To my wife and Andrew

Cast Iron: Physical and Engineering Properties

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First published in 1960 (as *Physical and Engineering Properties of Cast Iron*) by the British Cast Iron Research Association

Second Edition 1976
Reprinted 1978

ISBN 0 408 70933 2

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Library of Congress Cataloging in Publication Data

Angus, Harold T
Cast iron: physical and engineering properties.

Published in 1960 under title: Physical and engineering properties of cast iron.
Includes index.
1. Cast-iron.
TA474.A6 1975 620.1 7 75–31960
ISBN 0–408–70933–2

Printed in Great Britain by
Butler & Tanner Ltd, Frome and London

Preface

The first edition of this book, published in 1960 by the British Cast Iron Research Association, originated in a private collection of data which the author had acquired during a fairly wide range of industrial experience and from problems submitted to the British Cast Iron Research Association. There is probably no single item recorded that has not been required at some time or another, and it appeared probable, therefore, that such a collection of data and experience might be of value to others in the foundry and engineering industries. The task of assembling, correcting and checking would have been impossible as a spare time occupation without the unstinted assistance of the staffs of the Advisory Services and Investigations Departments of the Association under Mr. S. W. Palmer and Mr. I. C. H. Hughes, and of course the Information Department under Mr. M. R. Thibault. Much of the work quoted is taken directly from published and unpublished researches and reviews by members of all three departments of the Association.

There must be few industrial countries of the world which have not contributed notably to some section of the data quoted. The lists of references, however, are not intended to be exhaustive, and in many cases reference is made to more extensive works from which such lists can be obtained. *The Gray and Ductile Iron Castings Handbook*, published by the Gray and Ductile Iron Founders' Society of America, contains very full lists of references on many subjects. The data were originally presented primarily in Imperial units, pound and ton units (2240 lb) and in some cases in metric (c.g.s. units). The Système Internationale (SI) units were introduced after the main text had been prepared and although SI units have not yet been adopted universally—and notably not in the USA—nevertheless their use is likely to increase and where possible in tables, text and figures in this book, the SI units have been added to the conventional units. However, as the whole of the data printed in the book was originally carried out and published in Imperial or metric units it would have been wrong to issue the values in SI units only as it would make reference back to the original work almost impossible.

The manner in which the information is arranged is due to the author's

personal taste and has often depended upon the way in which a particular subject has been presented as an industrial problem.

In many cases a good deal of background investigation detail has been given with the data so that the reader can form his own conclusions on the validity of the data in relation to his own circumstances.

No attempt has been made to provide a detailed treatment of all the subjects dealt with, but an attempt has been made to summarise the latest information and research work and then to make reference to the various authorities for fuller information.

In particular, the papers of Morrogh on the metallurgy of cast iron, and of Gilbert on the mechanical properties of cast iron, must be studied by anyone wishing to be fully acquainted with these subjects.

Dr. H. E. Merritt in his book on gears quoted Robertson Buchanan, who in 1808 submitted that '... a rule though not absolutely perfect is better in all cases than to have no guide whatever' and added that '... clothing rules in algebraic garb may give them a deceptive appearance of truth'. The author agrees with both these statements.

The author is particularly indebted to the late Mr. G. R. Woodward and Mr. M. R. Thibault for the constant detailed help given in preparing this work for publication, and to Miss D. S. Drake, Mr. A. S. Morrison and Miss T. M. Jones for their untiring efforts in locating and assembling data and information from the most inaccessible sources.

Finally, it should be said that this book would not have been written without the constant encouragement and support of the late Dr. J. G. Pearce, Director of the British Cast Iron Research Association, 1924 to 1958, Dr. H. Morrogh, the present Director, and the Council of the Association.

H.T.A.

SI Units (Système Internationale d'Unités)

This is an international system of units which has been adopted with slight modifications in the legislation of most countries that use the metric system and it is the system being adopted by the UK.

SI units form a coherent system, i.e. the product or quotient of any two unit quantities is the unit of the resultant quantity, for example:

Unit area = unit length × unit length
Unit velocity = unit length ÷ unit time
Unit force = unit mass × unit acceleration

The SI base and derived units which are important to the contents of this book are given below, together with further information on some of the units and a table of conversion factors.

Some base and derived SI units

Quantity	*Name of SI unit*	*Symbol*	*Expressed in terms of SI base or derived units*	*Imperial* ft lb s	*CGS and other metric* cm g s
Base units:					
Length	metre	m		foot	centimetre
Mass	kilogram	kg		pound	gram
Time	second	s		second	second
Derived units:					
Force	newton	N	$1\ N = 1\ kg\ m/s^2$	pound force (lbf)	kilogram force (kgf or kp)
				poundal (pdl)	gram force (gf)
					dyne (dyn)
Work, energy, quantity of heat	joule	J	$1\ J = 1\ Nm$	horsepower hour (hp h)	kilowatt hour (kWh)
Power	watt	W	$1\ W = 1\ J/s$	foot pound force (ft. lbf)	kilogram force metre per second (kgf. m/s)

Conversion Factors From British Units to SI, CGS and Other Metric Units

Physical quantity	*Unit*	*Symbol*	*Definition in SI*	*CGS and other metric units*
Length	thou	10^{-3} in	25·4 μm	0·002 54 cm
	inch	1 in	25·4 mm	2·54 cm
	foot	1 ft	0·304 8 m	30·48 cm
	yard	1 yd	0·914 4 m	91·44 cm
	mile	1 mile	1·609 34 km	1·609 km
Area	square inch	1 in^2	645·16 mm^2	6·452 cm^2
	square foot	1 ft^2	0·092 903 m^2	92·9 cm^2
	square mile	1 $mile^2$	2·589 99 km^2	2·59 km^2
Volume	cubic inch	1 in^3	16·387 1 cm^3	16·387 cm^3
	cubic foot	1 ft^3	28·316 8 dm^3	28 317 cm^3
	UK gallon	1 gal	4·546 092 dm^3	4·546 litres
Mass	pound	1 lb	0·453 592 37 kg	453·6 g
	ton	1 ton	1016·05 kg	1 016·1 kg
Velocity	mile per hour	1 mph	1·609 34 km. h^{-1}	1·609 km/h
	foot per second	1 ft/s	0·304 8 m. s^{-1}	0·304 8 m/s
Density	pound per cubic inch	1 lb/in^3	27·6799 Mg. m^{-3}	27·7 g/cm^3
	pound per cubic foot	1 lb/ft^3	16·018 5 kg. m^{-3}	0·016 02 g/cm^3

Physical quantity	*Unit*	*Symbol*	*Definition in SI*	*CGS and other metric units*
Force	pound force	1 lbf	4·448 22 N	453·59 gf 0·453 6 kgf
	ton force	1 tonf	9·964 02 kN	1 016·05 kgf
Pressure	pound per square inch	1 lbf/in^2	6894·76 N.m^{-2}	0·070 3 kgf/cm^2
	ton per square inch	1 tonf/in^2	15·444 3 MN.m^{-2} = 1·544 43 daN/mm^2 = 1·544 43 kN/cm^2 = 1·544 43 hbar	157·5 kgf/cm^2 1·575 kgf/mm^2
Energy	British thermal unit	1 Btu	1·055 06 kJ	252 cal 0·252 kcal
	therm (100 000 Btu)	1 Th	105·506 MJ	25 200 kcal
	foot pound force	1 ft.lbf	1·355 82 J	0·138 3 kgf.m
Power	horsepower	1 hp (33 000 ft.lb/min)	0·745 7 kW	1·014 metric hp
	foot pound second	1 ft.lbf/s	1·355 82 W	0·138 3 kgf.m/s
Heat flow	Btu per hour	1 Btu/h	0·293 071 W	252 cal/h 0·252 kcal/h
Specific energy	Btu per pound	1 Btu/lb	2 326 J.kg^{-1}	0·555 kcal/kg
Calorific value	Btu per cubic foot (30 inHg, 60°F, wet)	1 Btu/ft^3	37·258 9 kJ.m^{-3}	9·547 kcal/m^3 (0°C, 760 mmHg, dry) 9·487 kcal/m^3 (0°C, 760 mmHg, wet)
Specific heat and specific entropy		1 Btu/lb.°F	4 186·8 J.kg^{-1}K^{-1}	1 cal/g.°C
Thermal conductivity		1 Btu/ft.°F.h	1·730 73 W.m^{-1}K^{-1}	0·004 134 cal/cm.s.°C*
Thermal conductance		1 Btu/ft^2.h.°F	5·678 W.m^{-2}K^{-1}	0·000 136 cal/cm^2.s.°C

* 1 cal/cm.s.°C ≡ 1 cal/cm^2/sec/°C/cm

Prefix	*Symbol*		*Factor by which the unit is multiplied*
nano	n	=	10^{-9}
micro	μ	=	10^{-6}
milli	m	=	10^{-3}
deca	da	=	10
kilo	k	=	10^{3}
mega	M	=	10^{6}
giga	G	=	10^{9}

The unit of mass (or weight) is an artificial one, the kilogram, which is equal to the mass of the international prototype in the custody of the Bureau International des Poids et Mesures, Sèvres near Paris.

