irons. In the 15% Cr iron, the secondary carbide is of the  $M_7C_3$  type, while in the 28.8% Cr iron, the carbide type is shifted to the higher Cr type, known as  $M_{23}C_6$ . Although the matrix is slightly depleted of Cr during the formation of these carbides, the dominant effect is the depletion of carbon. In the as-cast condition, the austenite matrix is supersaturated in carbon. Upon heat treatment, this excess C goes toward the formation of secondary carbides, lowering the C content of the austenite. For example, in the 3.1% C, 15.3% Cr iron, the as-cast C content is around 1.2%. After heat treatment at 1000 C (1832 F), the equilibrium C content in the austenite is around 0.5%. The rejection of 0.7% C results in the formation of about 8%  $M_7C_3$  carbides (assuming a C content of 8.6% for the  $M_7C_3$  carbide). For this type of iron, the 8% increase in carbide is significant, when compared to its eutectic CVF of 20%.

Before leaving the subject of how heat treatment affects the microstructure of a high-Cr AR iron, Fig. 5.54 shows an example of the wide range of hardnesses that are possible in a high-Cr iron (compositional information was not provided in the original work). Tailoring the hardness of the iron through manipulation of the martensite to austenite volume fraction is a powerful tool in the hands of the knowledgeable designer, foundryman or tribologist.

Heat treatments can go awry. Subsequent to destabilization, the iron must be cooled fast enough to miss the pearlite

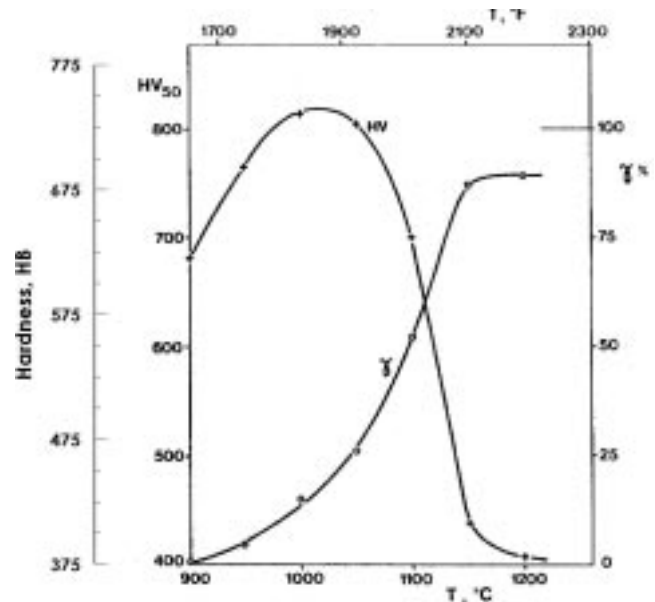
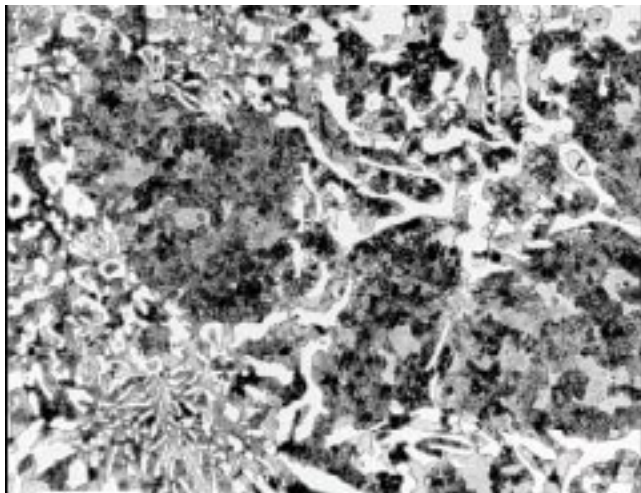


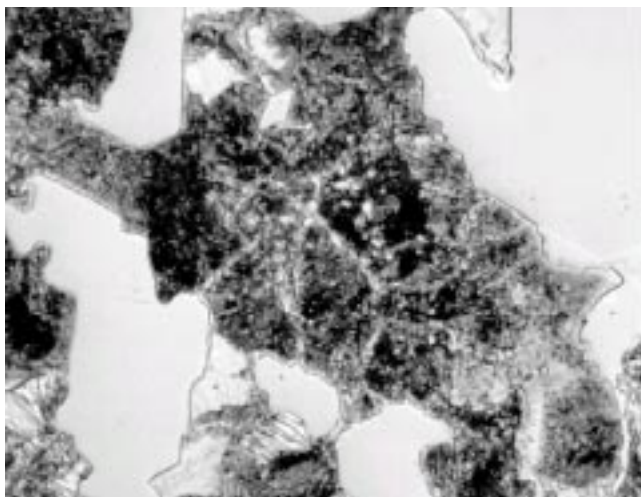
Fig. 5.54. The hardness of a typical high-Cr AR cast iron can be controlled by tailoring the ratio of austenite to martensite. In more direct terms, as the amount of retained austenite decreases, the hardness increases. In high-Cr irons, one way to control hardness is to use different destabilization temperatures. As the destabilization temperature is increased, the solubility of C in austenite similarly increases. Upon cooling, the increased C content lowers the martensite start temperature yielding a higher volume fraction of retained austenite at room temperature. (After Maratray and Poulalion, 1982)

nose. This can be a problem in thick-section castings. Figure 5.55a shows an example of a AR cast iron that has hit the pearlite nose while being cooled from its high-temperature destabilization treatment. The austenitic structure has almost completely transformed to pearlite. In Fig. 5.55b, secondary carbides can be seen within the former austenitic dendrite. The pearlite colonies are the light gray regions within the matrix. Bainite was also found and can be seen near the carbides in the upper right-hand and lower left-hand corners.

**Carbide Volume Fraction**—Figure 5.56 provides a simplified view of the relationships between CVF, hardness and compressive strength. Hardness trends follow that shown for other AR cast irons. For the most part, the hardness of the iron increases with CVF. The highest hardnesses are naturally found for irons having martensitic matrices. An upper bound-



(5.55a) 100X

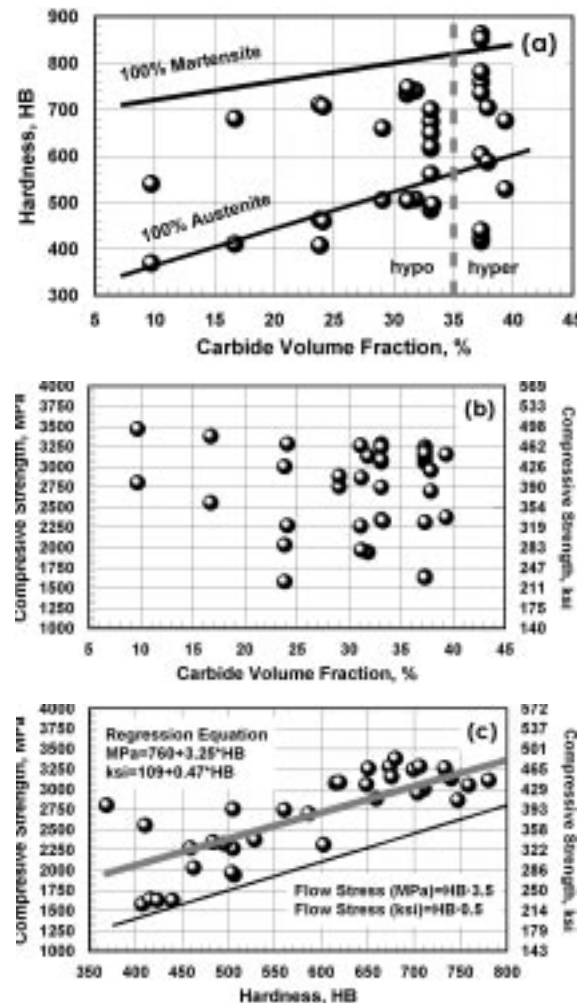


(5.55b) 1000X

**Fig. 5.55.** Destabilized high-Cr iron cooled through the pearlite nose. The composition of the iron is approximately 2.5% C, 18% Cr, 1% Mo, 1% Mn and 1% Si, and had a heat-treated hardness of 480 HB. The sample was obtained from the center of a 150-mm thick casting. Due to its much slower solidification rate, the structure shown in (a) is much coarser than that shown in Fig. 5.49a for the 75-mm dia ball. Etched in Villela's.

ary for hardness is shown by the line marked as 100% martensite. Most irons will be mixtures of austenite and martensite and their hardnesses will fall between the 100% martensite and 100% austenite boundary lines. Several of the irons show hardnesses much lower than that predicted by the 100% austenite line. This result is hard to explain. In some of the irons, the matrix is ferritic and, as a result, does yield softer irons but, overall, the 100% austenite line overpredicts the hardness.

Figure 5.56b plots compressive yield strength against CVF. It is difficult to discern any direct correlation. In Fig. 5.56c, the compressive yield strength was then plotted against hardness. The good correlation between compressive strength and hardness follows the proven relationship between flow stress and hardness as shown in Fig. 5.56c. A better fit is provided by the regression equation. The flow stress equation underpredicts the compressive strength in AR cast irons. This is understandable since the flow stress equation was developed for noncarbide-containing steels.



**Fig. 5.56.** Hardness and microstructures show a direct correlation. In (a), the hardness of the iron linearly increases with CVF. If one relies on just the CVF (b), little correlation is shown to predicting the compressive strength of the iron. Whereas, in (c), if we look at the combined effects of CVF and matrix microstructure in producing a soft or hard iron, a strong correlation is shown between hardness and compressive strength.

**Toughness**—Figure 5.57 provides some insight into the laboratory measured toughness of high-Cr AR cast irons. There exists no shortage of fracture toughness data for high-Cr irons; complete data is given in Appendix B. In the meantime, the data presented in Fig. 5.57a shows that the hardness of the iron is not a good indication of its fracture toughness. The iron presented in Fig. 5.57a has a base composition of 17% Cr, while the C content was varied from 1.2 to 3.6%. The iron was tested in the as-cast (austenitic) and heat-treated (martensitic) condition. Nothing is ever perfect; consequently, the resulting microstructures are neither totally austenitic nor totally martensitic. Nonetheless, the matrix microstructure in the as-cast iron is dominantly austenitic and in the heat-treated iron is dominantly martensitic (i.e., a mixture of martensite and secondary carbides). More details on these irons can be found in Diesburg and Borik (1974).

In general, you might say that as the hardness increases the toughness decreases, but this would only be partly right. For example, in Fig. 5.57a, the martensitic irons have much higher hardnesses and, in some cases, are just as tough as the austenitic irons. At lower C content (lower CVF), the austenitic irons are tougher. This result demonstrates the greater

importance of the matrix structure over that of the carbide structure. As the C content increases (higher CVF), the matrix microstructure becomes less important and the carbide structure starts to dominate.

With careful inspection, a better indicator of the iron's toughness is its CVF shown in parenthesis next to the iron's C content in Fig. 5.57a. Although austenitic irons are tougher than martensitic irons, in general, toughness is extremely sensitive to CVF. This statement is true for all AR cast irons. Naturally, an austenitic matrix is more ductile and tougher than a matrix microstructure consisting of a mixture of fibrous secondary carbides and martensite. The dominance of the eutectic carbide network on toughness is shown clearly in Fig. 5.57b. As the CVF increases the toughness decreases.

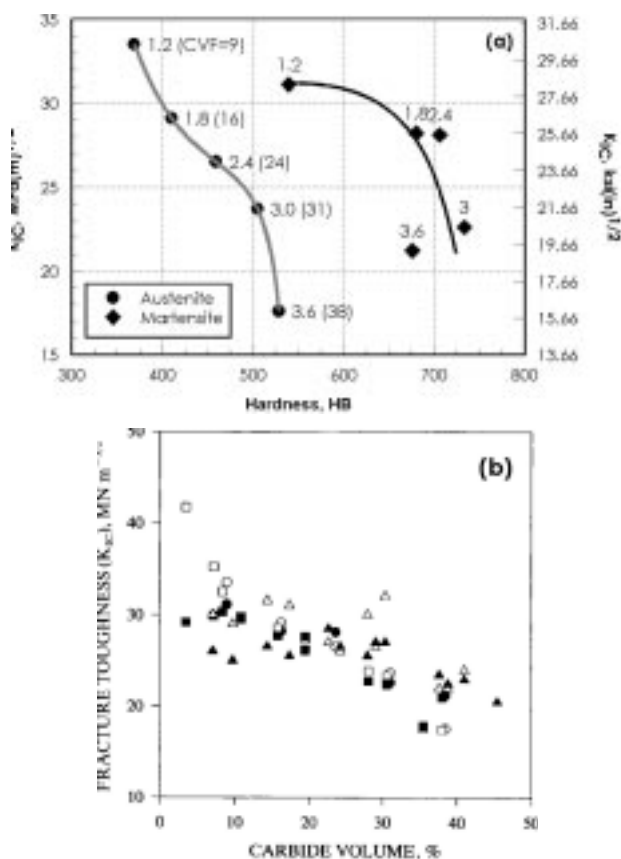
If we think of carbides as crack initiation sites or just as cracks, as we increase the amount of “cracks” in the iron, the toughness will decrease. This fundamental relationship is the bane of all AR cast irons. As the density of the carbide network increases, it is easier and easier for cracks to initiate and propagate. Hence, just knowing the hardness of the iron is not enough. To make an accurate prediction of the iron's toughness you must know the chemistry (i.e., C and Cr contents) and the final hardness.

Before getting too far away from this discussion of an iron's toughness, we should state emphatically that fracture toughness measurements (e.g., Charpy or  $K_{IC}$  or other impact tests) do not strictly translate into component “toughness.” For example, the user wants a casting that does not fail by breakage and will be used in various field conditions such as:

- Toughness to repeated impact loading, as in balls, *ball mills* (see Glossary), chute liners, hoppers, etc.
- Toughness to repeated bending loading, as in many crushers.
- Toughness to occasional shock (over) loading, as in some crushers or when mishandling equipment.
- Toughness to thermal shock loading, in steel mill rolls or by fires in a coal crusher.
- Toughness to slow bending in some crushers or during assembly.
- Toughness to withstand residual stresses from casting or heat treatments.

Consequently, the user must be careful when applying laboratory “toughness” measurements to real-world industrial applications.

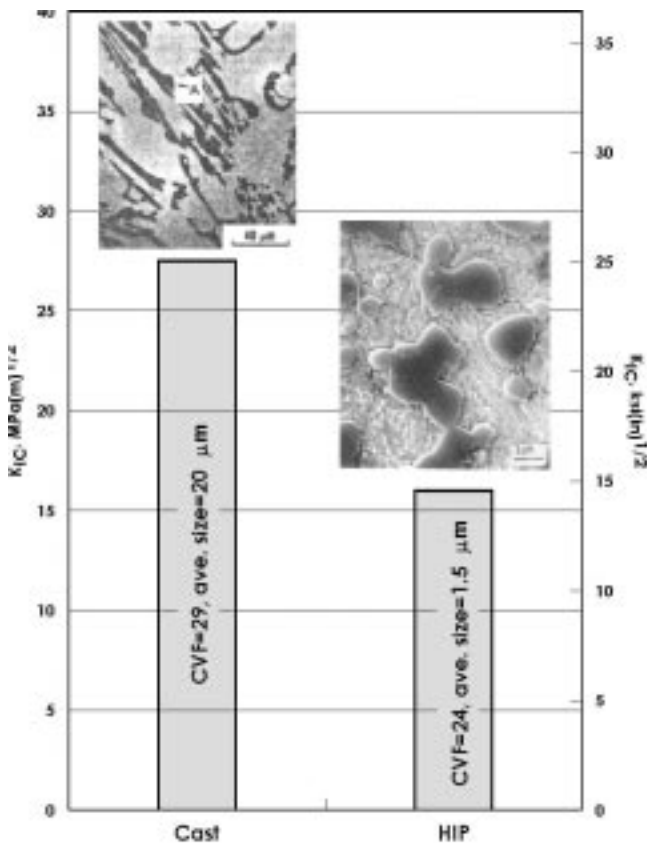
Several researchers have investigated the effects of carbide size and morphology on toughness. It has long been thought that a high-toughness iron could be made if only the carbide size was reduced. Such an experiment was performed by Kumar et al. (1984) on a cast and hot isostatic pressed (HIP) 2.5% C, 20% Cr, 1.5% Mo AR iron. Both irons were heat treated to form matrices consisting of secondary carbides and martensite. Both irons had hardnesses of around 550 HB (indicating that the authors had extensively tempered the martensite). Figure 5.58 summarizes the results of their findings.



**Fig. 5.57.** As shown in (a), matrix structure (austenitic or martensitic) and hardness correlates weakly with fracture toughness ( $K_{IC}$ ). In (a), the alloy's C content and estimated CVF is given to the right of the alloy's data point. As the alloy's CVF increases, its fracture toughness decreases. In (b), this trend of increasing CVF and decreasing fracture toughness is shown quite clearly in the data assembled by Tabrett (1997) from a broad variety of AR cast irons (symbols represent different types of AR irons).

Quite amazingly, the cast material with its coarse carbides is much tougher than the HIP-processed iron (Mohan et al., 1984). If the CVF is held constant, then as the carbide size decreases, the number of carbides increases. As the number of carbides increases, the mean free path or clear unobstructed distance between carbides decreases. In the cast structure, the mean free path is around 20  $\mu\text{m}$ , whereas in the HIP iron, this path was calculated to be approximately 2.2  $\mu\text{m}$ . From a simplified fracture mechanics viewpoint, it is just easier for a crack to jump from carbide particle to carbide particle in the HIP iron versus the much larger distances found in the coarse cast iron microstructure.

**Carbide Morphology**—Carbide morphology can also play a role in determining the hardness of the iron. As was shown for the low-alloy and Ni-Hard irons, the coarseness of the microstructure will increase or decrease the hardness of the iron. For cast irons, the coarseness of the microstructure is controlled by the solidification rate. More to the point, as the solidification rate increases, the average size of a carbide “rod” or “blade” will correspondingly decrease. This effect is most apparent when we compare hardnesses on the surface of the casting (high solidification rate) against hardnesses taken



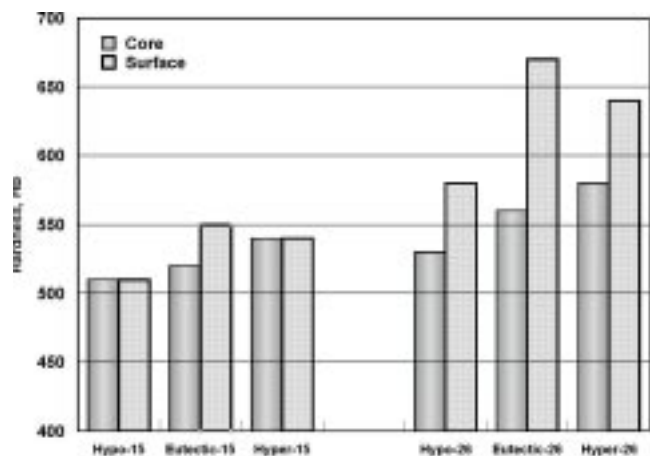
**Fig. 5.58.** The parable of the small spherical carbide for maximum toughness is not true. The best toughness arises from discrete coarse carbides. The coarseness of the carbide structure allows cracks to be blunted by islands of tough matrix. In the HIP structure (right), the mean free path between carbides is quite small, allowing cracks to jump directly from carbide-to-carbide. (Data courtesy of Kumar, 1984)

within the casting (slow solidification rate). Figure 5.59 plots hardnesses taken on the surfaces and cores of 50x50x180-mm sand-cast blocks. A range of carbide structures were studied from hypoeutectic to hypereutectic in irons containing 15% and 26% Cr (Laird and Dogan, 1996). As shown in Fig. 5.59, hardness values for the 26% Cr irons were very sensitive to the solidification rate. The eutectic iron showed the biggest change with a 110-point jump in hardness going from the core to the surface.

Interestingly enough, the lower Cr irons did not show a large response to solidification rate. This is explained by the effect of Cr on the size of carbide formed during solidification. To gain an appreciation of this mechanism, Figs. 5.60a and 5.60b show the eutectic-15 and eutectic-26 irons. It is clear that the higher-Cr iron has smaller carbides. Image analysis confirms this result, giving average carbide particle sizes of 4.3  $\mu\text{m}$  for the eutectic-15 and 2.8  $\mu\text{m}$  for the eutectic-26. It would appear that the larger carbides found in the lower Cr iron are not sufficiently refined by the higher solidification rate near the surface of the casting to affect the hardness measurement.

It should also be mentioned that the coarseness of the high-Cr microstructure (and, for that matter any AR cast iron microstructure) is strongly influenced by the section size of the casting. This can be seen by comparing the two microstructures shown in Fig. 5.49a and 5.55a. Both images were taken at 100X, yet the coarseness of the microstructures is significantly different. In brief, higher hardness will typically result in alloys cast in thin sections and, higher Cr content can result in higher surface hardnesses. All in all, before one can be assured of the iron's true hardness, one must consider the solidification rate (surface or core hardness) and alloy chemistry.

Improvements in yield strength have been shown by breaking up the continuity of the carbide matrix in low-alloy irons ( $\text{M}_3\text{C}$  carbides) by hot working (Hong et al., 1993). The

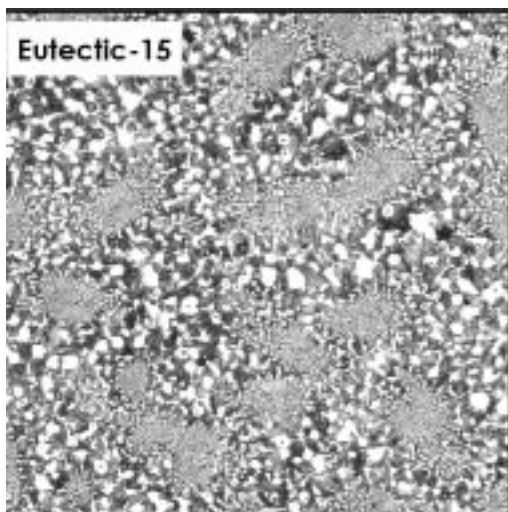


**Fig. 5.59.** Depending on the alloy type and the location of the hardness measurement, hardness values can vary significantly. This behavior is shown most dramatically in the 26% Cr irons, where the hardness difference between the surface and core can be as much as 110 points HB.

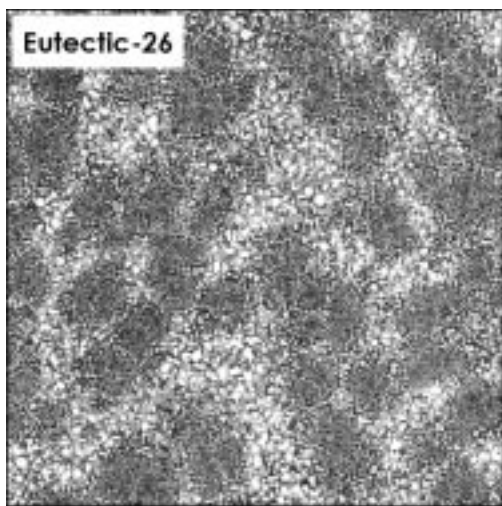
feasibility of hot working large AR castings is very low and this avenue of research has not been pursued. Moreover, this work was inconclusive and indicated that, even with extensive hot working, large  $M_3C$  carbides still remained intact. Nonetheless, we would suggest that for any given level of CVF, the toughest iron would consist of discrete carbides that are tens of microns in size (e.g., 10, 20, 30  $\mu m$  in size).

Although increased CVF decreases the toughness of the iron, it does increase the abrasion resistance. Figure 5.61a shows this relationship between CVF and dry abrasion resistance using a pin abrasive test. The importance of abrasive wear to the industrial community using wear resistance irons justifies an entire chapter devoted to this subject matter. However, only the briefest of introductions is presented to illuminate the relationship between microstructure and abrasion resistance.

In general, dry abrasive wear can be shown to decrease with an increase of CVF. However, even this simple statement



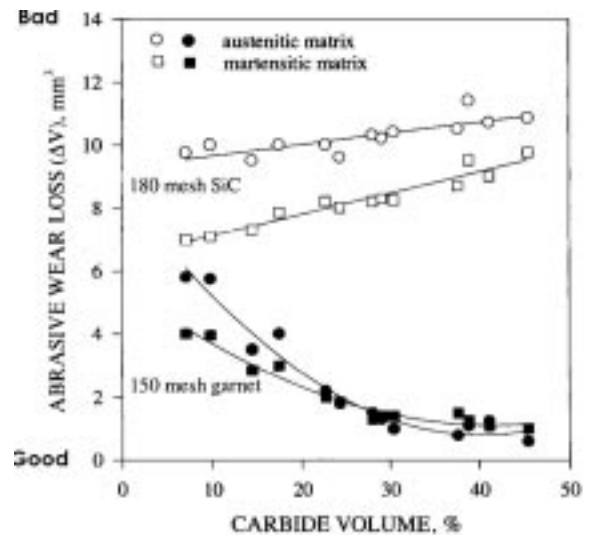
(5.60a) eutectic-15



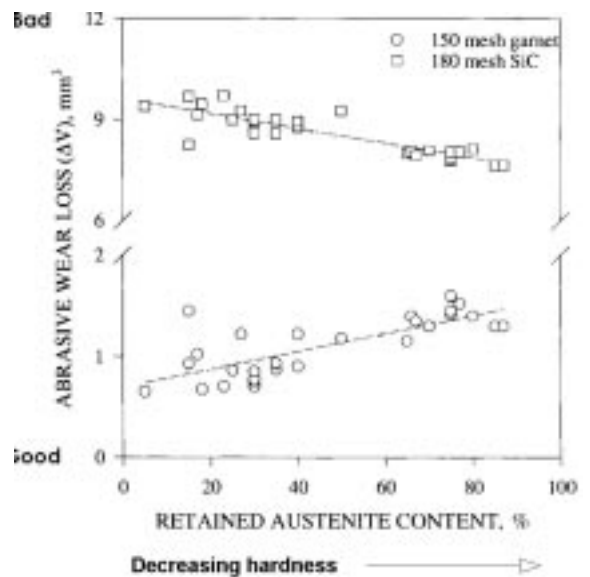
(5.60b) eutectic-26

Fig. 5.60. As the Cr content is increased, the carbide structure is refined. Both images were taken at 80X.

is not always true. For example, in Fig. 5.61a, different abrasives produce different wear results. In this case, the harder and more fracture-resistant SiC abrasive can gouge through the high-Cr microstructure, whereby the softer and more brittle garnet abrasive does not aggressively attack the same microstructure. Consequently, as the CVF increases, the SiC abrasive is able to rip larger and larger chunks of carbide from the microstructure (increasing the wear rate), while the garnet abrasive is “blunted” by the increasing CVF and the wear rate decreases.



(5.61a)



(5.61b)

Fig. 5.61. Making broad statements about the relationships between microstructure and abrasion resistance is tricky. Abrasive environments vary widely and, depending on the abrasive type and loading, the same alloy can perform well under one set of conditions and poorly under another set. In the most general terms, (a) high CVF and high hardness favor sliding abrasive wear (e.g., 150-mesh garnet). Tougher iron with a lower CVF and a more ductile matrix (b) favors very aggressive abrasive environments where the abrasive gouges the material (e.g., 150 mesh SiC). (Figures created by Tabrett, 1996, using data from Zum Gahr and Eldis, 1980)

From a matrix viewpoint, Fig. 5.61b shows that just the opposite is true. The expected result is shown for the garnet abrasive, namely, a decrease in wear resistance as the matrix becomes softer (increasing retained austenite). Conversely, we find the unexpected result for the SiC abrasive. In the latter case, the harder SiC abrasive plows through the microstructure causing extensive work hardening. Moreover, the more ductile matrix is able to hold or retain the carbides more effectively as the SiC particle rips through the microstructure.

From a microstructural viewpoint, the hardest iron will typically provide the best abrasion resistance. As shown in Fig. 5.56a, such an iron is obtained by maximizing the CVF and having a 100% martensitic matrix. The downside to obtaining such a structure is that the fracture toughness will be minimal (Fig. 5.57). The challenge facing the AR iron producers is balancing abrasion demands with toughness demands.

### Specialty Eutectic Carbide Category

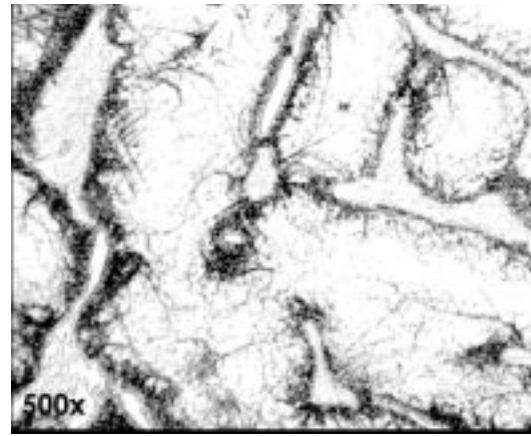
If we take a broad view of our definition of AR cast irons, it is any type of ferrous material that forms carbides via a eutectic reaction of Liquid  $\rightarrow$  Carbide + Dendrite (austenite or ferrite). Under this definition we have alloys such as 440C (1.1% C, 17% Cr) and D2, and more exotic alloys containing high levels of vanadium or niobium. Since material property data for these alloys is almost nonexistent, we will concentrate on a few selected microstructures to show their key features. For completeness, it should be mentioned that titanium has also been used in AR irons. Titanium, like Nb, forms its own separate carbides although their morphology is quite different forming very small and needle-like carbides versus niobium carbide's more blocky-like shapes.

#### 440C (1.1% C, 17% Cr) AR Cast Iron

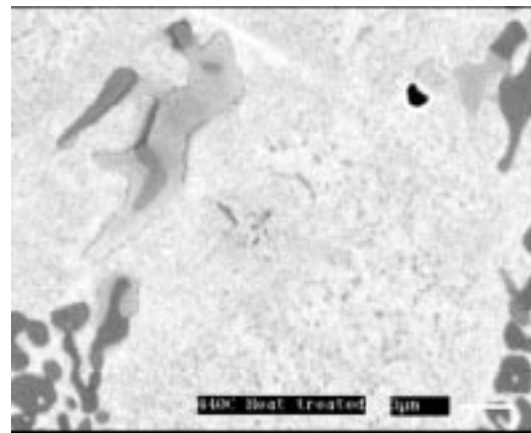
This alloy is often thought of as a corrosion-resistant *steel*. In truth, it is an *iron*. Its alloy content places it to the right of the austenite eutectic solidus line and, as shown in Fig. 2.32, it undergoes a eutectic reaction forming  $M_7C_3$  carbides. The microstructure of this iron is shown in Figs. 5.62a–5.62c. With a CVF of around 10%, the microstructure is dominated by the ferrous matrix.

This iron has a very unique carbide microstructure. Due to the alloy's high ratio of Cr to C, the austenitic matrix contains around 14% Cr after solidification. At this Cr level, the preexisting  $M_7C_3$  carbides will partially transform to  $M_{23}C_6$  carbides along the carbide/matrix interface. These duplex carbides are shown in Fig. 5.62b as cores of dark  $M_7C_3$  surrounded by lighter shells of  $M_{23}C_6$ . The different contrasts shown in Fig. 5.62b, using backscatter SEM, arise from the different densities of the two carbides: the lighter  $M_7C_3$  (~8.9% C) carbide absorbs more electrons than the heavier  $M_{23}C_6$  (5.7% C) carbide.

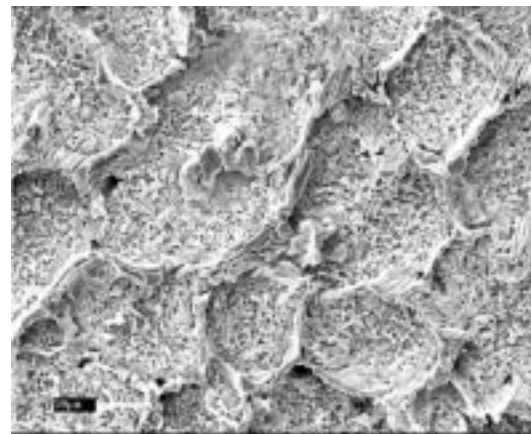
This solid-state transformation was first noted by Pierce (1986) in an iron containing 2.4% C and 31% Cr after being destabilized at 1100C (2012F) and in the as-cast state. In the



(5.62a)

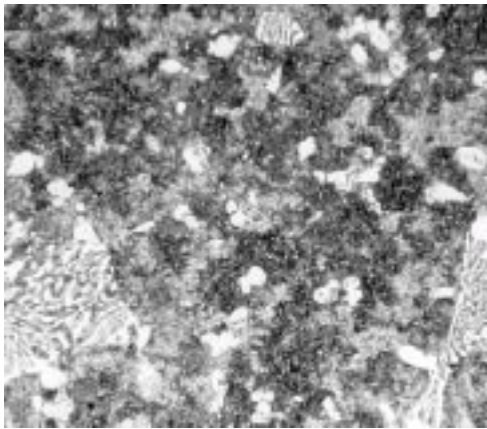


(5.62a)

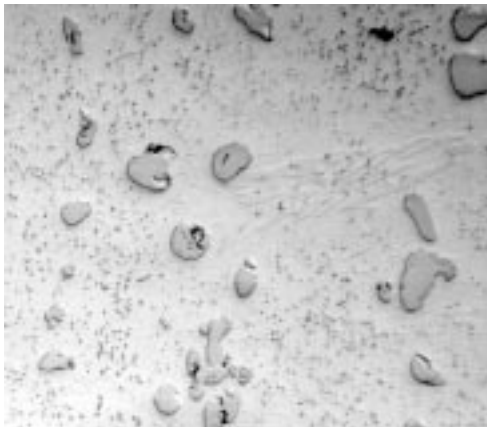


(5.62c)

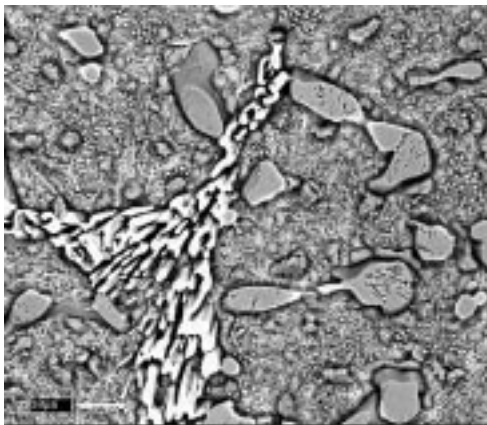
**Fig. 5.62.** Hypoeutectic low-C, high-Cr AR cast iron in the heat-treated condition. The iron's composition is 1.1% C, 17% Cr and was destabilized at 100 C (1832F) yielding a hardness of 550 HB. In (a), the eutectic  $M_7C_3$  carbide network is shown surrounded by small  $M_{23}C_6$  secondary carbides. The dark regions in (a) are martensite, however, the bulk of the matrix is austenite. If the eutectic  $M_7C_3$  carbides are inspected using SEM, it can be observed in (b) that the eutectic carbides have duplexed, forming shells of  $M_{23}C_6$ , subsequent to the solidification reaction. These duplex carbides are identified in (b) by their dark cores ( $M_7C_3$ ) and lighter shells ( $M_{23}C_6$ ). In (c), the matrix is deep-etched to reveal the  $M_7C_3$  carbide morphology (SEM image using secondary electron mode). ((b) and (c) courtesy of K. Collins, Department of Energy, Albany Lab.)



(6.63a)



(5.63b)



(6.63c)

**Fig. 5.63.** In V-based AR cast iron, the iron's composition is roughly 2.5% C, 5% V and 5% Cr. In the as-cast condition, the hardness can exceed 650 HB. The microstructure is a mixture of eutectic MC carbides (blocky shapes) and eutectic  $M_7C_3$  carbides (blade-like shapes). In (b), the microstructure is viewed under polarized light to emphasize the carbide structure. The MC carbides are much harder than the  $M_7C_3$  carbides and stand proud above the matrix. The small "dots" are MC secondary carbides that have precipitated within the matrix as the casting cooled, subsequent to solidification. Both images were taken at 500X. Figure (c) is a deep-etched view of the structure under backscattered SEM mode. The  $M_7C_3$  carbides appear lighter than the VC carbides. ((c) courtesy of Department of Energy)

case of the 440C iron, duplex carbides were found in both the as-cast state and after destabilization at 1000C (1832F).

From a mechanical property perspective, the formation of duplex carbides in this alloy is believed to be of little commercial importance. Although the  $M_{23}C_6$  carbide is softer (1000 DPH) versus the harder  $M_7C_3$  (1250–1500 DPH), this difference is fairly minor. Moreover, considering that the iron contains only about 10% CVF, this difference in hardness between the two carbides is of negligible importance.

The mechanical properties of this iron are controlled by the matrix. In the as-cast state, the microstructure is composed of austenite and  $M_7C_3$  /  $M_{23}C_6$  carbides, yielding a hardness of around 350 HB. The fracture toughness is quite good and, to a reasonable approximation, is given in Fig. 5.57b. As we would expect, if the matrix is partially transformed to martensite (as shown in Fig. 5.62a), the toughness decreases while the hardness increases.

This alloy finds many uses where corrosion resistance is critical yet abrasion resistance and toughness are also required. The iron's low CVF allows good toughness while providing some abrasion resistance. The real utility of this iron is its very good corrosion resistance due to its high-Cr content. As mentioned earlier, when the iron cools to room temperature, the carbides will coarsen, depleting the surrounding matrix of C and Cr. In most high-Cr alloys, this depleted region is extremely susceptible to corrosive attack. However, in the current alloy, the matrix's 14% Cr content is high enough to still provide corrosion resistance, even within the depleted regions, which probably have a Cr content of around 11% and, therefore, still meet the minimum requirements for corrosion protection.

#### **Vanadium-Alloyed (2.5% C, 5% Cr, 5% V) AR Cast Iron**

Eutectic vanadium carbides form the backbone of these irons. As detailed in Chapter 2 on Solidification, these alloys solidify, forming eutectic MC (where  $M = V, Fe, Cr$ , etc.) and eutectic  $M_7C_3$  carbides. Figures 5.63a–5.63c shows a typical as-cast microstructure of this iron. The MC carbide morphology is distinctly different than that of the  $M_7C_3$  carbides shown in Fig. 5.63a. On a carbide volume basis, the MC carbides are around 8% while the  $M_7C_3$  are around 4%, providing a total CVF of 12%.

The super hard MC carbides (2500 DPH) are more clearly shown in Fig. 5.63b. When viewed under polarized light, the topography of the microstructure can be easily observed. The much harder MC carbides are shown to rise above the softer  $M_7C_3$  carbides. One can also see small MC secondary carbides that have precipitated within the matrix as the iron cooled subsequent to solidification.

Vanadium has very low solubility in austenite. As the iron cools, any V present within the C-rich austenite dendrite will readily form MC secondary carbides. Under the right alloying and cooling conditions, it is possible to obtain as-cast structures that are dominantly martensitic with hardnesses in excess of 650 HB. That is, during cooling, the V will destabi-

lize the austenite dendrite sufficiently to raise the martensite start temperature above ambient.

Vanadium-based alloys can present many challenges to the foundry. Besides their unique microstructures, their hardenability is very poor, as compared to the Ni-Hard family or a standard 15% Cr, 3% Mo iron. Consequently, when cast in thick sections (say in excess of 25 mm), the base composition must be alloyed with sufficient quantities of Mo, Ni, Mn or Cu to avoid the pearlite nose upon cooling to room temperature.

Very little wear data on these irons has been published in the current literature. At present, V-based irons are finding some usage as roll materials in hot strip rolling mills. Data presented by Goto et al. (1993) indicate that V-alloyed irons provide superior resistance to hot cracking, as compared to a 2.7% C, 17.5% Cr iron. This seems reasonable considering that hot cracking is basically a tensile stress-induced failure, albeit, highly localized, where the hot strip has overheated the roll material.

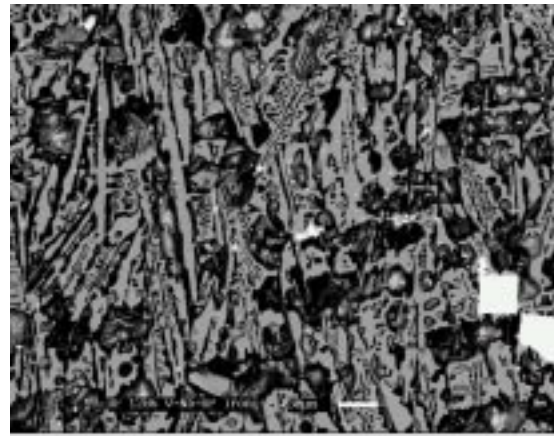
In the high-Cr iron, the CVF is around 25% while in the V iron, the CVF is around 12%. Since toughness is proportional to the CVF, it is no wonder that the V-based iron is much more resistant to hot cracking. Nonetheless, V-based irons are finding favor in many mill roll applications for their combination of good toughness and good abrasion resistance, which is due to their hard MC carbides. This is an emerging field.

#### **Niobium-Alloyed (3% C, 2% Cr, 2% Nb) AR Cast Iron**

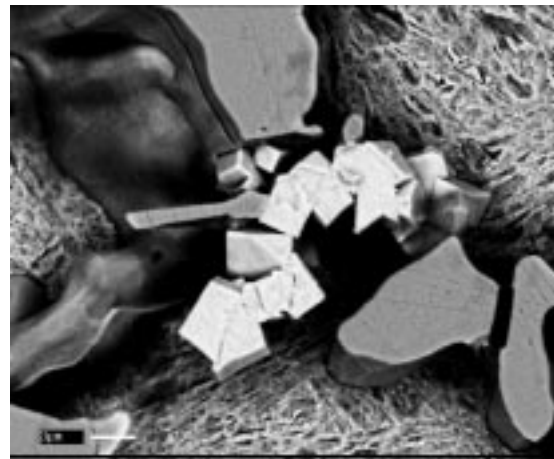
Pure eutectic niobium AR cast irons are very, very rare. In most cases, Nb is added to a base AR cast iron composition to create a few extra hard Nb carbides (2500 DPH). As an example, Figs. 5.64a–5.64c show a low-alloy iron that was alloyed with 2% Nb. The NbC carbides form as discrete carbide particles within the austenite dendrite. Since Nb has no solubility in other types of carbides ( $M_3C$  or  $M_7C_3$ ) or in austenite, it forms as a pure NbC in wear irons.

Several researchers have studied the effect of niobium (actually NbC) on AR cast irons. Work by Guesser (1985) showed that a modest improvement in wear resistance was made possible by the addition of around 1% Nb in a 3% C, 18% Cr iron. Interestingly enough, additional Nb produced no additional improvement. This result suggests that factors other than just the presence of NbC carbide were influencing the wear results. For the most part, the literature is inconclusive on the benefits of adding Nb to AR cast irons.

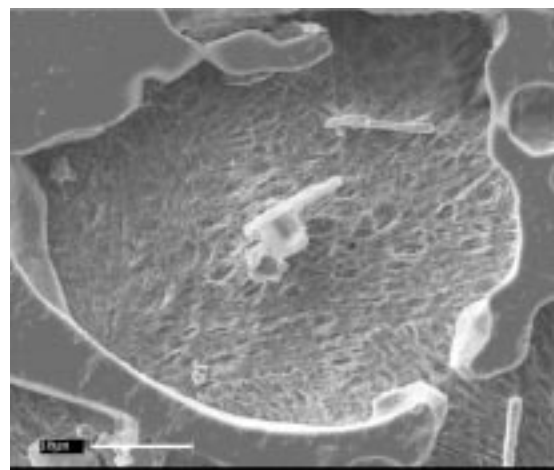
One benefit of adding Nb may be as an inoculant to refine the carbide structure. Since NbC will form before austenite or any eutectic carbide, the presence of these crystallized carbides within the melt can act as nucleation sites for the austenite and for the subsequent eutectic colonies (assuming we are casting a hypoeutectic iron). This result was indirectly mentioned by Chen et al. (1993). Nonetheless, the usage of Nb-alloyed AR cast irons is not expected to revolutionize the wear-resistance world.



(5.64a)



(5.64b)



(5.64c)

**Fig. 5.64.** In Nb-alloyed AR cast iron, the iron's composition is roughly 2.5% C, 2% Nb, 2% Cr and 3% Ni. The image was taken using backscatter SEM. The NbC carbides appear white, due to their much greater density than the  $M_3C$  carbides and surrounding matrix. The sample was etched in 2% nital, prior to viewing. The etch has attacked the martensite plates creating the striations. Figures (b) and (c) show the NbC a little more clearly after deep-etching. Figure (a) was taken in backscatter mode while (c) was taken in secondary electron mode. (Courtesy of Department of Energy)

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# Chapter 6

## Quality Control, Inspection and Specifying

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### Quality Control

Quality requirements for AR cast iron parts are high. The application for most components is arduous with normal operating conditions involving abrasive wear and impact fatigue. Reliability is paramount since premature failure can cause damage to the plant and equipment, and invariably results in expensive, unscheduled downtime. Dimensional accuracy is also vitally important because such parts are frequently supplied as both original equipment and as spare parts to be directly replaced in the field.

Quality control and inspection (testing) of the castings is, therefore, essential. Specifications must be set forward by the customer, and agreed on by both the user and producer, to achieve optimum performance of the component at the lowest feasible cost.

In the production of gray and ductile cast irons, as well as steel castings, chemical composition and mechanical properties are set down in “standards” and “specifications.” These standards and specifications are then routinely determined by the foundry on separate test samples and/or on test coupons attached to the casting. Standardized mechanical properties comprise hardness, tensile strength, yield strength, ductility and impact resistance. Sometimes, additional tests are agreed on, which are relevant for the application of the component. Most things that have gone wrong in the production of the castings show up in these tests.

For AR cast irons, due to their hard and brittle nature, most of these tests are either impractical, impossible or so expensive that they cannot routinely be done in a production foundry. The only routine means for quality control of the material are chemical analysis and hardness testing. For quality control, these tests have to be supplemented by casting inspection and process control.

Process control in AR cast iron production is much more important and has to be much more extensive than in normal gray or ductile iron production. It has to cover all stages, from raw material to shipping. Abrasion-resistant cast irons are much less tolerant to poor production practice and errors than are gray irons or most cast steels. Excellent process control is, therefore, essential for a successful production of high-performance AR cast iron components.

### Inspection

#### Chemical Analysis

The properties of AR iron castings are critically dependent on their chemical composition: carbon, silicon and alloying elements. This is true both for as-cast and heat-treated alloys. The complex interactions between as-cast structure, hardness, properties, heat-treating cycle and heat-treatment response have been described in the preceding chapters.

The chemical composition can be determined by wet chemical analysis (involves remelting) and/or with the various x-ray or spectrographic instruments common in iron and steel foundries, using the usual chill-cast test samples. The concept that the structure is “white” anyhow, and, therefore, samples from ingates, runners or risers, etc. can be used, is not valid. The carbide structure is coarser than in the chill-cast samples and, since the elements are unevenly distributed between carbides and matrix, the results of the spectrographic analysis will be inaccurate. Such samples can only be used to determine or verify the type of alloy. For a more accurate and reliable analysis, the sample will have to be remelted, or a wet analysis has to be performed.

For the calibration of the instrument, the available cast iron standards are often not fully suitable, especially for the high-Cr irons. Also, stainless steel standards may be inaccurate due to the higher C content of the AR irons. In case of doubt, the problem should be discussed with the instrument supplier and suitable standards should be obtained from the usual sources.

In iron foundries, thermal analysis is a common, quick and accurate means to determine the CE-value, the C and the Si contents. The alloying elements in AR irons displace the liquidus line so that the available calibration curves are either not useful or too inaccurate, even for low-alloy irons. In the literature, some regression formulas have been published to calculate the effect of the alloying elements. However, a foundry wishing to use this method will have to establish its own calibration curve and regression formula for a range of the type of alloy it uses.

Chill testing by means of a suitable test casting is a common way to control unalloyed or low-alloyed irons, which can solidify white or gray (ASTM A367). The usual test sample for AR irons is a rectangular vertical plate, molded in

sand with a chill block at the bottom end. The plate will be broken, lengthwise, along a cast-in notch so that the extension of the fully white and the mottled zone can be determined. Interpretation of the results, however, is difficult and takes a lot of experience. For irons that solidify completely white, like the Ni-Hards or high-Cr irons, this testing method is, of course, not possible.

## Material Properties Testing

In normal control of AR iron castings, mechanical testing is restricted to hardness testing. The tensile testing common in the production of gray or ductile cast irons and steel is unfeasible and meaningless for AR cast irons. Preparation of tensile test specimens from the hard and brittle material is difficult and expensive. It involves a high risk of creating small cracks that will cause erratic failure of the test bar. With most tensile machines in foundries, slight misalignments are possible that will also cause premature failure of the test bar. Results, therefore, represent more the *quality* of test conditions than the *properties* of the material. Tensile and yield strength fall together, and elongation is not measurable. In any case, there is no simple and straightforward relationship between tensile properties and structure, quality or service performance of the casting.

Bending tests are easier, but the results from a separate cast test bar have no useful relation to the properties in the actual casting.

Normal impact testing is meaningless. Apart from the problem of preparing suitable test pieces, the impact energy is in the same range as the energy consumed by friction and throwing the fractured pieces of the sample away. Tests with broken samples put together with adhesive tape have given the same results as unbroken samples. Nevertheless, current work by the AFS Special Irons Committee (5-D) using much larger impact bars (50x50x180 mm) have shown some success. Using this larger impact bar it has been possible to differentiate between irons of different heat treatments, CVFs and matrix microstructures. A review of this new impact test is given by Laird (1995).

Fracture toughness can be measured and, in research, a large number of data have been determined (see Appendices B and C for details). The relation to structure, thermal history and other properties of the specific material is, however, rather loose and is a matter of discussion.

The consequence is that mechanical testing, except for that of hardness testing, cannot be used for specification, quality control or acceptance criteria. If data are needed, they can be taken from the Appendix C of this book. If data are absolutely required for a specific type of AR iron, they should be determined by a laboratory specializing in testing hard and brittle materials.

## Hardness Testing

Hardness testing is the only valid method to determine the properties of an AR iron casting. Hardness is specified in all standards and is also used for ordering and as acceptance criteria. The hardness of the casting has to be determined, either as laid down in the standards or as agreed on by customer and foundry.

Contrary to steel, hardness does not correlate very well to tensile strength or toughness. Hardness testing on separate test pieces of different section sizes and heat treatments is usually meaningless, because the as-cast and heat-treated microstructures are extremely different.

### Static Hardness Testing

The ASTM A532 standard gives three different methods to measure “static” hardness:

1. Rockwell (HRC)
2. Vickers (HV)
3. Brinell (HB)

**Rockwell Hardness (HRC)**—Measurement of the Rockwell hardness involves the application of a spherical tipped conical diamond of 120° angle, with a defined force, against the surface of the specimen. The Rockwell hardness is then determined from the difference in the penetration between the initial and the main load. On modern hardness measuring equipment, this can be read, directly from a C-scale dial (B and A scales are used for other materials). The hardness is then listed as, for instance, 65 HRC (Hardness Rockwell C-scale). Standards are ASTM E 18, ISO 6507-1, DIN 50 103.

**Vickers Hardness (HV or DPH)**—Measurement of the Vickers hardness (HV) involves the application of a diamond in the form of a square pyramid with an apical angle of 136°, with an adjustable test force against the surface of the specimen. The use of a diamond pyramid has led to the often found abbreviation DPH for Diamond Pyramid Hardness. When the force is relieved, the diagonal of the produced impression is determined with a measuring microscope. Mathematical relationships lead to the computation of the Vickers hardness, which can be read from tables as a function of the size of the measured diagonals (mean value) and the selected test force. The hardness is then listed as, e.g. 820 HV50. Standards are ASTM E 92, ISO 6507-1, DIN 50 133. For AR cast irons, the most appropriate test forces are HV50 or HV100. Generally speaking, the highest feasible load should be used because the higher the load, the lower the scatter.

**Brinell Hardness (HB)**—Brinell hardness is determined by forcing a hardened steel sphere with a defined load into the specimen surface and measuring the indentation of the sphere that is left behind after the test. Tables are used to read off the Brinell hardness as a function of the size of sphere used and the defined test load. For AR cast irons, the most useful conditions

are a 10-mm dia ball and a force of 3000 kPa (exact: 29,420 N). The hardness is then listed as, e.g. 200 HB 10/3000 or HB30 (HB being Hardness Brinell). Standards are ASTM E 10, ISO 6507-1, DIN 50 351.

### ***Dynamic or Rebound Hardness Testing***

Besides the static hardness measuring methods, there exist a few dynamic methods, which are based on the rebound of a body impacting the surface of the casting. Two such methods are:

1. Shore Scleroscope
2. Equotip

***Shore Scleroscope Hardness***—In the past, the Shore Scleroscope hardness tester was very popular, especially in steel roll foundries. A small ball guided in a tube was dropped on the surface and the rebound height was read and used as indicator of the hardness. The device was cheap, pocket size and portable and could be used on large castings on the shop floor, in the cleaning room, in the machine shop or on site. The results were accurate enough if a mean value of a larger number of readings was taken.

***Equotip Hardness***—A modern adaptation of the rebound principle is the Equotip instrument, which was invented in the 1970s. The Equotip measurement technique uses a spherical projectile (tungsten carbide or diamond) and derives the hardness value (L) from the impact and rebound velocities of the indenter (see Appendix D). The instrument automatically measures the two velocities and calculates the L number as an average of five impact events. The hardness is then listed as, e.g.,  $L_D$  or  $L_E$ , depending on the projectile used: tungsten carbide or diamond.

The rebound hardness value is a function of the elastic modulus, yield strength and work-hardening characteristics of the material. Comparison of hardness values is, therefore, only meaningful for similar materials and conversion tables are subject to the same restrictions. Additional factors can be the size and weight, internal stresses, support of the casting, surface roughness, work hardening and inclination of the surface. For example, at hardness levels of above 700 HB, the Equotip gives meaningless numbers (Laird et al., 1995). Of advantage are the ease of the measurement, the portable instruments and the fact that the small indentations do not really damage the surface of the casting.

### ***Conversion and Interpretation of Hardness Values***

Often, it is of interest or necessary to convert hardness values. For this purpose hardness conversion tables are being used (see Appendix D). The hardness conversion tables for steels are, however, not applicable to AR cast irons. ASTM A532 and EN 12 513 show suitable conversion tables. These are the results of regression curves of the scatter band of experimental work (Parks, 1979). For the Equotip and other dynamic

methods, there exist similar conversion tables (Laird, 1995; Parks, 1979; Behrens, 1994). Basically, each type of AR cast iron and each type of microstructure gives a slightly different conversion factor because the response to the various types of test methods is different. The result is that, in a conversion, the normal scatter of a hardness determination is multiplied by the inaccuracy of the conversion factor. This can create difficulties and frustrating discussions when a certain specified hardness has to met. It is, therefore, advisable to agree on the hardness-measuring method beforehand and stick to it at the foundry and at the customer.

In martensitic AR cast irons, the hardness values can sometimes be ambivalent, and it is difficult to find out about their significance. In a medium-C ASTM A532 Class II B iron, which can obtain a hardness of 750 HB in a fully hardened condition, a low hardness of 610 HB, which still meets the specification, can have several causes.

1. The casting may have been stress relieved or tempered at a higher-than-normal temperature;
2. The alloy content and, therefore, the hardenability may have been insufficient for the section and/or the quench rate, resulting in pearlite or bainite formation;
3. The austenitization temperature may have been too low, resulting in low-C, soft martensite or it may have been too high, producing excessive amounts of retained austenite.
4. Overalloying can also result in retained austenite and, consequently, lower hardness.

Each microstructure will perform in a different way when in service. An idea of what has happened, which microstructure to expect and what needs to be changed can be obtained by examining the furnace and heat-treatment documentation. The prior comments demonstrate the importance of strict and well-documented quality/process controls in the production of high-quality AR cast irons.

### ***Microstructure Testing***

An examination of the microstructure in a foundry lab is often not very informative. With normal means, the distribution, size and shape of the eutectic carbides and the presence or absence of secondary carbides can be seen. With some more effort in sample preparation, the presence of pearlite or of graphite in unbalanced Ni-Hard compositions can be detected. Determining the amount of bainite and retained austenite is possible only with sophisticated research equipment and a high degree of experience.

The austenite content has a significant influence on performance, and it is often of great interest to know the exact content in a casting. Unfortunately, the usual austenite determination by x-ray diffraction gives meaningless results, due to the textured microstructure of AR cast irons. In order to

overcome this problem, a special method (Miller, 1968; Kim, 1979) was proposed wherein the specimen is tilted and rotated, in order to overcome the effect of the texturing. However, this still gave unsatisfactory results, so an even more sophisticated method was developed (Fillit et al., 1982). In any case, x-ray diffraction is a time-consuming process that can only be carried out on a laboratory basis, using very small specimens.

The fact that austenite is not magnetic (while ferrite is), is exploited by various instruments, like Magna-Gage, FerritScope or Permascope, to determine the ferrite content in stainless steels or stainless steel welds, as in ASTM A799 and A800. A similar approach was introduced to measure the austenite content in AR cast irons as a shop-floor quality-control tool. Calibration was made by the x-ray method (LaCount, 1980; Morton et al., 1995; Subramanyam, 1985). There were even some specifications written asking for, e.g., 5% austenite maximum, determined by the FerritScope. Unfortunately, the magnetic properties depend on the distribution and the chemical composition of the various phases, which, in turn, are a function of the alloy content, the section and the thermal history of the casting. The results turned out to be fairly unreliable. Meanwhile, the producer of the FerritScope ceased to offer the instrument for this purpose. Also, the stainless steel experts are suspicious about the accuracy of the magnetic determination of the ferrite content e.g., in weldments. The simple magnetic instruments may, however, be used as a quick, in-house check to control heat treatment.

### ***Destructive Testing***

A number of technological tests have been developed or proposed to simulate actual service conditions of castings, in a controlled way. Because the castings are destroyed in the process, these methods are used for research and development purposes. This subject is more fully covered in Chapter 8, Abrasive and Repetitive-Impact Wear Performance.

The most well-known method is the drop-ball test, where grinding balls fall (from a set height) on an anvil, until they spall or fracture (Dixon, 1961; Blickensderfer et al., 1983; Laird, 1995). This test simulates service in a ball mill. In another test, a plate-shaped casting is fixed at one end and a weight is dropped repeatedly (from a preset height) on the free end. This low-cycle impact-fatigue test is supposed to simulate conditions in an impact crusher. Other tests belonging to this category use small ball mills, impact or hammer crushers, bowl mills etc.

As of this date, several laboratory abrasion and impact-abrasion tests have been found to be of some use in determining the abrasive-wear behavior of castings. Albeit, great care must be exercised in applying this laboratory data toward field application performance. These tests include the dry-sand, rubber-wheel (DSRW) scratching abrasion test, the pin-on-drum abrasion test, the impeller-in-drum impact-abrasion test and the jaw crusher gouging abrasion test. For classes of alloys, such as high-Cr white cast irons, ductile irons, steels,

these tests, in conjunction with hardness measurements, provide quantitative information that can be used both to determine equations of the wear rate as a function of hardness and to rank materials according to their abrasive-wear behavior (Hawk et al., 1999). These tests can also be combined with field wear tests to determine the wear behavior of classes of alloys, in terms of normalized wear rates (Tylczak et al., 1999). Once the empirical relationships have been established, subsequent accelerated laboratory wear tests on alloys of the same class of materials can be used to predict field wear behavior.

### ***Nondestructive Testing***

Three nondestructive testing techniques are commonly used for the quality control of alloyed AR iron castings.

1. x-ray or ultrasound;
2. die penetrant;
3. magnetic particle.

X-ray and ultrasonic testing are used to assess internal soundness. Die penetrants and magnetic particle testing are used for surface crack detection and to locate surface discontinuities.

Internal soundness should be checked on a routine basis at the preproduction stage of all new and “remethoded” components (i.e., castings that have had their patterns changed). In some instances, an ongoing test is carried out either by customer demand or as an internal discipline within the foundry. In the most critical cases, 100% of all production is checked.

Both x-ray and ultrasonic techniques are standard, but some experience is required in the interpretation of ultrasonic response due to the dispersal effect of the carbides present in the microstructure. The use of transmission techniques in addition to normal reflection is a useful means of quantifying response. Lower frequencies are required for thicker sections. With suitable equipment, very heavy section castings can be sonic tested. Acceptance criteria, sensitivity, resolution and standards for synthetic flaws have to be carefully discussed and defined between foundry and customer, because the charts used for steel castings are not applicable to AR cast irons, with their coarser and less homogeneous structure.

Die penetrants can be used for surface crack detection and are widely used as a control tool. Detection can be done either by dye penetrant methods for localized examination, or by electrostatic deposition for more general testing. Dye penetrant methods also reveal surface porosity.

A less-convenient but more-sensitive alternative is the magnetic particle technique, which can be applied to ferromagnetic alloys. Very shallow surface cracks are indicated if the magnetic field direction intercepts the major dimension of the discontinuity. Also, near-surface defects can be detected, to some extent. After testing, the casting has to be demagnetized. No specialized adaptation of these techniques is required to examine alloyed AR cast irons.

## Specifying

When ordering AR iron castings, it is important to remember that high-performance castings are customized products, made to withstand demanding environments. A high level of communication between the foundry and purchaser is necessary for the timely delivery of cost-effective castings that are capable of meeting application requirements.

It is important that the engineers for both the foundry and the purchaser begin consultation at an early stage and that the complete design details and dimensions are provided. The cast design can often be optimized to improve performance and reduce costs.

### International Standards

A good place to start when specifying abrasion-resistant cast irons is with the standards provided by ASTM A532, EN 12513 or other international standards (see Appendix E). These standards provide guidelines for the basic requirements and specifications, such as grade, chemical composition, minimum hardness and hardness level, heat-treat condition and use of chills.

The standards give a wide margin for the chemical composition of each grade. This is necessary because one grade can be used to produce very thin to very thick castings; e.g., 10-mm thick mixing paddle liners versus 300-mm section crusher rolls. The wide range of permissible chemical composition also allows the foundry to select the most economic charge mix for the casting by price-conscious selection of raw materials and alloy additions.

Provided that the requirements and specification of properties are met, the choice of chemical composition within the frame of the grade is best left to the foundry—unless there are specific reasons or experiences.

The fact that a product obtained from another supplier previously performed well is no reason to demand exactly the same composition. This composition could have been used by chance or in connection with the specific process conditions at that foundry.

The production process of high-quality AR cast irons, from raw materials to final heat treatment, is very complex and is slightly different from one foundry to another. Each foundry is like an individual and it is virtually impossible to exactly copy the process technology and transfer it to another shop.

During the bidding and design process, emphasis of the discussions should be on the required properties and specifications of the finished casting. The foundry should be provided with details of the service environment; such as type and expected use of the equipment, type of abrasive, static and dynamic loading and eventual corrosion. Abrasion-resistant cast iron foundrymen usually have a lot of experience and know how their products perform. A casting designer is wise to make use of that experience and knowledge.

## Heat Treatment Specification

The heat-treated condition of the finished casting should be discussed and understood. Common heat-treated conditions can be as follows:

- as cast;
- as cast and stress relieved (tempered),
- hardened,
- hardened and stress relieved (tempered);
- soft annealed for machining;
- soft annealed and hardened;
- soft annealed, hardened and stress relieved (tempered).

In addition there are several special heat treatments:

- as-cast and subcritical tempering for hardening;
- hardening and subcritical tempering in various cycles to reduce the austenite content or to achieve a secondary hardening;
- cryogenic hardening.

The term tempering is not well defined. According to EN 12513 it is a heat treatment between 200 and 550°C (392–1022°F). This comprises the normal temper at 220–270°C (428–518°F) without significant structural changes, as well as the subcritical tempering that transforms retained austenite and brings about a significant change of matrix structure and properties.

For pearlitic chilled irons, the depth of chill, the degree of mottle and, eventually, the method of determining these, should be agreed on.

### Hardness Specification

An important and easily measurable property is hardness. The AR cast iron standards specify a fairly low minimum hardness. Actually, in some applications, castings give their best performance at maximum attainable hardness, while in others, the best results and/or the best cost-performance ratios are achieved with a comparatively low hardness.

In discussing hardness, however, it should not be assumed that a lower hardness means a higher toughness. This feeling, based on experience with quench-and-temper (Q+T) steels, is completely wrong for AR iron castings. On the other hand, specifying extremely high hardness levels is often detrimental, if it involves very high quench rates with the risk of cracking or high internal stresses, or if the tempering has to be omitted.

The location and the number of hardness tests should be agreed on, as well as the type of test to be used. As mentioned before, hardness conversions involve a certain inaccuracy. If hardness measurements are to be made as a means of quality control during the processing (as-cast, as-quenched, after tempering), there has to be agreement to, as well as a documentation of, the heat-treatment cycle.

## General Specification Advice

Testing for soundness and external discontinuities has to be specified together with the testing method to be used, the test location and the acceptance criteria.

The primary requirement for AR iron casting is, of course, a long and satisfactory service life. In some applications, it has become customary to give guaranties for a minimum life. This is a significant commercial risk that the foundry can take, if sufficient experience and service data are available.

In ordering and judging performance of AR iron castings, it has to be kept in mind that abrasion resistance is not a property of the material but is a quality that is subject to the service conditions. Seemingly minor changes can have dramatic consequences for the performance of the wear-resistant part. Relaying information to the supplier about adverse changes in service conditions is a good policy. The supplier can often neutralize such adversities by changing the alloy composition or the heat treatment of the cast parts with little or no additional costs.

A continuous relationship and exchange of information between foundry and end user is essential for the improvement of wear-resistant parts and the development of AR iron castings that give optimum performance.

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# Chapter 7

## Production Practices

Abrasion-resistant white cast irons are distinctly different from the usual gray or nodular cast irons. In many respects, they are more similar to high-strength steels. In fact, iron foundries who have neglected this fundamental difference have often experienced much greater difficulties when starting the production of high-quality AR iron castings than have steel foundries.

The main difference from normal cast irons is that AR cast irons solidify without graphite. This dramatically changes their feeding characteristics. The volume shrinkage at solidification of AR cast irons is not compensated for by the volume expansion of the graphite precipitation as it is in gray or nodular irons. Typically, a solidification shrinkage of about 4–5% will occur. This is almost identical to the shrinkage found in steel castings. The risering and feeding design of the mold is, therefore, similar to that of steel castings.

However, in contrast to steel, AR irons have little or no ductility. Their solidification mode, their carbide structure and their various transformations at relatively low temperatures make AR irons prone to tearing and cracking in all stages of processing. Furthermore, martensitic AR irons of the Ni-Hard family and the high-Cr irons are high-alloy materials. Such high-alloy materials require special process controls in both melting and founding.

In Chapter 1, AR cast irons were classified into five *broad* groups: pearlitic, Ni-Hard or Ni-Cr, Ni-Hard 4, high-Cr and specialty. From the viewpoint of *foundry production practice*, the AR irons can be divided into the following five groups:

1. **Fully White Pearlitic**—Abrasion-resistant irons that are white throughout the whole section, with pearlitic matrices in the as-cast condition.
2. **Chilled**—Abrasion-resistant irons for castings that have a white iron working layer and a gray iron backing. These irons are usually produced by casting against a chill and have pearlitic matrices in the as-cast condition.
3. **Ni-Hard 1 and Ni-Hard 2**—Nickel-chromium AR irons of the Ni-Hard 1 and 2 type, which are martensitic as-cast.
4. **Ni-Hard 4**—High-alloyed Ni-Cr AR iron with an austenitic-martensitic matrix in the as-cast condition; often heat treated to obtain a more martensitic matrix.
5. **High-Chromium**—High-Cr cast irons with a wide variety of as-cast structures. These irons are usually heat treated to obtain maximum hardness or toughness, or resistance to repeated impact.

### Fully White Pearlitic AR Cast Irons

This group of pearlitic AR irons was the only abrasion-resistant white iron available for a long time and was produced in large quantities for all types of wear parts (Poetter, 1953; Meyer, 1954).

The structure is similar to as-cast malleable iron, but the carbon content is usually higher. Due to their low toughness and an abrasion resistance that is inferior to the alloyed martensitic grades, the importance of pearlitic AR irons has decreased steadily. Today, their main uses are for low-cost, small grinding media and some unstressed wear plates and liners of modest section thickness.

With cost being the foremost consideration, requirements are often limited to obtaining a minimum hardness. This requirement is often met by the sole adjustment of the alloy's C content. The other elements may vary in a wide range, provided a white solidification is assured. This is often done by chill casting, especially for grinding media.

### Chemical Composition

These materials are classed as carbon irons, with carbon being the basic alloying element. The carbide formed is essentially  $\text{Fe}_3\text{C}$ .

### Melting

In view of the low demands for compositional accuracy and metallurgical quality, any melting unit can be used. Usually, cupola melting is the most economical way, provided suitable coke is available at a reasonable price. Charge materials consist of steel scrap, cast iron scrap, foundry returns and, if unavoidable, ferroalloy additions. The coke in the cupola provides carburization. A sulfur pickup is helpful for chilling. While the C content is kept near eutectic or as high as possible in view of cracking, the Si content is kept below 1%, usually about 0.5%, in order to avoid oxidation of the melt and graphite formation.

Too high a nitrogen pickup of the steel scrap charge can sometimes be a problem resulting in gas porosity. As long as N does not cause porosity, it is beneficial because it increases hardness and suppresses graphite formation. The solubility limit of N depends mainly on the C content and the type of molding or the solidification rate. It can be controlled by the addition of Ti-microalloyed steel scrap or ilmenite ore in the charge.

*(Fully White Pearlitic, continued)*

The only useful alloy addition is Cr to provide chill in heavier sections, but this is avoided for cost reasons, whenever possible. Usually the unsorted steel scrap introduces a certain amount of carbide stabilizers. Sometimes, some steel-making pig iron is used in the charge. It has a low Si and a high C content and introduces Mn, which helps a little for chilling and increases matrix hardness.

The other melting units, such as rotary furnaces, arc furnaces or induction furnaces, can be used, especially if discontinuous production or the energy supply does not justify cupola melting. All these melting methods require the addition of a carburizer to the steel scrap charge. Pig iron is usually too expensive. It should be kept in mind that the graphite of gray pig irons, including steelmaking iron, as well as some carburizers, can act as an inoculant and bring about an undesirable graphite formation. Higher superheat, longer holding time or the addition of Cr can counteract this. Melting, superheating and tapping temperatures should be kept as low as feasible to reduce costs, however, in practice, are often fairly high when a lot of small castings have to be poured.

### **Molding, Gating and Riser**

The patternmaker's allowance for AR iron castings is typically between 1.8 and 2% (1/4 in./ft), with the usual dependency on the type and shape of the casting.

Grinding media constitute the major application and use of pearlitic AR irons. A common manufacturing route is to cast the iron into chill molds made of high-C gray cast iron. The chilling power of the metal mold, coupled with the small diameter of the grinding media, is often more than sufficient to ensure a fully white solidification.

The castings can also be produced in sand molds. All of the common techniques and binder systems are suitable. The main precaution is to avoid graphite formation due to slow cooling rates in the eutectic region. Chill plates or blocks are most helpful. Chills have to be kept in good order, to avoid blowhole defects resulting from surface oxides or humidity trapped in fissures. When split chills or several chill are used, there is a risk of cracks at the parting lines due to steep thermal gradients. The parting-line zone of the casting is subject to hot cracking because it freezes later than the chilled section and, hence, is under a concentrated tensile stress during its solidification. Chills are coated and many different chill coatings are used. The purpose of chill coatings is to avoid burning-in while, at the same time, not insulating the molten metal from the chill. They are also instrumental in producing smooth, clean, fine-grained surfaces.

Allowance should be made for the warping of castings that are not symmetrically chilled. For example, a simple rectangular casting made in a horizontal position with the drag face chilled, will be convex on the chilled face and concave on the opposite face. Counterwarping the pattern and the chill, if one

piece, will overcome this difficulty. Warpage is determined by the thermal gradient within the filled mold. It varies with pouring temperature, pouring speed, thickness of the casting and thickness of the chill. With these and other factors coming into play, it is obvious that warpage figures can only be determined for particular patterns and conditions.

Runners, ingates and risers should be generous and should allow a directional solidification. No fixed rule can be offered for the number, size, height or positioning of risers, as these conditions will be determined by the individual case. However, they should be kept away from the critical wear surfaces and the parting line. The solidification shrinkage depends on the composition and is between 4 and 5 volume percent. It decreases with increasing C content.

The riser modulus can be estimated at 1.2 times the casting modulus of the section to be fed. The mass of a riser is intended to slow and direct the solidification front. With this slower solidification rate near the riser, the risk of graphite formation beneath it can become an issue. Whenever practical, risers should be surrounded with exothermic sleeves or insulating refractories to increase their effectiveness and improve casting yield. When the mold is poured, the risers should be covered with some type of insulating or hot top material to keep them open to the action of atmospheric pressure. With large castings, it is feasible to feed liquid metal from a ladle through an open riser fairly late after pouring the casting. In order to facilitate riser removal notching or breaker cores can be employed.

Abrasion-resistant irons are materials with no ductility. A casting hung by risers in the mold and restrained during shrinkage runs a risk of tearing. This risk can be avoided or minimized by suitable positioning of the risers, by the use of a collapsible mold or by loosening the sand around the risers after solidification. The thermal gradients created by a heavy riser can result in internal stresses, causing cracks. The hot tearing propensity decreases with increasing carbon equivalent (CE) above 3.5%. It is generally increased with higher pouring temperature. There are observations that relate it to the solidification mode of primary dendrites and to melt processing and charge mix.

*Reinforced Pearlitic AR Iron Castings*—Reinforced Pearlitic AR iron castings can be made by inserting steel rods; e.g., rebars in the mold and pouring the iron around them. The intention is not to obtain a bonding by fusion or welding, because in this case the steel will be embrittled by carbon pickup and the cracks will pass through it. Therefore, the steel is coated to avoid bonding. The reinforcement does not really prevent eventual cracking of the AR iron but prevents the parts splitting apart. Sometimes liner plates for chutes and bins have been produced in blast furnace plants in this way by using liquid pig iron in this way.

### **Pouring**

No special information.

## Postcasting Processes

### ***Cleaning/Finishing***

In handling, cleaning and finishing, it should be recognized that AR iron castings are much more brittle than steel, nodular or gray iron. Heavy risers should not be removed with a cutting torch because of the hazard of thermal cracking. Also when grinding, excessive pressure has to be avoided in order to avoid heat checks.

### ***Heat Treating***

The only heat treatment is a stress relief that can be done at temperatures up to about 600C (1112F) without risk of graphitization; however, some loss of hardness may occur due to tempering of the pearlite.

A hardening treatment is rarely used. It involves higher temperatures that go into the austenite range, above 780C (1436F). This brings about a graphitization unless the carbide is stabilized with at least 2% Cr. Chilled, fine-grained sections graphitize faster than do coarser structures. Liquid quenching is unfeasible, except for small grinding media, because of quench cracking. Slower cooling, to avoid cracks, does not provide sufficient hardening.

### ***Welding***

Abrasion-resistant irons are not weldable. The brittle structure is very sensitive to heat checking; therefore, cracks under the weldment are unavoidable, unless extremely high preheat temperatures (above 750C/1382F) are used. Such high temperatures are commercially impractical and involve the risk of graphitization. Fabrication and repair weldings are, therefore, not possible, and attempts to weld-repair defective castings usually aggravate the situation. They can, at best, be a cosmetic operation.