The SI unit of force is the newton, which is the force required to give unit acceleration (1 m/s^2) to unit mass (1 kg). Thus 1 newton = 1 kg m/s^2 = 10^5 dyn = 0·224 8 lbf = 0·101 97 kgf or as Newton might have appreciated = weight of an apple (about 3·6 oz) approximately.

The Imperial pound force (*lbf*) gives to 1 pound mass the gravitational acceleration of 32·174 ft/s^2. 1 lbf = 32·174 pdl = 4·448 N = 453·59 gf.

The poundal (*pdl*) gives to 1 pound mass an acceleration of 1 ft/s^2 = 0·138 3 N.

In metric technical units the kilogram force (kgf or kp) will give to 1 kg (mass) the gravitational acceleration of 9·806 7 m/s^2 and in equivalent to 2·205 lbf.

The dyne (*dyn*) gives to 1 gram mass an acceleration of 1 cm/s^2 = $2{\cdot}248 \times 10^{-6}$ lbf.

The SI unit of pressure and stress is the pascal (Pa) which is the special name for newton per square metre (N/m^2). This unit is very small for much practical use (1 N/m^2 = 0·000 145 lbf/in^2) and newtons per square millimetre which is the same value as meganewtons per square metre (MN/m^2) may be used. 1 MN/m^2 = 1 N/mm^2 = 145 lbf/in^2.

The standard atmospheric pressure (760 mmHg at 0°C)
1 atm = 1·033 2 kgf/cm^2 = 1 013 250 dyn/cm^2 = 14·696 lbf/in^2 = 1·013 bar = 29·92 inHg at 0°C = 33·9 ft water.

The technical atmosphere (736 mmHg at 0°C)
1 at = 1 kgf/cm^2 = 14·223 lbf/in^2 = 0·968 standard atmosphere (atm).

The bar (750·1 mmHg at 0°C)
1 bar = 1·019 7 kgf/cm^2 = 14·504 lbf/in^2 = 0·986 9 standard atmosphere (atm).

The torr = 1 mmHg at 0°C.

The hardness Vickers, 1(*HV*) = 1 kgf/mm^2 = 1 422·3 lbf/in^2 = 9·806 7 N/mm^2.

1 lbf/in^2 = 7·03 kgf/mm^2 $\times 10^{-4}$ = 6895 N/m^2.

1 tonf/in^2 = 1·574 9 kgf/mm^2 = 15·44 MN/m^2 = 15·44 N/mm^2.

Distance, roughness
1 in = 25·400 mm.
40 micro inches (39·4μin) = 1 micron = 1 μ metre.
1 thousandth of an inch (10^{-3} in) = 0·025 4 mm = 25·4 μ metres.

Heat
1 British thermal unit (Btu) = 252 cal = 0·252 kcal = 1055 joule = 778 ft.lbf = 107·6 kgf.m.
1 therm (Th) = 100 000 Btu = 25 200 kcal = 105·506 MJ.
1 kilocalorie (kcal) = 3·968 Btu = 4 186·8 joule = 3088 ft.lbf = 426·8 kgf.m.
1 joule = 1 watt second = 0·000 948 Btu = 0·737 5 ft.lbf = 0·101 9 kgf.m.

*Thermal conductivity**
1 cal/cm.s. °C = 418·68 W/m.K = 0·8063 Btu/ft^2/sec/°F/inch thickness.
1 W/m.K = 0·002 39 cal/cm.s.°C.

Heat flow rate
1 watt (W) = 0·238 8 calories per second (cal/s) = 3·412 Btu/h.

* 1 cal/cm.s. °C ≡ 1 cal/cm^2/cm/sec/°C

1

Constitution and structure of cast iron

Cast iron is essentially an iron–carbon alloy containing other important elements such as silicon, manganese, sulphur and phosphorus, which modify the structure and properties of the resulting alloy markedly.

The basic equilibrium diagram of the pure iron–carbon alloys is shown in *Figure 1*.

Melting point

Pure iron melts at 1535°C[1] and the eutectic of iron and carbon occurs with a carbon content between 4·2 and 4·3 per cent with a melting point of 1153°C. The eutectic melting temperature for cast iron varies slightly with silicon content and usually falls between 1135°C and 1150°C.

Additions of silicon lower the percentage of carbon in the eutectic by about 0·33 per cent for each 1 per cent of silicon added, so that with 1 per cent of silicon the eutectic would occur at approximately 3·97 per cent of carbon.

Additions of phosphorus affect the eutectic carbon value to about the same amount as silicon, so that with an iron containing 1 per cent silicon and 1 per cent phosphorus, the eutectic would occur with approximately 3·6 per cent carbon.

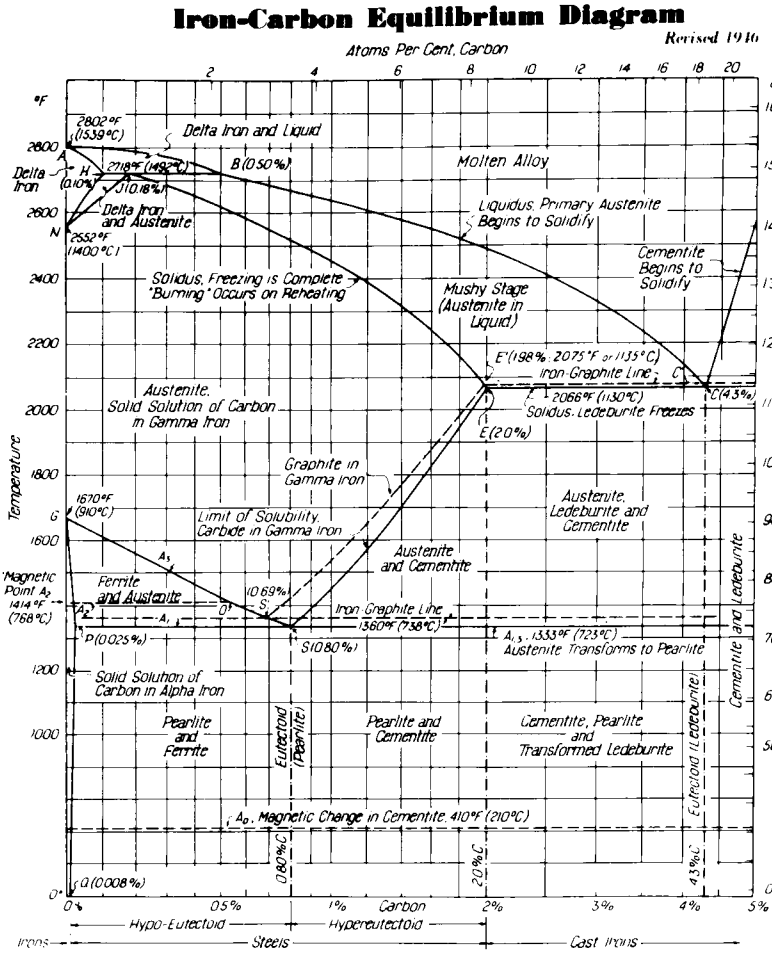
The other elements in the amounts normally present have a much smaller effect on eutectic carbon value, but have other significant effects referred to later.

For each 1 per cent addition the following figures have been given[2]: silicon up to 5·5 per cent lowers the carbon content of the eutectic by about 0·31 per cent; phosphorus up to 3·0 per cent lowers the carbon content by about 0·33 per cent; nickel up to 8·0 per cent lowers the carbon content by about 0·05 per cent; copper up to 3·8 per cent lowers the carbon content by about 0·075 per cent; chromium up to 9·0 per cent increases the carbon content by about 0·06 per cent; manganese up to 2·5 per cent increases the carbon content of the eutectic by about 0·03 per cent; vanadium up to 3·4 per cent increases the carbon content by about 0·1 per cent.

A cast iron may have a composition which makes it hypo-eutectic, eutectic,

or hypereutectic with regard to carbon, and its position in this respect is of importance, since not merely does the initial solidification point (liquidus) depend upon it, but also the manner of solidification and thereby also the structure.

The carbon contents of most commercial irons fall between 2·5 and 4·0 per cent. The iron–carbon diagram shows only the iron carbide/austenite (ledeburite) system in the high carbon region, and such irons, when commercially produced, are known as white or chilled irons and are hard, brittle and unmachinable. Grey irons, however, solidify with the formation of graphite instead of iron carbide, and are reasonably strong and machinable. An intermediate form known as mottled iron may also be formed, in which both



Principal points and lines, especially *G-S-E-J*, adjusted from positions in former edition of *Metal Progress* data sheet, to agree with diagram prepared for the subcommittee on phase diagrams, Kent R. Van Horn, chairman, John S. Marsh and F. N. Rhines, for inclusion in the 1947 ● Metals Handbook. Point *Q* according to J. H. Whiteley, *Journal* of the Iron and Steel Institute, 1936.

Figure 1. Iron–carbon equilibrium diagram (reproduced by courtesy of Metal Progress)

graphite and iron carbide are formed during solidification; such irons are harder and less readily machinable than the grey irons.

Carbon equivalent value

An iron may solidify grey, white or mottled, depending mainly upon the eutectic value and the rate of cooling, but before attempting to evaluate the effect of cooling rate, the effect of composition must be simplified. The most usual simplification employed in this country is the carbon equivalent value[3] (CE) in which

$$\text{CE} = \text{Total Carbon } \% + \frac{\text{Silicon } \% + \text{Phosphorus } \%}{3}$$

and it is this value, rather than the actual total carbon content, which applies with commercial irons when using the iron–carbon diagram in *Figure 1*.

The carbon equivalent value tells immediately whether an iron is hypo-eutectic or hypereutectic, and by how much. Thus, an iron with total carbon 3·2 per cent, silicon 2·0 per cent and phosphorus 0·4 per cent, has a CE value of 4·0 and is hypo-eutectic. An iron with total carbon 3·2 per cent, silicon 2·0 per cent and phosphorus 1·3 per cent, has a CE value of 4·3 and is eutectic. An iron with total carbon 3·2 per cent, silicon 2·9 per cent and phosphorus 1·3 per cent has a CE value of 4·6 and is hypereutectic.

The solubility of carbon in molten iron is dependent upon temperature and other elements present notably silicon and phosphorus[4], and *Figure 2* demonstrates this for carbon content and carbon equivalent value. Irons which are hypereutectic (carbon equivalent more than 4·2–4·3) will precipitate kish graphite on solidification with normal cooling rates.

In general, the lower the carbon equivalent value, the greater is the tendency for an iron to solidify white or mottled.

The solidification period determines whether a given iron becomes grey, mottled, or white, and the form of the graphite is very little altered by the subsequent cooling period.

Melting point (solidus)

Many workers have shown that silicon causes an increase in eutectic temperature in iron–carbon–silicon alloys. A rise from 1150°C to 1165°C when silicon rises from 0 to 3·0 per cent has been indicated[5]. For commercial cast irons, melting usually begins between 1135°C and 1150°C. Where phosphorus is present (above about 0·06 per cent) a ternary eutectic of iron, carbon and phosphorus is formed which melts at about 930°C.

Liquidus

The effect of silicon and phosphorus upon the liquidus temperature differs slightly from their effect on the carbon content of the eutectic and in the

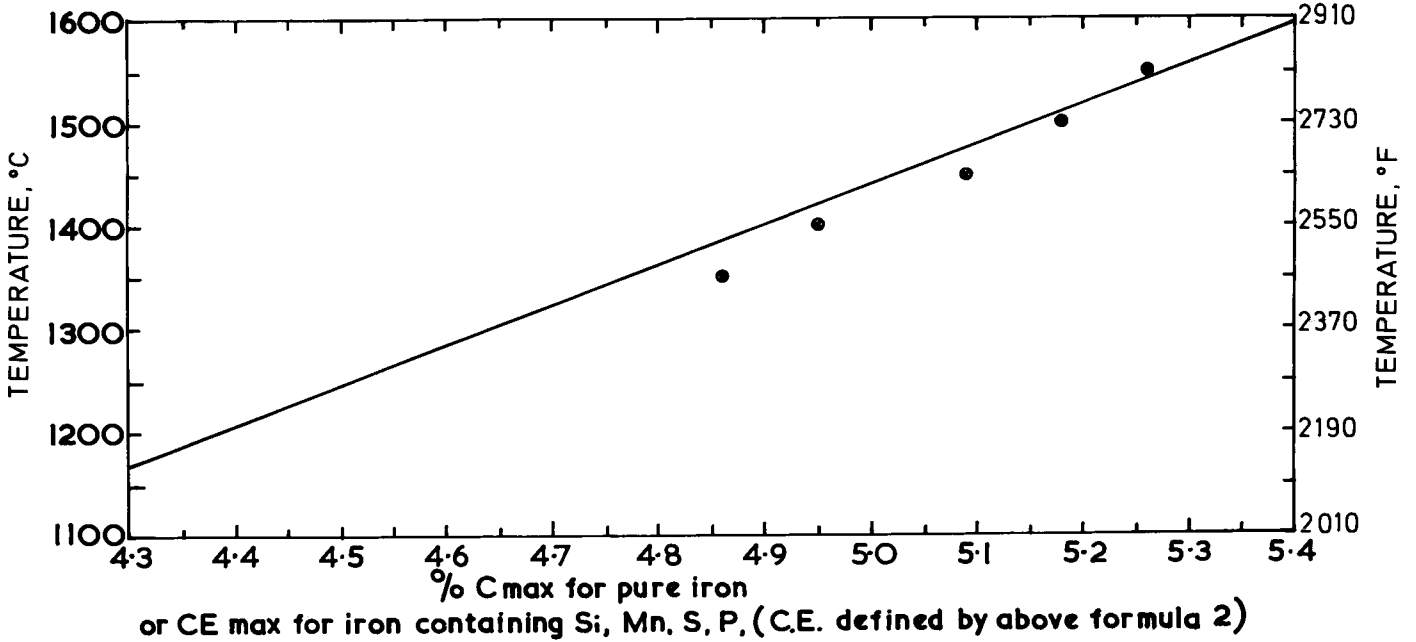


Figure 2. Solubility of carbon in molten iron: (1) $C_{max} = 1{\cdot}30 + 0{\cdot}002\,57\,T°C - 0{\cdot}31 \times \%\,Si - 0{\cdot}33 \times \%P + 0{\cdot}027 \times \%\,Mn - 0{\cdot}4 \times \%S$ *(2)* $CE_{max} = C + 0{\cdot}31 \times \%\,Si + 0{\cdot}33 \times \%\,P - 0{\cdot}027 \times \%\,Mn + 0{\cdot}4 \times \%S = 1{\cdot}30 + 0{\cdot}002\,57\,T°C$. $\otimes$ *= BCIRA experimental points (10 cwt HF furnace melt of 3·86% Si, 0·36% Mn, 0·03% S & P)*

normal grey and whiteheart malleable cast iron range a good indication of the liquidus temperature of hypo-eutectic irons is given by the formula[5]:

$$1669°C - 124\left(\text{Carbon }\% + \frac{\text{Phosphorus }\%}{2} + \frac{\text{Silicon }\%}{4}\right)$$

For very hypo-eutectic irons, such as blackheart malleable, where the carbon equivalent (liquidus) value

$$\left(\text{Carbon }\% + \frac{\text{Phosphorus }\%}{2} + \frac{\text{Silicon }\%}{4}\right)$$

is less than 3·2, the formula:

$$1570°C - 95\left(\text{Carbon }\% + \frac{\text{Phosphorus }\%}{2} + \frac{\text{Silicon }\%}{4}\right)$$

is a close approximation[6]. Copper up to 2 per cent and chromium up to 4·5 per cent have little effect on the liquidus temperature. Nickel lowers the liquidus by about 5°C for each 1 per cent addition up to 5 per cent.

White irons

If the iron is sufficiently below the eutectic value, has a very low silicon content, contains appreciable quantities of carbide, stabilising elements, or if the cooling rate is sufficiently rapid, the formation of graphite does not occur, solidification taking place first by the formation of austenite dendrites, and the interdendritic areas will remain high in carbon and solidify as a eutectic of iron carbide and austenite, known as ledeburite. As the solid casting cools down, at a temperature of 720–750°C, the austenite, which is γ iron with carbon in solution, transforms to α iron, in which carbon is substantially insoluble. Carbon is then rejected as iron carbide in the form of lamellae alternative with lamellae of α iron to form a matrix of pearlite. The initially formed iron carbide, however, remains unchanged.

The structure of a white iron at room temperature, therefore, consists of primary dendrites of pearlite with interdendritic areas of transformed ledeburite, which is a eutectic structure of iron carbide and pearlite, the pearlite areas having been formed from the original austenite (*Figure 20*). In some cases, the austenite produced during the final eutectic solidification of iron carbide and austenite is deposited on the already existing austenite dendrites, and the interdendritic area will be predominantly carbide. The final structure will then consist of dendrites of pearlite surrounded by cementite containing very little transformed eutectic pearlite[7].