### ***Machining***

A common and obvious way to machine AR irons is by grinding. Pearlitic AR irons can be machined with carbide, coated or ceramic tipped tools and, especially, with cubic boron nitride (CBN) tools. In order to avoid vibrations and misalignments of the tool, which may break the casting, sturdy and very stiff tool machines are required. Milling and turning pose no major problems.

Drilling small holes and cutting threads is difficult due to the problems of chip removal. In this case, mild steel inserts or cores can be used. Mild steel inserts should be well away from the corners and edges, in order to avoid cracks due to the different thermal expansion coefficients. Since they should have a metallic bonding to the surrounding metal, they must be absolutely clean and dry when placed into the mold. Modern coremaking allows even small holes and threads to be cast.

Soft annealing is not possible because the structure will be graphitized and cannot be rehardened. That is, once the existing carbides have transformed to graphite, you then have a malleable iron.

## Chilled AR Cast Irons

Castings that have an AR iron working layer and a gray iron backing are produced by casting against a chill. Typical castings are some types of rolling mill rolls, camshafts and tappets, mixing rollers for clay, hammers for some types of coal crushers and various liner plates. Such castings can, in fact, be regarded as compound or bimetallic designs.

Requirements for chilled castings are as follows; a graphite-free AR outer layer having a controlled and specified thickness, a defined mottled intermediate layer and a gray iron backing that is free from carbides to provide resistance against fracture of the whole casting and to allow machining. Meeting these demands calls for a close control of the chemical composition, graphite nucleation and chilling effect. Carbon, silicon, sulfur, manganese, phosphorus and eventual additions of chromium and other elements must be well balanced and maintained within close limits. The same applies for all factors that influence graphite nucleation such as raw materials, melt processing and inoculation, thickness and coating of the chill and pouring temperature.

*Chilled Nodular Cast Iron*—The toughness and strength of gray cast iron is limited, and substituting it with nodular iron should offer many advantages. Unfortunately, it is virtually impossible to avoid fine nodules forming in the chilled iron layer. The layer may have a white fracture but under the microscope nodules are visible at sufficient magnification. These graphite precipitates greatly reduce abrasion resistance. Adding Cr or any other carbide former to suppress them will result in free carbides in the nodular iron backing and embrittle it to and below the level of gray iron and make it very difficult to machine, too.

The high elastic modulus of the nodular cast iron will also result in very high internal stresses in most castings and increase the risk of cracks. For example, the elastic modulus of typical gray irons is around 112 GPa (16 Mpsi) while that for nodular iron is around 175 GPa (25 Mpsi).

### **Chemical Composition**

These materials are classed as carbon irons, with C being the basic alloying element. The carbide formed is essentially Fe<sub>3</sub>C.

### **Melting**

Chilled cast irons can be melted in any furnace that allows the required control of chemical composition and nucleation. At one time, this was achieved in a cupola, reverberatory or open hearth furnace. The latter have been discontinued for well over 30 years due to poor energy efficiency. The cupola has to operate very consistently with a charge of special pig iron, returns, selected steel scrap and ferroalloys. Results can be excellent, but this way of melting is fairly expensive, and orders are rarely large enough to allow for a cost-effective, consistent melting campaign.

(Chilled, continued)

Electric melting is now the standard and usual practice. Among the electric melting choices, arc furnace melting is not the first choice, since it affords limited control over nucleation. Mains and medium-frequency induction furnaces are the preferred alternative. Charge materials—pig iron, returns, cast iron and steel scrap, ferroalloys—should be of good and consistent quality. In fact, the requirements are higher than for melting normal gray iron. The melting cycle, with regard to temperatures, sequence and time of additions and holding periods, should be kept constant, because the control and adjustment of chill depth (distance into the casting to which white iron extends) depends on experience, to a large extent.

Rotary furnaces are suitable only if the required close control of composition and melting conditions is achievable.

The chill depth of the melt can be determined by means of a test casting. It is usually a rectangular plate (e.g., 200 mm H x 100 mm L x 20 mm W) with its width sitting on a chill block. It gives better information than the wedge test, for this purpose. Depth of chill can be adjusted, to a limited extent, in the ladle: additions of S or Cr increase it and FeSi, as an inoculant, decreases it. However, a melt that is way out of specification is difficult to save, after the fact.

### **Molding, Gating and Riser**

Riser is normally not a problem in chill-cast irons, because the AR iron will be fed from the gray iron side. This allows the foundry to use the gray iron as the riser with all the attendant feeding and riser benefits of this material. There is obviously a pronounced directional solidification. Suitable risers can be placed on the gray iron part if necessary.

The gray iron backing can be molded by any of the common sand systems. Ingates should be positioned well away from the chill to avoid its heating up by hot metal flowing across it.

Chills are critical. They must be kept under constant surveillance and be well maintained. Worn chills with deteriorated surfaces not only involve the risk of surface defects in the casting due to reactions with oxides, rust and humidity, but have also an uncertain chilling power. Chills are always coated and many different coatings are available. The purposes of chill coatings are to avoid the burning-in defect and produce a smooth, clean, fine-grained surface. Coatings are also used to control the chilling effect, to some extent, via their thermal properties. Split chills involve a risk of cracks at the parting line and should be avoided whenever feasible. Because the casting is chilled from one side only, warping is expected to occur and has to be allowed for.

Hot tearing can be a problem. The AR white iron shrinks during solidification while the gray iron section expands. This can easily result in tears within the chilled layer. Contrary to fully white iron, a higher C content is not beneficial because more graphite is precipitated in the gray part and increases expansion. A common countermeasure is to increase the P

content of up to 0.6%, even if this reduces the impact properties and machinability of the gray part. Phosphorus does not prevent hot tearing, but the phosphide eutectic remaining at the end of the solidification has a low melting point and, consequently, fills and seals the hot tears in the white layer. This is just a cosmetic fix to an underlying problem.

**Bimetallic Castings**—Centrifugal casting is a very common and direct way to produce compound or bimetallic pieces. The AR iron melt is cast into the rotating mold first and then after a short time the gray iron melt is poured to form the inner layer. Provided timing and fluxes allow a good bonding without too much mixing of the two materials, excellent results can be achieved. The same method can be applied for nodular cast iron. Obviously, a fairly wide range of melt compositions and types can be combined once the suitable parameters have been established.

### **Pouring**

No special information.

### **Postcasting Processes**

#### **Cleaning/Finishing**

Due to the bimetallic nature of the casting and the effect of the chill, the castings may contain fairly high internal stresses. This should be kept in mind when handling and cleaning. Severe shocks and dynamic loading, as well as sudden temperature changes, should be avoided. Tool machines and fixtures should be rigid to avoid bending stresses or vibrations. There may be a small risk that reducing the section of the gray iron part by machining, grinding or arc cutting will alter the stress pattern in such a way that the whole casting will crack or even explode. Special care, in this respect, has to be exercised when removing large risers. Internal stresses can be reduced, either by slower cooling in the mold until about ambient, or by stress-relief heat treatment.

#### **Heat Treating**

Stress relief can be given at temperatures up to about 600C without risk of graphitization of the white iron and decomposition of the matrix of the gray part, but some loss of hardness due to tempering of the pearlite may occur.

A hardening treatment is unusual and difficult to give, due to the bimetallic nature of the casting. An exception are camshafts, where the white and mottled cams are surface-hardened.

#### **Welding**

While the AR white iron part is not weldable, limited repair or fabrication welding can be done on the gray iron part in the same way as for gray iron castings. Attention should be paid to fact that the gray part is connected to a brittle and crack sensitive white part.

If larger weldments have to be done, preheating the whole casting to 300–500C is advisable, but often not feasible. The purpose of preheating is to avoid the formation of martensite

or bainite, as in usual gray iron welding, and also to minimize thermal stresses between the gray and white iron parts.

When welding with NiFe or soft steel consumables, without preheat, heat input must be low enough to avoid thermal distortion and significant temperature gradients, especially across the gray-white boundary. Prior to welding, it is usually advisable to stress relief the castings.

### **Machining**

Machining the gray cast iron part of the casting is the same as for normal gray iron castings. The AR white iron part is rarely machined. If it is necessary, the same principles apply as for fully pearlitic cast irons.

## **Ni-Hard 1 and Ni-Hard 2 (Ni-Cr) AR Cast Irons**

When considering production practice, Ni-Cr AR irons of the Ni-Hard 1 and 2 type (as well as other grades that are alloyed in a similar way to be martensitic as-cast) have to be regarded as high-strength ferrous materials similar to alloyed steels, and must be processed accordingly (NiDI, 1996; Spear, 1990). Charge materials, composition, melting practice, molding, internal and external soundness and surface finish determine the quality of the castings and their service performance and must be carefully controlled. Poor production technology and insufficient process control will result in castings that give substandard performance or cause excessive scrap rates within the foundry.

Much of the material covered in this Ni-Hard 1 and Ni-Hard 2 section also applies to Ni-Hard 4 AR cast irons.

### **Chemical Composition**

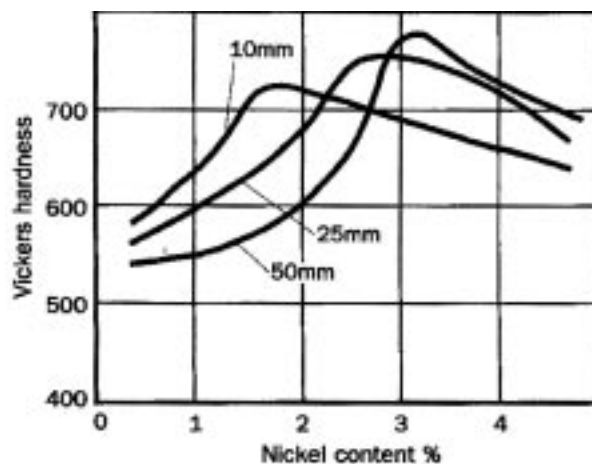
For the chemical composition of Ni-Hard 1 and 2, there are three rather important guidelines:

1. *Regarding the carbon content*—A lower C content results in higher impact resistance, but lower abrasion resistance, and vice versa. Therefore, the C content has to be adjusted to the application of the casting.
2. *Regarding the graphite content*—Excessive C and Si contents, combined with insufficient Cr content and inoculation by late additions of Si and other graphitizing elements may provoke graphite precipitation in heavier sections. Graphite is detrimental to toughness and abrasion resistance, but a small graphite content may not lower the hardness too much. Sometimes, a ladle addition of about 0.015% Bi is made to suppress graphite formation. Silicon is usually kept between 0.4 and 0.7%. Lower Si content involves the risk of slag defects and coarse structure. Sometimes a Si content of up to 1.5% is used to favor a matrix with less retained austenite and more martensite. However, they have to be compensated for by an increased Cr level and an eventual Bi addition.

3. *Regarding the Ni-Cr content*—The Ni and Cr levels have to be adjusted to the casting section size, in order to achieve the highest hardness and abrasion resistance. This has been discussed in the preceding chapters. Too low a hardenability (e.g., low Ni content) can result in high-temperature transformation products of austenite to pearlite and/or bainite, giving low hardness; too high a hardenability (e.g., too much Ni) will cause excessive austenite levels, which impair hardness and resistance to repeated impact. Figure 7.1 shows the relationship between hardness and Ni content for three sections. Table 4.1 (in Chapter 4) gives some guidelines for the chemical composition for various sand and chill cast sections.

In general, some guidelines can be provided for the common alloying elements.

- Manganese is best kept at 0.5–0.8%, but sufficiently high to combine with all sulfur. However, medium- and heavy-section castings have been produced successfully with much higher Mn content.
- Copper has sometimes been used to replace Ni for economic reasons. Up to 2% Cu has been used successfully. However, there are some indications that Cu may be slightly detrimental to toughness. This can be especially true when returns are melted or when the melt is oxidized.
- The Cr:Si ratio should be about 2.7:1 for thin-section castings and about 2.9:1 in medium and heavy sections. An increase of Si also demands an increase of Ni, because Si reduces hardenability, which may result in pearlite formation. The ratio should be  $\%Ni = 1.3\% Si + 1.6$  for 25-mm (1-in.) sections and  $\%Ni = 1.3\% Si + 2.8$  for 200 mm (8-in.) sections.
- Phosphorus should be kept below 0.25% or, better, below 0.15%; therefore, some pig irons and chilled iron scrap should be limited in the charge mix.



**Fig. 7.1. Relationship between section size, hardness and Ni content for Ni-Hard 1 castings (NiDI, 1996).** A low hardness at high Ni content for a given section is due to too much retained austenite; a low hardness at low Ni content is the result of pearlite formation.

## Melting

Ni-Hards can be melted in the cupola, cupola plus duplexing, rotary, arc or induction and crucible furnaces. The cost and availability of fuel and electric power and the size and nature of the castings to be produced, are the usual factors that determine the type of melting unit to be used. The relative difficulty in cupola-melting Ni-Hards has led many foundries to use electric melting; however, cupola melting is still of importance.

Because much of this text also applies to Ni-Hard 4, this Melting section contains more detailed information than the other Melting sections in this chapter.

### Cupola Melting (INCO, 1949)

The original Ni-Cr AR iron compositions were developed for iron foundry production at a time when cupolas were the only available melting units in iron foundries. Cupola melting restricts Ni-Cr iron to narrow chemical ranges, and a high level of operating skill is required to produce uniform quality. It is essential that the alloyed product be completely isolated from other irons that are being melted. This can be done by any of the following ways:

- Melting Ni-Hard in an independent cupola.
- Melting only Ni-Hard in the cupola, then rebuilding the bed and melting other iron.
- Melting other irons first, draining the cupola of all metal and slag, rebuilding the bed, charging and melting Ni-Hard.

Experience has repeatedly demonstrated that efforts to separate Ni-Hard from other irons in the cupola by using coke spreader charges of 50, 75 and even 100 cm (20, 30 and 40 in.) invariably leads to dilution of the Ni-Hard composition and, therefore, insufficient quality. This practice should not be used.

Properly sized coke, melting stock and charges must be used in order to consistently produce quality Ni-Hard. The size of the metal charge should be such that the corresponding coke charge has a depth of approximately 15 cm (6 in.).

The alloy material is too valuable to waste as first iron of poor quality that cannot be used for castings. A good preheating of the coke bed and the cupola refractories (to a high-temperature prior to the beginning of the heat) is required.

Charge materials consist of Ni-Hard scrap or returns, steel scrap, sometimes pig iron and/or alloy pig and alloy additions. A typical charge would be 50% Ni-Hard scrap or returns, with steel scrap and alloys making up the remainder.

Alloyed Ni-Cr pig should be viewed with great suspicion. Such pig is often melted from mixed scrap and residues of doubtful origin that cannot be used in any other way and may contain harmful tramp elements. High-alloy steel scrap can be used to introduce Ni and Cr instead of Ni or FeNi and charge chromium.

The recommended order of charging materials in the cupola is as follows:

1. Steel scrap, alloyed or unalloyed
2. Pig, Ni-Hard scrap, etc.
3. Returns
4. Alloys

The usual oxidation losses encountered in melting Ni-Hard in the cupola are:

- Silicon . . . . . 0–10%
- Manganese . . . . . 20–30%
- Nickel . . . . . nil
- Chromium . . . . . 10–15%~

On short heats, it is safe to plan on no loss of silicon. For heats of an hour or more, a loss of approximately 10% Si will be experienced over the latter half of the heat. It must be remembered that Si (or any other chill-reducing element) should never be added to Ni-Hard as a ladle addition because it will act as an inoculant and produce a mottled structure. An Si content of below 0.4% is not recommended.

Uniformity of alloy content is absolutely essential in Ni-Hard production. In order to maintain the Ni content at limits of  $\pm 0.4\%$  and Cr at  $\pm 0.15\%$  in cupola production, the following precautions must be observed:

- Strictly adhere to the recommendations given previously for metal and coke size; carefully charge all components to insure a uniform, horizontal distribution of materials and avoid bridging.
- Accurately weigh charges, with special emphasis on the alloy components.
- Use a mixing ladle with the metal level always held at a minimum of four charges before metal is removed for pouring; or, alternately, collect a volume of metal equivalent to at least four charges in the pouring ladle before it is sent to the pouring floor.

The C content from the cupola, particularly for the first tap can be rather high (around 3.3–3.6%). Impact properties of such high-C Ni-Hard are mediocre, and there is a risk of graphite formation if the Si is on the high side. Permanent mold casting in chill molds takes care of these problems; in fact, cupola melting is best suited for this type of production. Typical castings are grinding balls and mill liners.

Other problems in cupola melting can be a high sulfur content of up to 0.15%, which is detrimental to abrasion resistance and impact properties and can be a cause of tearing. It may be necessary to desulfurize the melt.

**Duplexing**—Duplexing the cupola iron into an electric furnace is the easiest way to avoid the typical problems of high-C, high-S and alloy variations. The C and other alloy content can be easily adjusted by the addition of steel scrap to the electric furnace. Additionally, there is usually sufficient temperature reserve for desulfurization.

### **Rotary Furnace Melting**

Melting Ni-Hard 1 and 2 in modern rotary furnaces with oxygen/gas or oxygen/oil burners has given reasonable results, provided the burn-off of alloys and the temperature can be controlled. Carbon has to be introduced as low-Si pig iron or Ni-Hard scrap and returns. This is because steel scrap is difficult to recarburize by the usual carburizers used in induction or arc furnace practice. Gray or ductile iron scrap can provoke graphite precipitation.

### **Induction Furnace Melting**

Electric melting, especially in mains or medium-frequency induction furnaces, is the preferred way to produce high-quality Ni-Hard castings (Spear, 1990). Refractories can be acid, neutral or basic, but acid refractory linings are most common. Modern, computer-controlled, medium-frequency induction furnaces with high-energy density are especially suited, because they allow the essential complete control of all melting parameters and short melting times.

The charge materials consist of steel scrap, Ni-Hard returns and scrap, eventually white pig iron with low Si, S and P contents, and suitable alloy additions. Gray or ductile iron scrap can lead to graphite precipitation. The Si content has to be carefully controlled and an eventual 0.2–0.4% Si pickup from an acid lining must be considered.

The quality of induction-melted Ni-Hard depends on careful selection of raw materials with limited use of returns and Ni-Hard scrap to allow room for virgin material in the charge and a melting practice that produces uniformly low hydrogen and oxygen contents. If the amount of returns and scrap in the charge is increased at the expense of primary materials, some loss in quality is inevitable.

An increasing amount of returns will increase the oxygen content in the metal. If the ratio of returns and virgin material is kept constant, a constant equilibrium oxygen content will be reached. Oxygen is detrimental to toughness, so breakage can occur during handling, heat treatment, or while in service. It may also be a cause of hot tearing, especially in conjunction with sulfur. Oxygen interacts with residual Cu to form embrittling needle-shaped precipitates in the martensitic matrix. The dangerous oxygen content seems to be above 200 ppm, but 600 ppm are not uncommon. Superheating and holding at temperatures of above 1600C should be avoided.

Hydrogen is usually introduced by wet, oily or rusty material charged onto a heel of molten metal in the furnace. A slowly increasing H level may eventually cause subsurface pinholes in the castings (appearing after blast cleaning) and later large blowholes. Hydrogen may also be a cause for cracking in the foundry or for erratic breakage of castings in service by a mechanism that seems to be similar to H embrittlement in high-strength steels. Badly gassed metal in induction melting can be purged by argon, introduced via a lance.

Pinholes can have several causes; one is H that is introduced via the charge. Other causes are wet ladle linings or H

pickup from green molding sand. In the latter case, Al is often accused to favor H pickup from the sand.

Oxygen content is another cause. Oxidic slags can react with the C in the melt to form gas bubbles. This is favored by low pouring temperatures and low Si content. The bubbles can be open to the surface and are oxidized or they remain subsurface and metallic. A sufficient seacoal content in the green sand seems to be beneficial.

Yet another cause of pinholes can be N, which is either introduced by the charge materials or picked up from molding sands. Organic binders contain appreciable N content, which can be accumulated during regeneration. Charge materials that introduce N can be carburizers or ferroalloys.

Charge materials for induction melting are:

|                        |                      |
|------------------------|----------------------|
| low-Si charge Cr       | high-C FeMn          |
| steel scrap            | Ni                   |
| Ni-Cr iron returns     | C raiser             |
| alloy scrap            | Ni-Mg addition alloy |
| low-Si, low-S pig iron |                      |

For high-quality martensitic AR iron, alloy scrap usage should be limited to materials that contain only specified elements. Use of Ni-Hard or Ni-Cr ingot pigged from miscellaneous scrap that may contain unspecified and harmful tramp elements should be limited to casting applications involving light abrasion loading without impact.

Following are some guidelines for cold charging and back charging in an induction furnace:

#### *Melting Practice for Cold Charging in Empty Furnace*

1. Charge in the following order:
  - pig iron
  - steel
  - Ni
  - Ni-Cr AR iron returns
2. Melt down, superheat to 1510C (2750F).
3. Add to furnace:
  - charge Cr
  - high-C FeMn
4. Superheat to temperature required for handling and pouring.
5. Tap onto 0.2% foundry grade 75% FeSi or about 0.4% NiMg15 in the bottom of a hot clean ladle. (The Mg alloy (NiMg15) should be at about 0.4% to obtain an magnesium content of between 0.02 and 0.03%.) After the Mg treatment, the melt should be inoculated with FeSi.

The Mg treatment, followed by inoculation with FeSi, has a favorable effect on Ni-Cr AR irons in foundry processing and in service. It is highly effective in reducing or eliminating breakage in blast cleaning, gate and riser removal and in service. The Mg addition makes the carbides more compact, thereby increasing the toughness of the alloy. It also seems to improve martensite formation.
6. Pour quickly at minimum practical temperature for best results in wear and toughness.

### *Melting Practice for Back Charging on Molten Heel*

#### Standard Practice:

1. Preheat *all* charge materials to 320C (600F) minimum to eliminate H pickup from wet, oily or rusty scrap. Water evaporates below 100C (200F); oil and grease evaporate at temperatures up to 200C (400F). Rust requires a minimum preheat temperature of 320C (600F) to drive off the water of hydration. Reheat should, however, not be so high as to cause scaling.
2. Follow steps 2 through 6 of Cold Charging in Empty Furnace.

#### Alternate Practice:

1. Charge in order:
  - pig iron
  - steel
  - Ni
  - returns
2. Melt down completely. At 1370C (2500F), purge melt with argon, using graphite lance or steel pipe. Argon should be introduced 25–30 cm (10–12 in.) below the surface as finely divided bubbles, with furnace power on low to roll metal. Purging time depends on heat size and amount of Ar. Argon volume should be sufficient to roll the metal strongly, without actually throwing metal. A one-minute purge in a 140-kg (300-lb) furnace and a ten-minute purge in a 2.8-t (6000-lb) induction furnace have been ample for eliminating H effects from badly gassed metal.
3. Follow steps 3 through 6 of *Cold Charging in Empty Furnace*.

*Effect of Recycling Returns on Melt Quality*—Recycling returns increases the amount of dissolved O in the melt. Charges made up of a constant ratio of returns to primary materials (i.e., pig iron, steel scrap) will eventually reach an O equilibrium. The higher the ratio of returns to primary materials in the charge, the higher the equilibrium of oxygen. The lower the ratio of returns to primary materials in the charge, the better the melt quality will be. Making up the charge with a constant ratio of returns to primary materials will promote uniform quality in the metal and the castings.

Copper is sometimes used as an alloying addition to replace part of the Ni in martensitic AR irons. This can have an embrittling effect, when large percentages of returns are recycled in induction melting, or when metal is held in the furnace for long periods of time. With Cu content is over 0.25% in heavily recycled material, the Cu effect shows up as bands or needles of carbide extending across the martensitic matrix. The embrittling effect can result in casting breakage during handling or erratic casting breakage during service. Copper has no detrimental effect on an initial induction furnace melt made with primary materials, or in arc furnace heats that are refined by oxygen lancing.

A high-O content in the Ni-Cr irons has a detrimental effect on toughness. Superheating and holding at temperatures over

1640C (3000F), in arc and induction furnaces, to get rid of hydrogen gassing, has resulted in erratic premature failures of grinding mill liners. Long holding times for molten metal in channel or large coreless induction furnaces also results in high O content in the cast metal. Oxygen content of 200–600 ppm are common in cases of premature failure of mill liners.

Hydrogen is a detrimental element, even when it is present at below the H level that can cause gas defects in castings. Hydrogen atoms will migrate to stressed or work-hardened areas in castings. Hydrogen gas forms at inclusions or in porous areas, rupturing the castings. This can occur as delayed cracking before the casting goes into service or as erratic early failure while in service.

Sulfur content can become a problem with increasing O content in the Ni-Cr irons. Sulfur trouble first shows up with increasing O content. Sulfide inclusions appear in a Chinese-script form, causing a substantial increase in impact wear rate. At higher O content (over 200 ppm) the Ni-Cr iron can become sensitive to hot tearing, subject to cracking during heat treatment and suffer erratic breakage in service.

Low-melting metals—tin, lead, arsenic, antimony—segregate to the cell boundaries during solidification of martensitic AR irons and cause embrittlement. Lead contamination also greatly increases susceptibility of Ni-Cr AR iron to hydrogen gassing when returns are recycled by induction melting.

### **Arc Furnace Melting**

The basic principles for arc furnace melting of Ni-Hards are the same as for induction melting. However, the arc furnace offers the possibility of refining the melt by a carbon boil or oxygen lancing. A boil-out of about 0.3% C is usually sufficient. In this way, H can be removed from the melt, and problems due to high contents of returns in the charge mix or due to Cu can be eliminated.

### **Molding, Gating and Riser**

Much of the information for Ni-Hard 1 and 2 is the same as for Chilled Cast Irons.

All sands and binder systems used for gray and ductile iron castings or steel castings can be used to make molds and cores for the Ni-Hard castings. The patternmaker's shrinkage rule is between 1 and 2% (1/8 to 1/4 in. per ft). Typical values for Ni-Hard 1 and 2 are 1.25% for restricted and 1.95% for unrestricted shrinkage. Chills and insulating or exothermic riser sleeves can be employed in a similar way as for cast steel in order to improve feeding and casting yields. Break-off cores at riser contact are useful in removing risers. A 60-degree angle is usual.

Chills have to be kept in good order to avoid blowhole defects resulting from surface oxides or from humidity trapped in fissures. Chills are coated and many different chill coatings are used. The purpose of chill coatings is to avoid burn-in, while at the same time not insulating the molten metal from the chill. They are also instrumental in producing smooth, clean,

fine-grained surfaces. For heavy castings, like coal crusher rolls, zirconia wash is often applied, which may be topped by a graphite paint.

When split or several chills are applied, there is a risk of cracks at the parting lines, due to steep thermal gradients. The parting line zone of the casting is subject to hot cracking because it freezes later than the chilled section and, hence, is under a concentrated tensile stress during its solidification. If individual or spot chills are used to enforce a directional solidification, they should have rounded edges or even be circular to avoid stress concentrations. A distance of 20–40 mm (3/4–1-1/2 in.) should be kept between adjacent chills.

Ni-Hard 1 and 2 castings of simple shape without cores can be produced in chill or permanent molds. Casting Ni-Hard 1 or 2 against a chill or in chill molds gives a high cooling rate and provides a fine-grained structure. Chill casting also offers the possibility to utilize higher C and Si contents without risk of graphite formation, which significantly improves abrasion resistance. This is in contrast to high-Cr irons or Ni-Hard 4, where chilling does not provide any benefit but seems rather to have a detrimental effect on abrasion resistance. Permanent mold casting is of special benefit with cupola melting, where some of the disadvantages (with regard to chemical composition, especially the high C content) can be overcome.

The beneficial effect of chill on abrasion resistance and mechanical properties of Ni-Hard 1 and 2 is frequently used by casting the wear side of the part against a chill plate. If one side of the casting is cast against a chill, distortion can occur due to the different cooling rates and warpage has to be allowed for. For example, a simple rectilinear casting made in a horizontal position with the drag face chilled will be convex on the chilled face and concave on the opposite face. Counterwarping the pattern and the chill (if one-piece) will overcome this difficulty. Warpage is determined by the thermal gradients within the filled mold. It varies with pouring temperature, pouring speed, thickness of the casting and thickness of the chill. With these and other factors coming into play, it is obvious that warpage figures can only be determined for particular patterns and conditions.

The Ni-Hard cast irons do not contain graphite and, therefore, their shrinkage behavior at solidification is similar to steel.

Runners, ingates and risers should be generous and should promote a directional solidification. The solidification shrinkage depends on the composition and is between 4 and 5 vol.% and decreases with increasing C content. Pouring through the riser or ingates into or under the riser are often advantageous. The feeding distance for Ni-Hard is similar to or somewhat greater than found in the steel casting of similar section thickness and configuration. A first approximation is that the feeding distance be 4–7 times the riser diameter, but there is some influence of the melt quality. No fixed rule can be offered for the number, size, height or positioning of risers, as these conditions will be determined by the individual case. However, they should be kept away from the critical wear surfaces. The riser modulus can be estimated at 1.2 times the casting modulus of the section to be fed. Figures 7.2 and 7.3

show two examples of the feeding and ingating of high-performance castings for coal crushers.

Whenever practical, risers should be surrounded with exothermic sleeves or insulating refractories to increase their effectiveness and improve casting yield. When the mold is poured, the risers should be covered with some type of insulating or hot top material to keep them open to the action of atmospheric pressure. With large castings, it is feasible to feed liquid metal from a ladle through an open riser fairly late after pouring the casting.

The mass of a riser is intended to slow solidification, which involves the risk of graphite formation beneath it. The temperature differences between the riser and the thinner sections of the casting can give rise to severe thermal and transformation stresses.

In order to facilitate riser removal notching or breaker cores can be employed. A 60-degree notch gives a clean break-off.

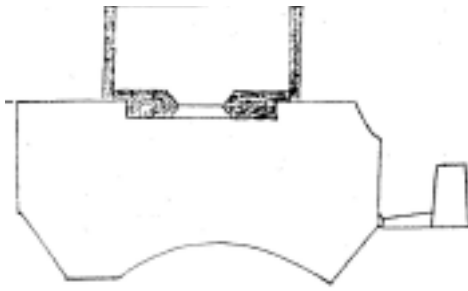
Shrinkage depends on the amount of retained austenite in the matrix because austenite has a higher shrinkage coefficient than martensite, pearlite or ferrite. This results in a peculiarity that must be kept in mind. The casting has an austenitic matrix after solidification, and shrinks with a factor close to 3% until the onset of the transformation of austenite to martensite. The transformation starts at about 250C (482F) for Ni-Hard 1. This transformation is accompanied by a volume increase so that the shrinkage stops or even turns into an expansion. The difference of the shrinkage curve between unalloyed, white pearlitic AR iron and Ni-Hard is shown in Fig. 7.4.

Temperature differences and associated differences in austenite transformation can create high stresses, in this way, which can either remain as internal stresses or cause the casting to rupture.

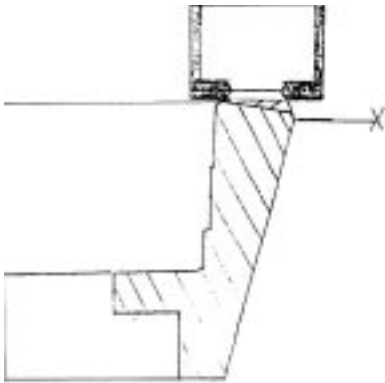
A casting hung by risers in the mold and restrained during shrinkage runs a risk of tearing. This risk can be avoided or minimized by a suitable positioning of the risers, by the use of a collapsible mold, or by loosening the sand around the risers after solidification. Especially, internal cores should be collapsible. The stresses are higher than in steel castings because AR irons are stronger, have little ductility and Ni-Hard irons shrink longer and to lower temperatures. In Fig. 7.4, the shrinkage of steel would be similar to pearlitic AR iron.

The cooling rate should be as slow as possible to below the critical range of martensite formation; that is, to below about 100C (212F). Small castings should be allowed to cool slowly in the sand mold to below 50C (122F). Large castings can be shaken out early, allowed to cool to dull red heat and then covered with insulating material to allow them to cool slowly through the critical range between 300 and 100C (572–212F) and almost to ambient. Sometimes, critical castings are transferred into a furnace for controlled cooling. These recommendations do not apply for chill-cast grinding media; for them, a hot shake-out and air cooling to below 400C (800F), followed by slow cooling, is recommended (INCO, 1949).

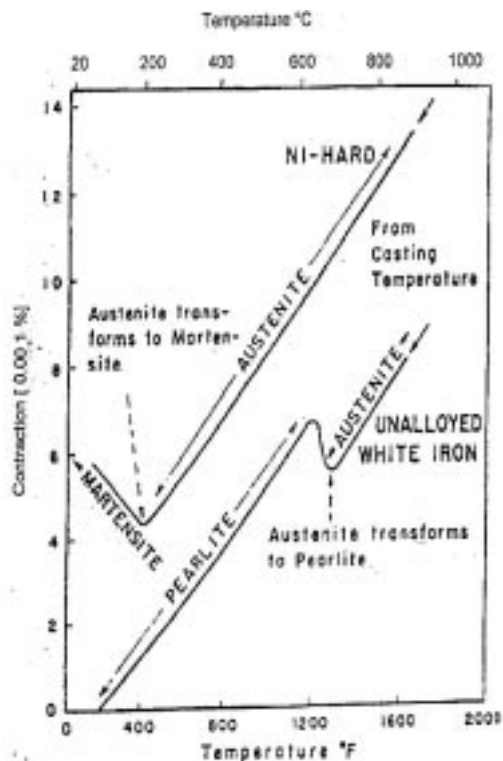
(Ni-Hards 1 and 2, continued)



**Fig. 7.2. Feeding and ingating for an E10 track ring of a coal crusher. It consists of twelve 280-mm (11-in.) dia feeders; 18 ingates 57x10 mm (2-1/4x3/8 in.) with 3-mm (1/8-in.) taper on each side; 75-64x100-mm (3-2-1/2x40-in.) runner bars and 6 downsprues 50 mm (2 in.) diameter.**



**Fig. 7.3. Feeding and ingating of a crusher roll with four 200-mm (8-in.) dia feeders with exothermic sleeves; 3 ingates 75x100 mm (2-1/2x3/8 in.) running in at the joint line; runner bars, fifty 38x50 mm (2-2-1/2x2 in.).**



**Fig. 7.4. Shrinkage curves of unalloyed, white pearlitic AR iron and Ni-Hard.**

**Bimetallic Ni-Hard Castings**—The idea to combine the high abrasion resistance of Ni-Hard with the toughness and machinability of steel or graphitic cast iron has led to the development of various techniques to produce bimetallic or compound castings.

Ni-Hard 1 and 2 are especially suited for bimetallic castings because Ni-Hard attains its abrasion resistance in the as-cast condition or after only a low-temperature heat treatment, thereby avoiding problems due the different transformation behavior and expansion coefficients of the two different materials.

A simple and efficient way is to reinforce Ni-Hard by means of cast-in steel bars or rods. The diameter of the rods is adapted to the section of the casting and they are placed well removed from the working face. The reinforcement extends both longitudinally and laterally in the casting and can, in some cases, have the form of a lattice or spiral arrangement.

Care is taken to avoid a metallic bond of the steel rods to the cast metal. This is done to avoid embrittlement by carburization of the steel, high stresses due to different expansion coefficients and defects due to reactions between melt and rust on the steel rod. Usually a refractory coating or a thin sand layer is used. Obviously, the reinforcement cannot completely avoid cracking of the Ni-Hard under severe dynamic stresses, but it prevents complete breakage and allows the casting to remain in service until it is worn out.

True bimetallic castings are produced by pouring two melts. The problem is to obtain a good bonding at the desired position in the casting without over mixing the two materials. This requires careful control of timing, temperatures, quantity, gating and pouring velocities. The technique has a long tradition in rollmaking, where rolling mill rolls, consisting of an AR iron working shell and a gray or ductile iron core and journals are common. The vertical cylindrical chill mold with sand-molded journals is first completely filled with Ni-Hard melt. The solidification starts from the chill, progressing toward the center of the body. After a specific time, only the center and the sand-molded journals are still liquid. Now the second melt is poured through the same gate at the bottom of the mold. Rising in the mold, it pushes the remaining Ni-Hard melt out through the top journal, until the core and the journals are filled with softer material.

This deceptively simple technique is obviously not applicable to castings of different and more complex shape. Here, more sophisticated methods have to be applied to achieve bonding without too much mixing. One way is to separate the two melts for a specific time interval by means of a sheet of steel or nickel in the mold cavity. Another is to employ two pouring systems.

Centrifugal casting is a technique well suited to the production of bimetallic Ni-Hard castings. It has been used extensively in rollmaking and in the production of other castings of symmetrical and cylindrical shape, such as various crusher rolls with an outer Ni-Hard working shell, or pipes where the Ni-Hard is on the inside. Provided that process parameters—

such as temperatures, timing and fluxes—are determined and allow a good bonding without too much mixing of the two materials, excellent results can be achieved. The same method can be applied with nodular cast iron or steel. Obviously, a fairly wide range of melt compositions and types can be combined, once the suitable parameters have been established.

## Pouring

Pouring temperatures should be as low as feasible with regard to the type and section of the casting. For heavy sections 1350C (2462F) is fully sufficient. Temperatures should not exceed 1450C (2642F), in order to obtain a fine-grained carbide structure. For low C content, the diagrams showing the variation of liquidus temperature in Chapter 2, Solidification (Fig. 2.8) can be consulted. Pouring should be fast to avoid the formation of an oxide skin on the melt surface. A pressurized gating system minimizes air aspiration and helps to avoid slag defects. A large pouring basin with stopper is helpful to keep the running system full and prevent slag and dross from entering the casting through the downsprue.

## Postcasting Processes

### Cleaning/Finishing

Risers and ingates are knocked off. Care must be taken to avoid breaking into the body of the casting. The as-cast Ni-Hards are not only harder and more brittle than steel or gray iron but also more sensitive to heat checking due to their austenite content. Risers should not be removed with a cutting torch because of the hazard of thermal cracking. Surface cleaning is done in the normal way. In the grinding of riser stumps, the temperature generated during grinding must be controlled. Excessive grinding pressures must be avoided in order not to overheat the surface and cause grinding cracks. Comparatively soft grinding wheels are normally used, often of the zirconia-alumina type with medium abrasive grain size. Complex castings may contain significant internal stresses. In critical cases grinding is best be done after stress relief heat treatment to minimize the risk of heat checking, grinding cracks or gross cracking.

### Heat Treating

Ni-Hard 1 and 2 castings are usually given a stress relief heat treatment at 225–275C (440–530F) for 4 to 12 hours, followed by furnace cooling. Other heat treatments are possible and have been described in Chapter 3, Heat Treatment.

### Welding

Like other AR irons, Ni-Hard 1 and 2 are not weldable. The brittle structure is very sensitive to heat checking so that cracks under the weldment are unavoidable. Fabrication and repair weldings are not possible, and attempts at weld repairing defective castings usually make the defect worse. They can, at best, be a cosmetic operation.

In some cases, cast-in low-C steel inserts are used to allow welding of the Ni-Hard castings to other parts of a fabricated assembly.

Hardfacing is possible, and is widely employed to recover worn crusher rolls. However, in this operation, the aim is not to produce a crack-free weld overlay. The hardfacing layer always contains numerous small cracks. These do not penetrate deeply into the underlying Ni-Hard casting. The hardfacing is done under carefully controlled conditions, where the casting is not allowed to heat up over a certain low temperature. Rolls, e.g., are in fact cooled during welding from the inside surface and are sometimes placed in a shallow water basin. (An example of hardfacing can be found in Chapter 10, Applications.)

### Machining

Ni-Hard castings can be machined, despite their high hardness. Due to its lower carbide content Ni-Hard 2 is easier to machine than is Ni-Hard 1. Soft annealing is not possible. The matrix structure cannot be sufficiently transformed to pearlite without risk of graphitization.

An obvious method of machining Ni-Hard is grinding. In order to avoid heat checking, wet grinding is preferred. On modern wet-grinding machines, large surfaces of Ni-Hard castings can be machined in a very economic way, with close tolerances.

An essential requirement for machining with cutting tools is to maintain maximum rigidity, both in the workpiece and in the cutting tool. Lathes and mills must have heavy head stocks that are robust and in good condition.

Sintered carbide tools can be used, but wear is rapid. Modern ceramic tools, especially cubic boron nitride (CBN), give much better results and allow machining by turning and milling without losing surface finish and accuracy due to tool wear. Machining pump casings and rollers for coal mills with CBN is an established practice today (Blackman, 1990; DeBeers). Table 7.1 contains suggested machining conditions for turning Ni-Hard castings. Typical data for machining 946-mm OD mud pump casings of Ni-Hard 2 with casting skin on vertical boring mill with CBN are as follows (DeBeers):

Surface speed of the workpiece . . . . 40 m/min  
Depth of cut . . . . . 1–1.5 mm  
Feed rate during roughing . . . . . 0.4 mm  
Feed rate during finishing . . . . . 0.25 mm

The risk of breaking the expensive CBN inserts on defects in casting surfaces is often thought to be too high. In such cases, less expensive and tougher CBN-containing-silicon nitride tools can be used, despite their lower wear life.

**Table 7.1**  
**Suggested Machining Conditions for**  
**Turning of Ni-Hard With 570–750 HBN with CBN (DeBeers)**

| Machining Operation | Surface Speed [m/min] | Maximum Depth of Cut [mm] | Feed Rate [mm/rev] |
|---------------------|-----------------------|---------------------------|--------------------|
| Finishing           | 40–80                 | 1.3                       | 0.0–0.15           |
| Light roughing      | 40–80                 | 2.0                       | 0.05–0.20          |
| Rough turning       | 40–80                 | 3.2                       | 0.05–0.40          |

Drilling holes and tapping is also possible with CBN tools, provided the diameter is large enough. For small holes, often cast-in mild steel inserts are employed. The inserts must be clean and dry in order to avoid formation of defects by reactions with the melt. Mild steel inserts should be well away from the corners and edges, to avoid cracks due to the different thermal expansion coefficients or during machining the insert. The inserts must also be located sufficiently far away from the wear surface so that they cannot become exposed after some time in service. Since they should have a metallic bond to the surrounding metal, they must be absolutely clean and dry when placed into the mold.

In many cases, holes and even threads can be cast-in by suitable cores with sufficient accuracy. In such precast holes, steel inserts can be screwed-in and, later, tapped. The remaining Ni-Hard section must be thick enough to avoid bursting when the steel insert is put in.

Small holes can also be cast-in by means of quartz glass tubes, available in lab supply shops.

## Ni-Hard 4 AR Cast Irons

Ni-Hard 4 is a 9Cr-6Ni-2Si AR cast iron. It is designed to give a structure with rod- or blade-like eutectic  $M_7C_3$  carbides and a matrix that, in the as-cast condition, is free from pearlite. It is predominantly austenitic, even in very heavy-section castings. After heat treatment, it is predominantly martensitic.

Details of the microstructures have been discussed in Chapters 2, 3 and 5 on Solidification, Heat Treatment and Morphology, respectively. The desired carbide structure is a result of the joint effect of the Cr, Ni and Si contents at a eutectic or slightly hypoeutectic C content. This means that control of the chemical composition is essential for high-performance castings. Besides their effect on the carbide type, the elements have important additional effects on structure and properties.

*Bimetallic Castings*—Bimetallic castings can be produced in the same way as described for Ni-Hard 1 and 2, but are rather rare. The heat treatment of two different materials is difficult and Ni-Hard 4 in the as-cast condition offers no advantages over Ni-Hard 1 or 2.

## Chemical Composition

Nickel is added to ensure a pearlite-free matrix in the as-cast condition. The exact Ni content depends primarily on the section and type of casting, but a minimum of about 5% Ni should be maintained to avoid pearlite formation during slow cooling in the mold or during heat treatment. Higher Ni content than 6.5% will stabilize the austenite too much, and if a higher hardenability is thought necessary, an addition of Mo is a better solution than a further increase of the Ni content.

Silicon is an essential element to ensure the correct carbide structure. The minimum Si content can be lowered if a higher Cr content combined with a lower C content is used. Contrarily, the Si content must be increased if higher C compositions are used. However, raising the Si content above 2% involves the risk of graphite precipitation in heavy sections, especially if the Cr content is on the low side and the C content is on the high side of the range. Therefore, it is recommended to aim for an Si content of between 1.8 and 2.0%. Another effect of Si is that it favors the transformation of austenite to martensite, thereby increasing hardness.

The effect of Si on graphite formation has to be compensated for by a sufficient amount of Cr, and Cr also increases hardenability, even if the major part is concentrated in the carbide phase. However, increasing the Cr content above 10% offers no advantage but will make it more difficult to transform the as-cast austenite to martensite.

The C content of Ni-Hard 4 gives a eutectic or slightly hypoeutectic composition. The C content determines the volume fraction of eutectic carbides, which is about 20% at 2.5% C and 30% at 3.5% C, respectively. Usually the C content is kept between 2.9 and 3.2% as a reasonable compromise between demands for abrasion resistance and toughness. At higher C content, there is also a risk of graphite formation in heavy sections, if the Si content is high.

Manganese increases hardenability but, even more, increases austenite stability and should, therefore, be kept in the recommended range of 0.4–0.6%. Alloys with higher Mn contents have been proposed several times but their structures are extremely difficult to control.

Molybdenum will improve hardenability without undue stabilization of the austenite; therefore, 0.5–1.5% Mo can be added to increase the hardenability in heavy and very slowly cooled castings. This is preferred to raising the Ni content above 6.5%.

Copper has been proposed as substitute for Ni, but results have been unsatisfactory.

Sulfur and phosphorus are detrimental to the toughness of Ni-Hard 4 and should be kept to the lowest feasible level. Sulfur has been shown to reduce the abrasion resistance.

## Melting

Ni-Hard 4 is best melted in electric furnaces. Cupola melting does not allow the required close control of the chemical composition. Cupola melting plus duplexing has been done successfully, however. Rotary furnaces also pose the problem of controlling composition and temperatures.

Electric melting practice of Ni-Hard 4 is almost identical to the practice described in detail for Ni-Hard 1 and 2.

The charge materials consist of Ni-Hard returns, which can be either Ni-Hard 4 or Ni-Hard 1 and 2, steel scrap and,

eventually, pig iron with low S and P contents. Stainless steel scrap can be used to introduce Ni and Cr instead of ferroalloys or metallic Ni. Silicon can be increased by FeSi additions. The same precautions as described for Ni-Hard 1 and 2 apply.

Little advantage has been found in Mg treatment or inoculation of Ni-Hard 4. In any case, inoculation cannot improve a poor carbide structure brought about by a previous excessive overheating of the melt.

Tapping temperature is about 1450C (2642F). As can be seen in Fig. 2.17 in Chapter 2 on Solidification, Ni-Hard 4 has about the same liquidus temperature as Ni-Hard 1 and 2.

### **Molding, Gating and Riser**

All sand and binder systems used for gray and ductile cast irons or steel castings can be used to make molds and cores for Ni-Hard 4.

The patternmaker's shrinkage rule is between 1.5% and 2.5%. Typical values are 1.80% for restricted and 2.25% for unrestricted shrinkage. When the castings are heat treated, they undergo a change of matrix structure from predominantly austenitic to predominantly martensitic. This transformation accounts for a positive volume change. This has to be considered in the shrinkage allowance of the finished casting.

Molding, gating and riser methods for Ni-Hard 4 are the same as for Ni-Hard 1 and 2.

While chills can be used in the same way as for Ni-Hard 1 and 2, to ensure directional solidification, chill or permanent mold casting is not useful for Ni-Hard 4. The grain-refining effect does not improve properties or abrasion resistance.

The feeding distance of risers for Ni-Hard 4 is often quoted to be 10 times the riser diameter, but this is not always the case. In some instances, much shorter feeding distances are needed and feeding problems have been observed, which seem vaguely related to metal quality, pouring temperature and riser height.

Heavy risers can create significant stresses in relatively thinner sections, because the slow cooling riser will form more martensite than the thinner section and thus expand more. This can then crack the thin section of the casting.

### **Pouring**

No special information.

### **Postcasting Processes**

#### ***Cleaning/Finishing***

For riser removal, cleaning and finishing, the same procedures apply as for Ni-Hard 1 and 2.

#### ***Heat Treating***

Ni-Hard 4 castings are almost always given a hardening heat treatment consisting of soaking at 780–850C (1436–1562F),

followed by furnace cool and a stress relief. Details are described in Chapters 3 and 4 on Heat Treatment and Hardenability, respectively.

The hardening mechanism is a destabilization of the austenite present in the as-cast condition which allows a transformation to martensite during slow cooling below about 500C (932F). This is completely different from the hardening treatment for steels or high-Cr irons, where a rapid cooling or quench is necessary to avoid a transformation to pearlite or bainite at the expense of martensite. If a Ni-Hard 4 casting has an insufficient hardness, it is usually due to high retained austenite, not pearlite or bainite. Faster cooling will reduce the hardness by maintaining more austenite. A hardness increase is obtained by slower cooling below 500C (932F). Cooling is, therefore, done in a furnace, either immediately from the destabilization temperature or after still air cool to about 600C (1112F).

#### ***Welding***

Like the other AR irons, Ni-Hard is not weldable without cracks occurring under the weldment and in the heat-affected zone. Repair weldings are, therefore, not possible and attempts at weld repairing defective castings should not be undertaken. They can, at best, be a cosmetic operation.

Hardfacing is possible and widely used to recover worn crusher rolls. Even new rolls are hardfaced, in order to extend service life. The Ni-Hard 4 roller serves as backup, if and when the weld overlay is worn. A steel or ductile iron would be destroyed in a very short time, forcing an unplanned plant shutdown. The weld overlay is not intended to be crack free. The hardfacing layer always contains numerous small cracks; however, they do not penetrate into the underlying casting. Welding is done under carefully controlled conditions where the casting is not allowed to exceed a certain low temperature. During welding, rolls must be cooled from the inside surface and, sometimes, be cooled in a water basin. (An example of hardfacing can be found in Chapter 10, Applications.)

#### ***Machining***

Ni-Hard 4 is machined in the same way as Ni-Hard 1 and 2, especially with cubic boron nitride (CBN).

## **High-Chromium AR Cast Irons**

High-Cr irons cover a range of compositions between 12 and over 40% Cr and 1–4% C, plus several additional elements like Mo, Ni, Cu, V, etc. This range covers strongly hypoeutectic alloys, which are often regarded as steels, as well as eutectic and hypereutectic irons. This wide composition range together with various possible heat treatments allows one to optimize them for different applications.

*Bimetallic Castings*—Bimetallic castings can be made in the same way as Ni-Hard 1 or 2. Heat treatment of the bimetallic castings creates problems, so the high-Cr iron is usually used as-cast and stress relieved with a predominantly austenitic

(High-Chromium, continued)

matrix. In the steel industry, bimetallic or duplex high-Cr castings are quite common, and extensive research and development has enabled many of these roll manufacturers to develop heat treatments that produce predominantly martensitic matrices.

## Chemical Composition

The AR grades are usually in the range of 12–26% Cr and 2–3.5% C. Their contents of C, Cr and other alloys, like Ni, Cu, V, etc., are selected, based on the application and previous experience. The basic factors for the composition are the requirements for toughness, hardenability, heat treating response and alloy cost.

Basic hardenability is provided by the Cr available in the matrix, which is defined by the C:Cr ratio. Additional hardenability is provided by Mo, Ni and or Cu. Manganese promotes hardenability at low cost but makes it more difficult to control hardness and retained austenite in castings of varying sections.

The low-C alloys with Cr content above about 28% are used mainly for applications where attack by corrosion and abrasion is combined. Their composition leads to higher liquidus and solidus temperatures and, therefore, requires higher melting and casting temperatures. In general, they are more difficult to melt and process than the more common AR grades.

Most low-C irons with over 28% Cr are ferritic as-cast and cannot be hardened to a fully martensitic matrix. They are put in service after stress relief. At very high Cr levels, a special heat treatment is often necessary to eliminate the sigma ( $\sigma$ ) phase. Additions of Ni and or Cu can be made to obtain austenitic or austenitic-ferritic structures, as-cast or after heat treatment.

## Melting

The high-Cr AR irons are readily melted in induction or arc furnaces. Most production is in acid-lined furnaces, but basic linings and neutral linings may be used. The cupola is unsuitable because of the high and uncontrollable carbon pickup due to the high-Cr content. Rotary furnaces pose problems with temperature control, slag formation, high-Cr loss of 10–25% and excessive lining wear.

Since there is little or no Ni-free high-Cr steel scrap available, the Cr AR irons are generally produced using foundry returns and virgin materials. Limiting furnace charges to materials of known composition and quality is a great advantage in producing a product consistent in composition and quality. The high-Cr AR irons are not suitable for holding in the molten state for long periods of time, because the reactivity of the Cr results in excessive increase in oxygen and hydrogen contents. Here, modern high-powered medium-frequency induction melting with computer control is a distinct advantage.

## Induction Furnace Melting

Suggested raw materials and air induction melt practice are as follows (Spear, 1990):

### Charge Materials

|                                             |                                      |
|---------------------------------------------|--------------------------------------|
| Low-C steel scrap (BOF* steel, if possible) |                                      |
| Cr AR iron returns                          | High-C FeMn                          |
| Low-Si pig iron                             | FeMo                                 |
| Low-Si charge Cr                            | Foundry grade 75% FeSi               |
| Low-C FeCr                                  | Ni or FeNi, Cu and other ferroalloys |

### \*Basic Oxygen Furnace

The quality of high-C ferroalloys is generally excellent with regard to cleanliness, low gas content (O and H) and the absence of significant amounts of unidentified and undesirable elements. A large late addition of high-C FeCr to an induction furnace melt acts as a powerful diluent and deoxidizer.

Molybdenum is usually added as FeMo, but moly oxide briquettes may be used if they are included in the solid charge and kept submerged in the melt.

Silicon content should be aimed at 0.7%. Less than 0.4% Si in the melt can cause difficulties with viscous slag; higher Si content can promote pearlite during heat treatment. In 26-Cr irons, higher Si improves fluidity and feeding distance and results in grain refining, with improved resistance to shock and higher bending strength. Above 1% Si, the iron becomes embrittled and the matrix structure may become partially ferritic.

The Mn content may range from 0.5 to 1.5%. Manganese on the high side of this range can erode acid furnace linings. Manganese content should be limited to 1% in meltdown and the remainder can be added to the ladle as crushed ferroalloy.

### Melting and Charging Practice

1. Charge in order:
  - pig iron
  - steel scrap
  - Cr AR iron returns
  - Mo, Ni and/or Cu
2. Melt down, superheat to 1480C (2696F) for normal C and Cr levels. For low-C irons with high-Cr content of 28% and above, temperatures over 1600C (2912F) are required to ensure solution of the FeCr
3. Add low Si charge Cr in small enough batches so that additions sink into the melt without freezing over the surface metal.
4. As soon as all the Cr has melted, heat to minimum superheat temperature required for handling and pouring.
5. Add high-carbon FeMn to the furnace.
6. Tap immediately onto makeup Si as foundry grade Fe75Si in the bottom of a hot clean ladle.
7. Pour castings at minimum practical temperature.

The melting practice is based on charging and melting the least oxidizable materials first, then progressively charging the more oxidizable materials, and finishing up with the most oxidizable (i.e., FeSi). Time and temperature are important factors in induction melting. Minimum melting and holding times reduce the amount of O taken into solution from air, slag and refractories, and H taken from moisture in the air.

Refining possibilities are limited in the induction melting of high-Cr AR irons. Acid refractories are reactive with the high-Cr melt; so, when melting on a silica lining, the metal should be superheated to the minimum superheat temperature required for handling and pouring, and then it should be immediately tapped.

Increasing oxygen in the metal is progressively more detrimental to castability and, probably, impact abrasion resistance. Scrap should never be preheated to point where it starts to form a scaled surface layer. In fact, some foundries blast clean returns and scrap, to avoid introducing oxygen into the melt.

Hydrogen, in addition to causing gas problems, can be responsible for spontaneous cracking of castings and erratic breakage in service. There is not much information available on the effect of increasing nitrogen content but, judging from limited experience, nitrogen seems to be beneficial. Content of 600–1100 ppm seem to be acceptable or even of an advantage.

An air induction furnace melt can be protected to varying degrees by an inert gas cover. This would limit oxygen, hydrogen and nitrogen pickup due to exposure of the melt to air. An Ar cover is quite effective, even in an open furnace when the melt meniscus is low in the furnace. Argon is denser than air, so a low-velocity stream of Ar will effectively displace the air in the furnace. As the melt meniscus rises in the furnace, however, both air entrainment and convection progressively reduce the effectiveness of the inert gas cover. A close fitting furnace cover will greatly increase the effectiveness of the Ar protection.

In induction melting, wet, oily or rusty materials should never be charged onto a heel of molten metal in the furnace. Wet, oily or rusty materials can be used in the initial charge to an *empty* furnace, because the volatiles will be eliminated long before a molten pool develops.

A gradual pickup of hydrogen will eventually result in subsurface pinholes that appear after blast cleaning castings. If the melt practice is not changed to take care of the pinholing at this point, large gas holes in the castings will follow. Hydrogen also appears to be a major cause of catastrophic cracking during processing castings in the foundry and erratic breakage of castings in service. Long, expensive heat-treatment cycles are sometimes used to diffuse hydrogen out of the castings and insure against hydrogen embrittlement.

### **Arc Furnace Melting**

Arc furnace melting follows similar rules but allows a refining of the melt.

## **Molding, Gating and Riser**

All sand and binder systems used for gray and ductile cast irons or steel castings can be utilized to make molds and cores for high-Cr irons. Mold washes should be non-carbonaceous because a carbon pickup can cause a hard, brittle skin on the casting surface. Zirconia washes usually give good results.

The patternmaker's allowance is indicated at an average of about 2% but varies in practice between 1.3 and 2.5%. The value depends, to a large extent, on the final matrix structure, which can be pearlitic, austenitic, ferritic or a mixture of all, depending on grade, composition and section size. An as-cast, austenitic structure has a higher shrinkage rate; a pearlitic structure has a smaller shrinkage rate. Add to this the effect of the casting geometry, and restricted and unrestricted shrinkage values are tacked on. Typical values for heavy-section 18% Cr iron are 1.5% for restricted and 2.0% for unrestricted shrinkage.

Castings that are heat treated will undergo a change of matrix structure to become predominantly martensitic, which is associated with a volume change. This also has to be considered in the shrinkage allowance of the finished casting.

Riser and feeding for the high-Cr irons follows the same rules as for the Ni-Hards. However, runner design and casting practice should take into account the viscous oxide film that tends to form on the surface of the molten metal. Therefore, adequate provision should be made in the system for trapping oxide dross, and the gating should provide rapid filling of the mold with a minimum of turbulence.

Castings can contain high internal stresses, which can result in cracking in the mold or during or after shakeout. This is especially the case for mixed matrix structures of austenite, martensite and pearlite, while fully pearlitic as-cast structures are much less risky. There are various ways to avoid cracking problems; i.e., long cooling times in the sand mold; or hot shakeout and transfer to a furnace or to an insulated slow cooling chamber; or covering the castings with insulating material.

While chills can be used in same way as for the Ni-Hards or steel to ensure directional solidification, chilled or permanent mold casting does not provide special benefits for high-Cr irons. The grain-refining effect does not improve the abrasion resistance and, in fact, may lower the iron's abrasion resistance (see Chapter 8, Abrasive and Repetitive-Impact Wear Performance).

The feeding distance for high-Cr irons depends on the chemical composition, and varies in a fairly wide range. The rules for steel casting of similar section thickness and configuration apply to a first approximation for high-Cr iron. However, even with a similar chemical composition, feeding distances can be inconsistent. Sometimes they are fairly long and, in other cases, on the same casting, very short. It seems that, especially, the 26% Cr iron can precipitate very long primary austenite dendrites, which choke the feeding channels. Experiments (Raynaud and Parent-Simonin, 1985 and

(High-Chromium, continued)

1986) with a 26% Cr iron. compared to unalloyed AR iron, in green sand molds showed the 26% Cr iron to have a much shorter feeding distance. Additions of up to 1% Si and 0.02% Ti improved the feeding distance, somewhat.

## Pouring

Low pouring temperatures reduce the incidence of metal penetration and burned-on sand and bring about a certain grain refinement. High pouring temperatures aggravate shrinkage under feeder heads and in other hot spots and can lead to microshrinkage and coarse dendritic structures.

Due to its surface oxide film, the melt appears very sluggish and viscous but is, in fact, quite fluid and able to fill thin sections. Pouring temperatures are seldom higher than 100°C (200°F) above the liquidus temperature. Liquidus temperatures are primarily determined by the C content and are increased by higher Cr, as discussed in Chapter 2 on Solidification.

Typical recommended pouring temperatures are:

|                     |                         |
|---------------------|-------------------------|
| 15% Cr–3% C grades: | 1300–1400C (2400–2550F) |
| 23–28% Cr grades :  | 1400–1500C (2550–2750F) |
| 30–35% Cr grades :  | 1450–1400C (2650–2830F) |

The oxide skin on the melt should be carefully skimmed off, or a bottom pouring ladle can be used.

## Postcasting Processes

### Cleaning/Finishing

For riser removal, cleaning and finishing, the same procedures apply as for the Ni-Hards. It is often useful to remove the risers or to grind the stumps after the heat treatment, despite the higher hardness, or at least after a stress relief in order to minimize the risk of grinding cracks. The precaution is especially true if the castings are fully or predominantly austenitic, as cast. Some castings may contain high internal stresses. If these become unbalanced during cleaning or machining, the casting can crack or even explode. Cracking can also happen if the untreated castings are stored outside in cold weather, or even during longtime storage inside.

Experiments have demonstrated that as-cast stresses of 700 N/mm<sup>2</sup> (100 ksi) were reduced to below 300 N/mm<sup>2</sup> (44 ksi) by holding the casting at 250C (480F) for eight hours and then cooling in the furnace.

### Heat Treating

High-chromium AR irons can be subject to a wide variety of heat treatments to fulfill the demands of various applications in a cost-effective way. This flexibility is one of the major advantages of this group of alloys. Heat treatments can be a stress relief for castings put in service with an as-cast structure; a hardening and temper heat treatment to obtain a martensite structure; a special heat treatment to minimize retained austenite for repeated impact loading; a hyperquench

to obtain a predominantly austenitic matrix; or soft-annealing to produce a fully pearlitic matrix. Details are described in Chapters 3 and 4 on Heat Treatment and Hardenability. For most heat treatments, the best structure to start with is a pearlitic matrix, which gives the most consistent and controllable results.

### Welding

Like the other AR irons, Cr irons, in general, are not weldable without cracks occurring under the weldment and in the heat affected zone. Repair or fabrication welding is, therefore, rarely possible. Successful welding has been done on low-C grades with high preheat. Welding is also possible in the austenite region, which implies temperatures above 850C (1560F). Some of the Ni- or Cu-containing grades have a region around 500C (932F), where the austenite does not transform for a long time after cooling from above 950C (1742F). These grades of castings have been welded in this interval using either FeNi, stainless steel or matching consumables.

Hardfacing is possible in a similar way as described for welding Ni-Hard 4.

### Machining

High-chromium AR iron castings can be machined, like the Ni-Hards, with CBN tools or by grinding. In contrast to the Ni-Hards, they can also be soft-annealed and rehardened after machining.

The aim of the soft annealing is to produce a fully pearlitic matrix. The hardness of the casting will then be less than 400 DPH. This transformation to pearlite is comparatively easy in alloys containing less than about 20% Cr and no significant amounts of Ni or Cu. Here, slow cooling from 950C (1742F) to below 500C (932F) is usually sufficient. Higher Cr and Ni or Cu contents stabilize the austenite and require longer and more elaborate heat treating cycles. (See Chapter 4 on Hardenability for details.) Low-C, soft-annealed high-Cr irons can be drilled and tapped.

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# Chapter 8

## Abrasive and Repetitive-Impact Wear Performance

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A common problem of many end-users dealing with abrasion-resistant and repetitive-impact irons is how to choose the right iron for the right application. For example, in structural, civil, transportation, marine and aeronautical fields, this problem is often solved by detailed specifications. Generally, in the case of abrasive and repetitive-impact wear, there are no detailed specifications.

It is impossible to create well-defined specifications that work well in all wear environments. The problem lies in the nature of the beast. Wear is a cacophony of competing and synergistic destructive mechanisms. In any one application, the wear process may be subjected to galling (metal-to-metal wear); corrosion; subsurface impact damage; surface microcracking (small bits of damage that coalesce into large cracks, leading to material loss); gouging (like a farmer tilling a new field, leaving larger furrows in his path); spalling (loss of large chunks of material, due to subsurface crack propagation); erosion (small chunks on the surface are removed by small abrasive particles playing a version of tribological touch and go); and, under some conditions, hydrogen embrittlement.

Testing for wear performance is a very individual thing. One test is not comparable to another. For example, the results of the “wear shoe on track” test cannot be compared to the results of the “Boyce/Avery” test, even though they are similar.

The study of wear goes back centuries. The ancient philosopher Lucretius (95-55 B.C.), is quoted as saying (Archard, 1980):

*“.....a ring is worn thin next to the finger with continual rubbing. Dripping water hollows a stone, a curved ploughshare, iron though it is, dwindles imperceptibly in the furrow. We see the cobble stones of the highway worn by the feet of many wayfarers. The bronze statues by the city gates show their right hands worn thin by the touch of all travellers who have greeted them in passing. We see that all these are being diminished since they are worn away. But to perceive what particles drop off at any particular time is a power grudged to us by our ungenerous sense of sight.”*

Since Lucretius’ time, the situation has improved only slightly. Although we can measure what is measurable and

quantify what is quantifiable, our ability to make accurate wear predictions based on this current understanding is rather pitiful. A contributing factor to this miasma is our dependence on laboratory tests that show only limited correlation to field wear processes. For a number of reasons, the bulk of wear research has concentrated on the development of laboratory tests, with little thought for eventual field correlation.

Many reasons for this concentration on laboratory work are given but, in general, laboratory work is more predictable, allows definitive conclusions to be generated and incurs little inconvenience to the people conducting the research. On the other side of the coin, field trials are very expensive, produce unpredictable results and can result in no publishable work if the industrial cooperator disagrees with the conclusions or deems the work too proprietary. This sad state of affairs leads to a ratio of one good “field trial” paper to roughly one hundred “laboratory wear” papers. Consequently, we are still looking for a good way to correlate the microstructural and mechanical properties of wear irons toward specific field applications.

In the meantime, users are still faced with the question of matching the right iron to the right application. In most cases, we are interested in macroscopic wear processes that remove centimeters, not nanometers of material. Here lies the true economic benefit of AR cast irons: superior abrasion resistance, whether near the surface or several centimeters deep within the casting. This is the special and particular niche that AR cast irons dominate and hold.

This chapter provides the tools to effectively match AR cast irons to a particular abrasive and repetitive-impact wear environment. We explore the correlation between wear tests and field wear data showing that, in general, abrasive and repetitive-impact wear tests provide only a qualitative look at the wear characteristics of the chosen materials, no matter how precisely they are conducted or how precisely the data is reported. Nevertheless, considering the nature of most industrial wear processes (according to one source, a tribological system may consist of over 80 factors), a qualitative assessment is perfectly adequate and allows one to quickly weed out materials of little benefit. Moving forward from this viewpoint, a few observations are presented on the interpretation of wear data and its application toward field environments.

Abrasive Wear Classification

Abrasive wear is well studied, and a number of classification schemes are available. Outside of the realm of wear specialists, very few people have heard of these schemes or care about whether their particular wear process is classified as two-body or three-body abrasive wear, or if their application is further classed as low-stress or high-stress. What is important to the typical user is a simple determination of the dominant wear process; i.e., abrasion, impact, fracture, corrosion, or corrosion/abrasion.

Very simply, abrasion is a form of wear that is mediated by hard particles. Moreover, these abrasive particles act to concentrate stress leading to severe plastic deformation at relatively low loads. Industrially, this process is shown in Fig. 8.1. The basic comminution principal is one of hard rock against hard iron. Although the hard iron is mostly victorious, the ploughing, cutting, fatigue and cracking action of the rock takes its toll. These abrasive processes are shown schematically in Fig. 8.2. It is these processes that define abrasive wear, i.e., the forcible removal of one material by another material.

Many laboratory tests have been devised to evaluate abrasion-resistant materials. Figure 8.3 shows just a sampling of these tests. Several of these tests attempt to simulate a combination of impact and abrasion (#1), abrasion and corrosion (#6 and #7) and microimpact and abrasion (#8). For our purposes, we will stick with a small array of these tests as approximately given in Fig. 8.4. Other definitions for abrasive wear are provided in Table 8.1.

Hardness and Abrasion Resistance

In order to cause ploughing, cutting or scratching associated with severe abrasion, the abrasive must be at least as hard as the metal. The schematic relationship between abrasive hardness, metal hardness and wear loss is shown in Fig. 8.5. It was first established by Wahl (1951) and has been confirmed in many laboratory and industrial tests.

As long as the abrasive is softer than the metal, wear is very low and mainly due to removal of corroded or oxidized surface layers, or to fatigue phenomena. The transition from low-level wear to high-level wear takes place at ratio of metal to abrasive hardness of about 0.8 for homogeneous metals, and spreads to 0.4–1.0 for metals with heterogeneous structures of phases of different hardness, like white irons (Wahl, 1951, Richardson, 1968).

In low-level as well as in high-level wear, harder metals suffer less abrasive loss, but the largest gain at relatively modest hardness increases can be achieved in the transition range, especially if the abrasion is moved into the low-level condition. Only a few minerals of technical interest have a hardness in the 40–200 HV range, like gypsum, high-purity limestone and low-ash coal, and are soft enough to cause low-level abrasion. Most minerals contain quartz or silicates with a hardness of 600–1250 HV and are in the transition range for white irons and hardened steels. Alumina, with a hardness of 1900 HV, and silicon carbide, with 2500–2800 HV, are much harder than even the carbides and cause high-level wear.

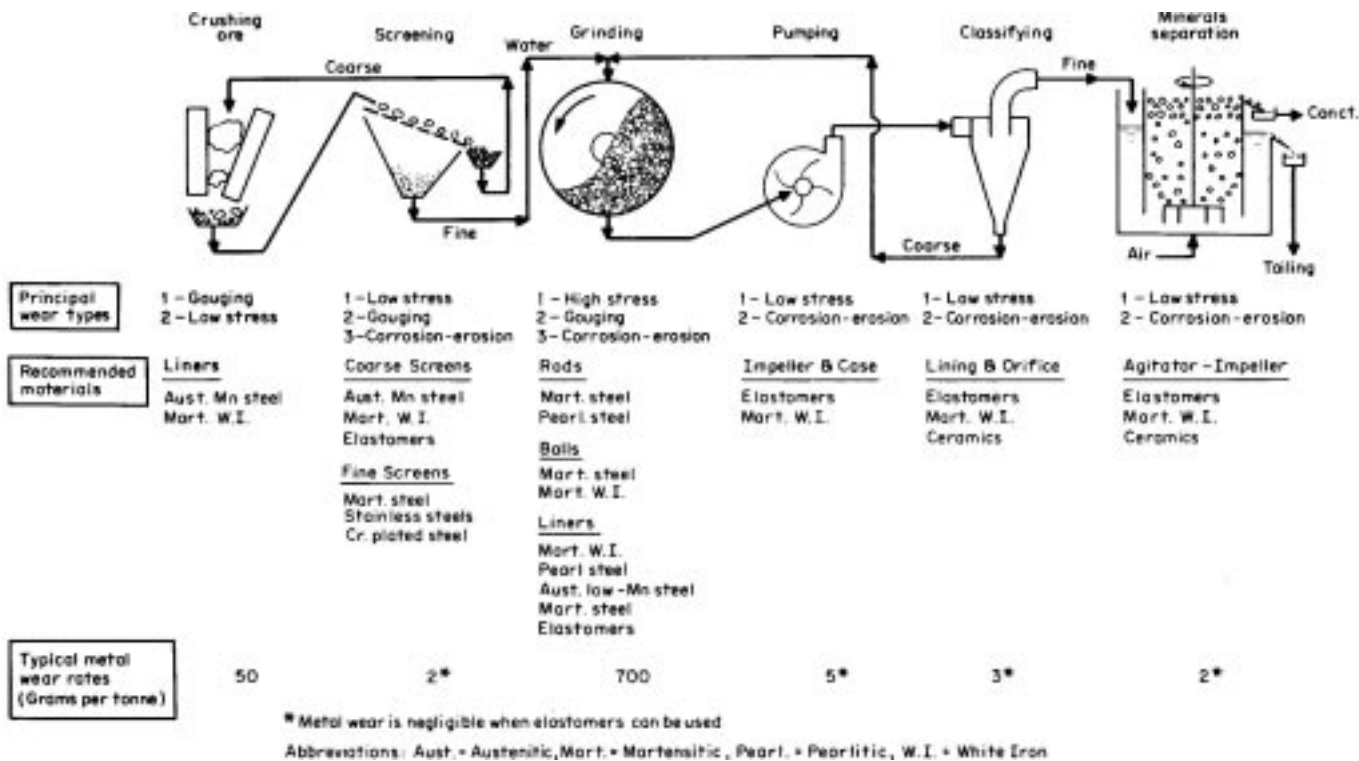


Fig. 8.1. Schematic of ore processing flowchart showing principal wear types, recommended materials and typical wear rates. (After Norman, 1980)

The effect of a change in the metal-to-abrasive hardness ratio can sometimes be quite dramatic. In grinding cement clinkers with a hardness of 450–550 HV, abrasive loss was reduced from about 300 g/t to 20 g/t when forged pearlitic steel balls with a typical hardness of around 400 HV were replaced by martensitic high-Cr iron with a hardness of above 650 HV (Röhrig, 1974).

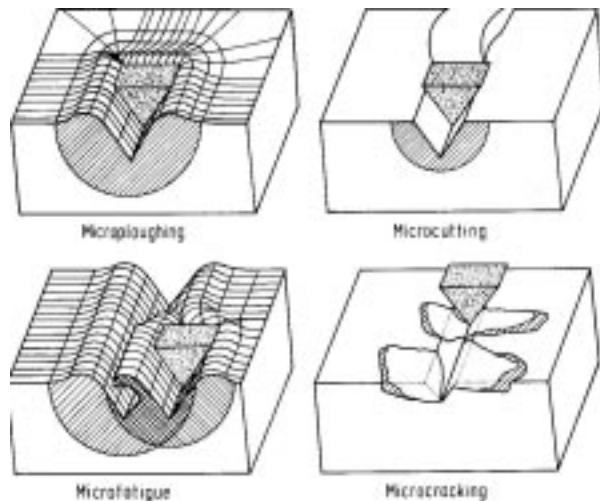


Fig. 8.2. Examples of abrasive wear processes. (After Zum Gahr, 1987)

In wet grinding quartzitic molybdenum ore, however, which has a hardness of around 1000 HV, a similar change of grinding ball alloy resulted only in a decrease of relative ball consumption from 114 to 82 (Norman and Hall, 1969). This is a very modest improvement, even when considering the fact that wet grinding causes much higher wear rates than dry grinding. Impact conditions are similar in both applications or even more severe in cement grinding. Changing to tungsten carbide balls for the experiment reduced the relative wear rate in the ore mill to 22.

Similar observations have been made in many other industrial applications, and usually the spread between a material with high- and low-abrasion resistance narrows when the abrasive conditions become more severe, due to higher mineral hardness, increased impact or the presence of water.