Grey irons

In hypo-eutectic irons, the first phase to be deposited from the melt is austenite—a solid solution of carbon in γ iron which crystallises in the form

of dendrites, the length and pattern of which depend upon the temperature gradients. The austenite dendrites continue to grow and the remaining liquid increases in carbon content until the eutectic temperature and concentration is reached. Solidification begins from a number of centres, each with an approximately spherical crystallisation front, and simultaneous deposition of graphite and austenite occurs. Ultimately, these eutectic cells meet and consume the remaining liquid. The austenite of the dendrites and that of the eutectic becomes continuous and the structure appears as a dispersion of graphite flakes in a matrix of austenite. After solidification, the eutectic cell structure and the random orientation of the primary austenite dendrites

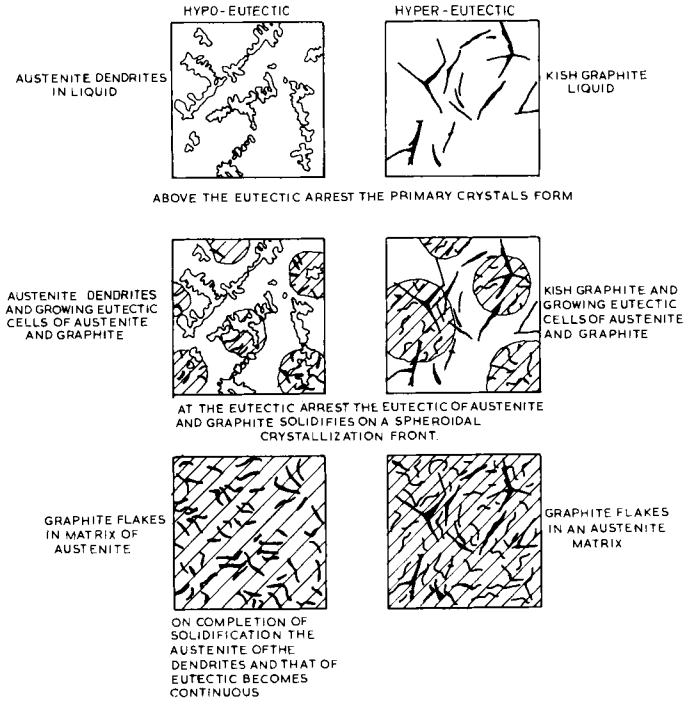


Figure 3. Diagrams illustrating the solidification of hypo-eutectic and hypereutectic irons

throughout the structure are essentially complete and cannot be subsequently modified on cooling or by subsequent treatment. The general pattern of solidification is illustrated in *Figure 3*[8]. Only in strongly hypo-eutectic iron is the austenite dendrite pattern clearly visible.

Hypereutectic irons solidify by the direct formation of graphite from the melt in the form of kish. Owing to its low relative density, kish tends to rise to the surface of the melt, but when entrapped by the metal it generally appears in the microstructure as characteristic long straight flakes or as lumpy flakes in rapidly cooled sections. This phase is deposited over a range of temperature, starting at the liquidus surface until the carbon content of the remaining melt is at the eutectic concentration when a simultaneous crystal-

lisation of graphite and austenite occurs. This eutectic graphite in hypereutectic irons as a general rule occurs in flake form on a finer scale than the primary kish.

The austenite, on cooling through the critical temperature range (in the neighbourhood of 720–830°C) transforms to pearlite (*Figures 1* and *6*). The products of this transformation depend upon the rate of cooling and composition of the material, but under normal conditions will consist either of pearlite, or of graphite and ferrite, or else mixtures of all three. Where the eutectoid transformation produces graphite and ferrite, the graphite is deposited on the already existing eutectic graphite flakes. The formation of ferrite is most likely to occur with slow rates of cooling, with high silicon contents that reduce the stability of iron carbide (*Figure 10*), or with high carbon equivalent values, or with fine undercooled graphite (*Figure 13*). The formation of fully pearlitic structures is more likely to occur with moderately rapid cooling rates or low carbon equivalent values.

If the iron is very rapidly cooled through the critical temperature, e.g. by oil-quenching, the transformation of austenite to pearlite or ferrite may be suppressed, and a hard matrix of martensite may be formed as in *Figure 14*. This treatment may be given deliberately where increased hardness is called for, and is dealt with under the section on heat-treatment.

Graphite

The form and distribution in which graphite is deposited depends on numerous factors, such as melting temperature, nucleation, cooling speed etc., but the basic forms are flake graphite, aggregate or temper carbon, and nodules or spheroids.

The ASTM has laid down a convenient standard for graphite (ASTM A247–47 and 67), defining shape, size and distribution, and this has been extended by the International Standards Organisation in ISO Recommendation 945, to cover additional forms. These are reproduced in *Figures 4a*, *4b*, *5a*, *5b*, *5c* and *5d*. The basic nomenclature and grading for flake graphite follows the generally accepted ASTM pattern.

FLAKE GRAPHITE

Flake graphite may form as (i) kish, which occurs in hypereutectic irons and may be present as straight and undistorted flakes (ASTM Type C; ISO Form I, distribution C) or very thick lumpy flakes; (ii) normal flake (Type A or Form I, distribution A); (iii) rosette (Type B or Form I, distribution B) *see Figure 12*; (iv) undercooled (Type D, Form I, distribution D) *see Figure 13*.

The undercooled or Type D graphite is normally associated with rapid rates of cooling and is particularly common in thin sections, but can be produced by other means (e.g. additions of titanium).

The rosette pattern is typical of slightly less rapid cooling and is not uncommon in the more rapidly cooled surfaces of castings with sections of $\frac{3}{8}$ in

or more. Both undercooled and rosette graphite are frequently associated with ferrite rather than pearlite, since the larger surface area of graphite

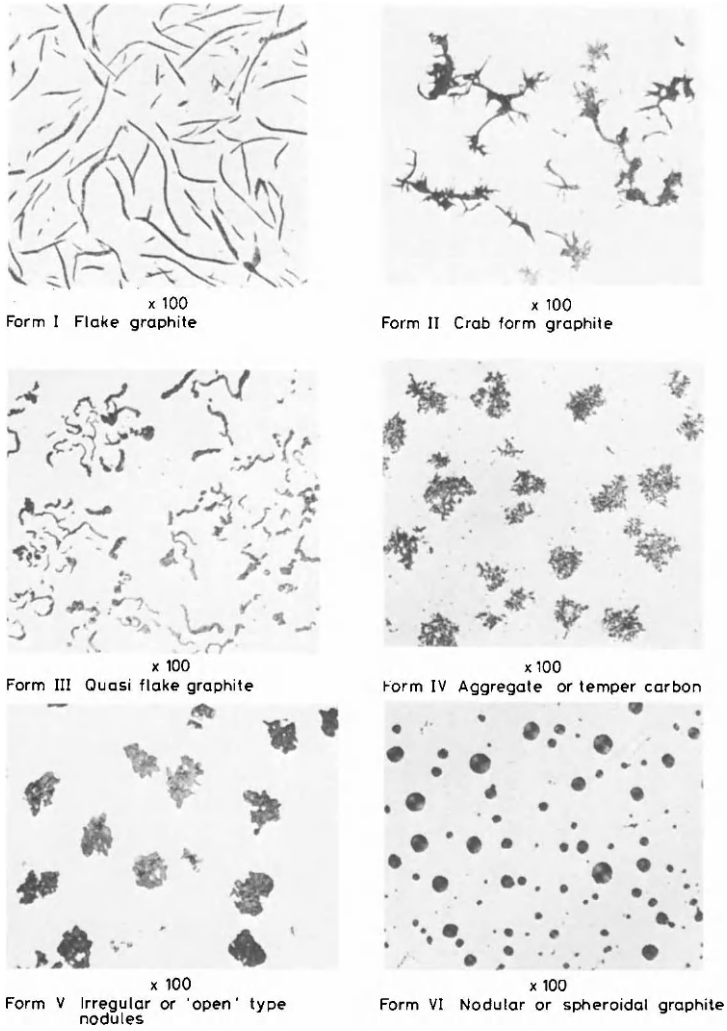


Figure 4(a). Graphite type or form (distribution A). ISO/R 945-1969(E)

available reduces the distance required for carbon diffusion during transformation from γ to α iron in the critical temperature range.

When ferrite is associated with random or coarse flake graphite, the iron is usually hypereutectic or the rate of cooling in the mould has been very slow.

Widmanstätten and mesh-form graphite (*Figures 34–37*) can occur when lead or tellurium are present and profoundly weakens the iron. When grey iron passes through the γ to α critical transformation range, graphite may be

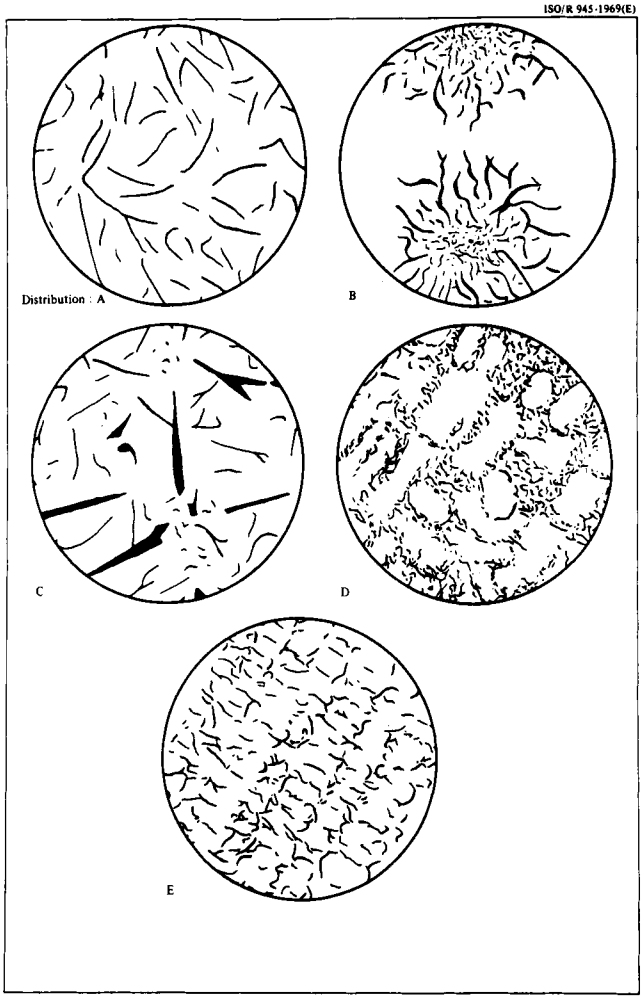
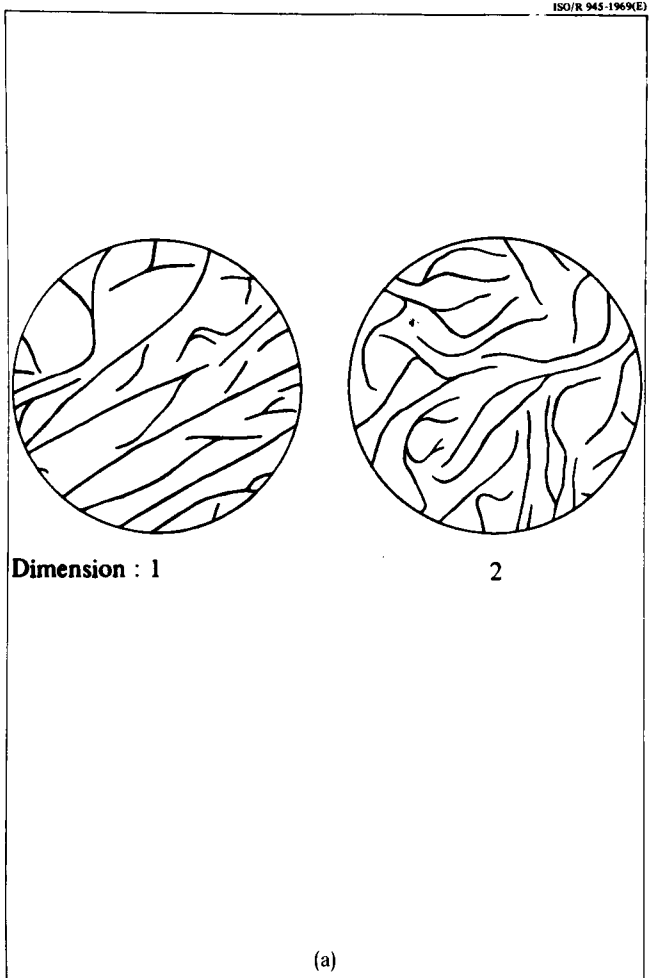
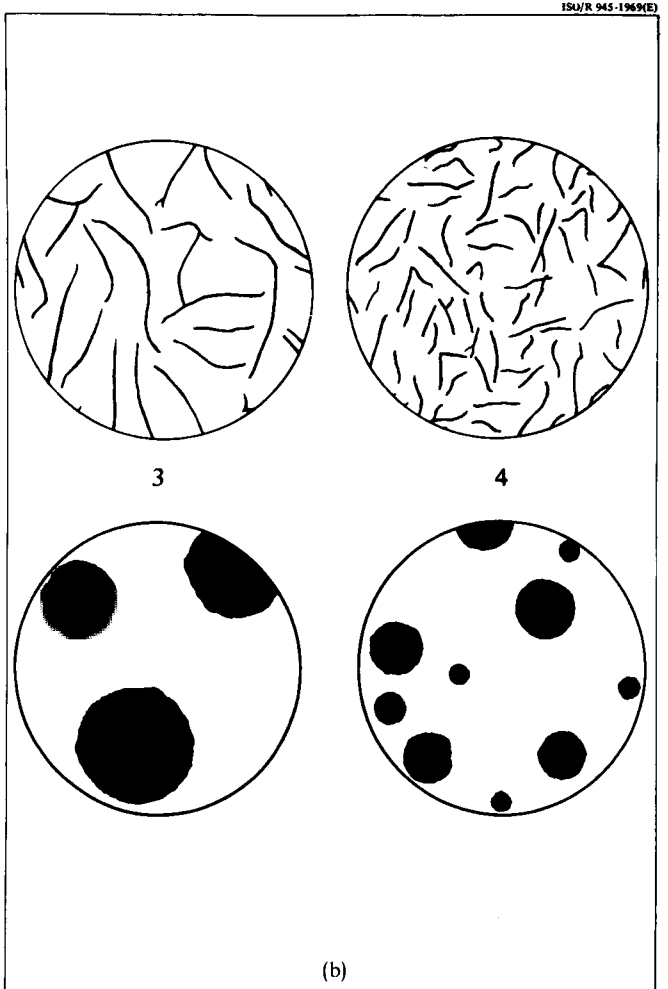


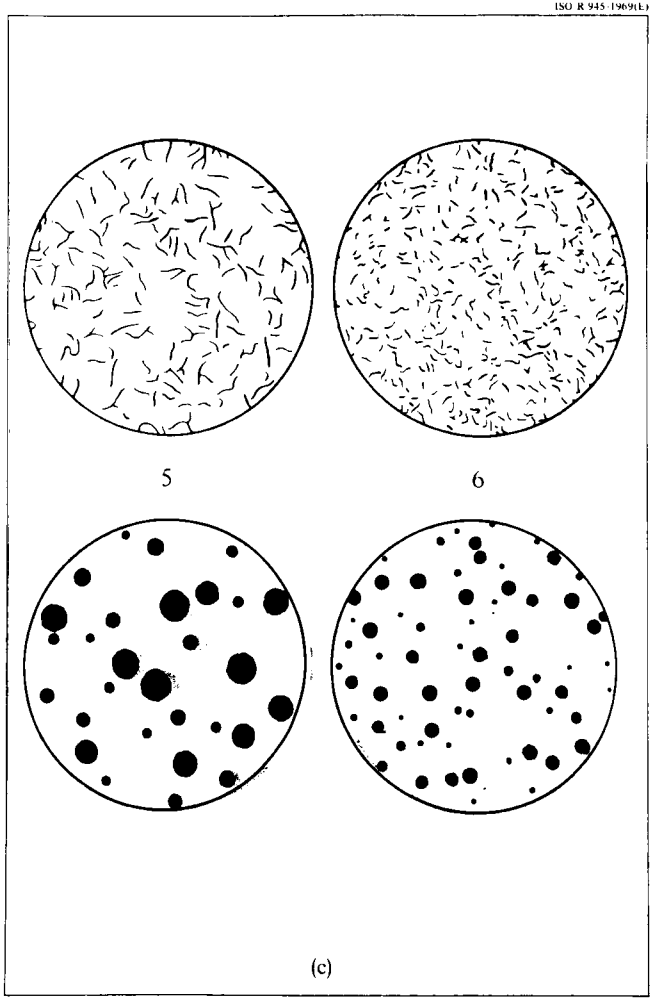
Figure 4(b). Reference diagram for the distribution of graphite (type or form I). The diagrams show only the outlines and not the structure of the graphite

rejected from the γ phase on to existing graphite flakes and form spiky irregularities on the graphite surface. This may occur during annealing and is shown in *Figure 11*.

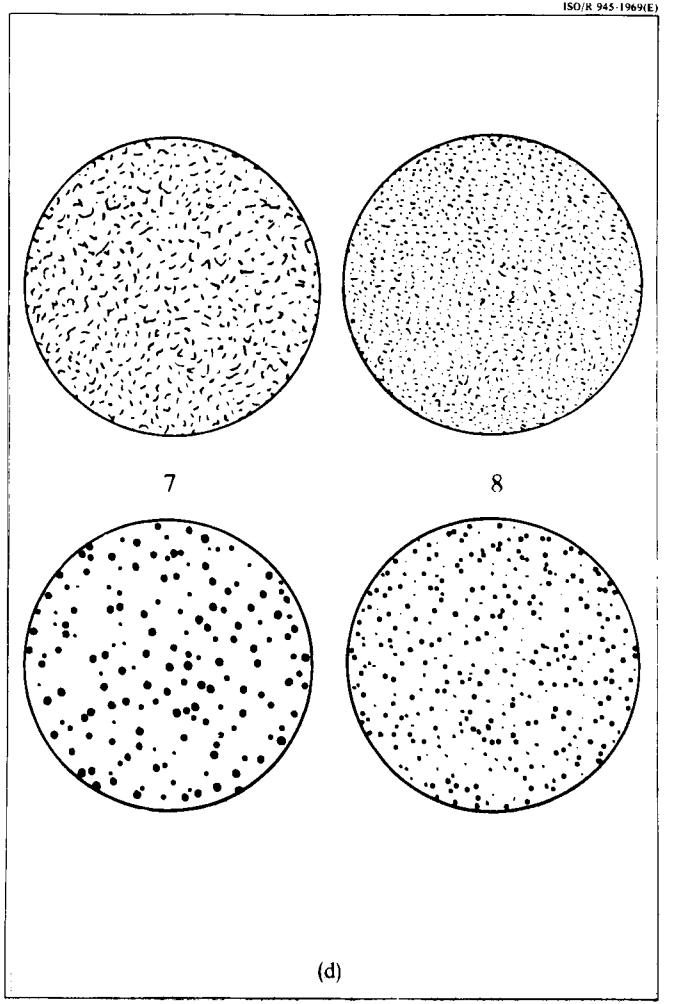




Figures 5(a) and (b). Reference diagram for the dimension of graphite forms I and VI and distribution A (× 50 magnifications). The diagrams show only the outlines and not the structure of the graphite (the original diagrams ISO/R 945-1969(E) were shown as × 100)



ISO/R 945-1969(E)



Figures 5(c) and (d). Reference diagram for the dimension of graphite forms I and VI and distribution A (× 50 magnifications). The diagrams show only the outlines and not the structure of the graphite (the original diagrams ISO/R 945-1969(E) were shown as × 100)

NODULAR GRAPHITE

The shape of the graphite precipitated on solidification can be markedly modified to the nodular or spheroidal form (ISO Form VI) by additions notably of magnesium and/or cerium.