The obvious conclusion from Fig. 8.5, that abrasion resistance is function of hardness, is not valid in a general sense. The most well-known example is austenitic manganese steel, which, despite its low hardness, shows excellent performance in many applications. Especially in the transition range of Fig. 8.6, where most industrial abrasion phenomena take place, other properties—like work-hardenability, ductility, resistance against crack initiation and propagation during the detachment of wear debris from the surface, chemical composition, and microstructure—are of importance. It is quite

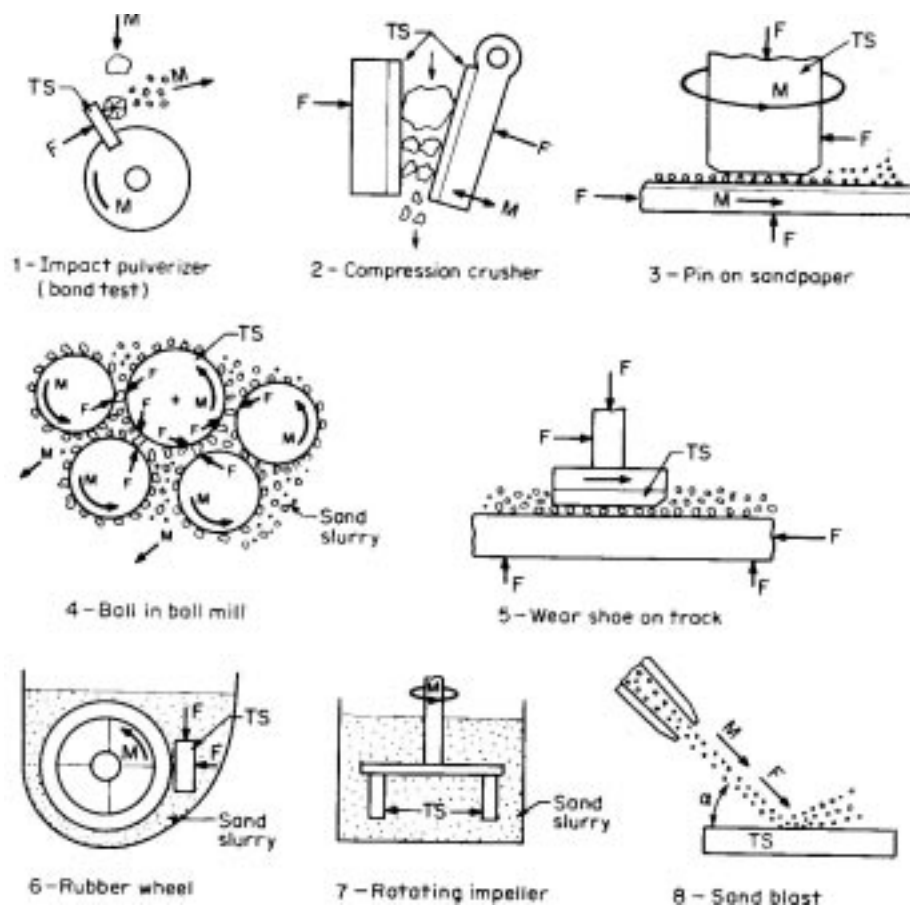


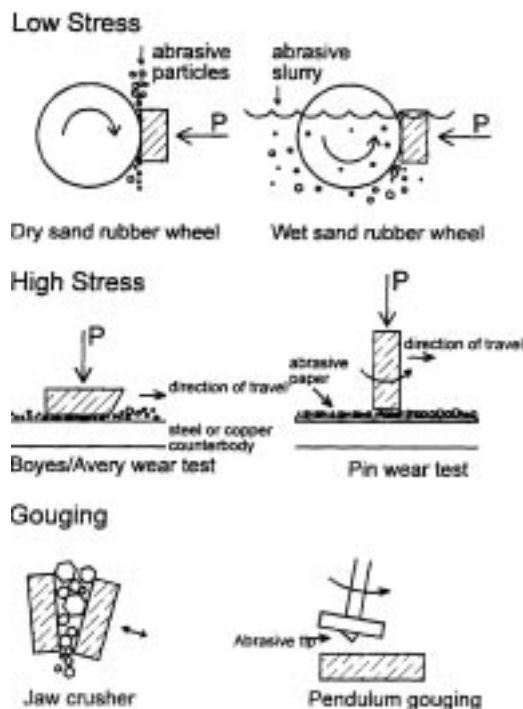
Fig. 8.3. Laboratory abrasion tests to simulate abrasive wear processes. *M* = motion, *F* = applied force, *TS* = test specimen. (After Norman, 1980)

**Table 8.1**  
**Abrasive Wear Classification<sup>(1)</sup>**

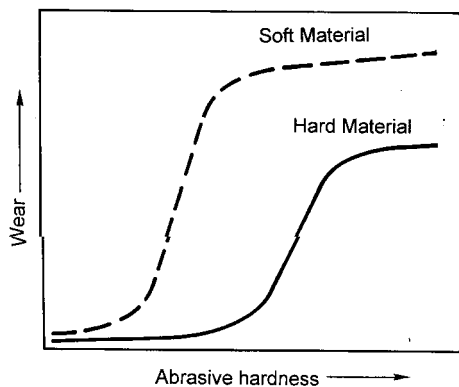
| Wear Mode           | Situation                                              | Manifestation                         |
|---------------------|--------------------------------------------------------|---------------------------------------|
| Impact-abrasion     | Repeated impact + abrasion                             | (not well characterized)              |
| Gouging abrasion    | Large abrasive particles,<br>extreme stresses          | Deep grooves,<br>gross deformation    |
| Grinding abrasion   | Stresses sufficient to fracture abrasive particles     | High wear rates,<br>visible scratches |
| Low-stress abrasion | Stresses not sufficient to fracture abrasive particles | Low wear rates,<br>fine scratches     |
| Corrosion-abrasion  | Corrosive + abrasive                                   | Varies;<br>accelerated corrosion      |

(1) Gates and Gore (1995)

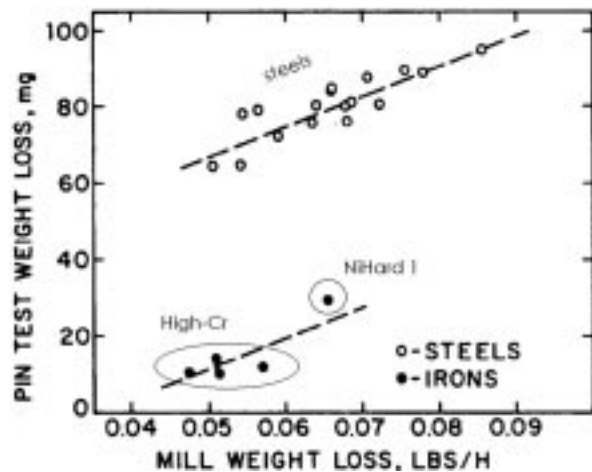
The bottom row represents the corrosive-abrasive phenomenon discussed in Chapter 9, Corrosive-Abrasive Wear Performance.



**Fig. 8.4. Wear terms (low stress, high stress, gouging) used to define laboratory abrasive wear tests. (After Tabrett, 1997)**



**Fig. 8.5. Schematic relation between abrasive hardness, metal hardness and wear loss. (After Wahl, 1951)**



**Fig. 8.6. The laboratory abrasive test shows good ranking ability among classes of materials (steels versus high-Cr vs. Ni-Hard 1) but is unable to simulate the actual field wear behavior. For example, several irons and steels have comparable mill weight loss but the pin-wear test would rank the irons as far superior materials. (After Albright and Dunn, 1983)**

common that materials of the same hardness show widely different abrasion resistance, and harder materials may even turn out to be less resistant than softer ones. (Some examples are shown in Fig. 8.7.)

A reasonable correlation between hardness and abrasion resistance can be observed for steels of the same or similar chemical composition, where different hardness levels have been obtained by heat treatment, especially by tempering a martensitic structure.

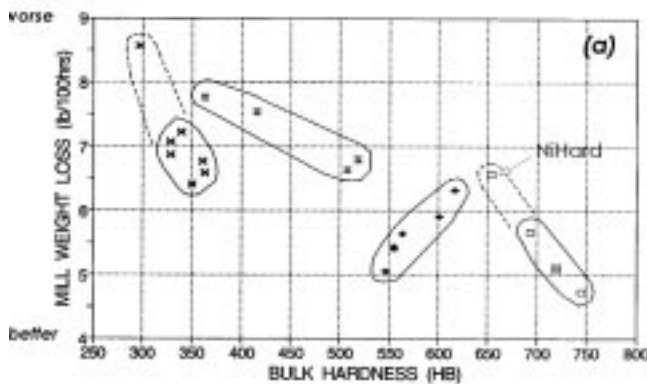
A better but not satisfactory correlation between hardness and relative abrasion resistance was found by Borik and Sponseller (1971) when they used the hardness of the worn work-hardened surface instead of the as-delivered bulk hardness. Testing was done in a small jaw-crusher with river gravel and materials as diverse as low-carbon wrought steel plate, austenitic manganese steels, and various low- and high-alloy white irons.

## Abrasive Wear Field Trials Vs. Laboratory Testing

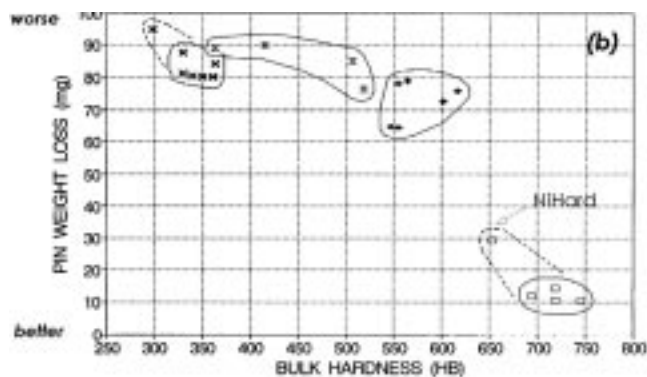
### Pin-On-Disk Abrasion Resistance Test

Figure 8.6 (Albright and Dunn, 1983) does a good job in showing the difficulties of correlating laboratory data with field trials data. Although the laboratory pin abrasion test did an excellent job in ranking the two classes of materials (irons and steels), it failed to show that, under field conditions, both iron and steel mill liners have somewhat equivalent wear rates within the mill (grate discharge ball mill, 2.9 m (114 in.) in dia and 2.44 m (96 in.) in length operated at Climax Molybdenum in Colorado). Based solely on the pin-wear test data, one would be led to believe that irons would have provided superior abrasion resistance when compared to the steel materials. As shown in Fig. 8.6, nothing could be further from the truth.

Albright and Dunn's work is worthy of deeper study. Table 8.2 gives the compositions and hardnesses of the alloys used in their study. The materials represent main-stream compositions and microstructures. Using the hardnesses given in Table 8.2 and other data provided in their paper,



(8.7a) field conditions



(8.7b) laboratory conditions

**Fig. 8.7. Under field conditions (a), the white cast irons are only slightly better than martensitic steel alloys even though they are much harder. In the laboratory (b), the white cast irons show superior abrasion resistance using the pin abrasive wear test. The outlier data point within the white cast iron group is the sole Ni-Hard 1 alloy given in Table 8.2, Liner V. (After Gates and Eaton, 1993)**

Figs. 8.7a and 8.7b were constructed by Gates and Eaton (1993). Figure 8.7 highlights the disparity between field and laboratory tests.

In the field trials (Fig. 8.7a), white cast irons are only slightly better than martensitic steels. While in the laboratory (Fig. 8.7b), the white cast irons show superior pin abrasion resistance. This disparity is hard to explain. One explanation would mention the effects of corrosion and impact damage to explain the difference. More simply, actual service environments involve damage over days and weeks, while laboratory tests involve damage over seconds and minutes.

As the abrasion-resistant material is abraded in the lab, micro and subsurface cracks are introduced. When exposed to atmospheric conditions over days and weeks, the fracture energy required to propagate these cracks is lowered. Hence, it is extremely difficult to match field trial data with short-term laboratory tests.

All is not lost. In general terms, Albright and Dunn's work did show a general correlation between field and laboratory tests as shown in Fig. 8.6. For example, the pin test did an adequate job in ranking the various steels. If we consider that the various steels have somewhat the same toughness and corrosion behavior, then, all things being equal, the pin test is measuring the relative effects of one component: severe abrasion (i.e., not corrosion, not impact).

If we focus on just the white iron group in Fig. 8.6, the pin test does a good job in ranking the behavior of the Ni-Hard 1 iron against the high-Cr iron in the same order as the ball mill field trial data. This result can be explained, microstructurally. Both the Ni-Hard 1 and high-Cr irons have microstructures that are composed of carbides and mixed proportions of martensite and austenite in the matrix. Hence, the materials fall within a general class of alloys. One significant difference between the two alloys is their carbide hardness. The Ni-Hard 1 carbides run around 900 DPH ( $M_3C$ ), while the high-Cr carbides run around 1250 DPH ( $M_7C_3$ ). For more information, see Appendix D.

Among just the white cast irons, the correlation with the ball mill field trial data is quite weak and is not statistically significant. It would appear that environmental conditions enforce a larger spread in wear data than the pin test's measurement of one wear variable: severe abrasion.

In general, we can make a few statements about the abrasive wear comparison between field data and laboratory data:

- The pin test provides useful ranking among classes of materials.
- Cross correlation between different material classes is difficult and, quite often, wrong.
- Field wear is a combination of many phenomena (e.g., perhaps up to 80 factors), while the laboratory pin abrasion test is a measure of just a couple of phenomena (e.g., less than five).

**Boyes/Avery Abrasion Resistance Test**

Norman and Loeb's (1948) paper is the classic of all field trial studies. The field trials were conducted at the Climax Molybdenum mine at Climax, Colorado, U.S.A. (same mine as in Albright and Dunn's work) in a 1.83x1.83 m (6x6 ft) ball mill, over a time span of seven years on over 200 metallurgical classifications of steel and iron balls (94 separate wear tests). It stands as a Herculean piece of investigative field work.

Avery's (1948) contribution was to compare laboratory test data against field trial data. The laboratory test was popularized by the American Brake Shoe Co. (ABEX) and is cited in several sources, e.g., Avery (1974). Abrasive action is modeled by moving a 3.8x5.7 cm (1.5x2.25 in.) sample, under pressure, over a Cu track for a distance of about 2.1 km (7000 ft), while the test sample is submerged in a mixture of water and quartz sand. Elapsed time for this test is about an hour.

Using identical samples to those evaluated in Norman and Loeb's field trial, Avery constructed the graph shown in Fig. 8.8. At first glance, the correlation seems exceedingly

good. Upon closer inspection, the only irons studied by Avery were of the chill cast and Ni-Hard 1 variety ( $M_3C$  carbide morphology). The other two classes of materials were Mn steels (austenitic) and alloy steels (martensitic and pearlitic). From this limited set, Avery showed that martensitic Ni-Hard 1 had superior laboratory abrasion resistance but only slightly better field trial abrasion resistance. Good correlation was shown among the steel group and the few Mn steels studied. In brief, Avery's work foreshadows the work of Albright and Dunn (1983), where good correlation was shown among the steel group but little cross correlation to other classes of materials.

Both Norman and Loeb (1948) and Avery (1948) felt that impact conditions within the ball mill were responsible for the relatively poor performance of the more brittle, low-alloy white cast irons. A similar statement was also made by Albright and Dunn (1983) regarding their comparison of high-Cr irons and alloy steels as liner materials. One tentative conclusion is that a combined measure of fracture toughness and laboratory abrasion resistance might be useful in predicting field wear performance (or abrasion resistance).

**Table 8.2**  
**Composition and Hardness of Ball Mill Liners<sup>(1)</sup>**

| Liner . . . . | Nominal Alloy Composition (Wt%) . . . . .    | Microstructure . . . . .                                          | Worn Surface Hardness, HB |
|---------------|----------------------------------------------|-------------------------------------------------------------------|---------------------------|
| A . . . . .   | 0.3C-1.1Mn-0.7Cr-0.5Mo-0.5Ni steel . . . . . | Martensite, bainite . . . . .                                     | 499                       |
| B . . . . .   | 0.4C-C-Mo-V steel <sup>(2)</sup> . . . . .   | Martensite . . . . .                                              | 547                       |
| C . . . . .   | 0.4C-Cr-Mo-V steel <sup>(2)</sup> . . . . .  | Martensite . . . . .                                              | 543                       |
| D . . . . .   | 0.5C-2.0Cr-0.5Mo steel . . . . .             | Martensite . . . . .                                              | 555                       |
| E . . . . .   | 0.5C-2.0Cr-0.5Mo steel . . . . .             | Martensite . . . . .                                              | 560                       |
| F . . . . .   | 0.5C-2.0Cr-0.5Mo steel . . . . .             | Martensite, bainite . . . . .                                     | 415                       |
| G . . . . .   | 0.5C-2.0Cr-0.5Mo steel . . . . .             | Martensite, bainite . . . . .                                     | 356                       |
| H . . . . .   | 0.5C-2.7Cr-0.3Mo steel . . . . .             | Pearlite . . . . .                                                | 298                       |
| I . . . . .   | 0.6C-3.0Cr-0.5Mo steel . . . . .             | Martensite, bainite, pearlite . . . . .                           | 507                       |
| J . . . . .   | 0.6C-2.2Cr-0.4Mo steel . . . . .             | Pearlite, spherical carbides <sup>(3)</sup> . . . . .             | 323                       |
| K . . . . .   | 0.6C-2.2Cr-0.4Mo steel . . . . .             | Martensite . . . . .                                              | 601                       |
| L . . . . .   | 0.6C-0.8Cr-0.5Mo steel . . . . .             | Martensite . . . . .                                              | 601                       |
| M . . . . .   | 0.8C-2.4Cr-0.4Mo steel . . . . .             | Pearlite, spherical carbides <sup>(3)</sup> . . . . .             | 352                       |
| N . . . . .   | 0.9C-2.5Cr-0.3Mo-0.3Cu steel . . . . .       | Pearlite, spherical carbides <sup>(3)</sup> . . . . .             | 337                       |
| O . . . . .   | 0.9C-2.5Cr-0.3Mo-0.3Cu steel . . . . .       | Pearlite, spherical carbides <sup>(3)</sup> . . . . .             | 341                       |
| P . . . . .   | 0.9C-2.5Cr-0.3Mo-0.3Cu steel . . . . .       | Pearlite, spherical carbides <sup>(3)</sup> . . . . .             | 359                       |
| Q . . . . .   | 0.9C-2.5Cr-0.3Mo-0.3Cu steel . . . . .       | Pearlite, spherical carbides <sup>(3)</sup> . . . . .             | 335                       |
| R . . . . .   | 2.6C-16.0Cr-2.5Mo iron . . . . .             | Martensite, $M_7C_3$ carbides, retained austenite (5%) . . . . .  | 712                       |
| S . . . . .   | 2.6C-16.2Cr-1.4Mo-0.7Cu iron . . . . .       | Martensite, $M_7C_3$ carbides, retained austenite (29%) . . . . . | 694                       |
| T . . . . .   | 2.6C-15.8Cr-1.5Mn-1.0Mo iron . . . . .       | Martensite, $M_7C_3$ carbides, retained austenite (8%) . . . . .  | 682                       |
| U . . . . .   | 2.7C-16.2Cr-1.4Mo-0.7Cu iron . . . . .       | Martensite, $M_7C_3$ carbides, retained austenite (30%) . . . . . | 659                       |
| V . . . . .   | 3.2C-3.3Ni-3.1Cr-0.4Cu iron . . . . .        | Martensite, $M_3C$ carbides, retained austenite (21%) . . . . .   | 627                       |

(1) Albright and Dunn (1983)

(2) Proprietary composition

(3) The term "spherical carbides" refers to the 2D interpretation of proeutectoid  $M_3C$  carbides subsequent to high-temperature heat treatment. When viewed in 3D, their morphology is more rod-like than spherical (Laird).

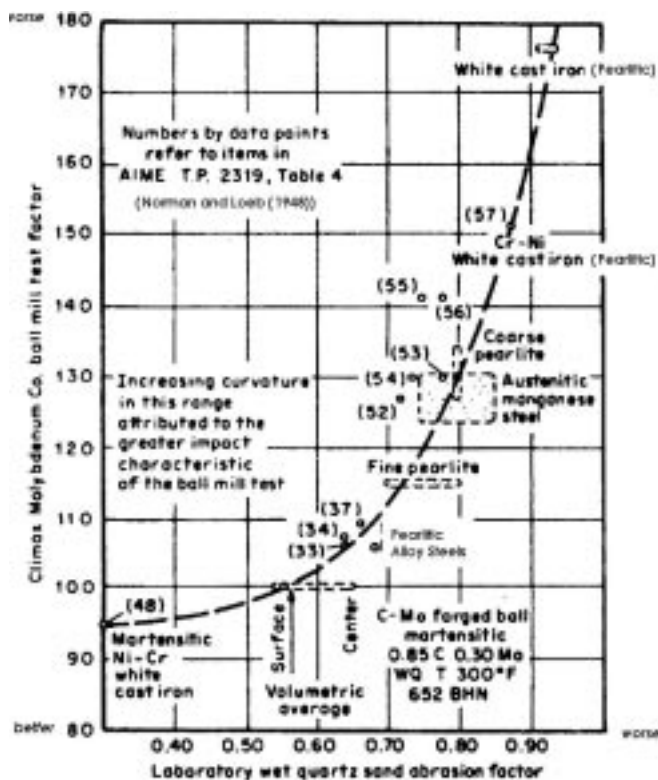


Fig. 8.8. Avery's laboratory wear test appears to give good correlation to field trial wear data. It should be noted that only chill-cast and Ni-Hard 1 irons were evaluated. Interestingly enough, martensitic Ni-Hard 1 is shown to have superior laboratory abrasion resistance over that of pearlitic and martensitic alloy steels but only slightly better field abrasion resistance. This is the same trend as that shown in Fig. 8.6, Albright and Dunn (1983), for their comparison of pin abrasion data against ball mill wear. (After Avery, 1948)

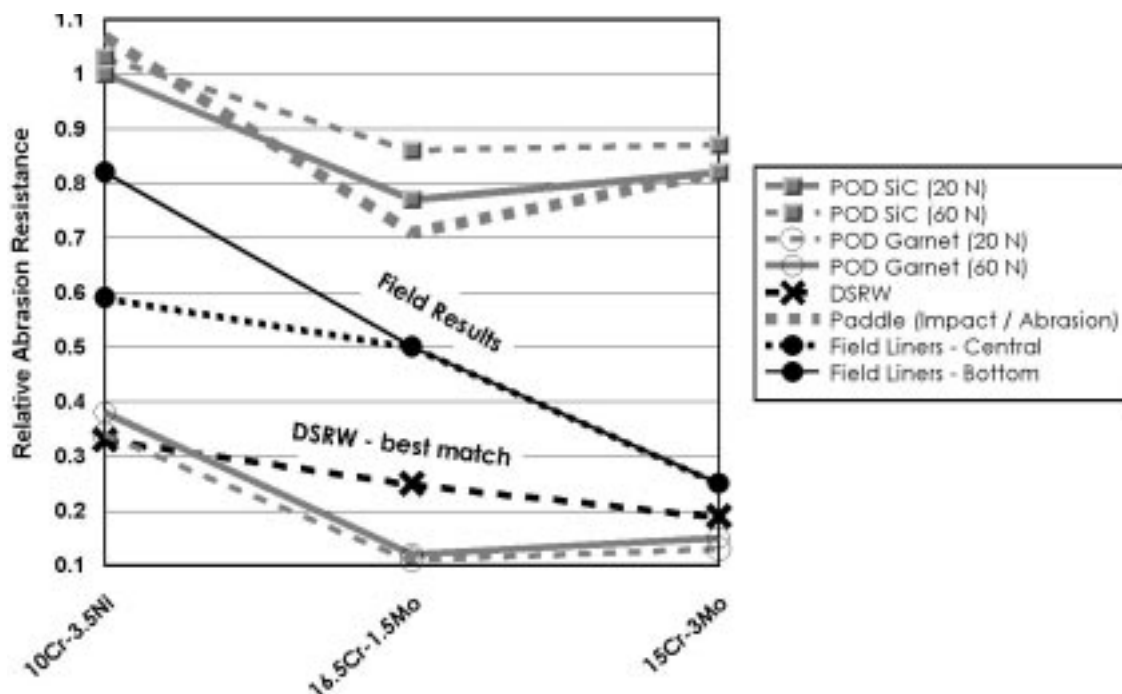


Fig. 8.9. On a qualitative basis, the dry sand rubber wheel (DSRW) test does the best job in predicting the field abrasion resistance of the three AR cast irons. (After Xu et al., 1993)

### Dry Sand Rubber Wheel Abrasion Resistance Test

A paper by Xu, Vose and St.John (1993) represents one of the few attempts to correlate field abrasive wear with a number of laboratory tests. Figure 8.9 presents a graphical summary of their work. The dry sand rubber wheel (DSRW) test (ASTM G 65-85 1985) shows the best correlation to their field liner test providing the correct ranking between the three alloys.

In this study, a paddle wheel test, or impact pulverizer (bond test) (see Fig. 8.3 (1)), is used to simulate combined impact and abrasive wear. Under the conditions used in their study, the pin-on-drum (POD) tests provided similar ranking as this paddle wheel test.

The field wear study in this work was carried out in an iron ore mine in Western Australia. The candidate materials were arranged along the side of a conveyor travelling at a speed of 3.8 m/s (12.4 ft/s). In this arrangement, ore impact is minimal, leaving the dominant wear mode as abrasive. Considering the field conditions, perhaps it is not surprising that the DSRW test provided the best correlation between field and lab tests. Nonetheless, since the scope of their work is so narrow (i.e., their sample set only covered three different irons), it is impossible to say whether their results are truly meaningful or just fortuitous.

### Qualitative Understanding of Field Wear Vs. Laboratory Testing

In many cases, the field wear process is qualitatively well understood by users of AR cast irons. They well aware that martensitic irons are far superior to pearlitic or austenitic irons in most abrasive environments. In most cases, the ideal microstructure consists of about 30% carbide volume fraction (CVF) within a matrix structure of about 50% martensite and 20% austenite.

Undocumented case histories abound of austenitic or pearlitic irons suffering severe abrasive wear while their martensitic cousins experience only moderate or minor wear. Troubleshooting, with regard to abrasion resistance, often consists of looking for pearlite within the matrix microstructure. This approach is borne out by the field wear testing conducted by Norman and Loeb (1948) (see Fig. 8.8) and many laboratory tests.

Figure 8.10 presents a comparison of laboratory pin-on-drum abrasion results among some pearlitic, austenitic and martensitic irons. The martensitic abrasion value is normalized to around 1.0, while the pearlitic and austenitic values are compared against this normalized value. This technique allows us to compare data from more than one source. It was found that the pearlitic irons show consistently poor abrasion resistance as compared to the austenitic or martensitic irons. This result corresponds to field trials and other anecdotal evidence.

In a few rare cases, pearlitic irons have been shown to give adequate abrasive wear resistance when severe impact prevents the use of austenitic or martensitic irons. This phenomenon reflects the stable nature of a fully pearlitic matrix. If the iron has been fully decomposed to pearlite, there is little or no austenite-to-martensite transformation and, hence, little chance of forming high internal stresses within the microstructure. Consequently, the iron is extremely ductile and resistant to impact. Repetitive-impact work using a ball-on-ball test (see Blickensderfer et al., 1989) shows that pearlitic balls yield superior repetitive-impact resistance over austenitic and martensitic irons but, as shown in Fig. 8.10, inferior abrasion resistance.

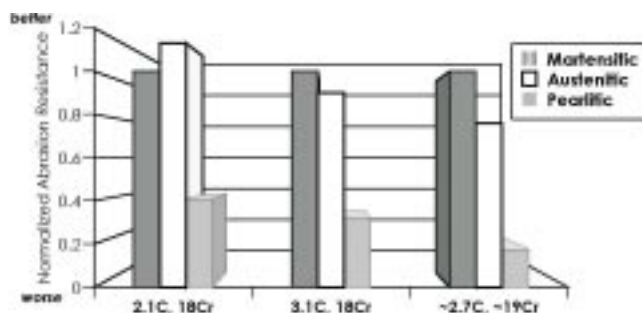


Fig. 8.10. Laboratory abrasion pin-on-drum tests consistently rank pearlitic AR cast irons near the bottom. (After Blickensderfer, 1989 and Gundlach and Parks, 1978)

A problem with the laboratory POD test results shown in Fig. 8.10 is that austenitic and martensitic irons are shown as having somewhat similar abrasion-resistant results. If we switch to the DSRW test, the data plotted in Fig. 8.11 show that martensitic irons provide better abrasion resistance than austenitic irons. To further muddy the waters, data from laboratory jaw crusher tests shown in Fig. 8.12 demonstrate that a mixed austenite and martensite matrix is best. The one thing that can be said with great confidence is that the variability of laboratory test results matches the variability in field wear results.

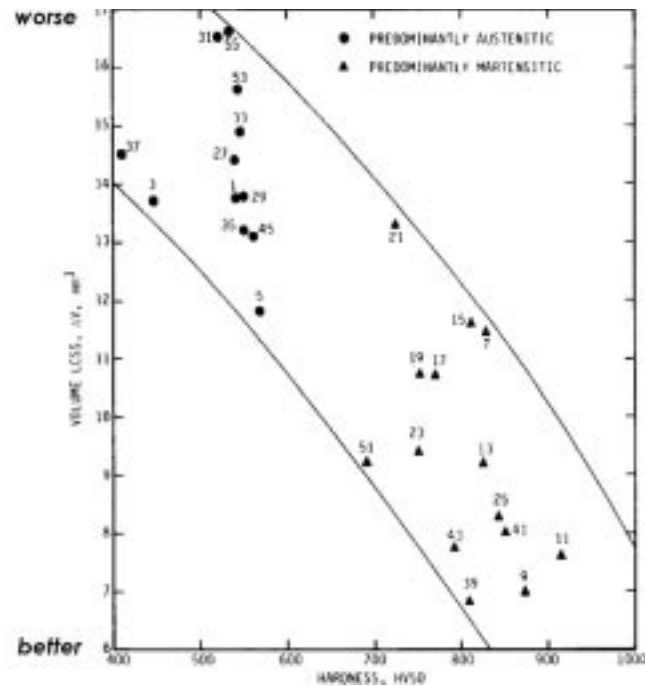


Fig. 8.11. Using the DSRW test, harder (i.e., martensitic) 2.9C, 19Cr, 2.4Mo and 0.9Cu irons give better abrasion results. (After Zum Gahr and Doane, 1980)

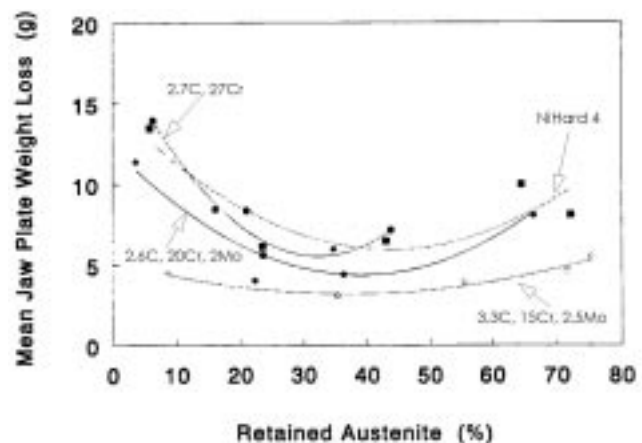


Fig. 8.12. Laboratory jaw crusher tests on a variety of alloys indicate that the best abrasion resistance is obtained with a mixed microstructure of martensite and austenite (between 20 and 30% retained austenite), with remaining portion of matrix being martensite. (After Sare and Arnold, 1989)

In general, most field wear abrasive applications respond favorably to a matrix microstructure that is neither 100% austenite nor 100% martensite. Laboratory wear tests will tend to favor certain matrix mixes. For example, the POD test favors austenitic structures, the DSRW test tends to favor martensitic structures and jaw-crusher tests favor a mixed microstructure leaning toward the martensitic side (this result is noted by Diesburg and Borik, 1974). Many other laboratory abrasion-resistance tests have similar biases. No one test is right and no one test is wrong.

The failure of many laboratory tests to correlate with field data has been attributed to the lack of work-hardening in the laboratory test (see Sare and Arnold, 1989 and Dodd, 1969). We know from field experience that all wear parts experience some degree of work-hardening during their service life (for a review of this subject see Avery, 1974). Depending on the severity of this work-hardening, we may see dominantly austenitic alloys performing better than dominantly martensitic alloys.

One point should be made crystal clear: as of this writing, no laboratory test has shown good or even moderate correlation with field wear behavior over a significant range of AR cast irons. Hence, the real purpose of laboratory testing is to gain insight into the relationships between microstructure and

abrasion-resistant performance, as defined by a particular test. For example, pin-on-drum abrasion tests favor austenitic alloys for a number of reasons. If our field application shows a similar trend, then a potential exists to use this laboratory test to explore improvements in this alloy, without incurring the excessive costs associated with extensive field trials. In brief, laboratory tests provide general insights into abrasion-resistant trends.

## Designing Microstructures for Abrasion Resistance

### Matrix Structure

Table 8.3 presents selected pin abrasion test data on alumina for several AR steels and AR cast irons, condensed from the work presented by Diesburg and Borik (1974). A complete set of their data can be found in Appendix C. The standard laboratory observations are easily made:

1. For cast steels, increasing hardness results in increasing wear resistance.
2. For Hadfield steels, hardness is not indicative of their true wear resistance.
3. For white cast irons, an austenitic matrix is superior to a martensitic matrix for pin abrasive wear.

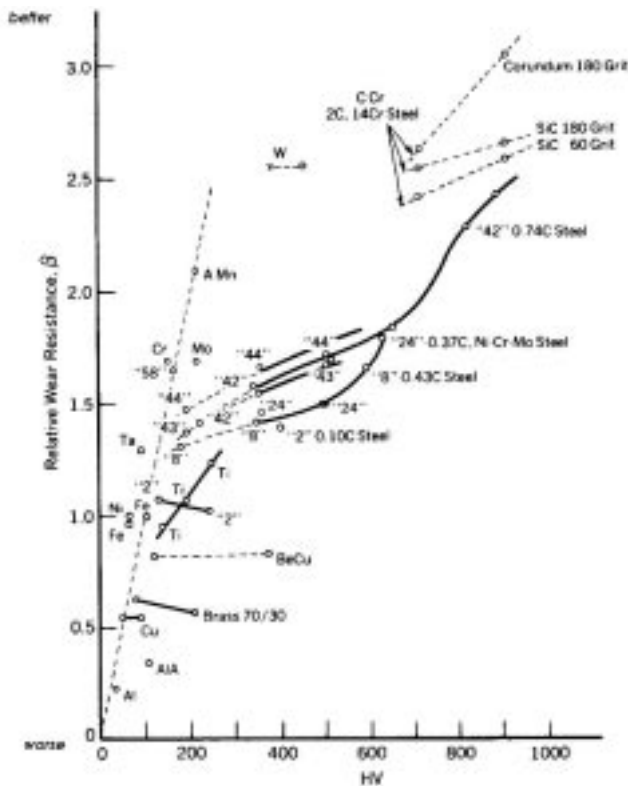
**Table 8.3**  
**Laboratory Pin Test Data for Common Steels and AR Cast Irons**

| Code No. <sup>(1)</sup>                                                                                                                                              | Type                        | %C   | %Cr  | %Ni  | %Si  | %Mo  | %Mn  | Other  | HB <sup>(3)</sup>  | Pin <sup>(2)</sup> (g) | Principal Microstructure Component |
|----------------------------------------------------------------------------------------------------------------------------------------------------------------------|-----------------------------|------|------|------|------|------|------|--------|--------------------|------------------------|------------------------------------|
| <b>Wrought Steels:</b>                                                                                                                                               |                             |      |      |      |      |      |      |        |                    |                        |                                    |
| 3 . . .                                                                                                                                                              | Mn-Mo-Cb                    | 0.08 | 0.04 | 0.09 | 0.03 | 0.25 | 1.33 | 0.03Cb | 187                | 0.156                  | F                                  |
| 7B . . .                                                                                                                                                             | T-1                         | 0.19 | 0.56 | —    | 0.28 | 0.18 | 0.84 | 0.02Ti | 376                | 0.124                  | M, B                               |
| <b>Cast Steels:</b>                                                                                                                                                  |                             |      |      |      |      |      |      |        |                    |                        |                                    |
| 26 . . .                                                                                                                                                             | Mn-Si-Cr-Mo                 | 0.43 | 0.83 | —    | 1.46 | 0.49 | 1.39 | —      | 366                | 0.088                  | M, B                               |
| 28 . . .                                                                                                                                                             | Mn-Si-Cr-Mo                 | 0.63 | 0.83 | —    | 1.48 | 0.49 | 1.44 | —      | 620                | 0.077                  | M, B                               |
| 29 . . .                                                                                                                                                             | Cr-Mo (0.63C)               | 0.63 | 2.30 | —    | 0.58 | 0.34 | 0.71 | —      | 280                | 0.098                  | P, c                               |
| <b>Tool Steel:</b>                                                                                                                                                   |                             |      |      |      |      |      |      |        |                    |                        |                                    |
| 31A . .                                                                                                                                                              | AlSi H-11                   | 0.35 | 5.0  | —    | —    | 1.5  | —    | 0.4V   | 658                | 0.073                  | M                                  |
| <b>Cast Austenitic Mn Steel (Hadfield):</b>                                                                                                                          |                             |      |      |      |      |      |      |        |                    |                        |                                    |
| 42 . . .                                                                                                                                                             | 12Mn-1Mo                    | 0.93 | —    | —    | 0.5  | 1.0  | 12.5 | —      | 188                | 0.087                  | A, p                               |
| <b>White Cast Irons:</b>                                                                                                                                             |                             |      |      |      |      |      |      |        |                    |                        |                                    |
| 62A . .                                                                                                                                                              | 15Cr-2Mo-1Cu <sup>(4)</sup> | 3.25 | 15.1 | —    | 0.49 | 1.95 | 0.66 | 1.08Cu | 483 <sup>(3)</sup> | 0.042                  | A, C, SC                           |
| 62F . .                                                                                                                                                              | 15Cr-2Mo-1Cu <sup>(5)</sup> | 3.25 | 15.1 | —    | 0.49 | 1.95 | 0.66 | 1.08Cu | 700 <sup>(3)</sup> | 0.047                  | A, C, M, SC                        |
| 78A . .                                                                                                                                                              | 18Cr-2Mo-1Cu <sup>(4)</sup> | 3.28 | 18.1 | —    | 0.57 | 2.00 | 0.75 | 0.98Cu | 512 <sup>(3)</sup> | 0.041                  | A, C, m                            |
| 78C . .                                                                                                                                                              | 18Cr-2Mo-1Cu <sup>(5)</sup> | 3.28 | 18.1 | —    | 0.57 | 2.00 | 0.75 | 0.98Cu | 763 <sup>(3)</sup> | 0.049                  | M, C, SC, a                        |
| Microstructure: A = austenite; M = martensite; F = ferrite; B = bainite; C = carbide; SC = secondary carbide; P = pearlite. (Lower case letters mean small amounts.) |                             |      |      |      |      |      |      |        |                    |                        |                                    |
| (1) Code numbers per Diesburg and Borik (1974).                                                                                                                      |                             |      |      |      |      |      |      |        |                    |                        |                                    |
| (2) Standard pin test against 80 $\mu$ m alumina paper (Muscara and Sinnot, 1972).                                                                                   |                             |      |      |      |      |      |      |        |                    |                        |                                    |
| (3) Hardness numbers converted from HRC measurements.                                                                                                                |                             |      |      |      |      |      |      |        |                    |                        |                                    |
| (4) Irons are in the as-cast condition.                                                                                                                              |                             |      |      |      |      |      |      |        |                    |                        |                                    |
| (5) Irons heat treated at 1000C (1832F).                                                                                                                             |                             |      |      |      |      |      |      |        |                    |                        |                                    |

The third statement needs some comment. The pin test can be done with silicon carbide, alumina or garnet cloth. Silicon carbide and alumina have such a high hardness that they can cut any conceivable matrix of iron base alloys as well as the carbides present in usual white irons. The effect of the bulk hardness or the hardness of the individual phases of white irons is reduced, and other material properties will become more important. A softer abrasive, like garnet (which is used because quartz cloth is not available any more), puts greater emphasis on hardness, and being closer to most commercial minerals can be expected to give a better representation of industrial conditions.

The wear behavior indicated in Table 8.3 can be discussed and explained along these lines. Interestingly enough, the majority of abrasive wear literature supports these simple conclusions.

The explanation for the wear behavior indicated in Table 8.3 is equally simple. As the hardness of standard, wrought and cast steels increases, resistance to abrasive gouging and other abrasive removal processes increases. In these steels, the microstructure is stable. That is, little or no metastable austenite exists to transform to hard martensite on experiencing severe plastic strain. For stable ferrous and nonferrous alloys, this trend of increasing hardness versus increasing laboratory abrasion resistance is well known and extensively documented. One example of this trend is shown in Fig. 8.13. In most field wear behavior situations, this trend is supported.



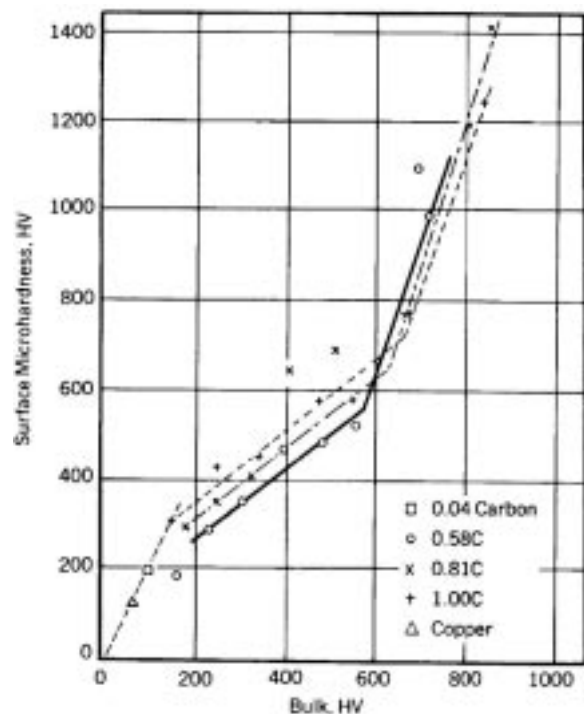
**Fig. 8.13. Relative wear resistance of pure ferrous and nonferrous metals. As the hardness increases, so does the pin abrasion resistance. (After Richardson, 1967)**

Naturally, this trend is reversed if severe impact is experienced. A prime example of this scenario is the use of pearlitic steels and AR cast irons for high-impact mill liners.

It has been mentioned that steels do work-harden, to some degree, under abrasive wear conditions. However, the type of work-hardening experienced by steel alloys is more along the lines of dislocation entanglement or cold work-hardening, than of transformation-based hardening. This effect was nicely demonstrated in work by Larsen-Badse (1966) and is shown in Fig. 8.14.

If we truly consider the results of work-hardening, then initial hardness is of little use in the prediction of wear life. This is demonstrated in Table 8.3 for the cast austenitic Mn steel. Although it is as soft or softer than all the other steels, it has a pin abrasion resistance that is equivalent to the hardest steel. Evidently, the pin abrasion test is creating sufficient plastic deformation to work-harden the Hadfield steel (i.e., cold work or induce a martensitic transformation).

For our AR material of choice, it is hard to beat a good AR cast iron for pin abrasive wear performance. The two irons shown in Table 8.3 provide a stark example of how the pin abrasion test on alumina favors austenitic irons (62A and 78A) over martensitic irons (62F and 78C). These two irons are representative of the bulk of high-Cr hypoeutectic irons in service today. The explanation for this interesting wear behavior lies in the strain-induced transformation of austenite to hard martensite during pin abrasion. This result has been confirmed by a number of researchers.



**Fig. 8.14. Relation between microhardness of abraded surface and bulk hardness for a range of hardened and tempered steels. Alloys were tested using a pin abrasion test with 400-grit SiC as the abrasive. (After Larsen-Badse, 1966)**

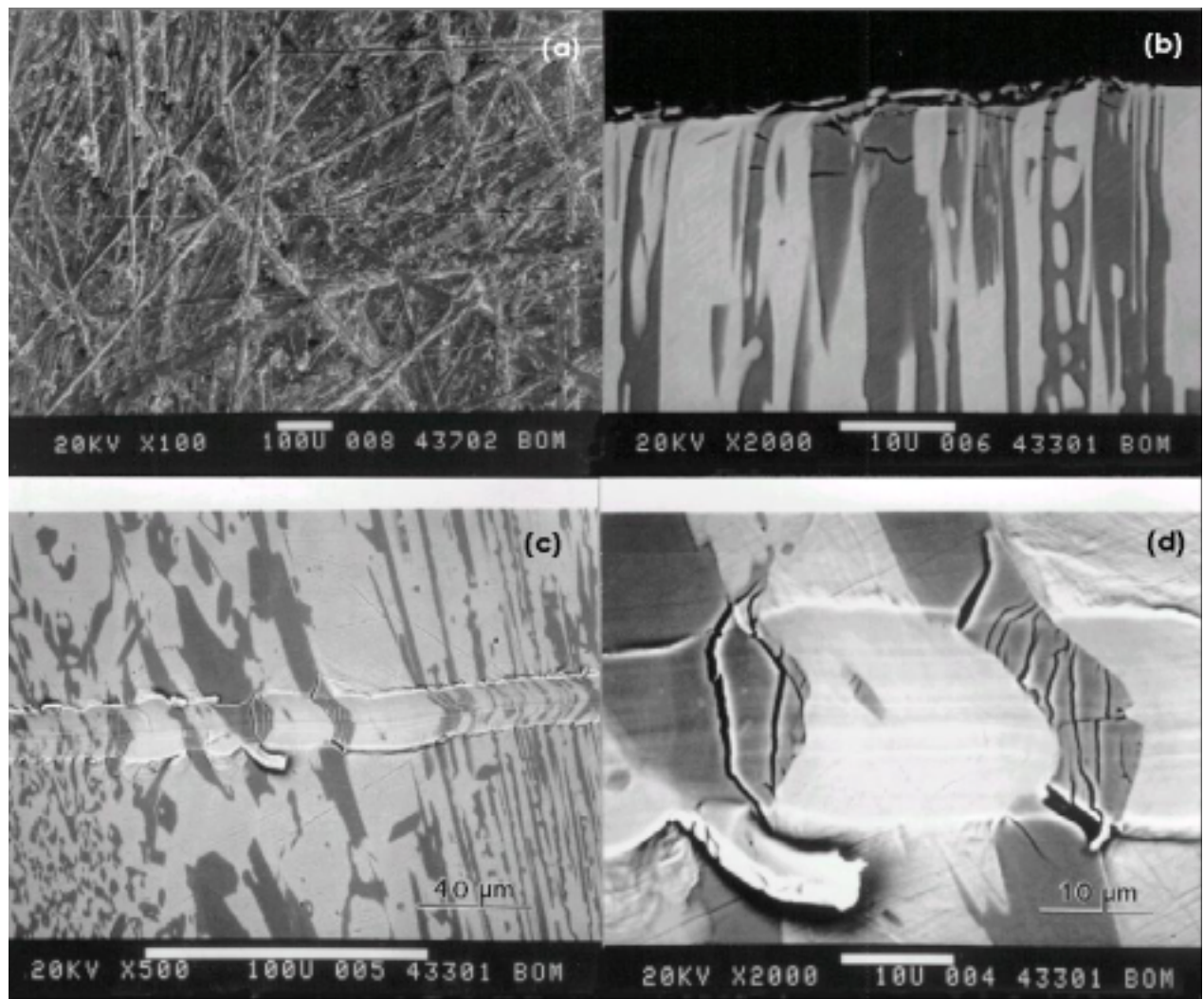
To gain some insight into the pin abrasion process, Fig. 8.15 provides a montage of abraded microstructures. The iron shown in Fig. 8.15 has a rough composition of 3% C and 25% Cr. Much ado has been made about wear scratches, to little avail. Typically, the iron is often shown to be severely deformed by the pin abrasion process. This would seem to be a natural conclusion for a hard material abrading a softer material. Nevertheless, such observations show that such severe deformations within an austenitic structure would lend itself naturally to the strain-induced-martensitic transformation process. Consequently, it seems reasonable that the pin abrasion test on alumina or SiC would rank austenitic irons as “more AR” than martensitic irons.

Pin abrasion tests on garnet cloth tend to favor predominantly martensitic matrix structures. Table 8.4 illustrates this situation. The high-Cr 1A and 1C and high-Cr 4A and 4B irons are less abrasion resistant in the austenitic as-cast condition than after heat treatment resulting in a predominantly martensitic matrix.

Unpublished work in Europe, using garnet cloth, has shown similar results. Here, it was possible to establish not only the correct ranking but also the relative spread for different high-Cr irons used for hard rock crushing in impact crushers.

Contrary to laboratory test results, the lion’s share of high-Cr AR cast irons are used in the heat-treated martensitic condition, i.e., in ball mill liners, gyratory crushers, pulp and paper disk refiners, hot strip mill rolls, etc. More information on usage is provided in Chapter 10, Applications. Obviously, most field abrasive applications do not create as severe abrasive conditions as those experienced within the pin abrasive test. In other words, the iron must have some initial hardness to resist the penetration and grooving of hard abrasive particles. Logically, the best AR iron is comprised of the right amount of carbide in a mixture of martensite and austenite.

The dry or wet sand rubber wheel test reverses this trend and ranks martensitic irons as having better laboratory abra-



**Fig. 8.15.** Pin abrasion tears and grooves on the surface of the test material (upper left), beneath the surface, the matrix and carbides are extensively deformed (upper right), a single scratch shows the ploughing behavior typical of hard abrasives in pin abrasion tests (lower left) and, at a higher magnification, eutectic  $M_7C_3$  carbides are shown to be fractured during the abrasive process (lower right). (After Dogan et al., 1995)

**Table 8.4**  
**Laboratory POD Test Data for Various AR Cast Irons<sup>(1)</sup>**

| Alloy        | %C   | %Cr  | %Ni | %Si | %Mo | %Mn | %Cu | H.T. <sup>(2)</sup> | HB  | POD <sup>(3)</sup><br>(g) | Principal<br>Microstructure<br>Component                                    |
|--------------|------|------|-----|-----|-----|-----|-----|---------------------|-----|---------------------------|-----------------------------------------------------------------------------|
| Forged Steel | 0.89 | 0.5  | 0.1 | 0.2 | 0.0 | 0.9 | 0.2 | Quenched            | 800 | 0.61                      | M                                                                           |
| Ni-Hard 1-1  | 3.1  | 2.6  | 4.4 | 2.0 | —   | 0.9 | —   | As Cast             | 620 | 0.46                      | M, A, C (M <sub>3</sub> C)                                                  |
| Ni-Hard 1-2  | 3.6  | 3.2  | 4.1 | 1.0 | 0.0 | 0.5 | 0.0 | As Cast             | 700 | 0.28                      | M, A, C (M <sub>3</sub> C), SC (M <sub>3</sub> C)                           |
| Ni-Hard 4-1  | 2.9  | 8.4  | 5.8 | 1.8 | 0.0 | 1.0 | 0.0 | As Cast             | 420 | 0.43                      | M, A, C (M <sub>7</sub> C <sub>3</sub> ), SC (M <sub>3</sub> C)             |
| Ni-Hard 4-2  | 3.5  | 8.6  | 5.8 | 2.0 | 0.1 | 0.4 | 0.2 | AH-1 + SR           | 550 | 0.37                      | A, m, C (M <sub>7</sub> C <sub>3</sub> ), SC (M <sub>3</sub> C)             |
| High-Cr 1A   | 3.1  | 15.3 | 0.4 | 0.4 | 0.8 | 0.6 | 0.8 | As Cast             | 420 | 0.27                      | A, C (M <sub>7</sub> C <sub>3</sub> )                                       |
| High-Cr 1C   | 3.1  | 15.3 | 0.4 | 0.4 | 0.8 | 0.6 | 0.8 | AH-2                | 780 | 0.24                      | M, C (M <sub>7</sub> C <sub>3</sub> ), SC (M <sub>7</sub> C <sub>3</sub> )  |
| High-Cr 4A   | 3.2  | 28.8 | 0.3 | 0.2 | 0.5 | 0.5 | 0.1 | As Cast             | 640 | 0.27                      | A, m, C (M <sub>7</sub> C <sub>3</sub> )                                    |
| High-Cr 4B   | 3.2  | 28.8 | 0.3 | 0.2 | 0.5 | 0.5 | 0.1 | AH-2                | 760 | 0.14                      | M, C (M <sub>7</sub> C <sub>3</sub> ), SC (M <sub>23</sub> C <sub>6</sub> ) |

*A = austenite; M = martensite; F = ferrite; B = bainite; C = carbide; SC = secondary carbide; P = pearlite. (Lower case letters mean small amounts.)*

*(1) Laird (1993)*

*(2) Heat treatment: AH = Air Hardened; 1= from 800C (1472F); 2 = from 1000C (1832F); SR = stress relieved at 205–270C (400–520F); and T = tempered.*

*(3) POD = pin-on-drum test using 150-mesh garnet paper.*

**Table 8.5**  
**Laboratory Wet Sand Test Data for Common Steels and AR Cast Irons**

| Code<br>No. <sup>(1)</sup> | Type            | %C   | %Cr  | %Ni  | %Si  | %Mo   | %Mn  | H.T. <sup>(3)</sup> | HB <sup>(3)</sup> | WSAF <sup>(2)</sup><br>(g) | Principal<br>Microstructure<br>Component     |
|----------------------------|-----------------|------|------|------|------|-------|------|---------------------|-------------------|----------------------------|----------------------------------------------|
| <b>Alloy Steels:</b>       |                 |      |      |      |      |       |      |                     |                   |                            |                                              |
| 1 . . . .                  | Aust. 13 Mn     | 1.2  | 0.5  | —    | 0.5  | —     | 13.0 | WQ                  | —                 | 0.80                       | A                                            |
| 2 . . . .                  | Pearlitic Cr-Mo | 0.83 | 2.3  | —    | 0.25 | 0.42  | 0.90 | AH + T              | 345               | 0.64                       | P                                            |
| <b>Ni-Hard 1:</b>          |                 |      |      |      |      |       |      |                     |                   |                            |                                              |
| 5 . . . .                  | Ni-Hard         | 3.18 | 2.32 | 3.93 | 0.51 | —     | 0.58 | SR                  | 600               | 0.56                       | M, A, C, SC (M <sub>3</sub> C)               |
| 6 . . . .                  | Ni-Hard         | 3.02 | 3.68 | 4.02 | 0.51 | —     | 0.90 | SR                  | 460               | 0.69                       | A, m, C (M <sub>3</sub> C)                   |
| <b>White Cast Irons:</b>   |                 |      |      |      |      |       |      |                     |                   |                            |                                              |
| 8 . . . .                  | HC-250          | 2.91 | 25.7 | 0.25 | 0.62 | 0.49  | 1.12 | As Cast             | 477               | 0.80                       | A, C (M <sub>7</sub> C <sub>3</sub> )        |
| 9 . . . .                  | HC-250          | 2.91 | 25.7 | 0.25 | 0.62 | 0.49  | 1.12 | AH-2                | 675               | 0.64                       | M, a, C, SC (M <sub>7</sub> C <sub>3</sub> ) |
| 10 . . .                   | 15-3            | 3.17 | 16.0 | —    | 0.90 | 3.174 | 0.94 | As Cast             | 444               | 0.87                       | A, C (M <sub>7</sub> C <sub>3</sub> )        |
| 11 . . .                   | 15-3            | 3.17 | 16.0 | —    | 0.90 | 3.174 | 0.94 | AH + SR             | 652               | 0.77                       | M, a, C, SC (M <sub>7</sub> C <sub>3</sub> ) |
| 13 . . .                   | Paraboloy       | 3.15 | 16.0 | 0.71 | 0.65 | 1.25  | 0.76 | AH-2                | 627               | 0.49                       | M, a, C, SC (M <sub>7</sub> C <sub>3</sub> ) |

*Microstructure: A = austenite; M = martensite; F = ferrite; B = bainite; C = carbide; SC = secondary carbide; P = pearlite. (Lower case letters mean small amounts.)*

*(1) Code numbers per Avery (1974).*

*(2) Wet Sand: sample is dragged and pushed in a sand slurry bath against a Cu track. More complete details can be found in Avery (1974).*

*(3) Heat Treatment: WQ = water quenched from about 1010C (1850F); AH = Air Hardened; 2 = from 1095C (2000F); SR = stress relieved at 205–270C (400–520F); and T = tempered.*

*(4) This number is suspicious since it is identical to that for the C content.*

sion resistance than austenitic irons. Figure 8.16 shows a side-by-side comparison of this laboratory abrasion behavior between a pin abrasion test and a rubber wheel test. The wet sand abrasive test (see Fig. 8.3, test 5) provides similar ranking to the rubber wheel test (test 6 in Fig. 8.3). For example, Table 8.5 provides a brief summary of some data for the wet sand abrasive test. More complete data can be found in Appendix C.

The dry or wet sand rubber wheel test reverses the trend obtained by the pin abrasion test on silicon carbide or alumina cloth and confirm the results of the pin abrasion test on garnet.

If we consider the data given in Table 8.5 to reflect actual field conditions, most commercial alloys would have compositions similar to Parabology (an alloy developed by the former American Brake Shoe Company (ABEX)). Obviously, there are some differences of opinion on this matter.

The abrasion resistance of martensite is reduced by tempering. This can be easily demonstrated for steels, like the work by Larsen-Badse (1966), which examined abrasion resistant steels with 0.58–1.28% C. Similar results have also been compiled by Béchet, Röhrig and Tither (1984). In white irons, the tempering effect is masked by the carbides, but often a loss in hardness can be observed. Heat treatments aimed at obtaining low austenite contents often involve tempering or a controlled quenching cycle that will temper the martensite and reduce its abrasion resistance. This could explain the lower abrasion resistance of the low-austenite structures shown in Fig. 8.12.

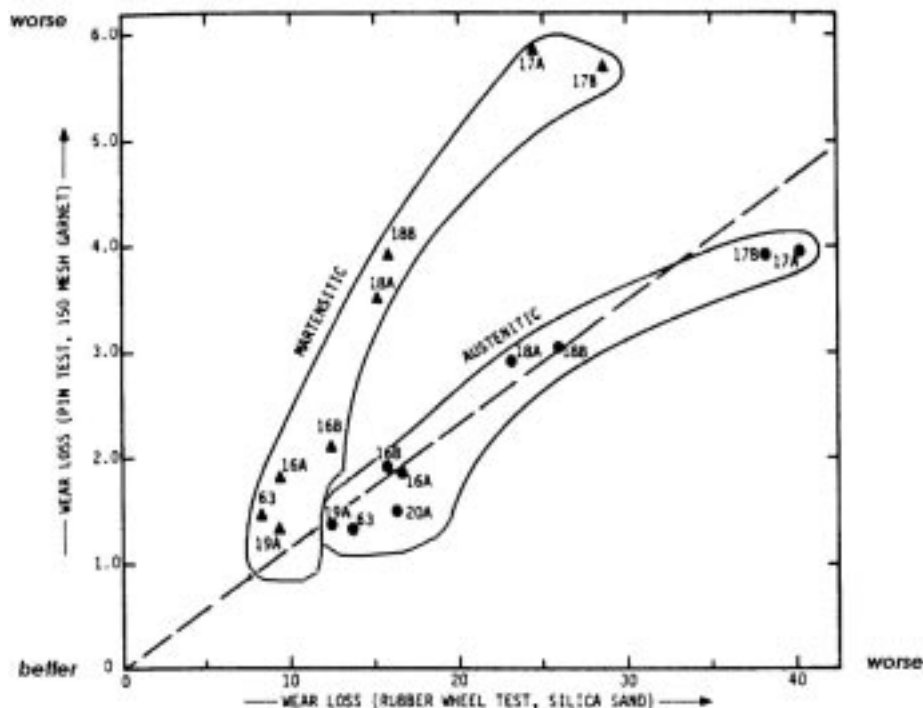
The tempering effect on abrasion resistance can be of commercial interest. For example, tempering high-Cr iron blow bars for impact crushers, even at a temperature as low as 240C resulted in a loss of abrasion resistance. In this case, a decision between longer wear life or increased risk of breakage had to be made before entering the highly competitive market.

### Carbon Content

Carbon is the main alloying element in AR iron-base alloys. Krushchov et al. (1960) has shown that the abrasion resistance of heat treated steels of equal hardness increased with increasing carbon content. Borik and Sponseller (1971) plotted the relative abrasion rates in the jaw-crusher test against carbon content and found a correlation in a scatter band that was, however, fairly wide.

These conclusions have been confirmed in many laboratory and industrial tests for a wide range of iron-base materials, under the condition that the matrix structure and the carbide type and morphology in white irons is taken into account.

Röhrig (1971, 1974) rearranged Borik and Sponseller's data together with additional test results according to matrix structures and carbide type. As shown in Fig. 8.17, abrasion resistance in the jaw-crusher test is a function of carbon content and microstructure. Martensitic or martensitic-austenitic cast irons and steels exhibit the highest abrasion resistance, followed by austenitic steels and irons. Pearlitic cast irons are extremely poor, which can be explained by the fact

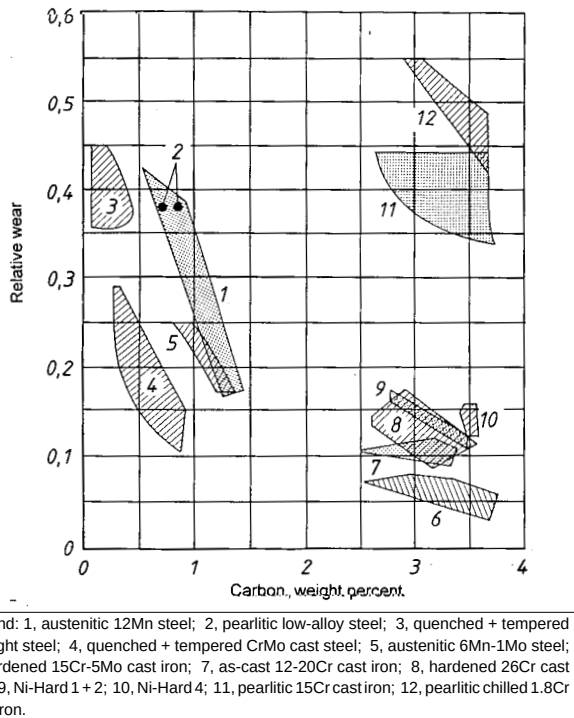


**Fig. 8.16.** An interesting comparison between two common laboratory abrasion tests. The rubber wheel abrasion test ranks martensitic irons as having consistently better abrasion resistance than austenitic irons, while the pin abrasion test ranks these alloys in the exact opposite manner, in most cases. (After Zum Gahr and Eldis, 1980)

that the pearlitic matrix is unable to sufficiently support the carbides against fracture and detachment under the high surface pressure in the jaw-crusher test. Austenitic matrix structures can be work-hardened to a high degree, but not to the level of an initially martensitic structure, which can also be work-hardened to some extent. Therefore, they are still less resistant. Mixed matrix structures of austenite and martensite in the as-cast high-Cr irons No. 7 are in between. The lower abrasion resistance of the 26Cr iron No. 8 results from its lower carbon martensite and a still fairly high austenite content. Ni-Hard 4 (No. 10) was not properly heat treated. This behavior of high-austenite-martensite structures agrees with the findings in Fig. 8.12. Figure 8.17 also shows the effect of the carbide type. With a comparable matrix structure, chromium carbides are more resistant than cementite, as shown by the comparison of the high-Cr-irons No. 6 and 7 to Ni-Hard 1 + 2 (No. 9).

Similar relationships between carbon content and abrasion resistance have been found by Röhrig (1971) after replotting existing published data. Figure 8.18 contains results of a slurry abrasion test with a mixture of water and silica sand, published by Miyoshi et al. (1966).

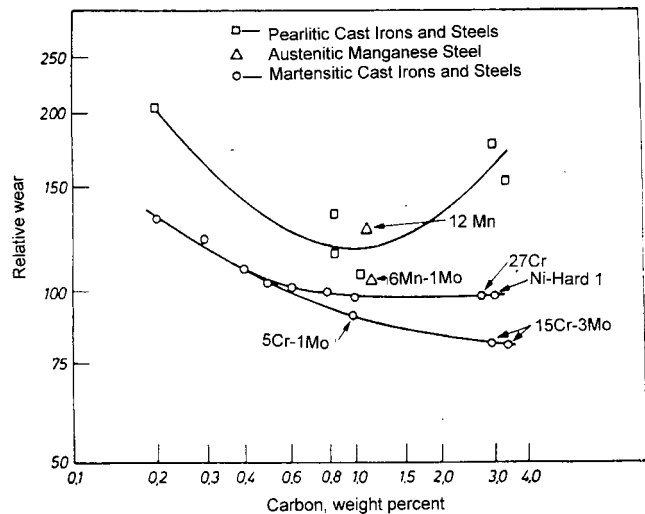
Martensitic cast irons and steels, pearlitic irons and steels, and austenitic manganese steels form three distinct scatter bands, and the influence of the carbide type is also evident. Austenitic manganese steel performs poorly under the low-stress abrasion conditions in this test.



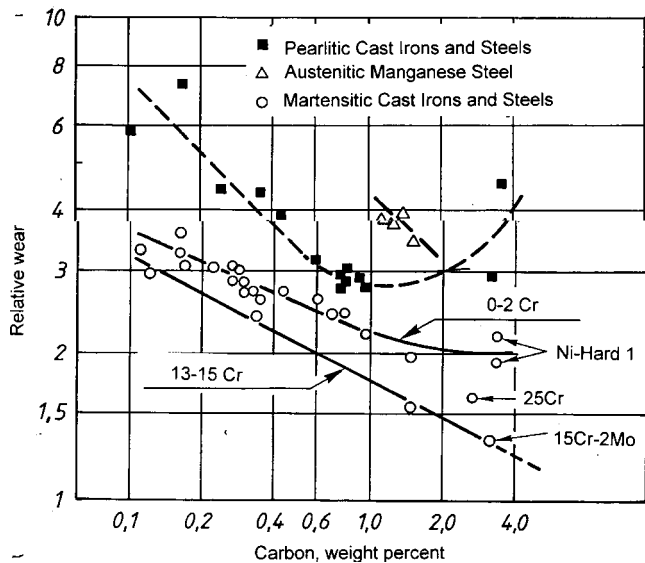
**Fig. 8.17. Relative abrasion rates of various iron base materials as a function of carbon content and microstructure. Testing was done in a small jaw-crusher crushing river gravel. The reference material is a low alloy HSLA steel Grade B ASTM A 517-68. (After Borik and Sponseller, 1971, and Röhrig, 1971, 1974)**

The results of mill testing grinding balls (Norman and Hall, 1969) in Fig. 8.19 demonstrate that the relationship between carbon content, matrix structure and abrasion resistance is also valid under industrial conditions. Here too, martensitic irons and steels perform better than pearlitic or austenitic materials.

Carbon seems to improve abrasion resistance by several mechanisms. The initial as-quenched hardness of martensite is increased by higher carbon content. This will have an effect in low-alloy steels as well as in white cast irons. The work-hardening of austenite—in manganese steels as well as in white cast irons—is increased by carbon. In white cast irons, carbon increases the amount of carbides (carbide volume fraction, CVF).



**Fig. 8.18. Relative abrasion rates of various steels and white cast irons as a function of carbon content and microstructure in a slurry abrasion test with a water and silica sand mixture. (After Röhrig, 1971, based on data by Miyoshi, et al., 1966)**



**Fig. 8.19. Relative abrasion rates of various steels and white cast irons as a function of carbon content and microstructure determined on grinding balls for grinding a hard quartz-rich molybdenum ore (After Röhrig, 1971, based on data by Norman and Hall, 1969)**

### Carbide Volume Fraction (CVF)

One of the more systematic papers on laboratory abrasion resistance is the work performed by Zum Gahr and Eldis (1980) on a broad range of high-Cr white cast irons. Table 8.6 presents the compositions of their alloys. The alloys were heat treated to have either austenitic or martensitic matrices. Abrasion results for just the martensitic matrices are presented in Fig. 8.20. In Fig. 8.20a, the rubber wheel test shows that, as the CVF is increased past 30–35%, the laboratory abrasion resistance decreases.

This is quite interesting. The eutectic CVF for high-Cr irons is around 35% and, as the iron goes hypereutectic, the abrasion resistance decreases. This result would tend to support empirical field evidence about hypereutectic irons. It would appear that the best abrasion resistance lies in irons having eutectic compositions.

Just to keep things confusing, Fig. 8.20b provides some perfectly contradictory results using the pin abrasion test. When the abrasive is garnet, the abrasive resistance reaches a plateau from about 35% CVF onward. As the iron becomes fully hypereutectic (>35% CVF), the garnet pin abrasion resistance is no better than regular eutectic irons (~35% CVF). This, in itself, demonstrates the all around robustness of AR irons having eutectic compositions.

**Table 8.6**  
**Chemical Composition and CVF of**  
**Experimental High-Cr Alloys<sup>(1)</sup>**

| Heat No. <sup>(2)</sup> | %C   | %Cr  | %Si                 | %Mo                | %Mn                | %Cu  | CVF <sup>(4)</sup>  |
|-------------------------|------|------|---------------------|--------------------|--------------------|------|---------------------|
| 17A                     | 1.41 | 11.6 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 1.24 | 7.1                 |
| 17B                     | 1.38 | 12.8 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 1.24 | 9.8                 |
| 18A                     | 2.00 | 15.8 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 1.14 | 14.4                |
| 18B                     | 1.89 | 17.8 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 1.14 | 17.4                |
| 16A                     | 2.58 | 17.6 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 1.03 | 24.3                |
| 16B                     | 2.48 | 18.7 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 1.03 | 22.7                |
| 19A                     | 2.87 | 20.0 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 0.94 | 29.1                |
| 19B                     | 2.79 | 21.0 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 0.94 | 30.4                |
| 63                      | 2.92 | 19.0 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 0.94 | 28.0                |
| 20A                     | 3.50 | 23.4 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 0.87 | 37.6                |
| 20B                     | 3.41 | 24.1 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 0.87 | 41.1 <sup>(5)</sup> |
| 21A                     | 3.93 | 24.6 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 0.76 | 45.4                |
| 21B                     | 3.81 | 25.7 | 0.58 <sup>(3)</sup> | 2.4 <sup>(3)</sup> | 1.6 <sup>(3)</sup> | 0.76 | 38.8 <sup>(5)</sup> |

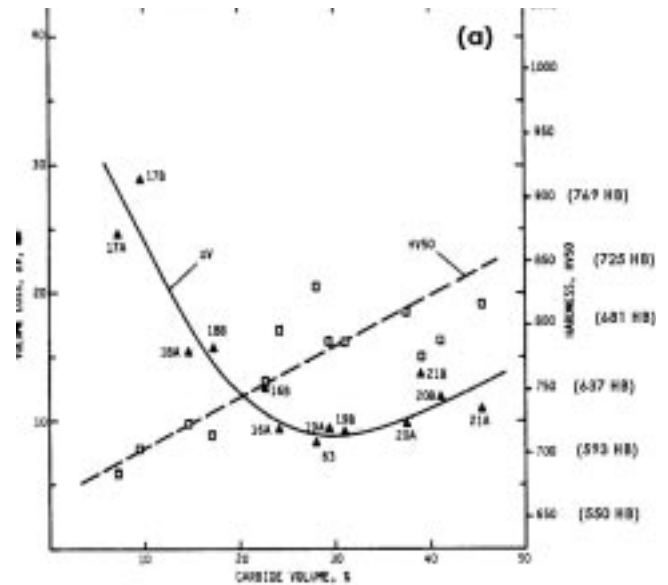
(1) Zum Gahr and Eldis (1980)

(2) Alloys were cast into 50-mm thick Y-blocks.

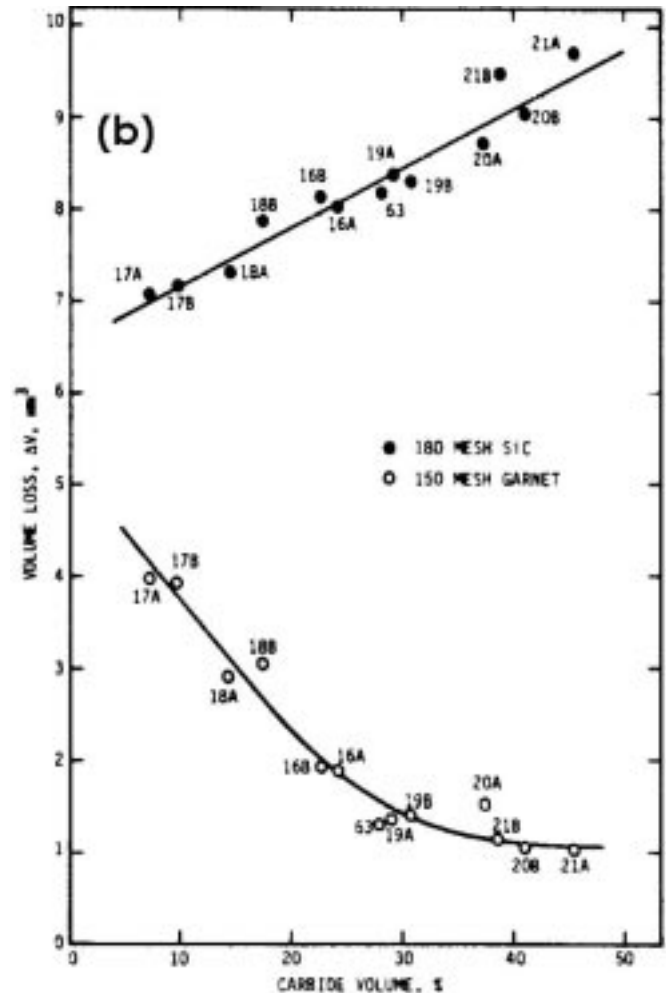
(3) Si, Mo and Mn contents are approximately the same for all alloys.

(4) Carbide volume fraction measured by image analysis using 50 fields at 250X.

(5) CVF measurement is worrisome. Based on phase diagram predictions. Heat No. 21B must have a higher CVF than heat No. 20B, and, based on other experimental work, its CVF should be in excess of 42% (see Laird and Dogan, 1996).



(8.20a) rubber wheel test

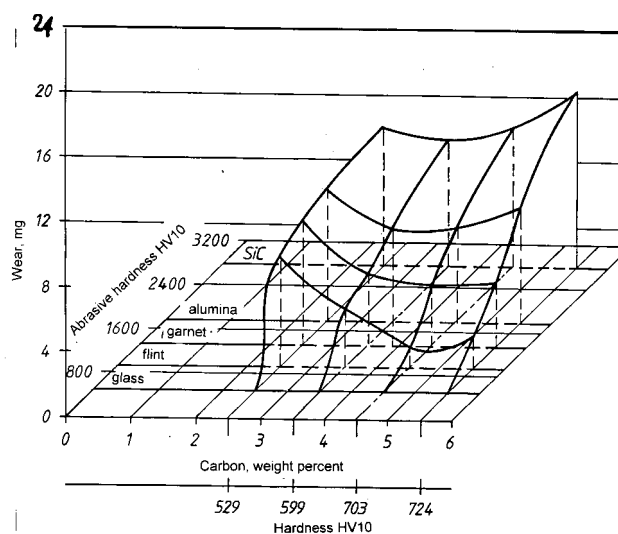


(8.20b) pin abrasion test

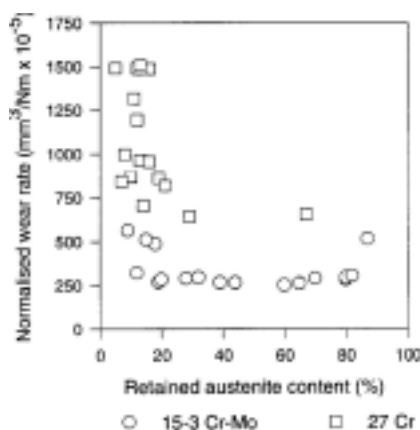
**Fig. 8.20.** A wide variety of laboratory abrasive behavior is demonstrated by the (a) rubber wheel test and the (b) pin abrasive test, using 180-mesh SiC and 150-mesh garnet abrasives. The irons have dominantly martensitic matrices (destabilized at 900C (1652F) and are then cryogenically treated at -78C (-108F)). (After Zum Gahr and Eldis, 1980).

If the abrasive is SiC, the pin abrasive results are nonsensical. As the CVF increases, the SiC pin abrasion resistance decreases. This phenomenon has been explained by a variety of mechanisms. Basically, the much harder and tougher SiC abrasive (2500 HV versus 1200 HV for garnet) rips apart the iron's microstructure, regardless of microstructure and carbide. As the CVF increases and the iron becomes more brittle, the SiC abrasive can rip out a larger chunk of the iron, increasing the laboratory wear loss. This laboratory phenomenon has found no counterpart in an industrial situation.

The effect of the carbon content, which translates into CVF in relation to abrasive hardness in the pin abrasion test, can be seen in Fig. 8.21. There is a distinct minimum amount of abrasion resistance that is displaced to higher carbon content when the abrasive hardness decreases. However, for garnet and flint, which can be regarded as similar to industrial minerals, the abrasive loss at hypereutectic carbon content is still much lower than at eutectic composition, which is around



**Fig. 8.21. Effect of the carbon content of a martensitic-austenitic 30 Cr iron on the abrasion loss in a pin abrasion test with abrasives of different hardness. (after Wahl, 1970)**



**Fig. 8.22. Laboratory pin abrasion data favors the 15% Cr iron, regardless of matrix microstructure (both irons have nearly the same CVF). Results show that the pin abrasion test favors larger, coarser carbides (15% Cr iron) over finer carbides (27% Cr iron). (After Tabrett, 1997)**

2.5% C in this 30Cr iron. Superior performance of strongly hypereutectic 35Cr irons has been reported by Sare (1977).

In industrial applications, a balance among abrasion resistance, resistance against breakage and processing has to be found. In most cases, hypoeutectic or eutectic irons are considered to be the best compromise. However, in some applications, hypereutectic cast irons give excellent performance. It should also be mentioned that hardfacing alloys contain very high carbide volume fractions.

### Carbide Size

Depending on the cooling rate of the casting or the chemical composition, the carbide size can vary significantly. For example, chill casting will create much finer carbides than are found in a slow-cooled sand cast iron. Moreover, as the Cr content increases in a high-Cr iron, the carbides will become smaller and smaller. This is shown most clearly in the carbide size distributions recorded for 15% Cr versus 25% Cr irons (see Chapter 4 on Hardenability for details).

Hidden away in Diesburg and Borik's (1974) work is a reference to the effect of cooling rate on pin abrasion resistance. They noticed that in a 3.2% C, 18% Cr iron higher pin abrasion resistance was found in the sample taken from the slower cooled iron. Their conclusion was that coarse, large carbides yielded increased laboratory pin abrasion resistance.

In another work by Avery, on thick mill liners, coarser structures yielded lower wet sand abrasive factors. His work used samples prepared from 50–50 mm (2–6 in.) thick mill liners. In most cases, the samples sectioned from the surface yielded better laboratory abrasion resistance than those sectioned from the interior of the liner. More complete results can be found in Appendix C. It would appear that Avery's wet sand abrasive test favors smaller carbides, while the pin abrasive test favors larger carbides.

Since alloy content influences the carbide size, it would be only natural to see some general effect based on Cr content. This trend is shown nicely in the work by Tabrett (1997). In his work, he compared the pin abrasion resistance of a 15% and 27% Cr white iron having nearly identical CVFs. Figure 8.22 gives a good idea of the differences in pin abrasion data for the two irons. Regardless of matrix microstructure, the 15% Cr iron had much better pin abrasion resistance than the 27% Cr iron. All things being equal, the only logical explanation is the differences in carbide particle size.

In Tabrett's work, a literature review is provided on the effects of carbide size on abrasion resistance. The literature is conflicting. Depending on the laboratory wear environment, the results can go either way. One safe statement is that carbide size does influence the abrasion resistance of the iron. Whether small or large carbides provide the "best" field abrasion resistance is unknown at this time.

Unpublished field experience with high-Cr iron blow bars on impact crushers in some European quarries has given indication that a finer grained carbide structure reduces wear performance.

### Carbide Morphology and Hardness

Two dominant carbide morphologies exist in the world of AR irons. The carbides are either of the  $M_3C$  plate-like morphology or of the  $M_7C_3$  rod- and blade-like morphology. It is that simple.

Field trials have consistently shown the  $M_7C_3$  carbide to provide better abrasion resistance than the  $M_3C$  carbide (e.g., Norman and Loeb, 1948 or Albright and Dunn, 1983). The  $M_7C_3$  carbide's improved abrasion resistance is attributed to two factors: 1) its discrete rod- and blade-like morphology and 2) its higher hardness. Both of these factors have been discussed at length in Chapter 5 on Morphology, and it seems natural that these attributes would translate into improved wear performance.

Laboratory wear tests abound in their quantitative estimates of the relative merits of irons based on the  $M_3C$  or the  $M_7C_3$  carbide. Once again, none of these laboratory tests are correlated with field performance. Table 8.5 provides a glimpse of how the  $M_3C$  carbide (Ni-Hard 1) stacks up against the  $M_7C_3$  carbide (HC-250, 15-3 and Paraboly). In some cases, the Ni-Hard 1 alloy provides better laboratory abrasion resistance than the low hardness HC-250 and 15-3 alloys. This result reflects the importance of matrix hardness in this laboratory test. Overall, if the matrix is of similar hardness, the irons containing  $M_7C_3$  carbides provide better laboratory abrasion resistance.

Pin abrasion tests provide similar comparisons between Ni-Hard 1 irons ( $M_3C$ ) and high-Cr irons ( $M_7C_3$ ). Examples of pin data are given in Table 8.4. Ni-Hard irons offer superior abrasion resistance, compared to high-hardness forged steel, but do not compare as well against the high-Cr alloys (e.g., high-Cr 1A). Microstructurally, the hard abrasives used in these laboratory tests are much harder than the  $M_3C$  carbide (e.g., 1400 versus 900 HV, respectively). Consequently, these abrasives can easily deform and gouge the Ni-Hard 1 microstructure. Since the  $M_3C$  carbide morphology is somewhat continuous and plate-like, it is easily fractured and “chipped” away during the laboratory abrasion test.

The higher Cr irons, with their harder  $M_7C$  carbides (e.g., ~1400 HV), resist, to a greater extent, the damage introduced by abrasives plowing through their microstructures. Since the carbides are much less continuous or somewhat discrete, the damage introduced by the hard abrasives is more localized and less prone to chipping.

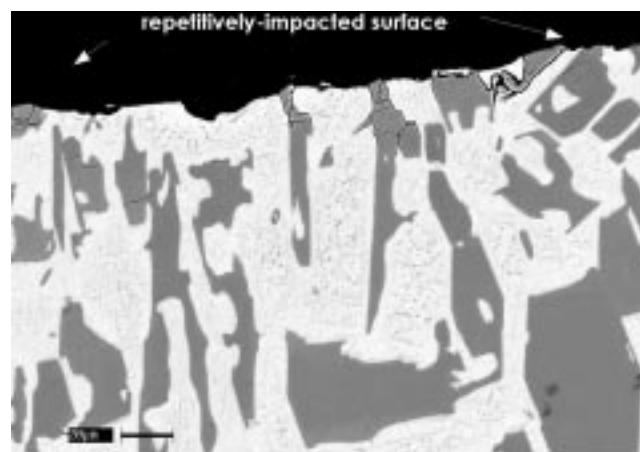
How well these mechanisms translate into field damage is debatable. In some applications, Ni-Hard 1 is the superior, cost-effective material, while, in other applications, a high-Cr material provides the competitive advantage. Nevertheless, understanding the limitations of the microstructure should help guide users in its use.

**Exotic Carbides**—In some select applications (e.g., hot strip steel mill rolls), the use of irons alloyed with NbC and VC carbides are finding acceptance. Although these carbides have hardnesses in excess of 2000 HV, they are often present in volume fractions less than 10% in most irons. At such a low

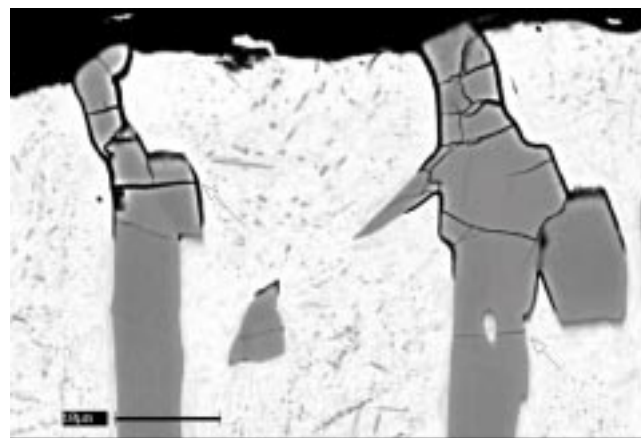
CVF it is difficult to attribute improved abrasive wear resistance to these carbides. Within the literature, only minor improvements in laboratory abrasion resistance has been shown for these exotic carbide additions.

For vanadium, it has been reported from field trials with dredge pumps and shot blast impeller blades that a higher wear life could be achieved with 15Cr-Mo irons if 1–2% V were added. It is not clear whether this effect is due to the presence of hard vanadium-rich carbides or to a reduction of retained austenite due to the carbon depletion of the matrix by the formation vanadium carbides.

Irons that contain only special hard carbides have improved abrasion resistance, as shown in Fig. 8.23. In pressing refractory bricks, the mold life was longer for irons containing Nb or V carbides instead of softer Cr or W carbides, according to Popov and Nagorny (1969). The matrix was austenitic because the refractory mass had a high alumina content.



(8.23a) horizontal cracks



(8.23b) enlargement of (8.23a)

**Fig. 8.23. Backscatter SEM images of subsurface cracking in a high-Cr impact hammer used in a high-speed vertical impact mill. The horizontal cracks shown in the  $M_7C_3$  carbides (arrows) are due to repetitive-impact damage. (Alloy courtesy of Eagle Foundry, Eagle Creek, Oregon, U.S.A. The image was taken by W. K. Collins, Albany Laboratory, DOE, Albany, Oregon, U.S.A.)**

## Repetitive-Impact Wear Classification

Repetitive-Impact wear exists, to some extent or another, in all large-scale abrasive wear processes. In its extreme form, the carbide structure of the iron is damaged, far below the surface of the iron. Such subsurface damage weakens large regions of the iron's microstructure, leading to a particular severe form of wear known as spalling. Figure 8.24 shows a good example of repetitive-impact damage in a heat-treated high-Cr AR cast iron (3.0% C, 25% Cr). Cracking parallel to the wear surface creates a subsurface fracture path that eventually links up to the surface. The result of this repetitive-impact damage mechanism is the formation of a spall or a convex-shaped wear fragment.

The mechanics of spall formation from repetitive-impact damage are quite complex but can be simplified to their essentials. For example, in the formation of spalls in high-Cr cast irons commonly used in vertical or horizontal impact mills, chunks of rock (5–25 cm (2–10 in.)), are quickly fractured into much smaller chunks for subsequent comminution or extraction processes. When the casting is impacted by a rock, tensile stresses are formed along the periphery of the

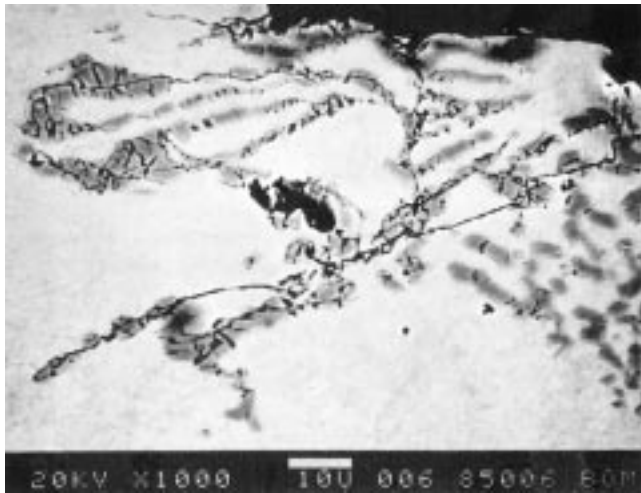


Fig. 8.24. Extensive subsurface cracking is shown in the backscatter SEM image taken on an experimental repetitive-impact-resistant high-Cr iron. (After Laird et al., 1988)

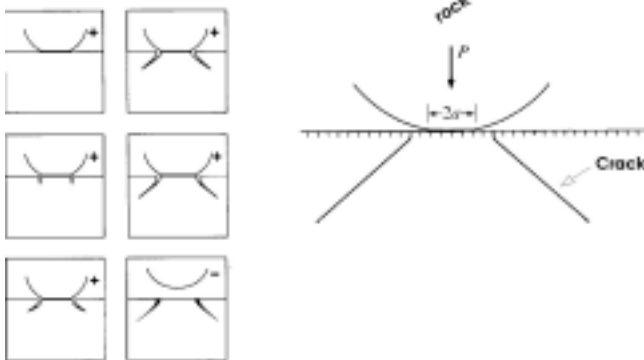


Fig. 8.25. Schematic of vertical crack formation due to impact. (After Lawn, 1993)

interface between the rock and casting. These tensile stresses, which develop due to the rock “punching” into the softer casting, drive the formation of vertical cracks that propagate from the surface of the casting downward into the material. Figure 8.25 shows a schematic of this type of crack formation.

As the rock makes full contact with the surface of the hammer, high shear stresses develop underneath the impact zone. As Fig. 8.26 shows, peak shear stresses develop at some distance beneath the surface of the casting. Consequently, the microstructure plastically deforms and weakens, as rock after rock hits the casting.

Eventually, these initially small cracks (as shown in Fig. 8.23) link up to form much larger cracks. Such severe repetitive-impact damage is shown in Fig. 8.24. After a few more impacts, chunks of the material are released in the form of spalls. Figure 8.27 provides a brief summary of the spalling mechanism.

Repetitive-impact damage can be extremely severe with rapid material loss due to large-scale spall formation. This wear mechanism should not be confused with catastrophic failure of the casting, which is more a characteristic of the

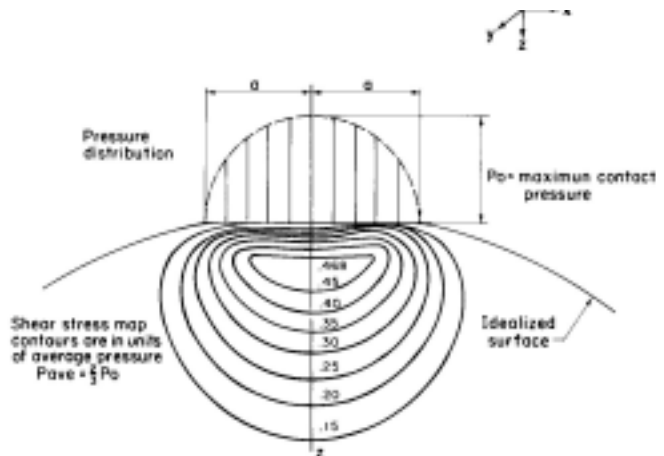


Fig. 8.26. Subsurface shear stress distribution due to localized contact by a hard particle. Similar distributions occur during impact loading. In all cases, maximum damage (i.e., max shear stress) occurs beneath the impacted surface.

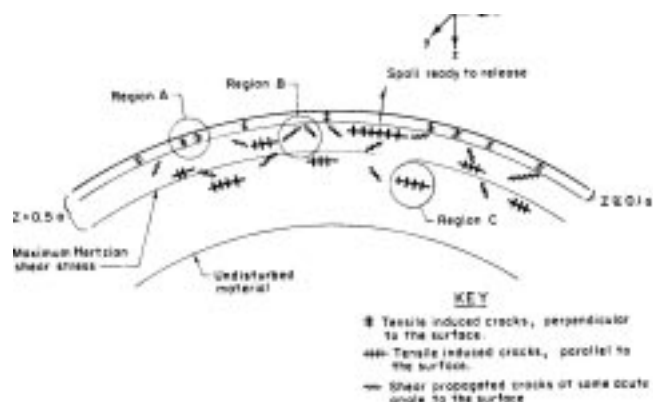
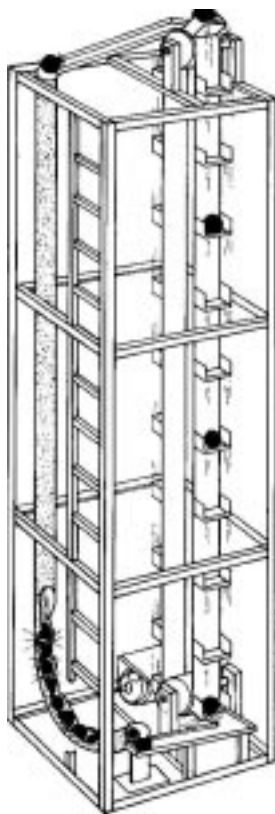


Fig. 8.27. Spalling mechanism in AR cast irons. Large spalls typically form after extensive subsurface damage from repetitive-impact. (After Laird et al., 1988)

casting's impact toughness, which can be related to the material's fracture toughness or simple-blow impact toughness (see Appendices B and C).

There is no simple test that allows an end user to quantitatively define the severity of the repetitive-impact component of their wear environment. In most mineral comminution applications, e.g., impact mills, repetitive-impact wear is dominant. In other applications it is more difficult to quantify. A prime example is in large-diameter ball mills. Under perfect conditions, the ball mill is always operated with a cushion of ore against the AR mill liners. With the ore cushioning the impact of the balls and falling ore, the impact component of the wear environment is rather mild. Nevertheless, mill conditions can vary, leaving little ore to cushion the impact of the grinding media, resulting in severe repetitive-impact conditions.

The ratio between abrasion and repetitive-impact wear for the complete life cycle of the liners and / or lifter bars is nearly impossible to calculate. Consequently, alloy selection is often best done by the person closest to the intended application. Following this logic, the goal in this section on repetitive-impact wear is to provide some clear guidelines as to why certain alloys perform the way they do under repetitive-impact and how best to combat repetitive-impact wear while still retaining good abrasion resistance.



**Fig. 8.28. Ball-on-ball impact machine.** A dropped ball hits a stack of balls contained within a J-shaped tube, creating multiple impacts of decreasing magnitude. The residual force from this impact causes the release of one or multiple balls from the end of the J-tube and into the conveyor system to repeat the process.

## Repetitive-Impact Field Trials Vs. Laboratory Testing

Repetitive-impact wear has been studied in the laboratory using a variety of test methods. We have the ball-on-ball and ball-on-block variants developed by the U.S. Bureau of Mines and a myriad of other testing apparatuses that have attempted to measure both repetitive-impact wear and abrasive wear, simultaneously. In all cases, none of these laboratory tests have been correlated to an industrial application that experiences repetitive-impact wear (e.g., impact hammers used in vertical and horizontal impact mills). Hence, our discussion of repetitive-impact field trials is extremely brief due to the paucity of published data.

### Ball-On-Ball Repetitive-Impact Test

Laboratory repetitive-impact testing is best exemplified by the ball-on-ball test developed by Blickensderfer and Tylczak (1983) at the U.S. Bureau of Mines. This test has only been casually correlated to marked ball data and ball mill liner data. However, in both cases, the marked ball and ball mill liner data has also shown correlation with several straight abrasion tests. Yet, strong belief still exists in the industry that the ball-on-ball test provides a good indication of the material's resistance to repetitive-impact field conditions.

Figure 8.28 shows a schematic of the ball-on-ball test. In this configuration, the test uses 75-mm (3-in.) dia balls and the drop height is about 3 m (9 ft). Other users have adopted this test to 150 mm (5 in.) dia balls and drop heights of 5 m (15 ft) or more. Each impact event produces a cascade of decreasing impacts. The test has two distinct benefits: a large number of impacts is produced in a very short time and a wide range of impact energies is imparted to the whole surface of the test specimen. Considering that the test imparts a spectrum of impact damage to the material, it perhaps more closely mimics field repetitive-impact conditions. However, this hypothesis has not been confirmed by field trials.

In this laboratory test, repetitive-impact testing is performed until the ball breaks or until it spalls excessively. An example of a heavily spalled ball and its detritus is shown in Fig. 8.29.



**Fig. 8.29. A heavily spalled ball and its detritus after extensive repetitive impacts in the ball-on-ball impact test.**

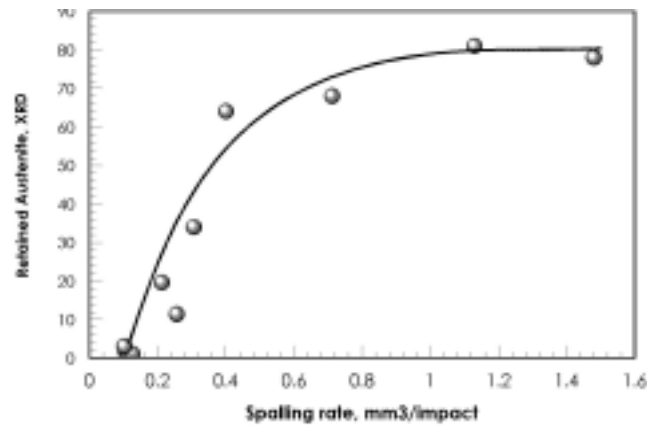
*(Repetitive-Impact Wear Classification, continued)*

A unique feature of this test is the ability to impart damage to the whole surface of the ball. As the ball is repetitively impacted, plastic flow occurs uniformly over the whole surface of the ball. Moreover, due to the random mixing of the experimental and standard balls during the course of the test, the impact distribution between hard or soft or chipped and spalled balls is consistently uniform. Table 8.7 presents an example of some ball-on-ball repetitive-impact data.

Analysis of the data given in Table 8.7 shows some clear trends. If the material is anything but austenitic (WCI-A1), then repetitive-impact toughness is extremely good. For example, the highest repetitive-impact resistances are provided by pearlitic (WCI-A6) materials followed by martensitic (WCI-A10) materials. In many ways, repetitive-impact damage is driven by the buildup of residual stresses due to the transformation of austenite to martensite. Figure 8.30 shows retained austenite data plotted against repetitive-impact (spalling) damage for a variety of AR cast irons. In some ways, the ball-on-ball repetitive-impact test can double as a tool to measure retained austenite, albeit, a very expensive and time-consuming tool.

Another interesting result offered by the ball-on-ball repetitive-impact test is that tempered martensitic materials perform slightly better than untempered irons. This is shown

in Table 8.7 in materials WCI-B9 through WCI B11. As the martensite is tempered (as demonstrated by the iron's decreasing hardness), the iron's laboratory repetitive-impact resistance increases. This is logical, based on standard metallurgical knowledge that tempering increases the toughness of most martensitic structures.



**Fig. 8.30. Retained austenite versus spalling data for AR cast irons. As the retained austenite increases, the spalling damage likewise increases hyperbolically. Trends indicate that below 10% retained austenite, little variation is found in repetitive-impact toughness as measured by the laboratory ball-on-ball test. (After Blickensderfer et al., 1989)**

**Table 8.7**  
**Laboratory Ball-On-Ball Test Data for Various AR Cast Irons and Steel**

| Alloy                       | Composition (Wt%) |      |      |      |      |      | Phase | Hardness<br>HB | Number of<br>Impacts       | Type of<br>Failure       | Volume Loss<br>mm³/Impact |
|-----------------------------|-------------------|------|------|------|------|------|-------|----------------|----------------------------|--------------------------|---------------------------|
|                             | C                 | Cr   | Mn   | Mo   | Ni   | Si   |       |                |                            |                          |                           |
| WCI-A1 <sup>(1)</sup>       | 2.0               | 18.4 | 0.54 | 0.0  | 0.38 | 0.29 | A     | 426            | 9,822                      | breakage                 | 1.13                      |
| WCI-A6 <sup>(1)</sup>       | 2.0               | 18.4 | 0.54 | 0.0  | 0.38 | 0.29 | P     | 466            | 220,000                    | spalling                 | 0.064                     |
| WCI-A10 <sup>(1)</sup>      | 2.0               | 18.4 | 0.54 | 0.0  | 0.38 | 0.29 | TM    | 629            | 100,000                    | spalling                 | 0.15                      |
| WCI-B1 <sup>(1)</sup>       | 3.1               | 17.7 | 0.63 | 1.10 | 0.45 | 0.22 | A     | 415            | <10,000                    | breakage                 | 1.48                      |
| WCI-B4 <sup>(1)</sup>       | 3.1               | 17.7 | 0.63 | 1.10 | 0.45 | 0.22 | P     | 486            | 230,000                    | spalling                 | 0.062                     |
| WCI-B9 <sup>(1)</sup>       | 3.1               | 17.7 | 0.63 | 1.10 | 0.45 | 0.22 | TM    | 737            | 50,000                     | spalling                 | 0.36                      |
| WCI-B10 <sup>(1)</sup>      | 3.1               | 17.7 | 0.63 | 1.10 | 0.45 | 0.22 | TM    | 655            | 115,000                    | spalling                 | 0.13                      |
| WCI-B11 <sup>(1)</sup> 1    | 3.1               | 17.7 | 0.63 | 1.10 | 0.45 | 0.22 | TM    | 618            | 140,000                    | spalling                 | 0.10                      |
| Ni-Hard 4-A1 <sup>(2)</sup> | 2.9               | 8.8  | 1.1  | 0.0  | 5.7  | 1.2  | M + A | 518            | 61,000                     | spalling/<br>breakage    | 0.21                      |
| Ni-Hard 4-A2 <sup>(2)</sup> | 3.5               | 9.6  | 1.1  | 0.0  | 6.5  | 2.0  | M     | 610            | 158,000                    | spalling                 | 0.082                     |
| Forged steel <sup>(3)</sup> | 0.89              | 0.50 | 0.85 | 0.0  | 0.10 | 0.21 | M     | 800            | 64,000<br>81,000<br>63,000 | spalling<br>0.22<br>0.21 | 0.23                      |
| Forged steel <sup>(4)</sup> | 0.73              | 0.20 | 0.73 | 0.0  | 0.13 | 0.15 | M + P | 354            | 300,000                    | flaking                  | 0.0087                    |

A = austenite, M = martensite, P = pearlite and TM = tempered martensite

(1) Data obtained from Blickensderfer et al. (1989).

(2) Data obtained from Laird II (1988).

(3) Data obtained from Laird II (1993).

(4) Data obtained from Blickensderfer et al. (1989).

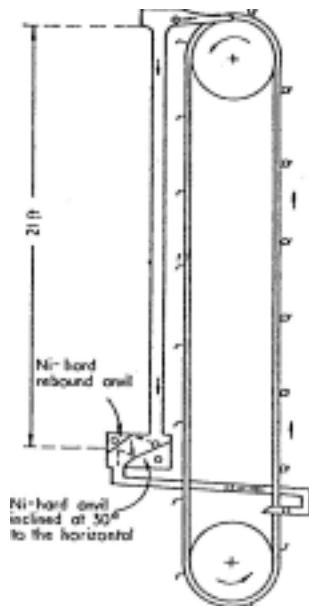
### Ball-On-Block Repetitive-Impact Test

One of the early laboratory testing programs for repetitive-impact resistance was carried out on ball-on-block test equipment. The most frequently cited work of this type is the paper by Dixon (1961). Figure 8.31 shows a schematic of Dixon's test. The test is very similar in form to that of the ball-on-ball except that the impact is much more severe against the heavy Ni-Hard block.

Dixon's work explores the heat treatment effects on one Ni-Hard 1 alloy having a nominal composition of 3.39% C, 0.63% Si, 0.60% Mn, 3.51% Ni, 1.40% Cr and balance Fe. Table 8.8 summarizes this work. From inspection of Table 8.8, it can be seen that the poorest repetitive-impact results are produced from irons that have significant amounts of austenite (as-cast M+A) or are excessively brittle (all martensite, M). This result parallels the data provided in Table 8.7 for high-Cr (WCI-A1 and WCI-B9) AR cast irons used in the ball-on-ball test. As the retained austenite content is reduced, the repetitive-impact resistance increases. The best results are produced when the structure is either tempered martensite or stable bainite. In both cases, we avoid transformation-induced stresses during repetitive-impact testing.

### Qualitative Understanding of Field Wear Vs. Laboratory Testing

In many respects, laboratory repetitive-impact tests provide a good measure of the iron's retained austenite content. If the iron contains any significant fraction of retained austenite, its laboratory repetitive-impact life is extremely short. This trend is shown in Fig. 8.30 and is commonly found in every laboratory repetitive-impact test.

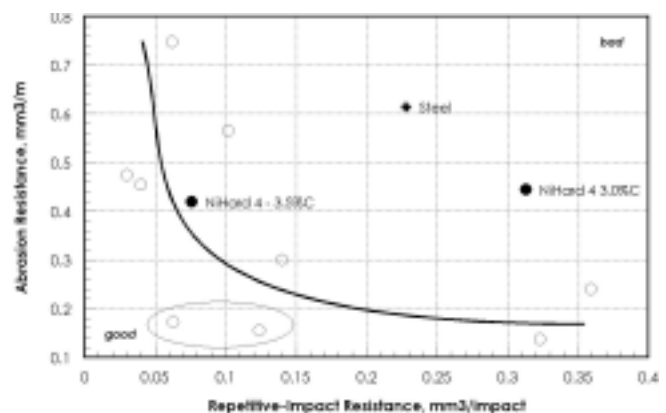


**Fig. 8.31. Ball-on-block repetitive-impact testing equipment.** A single ball drops approximately 21 feet on to a fixed wear test block. The test runs until the wear block fractures or wears significantly. Alternatively, balls can be evaluated in a similar manner (i.e., fracture or wear). In Dixon's test, balls were tested and had nominal diameters of 38 mm (1.5 in.) and were impacted until they fractured. (Schematic courtesy of Dixon, 1961)

Industrially, most users understand that high-volume fractions of retained austenite are extremely perilous in irons subjected to repetitive impact. In this regard, the laboratory repetitive-impact test has done a good job with simulating the premature failure of industrial components. These rather coarse, order-of-magnitude trends start to dissipate at the lower retained austenite contents.

For example, following the data shown in Tables 8.7 and 8.8 and Fig. 8.30, AR irons used in repetitive-impact environments should have extremely low retained austenite content. This is not exactly true, for a number of reasons. To obtain low retained austenite content in AR irons, a duplex heat treatment is required. Subsequent to the high-temperature destabilization stage, the resulting mixture of austenite and martensite must be subjected to a subcritical heat treatment that tempers the existing martensite and encourages the remaining austenite to further transform to martensite. The resulting structure is extremely low in retained austenite, but its hardness has suffered. For example, closer inspection of the data presented in Table 8.7 shows that high repetitive-impact resistance comes at the expense of hardness. As hardness suffers, abrasion resistance typically degrades, in concert.

For industrial users, it is a balancing act between repetitive-impact resistance and abrasion resistance demands of the application. Figure 8.32 shows an example of laboratory data from a standard pin-on-drum abrasion test and a ball-on-ball repetitive-impact test. In most cases, you either get good abrasion resistance or good repetitive-impact resistance. In a special case, both good abrasion and repetitive-impact resistance are shown for a couple of irons. Irons that exhibit these favorable laboratory characteristics have near eutectic compositions and have been duplex heat treated. What is most interesting is that similar trends have been shown industrially with many users switching to irons having compositions around the eutectic and low retained austenite contents for applications requiring significant repetitive-impact resistance.



**Fig. 8.32. Combining good laboratory abrasion resistance and repetitive-impact resistance is difficult.** In most cases, there will be either good abrasion resistance or good repetitive-impact resistance. The unmarked data points represent high-Cr AR irons. The two irons within the ellipse have a nominal composition of 3% C, 20% Cr and 2% Mo and were duplex heat-treated, to lower the retained austenite to below 10%. (After Laird, 1993)

**Table 8.8**  
**Laboratory Ball-On-Block Test Data<sup>(1)</sup>**

| Heat Treatment                              | Phase     | Estimated RA <sup>(1)</sup> | Hardness DPN (HB) | Average Life-to-Fracture in Drops |
|---------------------------------------------|-----------|-----------------------------|-------------------|-----------------------------------|
| As Cast                                     | M + A     | 50%                         | 715 (600)         | 487                               |
| 4 hr at 275C (527F)                         | M + B + A | 10%                         | 750 (630)         | 3070                              |
| 46 hr at 275C (527F)                        | TM + B    | 0%                          | 770 (650)         | 4350                              |
| 4 hr at 850C (1560F)                        | M         | 0%                          | 800 (681)         | 295                               |
| 4 hr at 450C (840F)<br>4 hr at 275C (527F)  | B + TM    | 0%                          | 670 (560)         | 7200                              |
| 4 hr at 750C (1380F)<br>4 hr at 275C (527F) | TM        | 0%                          | 790 (670)         | 5650                              |

*A = austenite; M = martensite; P = pearlite; B = bainite; TM = tempered martensite.*

*(1) Data obtained from Dixon (1961) on Ni-Hard 1 balls having a nominal composition of 3.39% C, 0.63% Si, 0.60% Mn, 3.51% Ni, 1.40% Cr and balance Fe. Balls were cast from one foundry heat into 37-mm (1.5-in.) dia green sand molds.*

*(2) Estimated retained austenite (RA) content is inferred from microstructural notes provided by Dixon (1961).*

## Designing Microstructures for Repetitive-Impact Resistance

### Matrix Structure

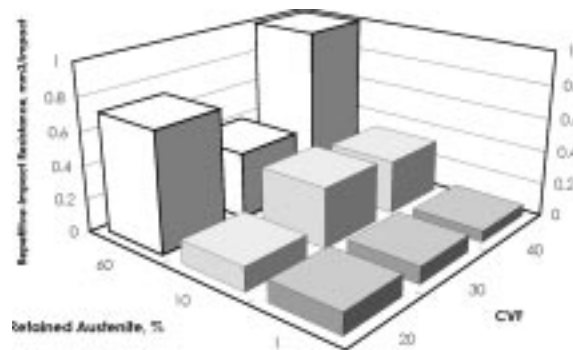
With regard to repetitive-impact resistance, retained austenite is everything. As mentioned, the real damage is done by the strain-induced phase transformation of the retained austenite to martensite as the iron is repetitively impacted. As this retained austenite transforms to martensite, the resultant 4% volume expansion creates high subsurface stresses within the iron. It is this volume expansion that is believed to cause severe subsurface cracking, leading to high spalling loss or catastrophic failure of the casting.

Laboratory repetitive-impact data tends to support these arguments that the matrix dominates the repetitive-impact performance of the material. It is this overwhelming reliance on the matrix structure over all other microstructural considerations that leaves little room for repetitive-impact optimization of AR cast iron structures.

In fact, if we plot repetitive-impact resistance against retained austenite and CVF, Fig. 8.33 leads one to believe that the CVF of the iron has little affect on its laboratory repetitive-impact resistance. This result confirms the importance of controlling the retained austenite content of irons subjected to severe repetitive-impact damage.

### CVF, Carbide Size and Morphology

Considering the dominance of the matrix structure in repetitive-impact situations, there is little comment concerning the carbide side of the equation. It is quite surprising considering the importance of CVF on fracture toughness, that carbide morphology would have so little affect on repetitive-impact damage. Nevertheless, laboratory data on repetitive-impact resistance is quite clear showing the nil affect of carbide morphology on repetitive-impact damage. It would appear that once the matrix has been stabilized (i.e., the retained austenite lowered to below 10%), almost any AR cast iron will survive successive impacts.



**Fig. 8.33. Regardless of the iron's CVF (CVF), the laboratory repetitive-impact resistance is governed by the iron's retained austenite content. (After Blickensderfer et al., 1989).**

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# Chapter 9

## Corrosive/Abrasive Wear Performance

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Corrosion can be defined as a chemical process that results in the gradual wearing of materials. Since AR irons are not typically used in applications where corrosion alone is an issue, we will ignore the issue of *pure* corrosion. Abrasion-resistant irons are typically placed into abrasive environments that, depending on the process, might be mildly or severely corrosive; nevertheless, the overall dominant wear mechanism is abrasion.

It is the synergism between corrosion and the abrasive process that makes the corrosive performance of AR irons so interesting.

It is the focus of this chapter to explore the relationship between corrosion and abrasion in AR irons and arrive at a better understanding of how composition and microstructure relates to corrosive/abrasive performance. (See Glossary for corrosion)

### Corrosion/Abrasion Relationship

If an iron or steel contains more than 12% Cr within its matrix, a very thin  $\text{Cr}_2\text{O}_3$  film will form over the surface of the casting. This protective film acts as a barrier to further corrosion or, in corrosion terminology, “passivates” the material. However, in an abrasive environment, this protective film is quickly removed, exposing the unpassivated material, which can then quickly corrode before subsequently passivating. In the end, under severe corrosive/abrasive conditions, a high-Cr iron or steel will wear no better than a less expensive low-Cr iron or steel.

In this context, it should be noted that even under so-called “noncorrosive” conditions, some corrosion takes place because air or the humidity in the air have a corrosive action. Laboratory experiments in the wear pot by Uetz and Föhl (1972, 1973) showed that the presence of air instead of inert gas increased wear of carbon steel up to three times. Increased wear was observed by these authors also when surface-active organic vapors were introduced. This phenomenon is difficult to explain by a simple corrosive action, and it indicates that factors other than corrosion play a role, too.

The presence of water in the abrasive is well known to increase wear dramatically, as shown not only in many laboratory experiments like those by Uetz and Föhl, but also in numerous crushing and grinding operations. In impact crushing hard rock to gravel, an increase of the water content by 1% results in 15% higher wear of the high-Cr iron impact bars (Röhrig, 1971), and wet milling consumes up to ten times as

much metal as dry grinding the same ore (Dodd and Dunn, 1983). The influence of the atmosphere can also be observed in applications such as ball mills. Grinding in inert atmosphere reduces corrosion and total wear rate. Grinding in oxygenated atmospheres increases corrosion as  $\text{O}_2$ -pressure increases (Remark and Wick, 1976, Dunn, 1982).

### Corrosion Resistance

When most people think of a corrosion-resistant material, they think of stainless steel. When it was first discovered in 1912, its stainless qualities were noted by a resistance to staining in a solution of nitric acid ( $\text{HNO}_3$ ). It wasn't until much later that scientists noted that this corrosion resistance was due to the formation of a very thin  $\text{Cr}_2\text{O}_3$  film over the surface of the material. The thickness of this film is in the realm of nanometers and it is transparent to the observer. This film acts as a barrier to the corrosive environment and prevents further subsurface dissolution. Hence, after a stainless material is polished, it will appear mirror-like for many days.

The key to the formation of this film is the presence of oxygen in the corrosive liquid, either as free oxygen or as a molecule within the liquid compound. If the liquid is free of oxygen, the protective  $\text{Cr}_2\text{O}_3$  film cannot form, and the material is free to corrode. In the case of ferrous-based materials, we are referring to the dissolution of iron atoms or the breaking of covalent bonds between the iron atoms via a chemical reaction. The end result is that the iron atoms are free to float off into the corrosive solution. Consequently, depending on the ability of the acid to create an environment poor or rich in free oxygen, different acids will have different corrosive rates.

### Oxidizing/Reducing Environments

Acids are classified by their ability to form oxidizing or reducing environments. Nitric acid is classed as an oxidizing acid since, when mixed with water or alcohol, it forms a solution rich with free oxygen as it attacks a ferrous material. Conversely, sulfuric acid ( $\text{H}_2\text{SO}_4$ ) and hydrochloric (HCl) are classed as reducing acids since they tend to form chemical solutions that are poor in oxygen. (More information on *Oxidizing and Reducing Environments* can be found in the Glossary.) As can be expected, stainless steels perform far better in oxidizing chemical solutions where there exists a surfeit of free oxygen that is readily available for the formation of a protective  $\text{Cr}_2\text{O}_3$  passive film over its vulnerable

ferrous substrate than they do in environments where the chemical solution is dominantly reducing.

On a practical basis, this behavior shows up in the use of various chemical etchants for metallographic sample preparation. Solutions of nitric acid (2% nital) work exceedingly well on low-Cr alloys to reveal microstructural details, while solutions of HCl are recommended for high-Cr and stainless alloys. In the former, the nitric acid will readily dissolve the martensitic or ferrite phase creating topological contrast that allows easy phase identification by the microscopist. In the latter, the protective passive film that forms on high-Cr and stainless alloys resists nitric acid attack, so HCl solutions should be used to break down this film to reveal the ferrous substrate.

Chromium Threshold

A threshold matrix Cr content of 12% is sufficient to make the steel stainless in air or water, that is to protect it against rust. For corrosion resistance in acidic or halide-containing environments, like sea water or chloride-contaminated waters,

higher Cr content is required. One of the most severe types of corrosion is localized attack by chlorides in the form of pitting or crevice corrosion, or stress-corrosion. During corrosion by pitting, chloride ions destroy the protective surface film locally and create a pit where an extremely corrosive, highly acidic environment persists, which will penetrate into the material. In order to make a steel resistant against this type of corrosion (e.g., stagnant sea water), a minimum of 30% Cr + 3% Mo in the matrix are required.

For moderate corrosion resistance, more than 12% Cr dissolved in the matrix is necessary. Since a substantial amount of Cr is bound in the carbide phase, a higher bulk Cr content is required. The higher the C content, the higher must be the total Cr content of the alloy. Küttner (1933) investigated this problem and found that a minimum %Cr of  $15 + 15 \times \%C$  is necessary to provide a reasonable corrosion resistance in most oxidizing environments.

For example, when the iron solidifies, regions surrounding the carbides will become slightly depleted in C and Cr. Likewise, during high-temperature heat treatment, the formation of small, secondary carbides will also cause a slight depletion of these alloying elements. In stainless steel terminology, the formation of these secondary carbides induces a form of “grain-boundary sensitization.” The matrix regions immediately adjacent to these carbides dips below the 12% Cr threshold and becomes prone to corrosive attack. In AR cast irons, carbides exist throughout the microstructure, regardless of any and all heat treatments. Consequently, to avoid this sensitization response, the practical matrix Cr content is probably closer to 14% than 12%.

These findings were translated into commercial cast high-Cr iron alloys and Roesch (1936) published a diagram indicating the limiting Cr and C contents for rust-free irons, and irons that are corrosion resistant in oxidizing acids. These border-

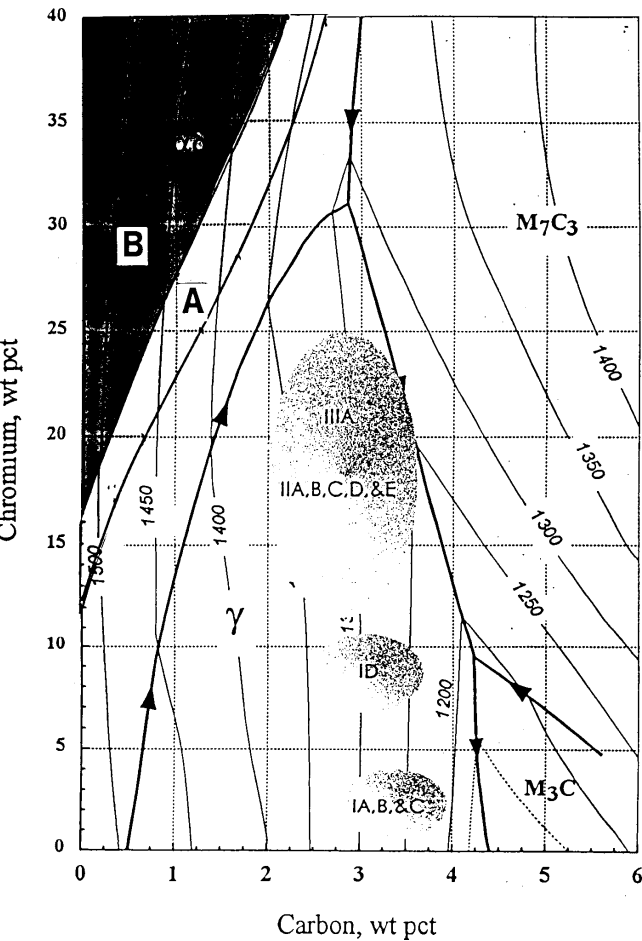


Fig. 9.1. Fe-Cr-C liquidus phase diagram with standard white iron composition ranges and regions of alloys that are rust-free (region A) or corrosion resistant in oxidizing acids (region B). (after Roesch, 1936)

Table 9.1  
Microchemical Analysis of Selected AR Cast Irons

| Alloy                                                    | —Melt— |       | —Matrix— |       |
|----------------------------------------------------------|--------|-------|----------|-------|
|                                                          | %C     | %Cr   | %C       | %Cr   |
| 440C                                                     | 1.1    | 17    | 0.3      | 13.5  |
| Low-Carbon<br>15% Cr Iron <sup>(1)</sup>                 | 1.87   | 14.69 | n.a.     | 12.47 |
| ASTM A532<br>Class II-D                                  | 2.2    | 18.4  | 0.6      | 13.0  |
| ASTM A532<br>Class II-D                                  | 3.0    | 20.2  | 0.8      | 11.0  |
| ASTM A532-A<br>Class III-A                               | 3.1    | 26.2  | 0.5      | 14.0  |
| Ferritic<br>High-Cr Iron                                 | 2      | 30    | 0.1      | 25    |
| n.a. = not analyzed                                      |        |       |          |       |
| <sup>(1)</sup> Analysis data after Kuwano et al. (1990). |        |       |          |       |

lines have been inserted into the liquidus phase diagram for Cr-C-alloys in Fig. 9.1.

According to Fig. 9.1, none of the standard white irons can be considered as really corrosion resistant. Table 9.1 provides direct microchemical analyses. It shows that, in fact, none of the irons would qualify as corrosion-resistant, based on their matrices' Cr content. However, the high-Cr content offers some protection against corrosion so that the corrosion rate decreases with increasing matrix Cr content. In general terms, a Class III Type A iron with 25% Cr can be expected to corrode less than a Class II Type B iron with 15% Cr, which, in turn, will be more resistant than a Class I Type B iron with only 4% Cr. The practical consequences depend upon the specific environment and abrasive situation.

Figure 9.2 shows an example of the effect of Cr content and abrasive, determined by the wear of grinding balls in a laboratory ball mill (Rajagopal and Iwasaki, 1992). The marked ball wear tests were conducted in a 203-mm diameter porcelain mill using various Cr-containing cast iron balls, all nominally 25.4 mm in diameter. The porcelain mill was charged with 126 balls including 14 marked ones. Before their use, the surface layer of the balls was removed by repeated wet grinding of quartzite samples for 48 hours. To control the atmosphere inside the mill, either nitrogen or oxygen was introduced through a hole drilled at the closed end of the mill.

Marked ball wear tests were performed on samples consisting of quartzite, quartzite plus 5% pyrrhotite or 5% chalcocopyrite crushed to -10 mesh. The mill charge consisted of a 600-g sample and 400 ml of 0.05M sodium sulfate solution at 60% solids. Each test was run for one hour. Wear under oxygen was higher than under nitrogen, especially in the mixtures containing pyrrhotite or chalcocopyrite at low Cr content. A marked transition can be observed in the region of 8–15% Cr, and wear tended to decrease with increasing Cr content over the whole range. The exact mechanism of the high wear rate under oxygen in the presence of pyrrhotite or chalcocopyrite is not quite clear. Maybe it was due to an interaction between these sulfidic minerals and oxygen.

In response to more severe corrosion requirements, some users specify a ferritic high-Cr iron. A typical composition is given at the bottom of Table 9.1. Although this iron forms a network of AR hard carbides ( $\text{Cr}_{23}\text{C}_6$ ), the matrix is quite soft (ferrite). Nevertheless, the iron is fairly corrosion resistant in oxidizing and some reducing environments, and provides a measure of abrasion resistance. Figure 9.3 shows a schematic of the corrosion resistance of a ferritic high-Cr AR cast iron (0.8% C, 29% Cr) in nitric and sulfuric acids. This iron meets the requirements of Fig. 9.1 for corrosion resistance and represents an example of the corrosion resistant behavior (its matrix contains over 20% Cr). It does give some idea of what is possible and not possible. Alloys of this type have been standardized, e.g., Germany in SEW 410, and have found extensive use, e.g., in phosphate processing (Desestret et al., 1994) and in some FGD installations. In general, if more

complete corrosion information is required, the literature on stainless steels can be consulted (e.g., *ASM Corrosion Handbook*).

It should also be mentioned that the super high-Cr ferritic wear irons are resistant to stress corrosion cracking. Most issues surrounding stress corrosion cracking relate to structural components that experience tensile stresses due to loading or faulty heat treating. Under such conditions, corrosive fluids can greatly lower the fracture toughness of the material allowing relatively low tensile stresses to cause catastrophic crack propagation. The mechanisms of stress-corrosion cracking are not entirely well understood. In brief, alloys that do not suffer from sensitization (i.e., a near-perfect passive film forms over the surface of the alloy) do not significantly suffer from stress-corrosion cracking in most corrosive environments.

The role of the metallic carbide in the corrosion process should be mentioned here. In Ni-Hard 4 and all high-Cr irons, the carbide matrix is of the  $\text{M}_7\text{C}_3$  type. As discussed in Chapter 5 on Morphology and Microstructure, these carbides contain a minimum of 25% Cr. Consequently a passive film will naturally form over the surface of these carbides under corrosive conditions. Nevertheless, if the matrix is low in Cr (a prime example is the Ni-Hard 4 irons where the matrix Cr content is around 3% Cr) the corrosion process can actually be accelerated by the formation of a galvanic couple between the passive carbides and the active (corroding) matrix. If we consider only irons that contain more than 12% Cr within their matrices (see Fig. 9.1) this problem is avoided and the iron can be considered truly corrosion resistant.

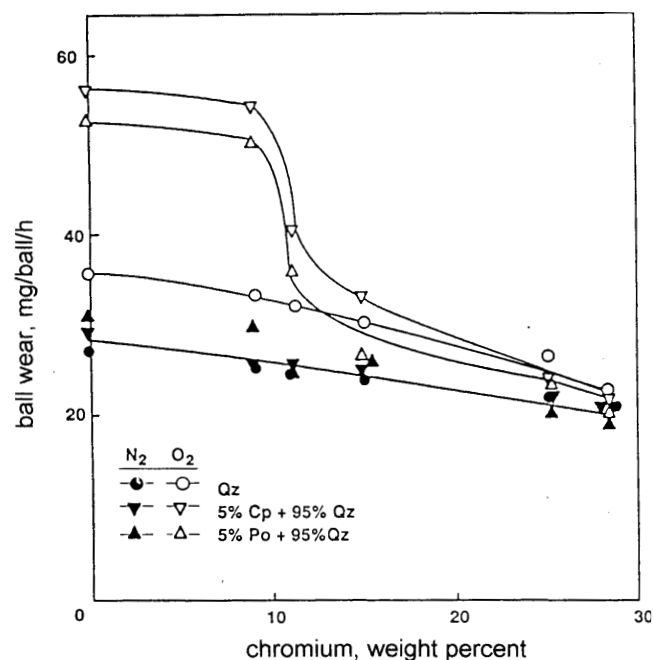


Fig. 9.2. Effect of Cr content and abrasive medium on the wear of grinding balls in a laboratory ball mill, under nitrogen or oxygen atmosphere. (After Rajagopal and Iwasaki, 1992)

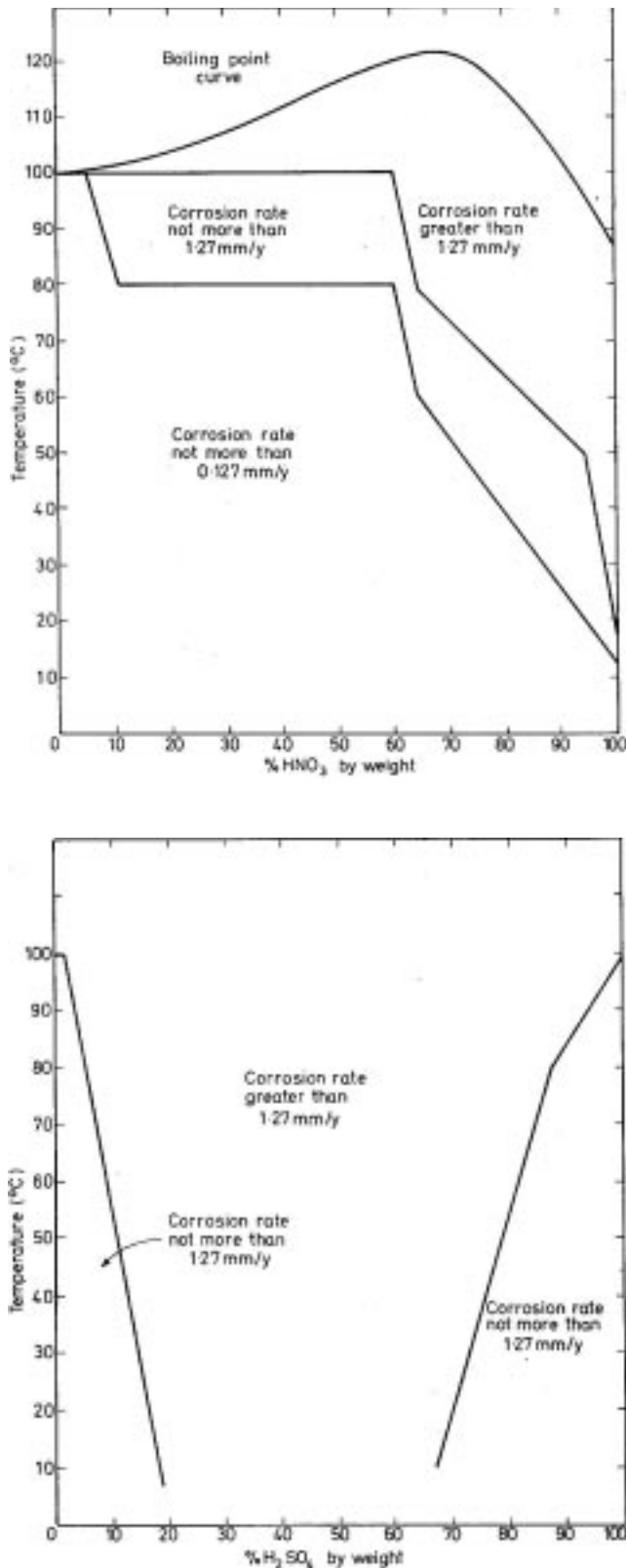


Fig. 9.03. Corrosive behavior of a very high Cr AR cast iron (0.8% C, 29% Cr). The corrosion rate is given in millimeters of surface volume lost per year. Although good corrosion protection is offered in an oxidizing acid, almost no corrosion resistance is offered when a reducing acid is used. This trend is identical to that shown by stainless steels, where corrosion protection is provided by a CrO passive film. (Courtesy of Bergische Stahl-Industrie, Remscheid, Germany, 1957)

## Behavioral Studies

### Passivation Behavior

For an iron to remain passive within a corrosive medium, the protective passive film must remain intact on the surface of the iron. This seems almost too simplistic to state but it is an important concept to focus on when we start discussing the effects of abrasion on the corrosive behavior of AR irons. For example, fundamental studies on the repassivation rates of 304 stainless steel have shown that light scratching doesn't always remove the passive film. That is, under some mild abrasive conditions the  $\text{Cr}_2\text{O}_3$  film can be quite tenacious. Figure 9.4 shows a few results from this study. Depending on the aggressiveness of the scratch (where the depth of the scratch is equal to its width), the corrosion behavior can remain completely passive. Logically, as the scratch gets wider and deeper it eventually starts to remove the passive film.

One way to measure the combined effects of corrosion and abrasion is the slurry pot wear test as shown in Fig. 9.5a. (This will be further discussed in the section on Measuring Wear Rates.) In this work, the slurry is pumped through the pot by a polymeric impeller. Samples and standard reference materials are placed round the perimeter of the pot, where they come in direct contact with the abrasive slurry (Fig. 9.5b). During operation, the samples' passivation behavior is monitored by a *potentiostat* (see Glossary). This instrument measures the electric current flowing through the sample into the slurry, while also measuring the voltage difference or potential between the sample and the corrosive liquid. By measuring the current flow and potential, one can determine whether the sample has passivated or not. That is, when a material passivates, the passive film forms an electrical insulating layer around the sample. When the sample is corroding, electrons can easily flow through the sample and into the slurry. When the sample is passivating (in the passive condition), the current flow is greatly reduced.

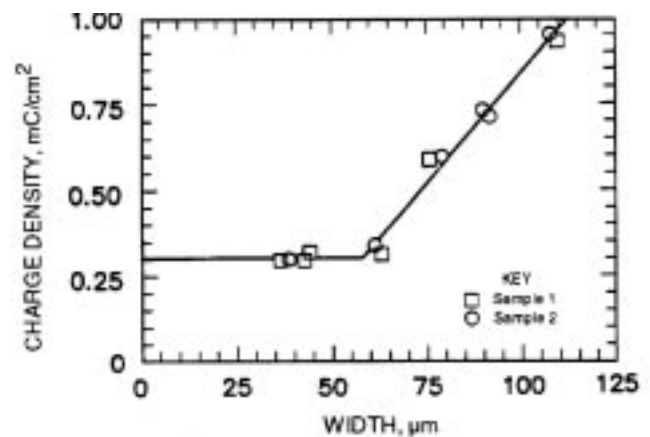


Fig. 9.4. The 304 stainless steel remains relatively passive (low charge density) under light scratching conditions. As the scratch width increases above 60 microns, the passive film is increasingly removed and the rate of corrosion increases (the charge density increases). (After Adler and Walters, 1993)

Regardless of the electrochemical measurements, the end result of a slurry pot wear test is a weight loss measurement. The weight loss or volume loss (dividing the weight loss by the density of the material) is then touted as the corrosive/abrasion resistance of the material to slurry wear. Figure 9.6 shows an example of test data collected using the slurry pot test. Very simply, the slurry pot wear test favors harder materials.

### Synergistic Behavior

It would seem from the data presented in Fig. 9.6 that corrosive wear plays only a small part of the corrosive/abrasive process. This is not exactly true. Laboratory and field data lend credence to the idea that a synergism occurs between the corrosive and abrasive processes that significantly affects the total wear resistance of the material.

This synergistic behavior was investigated by Madsen (1988) using a potentiostat-equipped slurry pot. By controlling the potential and, therefore, the current between the

sample and the corrosive liquid, the corrosion process was significantly slowed. Basically, Madsen was supplying the sample with enough electrons that the corrosive liquid had little effect on the sample (this is termed *cathodic protection*). By negating the corrosion process, the slurry pot was then measuring just the abrasive process.

By conducting pure abrasive and then standard corrosive/abrasive tests, the synergistic behavior that occurs between the corrosive and abrasive processes can be estimated. Figure 9.7 graphically presents this data. In all three alloys, the combined corrosive/abrasive wear rate is about one-third greater than the additive wear rates of straight corrosion and abrasion. This magnification factor is termed a synergistic effect.

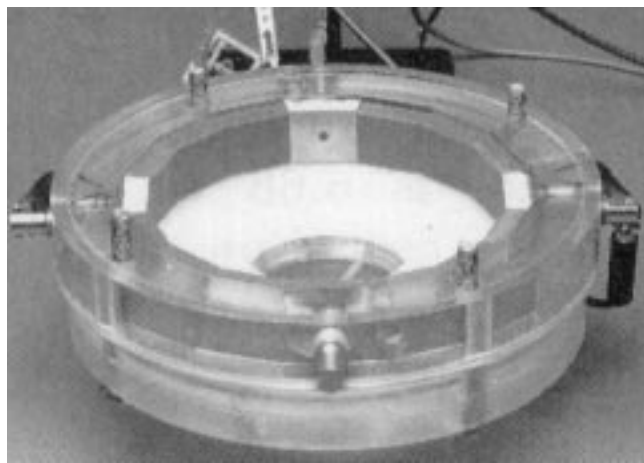
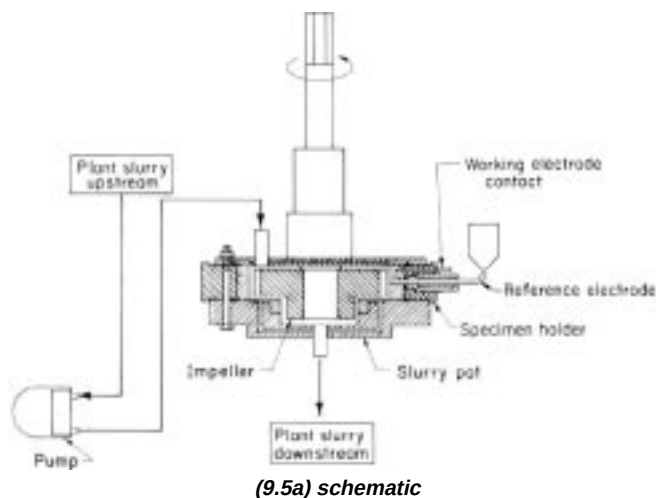


Fig. 9.5. Experimental apparatus to measure the corrosion and wear behavior of materials in a slurry environment (a). The samples have dimensions of about 25x25x10 mm (1x1x0.375 in.) and are arrayed around the perimeter of the slurry pot. Abrasive slurry is then introduced into the pot, where a polymeric impeller drives the slurry against the samples (b). (After Madsen, 1989)

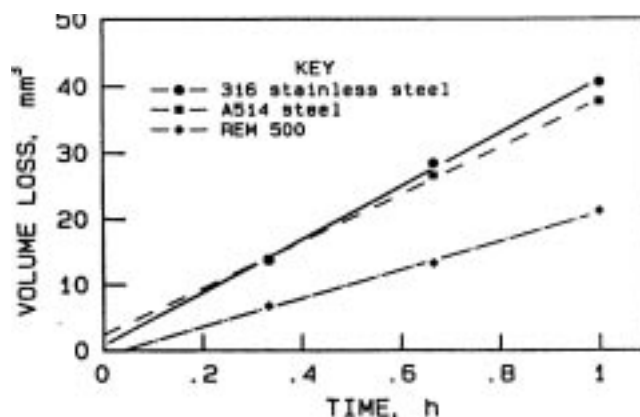


Fig. 9.6. Corrosive/abrasive wear of three alloys in a slurry pot, using a mixture of tap water and 2% silica sand. The hard martensitic steel (REM 500–HB 500) shows much better wear than the soft corrosion-resistant 316 stainless steel (HB 125). What is more surprising is that even the slightly harder A514 (HB 280) shows better corrosive/abrasive performance. (After Madsen, 1988)

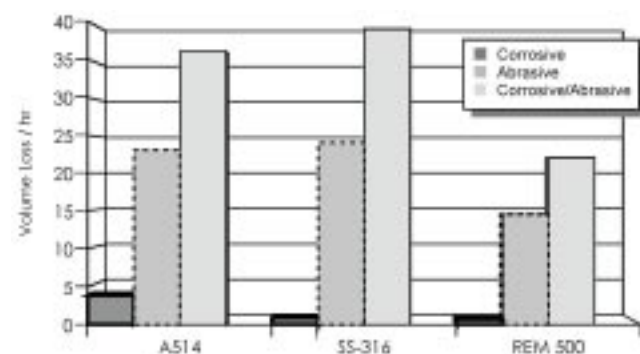


Fig. 9.7. A look at the synergistic behavior between corrosion and abrasion for three alloys evaluated in a slurry pot. The wear rate is given in cubic millimeters of material lost per hour, using a slurry consisting of a 2 wt% silica sand in a weak sodium sulfate solution (0.06 molar). A potentiostat was used to cathodically protect the ferrous samples and obtain the pure abrasive wear measurements. A synergistic effect is shown in all three samples (i.e., the corrosive/abrasive wear loss is greater than the combined wear effects of corrosion and abrasion). (After Madsen, 1988)

Although corrosive/abrasive synergism is known to be responsible for increased wear rates in many mineral processing environments, its mechanisms are quite complex. In general, corrosion processes are accelerated whenever a clean fresh surface is presented to the corrosive liquid. The converse of this behavior holds whenever a heavy scale is allowed to build up on the surface of the casting. In severely abrasive environments, the surface is constantly being scoured, presenting fresh unpassivated material to corrosive attack (whether or not the material contains more than 12% Cr within the matrix). As one can imagine, a combined corrosive/abrasive process will always produce greater wear rates than either straight corrosion or straight abrasion.

A more specific explanation of synergistic mechanisms in AR cast irons and steels is offered by Gates et al. (1995). Figure 9.8 provides a quick look at their proposed synergistic mechanisms. In brief, synergism occurs because the matrix suffers from corrosive attack and no longer supports the carbides against abrasive attack. This behavior was found to

be most severe in the low-Cr pearlitic steel, where no corrosion protection is offered and the ferritic phase between the  $M_3C$  carbide plates is rapidly corroded, leaving the carbide plates exposed to abrasive attack.

### Microstructural Behavior

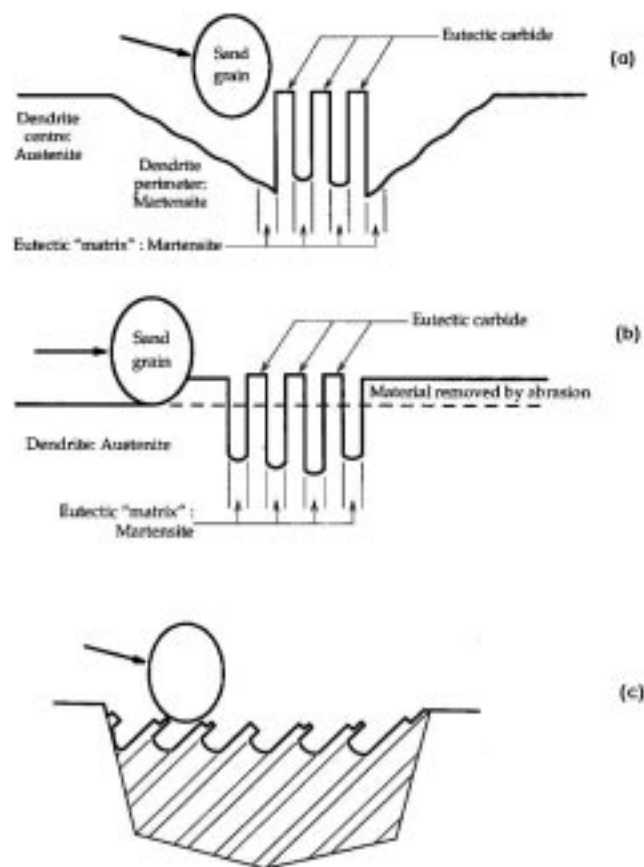
Even with the corrosion protection offered by the formation of a passive film in the high-Cr irons (1.4% C, 15% Cr), this film offered few benefits since it was quickly abraded away during the wear test. Interestingly, the work by Gates et al. showed that the martensitic phase was much more prone to corrosive attack than the austenitic phase. This was found to have a major influence on the combined corrosive/abrasive wear. That is, the austenitic iron was found to perform about 100% better than the same iron heat treated to have a dominantly martensitic matrix. It should be noted that their corrosion/abrasion test was quite severe, using a wet version of the ASTM dry sand rubber wheel test, where the liquid was a 10% nitric acid ( $HNO_3$ ) mixture.

This selected form of corrosive attack is sometimes referred to as *pitting corrosion*. In the case of AR irons, the martensitic phase will congregate around the eutectic carbides due to a number of microstructural factors. Selective corrosive attack of this phase leads us to the condition shown in Fig. 9.8a with the  $M_7C_3$  eutectic carbides standing unsupported above the matrix.

In summary, we can say that the corrosive/abrasive performance of an alloy is little influenced by whether or not it forms a passive film but is more influenced by standard microstructural factors such as CVF, hardness and various phase fractions of austenite, martensite and pearlite. That is, in aggressive abrasive environments, where AR cast irons are typically used, the abrasive quickly removes any passive film that forms on the surface of the material. Neglecting this effect leaves us with the standard microstructural mechanisms to explain why ferrous-based materials wear faster when subjected to corrosive/abrasive environments.

The most dominant mechanism for the difference between straight abrasive wear and combined corrosive/abrasive wear is just accelerated corrosive wear. Since the abrasive process is continually creating clean, unprotected surfaces, the corrosion process can proceed completely unhindered and, in the case of AR irons, selectively attack the martensitic phase, leading to even further accelerated corrosive/abrasive wear.

Further discussion will differentiate between abrasive/corrosive wear in rock crushing, especially ball mills, and in handling abrasive/corrosive slurries ... like in pumps, fittings and mixers. The overall tribological system in both groups of application is widely different with high abrasion rates and high impact and surface pressure in crushing, and low impact but potentially highly corrosive media in slurry handling.



**Fig. 9.8. Corrosive/abrasive processes in (a) martensitic, (b) austenitic AR cast irons containing more than 12% Cr in the matrix and (c) pearlitic steels. In martensitic irons, the martensitic phase is more aggressively attacked by the corrosive process, leaving the eutectic carbides exposed to abrasive impact. Austenitic irons provide better corrosion resistance around the eutectic carbides, resulting in better protection. In low-Cr pearlitic steels, the ferrite phase is aggressively attacked leaving the soft  $M_3C$  carbides unsupported and easily fractured during abrasive wear. (After Gates et al., 1995)**

## Measuring Wear Rates

### Ball Mill Wear

Although much of the market for grinding media is now dominated by martensitic 1% C steel, high-Cr AR cast irons were once considered for many ball mill applications. Even though high-Cr irons showed superior wear performance, their cost outweighed their wear performance. Nevertheless, laboratory and field data on their corrosive/abrasive wear behavior in ball mills is very illuminating as to the wear processes occurring in these environments.

Going back to the work by Norman and Loeb (1948) on grinding media, their wear performance data is perhaps a little bit more understandable. First, from a corrosion viewpoint, none of their materials had any corrosion resistance or, for that matter, resistance to corrosive pitting wear. Second, their AR cast irons were limited to the low-alloy (pearlitic) and Ni-Hard 1 types of mixed microstructures. Consequently, it does not come as much of a surprise that martensitic steel balls were somewhat comparable in service life to the various AR cast irons used in their work.

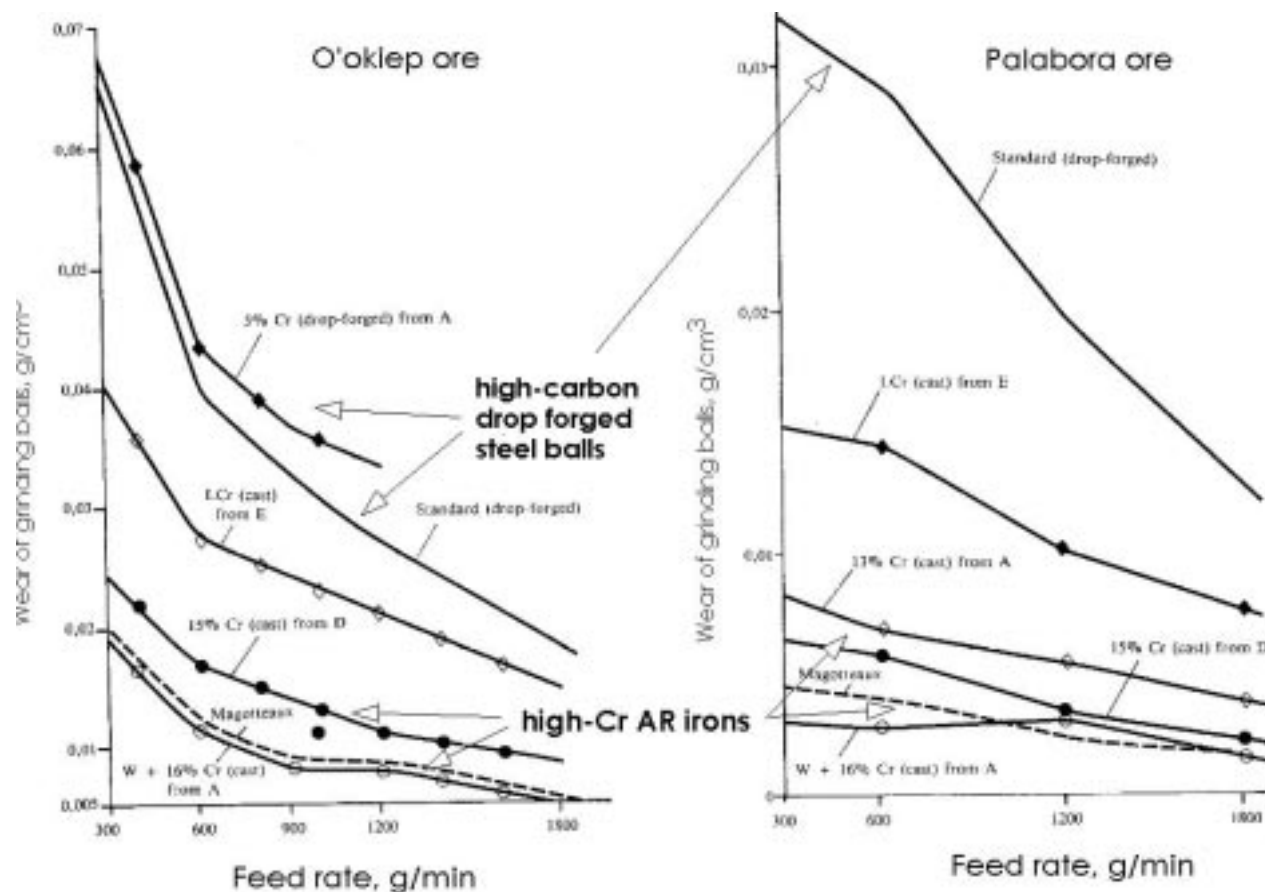
Modern laboratory work, using small-diameter ball mills, has shown that high-Cr irons do offer some benefits. One example is the work done by Howat (1983). In his work, the

ball mill had a diameter of 0.60 m (23.6 in.) and was charged with 200–250 kg (450–550 lb) of 50 mm (2 in.) balls. In this test, ore was fed into the ball mill at different rates, with the water feed adjusted to maintain a water/ore mixture that contained 75% solids. Figure 9.9 shows an example of the data gathered by Howat, using two different grades of ores. The data presented in Fig. 9.9 clearly indicate that high-Cr irons provide superior wear resistance over that of high-C drop forged balls. However, as mentioned by Howat (1983), economics preclude their use.

In other laboratory work aimed at simulating the ball mill grinding process, the corrosion rate was calculated while the sample was abraded by an ultrasonic vibrating steel rod. Table 9.2 presents a summary of Tolley's et al. (1984) work.

The abrasive/corrosive wear mechanism used in Tolley's work was an ultrasonic vibrating steel rod placed on top of the wear sample. The steel rod and sample were both immersed into a synthetic mill solution containing a mixture of various salts. Wear tests were then conducted while monitoring the electrochemical behavior of the wear sample with a potentiostat.