Intermediate or degenerate forms can also be produced. Thus, ISO Form II crab type, is a degenerate form of flake graphite occasionally found in magnesium-treated irons containing lead or titanium, and can be suppressed by additions of cerium.

Quasi-flake graphite (ISO Form III) is usually the result of insufficient magnesium additions to produce the fully spheroidal form.

Irregular open or 'exploded' type spheroids (ISO Form V) may occur with very pure raw materials treated with magnesium and cerium and misshapen spheroids tend to occur in heavy sections.

AGGREGATE OR TEMPER CARBON (ISO FORM IV)

This is found in blackheart malleable iron, is produced by heat treatment of white iron, but the spheroidal form may be found in whiteheart malleable when the sulphur content is not balanced by the manganese (i.e. manganese % less than 1·7 × sulphur %), *see Figure 32*. *Figure 31* shows the decarburised edge of whiteheart malleable iron.

In the nodular aggregate forms, the planes of weakness characteristic of flake graphite are absent and metal so produced has a higher tensile strength and much increased impact resistance and ductility.

Characteristic structures of this type are shown in *Figures 32* and *33*, and the properties of these irons are detailed in their separate sections.

The critical temperature range

This is the range of temperature in which γ iron transforms to α iron on cooling down, and vice versa on heating up. The change is of particular importance since carbon is soluble in γ iron and practically insoluble in α iron. The temperature at which this change takes place is markedly affected by the silicon content and may also vary appreciably with the rate of heating or cooling. Rehder[9] states that if the cooling rate is increased from 4°C per hour to 28°C per hour, the critical temperature is lowered by 14°C.

Morgan[10] quotes the figures given in *Table 1* for slow rates of heating and cooling for differing silicon contents.

These figures were obtained with iron–carbon–silicon alloys containing 0·3–0·6 per cent manganese, and with carbon contents from 0·09 to 4·06 per cent. They were determined on rising temperature (rate unspecified), since the authors state that the position of the change point is less affected by rate of heating than rate of cooling. These results are very close to those of Schneidewind and Wilder[11] derived from White and Elsea[12] and Rehder[13].

Figures derived from the work of Kriz and Poboril[14] are given in *Table 2*.

Table 1

Silicon, %	2·03	3·43	4·46	5·64
Total carbon, %	2·50	2·54	2·49	3·1
Critical temperature heating, °C	760	800	875	935
Critical temperature cooling, °C	685	740	810	920

Table 2

Silicon, %	0	1	2	3	4	5	6
Critical temperature, °C	720	750	785	820	862	910	963

Figures are given by Palmer[15] for cast irons with 2·8–3·15 per cent carbon, 1·23 per cent phosphorus, and silicon ranging from 0·96 to 2·77 per cent, with heating rates of 17°C per minute and cooling rates of 11·5°C per minute. His results straddle those of Kriz and Poboril, being 25–30°C higher on heating, and 20–25°C lower on cooling.

Figure 6 shows results from various investigations and the full curve which is based on results for steel by Kriz and Poboril[14] may well be the likely mean curve. Maitland and Hughes[16] have determined similar curves for nodular iron also shown in *Figure 6*. The lower dotted curve shows the commencement of α to γ transformation after 6 hours heating at temperature, and the upper curve shows the limit for complete γ transformation. One point for a banded steel (0·2 per cent carbon, 0·04 per cent silicon, 0·56 per cent manganese) is taken from recent work by Cairns and Charles[17] in which the pearlite bands were rendered martensitic by heating to 720°C for 4 seconds and quenching, without diffusion of the carbon into the adjacent ferrite band and this confirms the earlier work of Kriz and Poboril for steel. The dotted lines of Maitland and Hughes for nodular cast iron straddle the Kriz and Poboril curve above 1 per cent silicon, but are lower for silicon below 1 per cent.

Rehder's work[13, 18] with grey, malleable and nodular iron for material fully austenitised at 954°C and quenched from various lower temperatures, falls roughly within the limits of Maitland and Hughes. He also indicates the lowering of the critical temperature for varying rates of cooling[18].

These curves emphasise the importance in heat-treating by quenching of ensuring that the material is heated sufficiently above the critical temperature and for a sufficient period of time to ensure a fully austenitic structure before quenching. It would appear that for normal heating rates, the austenitising temperature should be at least 20°C above the upper curve of Maitland and Hughes or 40–80°C above the Kriz and Poboril curve.

The significance of this temperature range for stable equilibrium (α, γ, and graphite) and metastable equilibrium (α, γ, and carbide) has been discussed by Boyle[19] (*Figure 7*) and Burgess[20].

Some unpublished tests carried out by Boyes at BCIRA illustrate this point. The iron employed contained: total carbon 3·15 per cent; silicon 2·2 per cent; manganese 0·67 per cent; sulphur 0·11 per cent; phosphorus 0·24 per cent. Samples were soaked for 24 hours at various temperatures and

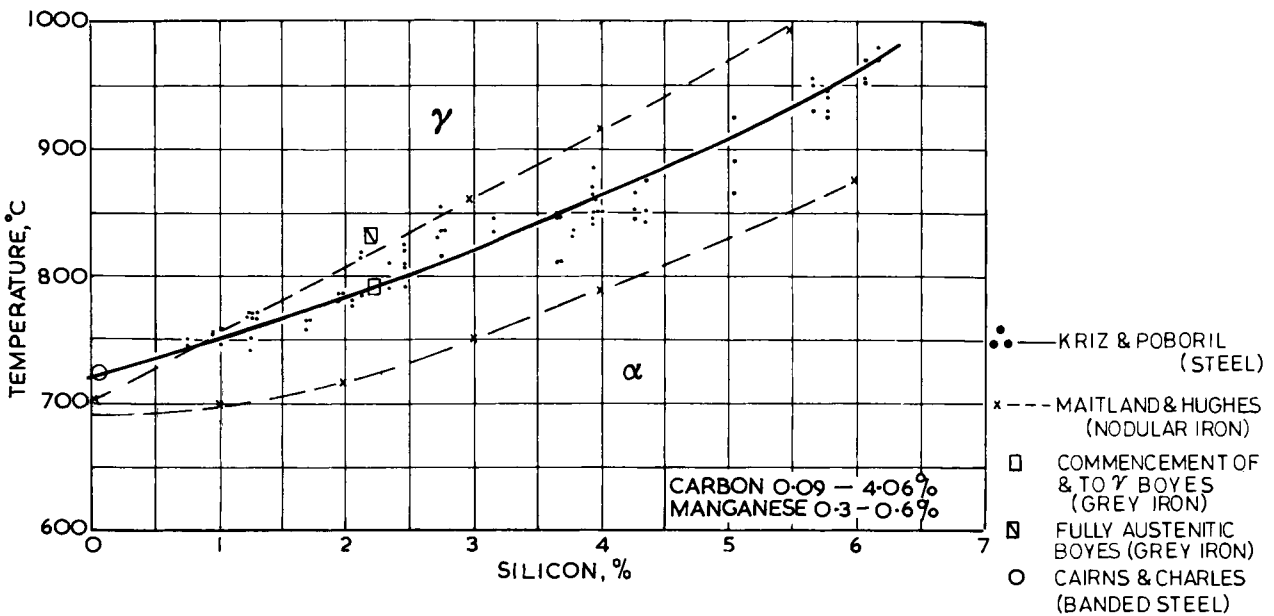


Figure 6. Effect of silicon on the α to γ transformation temperature of unalloyed steel and cast iron

water-quenched, and the results are summarised in *Table 3* and illustrated in *Figures 22–25*.