Results from Tolley's work showed that high-Cr irons did remarkably well. More interesting is the fact that the effects of CVF and microstructure were overridden, in favor of corro-



**Fig. 9.9.** Laboratory ball mill abrasion tests using different ores essentially yield the same ranking of materials. In all cases, the high-Cr AR cast iron has the best wear performance but was found not to be cost effective against the high-C drop forged steel balls. (After Howat, 1983)

**Table 9.2**  
**Simulated Ball Mill Abrasion Test Using Vibrating Steel Rod**

| Alloy          | HRC (HB) | Compositional Data |      |                                 | CVF <sup>(2)</sup> | Phase <sup>(3)</sup> | Electrochemical Data               |                                        |
|----------------|----------|--------------------|------|---------------------------------|--------------------|----------------------|------------------------------------|----------------------------------------|
|                |          | %C                 | %Cr  | %Cr <sub>m</sub> <sup>(1)</sup> |                    |                      | Corrosion Rate, mpy <sup>(4)</sup> | Corrosion Potential, mV <sup>(5)</sup> |
| Forged Steel-A | 61 (695) | 0.61               | 0.12 | 0.12                            | 0                  | M                    | 9.82                               | -452                                   |
| Forged Steel-B | 61 (695) | 0.70               | 0.45 | 0.45                            | 0                  | M                    | 8.63                               | -437                                   |
| Cast Steel     | 64 (764) | 0.79               | 0.23 | 0.23                            | 0                  | M                    | 8.80                               | -463                                   |
| AISI 4130      | 24 (247) | 0.31               | 1.12 | 1.12                            | 0                  | P                    | 4.65                               | -430                                   |
| Ni-Hard 1      | 60 (674) | 3.03               | 1.6  | 0.8                             | 22                 | M+A                  | 4.42                               | -379                                   |
| WCI-22% Cr     | 63 (740) | 3.0                | 22   | 11                              | 32                 | M                    | 2.05                               | -360                                   |
| WCI-29% Cr     | 60 (674) | 2.8                | 29   | 16                              | 35                 | M+A                  | 0.335                              | -292                                   |
| WCI-31% Cr     | 56 (595) | 2.35               | 31   | 20                              | 30                 | F                    | 0.141                              | -264                                   |

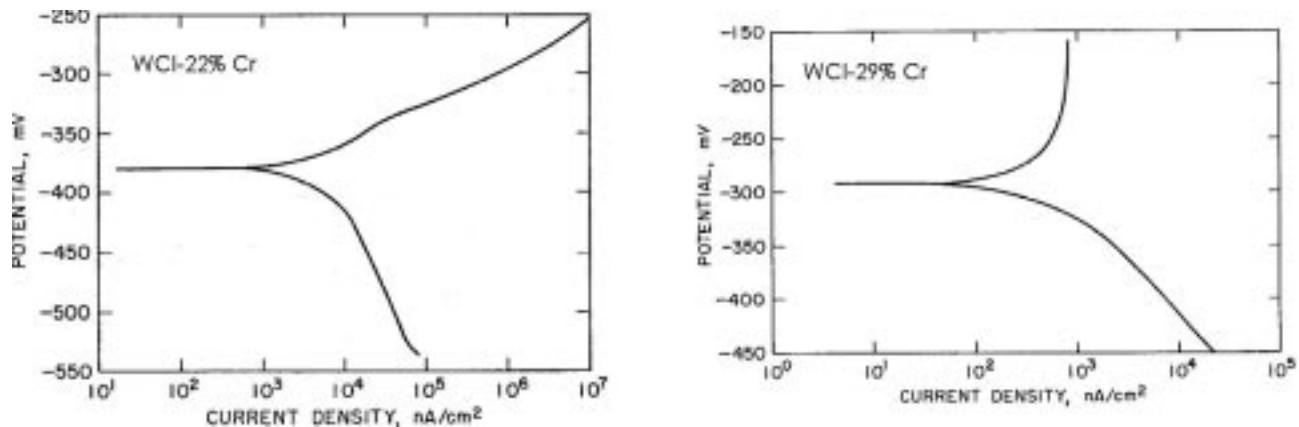
(1) Cr<sub>m</sub> is estimated Cr content in the matrix.  
(2) CVF is calculated carbide volume fraction.  
(3) Phase: M = martensite; P = pearlite; A = austenite; F = ferrite.  
(4) mpy is mils per year (0.001 in.) of material loss on the exposed surface per year of service. In this experiment the weight loss of the sample is measured and divided by its density, times its surface area, times the duration of the test.  
(5) mV is corrosion potential in millivolts. In electrochemical terms, the more negative the potential between the sample and the surrounding corrosive medium, the greater the corrosion rate of the sample. Basically, it reflects the amount of current or voltage potential required to keep the sample from corroding into the surrounding liquid.

sion resistance. The iron (WCI-31% Cr, in Table 9.2) that performed the best had the lowest CVF of the high-Cr irons and had a soft ferritic matrix. Nevertheless, its 20% Cr matrix content provided superior corrosion resistance and, most likely, was able to overcome the weak abrasive mechanism present in the test.

In general, the abrasive/corrosive test (Table 9.2) provided several surprising results. For example, the pearlitic steel (AISI 4130) with a hardness of HB 247 yielded better corrosive/abrasion results than all of the other hard steels, and came close to that of the Ni-Hard 1 alloy. This result contradicts other laboratory and field data that consistently rank the harder steels as more AR under conditions of combined

abrasion/corrosion. Moreover, ferritic steels and irons are perhaps the worst performers with regard to abrasive wear. These findings leads one to reevaluate the abrasive mechanism present in the ultrasonic test. In brief, the abrasive mechanism in this test was, most likely, causing severe plastic deformation over a localized region. Materials that exhibited more *ductility* (pearlitic steels and ferritic irons) fared better in ultrasonic tests than those materials that would tend to chip and spall (martensitic steels and irons). This is not characteristic of most abrasive processes or abrasion tests.

Although their corrosion/abrasion results are suspect, from a corrosion viewpoint, Tolley's et al. (1984) work provides some good data. Figure 9.10 shows two potentiodynamic



**Fig. 9.10. Potentiodynamic scans of two white cast irons given in Table 9.4. The figures reflect the corrosion behavior of the two irons. The formation of a passive film is indicated by the much lower current densities and the much lower potential required to passivate the material.**

scans on the WCI-22% Cr and WCI-29% Cr irons. Their results show that the 22% Cr (matrix Cr content ~11%) iron has little corrosion resistance while the 29% Cr iron (matrix Cr content ~14%) has stainless qualities. Other work by Pearce et al. (1989), using a variety of AR cast irons, confirms these results.

The preceding discussion would tend to negate the importance of corrosion resistance in the wear behavior of grinding media (balls); corrosive wear is often considered to only account for 1–5% of the total wear. This leads one to wonder why wear rates in wet grinding are sometimes 8–10 times greater than dry grinding. Basically, we are back to the synergistic relationship between corrosion and abrasion during corrosive/abrasive wear.

An example of this synergistic behavior is shown in Fig. 9.11, where a sequence of alloys was subjected to laboratory ball milling in a highly acidic mixture of phosphate rock and phosphoric acid. Right from the start, one will notice that the corrosion-resistant alloys perform far better than more traditional AR alloys. In fact, the AR cast irons perform the worst of all the materials studied. In very simple terms, the abrasive component in their small diameter ball mill (60 cm (24 in.)) was extremely weak, while the corrosive component was extremely harsh.

Looking at just the static corrosion rates shown in Fig. 9.11 it is possible to draw a pretty good conclusion that the corrosive/abrasive process in their laboratory grinding mill was dominated by the corrosive component. This piece of laboratory data provides good evidence that, if the abrasive process fails to remove the protective passive film, then corrosion protection will play a significant role in the wear behavior of the material.

The better performance of the commercial ball material over that of the high-Cr material (2.5% C, 13% Cr) can be explained by pitting corrosive wear around the eutectic carbides, leading to a greater corrosive/abrasive wear rate. Nevertheless, the data presented in this laboratory work emphasizes the synergistic nature of corrosive/abrasive wear, whether dominated by abrasion (Howat, 1983) or dominated by corrosion (Singleton and Blickensderfer, 1985). In practice, it is believed that ball mill wear is much more influenced by abrasive factors than corrosive factors.

This situation was discussed by Dodd and Dunn (1983) for ball milling and for other crushing operations by Desestret et al. (1994). They attribute the discrepancies between the results in lab scale testing and full size production mills by the effect of the high impact in the latter and other mechanisms associated with the presence of water. Water can significantly reduce strength and ductility. Tensile tests with many high-strength cast irons and steels, in water or with wet specimens, are known to result in lower values for strength and elongation, and fatigue strength is significantly lower. Water and surface-active agents tend to facilitate crack initiation and propagation that would facilitate detachment of wear particles from the surface. Dodd and Dunn (1983) discussed also that

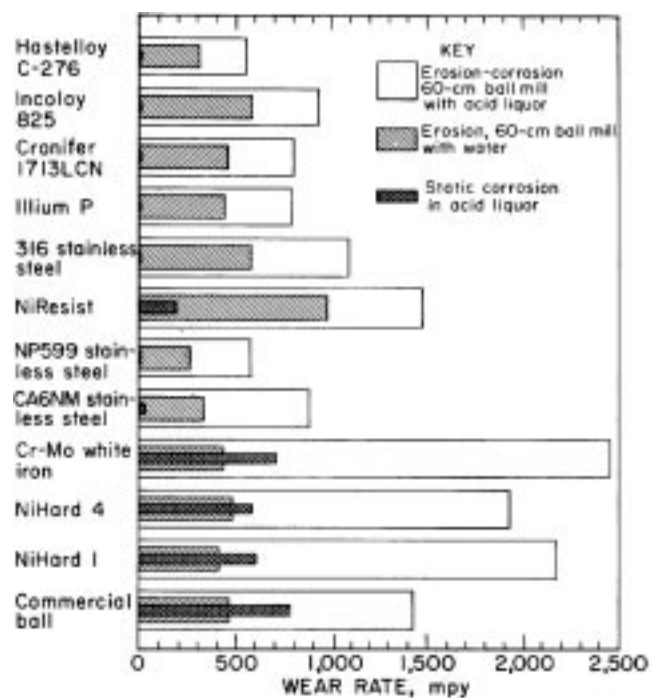
water entrapped between impacting balls could have a cavitation effect. Finally, there is the possibility that water will remove a protective layer of wear debris and crushed minerals from the surface.

## Slurry Wear

Slurry wear is of particular importance to many AR cast iron pump manufacturers and users. One laboratory test that has gained favor in predicting the slurry wear performance of various alloys is the slurry pot or wear cell test discussed in the prior section and shown in Fig. 9.5. An advantage of this test is that either laboratory or industrial slurries can be fed into the slurry pot. Samples and test standards are arrayed around the perimeter, allowing for multiple alloys to be investigated during one test. An example of laboratory and field data is presented in Table 9.3.

The wear rates presented in Table 9.3 are a bit confusing, at first glance. In the laboratory tests, the slurry consists of a dilute solution of sand and water (2 wt.% solids), while in field trials, the industrial slurries were in the range of 20 wt.%. To correlate the two data sets, the wear rate was based on a unit of abrasive solids pushed through the slurry pot. This procedure allows one to directly compare the wear rate of materials under neutral conditions (sand and tap water) with those under industrial conditions.

An experimental variable that has strong influence on the wear rate is the impeller speed of the slurry pot. At low speed and under laboratory conditions, polymers are by far the best



**Fig. 9.11. Corrosive/abrasive laboratory wear data using a 60 cm (24 in.) dia grinding mill. A mixture of phosphate rock and phosphoric acid waste water (combined pH between 1.6 and 3.0) was used in the grinding tests. The AR cast irons were the worst performers. (After Singleton and Blickensderfer, 1985).**

*(Measuring Wear Rates, continued)*

candidates to combat slurry wear. However, this finding is quickly thrown into doubt as the velocity of the slurry increases or the abrasiveness of the slurry increases. This finding correlates well with empirical data offered by experts in the field. Overall, regardless of slurry speed or slurry type, the data presented in Table 9.3 shows that the best alloy for slurry wear is the high-Cr AR cast iron 20-2-1 WCI in Table 9.4.

To focus on this finding to a little greater extent, Table 9.4 presents the microstructural data along with the slurry wear data for a subset of materials presented in Table 9.3. One piece

of data that jumps out is that none of the AR cast irons studied in this work are corrosion resistant. That is, all have a Cr content of less than 12% Cr in their matrices. Consequently we are looking at the corrosion/abrasion wear performance of noncorrosion-resistant irons in this study. What is most interesting is that, even without corrosion protection, the AR irons outperform all other materials due to the severity of the abrasive environment. This finding is more clearly shown by comparing the slurry wear rates of the 316 stainless steel and that for the much softer aluminum alloy.

Another way to view the data in Table 9.4 is by thinking of it as straight abrasive wear. For example, good abrasion

**Table 9.3**  
**Slurry Wear Rates in % Material Lost per 100 kg (225 lb) of Abrasive Solid<sup>(1)</sup>**

| Alloy            | HB   | Laboratory <sup>(2)</sup> |            | Asamera Float Cell <sup>(3)</sup> |             | Shiny Rock <sup>(4)</sup> |
|------------------|------|---------------------------|------------|-----------------------------------|-------------|---------------------------|
|                  |      | 8 m/s <sup>(5)</sup>      | 12 m/s     | 8 m/s                             | 12 m/s      |                           |
| 20-2-1 WCI       | 682  | 2.4                       | 19.5       | 0.1                               | 0.8         | 3.5                       |
| 18-1 WCI         |      |                           |            |                                   |             |                           |
| Austenitic       | 508  | 3.9                       | 18.1       | 0.1                               | 0.4         | n.a. <sup>(6)</sup>       |
| Martensitic      | 745  | 3.3                       | 22.0       | 0.2                               | 0.8         | n.a.                      |
| Ni-Hard 4        | 582  | 3.3                       | 25.3       | 0.1                               | 1.0         | 5.0                       |
| Ni-Hard 1        | 653  | 8.0                       | 35.4       | 0.2                               | 1.3         | n.a.                      |
| Pearlitic Iron   | 440  | 12.9                      | 70.9       | 2.0                               | 3.3         | 10.9                      |
| Nodular Iron     | 285  | 35.8                      | 172        | 2.5                               | 7.2         | n.a.                      |
| Martensitic      |      |                           |            |                                   |             |                           |
| Steel-A          | 476  | 22.6                      | 105        | 1.5                               | 4.4         | n.a.                      |
| Steel-B          | 534  | 18.0                      | 92.6       | 0.8                               | 2.8         | 10.7                      |
| Experiment Steel | 461  | 30.2                      | 139        | 0.7                               | 3.2         | n.a.                      |
| AISI 1095        | 444  | 15.9                      | 69.3       | 1.9                               | 2.6         | 9.9                       |
| AISI 52100       | 670  | 10.8                      | 21.8       | 1.2                               | 2.4         | 7.4                       |
| ASTM A514        | 269  | 43.7 ± 7.3 <sup>(7)</sup> | 217 ± 11.7 | 1.9 ± 0.3                         | 5.4 ± 1.4   | 15.8 ± 0.7                |
| 316 SS           | 140  | 39.1                      | 180        | 0.7                               | 2.8         | 13.9                      |
| Mn-N SS          | 229  | 37.7                      | 202        | 0.4                               | 2.8         | n.a.                      |
| 6061-T651 Al     | 0.92 | 193                       | 810        | 1.4                               | 18.1        | 83.3                      |
| HDPE             | <0.1 | <0.1                      | 26.0 ± 6.5 | 0.6 ± 0.1                         | 13.1 ± 12.6 | 20.7                      |
| UHMWPE           | <0.1 | <0.1                      | <0.1       | <0.1                              | 8.6 ± 8.5   | 6.5                       |
| CPVC             | <0.1 | 42.7                      | 170 ± 50   | 3.4 ± 0.8                         | 31.2 ± 16.2 | 118                       |

(1) After Madsen (1989).

(2) Slurry of 2 wt% of 212–300 µm (–50 to +70 mesh) AFS test sand particles and tap water.

(3) Asamera slurry was drawn from a float cell and contained 18–23 wt% solids and had a pH in the 9.0–9.7 range. Particle sizes ranged from 45 µm (325 mesh) to 300 µm (50 mesh). The rougher float cell is located at the Asamera Minerals, Wenatchee, WA gold mine.

(4) Shiny Rock slurry was taken from the rougher flotation cell used in a grinding/flotation mill for gold extraction. The slurry contained 19–25 wt% solids and had a pH in the 7.36–7.8 range. Particle sizes ranged from 45 µm (325 mesh) to 300 µm (50 mesh). The main mineral type was quartz and FeMg silicates.

(5) Slurry velocities are reported in meters per second.

(6) n.a., not analyzed

(7) Multiple samples were run in several tests under these conditions.

**Table 9.4**  
**Microstructural Data and Slurry Wear Rates**  
**for Selected Materials**

| Alloy        | HB   | CVF <sup>(1)</sup> | %Cr <sup>(2)</sup> | Laboratory<br>12 m/s <sup>(3)</sup> | Asamera<br>Float Cell<br>12 m/s |
|--------------|------|--------------------|--------------------|-------------------------------------|---------------------------------|
| 20-2-1 WCI   | 682  | 28                 | 11                 | 19.5                                | 0.8                             |
| 18-1 WCI     |      |                    |                    |                                     |                                 |
| Austenitic   | 508  | 35                 | 9                  | 18.1                                | 0.4                             |
| Martensitic  | 745  | 35                 | 9                  | 22.0                                | 0.8                             |
| Ni-Hard 4    | 582  | 35                 | 3                  | 25.3                                | 1.0                             |
| Ni-Hard 1    | 653  | 25                 | 1                  | 35.4                                | 1.3                             |
| Martensitic  |      |                    |                    |                                     |                                 |
| Steel-A      | 476  | 0                  | 0.98               | 105                                 | 4.4                             |
| AISI 52100   | 670  | 0                  | 1.39               | 21.8                                | 2.4                             |
| ASTM A514    | 269  | 0                  | 0.5                | 217 ±11.7                           | 5.4 ±1.4                        |
| 316 SS       | 140  | 0                  | 17.2               | 180                                 | 2.8                             |
| 6061-T651 Al | 0.92 | 0                  | 0                  | 810                                 | 18.1                            |
| HDPE         | <0.1 | 0                  | 0                  | 26.0 ±6.5                           | 13.1 ±12.6                      |
| UHMWPE       | <0.1 | 0                  | 0                  | <0.1                                | 8.6 ±8.5                        |
| CPVC         | <0.1 | 0                  | 0                  | 170 ±50                             | 31.2 ±16.2                      |

(1) CVF—Carbide volume fraction based on the alloys' C and Cr contents.  
(2) Estimated Cr content of the matrix.  
(3) Meters per second.

resistance is always offered when hard  $M_7C_3$  carbides are used at near-eutectic CVFs (i.e., 35%). This criterion is met by both the austenitic and martensitic 18-1 white cast irons given in Table 9.4, and both irons provide extremely good corrosive/abrasive wear. Of these two irons, the austenitic iron provides better corrosive/abrasive wear in the laboratory and significantly better wear in the field. This last result can be explained from a microstructural perspective or a corrosion perspective. For example, in many abrasion tests, austenitic matrices demonstrate better abrasion resistance since the austenitic matrix provides better toughness and will locally transform to hard martensite in response to the abrasive action. If this mechanism is not coming into play then one can rely on the austenitic matrix's resistance to corrosive pitting wear. In either case, high-Cr irons containing near-eutectic CVFs within austenitic matrices seem to provide the best possible resistance to corrosive/abrasive wear.

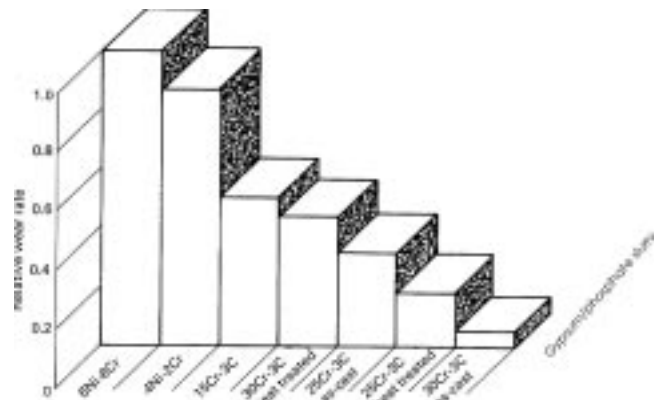
Interestingly enough, the worst performing AR cast iron is the Ni-Hard 1. Once again, if we focus on microstructure and abrasive wear principals (in this case its low CVF and its softer  $M_3C$  carbide type), its low wear rate is not as surprising as it may appear.

Since AR steels are commonly used in corrosive/abrasive wear environments, it is worthwhile to review the data presented in Table 9.4 on these alloys. Their wear rates correlate almost linearly with their hardness; i.e., the higher the hard-

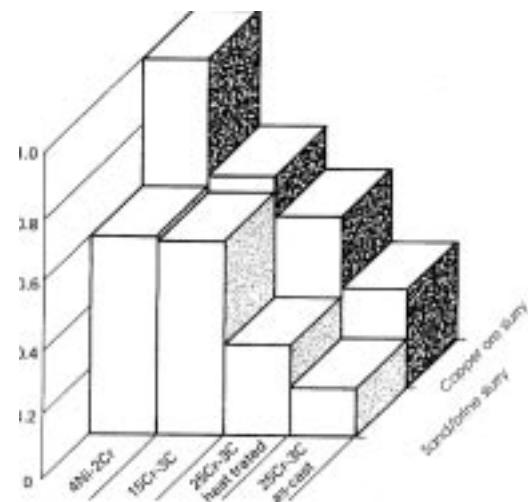
ness the lower the wear rate. Nevertheless, AR steel's wear performance in this slurry test is no match for any of the AR cast irons.

The absence of a strong corrosion/abrasion link in the slurry test is indicative of the aggressiveness of many mineral-laden slurries. If the protective passive film is being continuously removed by the abrasive particle, then it offers little in the form of corrosive/abrasive wear protection.

If one looks at some of the industrial literature on the corrosion/abrasion behavior of pump materials based on a microstructural focus, some of the results are more easily explainable. Figure 9.12 shows an example of some data collected in a laboratory wear test using two industrial slurries. In all cases, if we place emphasis on just the alloys' microstructure and its corresponding abrasive wear characteristics, the results are predictable. For example, the best wear performance is shown for alloys containing over 35% CVF of hard  $M_7C_3$  carbides (alloys with 25% Cr and 3% C) and having matrices that are dominantly austenitic (as-cast alloys).



(9.12a) gypsum/phosphate slurry



(9.12b) sand/brine and copper ore slurry

**Fig. 9.12. Relative wear rate due to corrosion/abrasive wear in two industrially based slurries. The best wear performance is always from an iron having the highest CVF and the most austenitic matrix. (After Elwell and Higginson, 1988)**

(Measuring Wear Rates, continued)

The situation changes if the pH of the slurry changes from about neutral to the acidic range. This effect is illustrated by the results of slurry abrasion tests performed in an impeller tester (which is similar to the wear pot) in Fig. 9.13.

The impeller tester consists of a polyethylene container, square in cross section, holding a slurry. The slurry is made of 20 kg quartz sand of uniform grain size (50–70 AFS silica test sand), 30 kg deionized water, and chemicals to maintain the slurry at a desired pH level. The stainless steel impeller consists of a shaft clamped in the chuck of a drill press and four equally spaced arms that carry abrasion test specimens secured in Teflon frames. The specimens have dimensions of 6.35x25.4x 57.5 mm (0.25x1.0x 2.25 in.).

The arms of the impeller rotate through the slurry, with the specimens attached, at a speed of 260 rpm, which represents an equivalent linear speed of 15 ft/sec (4.5 m/sec). The test continues for 300,000 revolutions (about 20 hours). The weight loss is determined as the difference of the specimen weight before and after the test. The chemical composition, the condition and the hardness are given in Table 9.5.

A pH change from neutral to basic, and especially to below 6, brings about a dramatic increase in wear loss.

As expected, the Ni-Hards and the tool steel suffer the highest wear and show a steep increase of wear at lower pH. The effect of the acidity on the high-Cr irons is less pronounced, except for the as-cast 20Cr irons. A ranking according to Cr content is not apparent. The heat-treated 20Cr-LC alloy, which was poor in the as-cast condition, comes out best. Obviously, there is an interaction between matrix structure, abrasion resistance and corrosion resistance.

These results, however, demonstrate the difficulty for a designer to select the best alloy for pumps and fittings operating in slurries with pH values below 7. Often, a decision has to be made between an abrasion resistant white cast iron and a stainless steel. An indication of what to expect provides the results of a wear pot tests. Table 9.6 contains comparative tests results on several white cast irons and stainless steels in mixtures of sand-water, sand plus dilute sulfuric acid, and dilute sulfuric acid alone.

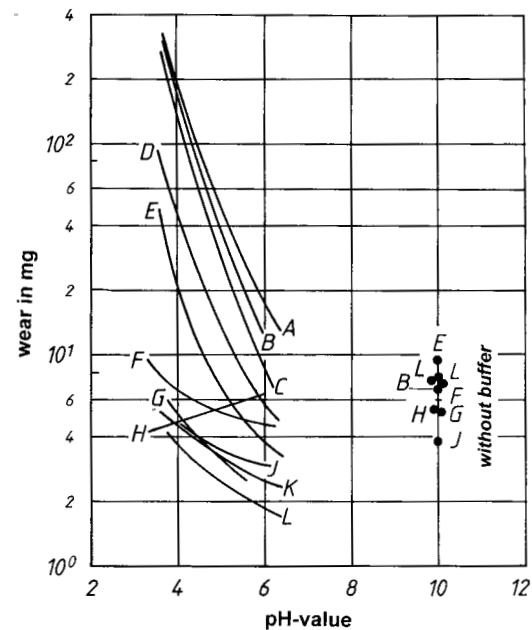


Fig. 9.13. Influence of the pH value on wear loss of various white cast irons and a tool steel in the impeller test (similar to the wear pot) with water, quartz and pH-stabilizers. The chemical composition, the condition and the hardness of the specimens are given in Table 9.5. (After Borik, 1976)

Table 9.5.  
Compositions and Conditions of Materials Used in Impeller Tester<sup>(1)</sup>

| Designation                    | %C   | %Mn  | %Si  | %Cr  | %Ni  | %Mo  | %Cu  | Condition    | Hardness<br>HRC |
|--------------------------------|------|------|------|------|------|------|------|--------------|-----------------|
| A Ni-Hard 1, Heat 1            | 3.47 | 0.70 | 0.45 | 3.20 | 3.97 |      |      | as-cast      | 57              |
| B Ni-Hard 1, Heat 2            | 3.55 | 0.94 | 0.41 | 3.42 | 3.46 |      |      | as-cast      | 60              |
| C M2 Tool Steel <sup>(2)</sup> | 0.85 |      |      | 4.00 |      | 5.00 |      | hardened     | 64              |
| D 20Cr-2Mo-1Cu, HC             | 2.47 | 0.67 | 0.40 | 21.0 |      | 1.86 | 1.18 | as-cast      | 58              |
| E 20Cr-2Mo-1Cu, LC             | 2.12 | 0.67 | 0.40 | 21.0 |      | 1.86 | 1.18 | as-cast      | 62.5            |
| F 20Cr-2Mo-1Cu, HC             | 2.47 | 0.67 | 0.40 | 21.0 |      | 1.86 | 1.18 | heat-treated | 62.6            |
| G 28Cr-1Mo, LC                 | 2.18 | 0.6  | 0.44 | 27.4 | 0.13 | 1.0  |      | as-cast      | 52.8            |
| H 28Cr-1Mo, HC                 | 2.64 | 0.63 | 0.20 | 28.0 |      | 0.94 |      | heat-treated | 56              |
| J 28Cr-1Mo, LC                 | 2.18 | 0.6  | 0.44 | 27.4 | 0.13 | 1.0  |      | heat-treated | 61              |
| K 28Cr-1Mo, HC                 | 2.64 | 0.63 | 0.20 | 28.0 |      | 0.94 |      | as-cast      | 52.5            |
| L 20Cr-2Mo-1Cu, LC             | 2.12 | 0.67 | 0.40 | 21.0 |      | 1.86 | 1.18 | heat-treated | 59              |

(1) After Borik (1976).

(2) Contains also 2% V and 6.2% W.

The corrosion-abrasion test essentially consisted of first rotating a sample in a slurry of 1% sulfuric acid + sand in an 800 ml beaker with a drill press and then determining the sample's weight loss. The test materials were machined into 47.6 mm (1-7/8 in.) diameter x 3.2 mm (1/8 in.) thick specimens having 8.33 mm (2-1/64 in.) diameter holes at their centers. The flat surfaces were finished by grinding on a dry 180-grit belt sander.

One specimen at a time was fastened to a threaded stainless steel shaft, insulated from the shaft with a polyethylene washer and fastened with two polyvinyl chloride nuts, the lower nut being closed at one end. The shaft itself was protected from the test medium with an organic coating. The drill press motor operating at 236 rpm rotated the specimen with a peripheral velocity of 5.88 m/s (99.3 fps). Air was bubbled into the slurry through a gas dispersion tube at a rate of 100 ml/min. White, 50-mesh sea sand was used as the abrasive. The slurry consisted of 300 g of sand plus 300 ml of 1% sulfuric acid.

The specimens were rotated in the slurry for two-hour test periods, with fresh abrasive and acid being used in each test period. The specimens were tested for five consecutive two-hour test periods and were washed, scrubbed, dried, and accurately weighed after each period. Temperature and pH of the slurry were determined before and immediately after each test period. No provision was made to control the temperature of the slurry. However, the initial temperatures of the test media were 22C, and increased to about 25C at the end of a two-hour test period.

In the presence of acid, stainless steel is obviously the best choice, while in neutral slurry the heat-treated high-Cr-Mo iron gives the expected superior wear resistance. Surprising is the extremely poor performance of the as-cast pearlitic high-Cr iron in acidic conditions.

**Table 9.6.**  
**Results of Wear Tests in Aerated 50% Silica Sand/50% Water Mixture and in Dilute Sulfuric Acid<sup>(1)</sup>**

| Alloy                        | Hardness<br>HRC | Sand +<br>water +<br>1% sulf. acid<br>—Weight Loss in mg/(dm <sup>2</sup> h)— | Sand +<br>water | 1%<br>sulfuric<br>acid |
|------------------------------|-----------------|-------------------------------------------------------------------------------|-----------------|------------------------|
| 15Cr-3Mo-3C,<br>as cast      | 50              | 1679                                                                          | 8,9             | 888                    |
| 15Cr-3Mo-3C,<br>heat treated | 60              | 840                                                                           | 4,9             | 589                    |
| 2Cr-4Ni-2.6C<br>(Ni-Hard 2)  | 52              | 778                                                                           | 6,7             | 712                    |
| Chilled iron<br>2.9C-1Cr     | 45              | 774                                                                           | 22              | 596                    |
| CF-8                         | —               | 125                                                                           | 112             | 0                      |
| CF-8M                        | —               | 169                                                                           | 171             | 0                      |

(1) Climax (1963).

Bess (1977) tested a series of white irons and stainless steels in an alumina-water slurry with additions of sulfuric acid or sodium chloride. According to Table 9.7, an ASTM A 532 Class III Type A iron with 25Cr-3.0C and Ni-Hard 4 perform best in the slurry without acid. Ni-Hard 4 seems to be sensitive to chlorides. All HC white irons fail at pH 2.5, while an LC 28Cr-2Ni iron is still reasonably resistant. Stainless steels are not wear resistant in abrasive neutral slurries. With decreasing pH, their weight loss increases but not to the extent as for the white irons. In any case, wear loss is much higher under abrasive-corrosive conditions than in a neutral slurry.

The discharge water from FGD scrubbers of coal-fired power stations is fairly abrasive and corrosive. While the main solid component is gypsum, limestone and fly ash introduce quartz and other hard minerals that cause a significant wear problem (Tischner, 1988). At the same time, the solutions contain high chloride levels and their pH ranges from 5 to below 2. At low pH only stainless steels, usually duplex steels, are suitable, but white cast irons can be used at the upper end of the pH range. Günther et al. (1988) investigated several cast alloys in an FGD plant as scrubber pump components and as a choke ring in the discharge pipe. The pH was 4–5 and the temperature around 45C. The results in Table 9.8 show that an LC-high-Cr iron gives the best overall performance and is superior to various stainless steels. A much harder HC-high-Cr iron, however, is completely unsuitable due to its low matrix Cr.

#### Service Performance in Slurries

In abrasive neutral slurries with pH around or above 7, white cast irons of high hardness and high CVF are the best choice. This has been proven in pumps for ore mines and similar applications in industrial scale tests (Parks, 1977, Madsen, 1989, KSB, 1994). Dredge pump casings and impellers cast of various white irons ranging from Ni-Hard 1 to high-Cr irons, like ASTM A 532 Class III Type A, have given excellent service in fresh water and in sea water, without any corrosion problems. Abrasion is obviously the dominant wear mechanism. Similar experiences are reported from iron and various copper and phosphate mines, as well as from bauxite process-

**Table 9.7.**  
**Results of Slurry Abrasion Testing of Two White Irons, Two Austenitic and a Duplex Stainless Steel in a Slurry With 20% Alumina<sup>(1)</sup>**

| pH of<br>Slurry | Alloy (Weight Loss in mm/a) |                      |                      |       |       |         |
|-----------------|-----------------------------|----------------------|----------------------|-------|-------|---------|
|                 | Ni-Hard 4                   | HC 25 <sup>(2)</sup> | LC 28 <sup>(3)</sup> | CF 8M | CN 7M | CD 4MCu |
| 11              | 0.20                        | 0.190                | 0.287                | 1.956 | 2.346 | 1.700   |
| 7               | 0.24                        | 0.234                | 0.306                | 1.757 | 1.656 | 1.661   |
| 7 + 2.5 %NaCl   | 0.47                        | 0.196                | 0.361                | 2.144 | 3.114 | 1.840   |
| 2.5             | 124.0                       | 95.844               | 1.905                | 4.038 | 5.250 | 3.830   |
| 0.5             | 173.9                       | 155.120              | 49.453               | 3.493 | 4.506 | 3.012   |

(1) After Bess (1977).  
(2) HC 250: ASTM A 532 Class III Type A iron with 25% Cr-3.0% C.  
(3) LC 28: 1.6% C, 28% Cr, 2% Mo, 2% Ni, 1% Cu.

**Table 9.8.**

**Results of Tests With Pump Components and Choke Rings in Flow Pipe in FDG Installation of Coal Fired Power Station Made of Various Cast Irons and Steels<sup>(1)</sup>**

|                                         | Hardness<br>HV     | Wear in % of Initial Weight<br>After 1000 Hours |        | Test Ring<br>(Choke) |
|-----------------------------------------|--------------------|-------------------------------------------------|--------|----------------------|
|                                         |                    | Impeller                                        | Lining |                      |
| HC-High-Cr iron,<br>4.5C-30Cr-1Mo       | 700                | 2.6                                             | —      | —                    |
| LC-High-Cr iron,<br>1.5C-28Cr-3Mo-3Cu-N | 500                | 0.7                                             | 2.3    | 1.3                  |
| High-C duplex SS<br>0.5C-25Cr-5Ni-3Mo   | 300                | —                                               | —      | 1.3                  |
| Duplex stainless steel<br>CD-4MCuN      | 340 <sup>(2)</sup> | 1.2                                             | 5.3    | 3.6                  |
| Duplex stainless steel<br>CD-4MCuN      | 250                | 1.1                                             | 5.6    | 3.6                  |
| CF-10M                                  | 200                | —                                               | —      | 3.7                  |
| Unalloyed steel, 0.45C                  | 200                | —                                               | —      | 68                   |

(1) After Günther et al. (1986).  
(2) Precipitation hardened.

ing (Day, 1982, Holdschick, 1986, KSB, 1988). With decreasing pH and increasing chloride levels, a higher Cr content and a higher Cr:C ratio become beneficial.

According to Tischner (1988), white cast iron grades with up to 15% Cr can be used in FDG applications at pH of 7 and above. If abrasion is low, chloride content must be very low. Grades with above 20% Cr and low carbon content can be employed until a pH of about 4 and are less sensitive against chlorides. At pH between 4, and 3, high-Cr irons can be used in some cases, provided temperatures and chlorides are not too high. Wear rates are fairly high but still lower than for duplex stainless steels.

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# Chapter 10

## Applications

This chapter is a pictorial guide through the myriad of applications in which abrasion resistant cast irons have found wide-ranging commercial utility. The emphasis of this chapter is to show the broad range of applications for these special-purpose irons and not to provide an endorsement of a specific alloy for a specific application. Please view this chapter as a “work in progress.” It is hoped that, in the second edition, these application examples can be expanded upon.

### Choosing the Proper AR Cast Iron

Selection and application of AR white cast irons are based on three considerations: (1) cost, (2) performance and (3) consistency/reliability, where all terms are used in the widest sense.

**Cost**—Cost is not only the price of the casting plus transport and stocking. It is also the cost of replacing worn parts and the associated downtime with eventual loss of production. These costs may be significantly higher than the cost of the part itself, so that the use of an expensive but long-living material is justified. Considerable efforts are made to facilitate changing wear parts and, thus, reduce labor and downtime expenses.

**Performance**—Performance means not only abrasion resistance but useful life. In many cases, wear parts are not replaced when worn, they are only replaced during a scheduled maintenance and repair program of the whole equipment or plant. A typical example showing both extremes are ball mills. Wear of grinding balls can be easily compensated for by adding new balls during normal operation; however, replacing worn liners is a major operation to be undertaken in a planned maintenance and plant shutdown period. In a plant that is shut down every weekend for maintenance and repairs, a wear part should last one week. An improvement of abrasion resistance to make it last one day longer is useless, but if the casting lasts only five days, an improvement of its wear life by one day will be a major achievement.

**Consistency/Reliability**—Consistency and reliability are major considerations. Accelerated wear or sudden failure by breakage can, at best, bring about unplanned downtime and loss of production. At worst, it can damage the equipment or even result in fatal accidents. The fear of breakage is often the decisive argument for preferring steel over AR cast irons.

Based on these considerations, users and suppliers have the choice between five large groups of white irons:

1. Pearlitic white or chilled cast irons
2. ASTM Class I Types A and B (Ni-Hard 1 and 2)
3. ASTM Class I Type D (Ni-Hard 4)
4. ASTM Class II Types A–D and III Type A (high-Cr and high-CrMo cast irons)
5. High-Cr, low-C grades, resistant to abrasion and corrosion.

### Discussion of Applications

Pearlitic white or chilled AR cast iron does not possess any outstanding properties and is inferior to the other martensitic grades, both in abrasion resistance and toughness. Its only advantage is low cost. For this reason, it is still used in applications involving modest abrasion, and in no- or low-impact applications, like small **liners or paddles** in mixers, **chutes** for fine-grained abrasive material and containers. A significant quantity is also produced as **small grinding media**, used, e.g., to produce silica dust (Fig. 10.1).

More often, the possibility to produce bimetallic castings (consisting of a chilled wear side and a gray iron backing) is considered. The gray iron provides a good backing against fracture and allows low-cost machining of the part. Typical examples are **hammers** for brown coal (lignite) crushers, where the head is cast in a chill mold and the shaft solidifies as gray iron in a sand mold (Fig. 10.2). In the case of the hammers, any other type of casting has turned out to be either more prone to breakage or less economical.

One of the more important applications of such bimetallic castings are **rolls** for paper making equipment, which are heated from the inside (Fig. 10.3).



**Fig. 10.1.** Small grinding media made of chilled white iron. (Courtesy of ZGV)



**Fig. 10.2.** Hammers for brown coal crushers, with white cast iron head cast in a chill mold and gray iron shaft cast in a sand mold. (Courtesy of ZGV)



**Fig. 10.3.** Chilled iron roll for papermaking equipment that is heated from the inside; 65 ton, 1425 mm dia total length 12,300 mm. (Courtesy of ZGV)



**Fig. 10.4.** Two-inlet, two-outlet camshafts for BMW 740d 8V cylinder twin-turbo diesel engine, cast hollow in order to reduce weight and save machining costs. The cams are chilled. (Courtesy of BMW)

Chilled iron components are extensively used for applications involving metal-to-metal wear, like **camshafts** (Fig. 10.4), **followers** and **tappets** in the valve train of internal combustion engines and for some types of **metal working rolls**.

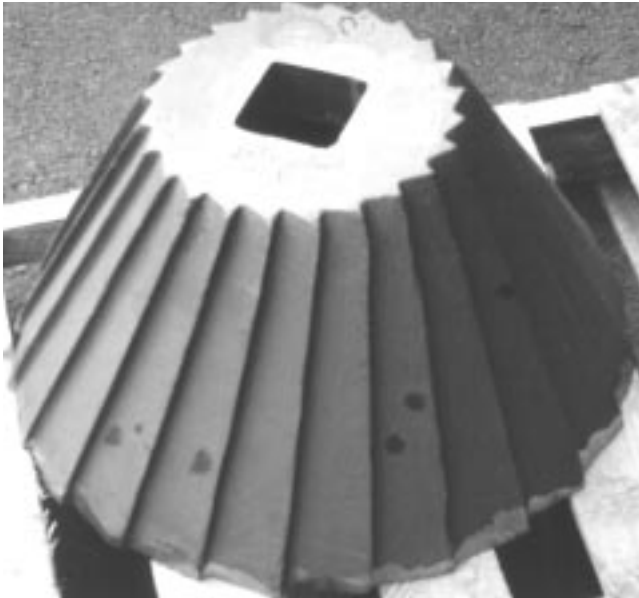
Martensitic white irons ASTM Class IA and IB (Ni-Hard 1 and 2) have good abrasion resistance, but their resistance against fracture is not very high. Their advantages are that they do not require a high-temperature heat treatment and can be put directly into service either as-cast or after a stress-relief heat treatment. Another competitive advantage of these alloys is their low alloy cost (although this benefit is somewhat negated when the price of Ni is high). Due to their softer carbides ( $M_3C$ ), their abrasion resistance is usually inferior compared to high-Cr irons ( $M_7C_3$  carbides), provided the latter are correctly heat treated. The Ni-Hards have found extensive applications for a wide variety of components of all sizes subjected to low or medium dynamic stresses, including **crusher parts**. Typical examples of a **mixing wheel**, **concrete mixing blade**, **grinder component**, **nozzles** and **road-paving auger** are shown in Figs. 10.5 through 10.9. Similar complex components can be made from high-Cr irons, as shown in Figs. 10.10 and 10.11.



**Fig 10.5a.** A variety of Ni-Hard 1 (ASTM IA) castings used in aggregate processing, concrete pipemaking, placement and mixing applications. The center mixing wheel is 38.1 cm (15 in.) in dia and weighs about 22 kg (50 lb). (Courtesy of Texaloy Foundry Co.)



**Fig. 10.5b.** Ni-Hard 1 (ASTM IA) mixer blade used in concrete mixer (shown on foundry alignment jig). Casting weighs about 191 kg (430 lb). (Courtesy of Texaloy Foundry Co.)



**Fig 10.6. Ni-Hard 1 (ASTM IA) grinder component.** The approximate weight of the casting is 18 kg (40 lb). (Courtesy of Weatherly Casting and Machine Co.)



**Fig 10.7. Ni-Hard 1 (ASTM IA) nozzles.** The approximate weight of one casting is 80 kg (180 lb). (Courtesy of Weatherly Casting and Machine Co.)



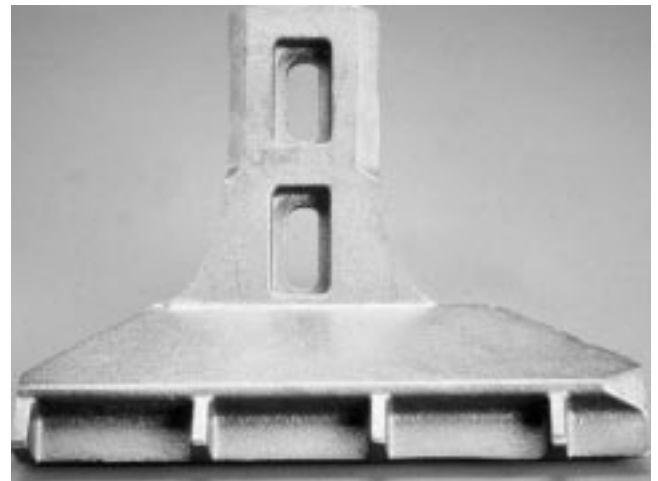
**Fig. 10.8. Mixer with blades and lining of Ni-Hard 1 (ASTM IA).** (Courtesy of BHS)

Ni-Hards 1 and 2 are also suitable for bimetallic castings or components cast with inserts, since heat treatment is unnecessary. That is, differences in coefficients of thermal expansion between the different materials and transformation-induced stresses will often cause cracking around cast-in inserts. Figure 10.12 shows an example of a **steel insert** that can be cast into an AR cast iron.

Another example of Ni-Hard 1 and Ni-Hard 2 usage is in the pump industry. Many smaller **pump parts** subjected to severe abrasion are cast in Ni-Hard 1 and Ni-Hard 2. An



**Fig. 10.9. High-alloy CrNi (ASTM ID, Ni-Hard 4) AR auger for road pavers.** Weight of mixer blade is 6 kg (13 lb). (Courtesy of Vautid)



**Fig 10.10. High-Cr (ASTM IIID) AR cast iron mixer blade for concrete mixing.** Weight is 11.5 kg (25 lb). (Courtesy of Vautid)



**Fig 10.11. High-Cr (ASTM IIIA) miscellaneous wear parts.** (Courtesy of Eagle Foundry)

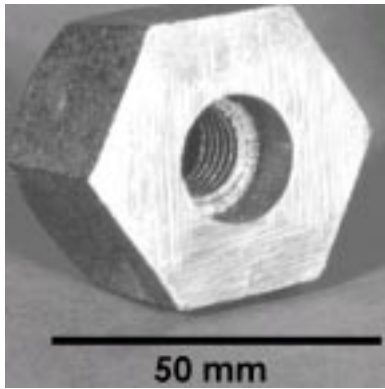


Fig. 10.12. Conventional steel nut as insert in a Ni-Hard 1 (ASTM IA) casting to be used to fasten liners in a shot-peening chamber. (Courtesy of ZGV)



Fig. 10.13. Pump for solids-bearing wastewater with Ni-Hard 1 (ASTM IA) impeller and housing. (Courtesy of DIA-Pumpen)



Fig. 10.14. Heavy medium distributor for the mining industry of Ni-Hard 1 (ASTM IA). Total weight 2200 kg (4851 lb). (Courtesy NiDI)

example of such a pump is shown in Fig. 10.13. Since Ni-Hard 1 is rather easy to cast and needs no heat treatment (involving a risk of distortion), fairly complex parts can be made, such as the **medium distributor** shown in Fig. 10.14.

Modern CBN or CBN-containing tool materials allow machining of martensitic white irons, eliminating the need for soft annealing. Figure 10.15 shows a Ni-Hard 1 **slurry pump housing** during machining. These tools have, in fact, widened the application range of Ni-Hard irons.

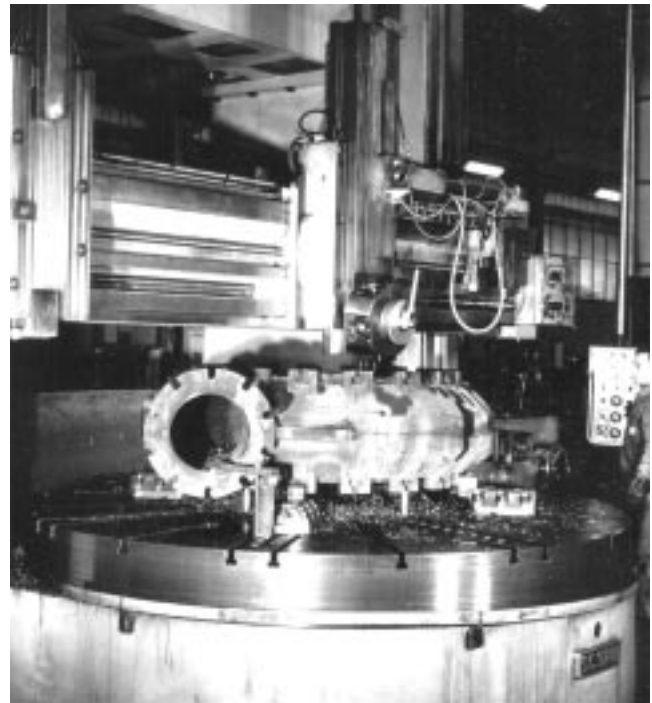


Fig. 10.15. Machining of Ni-Hard 1 (ASTM IA) slurry pump housing with CBN tooling. (Courtesy of DeBeers)



Fig 10.16. Modified Ni-Hard 1 (ASTM IA) mini-mill rolls for the structural steel industry. The roll is statically chill cast. The alloy has been modified to form a mixed microstructure of hard  $M_3C$  carbides and spheroidal graphite. (Courtesy of United Foundries)

A significant quantity of Ni-Hard-type irons is used for **rolling steel** and other metals (Fig. 10.16). The interdendritic graphite-containing grade is called indefinite-chill. High-hardness versions of such rolls of are still of importance in **continuous hot-strip mills**, even if a significant part has been substituted by high-Cr iron rolls. These rolls are usually produced as bimetallic castings by various methods, with a core of gray or ductile iron or steel. An example of bimetallic **high-Cr rolls** is shown in Fig. 10.17.

A large market for martensitic AR irons is rotary grinding mills, where they are used as **grinding media and liners**. Ni-



*Fig 10.17. High-Cr (ASTM IIB) duplex rolls for hot strip mill applications. The rolls were manufactured by centrifugally casting the high-Cr shell material and then statically casting the body and necks using a gray or nodule iron alloy. The castings shown in the photograph have been heat treated to a shell hardness of around 650 HB and then machined. (Courtesy of United Foundries)*



*Fig. 10.18. Installation of Ni-Hard 1 (ASTM IA) liner plates in a ball mill. (Courtesy of NiDI)*

Hards 1 and 2 liners can be used for low- and medium-impact ball mills, as shown in Fig. 10.18. **Liners** for high-impact mills are usually cast from high-Cr irons, ASTM Class II and III (Fig. 10.19). High-Cr irons with an optimized balance of heat treatment and chemical composition are also often preferred, due to their longer service life. **Grinding balls** are made of either high-Cr irons or steel. Wet grinding involves considerably higher wear losses than dry grinding. Only a small part of this loss is due to corrosion, however. Therefore, the use of a more corrosion-resistant material, like a high-Cr iron with higher Cr content, is rarely justified. As discussed in Chapters 8 and 9 on Abrasion and Corrosion, the main wear factors are straightforward abrasion resistance and hardness.

**Liners and grinding balls** of cement mills suffer high impact over a long period of time because cement clinker is not very abrasive. For this reason, low-C, high-Cr and high-CrMo irons, according to ASTM Class II Types A and B are used. They are then heat treated to obtain a low retained austenite content and, hence, a high resistance to repetitive-impact. An example of such an application is shown in Fig. 10.20.

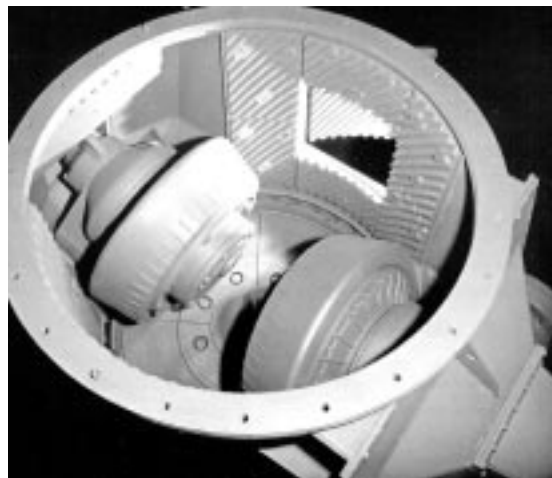


*Fig. 10.19. Lifter bars forming part of the lining of an auto-geneous mill, cast of CrMo irons (ASTM IID). (Courtesy of IMOA)*



*Fig. 10.20. The inside of a ball mill for grinding cement clinker, containing low-C high-CrMo (ASTM IIA) iron balls and liner plates. (Courtesy of IMOA)*

**Roller crushers** of various designs are used to grind coal in power stations, cement plants, etc. They are also used to grind limestone, cement, phosphate rock and other soft and medium-hard minerals. They consist of one or several **rollers and grinding tables** (Fig. 10.21). The castings are rather large and may weigh several tons. With some exceptions, rolls and tables can be made from ASTM Class I, II and III (Ni-Hard 1 or 2, Ni-Hard 4, or high-Cr and high-CrMo irons). The latter have the highest potential abrasion resistance, provided chemical composition and heat treatment are well balanced and optimized for the section and the specific application.



*Fig. 10.21. The inside of a large Loesche roller crusher used for grinding and drying coal, showing rolls and grinding table. Rolls and grinding table are cast from Ni-Hard 4 (ASTM ID) or high-Cr (ASTM IIB or D) cast irons. (Courtesy of Loesche)*



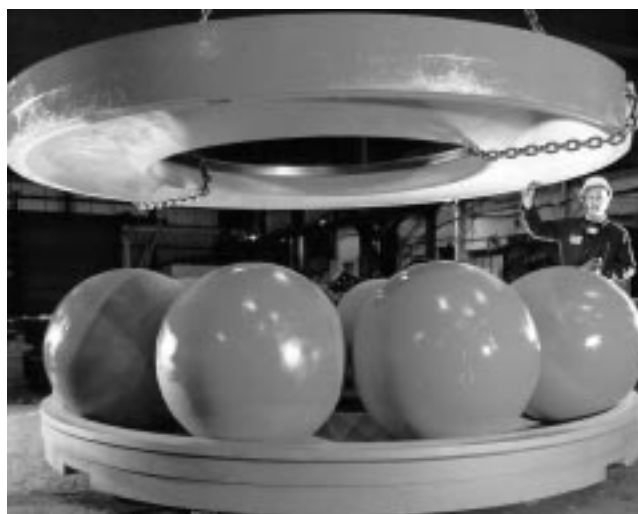
*Fig 10.22. High-Cr (ASTM IIIA) tire and wear track segment for coal pulverizing. (Courtesy of Columbia Steel Castings)*



*Fig 10.23. High-Cr (ASTM IIIA) wear track segment for coal crushing. The casting weighs approximately 444 kg (1000 lb). (Courtesy of Columbia Steel Castings)*



*Fig 10.24. High-Cr (ASTM IID) coal-crushing replacement rollers. The inside bearing races have been machined and are ready for field installation. (Courtesy of Firth Rixson Castings)*



*Fig 10.25. High-Cr (ASTM IID) coal-crushing wear parts shown as a completely matched set. (Courtesy of Firth Rixson Castings)*



*Fig 10.26. High-Cr (ASTM IIID) coal-pulverizer ring. (Courtesy of Georgia Iron Works)*

Ni-Hard 4 has the advantage of the simpler heat treatment of slow cooling in a furnace from 830C, which is less expensive and less prone to human errors. Also, the slow cool is supposed to produce lower internal stresses. For these reasons, especially in Europe, Ni-Hard 4 is often preferred. Nevertheless, service reports are sometimes contradictory about the performance of Ni-Hard 4. Ni-Hards 1 and 2 irons are less abrasion resistant, but this wear performance factor is compensated for by using a less-expensive alloy and eliminating the high-temperature heat treatment. Figures 10.22 through 10.27 show examples of **roller and table** castings for several types of crushers.

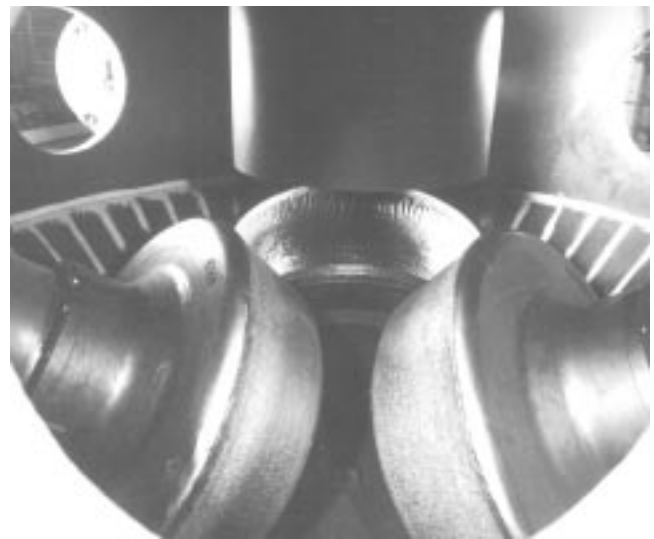
Worn rolls can be repaired by hardfacing. This is possible because the weld deposit contains many small cracks that do not grow into the white iron base material of the roll. **Hardfacing** can be done either in a special shop after dismantling the mill or in situ in the power plant if downtime is to be minimized. In the last few years, even new rolls for power stations are hardfaced, as shown in Fig. 10.28. This is done to extend their life and reduce downtime. The white iron base prevents accelerated wear that would occur if the hardfacing layer on a soft base (like steel or ductile iron) is worn. With a white iron base, it is possible to run the crusher until the next planned plant shutdown to change the rolls. Many other **crusher parts** are made of high-Cr cast irons or Ni-Hard 4 (Fig. 10.29).

**Hammers** for hammer crushers and especially **impact or blow bars** in impact crushers for hard rock crushing are usually cast from high-Cr cast iron grades ASTM Class II and III. High-Cr cast irons give the required combination of sufficient resistance to breakage under repeated and single heavy-impact and high-abrasion resistance. Examples of such castings are given in Figs. 10.30 through 10.33.

Wear parts of **shot blast machines** (Fig. 10.34) are an important application of high-Cr cast irons, ASTM Classes II and III. High-velocity shot and, especially, the mixture of shot and sand in blast cleaning installations are extremely abrasive.



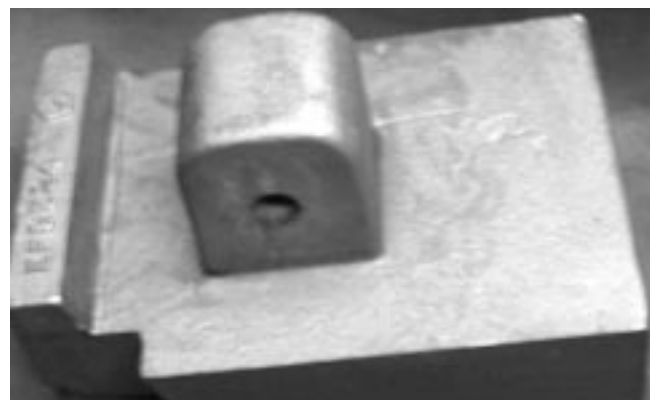
*Fig. 10.27. Roller tire for roller crusher cast of Ni-Hard 4 (ASTM ID). (Courtesy of BLK Babcock Giesserei)*



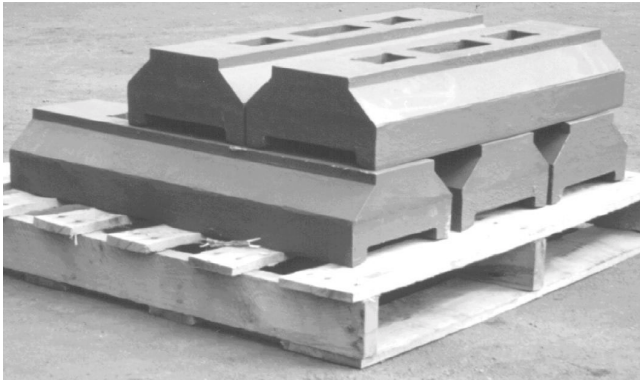
*Fig. 10.28. New Ni-Hard 4 (ASTM ID) rolls for a power station, hardfaced to obtain an extended life. (Courtesy of GEC Alsthom)*



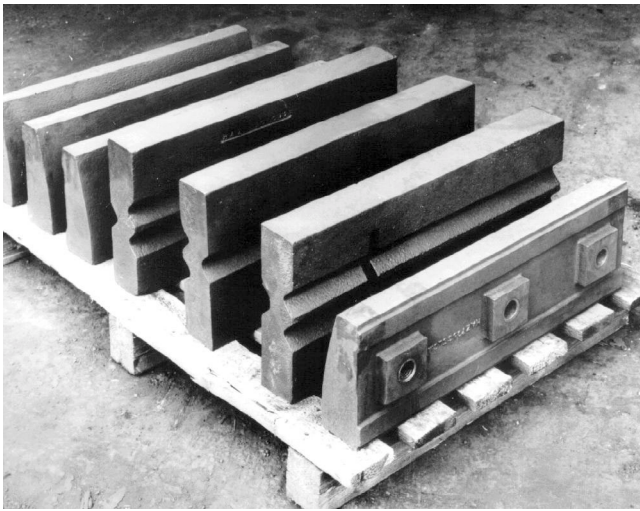
*Fig 10.29. High-Cr (ASTM IIIA) casting used to crush sand lumps. Casting is approximately 122x81 cm (48x32 in.) and weighs 89 kg (200 lb). (Courtesy of United Machine and Foundry)*



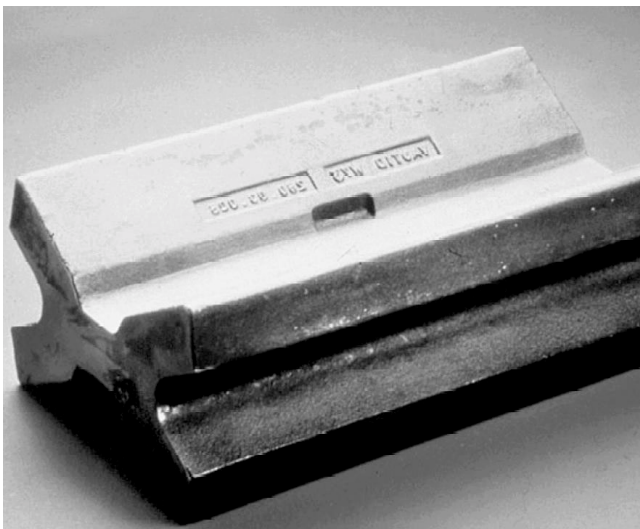
*Fig 10.30. Impeller shoe for a vertical shaft impact crusher, Canica Model 95. This part weighs 34 kg (76 lb) and is cast in ASTM A 532 Class III Type A. The part is one of five shoes that sits on a table that spins at approximately 900 rpm. The rock is deflected off the impeller shoe into a ring of anvils, which shatters or fractures the big rock into smaller rocks. (Courtesy of Eagle Foundry Co.)*



**Fig 10.31. High-Cr (ASTM IIB) impact hammers, heat treated to a hardness of 700 HB. The approximate weight of one casting is 142 kg (320 lb). (Courtesy of Weatherly Casting and Machine Co.)**



**Fig. 10.32. ASTM IIB impact bars (center) and beater blocks of impact crusher used to produce road building gravel from hard rock. (Courtesy of Hazemag)**



**Fig. 10.33. High-Cr AR cast iron (ASTM IID) blow bar for an impact crusher. The bar weights 280 kg (617 lb). (Courtesy of Vautid)**

**Large components of dredge and slurry pumps** can be cast from ASTM Class ID (Ni-Hard 4) or ASTM Class II and III (high-Cr and high-CrMo irons). Such applications are shown in Figs. 10.35 through 10.38. The advantage of Ni-Hard 4 is the heat treatment with slow cooling of the sometimes complex castings, reducing the risk of cracking and distortion. In most cases, performance of Ni-Hard 4 and high-Cr iron in neutral media is equal.

Corrosion does not play a significant role, as long as the water is neutral or slightly basic. Even **dredge pumps in sea water** do not show any corrosion damage or pitting attack in standstill periods. The main factor is abrasion resistance, and a higher or lower Cr content in the range of ASTM ID to ASTM IIIA does not have significant effect. Lowering the pH-value to below 7 into the slight acidic range, however, does increase the corrosion component, and higher Cr content becomes important. ASTM IIIA is superior to ASTM IID, which is superior to ASTM IIB and, especially, Ni-Hard 4 (ASTM ID). An example of the use of ASTM IIIA in corrosive/abrasive environments is that of sacrificial or protective **wear plates** (Fig. 10.39) that are bolted on the auger of a large sand screw. Another example is that of **drag head parts**, as shown in Fig. 10.40.



**Fig. 10.34. Wear parts of shot blast machines cast from high-Cr cast iron grade ASTM IIIA. (Courtesy of GF)**



**Fig 10.35. High-Cr (ASTM IIIA) 2.13-m (84-in.) dia slurry pump impellers. (Courtesy of Georgia Iron Works)**

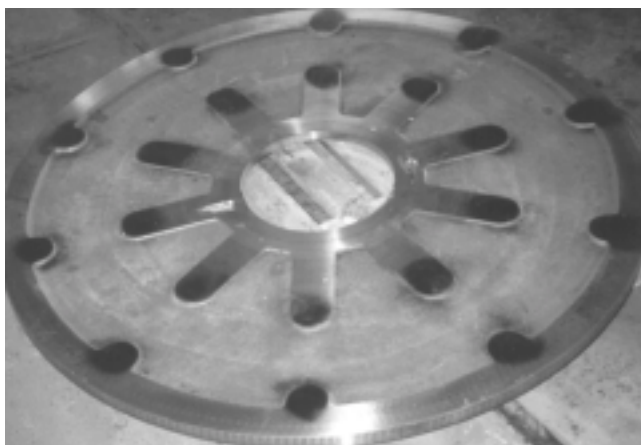
Castings of ASTM IIIA (Figs. 10.41, 10.42 and 10.43) are to be preferred in corrosive slurries. With regard to the castings in Figs. 10.42 and Fig. 10.43, it should be kept in mind that coal slurries can become acidic due to reactions of their sulfur content with water. The lowest pH-value, where high-Cr irons are useful, depends on the relative amounts of



**Fig 10.36. High-Cr (ASTM IIIA) sand and gravel pump housings. Casting weights range from 1100 to 2900 kg (2500 to 6500 lb). (Courtesy of Mobile Pulley and Machine Works)**



**Fig 10.37. High-Cr (ASTM IIIA) AR parts for the dredging industry. Weights of large castings can exceed 5780 kg (13,000 lb). (Courtesy of Mobile Pulley and Machine Works)**



**Fig 10.38. Ni-Hard 4 (ASTM ID) 183 cm (72 in.) dia pump head liner. The approximate weight of the casting is 575 kg (1300 lb). (Courtesy of Weatherly Casting and Machine Co.)**



**Fig 10.39. An auger flight shoe for a 112-cm (44-in.) Eagle sand screw is shown. This part weighs 5 kg (12 lb) and is cast in the ASTM A 532 Class III Type A. This part is one of approximately 96 pieces that is bolted onto an auger to protect the auger from wear. The screw is designed to wash gravel. (Courtesy of Eagle Foundry Co.)**



**Fig. 10.40. High-Cr (ASTM A532 IIIA) drag head parts for the sand and gravel business. Parts weigh from 15 to 125 kg (~25 to 300 lb). (Courtesy of Mobile Pulley and Machine)**



**Fig 10.41. High-Cr (ASTM IIIA) 135-cm (53-in.) dia impeller for corrosion/erosion pump slurry application. Casting is positioned on a two-plane dynamic balancing machine. The approximate weight of the casting is 1500 kg (3400 lb). (Courtesy of Weatherly Casting and Machine Co.)**

abrasion and corrosion. Chloride content increases wear loss of high-Cr irons in slightly acidic slurries. Eventually, special irons with high-Cr:C ratios or AR stainless steels have to be used. In any case, however, total wear loss will be much higher than the sum of abrasive and corrosive loss alone.



**Fig 10.42. High-Cr (ASTM IIIA) pipe elbow to transfer bauxite sludge.** The casting has a dia of 53.3 cm (21 in.) and weighs about 550 kg (1235 lb). (Courtesy of Texaloy Foundry Co.)

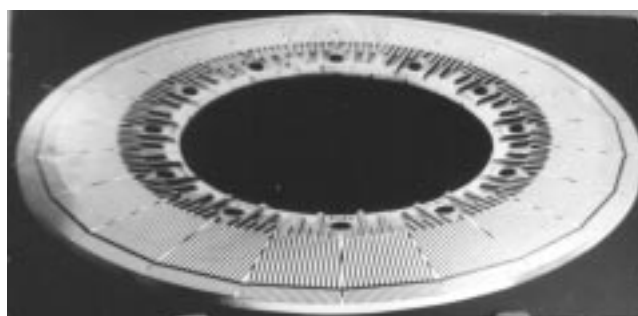


**Fig 10.43a. High-Cr (ASTM IIIA) casting used to de-water coal.** Casting is approximately 81x81 cm (32x32 in.) and weighs 290 kg (650 lb). (Courtesy of United Machine and Foundry)

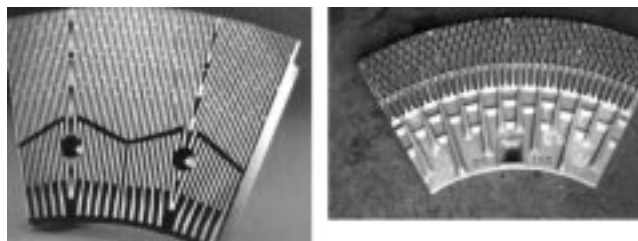


**Fig. 10.43b. High-Cr (ASTM IIIA) casting used in asphalt mixing.** Casting is approximately 74x46 cm (29x18 in.) and weighs 67 kg (150 lb). (Courtesy of United Machine and Foundry)

A classic example of this combined corrosive/abrasive wear behavior is that found in the **mechanical pulping of wood chips**. In this application, the corrosive medium is the wood pulp liquor and the abrasive is rock particles embedded within the wood chips. The refiner plate segments are mounted onto a disk and spun against a counter-rotating disk or a statically fixed disk. The wood chips are then fed between these rotating disks and processed into wood fibers. Figure 10.44 shows a circle of **refiner plates** and Fig. 10.45 shows several examples of separate refiner **plate segments** made from ASTM IIIA. Their high Cr content (25%) and high CVF (approximately 30%) provides superior corrosion/abrasive resistance. In some specialty applications, super high-Cr grades (30% Cr) are sometimes used as refiner plate alloys to combat corrosive wear.



**Fig 10.44. High-Cr (ASTM IIIA) AR refiner plate segments, laid out to form a complete circle prior to installation onto a refiner disk.** The refiner plate circle has a dia of 1.5 m (5 ft) and is composed of 12 refiner plate segments. (Courtesy of Durametal Corporation)



**Fig 10.45 High-Cr (ASTM IIIA) AR refiner plates for the pulp and paper industry.** The holes are about 35 mm (1.4 in.) dia and the castings weigh about 10 kg (22 lb) each. (Courtesy of Durametal Corporation)

## Appendix A 177

### Hardness of Minerals, Iron Phases and Carbides

Table A1. Hardness of Minerals, Iron Phases and Carbides

## Appendix B 177

### Mechanical & Thermophysical Properties

#### *Pearlitic White AR Cast Irons*

Table B1. Mechanical & Thermophysical Properties of a Representative Pearlitic White Cast Iron

Table B2. Effect of Carbon Content on Mechanical Properties of Pearlitic White AR Cast Iron

Table B3. Effect of Carbon Content on Thermal Properties of Pearlitic White AR Cast Iron

#### *Ni-Hard 1 & Ni-Hard 2 AR Cast Irons*

Table B4. Typical Range of Mechanical Properties of Ni-Hard 1 and Ni-Hard 2 AR Cast Irons

Table B5. Some Typical Physical Properties of Ni-Hard 1 and Ni-Hard 2

Table B6. Fracture Toughness of Ni-Hard 1 Alloys

Fig. B1. Hot hardness testing of a heavy-section Ni-Hard 1 casting.

#### *Ni-Hard 4 AR Cast Irons*

Table B7. Typical Range of Mechanical Properties of Ni-Hard 4 AR Cast Irons

Table B8. Some Typical Physical Properties of Ni-Hard 4

Table B9. Fracture Toughness of Ni-Hard 4 Alloys

#### *High-Chromium AR Cast Irons*

Table B10. Typical Range of Mechanical Properties of High-Cr AR Cast Iron

Table B11. Some Typical Physical Properties of High-Cr AR Cast Iron

Fig. B2. Typical values of compressive strength vs. hardness...

Fig. B3. Hot hardness of high-Cr AR cast irons.

Fig. B4. Elastic modulus ... as a function of composition and temperature.

Fig. B5. Thermal conductivity as a function of composition and temperature.

Fig. B6. Specific heat ... as a function of composition and temperature.

## Appendix C 183

### General Mechanical & Wear Properties

#### *Diesburg & Boric (1974) (text)*

Table C1. Type and Composition of Iron and Steel Alloys

Table C2. Heat Treatment of Iron and Steel Alloys

Table C3. Mechanical Properties and Abrasion Resistance of Iron and Steel Alloys

### Hardness and Abrasion Resistance

#### *Avery (1974) (text)*

Table C4. Composition and Heat Treatment of Grinding Mill Alloys.

Table C5. Abrasion Resistance of Thin and Thick Sections of Grinding Mill Liner Castings

Fig. C1a and C1b. Relationships between gouging wear, C content and various metals.

### Abrasion Resistance & Fracture Toughness

#### *Sare (1979) (text)*

Table C6. Alloy Compositions (Wt.%) and Abrasion Test Data

Table C7. Heat Treatments

Table C8. Abrasion & Fracture Toughness Test Results

Fig. C2. Schematic of shell liners studied.

Fig. C3. Description of wear and fracture tests.

## Appendix D 194

### Hardness Measurements

Table D1. Generalized Comparison of Errors in Hardness Measurements

Table D2. Hardness Conversion Equations

Table D3. Approximate Hardness Conversion Data for AR White Cast Irons

Fig. D1. Schematic of common hardness tests.

Fig. D2. Hypoeutectic, sand cast, white cast iron showing an  $L_D$  indentation.

Fig. D3. Comparison of curve-fitting errors between HB and HRC data.

Fig. D4. Comparison of curve-fitting errors between HB and  $L_D$  data.

References/Bibliography

## Appendix E 198

### Standards

Table E1. National Standard Designations Correlating Nearest Equivalent Grades

Table E2. France—Specifications

Table E3. Germany—Specifications

Table E4. United Kingdom—Specifications

Table E5. India—Specifications

Table E6. Sweden—Specifications

Table E7a. United States ASTM A 532/A532M–1993 Chemical Composition

Table E7b. United States ASTM A 532/A532M–1993 Hardness Requirements

Table E8. AR White Cast Irons According to European Standard 12 513

Table E9. Comparison of EN and ASTM Grades and Common Tradenames

## Appendix F 203

### Sample Preparation for Metallography

Table F1. General Metallographic Etchants to Reveal White Iron Microstructures

Table F2. Specific Metallographic Etchants to Reveal Various Carbide Structures

Table F3. Mesh and Grit Sizes for Abrasive Particles

Fig. F1. Photo showing a preferred method for sectioning and mounting a 5x25 mm (0.25x1.0 in.) pin in Bakelite™ or epoxy.

## Appendix G 205

### Calculation of Superheat

Table G1. Applicable Chemistries and Superheat Equations for Typical AR Cast Irons

Table G2. Approximate Austenite Liquidus Temperatures in °C and (°F)

## Appendix H 206

### Fluidity Measurements

Table H1. Experimental Raw Data

Tables H2–H4. Regression Analyses

Table H5. Relationships for Fluidity

Fig. H1. Design of fluidity spiral and gating system.

Figs. H2–H4. Plot of spiral length vs. pour temperature for 1–39% chromium irons.



# Appendices

## Appendix A

### Hardness of Minerals, Iron Phases and Carbides

**Table A1**  
**Hardness of Minerals, Iron Phases and Carbides<sup>(1)</sup>**

|                                                      | Hardness             |           |
|------------------------------------------------------|----------------------|-----------|
|                                                      | Knoop <sup>(2)</sup> | Vickers   |
| <b>Minerals:</b>                                     |                      |           |
| Talc                                                 | 20                   | —         |
| Carbon                                               | 35                   | —         |
| Gypsum                                               | 40                   | 36        |
| Calcite                                              | 130                  | 140       |
| Fluorite                                             | 175                  | 190       |
| Apatite                                              | 335                  | 540       |
| Glass                                                | 455                  | 500       |
| Felspar                                              | 550                  | 600-750   |
| Magnetite                                            | 575                  | —         |
| Orthoclase                                           | 620                  | —         |
| Flint                                                | 820                  | 950       |
| Quartz                                               | 1330                 | 1430      |
| Topaz                                                | 1360                 | —         |
| Garnet                                               | 1360                 | —         |
| Emery                                                | 1400                 | —         |
| Corundum                                             | 2020                 | 1800      |
| Silicon carbide SiC                                  | 2585                 | 2600      |
| Diamond                                              | 7575                 | 10,000    |
| <b>Iron Phases:</b>                                  |                      |           |
| Ferrite                                              | 235                  | 70-200    |
| Pearlite, unalloyed                                  | —                    | 250-320   |
| Pearlite, alloyed                                    | —                    | 300-460   |
| Austenite, 12% Mn                                    | 305                  | 170-230   |
| Austenite, low alloy                                 | —                    | 250-350   |
| Austenite, high Cr iron                              | —                    | 300-600   |
| Martensite                                           | 500-800              | 500-1010  |
| Cementite                                            | 1025                 | 840-1100  |
| <b>Carbides:</b>                                     |                      |           |
| Chromium Carbide (Fe,Cr) <sub>7</sub> C <sub>3</sub> | 1735                 | 1200-1600 |
| Molybdenum carbide Mo <sub>2</sub> C                 | 1800                 | 1500      |
| Tungsten carbide WC                                  | 1800                 | 2400      |
| Vanadium carbide VC                                  | 2660                 | 2800      |
| Titanium carbide TiC                                 | 2470                 | 3200      |
| Boron carbide                                        | 2800                 | 3700      |

(1) Data obtained from W. Fairhurst and K. Röhrig, "Abrasion-Resistant High-Chromium White Cast Irons," Foundry Trade Journal, vol. 136, No. 2999, pp. 685-98 (1974).

(2) Typical values.

## Appendix B

### Mechanical and Thermophysical Properties

Appendix B consists of tables and figures that are categorized by type of AR cast iron. They are as follows:

- Pearlitic White Cast Irons
- Ni-Hard 1 and Ni-Hard 2 AR Cast Irons
- Ni-Hard 4 AR Cast Irons
- High-Chromium AR Cast Irons

#### • Pearlitic White AR Cast Irons

**Table B1**  
**Mechanical and Thermophysical Properties**  
**of a Representative Pearlitic White Cast Iron<sup>(1)</sup>**

| Property                                                                                                                                                                                                                                                                                                                           | Sand Cast           | Chill Cast<br>(Perm. Mold) |
|------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------|---------------------|----------------------------|
| Hardness, HBN                                                                                                                                                                                                                                                                                                                      | 390-470             | 410-510                    |
| Tensile Strength<br>ksi (MPa)                                                                                                                                                                                                                                                                                                      | 30-60 (230-460)     | 35-60 (280-460)            |
| Elastic Modulus Mpsi (GPa)                                                                                                                                                                                                                                                                                                         | 30 (210)            | 30 (210)                   |
| Compressive Strength<br>psi (MPa)                                                                                                                                                                                                                                                                                                  | 195-225 (1500-1750) | n.a.                       |
| Bending Strength<br>psi (MPa) <sup>(2)</sup>                                                                                                                                                                                                                                                                                       | 55-90 (370-620)     | 70-105 (510-725)           |
| Density<br>lb/in. <sup>3</sup> (g/cm <sup>3</sup> )                                                                                                                                                                                                                                                                                | 0.27-0.28 (7.6-7.8) | 0.27-0.28 (7.6-7.8)        |
| Specific Heat<br>Btu/lb-F (J/kg-K)                                                                                                                                                                                                                                                                                                 | 0.135 (565)         | n.a.                       |
| Latent Heat of Fusion<br>Btu/lb (J/kg)                                                                                                                                                                                                                                                                                             | 84 (195,000)        | n.a.                       |
| Thermal Conductivity<br>Btu/hr-ft-F (W/m-K)                                                                                                                                                                                                                                                                                        | 7.5-17 (13-30)      | n.a.                       |
| Coefficient of Thermal Expansion:<br>0-528F (0-276C); 5.5-6 (10-11)x10 <sup>6</sup><br>528-1263F (276-684C); 8.8-9.1 (15.9-16.4)x10 <sup>6</sup>                                                                                                                                                                                   |                     |                            |
| n.a. = not available                                                                                                                                                                                                                                                                                                               |                     |                            |
| (1) The numbers presented in this table are representative of pearlitic unalloyed white cast irons cast into 30 mm (1.2 in.) dia bars. Compositions are between 2.5-3.5% C, 0.4-0.67% Si, 0.3-0.96% Mn, balance Fe. Data obtained from Angus, H.T., "Cast Iron: Physical and Engineering Properties," Butterworths, London (1976). |                     |                            |
| (2) Transverse rupture stress in a 30 mm (1.2 in.) dia bar using a 12 in. (305 mm) span.                                                                                                                                                                                                                                           |                     |                            |

**Table B2**  
**Effect of Carbon Content on Mechanical Properties of Pearlitic White AR Cast Iron<sup>(1)</sup>**

| Carbon Content (%)                            | 1.77 <sup>(3)</sup> | 2.00       | 2.22 <sup>(4)</sup> | 2.30       | 2.50       | 2.57 <sup>(5)</sup> | 3.00       | 3.50       |
|-----------------------------------------------|---------------------|------------|---------------------|------------|------------|---------------------|------------|------------|
| Tensile Strength <sup>(1)</sup> kpsi (Mpa)    | 64 (520)            |            | 80 (570)            |            |            | 77 (540)            |            |            |
| Tensile Strength <sup>(2)</sup> kpsi (Mpa)    |                     | 50 (350)   |                     | 55 (390)   | 60 (420)   |                     | 70 (490)   | 73 (510)   |
| Strain to Failure <sup>(1)</sup> %            | 0.51                |            | 0.49                |            |            | 0.51                |            |            |
| Compressive Strength <sup>(2)</sup> ksi (MPa) |                     | 190 (1350) |                     | 200 (1400) | 205 (1430) |                     | 215 (1500) | 225 (1575) |
| Elastic Modulus <sup>(1,2)</sup> Mpsi (GPa)   | 28-30               |            | (200-210)           |            |            |                     |            |            |

(1) Test bars were sand cast to 30 mm (1.2 in.). After casting, bars were tempered at 990( F (550( C) for 1 hr and then furnace cooled. Data obtained from: H.T. Angus, "Cast Iron: Physical and Engineering Properties," Butterworths, London (1976).

(2) Data obtained from Ruddeck, P., Über die Qualitätsbestimmung und Qualitätsentwicklung von unlegiertem Hartguß, Dr.-Ing. Dissertation, Freiberg (1966).

(3) Chemical composition was 1.77% total C, 0.96% Si, 0.22% Mn, 0.1% S, 0.07% P, balance Fe.

(4) Chemical composition was 2.22% total C, 1.12% Si, 0.22% Mn, 0.1% S, 0.05% P, balance Fe.

(5) Chemical composition was 2.57% total C, 1.03% Si, 0.21% Mn, 0.1% S, 0.07% P, balance Fe.

**Table B3**  
**Effect of Carbon Content on  
Thermal Properties of Pearlitic White AR Cast Iron<sup>(1)</sup>**

| Carbon Content (%) | Coefficient of Thermal Expansion <sup>(2)</sup> Strain / F (Strain / C)x10 <sup>6</sup> | Thermal Conductivity Btu/hr-ft-F (W/m-K) |
|--------------------|-----------------------------------------------------------------------------------------|------------------------------------------|
| 2.5                | 5.3 (9.5)                                                                               | 18 (31)                                  |
| 3.12               | 5.0 (9.1)                                                                               | 15 (26)                                  |
| 3.5                | 4.9 (8.8)                                                                               | 13 (23)                                  |
| 3.9                | 4.8 (8.6)                                                                               | 12 (20)                                  |

(1) Data obtained from Walzen Irle, GmbH, Netphen-Deuz, Germany.

(2) Coefficient of thermal expansion is given as an average value between the temperatures of 70–390F (20–200C).

• **Ni-Hard 1 and Ni-Hard 2 AR Cast Irons**

**Table B4**  
**Typical Range of Mechanical Properties of Ni-Hard 1 and Ni-Hard 2 AR Cast Irons<sup>(1)</sup>**

|                                | Brinell Hardness | Tensile Strength kpsi (MPa) | Transverse Strength <sup>(2)</sup> kpsi (MPa) | Deflection <sup>(2)</sup> in. (mm) | Modulus of Elasticity Mpsi (GPa) | Impact Energy <sup>(3)</sup> lb-ft (J) |
|--------------------------------|------------------|-----------------------------|-----------------------------------------------|------------------------------------|----------------------------------|----------------------------------------|
| <b>Ni-Hard 1 (3.0–3.6% C):</b> |                  |                             |                                               |                                    |                                  |                                        |
| Sand Cast <sup>(4)</sup>       | 550–690          | 40–50 (280–350)             | 70–90 (500–620)                               | 0.08–0.11 (2.0–2.8)                | 24–26 (169–183)                  | 38–56 (28–41)                          |
| Chill Cast <sup>(4)</sup>      | 600–730          | 50–60 (350–420)             | 80–125 (560–850)                              | 0.08–0.12 (2.0–3.0)                | 24–26 (169–183)                  | 48–75 (35–55)                          |
| <b>Ni-Hard 2 (max 2.9% C):</b> |                  |                             |                                               |                                    |                                  |                                        |
| Sand Cast <sup>(4)</sup>       | 530–630          | 45–55 (320–390)             | 80–100 (560–680)                              | 0.10–0.12 (2.5–3.0)                | 24–26 (169–183)                  | 48–65 (35–48)                          |
| Chill Cast <sup>(4)</sup>      | 580–680          | 60–75 (420–530)             | 100–125 (680–870)                             | 0.10–0.12 (2.5–3.0)                | 24–26 (169–183)                  | 65–105 (48–76)                         |

(1) Data obtained from Ni-Hard Material Data and Applications, Edited by K. Röhrig, NiDI, Ref. Book Series No. 11017 (1995).

(2) Transverse strength tests were performed on 30 mm (1.2 in.) dia test bars centered over a 300 mm (12 in.) span.

(3) Impact tests were conducted using an Izod test procedure on 1.2 in (30 mm) dia bars struck 76 mm (3 in.) above support.

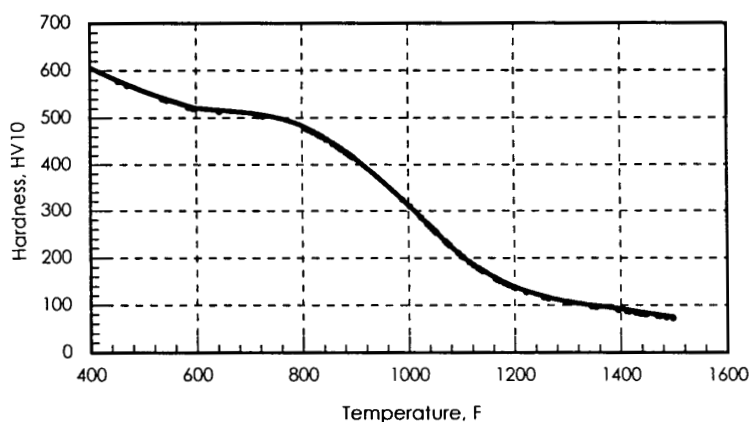
(4) Samples were cast into 30 mm (1.2 in.) dia bars and have nominal compositions for their alloy range.

**Table B5**  
**Some Typical Physical Properties of Ni-Hard 1 and Ni-Hard 2<sup>(1)</sup>**

|                                                                                                                              |                     |
|------------------------------------------------------------------------------------------------------------------------------|---------------------|
| Density lb/in. <sup>3</sup> (g/cm <sup>3</sup> )                                                                             | 0.27–0.28 (7.6–7.8) |
| Thermal Conductivity Btu/hr-ft-F (W/m·K):                                                                                    |                     |
| · Instantaneous at 70F (20C)                                                                                                 | 7.8–8.3 (14–15)     |
| · Instantaneous at 250F (120C)                                                                                               | 7.9 (14.2)          |
| · Instantaneous at 850F (450C)                                                                                               | 10.4 (18.8)         |
| Coefficient of Thermal Expansion $\epsilon/F$ ( $\epsilon/C$ ) $\times 10^6$ :                                               |                     |
| · Average between 50–200F (10–95C)                                                                                           | 4.5–5.0 (8–9)       |
| · Average between 50–480 F (10–250C)                                                                                         | 6.3–6.6 (11.3–11.9) |
| · Average between 50–800F (10–425C)                                                                                          | 6.8–7.1 (12.2–12.8) |
| $\epsilon$ = strength.                                                                                                       |                     |
| (1) Data obtained from Ni-Hard Material Data and Applications, Edited by K. Röhrig, NiDI, Ref. Book Series No. 11017 (1995). |                     |

**Table B6**  
**Fracture Toughness of Ni-Hard 1 Alloys**

| Alloy and Test Method                                                                                                                                                                                                                                       | Brinell Hardness | Fracture Toughness<br>kpsi·in. <sup>1/2</sup> (MPa·m <sup>1/2</sup> ) |
|-------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------|------------------|-----------------------------------------------------------------------|
| Ni-Hard 1 <sup>(1)</sup> - 3.7% C - Double Torsion Fracture Bend Test <sup>(1)</sup>                                                                                                                                                                        |                  |                                                                       |
| · Alloy a, as cast                                                                                                                                                                                                                                          | 610              | 23 (26)                                                               |
| · Alloy b, 4 hr 500F (275C), air cooled                                                                                                                                                                                                                     | 580              | 24 (26)                                                               |
| · Alloy c, 4 hr 1470F (800C), air cooled                                                                                                                                                                                                                    | 660              | 17 (19)                                                               |
| Ni-Hard 1 <sup>(2)</sup> - 3.2% C - ASTM E23-72 K <sub>IC</sub> Fracture Toughness Test <sup>(2)</sup>                                                                                                                                                      |                  |                                                                       |
| · Alloy 50A, as-cast                                                                                                                                                                                                                                        | 540              | 22 (24)                                                               |
| (1) Data obtained from I.R. Sare, "Abrasion Resistance and Fracture Toughness of White Cast Irons," Met. Technol., vol. 6, pp. 412-419 (1979).                                                                                                              |                  |                                                                       |
| (2) Data obtained from D.E. Diesburg and F. Borik, "Optimizing Abrasion Resistance and Toughness in Steels and Irons for the Mining Industry," in Symposium Materials for the Mining Industry, Vail, CO. July 30-31, 1974, pp. 15-42, Climax Molybdenum Co. |                  |                                                                       |



**Fig. B1. Hot hardness testing of a heavy-section Ni-Hard 1 casting. (Data obtained from Ni-Hard Material Data and Applications, Edited by K. Röhrig, NiDI, Ref. Book Series No. 11017, 1995)**

• **Ni-Hard 4 AR Cast Irons**

**Table B7**  
**Typical Range of Mechanical Properties of Ni-Hard 4 AR Cast Irons**

|                                            | <b>Brinell<br/>Hardness</b> | <b>Tensile<br/>Strength<br/>kpsi (MPa)</b> | <b>Transverse<br/>Strength<br/>kpsi (MPa)</b> | <b>Deflection<br/>in. (mm)</b> | <b>Modulus of<br/>Elasticity<br/>Mpsi (GPa)</b> |
|--------------------------------------------|-----------------------------|--------------------------------------------|-----------------------------------------------|--------------------------------|-------------------------------------------------|
| Ni-Hard 4 <sup>(1)</sup> (2.6-3.2% C)      | 550–700                     | 70–85 (500–600)                            | 85–115 (600–800)                              | 0.08–0.11 (2.0–2.8)            | 27–29 (190–200)                                 |
| Ni-Hard 4 <sup>(2)</sup> (3.3% C, 2.1% Si) | 640                         | not avail.                                 | 100 (695)                                     | 0.026 (0.66)                   | not avail.                                      |
| Ni-Hard 4 <sup>(2)</sup> (3.8% C, 1.6% Si) | 580                         | not avail.                                 | 77 (533)                                      | 0.018 (0.46)                   | not avail.                                      |
| Ni-Hard 4 <sup>(2)</sup> (3.6% C, 1.2% Si) | 530                         | not avail.                                 | 73 (505)                                      | 0.021 (0.53)                   | not avail.                                      |

(1) Data obtained from Ni-Hard Material Data and Applications, Edited by K. Röhrig, NiDI, Ref. Book Series No. 11017 (1995). Transverse strength tests were performed on 30 mm (1.2 in.) diameter test bars centered over a 300 mm (12 in.) span. Impact tests were conducted using an Izod test procedure on 1.2 in. (30 mm) diameter bars struck 76 mm (3 in.) above support. Samples were cast into 30 mm (1.2 in.) diameter bars and have nominal compositions for their alloy range.

(2) Data obtained from J.J. Fischer, "The Relationship of Microstructure to Properties in Ni-Hard 4," AFS Transactions, vol. 91, pp. 47-54 (1983). Other alloy elements had nominal values of 5.2% Ni, 7.8% Cr, 0.5% Mn. Bend tests were conducted on carefully machined and polished bars that were 127 mm (5 in.) long by 22 mm (0.9 in.) wide by 10 mm (0.4 in.) thick. A span distance of 82.5 mm (3.25 in.) was used. Bend samples were cut from casting such that the bending plane of tension was parallel to the chill face. Values reported are from samples that are estimated to be 10 mm (0.5 in.) from the chill face.

**Table B8**  
**Some Typical Physical Properties of Ni-Hard 4<sup>(1)</sup>**

|                                                                                                                              |               |
|------------------------------------------------------------------------------------------------------------------------------|---------------|
| Density lb/in. <sup>3</sup> (g/cm <sup>3</sup> )                                                                             | 0.27 (7.7)    |
| Thermal Conductivity Btu/hr-ft-F (W/m-K)                                                                                     | 7-9 (12.6–15) |
| Specific Heat at 70F (20C) Btu/lb-F (J/kg-K)                                                                                 | 0.119 (500)   |
| Coefficient of Thermal Expansion, $\epsilon/F$ ( $\epsilon/C$ )x10 <sup>6</sup> :                                            |               |
| · Average between 70–212F (10–100C)                                                                                          | 8.2 (14.8)    |
| · Average between 70–392F (10–200C)                                                                                          | 8.4 (15.2)    |
| · Average between 70–572F (10–300C)                                                                                          | 8.8 (15.8)    |
| · Average between 70–752F (10–400C)                                                                                          | 8.8 (15.9)    |
| · Average between 70–932F (10–500C)                                                                                          | 8.5 (15.3)    |
| $\epsilon$ = strength                                                                                                        |               |
| (1) Data obtained from Ni-Hard Material Data and Applications, Edited by K. Röhrig, NiDI, Ref. Book Series No. 11017 (1995). |               |

**Table B9**  
**Fracture Toughness of Ni-Hard 4 Alloys**

| <b>Alloy and Test Method</b>                                                       | <b>Brinell<br/>Hardness</b> | <b>Fracture Toughness<br/>kpsi·in.<sup>1/2</sup> (Mpa·m<sup>1/2</sup>)</b> |
|------------------------------------------------------------------------------------|-----------------------------|----------------------------------------------------------------------------|
| Ni-Hard 4 <sup>(1)</sup> - 3.5% C, 1.7% Si, 5.5% Ni, 9.6% Cr, 0.65% Mn, balance Fe |                             |                                                                            |
| Double Torsion Fracture Bend Test <sup>(1)</sup>                                   |                             |                                                                            |
| · Alloy a, as-cast                                                                 | 500                         | 27 (30)                                                                    |
| · Alloy b, 4 hr 1470F (800C), air cooled                                           | 700                         | 24 (27)                                                                    |
| · Alloy c, 4 hr 1470F (800C), air cooled, tempered 4 hr 525F (275C)                | 590                         | 24 (26)                                                                    |
| Ni-Hard 4 <sup>(2)</sup> - 3.4% C, 1.6% Si, 6.2% Ni, 8.6% Cr, 0.67% Mn, balance Fe |                             |                                                                            |
| ASTM E23-72 K <sub>IC</sub> Fracture Toughness Test <sup>(2)</sup>                 |                             |                                                                            |
| · Alloy 51B, 410F (210C), 1 hr                                                     | 510                         | 23 (25)                                                                    |

(1) Data obtained from I.R. Sare, "Abrasion Resistance and Fracture Toughness of White Cast Irons," Met. Technol., vol. 6, pp. 412-419 (1979).

(2) Data obtained from D.E. Diesburg and F. Borik, "Optimizing Abrasion Resistance and Toughness in Steels and Irons for the Mining Industry," in Symposium Materials for the Mining Industry, Vail, CO. July 30-31, 1974, pp. 15-42, Climax Molybdenum Co.

• **High-Chromium AR Cast Irons**

**Table B10**  
**Typical Range of Mechanical Properties of High-Cr AR Cast Iron**

|                                            | Brinell Hardness | Compressive Strength <sup>(2)</sup><br>kpsi (MPa) | Transverse Strength <sup>(3)</sup><br>kpsi (MPa) | Modulus of Elasticity <sup>(4)</sup><br>Mpsi (GPa) | Fracture Toughness <sup>(5)</sup><br>kpsi-in. <sup>1/2</sup> (Mpa·m <sup>1/2</sup> ) |
|--------------------------------------------|------------------|---------------------------------------------------|--------------------------------------------------|----------------------------------------------------|--------------------------------------------------------------------------------------|
| 56B <sup>(1)</sup> – 15Cr–3Mo (3.6% C)     | 410              | 238 (1640)                                        | 101 (696)                                        | 31–33 (215–230)                                    | 23.2 (25.5)                                                                          |
| 60 <sup>(1)</sup> – 15Cr–3Mo (3.6% C)      | 720              | 460 (3170)                                        | 71 (490)                                         | 31–33 (215–230)                                    | —                                                                                    |
| 81B <sup>(1)</sup> – 20Cr–2Mo–1Cu (2.9% C) | 740              | 416 (2870)                                        | 100 (689)                                        | 31–33 (215–230)                                    | 27.5 (30.2)                                                                          |
| 83B <sup>(1)</sup> – 27Cr (2.4% C)         | 650              | 419 (2890)                                        | 137 (945)                                        | 31–33 (215–230)                                    | —                                                                                    |
| 84B <sup>(1)</sup> – 27Cr (3.1% C)         | 700              | 429 (2960)                                        | 136 (938)                                        | 31–33 (215–230)                                    | —                                                                                    |

(1) Data obtained from D.E. Diesburg and F. Borik, "Optimizing Abrasion Resistance and Toughness in Steels and Irons for the Mining Industry," in Symposium Materials for the Mining Industry, Vail, CO. July 30-31, 1974, pp. 15-42, Climax Molybdenum Co.

(2) Compressive strength was measured following the ASTM E9-72 specification using 10 mm (0.4 in.) dia cylinders 20 mm (0.8 in.) long.

(3) Transverse strength tests were performed on 13 mm (0.5 in.) dia bars loaded in 4-point bending (span length is not important in the 4-point bend test).

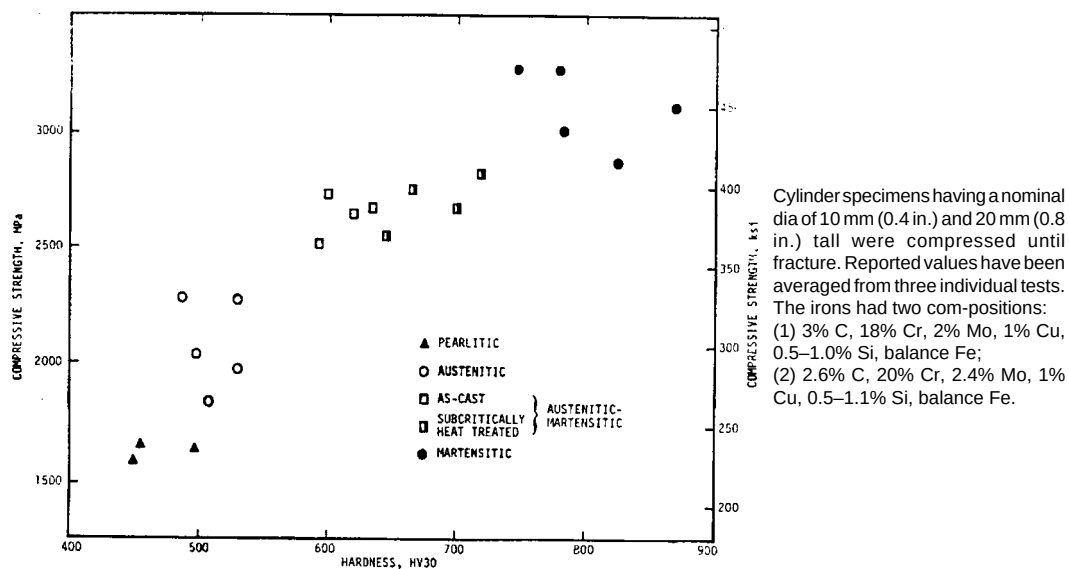
(4) Typical room temperature values from: P.E. Belyakova, et al., "Thermophysical Properties of Wear-Resisting Cast Irons," translated from Metallovedenie i Termicheskaya Obrabotka Metallov, No. 12 pp. 45-48 (1975).

(5) Fracture toughness tests were conducted following ASTM E399-72 using a compact specimen of 13x51x53 mm (0.5x2x2.125 in.).

**Table B11**  
**Some Typical Physical Properties of High-Cr AR Cast Iron<sup>(1)</sup>**

|                                                                                |                  |
|--------------------------------------------------------------------------------|------------------|
| Density lb/in. <sup>3</sup> (g/cm <sup>3</sup> )                               | 0.27 (7.6)       |
| Thermal Conductivity Btu/hr-ft-F (W/m-K)                                       | 8.6-17.3 (15–30) |
| Specific Heat at 70F (20C) Btu/lb-F (J/kg-K)                                   | 0.119 (500)      |
| Coefficient of Thermal Expansion $\epsilon/F$ ( $\epsilon/C$ ) $\times 10^6$ : |                  |
| · Average between 70–1112F (10–600C)                                           | 7.2 (13.0)       |
| $\epsilon$ = strength                                                          |                  |

(1) Data obtained from typical room temperature values from P.E. Belyakova, et al., "Thermophysical Properties of Wear-Resisting Cast Irons," translated from Metallovedenie i Termicheskaya Obrabotka Metallov, No. 12 pp. 45-48 (1975).



**Fig. B2. Typical values of compressive strength versus hardness for a series of high-Cr AR irons. (Figure obtained from J.L. Parks, "Characteristics of As-Cast and Subcritically Heat-Treated High-Chromium-Molybdenum White Irons for Thick-Section Castings, AFS Transactions, vol. 86, pp. 93-102, 1978).**

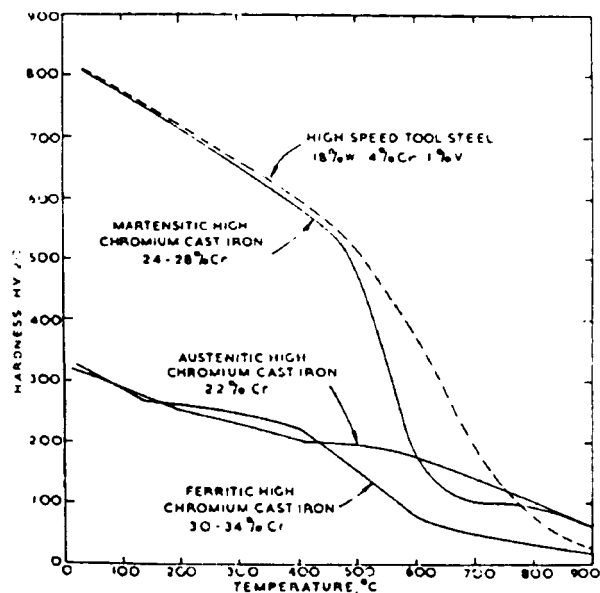
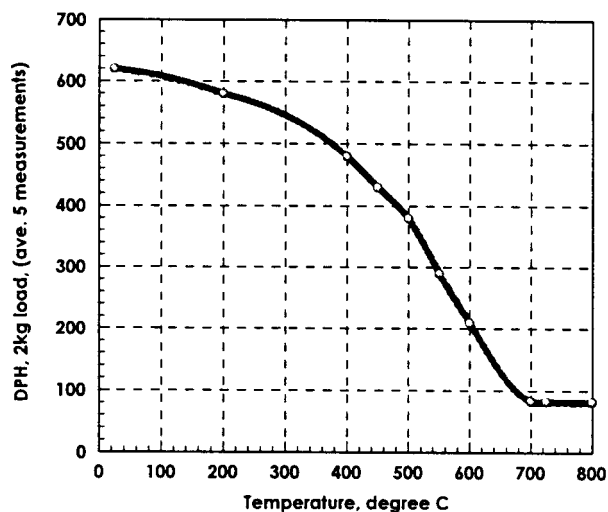


Fig. B3. Hot hardness of high-Cr AR cast irons. (Left) Hot hardness taken on a martensitically hardened 2.6% C, 16.9% Cr, 1.2% Si iron. (Courtesy of United Foundries) (Right) Hot hardness data taken on a variety of high-Cr AR cast irons. (Data obtained from J.W. Boyes and R.E. Phelps, "Hot Hardness Characteristics of High Chromium Cast Irons," BCIRA, vol. 12, pp. 54-63, 1964)

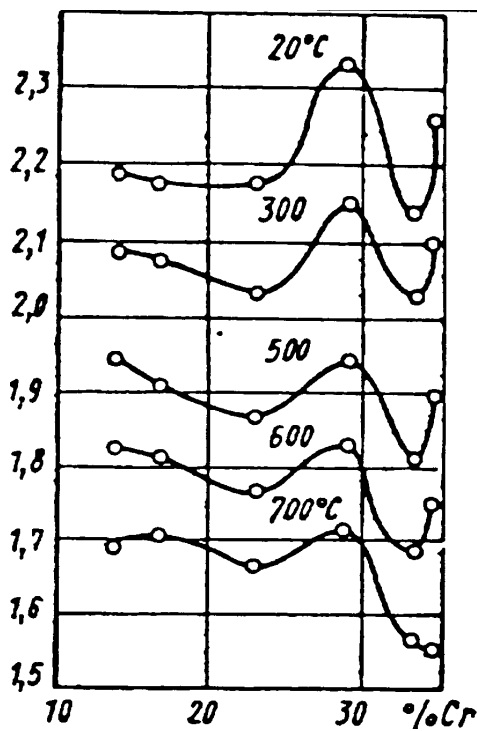
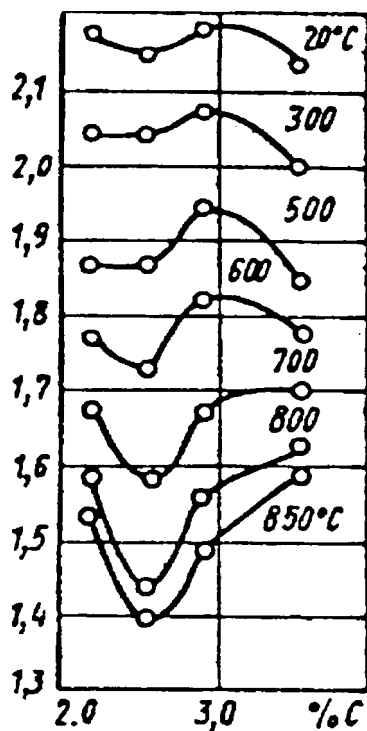
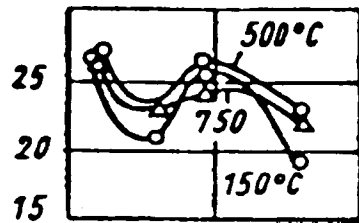


Fig. B4. Elastic modulus (GPa 100) in high-Cr irons, as a function of composition and temperature (left, carbon; right, chromium). (Data obtained from Typical room temperature values from P.E. Belyakova, et al., "Thermophysical Properties of Wear-Resisting Cast Irons," translated from *Metallovedenie i Termicheskaya Obrabotka Metallov*, No. 12 pp. 45-48, 1975).

$\lambda$ , W/m-deg C



$\lambda$ , W/m-deg C

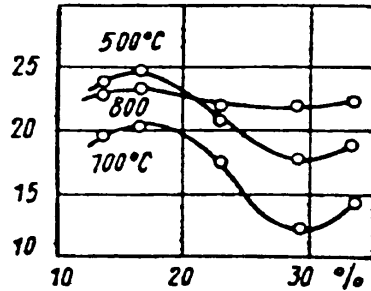


Fig. B5. Thermal conductivity as a function of composition (top, between 1.5 and 3.0% carbon; bottom, between 1.0 and 3.0% chromium) and temperature. (Data obtained from typical Room temperature values from P.E. Belyakova et al., "Thermophysical Properties of Wear-Resisting Cast Irons," translated from *Metallovedenie i Termicheskaya Obrabotka Metallov*, No. 12 pp. 45-48, 1975)

$C_p$ , kJ/kg-deg C

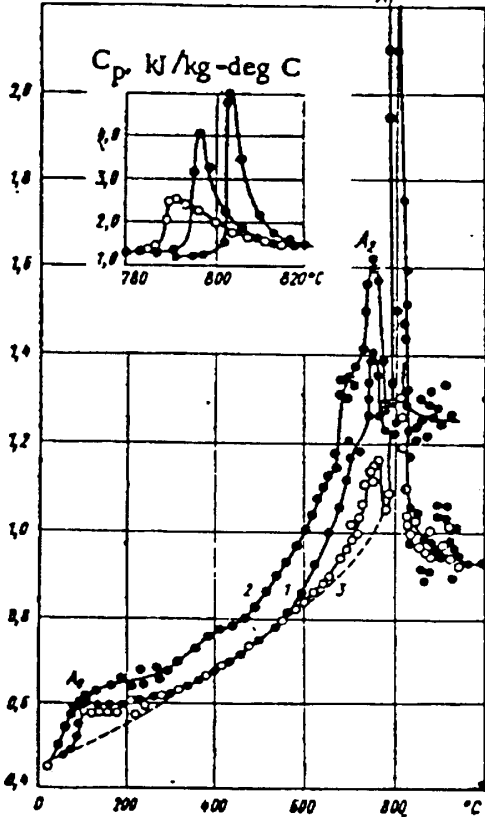


Fig. B6. Specific heat at carbon concentrations of 2.2, 2.9 and 3.6%, as a function of composition and temperature. (Data obtained from typical room temperature values from P.E. Belyakova et al., "Thermophysical Properties of Wear-Resisting Cast Irons," translated from *Metallovedenie i Termicheskaya Obrabotka Metallov*, No. 12 pp. 45-48, 1975).

## APPENDIX C

### General Mechanical and Wear Properties

#### Diesburg and Boric (1974)

The first broad survey of mechanical and wear property data is provided by D.E. Diesburg and F. Boric, "Optimizing Abrasion Resistance and Toughness in Steels and Irons for the Mining Industry," in *Symposium Materials for the Mining Industry*, Vail, Colorado, pp. 15-42 (1974). Their work is fairly comprehensive and provides a good starting point for the mechanical and wear characterization of AR cast irons. A caveat about their work is that no history is provided as to section size used for sample preparation or the surface finish used on the samples subjected to compressive and four-point bending tests. What their work does provide is a look at the mechanical and wear properties of a broad range of AR cast irons.

Tables C1 and C2 provide the chemical compositions and thermal histories of the irons. Table C3 gives the mechanical and wear data gathered on the irons described in Tables C1 and C2. For reference, a few steels have been included in the tables.

The list of irons covered in their work is exhaustive. What is particularly nice is the range of carbon contents (e.g., alloys 63 through 67) and heat treatments (e.g., alloys 78A through 78I) studied in their work. In some respects, there is something for everybody in their compilation.

Considering the laboratory basis of their work, one can guess that the toughness and abrasion testing samples were prepared from small sand cast blocks having a nominal thickness of 25 mm (1 in.). Using these fine-grained samples, several toughness and abrasion tests were performed on the samples.

The gouging test consists of a small laboratory scale jaw crusher. The test sample functions as the stationary jaw with a "standard" material used as the movable jaw. The standard material is the T1 grade of steel given at the end of Table C1. The test consists of crushing one ton of precrushed morainial rock having a size range of 38-50 mm (1.5- 2.0 in.). After the test, the plates are weighed, and the weight loss of the test sample is divided by weight loss of the standard. This numerical value was termed the "gouging ratio" by Diesburg and Borik. A low gouging ratio indicates good abrasion resistance.

The pin test abrades a 6.3 mm (0.250 in.) dia cylinder against 180 mesh alumina abrasive cloth. The pin is pressed against the cloth under a load of 63.75 Newton (15 lb<sub>f</sub>). As the pin is dragged across the cloth, it rotates at 20 rpm and travels a linear distance of 12.80 m (504 in.) in seven minutes. At the end of this test, the pin is weighed. Results presented in Table C3 were averaged from two tests. As can be guessed, a low weight loss indicates good abrasion resistance.

The rubber wheel test is covered in ASTM G 65 listed as "Standard Practice for Conducting Dry Sand/Rubber Wheel Abrasion Tests." Although Diesburg's and Borik's work was conducted prior to the implementation of this standard, the concept is similar. In their work, they press a small sample (22x22x50 mm) against a rotating rubber wheel while an abrasive flows between the sample and the wheel. The sample is held against the rubber wheel with a force of 222 Newton (50 lb<sub>f</sub>) while the 178 mm diameter wheel rotates at 240 rpm. The AFS 50- to 70-mesh silica sand provides the abrasive action. A test consists of 5000 revolutions. Some other complications are involved, but in brief, the weight loss of the sample is reported based on 5000 revolutions.

The toughness tests used by Diesburg and Borik follow ASTM Procedure E 23-72 for the Charpy and ASTM Procedure E399-72 for the fracture toughness tests. Compression tests were done using the ASTM Procedure E9-72. The specimens tested under compression were 10 mm diameter cylinders that were 20 mm long. Tensile data was evaluated by a four-point bend test using 13 mm dia unnotched bars. The surface finishes on the compression and bend bars were not

mentioned. This is a shame, considering the sensitivity of wear resistant cast irons to surface flaws. Consequently, the numbers quoted in Table C3 for compressive and bend strength should be used with care.

One of the strong features of their work is the comparison provided between wear resistant steels and irons. Figures C1a and C1b provide a summary of their work in this area. In these figures, the best abrasion resistance is provided by high-chromium irons. This is no big surprise, but it is nice to see it quantitatively graphed.

**Table C1**  
**Type and Composition of Iron and Steel Alloys<sup>(1)</sup>**

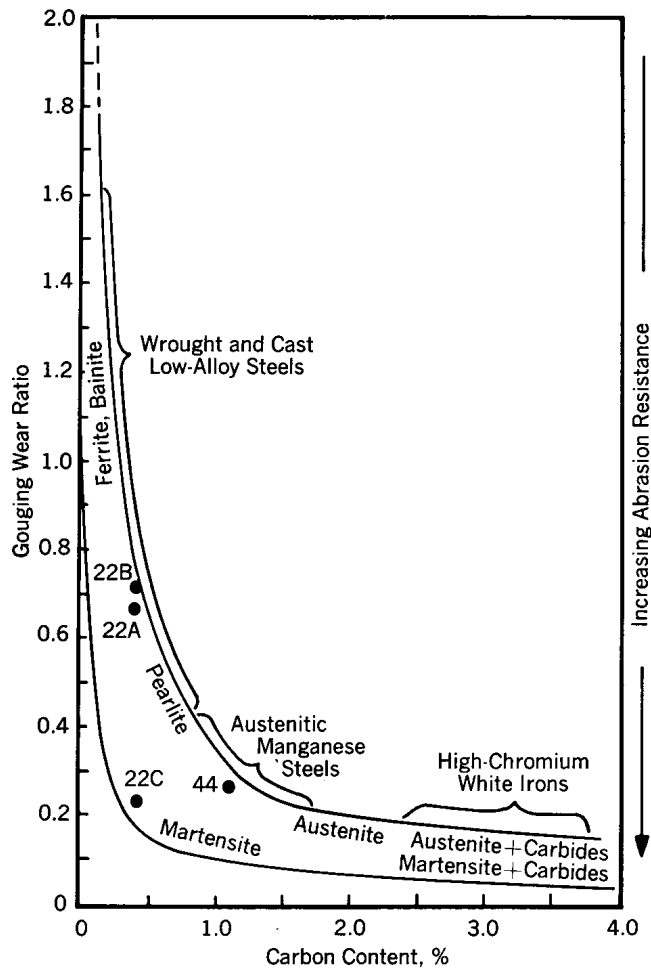
| Code No. | Irons                | C    | Mn    | Si   | Cr    | Ni   | Mo   | P     | S     | Other                  |
|----------|----------------------|------|-------|------|-------|------|------|-------|-------|------------------------|
| 50       | 2Cr-4Ni              | 3.19 | 0.64  | 0.57 | 2.20  | 3.94 | —    | 0.031 | 0.033 | —                      |
| 51       | 9Cr-6Ni              | 3.42 | 0.67  | 1.61 | 8.64  | 6.17 | —    | 0.029 | 0.036 | —                      |
| 52       | 12Cr-1Mo-1Cu         | 3.25 | 0.70  | 0.62 | 12.7  | —    | 0.68 | —     | —     | 1.20                   |
| 53       | 15Cr-3Mo             | 2.51 | 0.80  | 0.47 | 14.7  | —    | 2.62 | —     | —     | —                      |
| 54       | 15Cr-3Mo             | 3.59 | 0.8   | 0.4  | 15.0  | —    | 2.6  | —     | —     | —                      |
| 55       | 15Cr-3Mo             | 3.60 | 0.8   | 0.4  | 15.0  | —    | 2.6  | —     | —     | —                      |
| 56       | 15Cr-3Mo             | 3.60 | 0.84  | 0.42 | 14.95 | —    | 2.62 | —     | —     | —                      |
| 57       | 15Cr-3Mo             | 3.60 | 0.8   | 0.4  | 15.0  | —    | 2.69 | —     | —     | —                      |
| 58       | 15Cr-3Mo             | 3.60 | 0.8   | 0.4  | 15.0  | —    | 2.60 | —     | —     | —                      |
| 59       | 15Cr-3Mo             | 3.60 | 0.85  | 0.42 | 14.96 | —    | 2.61 | —     | —     | —                      |
| 60       | 15Cr-3Mo             | 3.68 | 0.8   | 0.4  | 15.0  | —    | 2.6  | —     | —     | —                      |
| 61       | 15Cr-3Mo             | 3.62 | 0.8   | 0.4  | 15.0  | —    | 2.6  | —     | —     | —                      |
| 62       | 15Cr-2Mo-1Cu         | 3.25 | 0.66  | 0.49 | 15.12 | —    | 1.95 | 0.039 | 0.029 | 1.08Cu                 |
| 63       | 17Cr-1.5Mo-1Cu       | 1.23 | 0.70  | 0.47 | 17.69 | —    | 1.49 | 0.044 | 0.025 | 1.08Cu                 |
| 64       | 17Cr-1.5Mo-1Cu       | 1.83 | 0.74  | 0.46 | 16.95 | —    | 1.56 | 0.040 | 0.029 | 1.10Cu                 |
| 65       | 17Cr-1.5Mo-1Cu       | 2.43 | 0.75  | 0.47 | 16.99 | —    | 1.58 | 0.040 | 0.029 | 1.00Cu                 |
| 66       | 17Cr-1.5Mo-1Cu       | 3.01 | 0.75  | 0.44 | 17.06 | —    | 1.60 | 0.040 | 0.029 | 1.06Cu                 |
| 67       | 17Cr-1.5Mo-1Cu       | 3.65 | 0.76  | 0.50 | 17.35 | —    | 1.60 | 0.039 | 0.025 | 1.13Cu                 |
| 68       | 18Cr-1Cu             | 2.89 | 0.78  | 0.57 | 18.0  | —    | 0.03 | 0.036 | 0.024 | 0.96Cu                 |
| 69       | 18Cr-1Cu             | 2.87 | 0.78  | 0.59 | 18.1  | —    | 0.04 | 0.029 | 0.029 | 0.99Cu                 |
| 70       | 18Cr-1Mo-1Cu         | 2.85 | 0.78  | 0.59 | 18.1  | —    | 1.10 | 0.039 | 0.025 | 0.99Cu                 |
| 71       | 18Cr-1Mo-1Cu         | 2.90 | 0.76  | 0.57 | 18.2  | —    | 1.02 | 0.028 | 0.096 | 1.00Cu                 |
| 72       | 18Cr-2Mo-1Cu         | 2.89 | 0.76  | 0.58 | 18.1  | —    | 2.12 | 0.025 | 0.097 | 1.00Cu                 |
| 73       | 18Cr-2Mo-1Cu         | 2.90 | 0.77  | 0.59 | 18.2  | —    | 2.13 | 0.026 | 0.028 | 1.00Cu                 |
| 74       | 18Cr-2Mo-1Cu         | 3.06 | 0.76  | 0.41 | 18.2  | —    | 1.96 | 0.036 | 0.028 | 0.98Cu                 |
| 75       | 18Cr-2Mo-1Cu         | 3.06 | 0.76  | 0.90 | 18.2  | —    | 1.96 | 0.036 | 0.028 | 0.98Cu                 |
| 76       | 18Cr-2Mo-1Cu         | 3.06 | 0.76  | 1.17 | 18.2  | —    | 1.96 | 0.036 | 0.028 | 0.98Cu                 |
| 77       | 18Cr-2Mo-1Cu         | 3.06 | 0.76  | 1.60 | 18.2  | —    | 1.96 | 0.036 | 0.028 | 0.98Cu                 |
| 78       | 18Cr-2Mo-1Cu         | 3.28 | 0.75  | 0.61 | 18.2  | —    | 2.00 | 0.042 | 0.029 | 0.98Cu                 |
| 79       | 18Cr-3Mo-1Cu         | 2.87 | 0.77  | 0.57 | 18.1  | —    | 3.02 | 0.037 | 0.026 | 0.98Cu                 |
| 80       | 18Cr-3Mo-1Cu         | 2.83 | 0.76  | 0.59 | 17.8  | —    | 3.14 | 0.039 | 0.027 | 1.00Cu                 |
| 81       | 18Cr-2Mo-1Cu         | 2.89 | 0.73  | 0.55 | 19.35 | —    | 2.08 | 0.015 | 0.016 | 1.02Cu                 |
| 82       | 27Cr                 | 2.68 | 0.83  | 0.36 | 27.60 | —    | —    | 0.009 | 0.008 | —                      |
| 83       | 27Cr                 | 2.38 | 0.8   | 0.4  | 26.67 | —    | —    | —     | —     | —                      |
|          | Steels               | C    | Mn    | Si   | Cr    | Ni   | Mo   | P     | S     | Other                  |
| 7        | T-1 Wrought Steel    | 0.19 | 0.84  | 0.28 | 0.56  | —    | 0.18 | 0.011 | 0.021 | 0.02Ti, 0.05V, 0.0032B |
| 21       | Cr-Ni-Mo Wrought     | 0.27 | —     | —    | 1.0   | 3.0  | 2.0  | —     | —     | —                      |
| 22       | AISI 4340            | 0.40 | 0.7   | 0.3  | 0.8   | 1.8  | 0.25 | —     | —     | —                      |
| 30       | Cr-Mo Cast Steel     | 0.88 | 0.95  | 0.72 | 2.44  | —    | 0.35 | 0.027 | —     | —                      |
| 31       | AISI H-11 Tool Steel | 0.35 | —     | —    | 5.0   | —    | 1.5  | —     | —     | 0.4V                   |
| 41       | 12Mn (0.95C)         |      |       |      |       |      |      |       |       |                        |
|          | Cast Austenitic Mn   | 0.93 | 12.97 | 0.50 | —     | —    | —    | —     | —     | —                      |
| 44       | 12Mn (1.1C)          |      |       |      |       |      |      |       |       |                        |
|          | Cast Austenitic Mn   | 1.1  | 12.5  | 0.5  | —     | —    | —    | —     | —     | —                      |

(1) Diesburg and Borik, 1974.

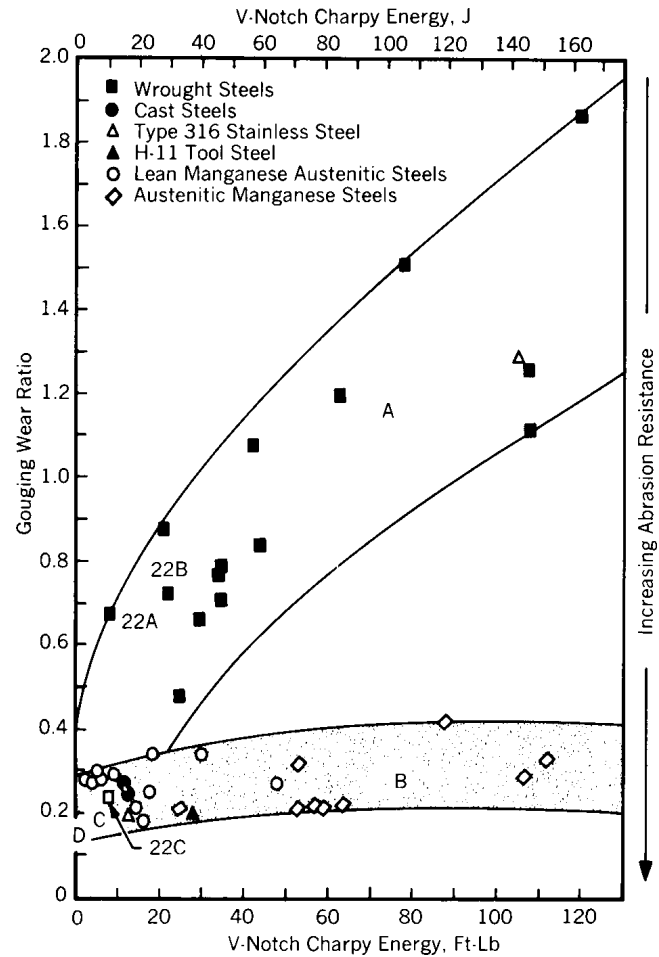
**Table C2**  
**Heat Treatment of Iron and Steel Alloys<sup>(1)</sup>**

| <b>Code No.</b> | <b>Irons</b>   | <b>Heat Treatment Schedule</b>                                                                                              | <b>Code No.</b> | <b>Irons</b>                          | <b>Heat Treatment Schedule</b>                                                                                                             |
|-----------------|----------------|-----------------------------------------------------------------------------------------------------------------------------|-----------------|---------------------------------------|--------------------------------------------------------------------------------------------------------------------------------------------|
| 50A             | 2Cr-4Ni        | None                                                                                                                        | 74B to          |                                       |                                                                                                                                            |
| 50B             | 2Cr-4Ni        | 410F (210C) - 1 hr                                                                                                          | 77B             | 18Cr-2Mo-1Cu                          | 1750F (955C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 410F (210C) - 2 hr                                                              |
| 51A             | 9Cr-6Ni        | 410F (210C) - 1 hr                                                                                                          | 78A             | 18Cr-2Mo-1Cu                          | 410F (210C) - 2 hr                                                                                                                         |
| 51B             | 9Cr-6Ni        | 410F (210C) - 2 hr                                                                                                          | 78B             | 18Cr-2Mo-1Cu                          | 1700F (925C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 410F (210C) - 2 hr                                                              |
| 52A             | 12Cr-1Mo-1Cu   | None                                                                                                                        | 78C             | 18Cr-2Mo-1Cu                          | 1750F (955C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 410F (210C) - 2 hr                                                              |
| 52B             | 12Cr-1Mo-1Cu   | 1750F (955C) - 1 hr, oil quenched, 400F (205C)                                                                              | 78D             | 18Cr-2Mo-1Cu                          | 1850F (1010C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 410F (210C) - 2 hr                                                             |
| 53A             | 15Cr-3Mo       | None                                                                                                                        | 78E             | 18Cr-2Mo-1Cu                          | 2100F (1150C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 410F (210C) - 2 hr                                                             |
| 53B             | 15Cr-3Mo       | 1850F (1010C) - 2 hr, air cooled, 400F (205C)                                                                               | 78F             | 18Cr-2Mo-1Cu                          | 2100F (1150C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 1750F (955C) - 4 hr, cooled as 2-in. (50 mm) plate in air, 410F (210C) - 2 hr  |
| 54              | 15Cr-3Mo       | 1650F (900C) - 6 hr, furnace cooled, 400F (205)                                                                             | 78G             | 18Cr-2Mo-1Cu                          | 2150F (1175C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 1750F (955C) - 4 hr, cooled as 2-in. (50 mm) plate in air, 410F (210C) - 2 hr  |
| 55              | 15Cr-3Mo       | 1650F (900C) - 1 hr, oil quenched, 400F (205)                                                                               | 78H             | 18Cr-2Mo-1Cu                          | 2200F (1205C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 1750F (955C) - 4 hr, cooled as 2-in. (50 mm) plate in air, 410F (210C) - 2 hr  |
| 56A             | 15Cr-3Mo       | None                                                                                                                        | 78I             | 18Cr-2Mo-1Cu                          | 2200F (1205C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 1850F (1010C) - 4 hr, cooled as 2-in. (50 mm) plate in air, 410F (210C) - 2 hr |
| 56B             | 15Cr-3Mo       | 1650F (900C) - 1 hr, furnace cooled, 400F (205)                                                                             | 79              | 18Cr-3Mo-1Cu                          | 410F (210C) - 2 hr                                                                                                                         |
| 56C             | 15Cr-3Mo       | 1650F (900C) - 1 hr, furnace cooled, 400F (205), 1850F (1010C) - 1 hr, air cooled, 410F (210C)                              | 80              | 18Cr-3Mo-1Cu                          | 1750F (955C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 410F (210C) - 2 hr                                                              |
| 57              | 15Cr-3Mo       | 1650F (900C) - 6 hr, oil quenched, 400F (205)                                                                               | 81A             | 18Cr-2Mo-1Cu                          | None                                                                                                                                       |
| 58              | 15Cr-3Mo       | 1800F (980C) - 1 hr, oil quenched, 400F (205)                                                                               | 81B             | 18Cr-2Mo-1Cu                          | 1750F (955C) - 1 hr, air cooled, 400F (205C) - 2 hr                                                                                        |
| 59A             | 15Cr-3Mo       | None                                                                                                                        | 81C             | 18Cr-2Mo-1Cu                          | 1750F (955C) - 1 hr, air cooled, 400F (205C), 1850F (1010C) - 1 hr, air cooled, 410F (210C) - 2 hr                                         |
| 59B             | 15Cr-3Mo       | 1800F (980C) - 1 hr, furnace cooled, 400F (205)                                                                             | 82              | 27Cr                                  | 410F (210C) - 2 hr                                                                                                                         |
| 60              | 15Cr-3Mo       | 1800F (980C) - 6 hr, oil quenched, 400F (205)                                                                               | 83A             | 27Cr                                  | None                                                                                                                                       |
| 61              | 15Cr-3Mo       | 1800F (980C) - 6 hr, furnace cooled, 400F (205)                                                                             | 83B             | 27Cr                                  | 1850F (1010C) - 2 hr, air cooled, 450F (230C)                                                                                              |
| 62A             | 15Cr-2Mo-1Cu   | 410F (210C) - 2 hr                                                                                                          | 84A             | 27Cr                                  | None                                                                                                                                       |
| 62B             | 15Cr-2Mo-1Cu   | 1850F (1010C) - 4 hr                                                                                                        | 84B             | 27Cr                                  | 1850F (1010C) - 2 hr, air cooled, 450F (230C)                                                                                              |
| 62C             | 15Cr-2Mo-1Cu   | 1850F (1010C) - 4 hr, air cooled, 410F (210C) - 2 hr                                                                        |                 |                                       |                                                                                                                                            |
| 62D             | 15Cr-2Mo-1Cu   | 1850F (1010C) - 4 hr, air cooled, 600F (315C) - 2 hr                                                                        | <b>No.</b>      | <b>Steels</b>                         | <b>Heat Treatment Schedule</b>                                                                                                             |
| 62E             | 15Cr-2Mo-1Cu   | 1850F (1010C) - 4 hr, air cooled, 800F (425C) - 2 hr                                                                        | 7A              | T-1 Wrought Steel                     | Hot-rolled 1-in. (25 mm) plate, 1700F (925C), water quenched, 1200F (650C)                                                                 |
| 62F             | 15Cr-2Mo-1Cu   | 1750F (955C) - 4 hr, air cooled, 410F (210C) - 2 hr                                                                         | 21A             | Cr-Ni-Mo Wrought Steel                | Hardened and tempered to 510 HB                                                                                                            |
| 62G             | 15Cr-2Mo-1Cu   | 2150F (1175C) - 4 hr, furnace cooled to 1850F (1010C) in 3 hr, held at 1850F (1010C) - 4 hr, air cooled, 410F (210C) - 2 hr | 22A             | AISI 4340 (0.4C)                      | 1750F (955C), air cooled                                                                                                                   |
| 62H             | 15Cr-2Mo-1Cu   | 1850F (1010C) - 4 hr, program cooled to form bainite, air cooled, 410F (210C) - 2 hr                                        | 22B             | AISI 4340 (0.4C)                      | 1600F (870C), oil quenched, 1200F (650C)                                                                                                   |
| 63A to          |                |                                                                                                                             | 22C             | AISI 4340 (0.4C)                      | 1600F (870C), oil quenched, 400F (205C)                                                                                                    |
| 67A             | 17Cr-1.5Mo-1Cu | 410F (210C) - 2 hr                                                                                                          | 30              | Cr-Mo Cast Steel                      | 1800F (980), air cooled, 1075F (580C)                                                                                                      |
| 63B to          |                |                                                                                                                             | 31A             | AISI H-11 Tool Steel                  | Hardened and tempered at 400F (205C)                                                                                                       |
| 67B             | 17Cr-1.5Mo-1Cu | 1850F (1010C) - 4 hr, air cooled, 410F (210C) - 2 hr                                                                        | 41              | 12Mn (0.95C) Cast Austenitic Mn Steel | 1900F (1040C) - 4 hr, water quenched                                                                                                       |
| 68              | 18Cr-1Cu       | 410F (210C) - 2 hr                                                                                                          | 44              | 12Mn (1.1C) Cast Austenitic Mn Steel  | 1900F (1040C) - 4 hr, water quenched                                                                                                       |
| 69              | 18Cr-1Cu       | 1750F (955C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 410F (210C) - 2 hr                                               |                 |                                       |                                                                                                                                            |
| 70              | 18Cr-1Mo-1Cu   | 410F (210C) - 2 hr                                                                                                          |                 |                                       |                                                                                                                                            |
| 71              | 18Cr-1Mo-1Cu   | 1750F (955C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 410F (210C) - 2 hr                                               |                 |                                       |                                                                                                                                            |
| 72              | 18Cr-2Mo-1Cu   | 410F (210C) - 2 hr                                                                                                          |                 |                                       |                                                                                                                                            |
| 73              | 18Cr-2Mo-1Cu   | 1750F (955C) - 4 hr, cooled as 2-in. (50-mm) plate in air, 410F (210C) - 2 hr                                               |                 |                                       |                                                                                                                                            |
| 74A to          |                |                                                                                                                             |                 |                                       |                                                                                                                                            |
| 77A             | 18Cr-2Mo-1Cu   | 410F (210C) - 2 hr                                                                                                          |                 |                                       |                                                                                                                                            |

(1) Diesburg and Borik, 1974.



**Fig. C1a.** Relationship between gouging wear ratio (a severe abrasive wear test) and the C content of various types of steels and irons. The numbers in the figure correspond to specific irons and steels provided in Tables C1, C2 and C3, authored by Diesburg and Borik (1974).



**Fig. C1b.** Relationship between resistance to gouging abrasion and toughness of various materials. Area A = wrought and cast low-alloy steels; Area B = austenitic Mn steels; Area C = variety of heat-treated steels; Area D = high-Cr irons. The numbers in the figure correspond to specific irons and steels provided in Tables C1, C2 and C3, authored by Diesburg and Borik (1974).

Table C3  
Mechanical Properties and Abrasion Resistance of Iron and Steel Alloys<sup>(1)</sup>

| Code No. | Irons                 | Mechanical Properties |     |                                          |                                      |                                       | Abrasion Resistance |                       |                                |            | Principal Components of Microstructure |
|----------|-----------------------|-----------------------|-----|------------------------------------------|--------------------------------------|---------------------------------------|---------------------|-----------------------|--------------------------------|------------|----------------------------------------|
|          |                       | HRC                   | HB  | Fracture Toughness<br>ksi√in.<br>(Mpa√m) | Compressive Strength<br>ksi<br>(MPa) | Bend Tensile Strength<br>ksi<br>(MPa) | Gouging Wear Ratio  | Pin Test (grams lost) | Rubber Wheel Test (grams lost) |            |                                        |
|          |                       |                       |     |                                          |                                      |                                       |                     |                       |                                |            |                                        |
| 50A      | 2Cr—4Ni               | 53.2                  | 547 | 22.0 (24.2)                              | —                                    | —                                     | —                   | —                     | 0.035                          | A,C,M      |                                        |
| 50B      | 2Cr—4Ni               | 53.2                  | 547 | —                                        | 384 (2650)                           | —                                     | —                   | —                     | —                              | A,C,M      |                                        |
| 51A      | 9Cr—6Ni               | 50.7                  | 509 | —                                        | —                                    | —                                     | 0.151               | —                     | 0.055                          | A,C,M      |                                        |
| 51B      | 9Cr—6Ni               | 50.3                  | 503 | 23.2 (25.5)                              | 314 (2160)                           | —                                     | —                   | —                     | —                              | A,C,M      |                                        |
| 52A      | 12Cr—1Mo—1Cu          | 50.5                  | 506 | —                                        | 282 (1940)                           | 94 (648)                              | —                   | —                     | —                              | P,C,a      |                                        |
| 52B      | 12Cr—1Mo—1Cu          | 63                    | 740 | —                                        | 455 (3140)                           | —                                     | 0.073               | —                     | 0.048                          | M,C,A,SC   |                                        |
| 53A      | 15Cr—3Mo              | 47.1                  | 462 | —                                        | 295 (2030)                           | —                                     | —                   | —                     | —                              | A,C        |                                        |
| 53B      | 15Cr—3Mo              | 61.7                  | 710 | 24.0 (26.4)                              | 436 (3000)                           | —                                     | 0.076               | —                     | 0.065                          | M,C,A,SC   |                                        |
| 54       | 15Cr—3Mo              | 41.3                  | 407 | —                                        | 229 (1580)                           | 100 (689)                             | 0.435               | —                     | 0.142                          | P,C,SC,S   |                                        |
| 55       | 15Cr—3Mo              | 67.9                  | 862 | —                                        | 472 (3250)                           | —                                     | 0.048               | —                     | 0.037                          | M,C,A,SC   |                                        |
| 56A      | 15Cr—3Mo              | —                     | —   | —                                        | 339 (2340)                           | 74 (510)                              | —                   | —                     | —                              | M,C,A,SC,P |                                        |
| 56B      | 15Cr—3Mo              | 42.4                  | 415 | 23.2 (25.5)                              | 238 (1640)                           | 101 (696)                             | 0.407               | —                     | 0.140                          | P,C,SC,S   |                                        |
| 56C      | 15Cr—3Mo              | 63.8                  | 758 | —                                        | 443 (3050)                           | —                                     | —                   | —                     | —                              | M,C,A,SC   |                                        |
| 57       | 15Cr—3Mo              | 67.5                  | 851 | —                                        | 465 (3210)                           | —                                     | 0.044               | —                     | 0.036                          | M,C,A,SC   |                                        |
| 58       | 15Cr—3Mo              | 64.7                  | 780 | —                                        | 452 (3120)                           | 83 (572)                              | 0.053               | —                     | 0.044                          | M,C,A,SC   |                                        |
| 59A      | 15Cr—3Mo              | 56.4                  | 602 | —                                        | 336 (2320)                           | 92 (634)                              | —                   | —                     | —                              | M,C,A,SC,P |                                        |
| 59B      | 15Cr—3Mo              | 43.3                  | 423 | —                                        | 236 (1630)                           | 71 (490)                              | —                   | —                     | 0.142                          | P,C,SC,S   |                                        |
| 60       | 15Cr—3Mo              | 62.8                  | 735 | —                                        | 460 (3170)                           | 94 (648)                              | 0.062               | —                     | 0.039                          | M,C,A,SC   |                                        |
| 61       | 15Cr—3Mo              | 45                    | 439 | —                                        | 236 (1630)                           | —                                     | 0.331               | —                     | 0.142                          | P,C,SC,S   |                                        |
| 62A      | 15Cr—2Mo—1Cu          | 48.8                  | 483 | 21.9 (24.1)                              | 341 (2350)                           | —                                     | 0.093               | 0.0424                | —                              | A,C,SC     |                                        |
| 62B      | 15Cr—2Mo—1Cu          | 60                    | 673 | 20.9 (23.0)                              | 477 (3290)                           | —                                     | —                   | 0.0553                | —                              | A,C,M,SC   |                                        |
| 62C      | 15Cr—2Mo—1Cu          | 58.9                  | 650 | 23.2 (25.5)                              | 473 (3260)                           | —                                     | —                   | 0.0448                | —                              | A,C,M,SC   |                                        |
| 62D      | 15Cr—2Mo—1Cu          | 57.1                  | 615 | 21.8 (24.0)                              | 447 (3080)                           | —                                     | —                   | 0.0444                | —                              | A,C,M,SC   |                                        |
| 62E      | 15Cr—2Mo—1Cu          | 58.8                  | 648 | 19.0 (20.9)                              | 444 (3060)                           | —                                     | —                   | 0.0454                | —                              | A,C,M,SC   |                                        |
| 62F      | 15Cr—2Mo—1Cu          | 61.2                  | 699 | 22.2 (24.2)                              | 471 (3250)                           | —                                     | 0.072               | 0.0474                | —                              | A,C,M,SC   |                                        |
| 62G      | 15Cr—2Mo—1Cu          | 57.3                  | 619 | 30.1 (33.1)                              | 448 (3090)                           | —                                     | —                   | 0.0399                | —                              | A,C,M,SC   |                                        |
| 62H      | 15Cr—2Mo—1Cu          | 54                    | 560 | 26.3 (28.9)                              | 399 (2750)                           | —                                     | —                   | 0.0538                | —                              | A,C,F      |                                        |
| 63B      | 17Cr—1.5Mo—1Cu (1.2C) | 52.7                  | 539 | 28.3 (31.1)                              | 504 (3470)                           | —                                     | —                   | 0.0568                | —                              | M,c,A,SC   |                                        |
| 64A      | 17Cr—1.5Mo—1Cu (1.8C) | 41.7                  | 410 | 26.5 (29.1)                              | 371 (2560)                           | —                                     | —                   | 0.0497                | —                              | A,C        |                                        |
| 64B      | 17Cr—1.5Mo—1Cu (1.8C) | 60.3                  | 680 | 25.7 (28.2)                              | 491 (3390)                           | —                                     | —                   | 0.0542                | —                              | M,C,A,SC   |                                        |
| 65A      | 17Cr—1.5Mo—1Cu (2.4C) | 46.8                  | 459 | 24.1 (26.5)                              | 330 (2280)                           | —                                     | —                   | 0.0470                | —                              | A,C,m      |                                        |

(Continued on next page.)

|                    |                       |      |     | Mechanical Properties                    |                                      |                                       | Abrasion Resistance |                          |                                   |                                        |
|--------------------|-----------------------|------|-----|------------------------------------------|--------------------------------------|---------------------------------------|---------------------|--------------------------|-----------------------------------|----------------------------------------|
| Code No.           | Irons                 | HRC  | HB  | Fracture Toughness<br>ksi√in.<br>(Mpa√m) | Compressive Strength<br>ksi<br>(MPa) | Bend Tensile Strength<br>ksi<br>(MPa) | Gouging Wear Ratio  | Pin Test<br>(grams lost) | Rubber Wheel Test<br>(grams lost) | Principal Components of Microstructure |
| 65B                | 17Cr—1.5Mo—1Cu (2.4C) | 61.5 | 706 | 25.6 (28.1)                              | 477 (3290)                           | —                                     | —                   | 0.0497                   | —                                 | M,C,A,SC                               |
| 66A                | 17Cr—1.5Mo—1Cu (3.0C) | 50.4 | 505 | 21.6 (23.7)                              | 329 (2270)                           | —                                     | —                   | 0.0445                   | —                                 | A,C,m                                  |
| 66B                | 17Cr—1.5Mo—1Cu (3.0C) | 62.7 | 733 | 20.6 (22.6)                              | 474 (3270)                           | —                                     | —                   | 0.0443                   | —                                 | M,C,A,SC                               |
| 67A                | 17Cr—1.5Mo—1Cu (3.6C) | 52   | 528 | 16.0 (17.6)                              | 345 (2380)                           | —                                     | —                   | 0.0417                   | —                                 | A,C,m                                  |
| 67B                | 17Cr—1.5Mo—1Cu (3.6C) | 60.1 | 675 | 19.3 (21.2)                              | 458 (3160)                           | —                                     | —                   | 0.0405                   | —                                 | M,C,A,SC                               |
| 68                 | 18Cr—1Cu              | 54.9 | 575 | 20.7 (22.8)                              | —                                    | —                                     | —                   | 0.0771                   | 0.1644                            | P,C,a                                  |
| 69                 | 18Cr—1Cu              | 35.9 | 377 | 24.9 (27.4)                              | —                                    | —                                     | —                   | 0.0939                   | 0.1966                            | S,C                                    |
| 70                 | 18Cr—1Mo—1Cu          | 48.8 | 483 | 28.1 (30.9)                              | —                                    | —                                     | —                   | 0.0434                   | 0.0773                            | A,C,m                                  |
| 71                 | 18Cr—1Mo—1Cu          | 63   | 740 | 22.9 (25.2)                              | —                                    | —                                     | —                   | 0.0508                   | 0.0605                            | M,C,SC,a                               |
| 72                 | 18Cr—2Mo—1Cu          | 49.4 | 491 | 27.4 (30.1)                              | —                                    | —                                     | —                   | 0.0421                   | —                                 | A,C,m                                  |
| 73                 | 18Cr—2Mo—1Cu          | 62.9 | 737 | 23.5 (25.8)                              | —                                    | —                                     | —                   | 0.0475                   | 0.0538                            | M,C,SC,a                               |
| 74A                | 18Cr—2Mo—1Cu (0.4Si)  | 50.5 | 506 | 23.8 (26.2)                              | —                                    | —                                     | —                   | 0.0428                   | —                                 | A,C,m                                  |
| 74B                | 18Cr—2Mo—1Cu (0.4Si)  | 63.7 | 756 | 24.5 (27.0)                              | —                                    | —                                     | —                   | 0.0486                   | —                                 | M,C,SC,a                               |
| 75A                | 18Cr—2Mo—1Cu (0.9Si)  | 50.7 | 509 | 29.0 (31.9)                              | —                                    | —                                     | —                   | 0.0435                   | —                                 | A,C,m                                  |
| 75B                | 18Cr—2Mo—1Cu (0.9Si)  | 62.9 | 737 | 26.3 (28.9)                              | —                                    | —                                     | —                   | 0.0512                   | —                                 | M,C,SC,a,s                             |
| 76A                | 18Cr—2Mo—1Cu (1.2Si)  | 51.8 | 525 | 31.0 (34.1)                              | —                                    | —                                     | —                   | 0.0431                   | —                                 | A,C,m,p                                |
| 76B                | 18Cr—2Mo—1Cu (1.2Si)  | 62.4 | 726 | 26.8 (34.1)                              | —                                    | —                                     | —                   | 0.0508                   | —                                 | M,C,SC,s                               |
| 77A                | 18Cr—2Mo—1Cu (1.6Si)  | 53.3 | 548 | 31.4 (34.5)                              | —                                    | —                                     | —                   | 0.0438                   | —                                 | A,C,m,p                                |
| 77B                | 18Cr—2Mo—1Cu (1.6Si)  | 55.4 | 584 | 29.1 (32.0)                              | —                                    | —                                     | —                   | 0.0596                   | —                                 | M,C,SC,s                               |
| 78A                | 18Cr—2Mo—1Cu          | 50.9 | 512 | 22.0 (24.2)                              | —                                    | —                                     | —                   | 0.0415                   | —                                 | A,C,m                                  |
| 78A <sup>(2)</sup> | 18Cr—2Mo—1Cu          | 53   | 544 | 23.2 (25.5)                              | —                                    | —                                     | —                   | 0.0526                   | —                                 | A,C,m                                  |
| 78B                | 18Cr—2Mo—1Cu          | 64.5 | 775 | 18.3 (20.1)                              | —                                    | —                                     | —                   | 0.0479                   | —                                 | M,C,SC,a                               |
| 78C                | 18Cr—2Mo—1Cu          | 64   | 763 | 20.0 (22.0)                              | —                                    | —                                     | —                   | 0.0494                   | —                                 | M,C,SC,a                               |
| 78D                | 18Cr—2Mo—1Cu          | 62.9 | 737 | 25.3 (27.8)                              | —                                    | —                                     | —                   | 0.0465                   | —                                 | M,C,SC,a                               |
| 78E <sup>(2)</sup> | 18Cr—2Mo—1Cu          | 59.9 | 671 | 29.2 (32.1)                              | —                                    | —                                     | —                   | 0.0496                   | —                                 | A,C,m,SC                               |
| 78F                | 18Cr—2Mo—1Cu          | 63   | 740 | 22.2 (24.2)                              | —                                    | —                                     | —                   | —                        | —                                 | M,C,SC,a                               |
| 78F <sup>(2)</sup> | 18Cr—2Mo—1Cu          | 63   | 740 | 21.4 (23.5)                              | —                                    | —                                     | —                   | 0.0596                   | —                                 | M,C,SC,a                               |
| 78G                | 18Cr—2Mo—1Cu          | 62.8 | 735 | 24.2 (26.6)                              | —                                    | —                                     | —                   | 0.0484                   | —                                 | M,C,SC,a                               |
| 78H                | 18Cr—2Mo—1Cu          | 63.6 | 754 | 20.0 (22.0)                              | —                                    | —                                     | —                   | 0.0425                   | —                                 | M,C,SC,a                               |
| 78I                | 18Cr—2Mo—1Cu          | 61.5 | 706 | 29.0 (31.9)                              | —                                    | —                                     | —                   | 0.0425                   | —                                 | M,C,SC,a                               |
| 79                 | 18Cr—3Mo—1Cu          | 50.4 | 505 | 23.0 (25.3)                              | —                                    | —                                     | —                   | 0.0414                   | 0.0702                            | A,c,m,sc                               |

(Continued on next page.)