Table 3

	24 h at			
	770°C	*790°C*	*810°C*	*830°C*
Structure after water quench	Fully ferritic	Some α to γ change, but predominantly ferritic	Considerable α to γ change, but some ferrite present	α to γ change complete. Fully martensitic

It is apparent, therefore, that if heat treatment by quenching is to be fully effective, the minimum temperature to which the casting must be heated for full austenitisation to take place is from 40°C to 80°C above the critical temperature indicated by the Kriz and Poboril curve. If quenching takes place

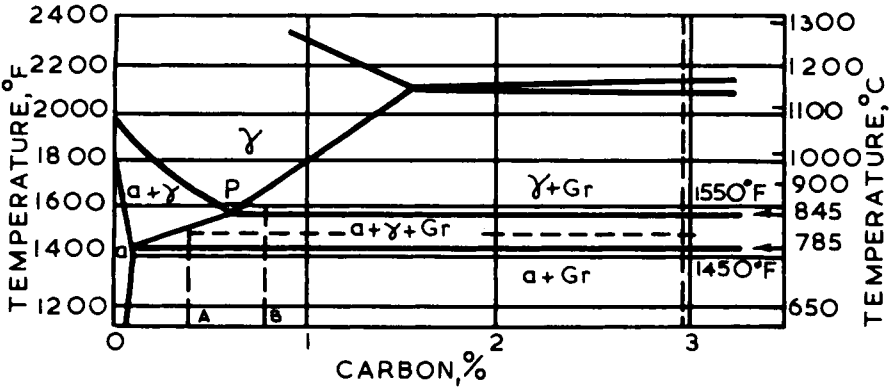


Figure 7. Reproduced from A. Boyle: 'The Structure of Cast Iron', by courtesy of the American Society of Metals

from the intermediate temperature range, full hardening cannot occur, as some free ferrite will remain.

Apart from silicon and manganese, the effect of the other normal elements in cast iron upon the critical temperature is not very great. According to Rehder[9], carbon between 2·5 and 3·50 per cent does not affect the temperature range, and manganese lowers the top and bottom of the range by 0·35°C and 1·4°C respectively for each 0·01 per cent manganese in the range 0·1 per cent to 0·7 per cent of free manganese. Burgess[20] gives an average figure of −25°C for each 1·0 per cent manganese addition and gives the formula:

$$\text{Critical temperature} = 730 + 28\,(\%\ \text{Si}) - 25\,(\%\ \text{Mn})$$

which conforms reasonably well to the curve in *Figure 6* in the range up to 2·0 per cent to 2·5 per cent silicon, but is low for the higher ranges.



Figure 8. Random graphite with a pearlitic matrix and low phosphorus. A typical structure of an engineering iron. The hardness may vary with strength and section from 180 to 240 Brinell (etched 4% picral, × 66)

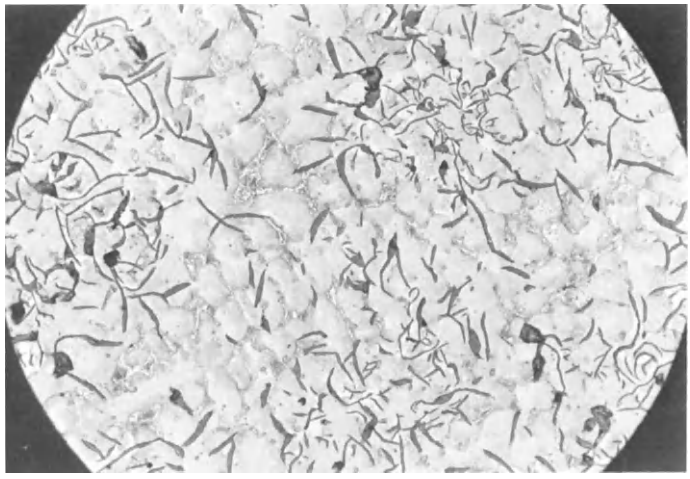


Figure 9. Grey iron with a high phosphorus content. The phosphorus is present as an interdendritic segregate of pseudobinary phosphide eutectic. A typical engineering or light castings iron with 0·7–1% phosphorus. The hardness may vary from 200 to 250 Brinell (etched 4% picral, × 66)

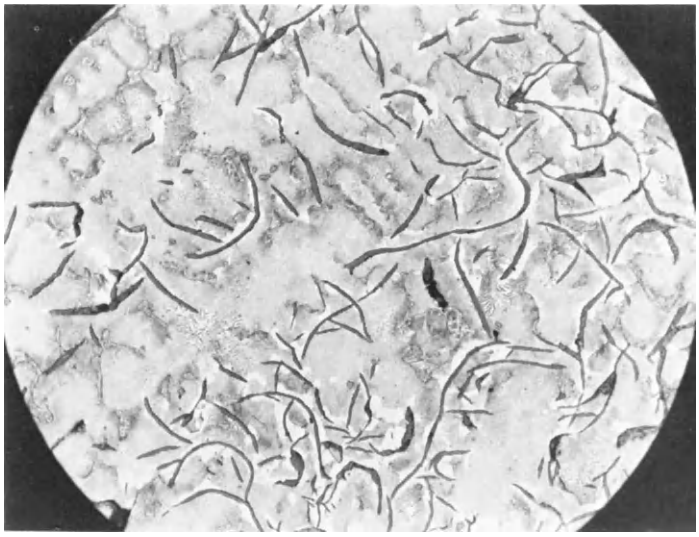


Figure 10. Mainly pearlitic grey iron with ferrite associated with normal graphite. A soft engineering iron with a hardness of 160–180 Brinell (etched 4% picral, × 66)



Figure 11. Ferritic grey iron produced by annealing a pearlitic grey iron. The form of the graphite is slightly altered by deposition of graphite following decomposition of carbide. This structure has been produced by annealing—usually for free machinability. The hardness of a low phosphorus annealed iron may be from 90 to 140 Brinell (etched 5% nital, × 66)



Figure 12. Rosette graphite (etched 4% picral, × 66)

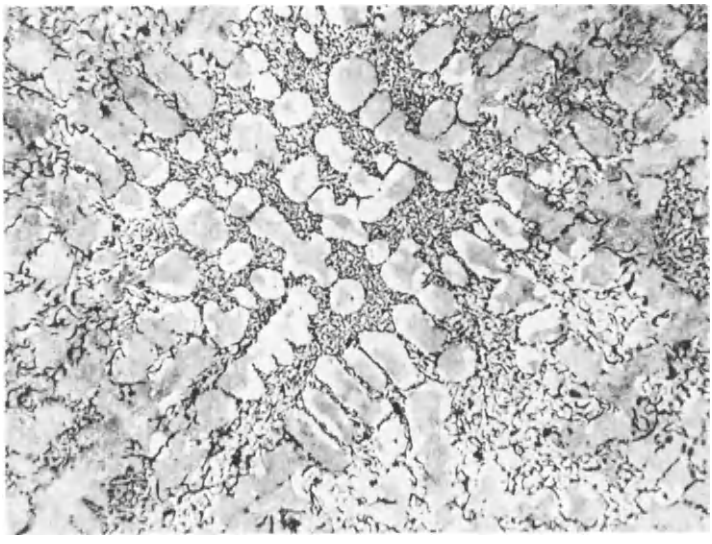


Figure 13. Undercooled graphite (etched 4% picral, × 66)



Figure 14. Martensitic grey iron. Unalloyed. Oil quenched from 850°C. Analysis: total carbon 3·51%, silicon 2·20%, manganese 0·65%, sulphur 0·072%, phosphorus 0·22%. Hardness 315–375 Brinell. Surface hardening may give figures up to 580 HV (etched 4% picral, × 600)

Rehder states that phosphorus raises both upper and lower levels by about 2·2°C for each 0·01 per cent phosphorus, at least up to 0·2 per cent phosphorus, but the effect is smaller above 0·2 per cent phosphorus. The work of Palmer[15] confirms a slight rise in the upper and lower limits with rapid rates of temperature change, but the effect does not exceed about 1·7°C for 0·1 per cent rise in phosphorus between 0·4 per cent and 1·1 per cent phosphorus, and it is uncertain whether with slow rates of temperature change the effect of phosphorus is significant.

Nickel depresses the range by about 17°C for a 1 per cent addition, and copper also depresses the range slightly, but there are anomalies with both elements[20].

Chromium raises the critical temperature by about 40°C for each 1 per cent addition.

The variation of the critical temperature with alloys must be remembered when heat treatment is carried out.

Phosphorus may vary from 0·04 to 1·7 per cent, and its effect on the carbon eutectic has already been mentioned. It forms an independent low melting point eutectic, solidifying intercellularly at about 930°C and forms an almost continuous intercellular network at about 0·4–0·7 per cent phosphorus (*Figure 9*) which markedly embrittles the iron and increases its hardness by 30–40 points Brinell for 1 per cent phosphorus addition.

Manganese may vary from 0·3 per cent to over 1 per cent and its chief function is to neutralise the sulphur content as manganese sulphide, and prevent the formation of the more harmful iron sulphide. This is achieved when the manganese content equals 1·7 × sulphur per cent + 0·2 or 0·3. When the sulphur content is not balanced by manganese, anomalous inverse chilling may occur, indicated by free carbide in the centre of an otherwise grey section. Any excess of manganese over the amount required to neutralise the sulphur will stabilise the pearlite and slightly increase the hardness, without, however, forming free carbide. One per cent manganese lowers the γ to α temperature by about 25°C.

Sulphur is usually present in amounts from 0·03 to 0·17 per cent and is a normal constituent of cast iron. When present in amounts less than 0·03 per cent, marked structural alterations from the conventional patterns may occur[21, 22].

The basic structures of grey and white irons are shown in the photomicrographs *Figures 8–13* and *20*.

More detailed microstructures have been published and should be consulted if additional information is required[23, 24].