(Appendix C, General Mechanical & Wear Properties, Diesburg & Borik, Table C3, continued)

| Code No. | Irons        | HRC  | HB  | Mechanical Properties                    |                                      |                                       |  | Abrasion Resistance |                       |                                |                                        |
|----------|--------------|------|-----|------------------------------------------|--------------------------------------|---------------------------------------|--|---------------------|-----------------------|--------------------------------|----------------------------------------|
|          |              |      |     | Fracture Toughness<br>ksi√in.<br>(MPa√m) | Compressive Strength<br>ksi<br>(MPa) | Bend Tensile Strength<br>ksi<br>(MPa) |  | Gouging Wear Ratio  | Pin Test (grams lost) | Rubber Wheel Test (grams lost) | Principal Components of Microstructure |
| 80       | 18Cr—3Mo—1Cu | 63   | 740 | 23.2 (25.5)                              | —                                    | —                                     |  | —                   | 0.0486                | 0.0541                         | M,C,SC,a                               |
| 81A      | 18Cr—2Mo—1Cu | 50.3 | 503 | 28.5 (31.4)                              | 286 (1970)                           | —                                     |  | —                   | —                     | —                              | A,C,sc                                 |
| 81B      | 18Cr—2Mo—1Cu | 63.3 | 747 | 27.5 (30.2)                              | 416 (2870)                           | 100 (689)                             |  | 0.081               | —                     | 0.051                          | M,C,SC,a                               |
| 81C      | 18Cr—2Mo—1Cu | 62.9 | 737 | 25.6 (30.2)                              | —                                    | —                                     |  | —                   | —                     | —                              | M,C,SC,a                               |
| 82       | 27Cr         | 49.7 | 495 | 27.7 (30.5)                              | 338 (2330)                           | —                                     |  | 0.133               | —                     | —                              | A,C,sc                                 |
| 83A      | 27Cr         | 50.4 | 505 | —                                        | 400 (2760)                           | 120 (827)                             |  | —                   | —                     | —                              | A,C,sc                                 |
| 83B      | 27Cr         | 59.3 | 659 | —                                        | 419 (2890)                           | 137 (945)                             |  | 0.127               | —                     | 0.099                          | M,C,A,SC                               |
| 84A      | 27Cr         | 55.5 | 586 | —                                        | 392 (2700)                           | 106 (731)                             |  | —                   | —                     | —                              | A,C,sc                                 |
| 84B      | 27Cr         | 61.4 | 703 | —                                        | 429 (2960)                           | 136 (938)                             |  | 0.099               | —                     | 0.066                          | M,C,A,SC                               |

|     | Steels                                   | HRC | HB  | Charpy V—Notch Impact Energy |       | Abrasion Resistance |                       |                                |                                         |
|-----|------------------------------------------|-----|-----|------------------------------|-------|---------------------|-----------------------|--------------------------------|-----------------------------------------|
|     |                                          |     |     | ft-lb                        | Joule | Gouging Wear Ratio  | Pin Test (grams lost) | Rubber Wheel Test (grams lost) | Principal Components of Microstructure* |
| 7A  | T-1 Wrought Steel                        | —   | 260 | 42                           | 57    | 1.08                | 0.1353                | 0.845                          | M,B                                     |
| 21A | Cr-Ni-Mo Wrought Steel                   | —   | 510 | —                            | —     | 0.36                | 0.0954                | 0.756                          | M                                       |
| 22A | AlSi 4340 (0.4C)                         | —   | 320 | 6.0                          | 8.1   | 0.67                | —                     | —                              | F,P                                     |
| 22B | AlSi 4340 (0.4C)                         | —   | 340 | 22.0                         | 29.8  | 0.72                | —                     | —                              | F,c                                     |
| 22C | AlSi 4340 (0.4C)                         | —   | 520 | 7.8                          | 10.6  | 0.23                | —                     | —                              | M                                       |
| 30  | Cr—Mo Cast Steel                         | —   | 343 | —                            | —     | 0.38                | 0.0942                | 0.460                          | P,c                                     |
| 31A | AlSi H-11 Tool Steel                     | —   | 658 | 28                           | 38    | 0.20                | 0.0726                | 0.710                          | M                                       |
| 41  | 12Mn (0.95C)<br>Cast Austenitic Mn Steel | —   | 185 | 102                          | 138   | 0.33                | —                     | —                              | A                                       |
| 44  | 12Mn (1.1C)<br>Cast Austenitic Mn Steel  | —   | 199 | —                            | —     | 0.28                | 0.0821                | —                              | A                                       |

Microstructure: A = austenite; M = martensite; S = spheroidal; F = acicular ferrite (bainite); C = primary carbide; SC = secondary carbide; P = pearlite. Lower case letters = small amounts.

(1) Diesburg's and Borik, 1974.

(2) Cast in graphite.

## Hardness and Abrasion Resistance

### Avery (1974)

H.S. Avery's work ("Work Hardening in Relation to Abrasion Resistance," in *Symposium Materials for the Mining Industry*, Vail, Colorado, pp. 43-78, 1974) represents another great compilation of hardness and abrasion data for AR steels and irons. Of particular note is his comparison of the hardness and abrasion resistance of castings having different section sizes (i.e., having fine- and coarse-grained structures). A portion of his work is summarized in Tables C4 and C5 and Fig. C2.

The abrasion test used by Avery was developed at Abex Corporation and is known as the wet sand abrasion test. The sample is

rotated against a copper track within a mixture of water and abrasive. The sample is pushed against the copper track under a nominal pressure of 54 psi (0.4 MPa) with a velocity of 119 ft/sec (0.60 m/sec). The sample is immersed in a slurry of 0.64/100 by weight or 0.24/100 by volume of sand to water. The test is run for 60 minutes and the sample weighed. The wet sand abrasion factor (WSAF) is then computed. A low value implies high abrasion resistance.

One of the interesting features of Avery's work is the comparison between thick- and thin-section castings. Tables C4 and C5 present part of the data compiled by Avery. In all cases, the best abrasion resistance is on the surface of the casting. This effect is most pronounced for heavy-section castings. Some alloys are more sensitive to this thickness effect than are others.

**Table C4**  
**Composition and Heat Treatment of Grinding Mill Alloys<sup>(1)</sup>**

| Item | Type                         | ASTM A-532 Grade | %C   | %Mn  | %Si  | %Cr  | %Mo  | %Ni  | Heat Treatment |
|------|------------------------------|------------------|------|------|------|------|------|------|----------------|
| 1    | Aust. 13 Mn                  | —                | 1.2  | 13.0 | 0.5  | 0.5  | —    | —    | WQ             |
| 2    | Pearlitic Cr-Mo              | —                | 0.83 | 0.9  | 0.25 | 2.33 | 0.42 | —    | AH & T         |
| 3    | Pearlitic Cr-Mo              | —                | 0.90 | 1.0  | 0.55 | 2.25 | 0.37 | —    | AH & T         |
| 4    | Mart. Cr-Mo                  | —                | 0.65 | 1.0  | 0.5  | 2.3  | 0.4  | —    | AH & T         |
| 5    | Ni-Hard                      | IA               | 3.18 | 0.58 | 0.51 | 2.32 | —    | 3.93 | SR             |
| 6    | Ni-Hard                      | IA               | 3.02 | 0.90 | 0.43 | 3.68 | —    | 4.02 | SR             |
| 7    | Ni-Hard                      | IA               | 3.00 | 0.73 | 0.77 | 2.25 | —    | 4.70 | SR             |
| 8    | HC-250                       | IIIA             | 2.91 | 1.12 | 0.62 | 25.7 | 0.49 | 0.25 | As Cast        |
| 9    | HC-250                       | IIIA             | 2.91 | 1.12 | 0.62 | 25.7 | 0.49 | 0.25 | AH2            |
| 10   | 15-3                         | IIC              | 3.17 | 0.94 | 0.90 | 16.0 | 3.17 | —    | As Cast        |
| 11   | 15-3                         | IIC              | 3.17 | 0.94 | 0.90 | 16.0 | 3.17 | —    | AH & SR        |
| 12   | Parabology                   | IIE              | 2.95 | 0.73 | 0.73 | 19.1 | 1.14 | 0.77 | AH2            |
| 13   | Parabology                   | IIE              | 3.15 | 0.76 | 0.65 | 19.7 | 1.25 | 0.71 | AH2            |
| 14   | 19Cr-Ni-Mo<br>(underalloyed) | IID              | 2.47 | 0.75 | 0.62 | 19.8 | 0.73 | 0.73 | AH2            |
| 15   | 19Cr-Ni-Mo<br>(overalloyed)  | IIE              | 2.83 | 0.71 | —    | 19.5 | 1.24 | 1.24 | AH2            |
| 16   | Parabology                   | IIE              | 2.70 | 0.68 | 0.54 | 18.1 | 1.22 | 0.65 | AH2            |
| 17   | Parabology                   | IIE              | 2.89 | 0.72 | 0.50 | 17.9 | 1.14 | 0.80 | AH2            |
| 18   | Parabology                   | IIE              | 2.89 | 0.72 | 0.50 | 17.9 | 1.14 | 0.80 | AH2            |

*Heat treatment: WQ = water quenched from about 1850F (1010C); AH = air hardened; 2 = from 2000F (1095C); SR = stress relieved at 400–520F (205–270C) - 4 to 10 hr.; T = tempered.*

(1) Avery, 1974.

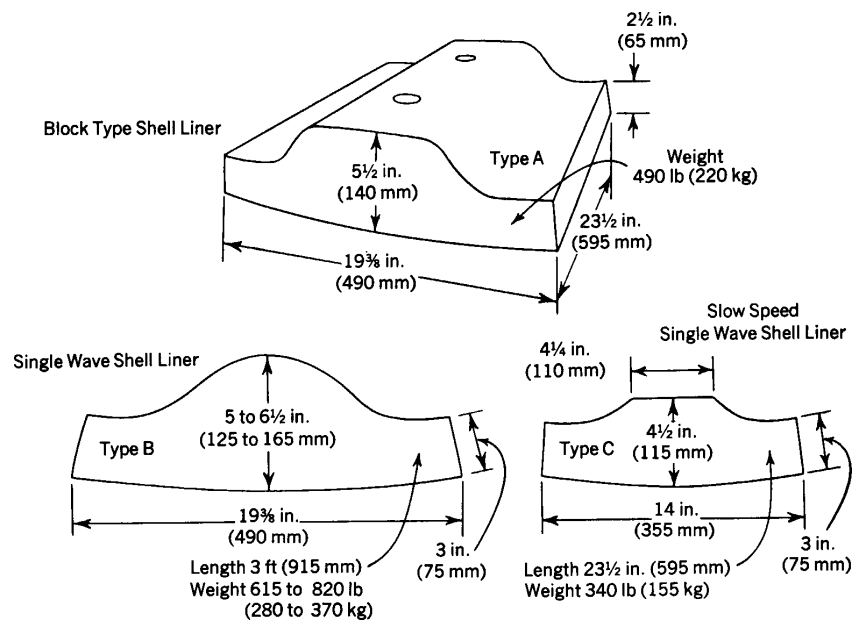
**Table C5**  
**Abrasion Resistance of Thin and Thick Sections of Grinding Mill Liner Castings<sup>(1)</sup>**

| Item | Type                         | Liner Type <sup>(2)</sup> | Thin Section                       |     |      |          |     |      | Thick Section                            |     |      |                    |     |      |
|------|------------------------------|---------------------------|------------------------------------|-----|------|----------|-----|------|------------------------------------------|-----|------|--------------------|-----|------|
|      |                              |                           | 2-1/2 to 3 in. (65 to 75 mm) Thick |     |      |          |     |      | 4-1/2 to 6-1/2 in. (115 to 165 mm) Thick |     |      |                    |     |      |
|      |                              |                           | Surface                            |     |      | Interior |     |      | Surface                                  |     |      | Interior           |     |      |
|      |                              |                           | HRC                                | HB  | WSAF | HRC      | HB  | WSAF | HRC                                      | HB  | WSAF | HRC                | HB  | WSAF |
| 1    | Aust. 13 Mn                  | —                         | —                                  | —   | —    | —        | —   | —    | —                                        | —   | —    | Expectancy is 0.80 |     |      |
| 2    | Pearlitic Cr-Mo              | —                         | —                                  | —   | —    | —        | —   | —    | —                                        | —   | —    | 36                 | 345 | 0.64 |
| 3    | Pearlitic Cr-Mo              | —                         | —                                  | —   | —    | —        | —   | —    | —                                        | —   | —    | 36                 | 370 | 0.62 |
| 4    | Mart. Cr-Mo                  | —                         | —                                  | —   | —    | —        | —   | —    | —                                        | —   | —    | Expectancy is 0.71 |     |      |
| 5    | Ni-Hard                      | B                         | 62                                 | 667 | 0.52 | 59       | 600 | 0.59 | 62                                       | 667 | 0.51 | 59                 | 600 | 0.56 |
| 6    | Ni-Hard                      | B                         | 52                                 | 447 | 0.58 | 51       | 447 | 0.63 | 52                                       | 447 | 0.63 | 49                 | 460 | 0.69 |
| 7    | Ni-Hard                      | B                         | 59                                 | 613 | 0.60 | 56       | 555 | 0.69 | 59                                       | 613 | 0.59 | 56                 | 555 | 0.67 |
| 8    | HC-250                       | A                         | 57                                 | 555 | 0.46 | 57       | 555 | 0.49 | 57                                       | 555 | 0.52 | 50                 | 477 | 0.80 |
| 9    | HC-250                       | A                         | 63                                 | 633 | 0.46 | 62       | 667 | 0.49 | 62                                       | 683 | 0.45 | 62                 | 675 | 0.64 |
| 10   | 15-3                         | A                         | 48                                 | 460 | 0.81 | 47       | 444 | 0.86 | 48                                       | 460 | 0.81 | 46                 | 444 | 0.87 |
| 11   | 15-3                         | A                         | —                                  | —   | —    | —        | —   | —    | —                                        | —   | —    | 60                 | 652 | 0.77 |
| 12   | Paraboloy                    | A                         | 60                                 | 600 | 0.36 | 58       | 555 | 0.39 | 63                                       | 652 | 0.36 | 61                 | 627 | 0.42 |
| 13   | Paraboloy                    | A                         | 63                                 | 652 | 0.35 | 63       | 627 | 0.38 | 63                                       | 667 | 0.33 | 63                 | 627 | 0.49 |
| 14   | 19Cr-Ni-Mo<br>(underalloyed) | A                         | 63                                 | 652 | 0.45 | 63       | 652 | 0.44 | 63                                       | 652 | 0.58 | 62                 | 627 | 0.60 |
| 15   | 19Cr-Ni-Mo<br>(overalloyed)  | B                         | 65                                 | 685 | 0.35 | 64       | 652 | 0.43 | 64                                       | 683 | 0.37 | 64                 | 652 | 0.53 |
| 16   | Paraboloy                    | C                         | 65                                 | 713 | 0.34 | 64       | 713 | 0.37 | 65                                       | 713 | 0.37 | 64                 | 698 | 0.49 |
| 17   | Paraboloy                    | B                         | 61                                 | 600 | 0.42 | 61       | 600 | 0.46 | 60                                       | 578 | 0.41 | 61                 | 627 | 0.45 |
| 18   | Paraboloy                    | A                         | 61                                 | 627 | 0.39 | 61       | 627 | 0.41 | 62                                       | 627 | 0.40 | 62                 | 627 | 0.46 |

HRC = Rockwell, C-scale hardness; HB = Brinell Hardness; WSAF = Wet Sand Abrasion Factor

(1) Avery, 1974.

(2) See Fig. C2, below.



**Fig. C2. Schematic of Type A, Type B and Type C shell liners studied.**  
 Results are contained in Table C5, above.

Abrasion Resistance and Fracture Toughness of White Cast Irons

Sare (1979)

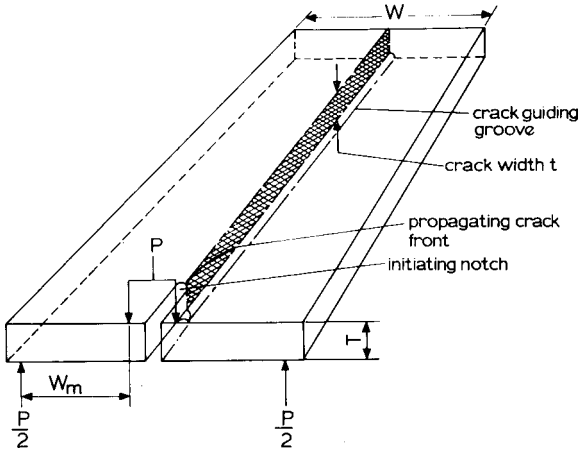
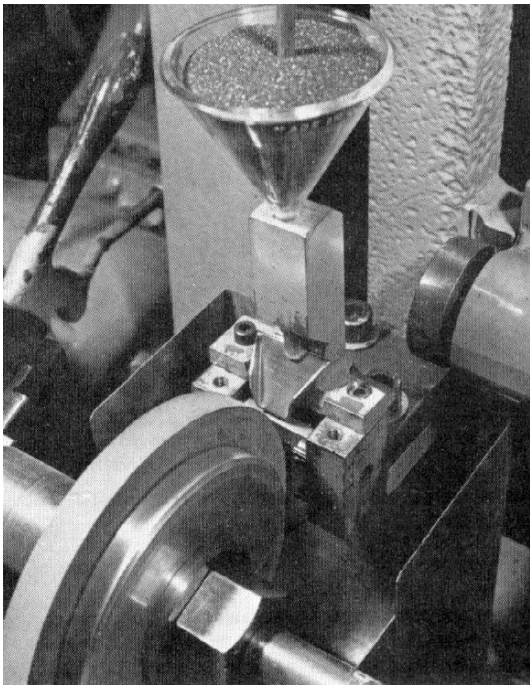
In “Abrasion Resistance and Fracture Toughness of White Cast Irons, *Met. Technol.*, vol. 6, pp. 412-419, 1979, I.R. Sare provides

an in-depth look at both abrasion resistance and fracture toughness in Tables C6, C7, C8 and Fig. C3. His represents an important piece of work for its completeness in covering a wide range of irons and a few common AR steels.

Table C6  
Alloy Composition (Wt%) and Abrasion Test Data<sup>(1)</sup>

| Alloy            | %C   | %Si  | %Mn  | %Cr  | %Mo  | %Ni  | %Cu  | Test Data                                            |
|------------------|------|------|------|------|------|------|------|------------------------------------------------------|
| Mild Steel       | 0.24 | 0.01 | 0.53 | 0.07 | 0.01 | 0.08 | —    | Reference material for abrasion tests                |
| 15Cr-3Mo A       | 3.4  | 0.30 | 0.28 | 15   | 2.85 | 0.05 | 0.01 | Abrasion test and K <sub>IC</sub> ; cast 5 mm thick  |
| 15Cr-3Mo B       | 3.5  | 0.65 | 0.76 | 14.9 | 2.76 | 0.19 | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| 15Cr-3Mo C       | 3.6  | 0.70 | 0.83 | 14.8 | 2.55 | 0.27 | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| 15Cr-3Mo D       | 3.8  | 0.94 | 0.59 | 15.2 | 2.90 | 0.13 | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| 15Cr-3Mo E       | 3.5  | 0.56 | 0.86 | 16.1 | 2.60 | 0.45 | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| 15Cr-3Mo F       | 3.6  | 0.75 | 0.63 | 15.3 | 2.50 | 0.19 | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| 15Cr-3Mo G       | 3.7  | 0.79 | 0.73 | 15.3 | 2.50 | 0.20 | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| 15Cr-3Mo H       | 3.54 | 0.99 | 0.60 | 17.2 | 3.0  | —    | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| 15Cr-2Mo-1Cu     | 2.75 | 0.35 | 0.64 | 16.6 | 1.83 | —    | 0.95 | K <sub>IC</sub> ; cast 12 mm thick                   |
| 27% Cr A         | 2.55 | 0.50 | 0.26 | 26.3 | 0.55 | 0.10 | 0.02 | Abrasion test and K <sub>IC</sub> ; cast 12 mm thick |
| 27% Cr B         | 2.95 | 0.74 | 0.86 | 28.8 | —    | —    | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| 27% Cr C         | 2.75 | 0.65 | 0.70 | 27.0 | —    | 0.2  | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| 35% Cr A         | 5.24 | 1.15 | 3.30 | 35.5 | 0.01 | 0.15 | 0.05 | Abrasion test and K <sub>IC</sub> ; cast 12 mm thick |
| 35% Cr B         | 4.42 | 0.35 | 0.40 | 33.2 | 0.01 | 0.08 | —    | Abrasion test; cast 12 mm thick                      |
| Ni-Cr 1          | 3.71 | 0.84 | 0.55 | 2.10 | 0.01 | 4.07 | —    | Abrasion test and K <sub>IC</sub> ; cast 12 mm thick |
| Ni-Cr 4          | 3.43 | 1.75 | 0.65 | 9.60 | 0.06 | 5.50 | —    | Abrasion test and K <sub>IC</sub> ; cast 12 mm thick |
| Ni-Cr-Mo steel A | 0.32 | 1.28 | 1.08 | 1.53 | 0.42 | 0.39 | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| Ni-Cr-Mo steel B | 0.33 | 1.52 | 0.72 | 1.49 | 0.46 | 0.24 | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| Ni-Cr-Mo steel C | 0.35 | 1.79 | 0.86 | 1.49 | 0.43 | 0.28 | —    | K <sub>IC</sub> ; cast 12 mm thick                   |
| Cr-Mo steel      | 0.50 | 0.58 | 1.02 | 0.99 | 0.39 | 0.11 | —    | K <sub>IC</sub> ; cast 12 mm thick                   |

(1) Sare, 1979.



↑ Double-Torsion Bend Test: Samples were cast to size, surfaces ground flat, and the crack-guiding groove and initiating notch were machined.

⇐ Rubber Wheel Abrasive Wear Test: All tests were conducted using –100, +150 mesh SiC grit. Results are reported as the weight loss of the mild steel standard divided by the weight loss of the wear test specimen.

Fig. C3. Description of abrasion and fracture tests (Sare, 1979).

**Table C7**  
**Heat Treatments<sup>(1)</sup>**

| Alloy                                                      | Heat <sup>(2)</sup><br>Treatment<br>Designation | Heat Treatment                                   |
|------------------------------------------------------------|-------------------------------------------------|--------------------------------------------------|
| 15Cr-3Mo A                                                 | a                                               | As cast                                          |
|                                                            | b                                               | 40 min, 1832F (1000C); AC                        |
|                                                            | c                                               | 40 min, 1832F (1000C); 30 min, 1247F (675C); AC  |
|                                                            | d                                               | 40 min, 1832F (1000C); 100 hr, 536F (280C); AC   |
| 15Cr-3Mo B-H                                               | b                                               | ~3 hr, 1832F (1000C); AC                         |
| 20Cr-2Mo-1Cu                                               | b                                               | ~3 hr, 1832F (1000C); AC                         |
| 27%Cr A                                                    | a                                               | As cast                                          |
|                                                            | b                                               | 40 min, 1832F (1000C); AC                        |
|                                                            | c                                               | 40 min, 1832F (1000C); 100 min, 1274F (690C); AC |
|                                                            | d                                               | 40 min, 1832F (1000C); 100 hr, 662F (350C); AC   |
| 27%Cr B                                                    | b                                               | 3 hr, 1920F (1050C); AC                          |
| 27% Cr C                                                   | b                                               | 3 hr, 1920F (1050C); AC                          |
| 35% Cr A                                                   | a                                               | As cast                                          |
|                                                            | b                                               | 40 min, 1832F (1000C); AC                        |
|                                                            | c                                               | 40 min, 1832F (1000C); 2 min, 1240F (670C); AC   |
|                                                            | d                                               | 60 min, 1832F (1000C); 100 hr, 535F (280C); AC   |
| NiCr 1                                                     | a                                               | As cast                                          |
|                                                            | b                                               | 4 hr, 527F (275C); AC                            |
|                                                            | c                                               | 4 hr, 842F (450C); AC; 4 hr 527F (275C); AC      |
| NiCr 4                                                     | a                                               | As cast                                          |
|                                                            | b                                               | 4 hr, 1472F (800C); AC                           |
|                                                            | c                                               | 4 hr, 1472F (800C); AC; 4 hr 527F (275C); AC     |
| Ni-Cr-Mo steel A                                           | b                                               | 1724F (940C); furnace cooled;                    |
| Ni-Cr-Mo steel B                                           | b                                               | 1796F (980C); AC; 6 hr, 608F (320C); AC          |
| Ni-Cr-Mo steel C                                           | b                                               | 1724F (940C); OQ; 6 hr, 608F (320C); AC          |
| Cr-Mo steel                                                | b                                               | 1742F (950C); AC; 1 hr, 752F (400C); AC          |
| <i>Heat treatment: AC = air cooled; OQ = oil quenched.</i> |                                                 |                                                  |
| <i>(1) Sare, 1979.</i>                                     |                                                 |                                                  |
| <i>(2) Shows relationship to the alloys in Table C8.</i>   |                                                 |                                                  |

**Table C8**  
**Abrasion Resistance and Fracture Toughness Test Results<sup>(1)</sup>**

| Alloy                  | Heat<br>Treat-<br>ment | Hard-<br>ness<br>HV50 | Abrasion<br>Resistance<br>Ratio | Fracture<br>Toughness<br>K <sub>IC</sub> , Mpa·in. <sup>1/2</sup> |
|------------------------|------------------------|-----------------------|---------------------------------|-------------------------------------------------------------------|
| 15Cr-3Mo A             | a                      | 557                   | 1.75±0.04                       | 29.4                                                              |
|                        | b                      | 829                   | 1.73±0.04                       | 22.2                                                              |
|                        | c                      | 809                   | 1.67±0.04                       | 21                                                                |
|                        | d                      | 735                   | 1.79±0.04                       | 21                                                                |
| 15Cr-3Mo B             | b                      | 890                   | —                               | 19.5                                                              |
| 15Cr-3Mo C             | b                      | 937                   | —                               | 20.8                                                              |
| 15Cr-3Mo D             | b                      | 841                   | —                               | 25.5                                                              |
| 15Cr-3Mo E             | b                      | 859                   | —                               | 25.5                                                              |
| 15Cr-3Mo F             | b                      | 925                   | —                               | 23.8                                                              |
| 15Cr-3Mo G             | b                      | 929                   | —                               | 21.3                                                              |
| 15Cr-3Mo H             | b                      | 955                   | —                               | 21.4                                                              |
| 20Cr-2Mo-1Cu           | b                      | 824                   | —                               | 24.5                                                              |
| 27%Cr A                | a                      | 545                   | 2.06±0.06                       | 30.3                                                              |
|                        | b                      | 868                   | 1.92±0.04                       | 22.6                                                              |
|                        | c                      | 459                   | 1.98±0.06                       | 27.4                                                              |
|                        | d                      | 840                   | 1.90±0.06                       | 22.7                                                              |
| 27%Cr B                | b                      | 697                   | —                               | 30.1                                                              |
| 27%Cr C                | b                      | 753                   | —                               | 30.6                                                              |
| 35%Cr A                | a                      | 729                   | 2.14±0.06                       | 27.1                                                              |
|                        | b                      | 884                   | 2.21±0.06                       | 21.4                                                              |
|                        | c                      | 885                   | 2.21±0.06                       | 22.2                                                              |
|                        | d                      | 870                   | 2.29±0.06                       | 20.4                                                              |
| NiCr 1                 | a                      | 719                   | 1.32±0.06                       | 25.6                                                              |
|                        | b                      | 691                   | 1.38±0.06                       | 26.2                                                              |
|                        | c                      | 788                   | 1.36±0.06                       | 19.1                                                              |
| NiCr 4                 | a                      | 604                   | 1.44±0.06                       | 30.6                                                              |
|                        | b                      | 820                   | 1.36±0.06                       | 26.6                                                              |
|                        | c                      | 704                   | 1.44±0.06                       | 26.2                                                              |
| Ni-Cr-Mo Steel A       | b                      | 540                   | —                               | 88.5                                                              |
| Ni-Cr-Mo Steel B       | b                      | 493                   | —                               | 70.1                                                              |
| Ni-Cr-Mo Steel C       | b                      | 548                   | —                               | 83.6                                                              |
| Cr-Mo steel            | b                      | 512                   | —                               | 50.9                                                              |
| <i>(1) Sare, 1979.</i> |                        |                       |                                 |                                                                   |

## Appendix D

### Hardness Measurements

#### Background Information

Hardness measurements are perhaps the single most prevalent and important metallurgical measurement made in today's foundry. Typically, a near homogeneous material is deformed by a much harder indenter, and the amount of plastic deformation in the sample is measured. This simple concept implies many assumptions about the material's behavior as it is plastically deformed (i.e., an indent is left in the sample after the hardness measurement). For example, the amount of plastic deformation produced by a given load is a direct function of the material's elastic modulus, yield strength and work hardening characteristics. Besides these interactions, a key tenet of all hardness measurements is that the material region "sampled" by the indenter is representative of the material in general. For macroscopically and microscopically heterogeneous AR cast irons, this criterion must be carefully reviewed for each type of hardness measurement technique employed.

#### Measurement Technique

Figure D1 shows a schematic of the three most common hardness measurement techniques: Brinell, Rockwell and Equotip. All three use a very hard indenter (tungsten carbide for the Equotip  $L_D$  and Brinell, and diamond for Equotip  $L_E$  and Rockwell), which is assumed to remain rigid during the indentation test.

Of the three hardness techniques, the Brinell and Rockwell tests both make direct measurements of the residual plastic deformation produced by their indenters. The Brinell test measures the diameter of indentation crater, while the Rockwell test measures the depth of the indentation crater. This is an important distinction. The depth of the indentation crater is directly affected by the elastic response of the undeformed material surrounding the plastically deformed region (elastic recovery), while the diameter of the crater remains relatively unaffected. Consequently, converting Brinell hardness measurements to Rockwell hardness measurements or vice versa is not recommended between materials having different elastic moduli (i.e., between aluminum and iron).

The Equotip measurement technique is a relatively new technique invented by D. Leeb in the late 1970s. The name Equotip is an abbreviation for **E**nergy-**q**uotient (Equo) and "tip" is derived from

the spherical projectile used in the indentation process. The reported hardness value is derived from the impact ( $V_{\text{impact}}$ ) and rebound ( $V_{\text{rebound}}$ ) velocities of a spherical indenter as:

$$L = 1000 \cdot V_{\text{rebound}} / V_{\text{impact}}$$

where L is in honor of the inventor D. Leeb.

The instrument automatically measures the two velocities and calculates the L number. Reported L numbers are always averages of five impact events. Depending upon the type of spherical indenter, whether tungsten carbide or diamond, the L number is denoted as  $L_D$  or  $L_E$ , respectively. The L value of this dynamic measurement is a function of the elastic modulus, yield strength and work-hardening characteristics of the material.

Since the rebound velocity of the indenter is a critical factor in calculating the L number, the indented material's elastic modulus will have a strong influence on the resulting number. For example, specimens of steel (210 GPa) and aluminum (70 GPa) were metallurgically designed to have Brinell hardnesses of 150 (i.e., similar yield strengths and work-hardening characteristics). Upon testing with the Equotip instrument, their  $L_D$  numbers varied from 414 for the steel specimen to 542 for the aluminum specimen—a variation of 30% (calculated as:  $128 / 414 \cdot 100$ ) (Laird et al., 1995). This behavior is simply due to the effect of the aluminum storing more elastic energy during impact than steel. If the yield strengths are equivalent, then:

$$\text{Energy}_{\text{steel}} = 70 \text{ GPa} / 210 \text{ GPa} \cdot \text{Energy}_{\text{aluminum}}$$

The steel specimen only stores about 1/3 the elastic energy of the aluminum specimen, and therefore produces a lower rebound velocity and a lower L number.

An important concern to the producers of AR cast irons is the precision and accuracy of the hardness measurement on heterogeneous microstructures. Both precision (the ability of the instrument to reproduce a certain reading) and accuracy (the deviation of the instrument reading from a known standard) are influenced by the region sampled during the indentation process. Since AR irons are used in macroscopic wear environments, where wear losses are measured in centimeters, an accurate hardness measurement must sample a macroscopic region of the casting.

A simple calculation shows that, at a hardness Brinell (HB) of 578, using a 3000 kgf static load and a 10-mm tungsten carbide ball,

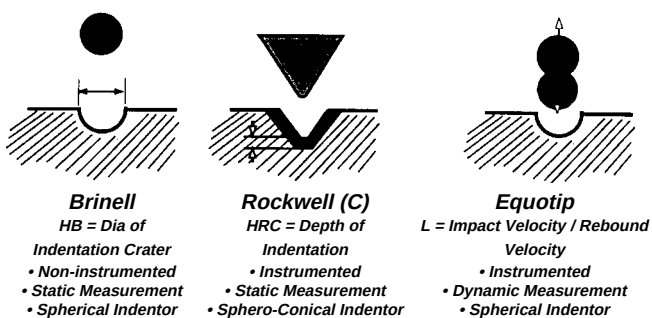


Fig. D1. Schematic of common hardness tests. In all cases, the hardness measurement is a measure of the resistance to plastic flow.

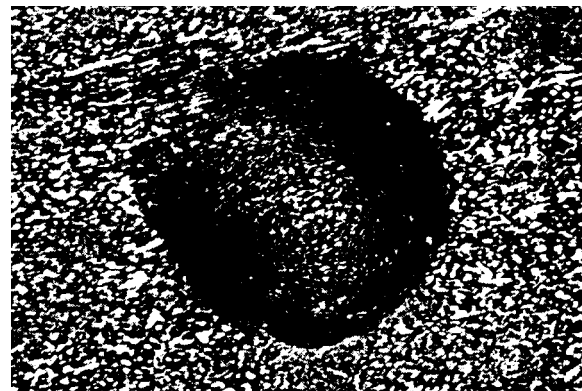


Fig. D2. Hypoeutectic, sand-cast, white cast iron showing an  $L_D$  indentation.

the indentation crater has a diameter of 2550  $\mu\text{m}$ . The size of this crater is far greater than that of the microstructural features found in white cast irons, where the carbides are from 10 to 50  $\mu\text{m}$  in diameter and the austenite grains are from 25 to 100  $\mu\text{m}$  in diameter.

In contrast, the Equotip measurement only samples a very small region of the sample. Figure D2 shows an Equotip indentation crater made in a high-Cr AR iron. The size of the indentation crater covers just a few eutectic colonies. As would be expected, for an accurate Equotip measurement, many readings must be made and statistically analyzed.

#### Conversion Guidelines

When working within any one hardness measurement system, questions arise as to how comparable the measurement errors to other hardness techniques are. In work done by the AFS 5-D committee, it was discovered that, under controlled conditions, measurement errors between the three common hardness techniques (Table D1) are virtually equivalent across a broad spectrum of AR cast irons. Nonetheless, when one converts hardness measurements from one system to another system, the measurement error has just doubled by using a conversion equation (Table D2).

A way to view this problem is shown in Fig. D3. This figure shows the correlation equation used to fit HRC data to HB data; it is not a perfect fit to the experimental data.

Another hidden problem with converting hardness numbers is in recognizing the limitations of each hardness technique. This is most apparent between HB and the Equotip. At hardnesses in excess of 700 HB, the HB measurement is still accurate (although it does not meet the ASTM specification, it still gives valid, accurate hardness numbers!), while the Equotip,  $L_D$  measurement gives meaningless numbers. This is clearly shown in Fig. D4. As the hardness of the AR iron increases, the  $L_D$  measurement does not effectively discriminate between irons of higher hardness.

With these caveats, conversion equations are introduced in Table D2. In all cases, the end user should have access to the original hardness measurement and the converted hardness number, and both should be reported. Tabulated conversions (Table D3) are also presented for convenience.

Since Vickers hardness measurements are prevalent, conversions in Table D3 are presented to convert between Brinell, Rockwell C and Vickers hardness values (Parks, 1979).

**Table D1**  
**Generalized Comparison of**  
**Errors in Hardness Measurements<sup>(1)</sup>**

| Technique | Example Measurement Value | Typical Measurement Error | Conversion Approximations for Typical Measurement Error in Other Hardness Units |           |            |
|-----------|---------------------------|---------------------------|---------------------------------------------------------------------------------|-----------|------------|
|           |                           |                           | $\pm L_D$                                                                       | $\pm HB$  | $\pm HRC$  |
| $L_D$     | 750                       | $\pm 6.6$                 | $\pm 6.6$                                                                       | $\pm 10$  | $\pm 0.66$ |
| HB        | 591                       | $\pm 9.6$                 | $\pm 6.0$                                                                       | $\pm 9.6$ | $\pm 0.60$ |
| HRC       | 55.7                      | $\pm 0.62$                | $\pm 5.9$                                                                       | $\pm 11$  | $\pm 0.62$ |

(1) For more details, see Laird, 1995.

**Table D2**  
**Hardness Conversion Equations**

| Equation                                                                     | Range         | $r^2/\text{Error}^{(1)}$ |
|------------------------------------------------------------------------------|---------------|--------------------------|
| <b>Equotip (<math>L_D</math>):</b>                                           |               |                          |
| $L_D^{(2)} = 380.6 + 0.626 \cdot HB$                                         | 340–700 HB    | $0.976 \pm 9.5 L_D$      |
| $L_D^{(2)} = 489.8 + 0.0848 \cdot HRC^{(3)}$                                 | 34.9–61.1 HRC | $0.967 \pm 11.0 L_D$     |
| <b>Brinell (10/3000):</b>                                                    |               |                          |
| $HB^{(2)} = -579.1 + 1.56 \cdot L_D$                                         | 593–819 $L_D$ | $0.976 \pm 15 HB$        |
| $HB^{(3)} = 717.8 - 22.515 \cdot HRC + 0.363 \cdot HRC^{(3)}$                | 38.0–66.8 HRC | $0.992 \pm 12.5 HB$      |
| <b>Rockwell (C):</b>                                                         |               |                          |
| $HRC^{(2)} = -129.7 + 0.395 \cdot L_D - 1.97 \times 10^{-4} \cdot L_D^{(3)}$ | 593–819 $L_D$ | $0.975 \pm 1.2 HRC$      |
| $HRC^{(3)} = -12.0 + 0.167 \cdot HB - 8.82 \times 10^{-5} \cdot HB^{(3)}$    | 385–848 HB    | $0.992 \pm 0.56 HRC$     |

(1) Squared multiple correlation coefficient / standard error of the estimate (i.e., the standard deviation of the actual data against the data generated from the regression curve).

(2) Equations from Laird et al. (1995).

(3) Equations from Parks (1979).

(Continued on next page.)

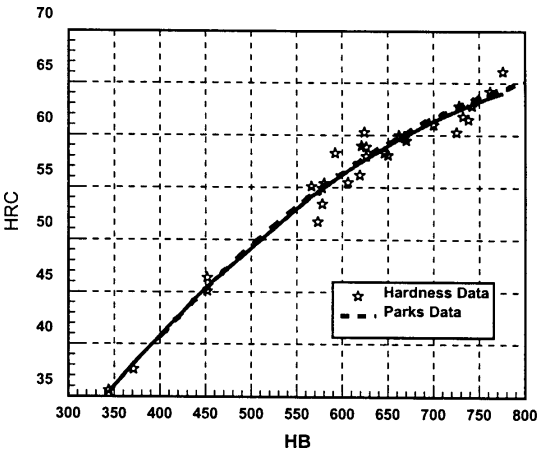


Fig. D3. Comparison of curve-fitting errors between HB and HRC data.

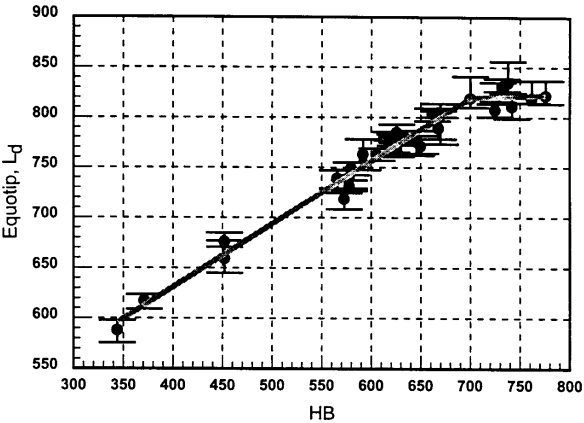


Fig. D4. Comparison of curve-fitting errors between HB and  $L_D$  data.

Table D3  
Approximate Hardness Conversion Data for AR White Cast Irons

| Equotip $L_D$ to Brinell and Rockwell C Hardness Numbers: |               |                 |                    |               |                 |
|-----------------------------------------------------------|---------------|-----------------|--------------------|---------------|-----------------|
| Equo-<br>tip $L_D$                                        | HB<br>10/3000 | Rock-<br>well C | Equo-<br>tip $L_D$ | HB<br>10/3000 | Rock-<br>well C |
| 820                                                       | 700           | 61.7            | 700                | 513           | 50.3            |
| 810                                                       | 685           | 61.0            | 690                | 497           | 49.1            |
| 800                                                       | 669           | 60.2            | 680                | 482           | 47.8            |
| 790                                                       | 653           | 59.4            | 670                | 466           | 46.5            |
| 780                                                       | 638           | 58.5            | 660                | 451           | 45.2            |
| 770                                                       | 622           | 57.7            | 650                | 435           | 43.8            |
| 760                                                       | 607           | 56.7            | 640                | 419           | 42.4            |
| 750                                                       | 591           | 55.7            | 630                | 404           | 41.0            |
| 740                                                       | 575           | 54.7            | 620                | 388           | 39.5            |
| 730                                                       | 560           | 53.7            | 610                | 373           | 37.9            |
| 720                                                       | 544           | 52.6            | 600                | 357           | 36.4            |
| 710                                                       | 529           | 51.4            | 590                | 341           | 34.8            |

| Brinell Hardness to Equotip $L_D$ and Rockwell C Hardness Numbers: |                     |                 |               |                     |                 |
|--------------------------------------------------------------------|---------------------|-----------------|---------------|---------------------|-----------------|
| HB<br>10/3000                                                      | Equo-<br>tip, $L_D$ | Rock-<br>well C | HB<br>10/3000 | Equo-<br>tip, $L_D$ | Rock-<br>well C |
| 700                                                                | 819                 | 61.1            | 500           | 694                 | 49.3            |
| 680                                                                | 806                 | 60.2            | 480           | 681                 | 47.7            |
| 660                                                                | 794                 | 59.2            | 460           | 669                 | 46.1            |
| 640                                                                | 781                 | 58.2            | 440           | 656                 | 44.4            |
| 620                                                                | 769                 | 57.2            | 420           | 644                 | 42.6            |
| 600                                                                | 756                 | 56.0            | 400           | 631                 | 40.8            |
| 580                                                                | 744                 | 54.8            | 380           | 618                 | 38.9            |
| 560                                                                | 731                 | 53.5            | 360           | 606                 | 36.9            |
| 540                                                                | 719                 | 52.2            | 340           | 593                 | 34.9            |
| 520                                                                | 706                 | 50.7            |               |                     |                 |

| Rockwell to Brinell and Equotip $L_D$ Hardness Numbers: |               |                     |                 |               |                     |
|---------------------------------------------------------|---------------|---------------------|-----------------|---------------|---------------------|
| Rock-<br>well C                                         | HB<br>10/3000 | Equo-<br>tip, $L_D$ | Rock-<br>well C | HB<br>10/3000 | Equo-<br>tip, $L_D$ |
| 61.5                                                    | 701           | 811                 | 46.5            | 463           | 673                 |
| 60                                                      | 672           | 795                 | 45              | 444           | 662                 |
| 58.5                                                    | 645           | 780                 | 43.5            | 427           | 650                 |
| 57                                                      | 619           | 765                 | 42              | 410           | 639                 |
| 55.5                                                    | 594           | 751                 | 40.5            | 395           | 629                 |
| 54                                                      | 570           | 737                 | 39              | 380           | 619                 |
| 52.5                                                    | 546           | 724                 | 37.5            | 367           | 609                 |
| 51                                                      | 524           | 710                 | 36              | 354           | 600                 |
| 49.5                                                    | 503           | 698                 | 34.5            | 342           | 591                 |
| 48                                                      | 482           | 685                 |                 |               |                     |

Table D3, continued)

| <b>Brinell to Vickers and Rockwell C (Parks, 1980):</b> |                       |                         |                       |                       |                         |
|---------------------------------------------------------|-----------------------|-------------------------|-----------------------|-----------------------|-------------------------|
| <b>HB<br/>10/3000</b>                                   | <b>Vick.<br/>HV50</b> | <b>Rock-<br/>well C</b> | <b>HB<br/>10/3000</b> | <b>Vick.<br/>HV50</b> | <b>Rock-<br/>well C</b> |
| 900 <sup>(1)</sup>                                      | 996                   | 69                      | 560                   | 610                   | 54                      |
| 880                                                     | 973                   | 69                      | 550                   | 599                   | 53                      |
| 860                                                     | 951                   | 68                      | 540                   | 587                   | 53                      |
| 840                                                     | 928                   | 67                      | 530                   | 576                   | 52                      |
| 820                                                     | 905                   | 66                      | 520                   | 565                   | 51                      |
| 800                                                     | 882                   | 65                      | 510                   | 553                   | 51                      |
| 780                                                     | 860                   | 65                      | 500                   | 542                   | 50                      |
| 760                                                     | 837                   | 64                      | 490                   | 530                   | 49                      |
| 740                                                     | 814                   | 63                      | 480                   | 519                   | 49                      |
| 720                                                     | 792                   | 62                      | 470                   | 508                   | 48                      |
| 700                                                     | 769                   | 61                      | 460                   | 496                   | 47                      |
| 680                                                     | 746                   | 60                      | 450                   | 485                   | 46                      |
| 660                                                     | 723                   | 59                      | 440                   | 474                   | 45                      |
| 640                                                     | 701                   | 58                      | 430                   | 462                   | 44                      |
| 620                                                     | 678                   | 57                      | 420                   | 451                   | 43                      |
| 610                                                     | 667                   | 57                      | 410                   | 440                   | 42                      |
| 600                                                     | 655                   | 56                      | 400                   | 428                   | 40                      |
| 590                                                     | 644                   | 56                      | 390                   | 417                   | 39                      |
| 580                                                     | 632                   | 55                      | 380                   | 406                   | 37                      |
| 570                                                     | 621                   | 55                      | 370                   | 394                   | 33                      |

(1) Brinell numbers above 630 are not recommended by ASTM. Nevertheless, Brinell numbers up to 700 HB are industrially accepted.

### References/Bibliography for Appendix D

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### Rockwell C to Brinell and Vickers (Parks, 1980):

| <b>Rock-<br/>well C</b> | <b>HB<br/>10/3000</b> | <b>Vick.<br/>HV50</b> | <b>Rock-<br/>well C</b> | <b>HB<br/>10/3000</b> | <b>Vick.<br/>HV50</b> |
|-------------------------|-----------------------|-----------------------|-------------------------|-----------------------|-----------------------|
| 68                      | 865 <sup>(1)</sup>    | 948                   | 51                      | 514                   | 563                   |
| 67                      | 839                   | 920                   | 50                      | 500                   | 546                   |
| 66                      | 813                   | 892                   | 49                      | 486                   | 530                   |
| 65                      | 788                   | 866                   | 48                      | 473                   | 515                   |
| 64                      | 764                   | 840                   | 47                      | 461                   | 501                   |
| 63                      | 740                   | 814                   | 46                      | 450                   | 487                   |
| 62                      | 717                   | 789                   | 45                      | 440                   | 474                   |
| 61                      | 695                   | 765                   | 44                      | 430                   | 461                   |
| 60                      | 674                   | 742                   | 43                      | 421                   | 450                   |
| 59                      | 653                   | 719                   | 42                      | 412                   | 439                   |
| 58                      | 633                   | 697                   | 41                      | 405                   | 428                   |
| 57                      | 614                   | 676                   | 40                      | 398                   | 419                   |
| 56                      | 595                   | 655                   | 39                      | 392                   | 410                   |
| 55                      | 578                   | 635                   | 38                      | 386                   | 401                   |
| 54                      | 560                   | 616                   | 37                      | 382                   | 394                   |
| 53                      | 544                   | 598                   | 36                      | 378                   | 387                   |
| 52                      | 529                   | 580                   | 35                      | 374                   | 381                   |

(1) Brinell numbers above 630 are not recommended by ASTM. Nevertheless, Brinell numbers up to 700 HB are industrially accepted.

### Vickers to Brinell and Rockwell C (Parks, 1980):

| <b>Vick.<br/>HV50</b> | <b>HB<br/>10/3000</b> | <b>Rock-<br/>well C</b> | <b>Vick.<br/>HV50</b> | <b>HB<br/>10/3000</b> | <b>Rock-<br/>well C</b> |
|-----------------------|-----------------------|-------------------------|-----------------------|-----------------------|-------------------------|
| 1000                  | 857 <sup>(1)</sup>    | 70                      | 680                   | 576                   | 57                      |
| 980                   | 840                   | 69                      | 660                   | 558                   | 56                      |
| 960                   | 822                   | 68                      | 640                   | 541                   | 55                      |
| 940                   | 804                   | 68                      | 620                   | 523                   | 54                      |
| 920                   | 787                   | 67                      | 600                   | 505                   | 53                      |
| 900                   | 769                   | 66                      | 580                   | 488                   | 52                      |
| 880                   | 752                   | 66                      | 560                   | 470                   | 51                      |
| 860                   | 734                   | 65                      | 540                   | 452                   | 50                      |
| 840                   | 717                   | 64                      | 520                   | 435                   | 48                      |
| 820                   | 699                   | 63                      | 500                   | 417                   | 47                      |
| 800                   | 681                   | 62                      | 480                   | 400                   | 45                      |
| 780                   | 664                   | 62                      | 460                   | 382                   | 44                      |
| 760                   | 646                   | 61                      | 440                   | 364                   | 42                      |
| 740                   | 629                   | 60                      | 420                   | 347                   | 40                      |
| 720                   | 611                   | 59                      | 400                   | 329                   | 38                      |
| 700                   | 593                   | 58                      | 380                   | 312                   | 35                      |

(1) Brinell numbers above 630 are not recommended by ASTM. Nevertheless, Brinell numbers up to 700 HB are industrially accepted.

## Appendix E

### Standards

Worldwide specifications are presented in Tables E1 through E7 (current as of 1999). These tables should be viewed as a good starting point and not the final word.

Table E8 presents the specifications for the new European Standard EN 12 513, which has been issued and will probably be effective from 2001 onward. From then on, all national standards of the EU member countries will not be officially valid anymore. Table E9 is a synoptic table, showing the comparisons of EN and ASTM grades.

**Table E1**  
**National Standard Designations Correlating Nearest Equivalent Grades**

| International<br>Trade Name | France<br>NF A 32-401 | W. Germany<br>DIN 1695                     | U.K.<br>BS 4844 | India<br>IS 7925                               | Sweden<br>SIS 14 04<br>XX<br>IS 4771<br>SIS 14 05<br>XX | U.S.A.<br>ASTM A532                  |
|-----------------------------|-----------------------|--------------------------------------------|-----------------|------------------------------------------------|---------------------------------------------------------|--------------------------------------|
| — —                         | —                     | —                                          | 1A              | 1                                              | —                                                       | —                                    |
| — —                         | —                     | —                                          | 1B              | 2                                              | —                                                       | —                                    |
| — —                         | —                     | —                                          | 1C              | 3                                              | —                                                       | —                                    |
| Ni-Hard Type 2              | FB Ni4 Cr2 BC         | G-X 260 NiCr 4 2                           | 2A              | Type 1a. NiLCr 30/500                          | 05 12                                                   | Class I, Type B                      |
| Ni-Hard Type 1              | FB Ni4 Cr2 HC         | G-X 330 NiCr 4 2                           | 2B              | Type 1a. NiLCr 34/550                          | 05 13                                                   | Class I, Type A                      |
| Ni-Hard Type 3              | FB A                  | —                                          | —               | —                                              | —                                                       | Class I, Type C                      |
| Ni-Hard Type 4              | FB Cr9 Ni5            | G-X 300 CrNiSi 9 5 2                       | 2C + 2D + 2E    | Type 1b. NiHCr                                 | 04 57                                                   | Class I, Type D                      |
| —                           | FB Cr12 Mo Ni         | —                                          | —               | —                                              | —                                                       | Class II. Type A                     |
| —                           | FB Cr15 Mo Ni         | G-X 300 CrMo 15 3<br>G-X 300 CrMoNi 15 2 1 | 3A<br>3B        | Type 2. CrMoLC 28/500<br>Type 2. CrMoHC 34/500 | —<br>—                                                  | Class II. Type B<br>Class II. Type C |
| —                           | FB Cr20 Mo Ni         | G-X 260 CrMoNi 20 2 1                      | 3C              | —                                              | —                                                       | Class II.<br>Types D + E             |
| —                           | FB Cr26 Mo Ni         | G-X 260 Cr 27<br>G-X 300 CrMo 27 1         | 3D<br>3E        | Type 3. HCr + HCrNi                            | 04 66                                                   | Class III. Type A                    |

**Table E2**  
**France — Specifications**

| Specification No. | Grade         | Chemical Composition, % |         |                        |                        |                        |                        |                        | Hardness HB <sup>(1)</sup> |
|-------------------|---------------|-------------------------|---------|------------------------|------------------------|------------------------|------------------------|------------------------|----------------------------|
|                   |               | C                       | Si      | Mn                     | Ni                     | Cr                     | Cu                     | Mo                     |                            |
| NF A 32-401-      | FB O          | 2.7–3.9                 | 0.4–1.5 | 0.2–0.8                | —                      | —                      | —                      | —                      | 350–500                    |
| April 1980        | FB A          | 2.7–3.9                 | 0.4–1.5 | 0.2–0.8                | 0.3–3.0 <sup>(2)</sup> | 0.2–2.0 <sup>(2)</sup> | 0.3–2.0 <sup>(2)</sup> | 0.1–1.0 <sup>(2)</sup> | 400–600                    |
|                   | FB Ni4 Cr2 BC | 2.7–3.2                 | 0.2–0.8 | 0.3–0.7                | 3.0–5.5 <sup>(3)</sup> | 1.5–2.5                | —                      | 0.0–1.0                | 450–650                    |
|                   | FB Ni4 Cr2 HC | 3.2–3.6                 | 0.2–0.8 | 0.3–0.7                | 3.0–5.5 <sup>(3)</sup> | 1.5–2.5                | —                      | 0.0–1.0                | 500–700                    |
|                   | FB Cr9 Ni5    | 2.5–3.6                 | 1.5–2.2 | 0.3–0.7                | 4.0–6.0                | 5.0 11.0               | —                      | 0.5 max                | 550–750                    |
|                   | FB Cr12 Mo Ni | 2.0–3.6                 | 0.2–0.8 | 0.5–1.0                | 0.0–2.0                | 11.0–14.0              | —                      | 0.5–3.0                | 500–800                    |
|                   | FB Cr15 Mo Ni | 2.0–3.6                 | 0.2–0.8 | 0.5–1.0                | 0.0–2.5                | 14.0–17.0              | —                      | 0.5–3.0                | 500–800                    |
|                   | FB Cr20 Mo Ni | 2.0–3.6                 | 0.2–1.2 | 0.5–1.5                | 0.0–2.5                | 17.0–22.0              | 0.0–1.5                | 0.5–3.0                | 500–800                    |
|                   | FB Cr26 Mo Ni | 1.5–3.5                 | 0.2–1.2 | 0.5–1.5 <sup>(4)</sup> | 0.0–2.5                | 22.0–28.0              | 0.0–1.5                | 0.5–3.0                | 450–650                    |

*(Hardness is specified while compositional values are for guidance)*

*(1) The highest hardness values apply to thin-section castings and/or shell-molded.*

*(2) Optional or, if necessary, as a combination.*

*(3) For these alloys, the Ni may be partly replaced by Cu.*

*(4) In order to obtain a hard austenitic structure, the percentage of Mn may be 4.0 max*

**Table E3**  
**Germany — Specifications**

| Specification No. | Grade                 | Chemical Composition, % |         |         |           |                        |         | Hardness                                                       |                                     |                             |
|-------------------|-----------------------|-------------------------|---------|---------|-----------|------------------------|---------|----------------------------------------------------------------|-------------------------------------|-----------------------------|
|                   |                       | C                       | Si      | Mn      | Cr        | Ni                     | Mo      | Special Heat Treatment or As-Cast <sup>(1)</sup> Min HV30 (HB) | Hardened and Tempered Min HV30 (HB) | Soft Annealed Max HV30 (HB) |
| DIN-              | G-X 300 NiMo 3 Mg     | 2.8–3.5                 | 2.0–2.6 | 0.2–0.5 | —         | 1.5–4.5                | 0.5–0.8 | 400 (380)                                                      | 550 (500)                           | 300 (290)                   |
| 1695              | G-X 260 NiCr 4 2      | 2.6–2.9                 | 0.2–0.8 | 0.3–0.7 | 1.4–2.4   | 3.3–5.0                | 0.5     | 450 (430)                                                      | 520 (480)                           | —                           |
| Sept.             | G-X 330 NiCr 4 2      | 3.0–3.6                 | 0.2–0.8 | 0.3–0.7 | 1.4–2.4   | 3.3–5.0                | 0.5     | 450 (430)                                                      | 550 (500)                           | —                           |
| 1981              | G-X 300 CrNiSi 9 5 2  | 2.5–3.5                 | 1.5–2.2 | 0.3–0.7 | 8.0–10.0  | 4.5–6.5                | 0.5     | 450 (430)                                                      | 600 (535)                           | —                           |
|                   | G-X 300 CrMo 15 3     | 2.3–3.6                 | 0.2–0.8 | 0.5–1.0 | 14.0–17.0 | 0.72                   | 1.0–3.0 | 450 (430)                                                      | 600 (535)                           | 400 (380) <sup>(3)</sup>    |
|                   | G-X 300 CrMoNi 15 2 1 | 2.3–3.6                 | 0.2–0.8 | 0.5–1.0 | 14.0–17.0 | 0.8–1.2 <sup>(2)</sup> | 1.8–2.2 | 450 (430)                                                      | 600 (535)                           | 400 (380) <sup>(3)</sup>    |
|                   | G-X 260 CrMoNi 20 2 1 | 2.3–2.9                 | 0.2–0.8 | 0.5–1.0 | 18.0–22.0 | 0.8–1.2 <sup>(2)</sup> | 1.4–2.2 | 450 (430)                                                      | 600 (535)                           | 420 (400)                   |
|                   | G-X 260 Cr 27         | 2.3–2.9                 | 0.5–1.5 | 0.5–1.5 | 24.0–28.0 | 1.2                    | 1.0     | 450 (430)                                                      | 550 (500)                           | 400 (380)                   |
|                   | G-X 300 CrMo 27 1     | 3.0–3.5                 | 0.2–1.0 | 0.5–1.0 | 23.0–28.0 | 1.2                    | 1.0–2.0 | 450 (430)                                                      | 600 (535)                           | 400 (380)                   |

*(1) These materials shall be supplied in the as-cast state only when no requirements are made about impact strength*

*(2) Can also be produced as a copper-containing alloy.*

*(3) Up to a max carbon content of 3.1%.*

*(Continued on next page.)*

**Table E4**  
**United Kingdom — Specifications**

| Speci-<br>fication<br>No. | Grade | Chemical Composition % |         |         |         |           |         |         |                  | Hardness <sup>(1)</sup><br>HB, min |               |
|---------------------------|-------|------------------------|---------|---------|---------|-----------|---------|---------|------------------|------------------------------------|---------------|
|                           |       | C                      | Si      | Mn      | Ni      | Cr        | Mo      | Cu      | P <sub>max</sub> | S <sub>max</sub>                   |               |
| BS 4844                   | 1A    | 2.4–3.4                | 0.5–1.5 | 0.2–0.8 | —       | 2.0 max   | —       | —       | 0.15             | —                                  | 400           |
| Part 1:                   | 1B    | 2.4–3.4                | 0.5–1.5 | 0.2–0.8 | —       | 2.0 max   | —       | —       | 0.5              | —                                  | 400           |
| 1974                      | 1C    | 2.4–3.0                | 0.5–1.5 | 0.2–0.8 | —       | 2.0 max   | —       | —       | 0.15             | —                                  | 250           |
| BS 4844                   | 2A    | 2.7–3.2                | 0.3–0.8 | 0.2–0.8 | 3.0–5.5 | 1.5–2.5   | 0.5 max | —       | 0.3              | 0.15                               | 500           |
| Part 2:                   | 2B    | 3.2–3.6                | 0.3–0.8 | 0.2–0.8 | 3.0–5.5 | 1.5–2.5   | 0.5 max | —       | 0.3              | 0.15                               | 550           |
| 1972                      | 2C    | 2.4–2.8                | 1.5–2.2 | 0.2–0.8 | 4.0–6.0 | 8.0–10.0  | 0.5 max | —       | 0.3              | 0.15                               | 500           |
|                           | 2D    | 2.8–3.2                | 1.5–2.2 | 0.2–0.8 | 4.0–6.0 | 8.0–10.0  | 0.5 max | —       | 0.3              | 0.15                               | 550           |
|                           | 2E    | 3.2–3.6                | 1.5–2.2 | 0.2–0.8 | 4.0–6.0 | 8.0–10.0  | 0.5 max | —       | 0.3              | 0.15                               | 600           |
| BS 4844                   | 3A    | 2.4–3.0                | 1.0 max | 0.5–1.5 | 0–1.0   | 14.0–17.0 | 0.0–2.5 | 0.0–1.2 | 0.1              | —                                  | 450 (as-cast) |
| Part 3:                   | 3B    | 3.0–3.6                | 1.0 max | 0.5–1.5 | 0–1.0   | 14.0–17.0 | 1.0–3.0 | 0.0–1.2 | 0.1              | —                                  | 500 (as-cast) |
| 1974                      | 3C    | 2.2–3.0                | 1.0 max | 0.5–1.5 | 0–1.0   | 17.0–22.0 | 0.0–3.0 | 0.0–1.2 | 0.1              | —                                  | 450 (as-cast) |
|                           | 3D    | 2.4–2.8                | 1.0 max | 0.5–1.5 | 0–1.0   | 22.0–28.0 | 0.0–1.5 | 0.0–1.2 | 0.1              | —                                  | 400 (as-cast) |
|                           | 3E    | 2.8–3.2                | 1.0 max | 0.5–1.5 | 0–1.0   | 22.0–28.0 | 0.0–1.5 | 0.0–1.2 | 0.1              | —                                  | 450 (as-cast) |

(1) Part 1: In castings with a section thickness of greater than 50 mm (2.0 in.), HB may be 50 less than specified.  
Part 2: In castings with a section thickness of greater than 125 mm (5.0 in.), HB may be 50 less than specified.

**Table E5**  
**India — Specifications**

| Specifica-<br>tion No. | Grade          | Chemical Composition, % |         |         |         |           |         |                  |                  | Hardness, HB, min <sup>(1)</sup> |          |          |
|------------------------|----------------|-------------------------|---------|---------|---------|-----------|---------|------------------|------------------|----------------------------------|----------|----------|
|                        |                | C                       | Si      | Mn      | Ni      | Cr        | Mo      | S <sub>max</sub> | P <sub>max</sub> | As-Cast                          | Hardened | Annealed |
| IS 7925 -              | 1              | 2.4–3.4                 | 0.5–1.5 | 0.2–0.8 | —       | 2.0 max   | —       | —                | 0.15             | 400                              | —        | —        |
| 1976 <sup>(2)</sup>    | 2              | 2.4–3.4                 | 0.5–1.5 | 0.2–0.8 | —       | 2.0 max   | —       | —                | 0.15             | 400                              | —        | —        |
|                        | 3              | 2.4–3.0                 | 0.5–1.5 | 0.2–0.8 | —       | 2.0 max   | —       | —                | 0.15             | 250                              | —        | —        |
| IS 4771 -              | <b>Type 1a</b> |                         |         |         |         |           |         |                  |                  |                                  |          |          |
| 1972                   | NiCr 30/500    | 2.7–3.3                 | 0.3–0.6 | 0.3–0.6 | 3.0–5.5 | 1.5–2.5   | 0.5 max | 0.15             | 0.3              | 500                              | —        | —        |
|                        | NiCr 34/550    | 3.2–3.6                 | 0.3–0.6 | 0.3–0.6 | 3.0–5.5 | 1.5–2.5   | 0.5 max | 0.15             | 0.3              | 550                              | —        | —        |
|                        | <b>Type 1b</b> |                         |         |         |         |           |         |                  |                  |                                  |          |          |
|                        | NiCr 27/500    | 2.5–2.9                 | 1.5–2.2 | 0.3–0.6 | 4.0–6.0 | 8.0–10.0  | 0.5 max | 0.15             | 0.3              | 500                              | —        | —        |
|                        | NiCr 30/550    | 2.8–3.3                 | 1.5–2.2 | 0.3–0.6 | 4.0–6.0 | 8.0–10.0  | 0.5 max | 0.15             | 0.3              | 550                              | —        | —        |
|                        | NiCr 34/600    | 3.2–3.6                 | 1.5–2.2 | 0.3–0.6 | 4.0–6.0 | 7.5–9.5   | 0.5 max | 0.15             | 0.3              | 600                              | —        | —        |
|                        | <b>Type 2</b>  |                         |         |         |         |           |         |                  |                  |                                  |          |          |
|                        | CrMoHC 34/500  | 3.1–3.6                 | 0.3–0.8 | 0.4–0.9 | 0.5 max | 14.0–18.0 | 2.5–3.5 | 0.15             | 0.3              | 500                              | 600      | 380 max  |
|                        | CrMoLC 28/500  | 2.4–3.1                 | 0.3–0.8 | 0.4–0.9 | 0.5 max | 14.0–18.0 | 2.5–3.5 | 0.15             | 0.3              | 500                              | 550      | 380 max  |
|                        | <b>Type 3</b>  |                         |         |         |         |           |         |                  |                  |                                  |          |          |
|                        | HCrNi 27/400   | 2.3–3.0                 | 0.2–1.5 | 1.5 max | 1.2 max | 24.0–28.0 | 0.6 max | 0.15             | 0.3              | 400                              | 550      | —        |
|                        | HCr 27/400     | 2.3–3.0                 | 0.2–1.5 | 1.5 max | 0.5 max | 24.0–28.0 | 0.6 max | 0.15             | 0.3              | 400                              | 550      | 380 max  |

(1) IS 4771 - 1972 for castings with a section thickness greater than 125 mm (5 in.), HB may be 50 less than specified.  
(2) IS 7925 - 1976 composition is specified; purchaser can request specified hardness values.

**Table E6**  
**Sweden — Specifications**

| Specification No. | Grade | Chemical Composition, % |         |         |           |         |                  |                  | Hardness (As-Cast) HB, min |
|-------------------|-------|-------------------------|---------|---------|-----------|---------|------------------|------------------|----------------------------|
|                   |       | C                       | Si      | Mn      | Cr        | Ni      | P <sub>max</sub> | S <sub>max</sub> |                            |
| SIS 14 05 12 1971 | 05 12 | 2.7–3.3                 | 0.3–0.6 | 0.3–0.6 | 1.5–2.5   | 3.0–5.5 | 0.3              | 0.15             | 500                        |
| SIS 14 05 13 1971 | 05 13 | 3.2–3.6                 | 0.3–0.6 | 0.3–0.6 | 1.5–2.5   | 3.0–5.5 | 0.3              | 0.15             | 550                        |
| SIS 14 04 57 1971 | 04 57 | 2.8–3.3                 | 1.5–2.2 | 0.3–0.6 | 8.0–10.0  | 4.0–6.0 | 0.3              | 0.15             | 550                        |
| SIS 14 04 66 1971 | 04 66 | 2.5 min                 | 1.0     | 0.5     | 24.0–30.0 | —       | —                | —                | 550                        |

**Table E7a**  
**U.S.A. - ASTM A 532 / A 532M - 1993 Compositional Specifications**

| Class | Type | Designation | Chemical Composition, % |         |         |         |           |         |         |          | S        |
|-------|------|-------------|-------------------------|---------|---------|---------|-----------|---------|---------|----------|----------|
|       |      |             | C                       | Mn      | Si      | Ni      | Cr        | Mo      | Cu      | P        |          |
| I     | A    | Ni-Cr-HC    | 2.8–3.6                 | 2.0 max | 0.8 max | 3.3–5.0 | 1.4–4.0   | 1.0 max | —       | 0.3 max  | 0.15 max |
| I     | B    | Ni-Cr-LC    | 2.4–3.0                 | 2.0 max | 0.8 max | 3.3–5.0 | 1.4–4.0   | 1.0 max | —       | 0.3 max  | 0.15 max |
| I     | C    | Ni-Cr-GB    | 2.5–3.7                 | 2.0 max | 0.8 max | 4.0 max | 1.0–2.5   | 1.0 max | —       | 0.3 max  | 0.15 max |
| I     | D    | Ni-HiCr     | 2.5–3.6                 | 2.0 max | 2.0 max | 4.5–7.0 | 7.0–11.0  | 1.5 max | —       | 0.10 max | 0.15 max |
| II    | A    | 12% Cr      | 2.0–3.3                 | 2.0 max | 1.5 max | 2.5 max | 11.0–14.0 | 3.0 max | 1.2 max | 0.10 max | 0.06 max |
| II    | B    | 15% Cr-Mo   | 2.0–3.3                 | 2.0 max | 1.5 max | 2.5 max | 14.0–18.0 | 3.0 max | 1.2 max | 0.10 max | 0.06 max |
| II    | D    | 20% Cr-Mo   | 2.0–3.3                 | 2.0 max | 1.0–2.2 | 2.5 max | 18.0–23.0 | 3.0 max | 1.2 max | 0.10 max | 0.06 max |
| III   | A    | 25% Cr      | 2.0–3.3                 | 2.0 max | 1.5 max | 2.5 max | 23.0–30.0 | 3.0 max | 1.2 max | 0.10 max | 0.06 max |

*Designation: HC = high carbon; LC = low carbon; GB = Great Britain*

**Table E7b**  
**U.S.A. - ASTM A 532 / A 532M - 1993 Hardness Requirements**

| Class    Type    Designation |     |           | As-Cast or<br>As-Cast and<br>Relieved |     |     | Sand Cast, min <sup>(1)</sup>               |     |     |         |     |     | Chill<br>Cast, min <sup>(2)</sup> |     |     | Softened,<br>max |     |     |
|------------------------------|-----|-----------|---------------------------------------|-----|-----|---------------------------------------------|-----|-----|---------|-----|-----|-----------------------------------|-----|-----|------------------|-----|-----|
|                              |     |           |                                       |     |     | Hardened or<br>Hardened and Stress Relieved |     |     |         |     |     |                                   |     |     |                  |     |     |
|                              |     |           |                                       |     |     | Level 1                                     |     |     | Level 2 |     |     |                                   |     |     |                  |     |     |
| HB                           | HRC | HV        | HB                                    | HRC | HV  | HB                                          | HRC | HV  | HB      | HRC | HV  | HB                                | HRC | HV  | HB               | HRC | HV  |
| I                            | A   | Ni-Cr-Hc  | 550                                   | 53  | 600 | 600                                         | 56  | 660 | 650     | 59  | 715 | 600                               | 56  | 660 | —                | —   | —   |
| I                            | B   | Ni-Cr-Lc  | 550                                   | 53  | 600 | 600                                         | 56  | 660 | 650     | 59  | 715 | 600                               | 56  | 660 | —                | —   | —   |
| I                            | C   | Ni-Cr-GB  | 550                                   | 53  | 600 | 600                                         | 56  | 660 | 650     | 59  | 715 | 600                               | 56  | 660 | 400              | 41  | 430 |
| I                            | D   | Ni-HiCr   | 500                                   | 50  | 540 | 600                                         | 56  | 660 | 650     | 59  | 715 | 550                               | 53  | 600 | —                | —   | —   |
| II                           | A   | 12% Cr    | 550                                   | 53  | 600 | 600                                         | 56  | 660 | 650     | 59  | 715 | 550                               | 53  | 600 | 400              | 41  | 430 |
| II                           | B   | 15% Cr-Mo | 450                                   | 46  | 485 | 600                                         | 56  | 660 | 650     | 59  | 715 | —                                 | —   | —   | 400              | 41  | 430 |
| II                           | D   | 20% Cr-Mo | 450                                   | 46  | 485 | 600                                         | 56  | 660 | 650     | 59  | 715 | —                                 | —   | —   | 400              | 41  | 430 |
| III                          | A   | 25% Cr    | 450                                   | 46  | 485 | 600                                         | 56  | 660 | 650     | 59  | 715 | —                                 | —   | —   | 400              | 41  | 430 |

*Designation: HC = high carbon; LC = low carbon; GB = Great Britain*

*(1) 90% of the minimum surface hardness level shall be maintained to a depth of 40% of the casting section, with any softer material being at the thermal center of the casting. A sampling procedure should be established by agreement between the supplier and the purchaser.*

*(2) Non-chilled areas of casting shall meet minimum hardness or sand-cast requirements.*

**Table E8**  
**AR White Cast Irons According to European Standard 12 513**

| Designation of<br>Mat'l. Grades <sup>(1)</sup>                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                    | Symbol        | Number  | ChemicalComposition % (Mass Fraction)<br>(Contents are max. unless range is indicated) |         |      |      |           |         |     |     | Vickers<br>Hardness<br>HV <sub>min</sub> |
|-----------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------|---------------|---------|----------------------------------------------------------------------------------------|---------|------|------|-----------|---------|-----|-----|------------------------------------------|
|                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                   |               |         | C                                                                                      | Si      | Mn   | P    | S         | Cr      | Ni  | Mo  |                                          |
| <b>High-Chromium:</b> <sup>(2,3,4)</sup>                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                          |               |         |                                                                                        |         |      |      |           |         |     |     |                                          |
| EN-GJN-<br>HV600(XCr11)                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                           | EN-JN<br>3019 | 1.8–3.6 | 1.0                                                                                    | 0.5–1.5 | 0.08 | 0.08 | 11.0–14.0 | 2.0     | 3.0 | 1.2 | 600                                      |
| EN-GJN-<br>HV600(XCr14)                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                           | EN-JN<br>3029 | 1.8–3.6 | 1.0                                                                                    | 0.5–1.5 | 0.08 | 0.08 | 14.0–18.0 | 2.0     | 3.0 | 1.2 | 600                                      |
| EN-GJN-<br>HV600(XCr18)                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                           | EN-JN<br>3039 | 1.8–3.6 | 1.0                                                                                    | 0.5–1.5 | 0.08 | 0.08 | 18.0–23.0 | 2.0     | 3.0 | 1.2 | 600                                      |
| EN-GJN-<br>HV600(XCr23)                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                           | EN-JN<br>3049 | 1.8–3.6 | 1.0                                                                                    | 0.5–1.5 | 0.08 | 0.08 | 23.0–26.0 | 2.0     | 3.0 | 1.2 | 600                                      |
| <b>Nickel-Chromium:</b> <sup>(2,4)</sup>                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                          |               |         |                                                                                        |         |      |      |           |         |     |     |                                          |
| EN-GJN-<br>HV520                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                  | EN-JN<br>2029 | 2.5–3.0 | 0.8                                                                                    | 0.8     | 0.10 | 0.10 | 1.5–3.0   | 3.0–5.5 | —   | —   | 520                                      |
| EN-GJN-<br>HV550                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                  | EN-JN<br>2039 | 3.0–3.6 | 0.8                                                                                    | 0.8     | 0.10 | 0.10 | 1.5–3.0   | 3.0–5.5 | —   | —   | 550                                      |
| EN-GJN-<br>HV600                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                  | EN-JN<br>2049 | 2.5–3.5 | 1.5–2.5                                                                                | 0.3–0.8 | 0.08 | 0.08 | 8.0–10.0  | 4.5–6.5 | —   | —   | 600                                      |
| <b>Unalloyed or Low-Alloy Chilled Iron:</b> <sup>(2)</sup>                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                        |               |         |                                                                                        |         |      |      |           |         |     |     |                                          |
| EN-GJN-<br>HV350                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                                  | EN-JN<br>2019 | 2.4–3.9 | 0.4–1.5                                                                                | 0.2–1.0 | —    | —    | 2.0       | —       | —   | —   | 350                                      |
| <p>(1) The material designation by symbol is in accordance with EN 1560 with additional information in brackets regarding their chemical composition. The material designation by number is in accordance with EN 1560</p> <p>(2) Both toughness and resistance to repeated impact increase as the carbon content decreases.</p> <p>(3) The microstructure of high-Cr AR cast irons depends on the cooling rate and is therefore sensitive to the casting thickness. To obtain specific properties (within the ranges in this table) some adjustments to the percentage contents of the alloying elements given may be necessary, depending on the casting dimensional variations.</p> <p>(4) In the case of castings with low alloy contents supplied in the as-cast condition and castings with greater thicknesses, it can be difficult to obtain the minimum hardness. The requirements for such castings should be agreed on between the manufacturer and the purchaser.</p> |               |         |                                                                                        |         |      |      |           |         |     |     |                                          |

**Table E9**  
**Comparison of EN and ASTM Grades  
and Common Tradenames**

| EN 12 513           | Number    | ASTM A532                 | Trade-name |
|---------------------|-----------|---------------------------|------------|
| EN-GJN-HV350        | EN-JN2019 |                           |            |
| EN-GJN-HV520        | EN-JN2029 | Class I Type B Ni-Cr-LC   | Ni-Hard 2  |
| EN-GJN-HV550        | EN-JN2039 | Class I Type B Ni-Cr-HC   | Ni-Hard 2  |
| EN-GJN-HV600        | EN-JN2049 | Class I Type D Ni-Hi-Cr   | Ni-Hard 2  |
| EN-GJN-HV600(XCr11) | EN-JN2019 | Class II Type A 12% Cr    |            |
| EN-GJN-HV600(XCr14) | EN-JN2029 | Class II Type B 15% Cr-Mo | 15-3       |
| EN-GJN-HV600(XCr18) | EN-JN2039 | Class II Type D 20% Cr-Mo | 15-2-1     |
| EN-GJN-HV360(XCr23) | EN-JN2049 | Class II Type A 25% Cr    | 20-2-1     |

## Appendix F

### Sample Preparation for Metallography

Metallography involves sectioning and polishing a metal sample, etching its surface to reveal its crystalline structure, and then analyzing the nature of this structure. Tables F1 and F2 should help in choosing the proper etchant.

Abrasion-resistant irons are usually some of the easiest materials to metallographically polish, but some of the most difficult to cut and section for metallography. This latter problem has slowly receded with the use of slow-speed tungsten-tipped band saw blades to section larger castings down to size, and then the use of electro-discharge-machining (EDM) to prepare plugs for subsequent mounting and polishing.

#### Sample Preparation and Mounting

It is difficult to surpass the efficiency of EDM for the preparation of metallographic samples from hard, carbide-containing materials. In most cases, a section of the casting can be broken or “torched” into a small enough chunk to fit within a standard EDM tank. From this point forward, a number of EDM technologies are available, from simple drill press configurations to 5-axis computer numeric control (CNC) machines.

For metallographic purposes, our preference is to EDM a small-diameter cylinder (i.e., 5 mm/0.25 in. dia by 25 mm/1.0 in. long) using a graphite/copper thin-wall tube electrode. This small cylinder can then easily be sectioned using a high-speed abrasive wheel, and mounted in Bakelite™ or epoxy for subsequent polishing. Figure F1 shows a photograph of how such a sample would be mounted.

The utility of mounting a sample in this configuration is that orthogonal planes of the alloy’s microstructure are presented within one mount. For AR cast irons, this type of metallographic approach can prevent the metallographer from making false assumptions about the alloy’s microstructure (e.g., spherical carbides).

#### Grinding and Polishing

Abrasion-resistant cast irons present some of the easiest ferrous materials to metallographically prepare. If only a quick and dirty metallographic analysis is required, grinding and polishing can be accomplished in four steps:

1. Coarse grind, using 150-mesh paper (see Table F3 for matching mesh to abrasive particle size classifications).
2. Subsequent grinding operation using 280- or 320-mesh papers.
3. Final grind using 600-mesh paper.
4. Simple polish using 6  $\mu\text{m}$  diamond paste on rotating cloth wheel.

This quick sequence produces acceptable metallographic surfaces in a matter of minutes. If high-resolution microscopy (above 500X) is desired, additional polishing steps are required. For example, the last 6  $\mu\text{m}$  polish stage is followed by 1  $\mu\text{m}$  and then 0.25  $\mu\text{m}$  polishing stages. Alternatively, vibratory polishing produces scratch-free surfaces if time is not of essence.

It should be kept in mind that there is no one “right” way to metallographically grind and polish samples. This brief section is merely meant to illustrate one popular technique that has been shown to be effective for AR cast irons.

**Table F1**  
**General Metallographic Etchants to Reveal White Iron Microstructures**

| Etchant           | Chemical Composition                                                                     | Remarks                                                      | Uses <sup>(1)</sup>                                                                                                                                                                                                 |
|-------------------|------------------------------------------------------------------------------------------|--------------------------------------------------------------|---------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------|
| Vilella's reagent | 5 ml HCl, 1 ml picric acid, 100 ml ethyl or methyl alcohol (95% or absolute)             | Swab or immerse from a few to 30 seconds to reveal structure | High-Cr irons: stains austenite a light yellow color and darkens martensite. Reveals $\text{Mo}_2\text{C}$ carbides along austenite grain boundaries.                                                               |
| 2% nital          | 2 ml $\text{HNO}_3$ , 98 ml ethyl or methyl alcohol (95% or absolute; also amyl alcohol) | Swab from a few to 15 seconds to reveal structure.           | Good general etchant for Ni-Hard 1, 2 and 4 alloys. Gives maximum contrast between pearlite and other phases. Is not recommended for as-cast high-Cr irons or other austenitic or ferritic grades of AR cast irons. |

(1) Any of these etchants can be used to reveal microstructure.

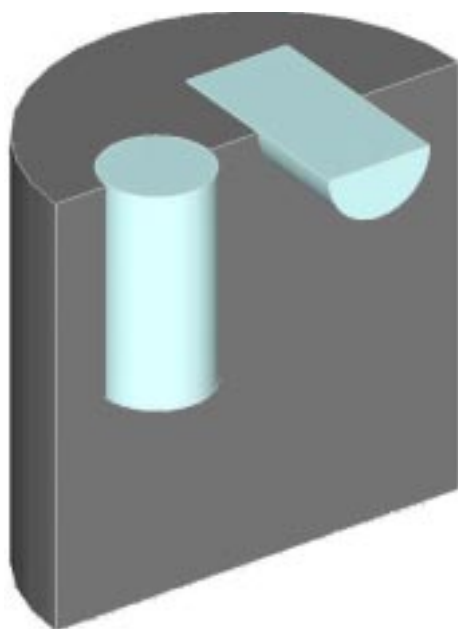
**Table F2**  
**Specific Metallographic Etchants to Reveal Various Carbide Structures**

| Etchant                                                | Chemical Composition                                               |                                     | M <sub>3</sub> C | M <sub>7</sub> C <sub>3</sub> | M <sub>23</sub> C <sub>6</sub> | M <sub>6</sub> C  | M <sub>2</sub> C |
|--------------------------------------------------------|--------------------------------------------------------------------|-------------------------------------|------------------|-------------------------------|--------------------------------|-------------------|------------------|
| Alkaline picrate                                       | picric acid<br>sodium hydroxide<br>water                           | 2 g<br>25 g<br>100 cm <sup>3</sup>  | A                | NA <sup>(1)</sup>             | NA <sup>(1)</sup>              | NA <sup>(1)</sup> | A                |
| Murakami <sup>(2)</sup>                                | potassium ferricyanide<br>potassium hydroxide<br>water             | 10 g<br>10 g<br>100 cm <sup>3</sup> | NA               | A                             | A                              | A                 | A                |
| Chromic acid electrolytic etch                         | chromic acid<br>water                                              | 1 g NA<br>100 cm <sup>3</sup>       | A                | NA                            | NA                             | A                 |                  |
| Sodium hydroxide saturated with potassium permanganate | 4% sodium hydroxide solution saturated with potassium permanganate |                                     | NA               | A                             | NA                             | A                 | A                |

*A = attacked; NA = not attacked.*

*(1) Lower chromium carbides are preferentially attacked.*

*(2) This stain does a superb job differentiating between the various carbides, especially the M<sub>3</sub>C and M<sub>7</sub>C<sub>3</sub> types.*



**Fig. F1.** Photograph showing a preferred method for sectioning and mounting a 5x25 mm (0.25x1.0 in.) pin in Bakelite or epoxy.

**Table F3**  
**Mesh and Grit Sizes for Abrasive Particles**

| Grit No.<br>(mesh) | U.S.A.<br>Sieve Series | Particle Size (RMS, μm) |                                  |          |
|--------------------|------------------------|-------------------------|----------------------------------|----------|
|                    |                        | Commercial              | FEFPA <sup>(1)</sup><br>F-Series | P-Series |
| 50                 | 350                    | 324                     |                                  |          |
| 60                 | 270                    | 274                     |                                  |          |
| 80                 | 190                    | 194                     |                                  |          |
| 120                | 115                    | 115                     |                                  |          |
| 150                | 80                     | 97                      |                                  |          |
| 180                | 70                     | 81                      |                                  |          |
| 220                | 62                     | 58                      |                                  |          |
| 240                | 54                     | 49                      | 49.3                             | 58.5     |
| 280                | 35                     | 41                      | 41.5                             | 52.2     |
| 320                | 29                     | 34                      | 34.4                             | 46.2     |
| 400                | 23                     | 29                      | 23.0                             | 35.0     |
| 600                | 17                     | 20                      | 14.3                             | 22.8     |

*(1) Federation Europeane des Fabricants des Produits Abrasifs.*

## Appendix G

### Calculating Superheat for AR Cast Irons

The purpose of this section is provide quick reference tables for calculation of liquidus start temperatures for standard commercial grades of AR cast irons. All equations were derived from the data provided in Arnold and Sare, "A Thermal Analysis Study of Alloy White Cast Irons," *The Foundryman*, vol. 80, pp. 330-335 (1987).

Table G1 gives the applicable chemistries and base equations used to generate the subsequent superheat tables. Tables G1 and G2 are specified by the iron's base chromium content. For irons that fall outside of this Cr content, liquidus temperatures can be directly calculated based on the Fe-Cr-C ternary phase diagram provided in Chapter 2, Solidification. Table G2 will provide the austenite liquidus temperatures.

**Table G1**  
**Applicable Chemistries and Superheat Equations for Typical AR Cast Irons<sup>(1)</sup>**

| Alloy Type                                                                                                                                                                                                                                | Applicable Chemistry Ranges (Wt%) |           | Liquidus Temperature <sup>(2)</sup><br>°C | Eutectic Temperature <sup>(3)</sup><br>°C (°F) |
|-------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------------|-----------------------------------|-----------|-------------------------------------------|------------------------------------------------|
|                                                                                                                                                                                                                                           | C                                 | Cr        |                                           |                                                |
| Ni-Hard 1 & 2                                                                                                                                                                                                                             | 2.5–4.3                           | 1.5–2.5   | 1556 – 94.4 · %C                          | 1150 (2100)                                    |
| Ni-Hard 4                                                                                                                                                                                                                                 | 2.0–3.7                           | 7.5–9.0   | 1486 – 85.3 · %C                          | 1170 (2138)                                    |
| ASTM II B–15% Cr                                                                                                                                                                                                                          | 2.0–3.5                           | 14.0–16.0 | 1500 – 80.0 · %C                          | 1220 (2228)                                    |
| ASTM II D–20% Cr                                                                                                                                                                                                                          | 2.0–3.4                           | 18.0–20.0 | 1446 – 60.7 · %C                          | 1240 (2264)                                    |
| ASTM III A–25% Cr                                                                                                                                                                                                                         | 2.0–2.8                           | 24.0–26.0 | 1485 – 75.0 · %C                          | 1275 (2327) <sup>(4)</sup>                     |
| (1) Data courtesy of B.K. Arnold and I.R. Sare, "A Thermal Analysis Study Of Alloy White Cast Irons," <i>The Foundryman</i> , vol. 80, pp. 330-335 (1987).                                                                                |                                   |           |                                           |                                                |
| (2) Accuracy of equations is estimated to be within 10°C of the equilibrium temperature.                                                                                                                                                  |                                   |           |                                           |                                                |
| (3) The ternary Fe-Cr-C liquidus surface (W.R. Thorpe and B. Chicco, "The Fe-Rich Corner of the Metastable C-Cr-Fe Liquidus Surface," <i>Metall. Trans. A.</i> , vol. 16A, pp. 1541-49, 1985) should be consulted for more accurate data. |                                   |           |                                           |                                                |
| (4) The 25% Cr iron has a very broad eutectic valley. According to Arnold and Sare (1987), the eutectic valley stretches from 2.8 to 3.2% carbon.                                                                                         |                                   |           |                                           |                                                |

**Table G2**  
**Approximate Austenite Liquidus Temperatures in °C (°F)**

| Carbon % | Ni-Hard 1 & 2<br>1.5–2.5% Cr | Ni-Hard 4<br>7.5–9.0% Cr | ASTM II B, 15% Cr<br>14.0–16.0% Cr | ASTM II D, 20% Cr<br>18.0–20.0% Cr | ASTM III A, 25% Cr<br>24.0–26.0% Cr |
|----------|------------------------------|--------------------------|------------------------------------|------------------------------------|-------------------------------------|
| 2        | —                            | 1315 (2400)              | 1340 (2444)                        | 1325 (2416)                        | 1335 (2435)                         |
| 2.1      | —                            | 1307 (2385)              | 1332 (2430)                        | 1319 (2405)                        | 1328 (2422)                         |
| 2.2      | —                            | 1298 (2369)              | 1324 (2415)                        | 1312 (2395)                        | 1320 (2408)                         |
| 2.3      | —                            | 1290 (2354)              | 1316 (2401)                        | 1306 (2384)                        | 1313 (2395)                         |
| 2.4      | —                            | 1281 (2339)              | 1308 (2386)                        | 1300 (2373)                        | 1305 (2381)                         |
| 2.5      | 1320 (2408)                  | 1273 (2323)              | 1300 (2372)                        | 1294 (2362)                        | 1298 (2368)                         |
| 2.6      | 1311 (2391)                  | 1264 (2308)              | 1292 (2358)                        | 1288 (2351)                        | 1290 (2354)                         |
| 2.7      | 1301 (2374)                  | 1256 (2293)              | 1284 (2343)                        | 1282 (2340)                        | 1283 (2341)                         |
| 2.8      | 1292 (2357)                  | 1247 (2277)              | 1276 (2329)                        | 1276 (2329)                        | 1275 (2327)                         |
| 2.9      | 1282 (2340)                  | 1239 (2262)              | 1268 (2314)                        | 1270 (2318)                        |                                     |
| 3        | 1273 (2323)                  | 1230 (2247)              | 1260 (2300)                        | 1264 (2307)                        |                                     |
| 3.1      | 1263 (2306)                  | 1222 (2231)              | 1252 (2286)                        | 1258 (2296)                        |                                     |
| 3.2      | 1254 (2289)                  | 1213 (2216)              | 1244 (2271)                        | 1252 (2285)                        |                                     |
| 3.3      | 1244 (2272)                  | 1205 (2200)              | 1236 (2257)                        | 1246 (2274)                        |                                     |
| 3.4      | 1235 (2255)                  | 1196 (2185)              | 1228 (2242)                        | 1240 (2263)                        |                                     |
| 3.5      | 1226 (2238)                  | 1187 (2170)              | 1220 (2228)                        |                                    |                                     |
| 3.6      | 1216 (2221)                  | 1179 (2154)              |                                    |                                    |                                     |
| 3.7      | 1207 (2204)                  | 1170 (2139)              |                                    |                                    |                                     |
| 3.8      | 1197 (2187)                  |                          |                                    |                                    |                                     |
| 3.9      | 1188 (2170)                  |                          |                                    |                                    |                                     |
| 4        | 1178 (2153)                  |                          |                                    |                                    |                                     |
| 4.1      | 1169 (2136)                  |                          |                                    |                                    |                                     |
| 4.2      | 1160 (2119)                  |                          |                                    |                                    |                                     |
| 4.3      | 1150 (2102)                  |                          |                                    |                                    |                                     |

## Appendix H

### Fluidity Measurements

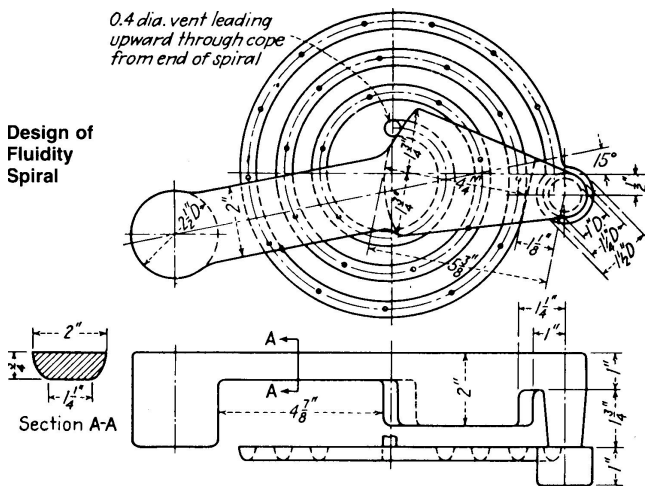
This appendix was developed from work authored by B. T Hodge of the Duramet Corporation in Muncy, Pennsylvania in cooperation with the AFS 5-D Technical Committee. The complete work can be found in *AFS Transactions*, vol. 107, pp. 743-749, 1999.

An alloy's fluidity is often thought of as a physical property, much akin to its liquidus and solidus temperatures. In foundry practice, fluidity has no simple clean definition. Over the years, an alloy's fluidity has come to denote its ability to flow within the confines of a narrow horizontal channel. This foundry definition has little to do with the more classical definition of fluidity as the reciprocal of viscosity. For example, molten steel's viscosity varies only slightly with temperature yet its fluidity is known to vary significantly with temperature (SCRATA).

When foundrymen ask about an alloy's fluidity, they want know how the alloy will flow or, more directly, whether or not it will fill their mold. To address these concerns, the concept of maximum fluidity length (Feliu et al., 1962) was developed to quantitatively indicate the maximum distance the metal can flow within a channel. In the tests described in this report a standard fluidity spiral (AFS, 1989) is used to measure this maximum fluidity length (Fig. H1). This length is then reported as the fluidity of a specific abrasion resistant cast iron.

Fluidity measurements were made on a wide range of abrasion resistant irons. Most of the samples have compositions which fall within the range of the grades listed in ASTM 532 (1993). However, some samples were nonstandard compositions. Measurements were made at participating foundries under foundry conditions using sand molds created from the same fluidity spiral pattern that was passed from foundry to foundry.

Foundries participating in the project include: Duramet Corporation, Muncy, PA, Eagle Foundry Co., Eagle Creek, OR, Columbia Steel, Portland, OR, Weatherly Casting and Machine Co., Weatherly, PA, Texaloy Foundry, Floresville, TX, and GIW Industries, Inc., Grovetown, GA. Additional technical and analytic support was provided by the National Research Council, Albany, OR, Climax Research Services, Farmington Hills, MI, and Predictive Engineering, Inc. Corvallis, OR.



**Fig. H1. Design of fluidity spiral and gating system.**

Multiple-linear regression analysis was used to describe correlation between chemistry, pouring temperature and fluidity. When data was grouped into sufficiently narrow ranges of composition, the resulting equations showed outstanding predictive capability. For ease of use, these equations are presented in a format that allows the calculation of fluidity based on pouring temperature and chemistry.

Table H1 gives a list of the raw data collected during this experiment. A total of 41 spirals were poured. Chromium content of these alloys ranged from 1% to 39%. For brevity, phosphorus and sulfur are not included in the report. However, in all samples, phosphorus was below 0.05%. Sulfur was not reported for all samples, but is assumed to be below the limits specified in ASTM A532.

The information in this Appendix should be useful to foundrymen who must make castings with abrasion resistant irons. It would be most desirable to predict the fluidity of an alloy in the simplest possible way. An initial attempt was made to correlate fluidity to superheat and freezing range, because they have an intuitive relationship to fluidity. Superheat (temperature above freezing point) would seem more directly related to fluidity than absolute pouring temperature. Freezing range (liquidus minus solidus temperatures) would seem to be a good indicator of how close the composition is to a eutectic "spike" in the fluidity curve. This approach also helps solve the rather messy problem one has to deal with by having so many elements of significance in AR iron compositions; they are all factored in to the liquidus and solidus temperatures.

There remains the problem, however, of determining liquidus and solidus temperatures. Few, if any, production foundries have DTA equipment to determine liquidus and solidus temperatures. Literature is available that can be used to determine liquidus and solidus temperatures from C and Cr composition (Arnold and Sare, 1987). However, many of the A532 grades contain significant amounts of elements other than C and Cr. This approach would also require a three-step process for the foundryman to determine fluidity: (1) determine liquidus and solidus temperatures from composition; (2) determine freezing range and superheat; and (3) determine fluidity.

After additional analysis, it was determined that correlation as good or better could be obtained directly by using chemistry and pour temperature as independent variables, and fluidity as the dependent variable. Chemistry and pour temperature are the variables most easily and directly measured by the foundry, and thus, presenting the relationships in this way seems to be most useful from a user point of view.

As previously suggested, multiple variable, linear regression analysis was used to determine the relationships between composition, pour temperature and fluidity. The specific software package used was Microsoft Excel '97. Empirically, it was determined that the best correlation could be obtained by splitting the data set into three subgroups: 1–4% Cr, 7–15% Cr and 20–39% Cr. Empirically, Mo and Cu appeared to have very little effect on fluidity, so the final regression analysis was made using pouring temperature, C, Cr, Si, Mn and Ni as the independent variables, and spiral length the dependent variable.

Calculations showed a correlation ( $R^2$ ) of between 95% and 98% for each of the three data groups. However, given the limited number of data sets in each group, the 95% confidence interval for each of the coefficients is quite large. In fact, the number of data points in the 7–15% Cr group was only seven. This is only one larger than the minimum number needed to calculate an "exact" solution. (Recall from basic linear algebra that the minimum number of data sets

needed to solve a system of linear simultaneous equations is equal to the number of independent variables.) For this set of data, a confidence interval could not even be calculated due to the minimal number of data sets.

Results of the regression analyses are shown in Tables H2–H4. Figures H2–H4 show plots of spiral length versus pour temperature for both the actual spiral length and the spiral length predicted by the regression equations.

The relationships in Table H5 are offered for fluidity of white irons within the range of chemistries shown. Caution should be used, however, given the limited amount of data used in calculating these coefficients, especially for the 7–15% Cr compositions.

As has been emphasized, the 95% confidence interval of the coefficients in Table H5 is largely due to the limited number of sample

points available for analysis. It is suggested that additional data would help improve the accuracy of the coefficients in the equations.

Probably the most difficult part of collecting accurate data is in obtaining reliable temperature readings. As experienced foundrymen know, ladle transfer and pouring must occur quickly in order to prevent significant heat loss. Moreover, the thermal gradients within a ladle of iron can be significant, on the order of 25–50°C, especially if teapot ladles are used. Furthermore, fluidity length is very sensitive to pouring temperature. Given all this, it is important to take pains that accurate temperature measurements are taken when pouring spirals. It is suspected that temperature measurement technique is probably a major source of what variations there are in the data. This might be an area for further investigation. For instance a Gage R&R study might be devised to determine the repeatability of temperature and spiral length at each foundry.

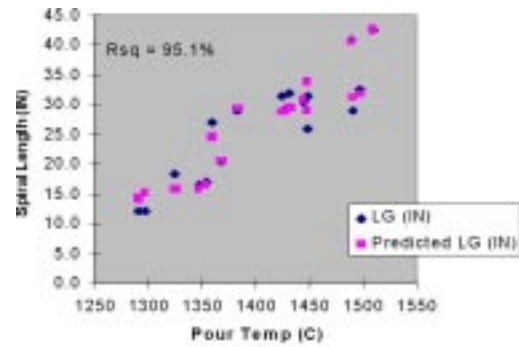
**Table H1**  
**Experimental Raw Data**

|    | Fdy. | Alloy    | C    | Si   | Mn   | Cr    | Mo    | Ni    | Cu    | PT   | LT   | ST   | SL (in.) |
|----|------|----------|------|------|------|-------|-------|-------|-------|------|------|------|----------|
| 1  | A    | IIIA (1) | 2.52 | 0.74 | 0.72 | 28.18 | 0.650 | 0.770 | 0.070 | 1498 | 1283 | 1272 | 30.0     |
| 2  | A    | IIIA (1) | 2.52 | 0.74 | 0.72 | 28.18 | 0.650 | 0.770 | 0.070 | 1509 | 1283 | 1272 | 35.5     |
| 3  | A    | IIE      | 3.09 | 0.64 | 0.66 | 20.05 | 1.980 | 0.180 | 0.970 | 1461 | 1255 | 1238 | 33.5     |
| 4  | A    | IIE      | 3.09 | 0.64 | 0.66 | 20.05 | 1.980 | 0.180 | 0.970 | 1462 | 1255 | 1238 | 34.0     |
| 5  | A    | IIE      | 3.09 | 0.64 | 0.66 | 20.05 | 1.980 | 0.180 | 0.970 | 1472 | 1255 | 1238 | 35.5     |
| 6  | A    | IIIA (2) | 2.85 | 0.68 | 0.66 | 25.69 | 0.210 | 0.300 | 0.100 | 1468 | 1285 | 1268 | 30.0     |
| 7  | A    | IIIA (2) | 2.85 | 0.68 | 0.66 | 25.69 | 0.210 | 0.300 | 0.100 | 1492 | 1285 | 1268 | 30.0     |
| 8  | A    | IA       | 3.28 | 0.64 | 0.41 | 2.11  | 0.190 | 4.420 | 0.044 | 1426 | 1240 | 1120 | 29.0     |
| 9  | A    | IA       | 3.28 | 0.64 | 0.41 | 2.11  | 0.190 | 4.420 | 0.044 | 1445 | 1240 | 1120 | 30.5     |
| 10 | B    | IIIA     | 2.59 | 0.86 | 0.96 | 26.21 | 0.260 | 1.060 | 0.120 | 1452 | 1268 | 1260 | 32.5     |
| 11 | B    | IIIA     | 2.59 | 0.86 | 0.96 | 26.21 | 0.260 | 1.060 | 0.120 | 1443 | 1268 | 1260 | 28.5     |
| 12 | B    | IIIA     | 2.59 | 0.86 | 0.96 | 26.21 | 0.260 | 1.060 | 0.120 | 1375 | 1268 | 1260 | 22.5     |
| 13 | B    | IIIA     | 2.59 | 0.86 | 0.96 | 26.21 | 0.260 | 1.060 | 0.120 | 1333 | 1268 | 1260 | 17.5     |
| 14 | B    | Non-Std. | 3.96 | 0.6  | 2.63 | 39.09 | 0.330 | 0.500 | 0.044 | 1518 | 1399 | 1269 | 10.0     |
| 15 | B    | Non-Std. | 3.96 | 0.6  | 2.63 | 39.09 | 0.330 | 0.500 | 0.044 | 1474 | 1399 | 1269 | 6.0      |
| 16 | B    | Non-Std. | 3.96 | 0.6  | 2.63 | 39.09 | 0.330 | 0.500 | 0.044 | 1401 | 1399 | 1269 | 3.5      |
| 17 | B    | ID       | 3.24 | 1.92 | 0.59 | 9.15  | 0.270 | 5.520 | 0.260 | 1376 | 1203 | 1168 | 35.5     |
| 18 | B    | ID       | 3.24 | 1.92 | 0.59 | 9.15  | 0.270 | 5.520 | 0.260 | 1244 | 1203 | 1168 | 12.5     |
| 19 | B    | ID       | 3.24 | 1.92 | 0.59 | 9.15  | 0.270 | 5.520 | 0.260 | 1228 | 1203 | 1168 | 9.5      |
| 20 | C    | IA       | 3.74 | 0.57 | 0.45 | 1.85  | 0.320 | 3.350 | 0.046 | 1383 | 1200 | 1110 | 29.0     |
| 21 | C    | IA       | 3.74 | 0.57 | 0.45 | 1.85  | 0.320 | 3.350 | 0.046 | 1509 | 1200 | 1110 | 42.5     |
| 22 | C    | IA       | 3.74 | 0.57 | 0.45 | 1.85  | 0.320 | 3.350 | 0.046 | 1489 | 1200 | 1110 | 41.0     |
| 23 | C    | IA       | 3.64 | 0.53 | 0.44 | 1.86  | 0.280 | 3.440 | 0.055 | 1431 | 1125 | 1066 | 32.0     |
| 24 | C    | IA       | 3.64 | 0.53 | 0.44 | 1.86  | 0.280 | 3.440 | 0.055 | 1424 | 1125 | 1066 | 31.5     |
| 25 | C    | IA       | 3.64 | 0.53 | 0.44 | 1.86  | 0.280 | 3.440 | 0.055 | 1297 | 1125 | 1066 | 12.0     |
| 26 | C    | IA       | 3.64 | 0.53 | 0.44 | 1.86  | 0.280 | 3.440 | 0.055 | 1291 | 1125 | 1066 | 12.0     |
| 27 | C    | IA       | 3.55 | 0.56 | 0.43 | 1.79  | 0.300 | 3.360 | 0.057 | 1497 | 1166 | 1067 | 32.5     |
| 28 | C    | IA       | 3.55 | 0.56 | 0.43 | 1.79  | 0.300 | 3.360 | 0.057 | 1490 | 1166 | 1067 | 29.0     |
| 29 | C    | IA       | 3.55 | 0.56 | 0.43 | 1.79  | 0.300 | 3.360 | 0.057 | 1354 | 1166 | 1067 | 17.0     |
| 30 | C    | IA       | 3.55 | 0.56 | 0.43 | 1.79  | 0.300 | 3.360 | 0.057 | 1347 | 1166 | 1067 | 16.5     |
| 31 | D    | Non-Std. | 3.38 | 1.32 | 0.49 | 3.40  | 0.040 | 4.410 | 0.110 | 1448 | 1212 | 1088 | 31.5     |
| 32 | D    | Non-Std. | 3.38 | 1.32 | 0.49 | 3.40  | 0.040 | 4.410 | 0.110 | 1360 | 1212 | 1088 | 27.0     |
| 33 | D    | Non-Std. | 1.79 | 1.46 | 0.8  | 35.02 | 0.740 | 1.460 | 0.430 | 1487 | 1337 | 1285 | 16.5     |
| 34 | D    | Non-Std. | 1.79 | 1.46 | 0.8  | 35.02 | 0.740 | 1.460 | 0.430 | 1432 | 1337 | 1285 | 11.0     |
| 35 | D    | IA       | 3.37 | 0.67 | 0.54 | 2.88  | 0.020 | 4.180 | 0.110 | 1448 | 1219 | 1095 | 26.0     |
| 36 | D    | IA       | 3.37 | 0.67 | 0.54 | 2.88  | 0.020 | 4.180 | 0.110 | 1368 | 1219 | 1095 | 20.5     |
| 37 | D    | IA       | 3.37 | 0.67 | 0.54 | 2.88  | 0.020 | 4.180 | 0.110 | 1324 | 1219 | 1095 | 18.5     |
| 38 | D    | ID       | 3.11 | 1.68 | 0.61 | 7.12  | 0.200 | 5.380 | 0.160 | 1484 | 1227 | 1150 | 30.0     |
| 39 | D    | ID       | 3.11 | 1.68 | 0.61 | 7.12  | 0.200 | 5.380 | 0.160 | 1365 | 1227 | 1150 | 16.0     |
| 40 | D    | IIB      | 3.14 | 0.66 | 0.98 | 15.32 | 0.810 | 1.030 | 0.860 | 1434 | 1222 | 1216 | 29.0     |
| 41 | D    | IIB      | 3.14 | 0.66 | 0.98 | 15.32 | 0.810 | 1.030 | 0.860 | 1354 | 1222 | 1216 | 8.0      |

*All temperatures are Celsius. PT = Pour Temperature; LT = Liquidus Temperature; ST = Solidus Temperature. SL = Spiral Length.*

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**Fig. H2. Line fit plot of spiral length vs. pour temperature for 1–4% Cr irons.**

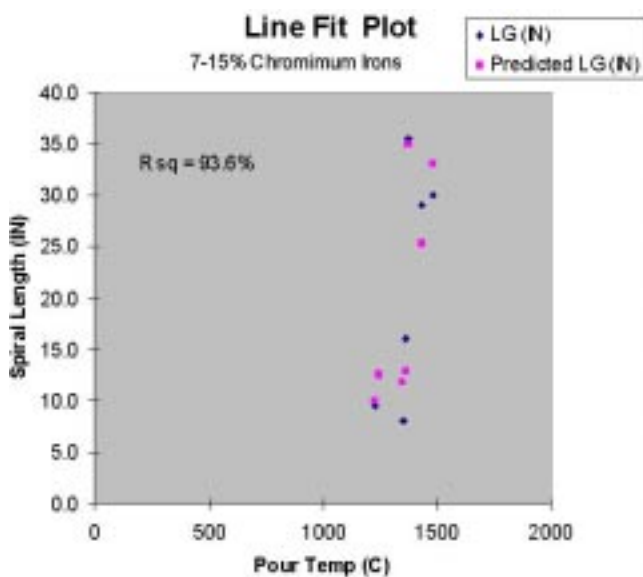
**Table H2**  
**Regression Analysis for 1–4% Chromium Irons**

| Regression Statistics:    |                        |                     |            |          |                |           |
|---------------------------|------------------------|---------------------|------------|----------|----------------|-----------|
| Multiple R                | 0.975406901            |                     |            |          |                |           |
| R Square                  | 0.951418623            |                     |            |          |                |           |
| Adjusted R Square         | 0.924919691            |                     |            |          |                |           |
| Standard Error            | 2.422201474            |                     |            |          |                |           |
| Observations              | 18                     |                     |            |          |                |           |
| ANOVA:                    |                        |                     |            |          |                |           |
|                           | df                     | SS                  | MS         | F        | Significance F |           |
| Regression                | 6                      | 1263.906785         | 210.6511   | 35.90404 | 1.33687E-06    |           |
| Residual                  | 11                     | 64.53765981         | 5.86706    |          |                |           |
| Total                     | 17                     | 1328.444444         |            |          |                |           |
|                           | Coefficients           | Standard Error      | t Stat     | P-value  | Lower 95%      | Upper 95% |
| Intercept                 | −381.2112442           | 53.74314735         | −7.09321   | 2.01E-05 | −499.4991738   | −262.9233 |
| Mn                        | 93.36435486            | 161.2915209         | 0.578855   | 0.574349 | −261.6360686   | 448.36478 |
| Ni                        | 21.44902999            | 6.95930093          | 3.082067   | 0.010433 | 6.131704168    | 36.766356 |
| Si                        | 16.69373903            | 25.04360801         | 0.666587   | 0.518773 | −38.42689844   | 71.814377 |
| C                         | 47.6417703             | 10.50663311         | 4.534447   | 0.000852 | 24.51681505    | 70.766726 |
| Cr                        | −15.41612408           | 22.05887485         | −0.69886   | 0.499148 | −63.96740482   | 33.135157 |
| PT (C)                    | 0.106609928            | 0.009911325         | 10.75637   | 3.55E-07 | 0.084795238    | 0.1284246 |
| RESIDUAL OUTPUT:          |                        |                     |            |          |                |           |
| Observation               | Predicted Length (in.) | Actual Length (in.) | Difference |          |                |           |
| 1                         | 31.7                   | 32.5                | 0.8        |          |                |           |
| 2                         | 31.0                   | 29.0                | −2.0       |          |                |           |
| 3                         | 16.6                   | 17.0                | 0.4        |          |                |           |
| 4                         | 15.7                   | 16.5                | 0.8        |          |                |           |
| 5                         | 29.2                   | 29.0                | −0.2       |          |                |           |
| 6                         | 42.7                   | 42.5                | −0.2       |          |                |           |
| 7                         | 40.6                   | 41.0                | 0.4        |          |                |           |
| 8                         | 29.4                   | 32.0                | 2.6        |          |                |           |
| 9                         | 28.7                   | 31.5                | 2.8        |          |                |           |
| 10                        | 15.0                   | 12.0                | −3.0       |          |                |           |
| 11                        | 14.4                   | 12.0                | −2.4       |          |                |           |
| 12                        | 28.7                   | 29.0                | 0.3        |          |                |           |
| 13                        | 30.8                   | 30.5                | −0.3       |          |                |           |
| 14                        | 29.0                   | 26.0                | −3.0       |          |                |           |
| 15                        | 20.4                   | 20.5                | 0.1        |          |                |           |
| 16                        | 15.7                   | 18.5                | 2.8        |          |                |           |
| 17                        | 34.0                   | 31.5                | −2.5       |          |                |           |
| 18                        | 24.5                   | 27.0                | 2.5        |          |                |           |
| (Continued on next page.) |                        |                     |            |          |                |           |

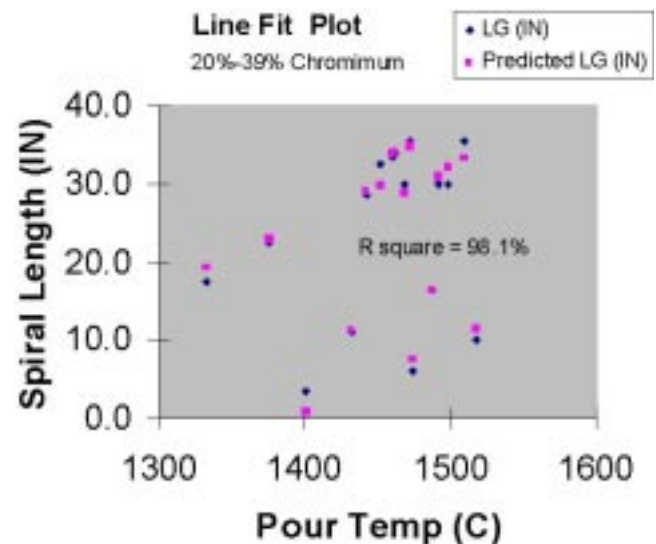
(Continued on next page.)

**Table H3**  
**Regression Analysis for 7–15% Chromium**

| Regression Statistics: |                        |                     |            |         |                |             |
|------------------------|------------------------|---------------------|------------|---------|----------------|-------------|
| Multiple R             | 0.967534544            |                     |            |         |                |             |
| R Square               | 0.936123093            |                     |            |         |                |             |
| Adjusted R Square      | 65535                  |                     |            |         |                |             |
| Standard Error         | 0                      |                     |            |         |                |             |
| Observations           | 7                      |                     |            |         |                |             |
| ANOVA:                 |                        |                     |            |         |                |             |
|                        | df                     | SS                  | MS         | F       | Significance F |             |
| Regression             | 6                      | 699.9526098         | 116.6588   | 0       | #NUM!          |             |
| Residual               | 0                      | 47.76167593         | 65535      |         |                |             |
| Total                  | 6                      | 747.7142857         |            |         |                |             |
|                        | Coefficients           | Standard Error      | t Stat     | P-value | Lower 95%      | Upper 95%   |
| Intercept              | −161.0945519           | 0                   | 65535      | #NUM!   | −161.0945519   | −161.094552 |
| Mn                     | −4.513060057           | 0                   | 65535      | #NUM!   | −4.513060057   | −4.51306006 |
| Ni                     | −14.25149167           | 0                   | 65535      | #NUM!   | −14.25149167   | −14.2514917 |
| Si                     | 82.24104235            | 0                   | 65535      | #NUM!   | 82.24104235    | 82.2410423  |
| C                      | −48.80118561           | 0                   | 65535      | #NUM!   | −48.80118561   | −48.8011856 |
| Cr                     | 4.800048584            | 0                   | 65535      | #NUM!   | 4.800048584    | 4.80004858  |
| PT (C)                 | 0.169853857            | 0                   | 65535      | #NUM!   | 0.169853857    | 0.16985386  |
| RESIDUAL OUTPUT:       |                        |                     |            |         |                |             |
| Observation            | Predicted Length (in.) | Actual Length (in.) | Difference |         |                |             |
| 1                      | 33.1                   | 30.0                | −3.1       |         |                |             |
| 2                      | 12.9                   | 16.0                | 3.1        |         |                |             |
| 3                      | 35.0                   | 35.5                | 0.5        |         |                |             |
| 4                      | 12.6                   | 12.5                | −0.1       |         |                |             |
| 5                      | 9.9                    | 9.5                 | −0.4       |         |                |             |
| 6                      | 25.2                   | 29.0                | 3.8        |         |                |             |
| 7                      | 11.8                   | 8.0                 | −3.8       |         |                |             |



**Fig. H3.** Line fit plot of spiral length vs. pour temperature for 7–15% Cr irons.



**Fig. H4.** Line fit plot of spiral length vs. pour temperature for 20–39% Cr irons.

**Table H4**  
**Regression Analysis for 20–39% Chromium**

|                               |       |             |  |  |  |  |
|-------------------------------|-------|-------------|--|--|--|--|
| <b>Regression Statistics:</b> |       |             |  |  |  |  |
| Multiple R                    | ..... | 0.990691769 |  |  |  |  |
| R Square                      | ..... | 0.981470182 |  |  |  |  |
| Adjusted R Square             | ..... | 0.96911697  |  |  |  |  |
| Standard Error                | ..... | 1.957821908 |  |  |  |  |
| Observations                  | ..... | 16          |  |  |  |  |

|               |           |             |           |          |                       |  |
|---------------|-----------|-------------|-----------|----------|-----------------------|--|
| <b>ANOVA:</b> |           |             |           |          |                       |  |
|               | <b>df</b> | <b>SS</b>   | <b>MS</b> | <b>F</b> | <b>Significance F</b> |  |
| Regression    | 6         | 1827.236775 | 304.5395  | 79.45061 | 2.78228E-07           |  |
| Residual      | 9         | 34.49759962 | 3.833067  |          |                       |  |
| Total         | 15        | 1861.734375 |           |          |                       |  |

|           |                     |                       |               |                |                  |                  |
|-----------|---------------------|-----------------------|---------------|----------------|------------------|------------------|
|           | <b>Coefficients</b> | <b>Standard Error</b> | <b>t Stat</b> | <b>P-value</b> | <b>Lower 95%</b> | <b>Upper 95%</b> |
| Intercept | –84.3507223         | 460.3151644           | –0.18325      | 0.858667       | –1125.656762     | 956.95532        |
| Mn        | –11.09433998        | 109.5800039           | –0.10124      | 0.921576       | –258.9817196     | 236.79304        |
| Ni        | 14.6372801          | 79.061096             | 0.185139      | 0.857225       | –164.2114808     | 193.48604        |
| Si        | –17.37965317        | 28.98845112           | –0.59954      | 0.563601       | –82.95613549     | 48.196829        |
| C         | 6.8966554           | 133.2645095           | 0.051752      | 0.959857       | –294.5688391     | 308.36215        |
| Cr        | –0.908973242        | 4.004658811           | –0.22698      | 0.825512       | –9.96814776      | 8.1502013        |
| PT (C)    | 0.089710204         | 0.014332594           | 6.259174      | 0.000148       | 0.0572876        | 0.1221328        |

|                         |                               |                            |                   |
|-------------------------|-------------------------------|----------------------------|-------------------|
| <b>RESIDUAL OUTPUT:</b> |                               |                            |                   |
| <b>Observation</b>      | <b>Predicted Length (in.)</b> | <b>Actual Length (in.)</b> | <b>Difference</b> |
| 1                       | 11.4                          | 10.0                       | –1.4              |
| 2                       | 7.4                           | 6.0                        | –1.4              |
| 3                       | 0.8                           | 3.5                        | 2.7               |
| 4                       | 16.2                          | 16.5                       | 0.3               |
| 5                       | 11.3                          | 11.0                       | –0.3              |
| 6                       | 32.3                          | 30.0                       | –2.3              |
| 7                       | 33.2                          | 35.5                       | 2.3               |
| 8                       | 29.8                          | 32.5                       | 2.7               |
| 9                       | 29.0                          | 28.5                       | –0.5              |
| 10                      | 23.0                          | 22.5                       | –0.5              |
| 11                      | 19.2                          | 17.5                       | –1.7              |
| 12                      | 28.9                          | 30.0                       | 1.1               |
| 13                      | 31.1                          | 30.0                       | –1.1              |
| 14                      | 34.0                          | 33.5                       | –0.5              |
| 15                      | 34.1                          | 34.0                       | –0.1              |
| 16                      | 35.0                          | 35.5                       | 0.5               |

**Table H5**  
**Relationships for Fluidity**

| <b>Fluidity (in.) =</b> | <b>Intercept</b> | <b>+%Mn × Coefficient</b> | <b>+%Ni × Coefficient</b> | <b>+%Si × Coefficient</b> | <b>+%C × Coefficient</b> | <b>+Cr × Coefficient</b> | <b>+ PT (C) × Coefficient</b> |
|-------------------------|------------------|---------------------------|---------------------------|---------------------------|--------------------------|--------------------------|-------------------------------|
| 1–4% Cr alloys          | –381.2           | 93.4                      | 21.4                      | 16.7                      | 47.60                    | –15.4                    | 0.107                         |
| 7–15% Cr alloys         | –161.1           | –4.5                      | –14.2                     | 82.2                      | –48.80                   | 4.8                      | 0.170                         |
| 20–39% Cr alloys        | –84.3            | –11.1                     | 14.6                      | –17.4                     | 6.90                     | –0.909                   | 0.090                         |

# Glossary

## Terms in Glossary

*(First, and sometimes second, citing in parentheses)*

|                                                                  |                                                                                                                               |
|------------------------------------------------------------------|-------------------------------------------------------------------------------------------------------------------------------|
| Alpha Ferrite ( <i>see</i> : Delta Ferrite)                      | Hypoeutectic (Ch. 2)                                                                                                          |
| Ball and Tumbling Mills (Ch. 5)                                  | Ledeburite (Ch. 1)                                                                                                            |
| Carbon Equivalent (Ch. 2)                                        | Malleable Iron (Ch. 5)                                                                                                        |
| Cementite (Ch. 1)                                                | Metastable Phase Diagrams (No citing exactly) ( <i>see</i> Ch. 5 under Transformation Diagrams; only says “Metastable Phase”) |
| Columnar or Equiaxed Growth (Ch. 2)                              | Oxidizing and Reducing Environments (Ch. 9)                                                                                   |
| Constituents and Phases (Ch. 1)                                  | Pearlite (Ch. 1)                                                                                                              |
| Corrosion (Ch. 1 and Ch. 9)                                      | Peritectic Reactions (Ch. 2)                                                                                                  |
| Delta and Alpha Ferrite (Ch. 2, Ch. 3)                           | Phase Diagram (Ch. 2)                                                                                                         |
| Differential Thermal Analysis ( <i>see</i> : DTA Thermogram)     | Plastic Strain ( <i>see</i> : Dislocation Movement)                                                                           |
| Dislocation Movement (Ch. 5)                                     | Potentiostat (Ch. 9)                                                                                                          |
| DTA Thermogram (Ch. 2)                                           | Primary Carbides ( <i>see</i> : Proeutectic Austenite)                                                                        |
| Elastic Strain ( <i>see</i> : Dislocation Movement)              | Proeutectic Austenite and Primary Carbides (Ch. 2)                                                                            |
| Electron Microprobe (Ch. 2)                                      | Scanning Electron Microscopy (Ch. 1)                                                                                          |
| Epsilon Carbide (Ch. 5)                                          | Stainless Steels (Ch. 1, Ch. 5)                                                                                               |
| Equiaxed Growth ( <i>see</i> : Columnar or Equiaxed Growth)      | Stoichiometry (Ch. 2)                                                                                                         |
| Equilibrium Phase Diagram ( <i>see</i> Metastable Phase Diagram) | Superheat (Ch. 1)                                                                                                             |
| Eutectic (Ch. 1)                                                 | Transformation Diagram ( <i>see</i> : Phase Diagram)                                                                          |
| Eutectic Colony (Ch. 2)                                          | Tumbling Mills ( <i>see</i> : Ball and Tumbling Mills)                                                                        |
| Gibbs Free Energy                                                | Wavelength Dispersive Spectrometry (WDS) Analysis ( <i>see</i> : Electron Microprobe)                                         |
| Hypereutectic ( <i>see</i> : Hypoeutectic) (Ch. 2)               |                                                                                                                               |

See page 213 for complete Glossary →



## Glossary

**Alpha Ferrite** (see: Delta Ferrite)

**Ball and Tumbling Mills:** A tumbling mill is essentially a large barrel, rotating on a horizontal axis. Ball mills use the momentum of steel grinding balls (up to 14 kg) to crush the ore. In autogenous (AG) mills, size reduction is achieved by the momentum of the larger pieces of ore, while in semiautogenous (SAG) mills, this action is assisted by a small volume fraction of balls. AG and SAG mills tend to be very large, commonly 10 m in diameter. In grinding moisture-sensitive materials, such as cement, grinding is performed dry, but in most mineral operations, the ore is introduced as a slurry.

**Carbon Equivalent (CE):** Metallurgists have long known that certain alloying elements can mimic the effect of carbon in graphitic irons. This effect can be seen graphically in Fig. 1, provided by Neumann (1968), showing the effect of various elements on the change in C solubility. As the P or Si content is increased, the solubility of C in the molten iron decreases. Conversely, the exact opposite effect occurs when V and Cr are added. (We could also say that P and Si increase the activity coefficient of C in iron while V and Cr decrease its activity coefficient.) In short, the effect of P and Si can be taken into account for graphitic irons by the simple formula:  $CE = \%C + 1/3\% Si + 1/4\% P$ . For Fe-C iron alloys that solidify white or carburitic:  $CE = \%C + 1/9\% Si + 1/3.5\% P$  (A. Moore, "Carbon Equivalent of White Cast Irons," *AFS Cast Metals Research Journal*, vol. 8, pp. 15-19, 1972.) We can also give another equation for the effect of V and Cr. In this case, the equation is:  $CE = \%C - 1/10\% V - 1/20\% Cr$ . The whole point is that the CE formula must be used carefully and appropriately. For example, due to solidification dynamics, the CE formula for the effect of Si is not applicable to low-alloy AR cast irons (basically mixtures of iron and carbon) and, at high Cr content, the effect of Si is completely masked by Cr's countereffect.

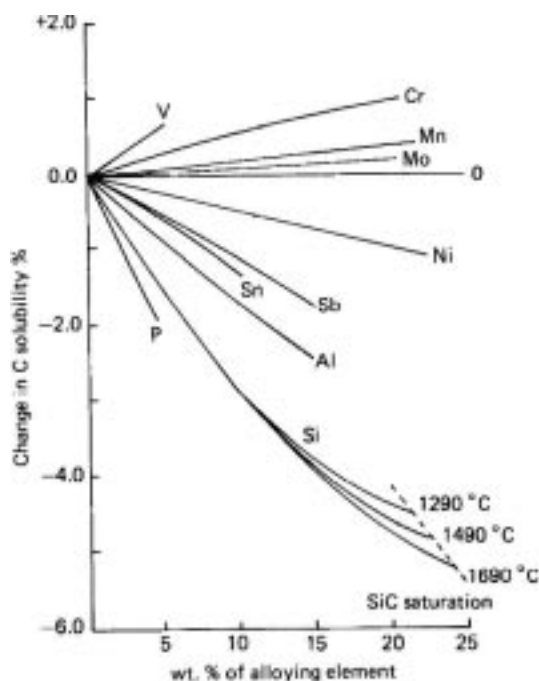


Fig. 1. Influence of various alloying elements on the solubility of carbon in liquid iron. (Courtesy of Newman, 1968)

**Cementite:** This is the common metallurgical name for the eutectic  $M_3C$  carbide that forms in low-alloy AR cast irons, as a product of the eutectic solidification reaction of  $L \Rightarrow \gamma Fe + M_3C$ . Pearlitic, chill-cast and Ni-Hards 1 and 2 irons all contain this carbide. Some sources give the formula for this carbide as  $Fe_3C$ . However, common alloying elements (e.g., Cr, Mo and Mn) are soluble within the carbide crystalline structure and, therefore, the correct formula is  $M_3C$ , where M indicates a mixture of Fe and the other soluble alloying elements.

**Columnar or Equiaxed Growth:** Using Fig. 2, we can take a quick look at how the macrostructure is formed during solidification. As the hot metal flows into the mold, solidification occurs almost immediately along the cold sand. If the flow is brisk, these growing dendrites will break off and move into the melt as "seeds" for later growth. If the melt is somewhat cool, these detached small chunks of frozen metal, will stay frozen and initiate the growth of eutectic colonies or other dendrites. Since these new growths are not constrained, they can grow pretty much anyway they want to. In the case of eutectic cell growth, the equiaxed growth will result in many fine eutectic grains that are actually more cylindrical than spherical (see e.g., Fig. 2.4). If there is high superheat, the little frozen chunks will remelt and the initial growth from the mold wall will continue inward, toward the center of the casting. This type of unidirectional growth is called *columnar* growth, while unconstrained free growth is called *equiaxed* growth. Besides superheat, alloy composition and the use of tiny bits of solid particles (e.g., inoculates) can hinder or promote columnar or equiaxed growth in a casting.

**Constituents and Phases:** Classically, constituent means "an essential part," and this is a good place to start the definition. Metallurgically, *constituent* is used to denote a structure within a material that has a distinct morphology, crystal structure and chemistry. It is a rigorous definition that is rarely met in many alloys. For example, the presence of ferrite grains within an austenitic matrix does not imply that the ferrite is a separate constituent. That is, ferrite does not have a distinct morphology or distinct chemical differences over that of austenite. In this case, ferrite is noted to be a different phase within the iron but not as a distinct constituent; whereas, carbides are classed as distinct constituents within the iron

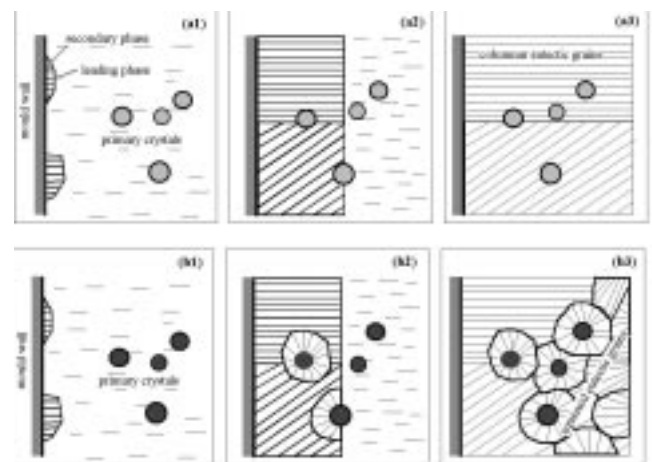


Fig. 2. Columnar vs. equiaxed growth mechanisms. (Courtesy of Springer-Verlag)

since they are of distinct morphology, crystal structure and chemistry. This is analogous to composite terminology, where the fiber reinforcement is considered a distinct constituent from the epoxy matrix. This definition may seem a little pedantic, but it helps to clarify the importance of having distinct constituents versus distinct phases within your iron.

**Corrosion:** Corrosion is the process of gradual wearing away of a material by chemical means. For ferrous-based materials, it is the disassociation of the metallic bonds between the Fe atoms, allowing the release of free Fe atoms into the chemical solution. A by-product of this reaction is the chemical reaction of these free Fe atoms with O, to form rust. Rust is a complex compound of  $\text{Fe}_x\text{O}_y$  atoms in non-stoichiometric parts.

**Delta ( $\delta$ ) and Alpha ( $\alpha$ ) Ferrite:** When pure iron solidifies at 1539C (2802F), it forms as a ferrite body centered cubic (BCC) crystalline structure. As the pure iron cools, it remains ferritic down to 1400C (2552F). At this temperature, it transforms to austenite ( $\gamma$ ) (i.e.,  $\delta\text{Fe} \Rightarrow \gamma\text{Fe}$ ), which remains stable down to 910C (1670F), where it transforms back to ferrite (i.e.,  $\gamma\text{Fe} \Rightarrow \alpha\text{Fe}$ ). Historically, when these transformations were detected, it was not known that the crystalline structure between 1539 and 1400C was the same as the one below 910C. Hence, they got two different names:  $\delta\text{Fe}$  for the high temperature form (liquid  $\Rightarrow \delta\text{Fe}$ ) and  $\alpha\text{Fe}$  for the low temperature form ( $\gamma\text{Fe} \Rightarrow \alpha\text{Fe}$ ). When the composition of the alloy is such that the initial solidification reaction of liquid  $\Rightarrow \delta\text{Fe}$  is all that occurs as the alloy cools to room temperature, then the correct name for the ferrite phase is delta ( $\delta\text{Fe}$ ) and not alpha ( $\alpha\text{Fe}$ ) ferrite.

**Differential Thermal Analysis** (see: DTA Thermogram)

**Dislocation Movement:** There are many fine metallurgy books that provide extensive information about dislocations and dislocation movement (e.g., R.E. Reed-Hill, 1973). In the meantime, think of dislocations as small movements within the crystal lattice that allow the accommodation of plastic strain. For example, when a load or force is applied to a material, the material can behave *elastically* or *plastically*. If loaded *elastically*, the material will rebound or revert to its original shape after the load is applied. If loaded *plastically*, the material will “plastically” flow and will not revert to its original shape. This mechanism of plastic flow is seen in every hardness measurement. For plastic flow to occur, there must be dislocation movement. Dislocations can be defined in many different ways. One way to view a dislocation is to think of one atom slipping over another atom within the crystal lattice. That is, there is no such thing as a dislocation; it is only a name given to the physical process of one atom slipping past another atom. Naturally enough, depending on the crystal structure of the material, it is easier for an atom to slip in some directions than in others. How easily atoms slip past each other determines the yield strength and, of course, the hardness of the material. If the crystal lattice is highly strained, as it is in high-C martensite, it becomes exceedingly difficult for dislocation movement (i.e., for atoms to slip). Since the crystal structure is more resistant to plastic flow, the indenter will create a smaller crater for any given load and a higher hardness will be measured. For more detailed information on hardness measurements, see Appendix D.

**DTA Thermogram:** Differential thermal analysis (DTA) is a very effective tool for understanding how metals (and organic

compounds) melt and solidify. It is called *differential* because temperatures are measured simultaneously from two small crucibles. One crucible will contain the alloy of interest and the other crucible will be empty or contain a material of known thermophysical properties. Both crucibles are placed within a programmable, radiant furnace and heated or cooled at a controlled rate. The differential temperature between the two crucibles is then plotted as the alloy undergoes melting or freezing. With some care, one can measure the heat of fusion that is released during the liquid  $\Leftrightarrow$  solid phase change. More importantly, we can determine the exact start and stop temperatures of all phase change reactions (cooling curve data, which simply measures the temperature of the metal as it cools from liquid  $\Rightarrow$  solid, does not typically yield such good numbers). For example, using a DTA instrument we can create a graphical record of the iron as it solidifies, pinpointing the austenite start temperature and the start of the eutectic reaction. Such data allows a quick determination of whether the iron is a short or long freezing alloy.

**Elastic Strain** (see: Dislocation Movement)

**Electron Microprobe:** This is another very useful form of electron microscopy that is similar to SEM but with enhanced capabilities. The electron microprobe allows the direct chemical analysis of micron-sized phases with extremely good accuracy. The electron beam is focused through a series of electromagnetic lenses (just like the SEM) onto the phase under analysis. X-rays then cascade from the bombarded phase and are filtered and measured via wavelength dispersive spectrometers (WDS). Hence, electron microprobe analysis is also known as WDS analysis. Quantitative work is performed by analyzing known standards (e.g.,  $\text{Cr}_7\text{C}_3$  carbide or NBS standards) and then hitting your unknown phase with the electron probe.

**Epsilon ( $\epsilon$ ) Carbide:** This carbide is extremely small and only forms during the tempering (below 200C/392F) of medium- to high-C martensite (C content in excess of 0.2%). It has an unusual carbide structure of hexagonal close-packed and an unusual chemistry of  $\text{Fe}_{2.4}\text{C}$  (it is not cementite). It can only be seen by using very high-resolution electron microscopy. In fact, for many years there existed some controversy over its structure and chemical composition. As mentioned, during martensitic decomposition at low temperatures, epsilon carbide will form. This process allows the body-centered tetragonal (BCT) crystal lattice to relax or become less distorted. The end result of this process is that medium- and high-C martensite is replaced with a mixture of epsilon carbide and lower carbon martensite that is of lower hardness but much tougher.

**Equiaxed Growth** (see: Columnar or Equiaxed Growth)

**Equilibrium Phase Diagram** (see: Metastable Phase Diagram)

**Eutectic Colony:** One definition of a colony is a group of things having a common characteristic. In this case, a common characteristic of a eutectic colony is its growth direction. Once solidification has started (whether near the wall of the mold or on some small fragment of solid) growth will occur in a preferred direction, determined, in part, by the thermal gradient and, in part, by the nature of the carbide crystal. All carbide crystals grow faster along some directions than along other directions. In the case of a eutectic colony, we have cooperative or coupled growth between the austenite and the carbide phase in one preferred direction. Each colony originates from

a single nucleus and is typically comprised of only a few carbide and austenite crystals. The size of a eutectic colony is influenced by the number of colonies competing for the available liquid. If we have many, many nucleation sites (and hence, many small growing eutectic colonies), the iron will have a network of fine eutectic carbides. Conversely, if only a few eutectic colonies form (e.g., slow cooling favors growth over nucleation), they will be large, resulting in large eutectic carbides.

**Eutectic:** When an alloy solidifies through a eutectic reaction, it goes from a liquid to two solids. In the case of AR irons, one of the solids is always a metallic carbide, while the other solid is the iron phase (austenite or ferrite). A nice characteristic of eutectic alloys is that solidification occurs over a very tight temperature range (giving good fluidity) and, for any given alloy family, at the lowest temperature. This is why the word *eutectic*—coming from the Greek word meaning *most fusible* or *most liquid*—was chosen to represent this type of metallurgical behavior. That is, eutectic alloys are extremely fluid or most liquid during the casting process.

**Gibbs Free Energy:** This is a thermodynamic term referring to the energy that is released when chemical compounds react and form other compounds. A concise definition is: “For any spontaneous process in which the products and reactants are at constant temperature and pressure and, in which no work is done other than against pressure, the Gibbs free energy will decrease,  $\Delta G < 0$ .” Thermodynamically, this implies that, as the Gibbs free energy becomes more and more negative, a higher potential exists that the reaction will occur spontaneously. Additionally, the products of high negative Gibbs free energy reactions are more thermodynamically stable.

**Hypoeutectic and Hypereutectic:** The prefixes *hypo* and *hyper* are Greek, meaning under and over, respectively. In chemistry, hypo means the lowest while hyper means the highest. Consequently, hypoeutectic irons and hypereutectic irons have compositions that are respectively, under and over the eutectic composition. If we base our definition on just the carbon content of the iron and assume that other alloying elements are being held constant (e.g., Cr), a hypoeutectic iron will have a lower carbon content than the eutectic and a hypereutectic iron will have a higher carbon content. Besides these differences in chemistry, a hypoeutectic iron starts solidification with the formation of ferrous dendrites, followed by a eutectic reaction while a hypereutectic iron starts solidification with the formation of large, primary metallic carbides and ends by solidifying as a eutectic.

**Ledeburite:** A metallurgical term for the eutectic mixture of austenite and  $M_3C$  carbide found in low-alloy AR cast irons.

**Malleable Iron:** The key requirement in the manufacturing of malleable cast iron is keeping the chromium content low. Typically, it is recommended that the Cr content never exceed 0.1%. Nevertheless, the iron must be induced to solidify via the carbide eutectic and end up with a microstructure that is composed of  $M_3C$  ( $\sim Fe_3C$ ) carbide and austenite. Upon cooling to room temperature, the austenite will readily transform into pearlite, due to the iron’s low hardenability (typically alloy contents are 1% Si, 0.5% Mn, and a broad range of S and P). Since the carbide is very unstable (low Cr content), the  $M_3C$  carbide will decompose into graphite upon heating to temperatures above 900C (1652F). One of the nice things

about malleable iron is that a ferritic outer layer can be formed by heat treating the iron in a decarburizing atmosphere. In this process (called a combined decarburization and graphitization heat treatment), the carbides are decomposed into graphite and the outer layer of the iron is decarburized. Upon cooling to room temperature, it will have a microstructure of soft ferrite on the surface and a mixture of austenite, pearlite and martensite within the core (all depending on the iron’s alloy content). The main reasons for selecting malleable irons are low cost and ease of machining. As can be expected, malleable irons have limitations in section size, damping capacity, impact resistance and hot and cold cracking and warpage. For more information, see Elliott (1988).

**Metastable Phase Diagrams:** Scientists enjoy equilibrium, i.e., a constant unchanging reference state. To obtain a true equilibrium state is often nearly impossible for many materials. For example, at equilibrium, all Fe-C alloys should be composed of ferrite and graphite. This can happen over a geological time scale but, typically, the steel or iron rusts into nothingness first. In the construction of the metastable liquidus surface, the irons were cooled at a fixed rate. This fixed rate is several orders of magnitude faster than what true equilibrium would require and hence, the term *metastable* is used to describe the surface. All in all, this suits the need, since all castings are cooled at some dynamic, non-equilibrium rate. The obvious question is: How does the metastable phase diagram differ from the equilibrium diagram? A good example of where metastable and equilibrium phase diagrams do differ is in the binary Fe-C system. That is, the metastable diagram predicts carbide formation (fast cooling rate) while the equilibrium diagram predicts the formation of graphite (slow cooling). In the case of the ternary Fe-Cr-C system, there exists an equilibrium liquidus surface and a metastable liquidus surface. We prefer to use only the metastable surface, since the equilibrium phase diagram is not applicable to castings. A good review of the ternary Fe-Cr-C system is given by Rivlin (1984).

**Oxidizing and Reducing Environments:** When an element is *oxidized*, it loses electrons and when it is *reduced*, it gains electrons. Most acids, by their very nature, will oxidize metals. When the term oxidization is used, it is referring to the disassociation of metal atoms from the parent substrate. From a corrosion viewpoint, acids can be further defined as oxidizing or reducing in nature. When an oxidizing acid is added to an environment, it tends to promote oxidization of the metal while increasing the availability of free oxygen within the liquid. Conversely, when a reducing acid is used, there is little change or sometimes a decrease in the availability of free oxygen. For phases and constituents containing large amounts of Cr, an oxidizing environment will cause Cr atoms (and likewise, other metallic atoms, e.g., Fe, Mn, Si, etc. ) to disassociate from the parent substrate and become positively charged (it loses electrons). The positively charged Cr atoms are now highly reactive and will eagerly bond with the available oxygen. This reaction forms a protective layer of  $Cr_2O_3$  over the surface of the parent material and hinders further oxidization or, in corrosion parlance, the material becomes passivated. Examples of strong oxidizing acids are  $HNO_3$  and  $H_2O_2$ . The addition of these acids to an environment has little effect on high-Cr ferrous materials, since the oxidized material will readily passivate. In more natural environments, Cr atoms are naturally oxidized through electrochemical means or through the presence of salts or other

simple oxidizing chemicals and, since free O is readily available in most environments, a protective  $\text{Cr}_2\text{O}_3$  film can easily form. With reducing acids such as HCl or  $\text{H}_2\text{SO}_4$ , metallic (Fe, Cr, etc.) atoms are still disassociated from the matrix, but without the oxygen component, the protective film never forms, and the phase or constituent continues to dissolve away.

**Pearlite:** A eutectoid mixture of alpha ferrite ( $\alpha\text{Fe}$ ) and  $\text{M}_3\text{C}$ . Pearlite is a decomposition product of iron that has been alloyed with some fraction of carbon. After solidification, as the iron cools through the eutectoid temperature, pearlite is formed via the following reaction:  $\gamma\text{Fe} \Rightarrow \alpha\text{Fe} + \text{M}_3\text{C}$ . Its morphology is often characterized as a laminar structure of alternating plates of ferrite and carbide. Optically it has the appearance of mother-of-pearl.

**Peritectic Reactions:** The Fe-Cr-C system has two peritectic reactions. In peritectic reactions a liquid and a solid phase combine to solidify as two solid phases. A method to locate these peritectic reactions is to look for the meeting point of three liquidus surfaces. One such intersection is formed by the  $\delta$ ,  $\gamma$ , and  $\text{M}_7\text{C}_3$  liquidus surfaces. This peritectic point (fixed in composition and temperature) is reached by ferritic irons, for example, corrosion-resistant irons having a base composition of 2% C, 30% Cr. As this iron solidifies, initially  $\delta$  forms. The remaining liquid then moves toward the peritectic point at 2.8% C and 32% Cr at 1289C (2352F). Since this is the intersection of three liquidus surfaces, all three phases will coexist in the solid form. The actual reaction is:  $\text{L} + \alpha\delta\text{Fe} \Rightarrow \gamma\text{Fe} + \text{M}_7\text{C}_3$ . At the other end of the phase diagram, we find a low-Cr peritectic point being the intersection of the  $\gamma$ ,  $\text{M}_7\text{C}_3$ , and  $\text{M}_3\text{C}$  liquidus surfaces. If the iron is a Ni-Hard 4 containing 3% C, 9% Cr, and an Si <1.5%, then it will undergo the peritectic reaction located at 4.2% C, 9.7% Cr at 1183C (2161F), where  $\text{L} + \text{M}_7\text{C}_3 \Rightarrow \gamma\text{Fe} + \text{M}_3\text{C}$ . Peritectic reactions are often incomplete because a solid phase must transform into another solid phase via a diffusion-dependent route. Typically, as the iron continues to cool, it drops through the peritectic point into the eutectic valley.

**Phase Diagram:** Understanding phase diagrams is not that difficult. Think of phase diagrams as simple maps to describe the iron or alloy behavior in terms of temperature and composition. We generally check and/or measure the properties of an iron at room-temperature properties; phase diagrams, however,

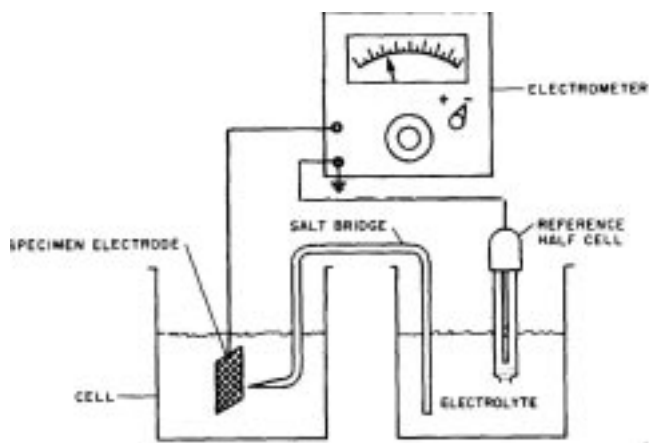


Fig. 3. Schematic of a simple potentiostat.

tell us what phase the iron is in at any temperature or at any composition. The diagrams are constructed on the basis of the metallurgical analysis of dozens and dozens of alloys. For example, to determine the various phases that are present in the iron as a function of temperature, one procedure is to melt a small sample and cool it very slowly to the desired temperature and then quench it in water. The sample is then metallurgically examined to determine the phases present and their relative volume fractions. This procedure is repeated for different temperatures and many compositions until enough data is gathered to construct the phase diagram. So why guess? If you really want to know what your iron is doing, during solidification or heat treating, consult a phase diagram. You will avoid doing a lot of work that others have already done. (See also: Equilibrium and Metastable Phase Diagrams)

**Plastic Strain** (see: Dislocation Movement)

**Potentiostat:** This is an electrical device used to determine the corrosion characteristics of metals. A schematic of a simple potentiostat is shown in Fig. 3. By maintaining an electrical bridge to a neutral potential, the electrometer can then measure any changes in the flow of electrons, from the sample into the liquid or vice versa. More advanced units can vary the potential between the sample and the neutral cell, while measuring the current flow. In such a fashion, the onset of passivation can be measured. That is, once the material has passivated the current flow will drop appreciably due to the formation of the protective and insulating passive film. The potential at which this film forms determines whether the material will passivate or corrode under oxidation conditions. Data from such experiments are plotted as potential (mV) versus current density ( $\text{nA}/\text{cm}^2$ ) and are known as potentiodynamic plots. More general information can be found in corrosion textbooks or in ASTM G-3, Conventions Applicable to Electrochemical Measurements in Corrosion Testing.

**Primary Carbides** (see: Proeutectic Austenite)

**Proeutectic Austenite and Primary Carbides:** Metallurgical terminology is sometimes not as straightforward as it may appear. In *hypoeutectic* irons, *proeutectic austenite* is the first phase to solidify and it occurs before the eutectic, hence the prefix *pro*. In *hypereutectic* irons, *primary carbides* are the first phase to solidify and their formation occurs before the eutectic. It would seem logical that primary carbides might also be known as proeutectic carbides. However, the terminology proeutectic austenite and primary carbides prevails.

**Reducing Environment** (see: Oxidizing and Reducing Environments)

**Scanning Electron Microscopy (SEM):** A wonderful tool that has brought material science (what used to be called *metallurgy*) out of the dark ages. If one thinks of how an optical microscope works—bending and focusing optical waves (which are made up of photons) through a series of glass lenses—we can use the same process for other wave forms. The electron microscope simply uses electrons instead of photons as their light source, and uses magnetic lenses to bend and focus the electron waves. Electrons are quite a bit larger than photons and their characteristic wavelength is much, much smaller; consequently, electrons make much smaller waves than pho-

tons. The smaller wavelengths yield greater resolution (i.e., magnifications in excess of 100,000X), whereas optical microscopes have a hard time resolving images above 2000x. The SEM is completely electronic (the first commercial models did not reach the marketplace until around 1965) and imaging output can be electronically processed in a number of ways, yielding many different types of microstructural information.

**Stainless Steels:** Stainless steels were so named at the time of their discovery in 1912, when Harry Brearly, a British metallurgist, found that he was unable to etch steel alloys containing 12% Cr using a conventional dilute nitric acid etchant (e.g., a solution of 2% HNO<sub>3</sub> in an alcohol base). At that time, etching was more commonly referred to as “staining,” and so he named the steels *stainless* because of their resistance to staining.

**Stoichiometry:** This is the statement of the quantitative chemical composition of a substance. In general, stoichiometry implies a law of definite proportions. This is very applicable to the carbides found in AR irons. For example, the M<sub>3</sub>C carbide will unvaryingly be composed of three metal atoms to one carbon atom. Likewise, for the higher alloyed irons containing large amounts of chromium (say greater than 8% Cr), the carbide is composed of seven metal atoms to three carbon atoms, yielding a stoichiometry of M<sub>7</sub>C<sub>3</sub>. Hence, the use of the word “stoichiometry” to help us describe the chemical composition of the various carbides found in AR irons is accurate and illuminating.

**Superheat:** It is impossible to cast an alloy without some superheat.

To be exact, the difference between the alloy’s pouring temperature and the temperature at which solidification starts is termed the superheat. For example, if a hypoeutectic alloy starts to solidify at 1371C (2500F) and finishes some 111C (200F) later at 1260C (2300F), the superheat calculation would be based on the 1371C (2500F) temperature and not the lower temperature. Hence, given a pouring temperature of 1538C (2800F), the superheat would be calculated as: 2800 – 2500 = 167C (300F).

**Transformation Diagram** (see: Phase Diagram)

**Tumbling Mills** (see: Ball and Tumbling Mills)

**Wavelength Dispersive Spectrometry (WDS) Analysis** (see: Electron Microprobe)

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