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TYPICAL MICROSTRUCTURES

The graphite forms have been described and defined in the section on constitution and microstructure. The matrix structure of cast irons usually contains one or more of the following constituents:

(1) Ferrite in cast iron is essentially a single phase solid solution of silicon in amounts varying with graphite structure, cooling rate and silicon content, tending to increase in amount as the cooling rate decreases, the silicon content increases, and the graphite approaches the undercooled form (*Figure 13*). Fully ferritic structures are normally only obtained by annealing.

(2) Pearlite, consisting of alternate lamellae of ferrite and cementite which can be distinguished in *Figures 8* and *10*. At low magnification it appears as a half-tone colour. This structure is formed by the transformation of austenite during normal cooling in the mould or in the air through the critical range (720–900°C).

(3) Cementite (Fe_3C) in the massive eutectic form is a hard white constituent formed during solidification as in mottled (*Figure 15*) or white irons (*Figure 20*), and in the lamellar form in pearlite it is formed by the transformation of austenite through the critical temperature.

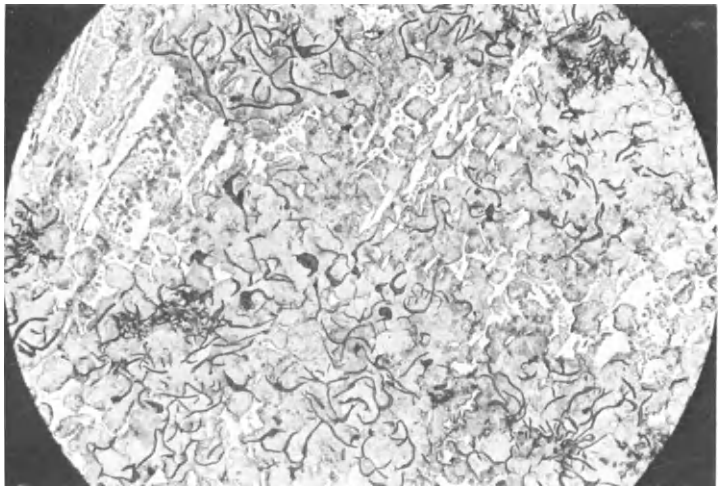


Figure 15. Mottled iron. Graphite, pearlite, cementite and phosphide. Hardness 260–350 Brinell (etched 4% picral, × 66)

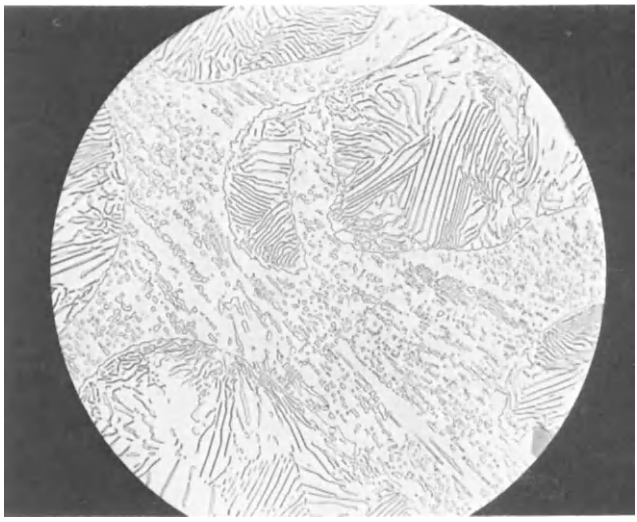


Figure 16. Ternary phosphide eutectic (etched 4% picral, ×400). See p. 27

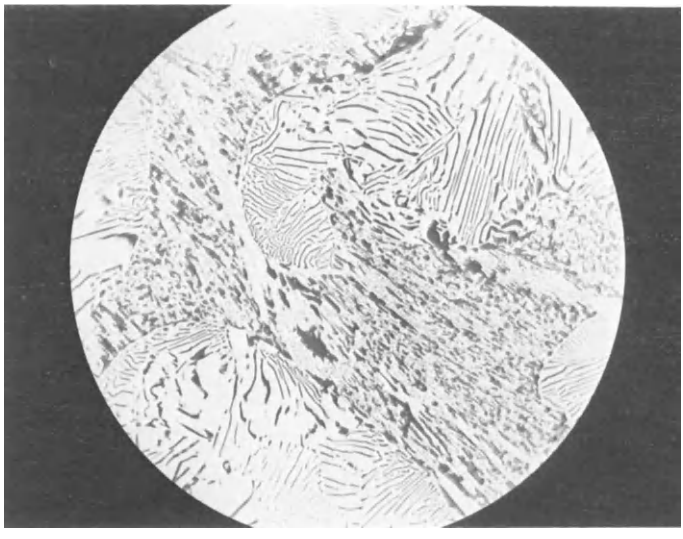


Figure 17. Ternary phosphide eutectic (etched Murakami's reagent the phosphide etched in black, ×400). See p. 27

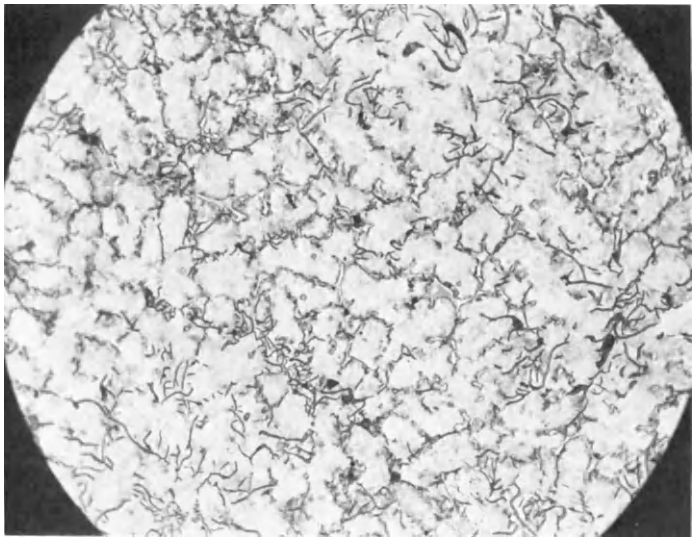


Figure 18. Ternary phosphide eutectic (etched 4% picral, × 66)



Figure 19. Ternary phosphide eutectic (etched 4% picral, × 40)

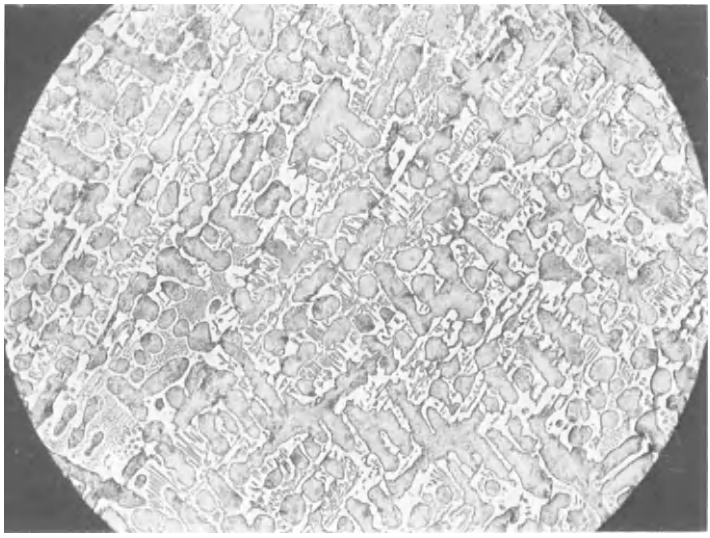


Figure 20. White iron pearlitic. The original dendrites of austenite have transformed to pearlite (half-tone) and in the interdendritic eutectic structure of cementite and austenite, the austenite has transformed to pearlite, leaving the cementite in its original form. Hardness 400–500 HV (etched 4% picral, × 66)

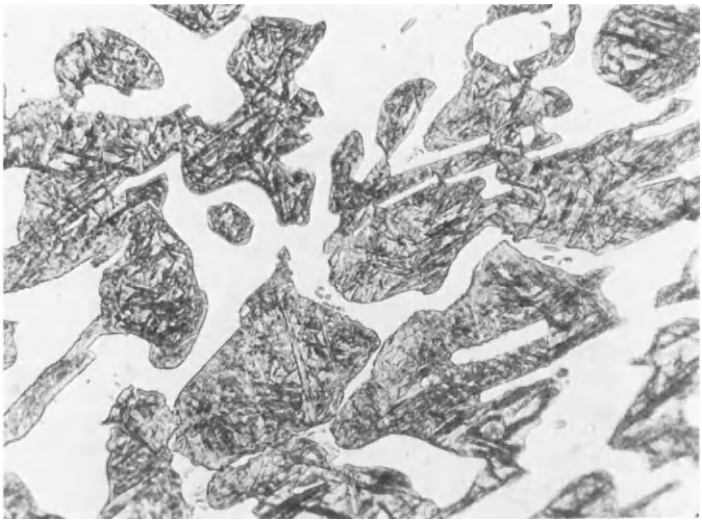


Figure 21. White iron martensitic in which the original austenite dendrites have transformed to martensite instead of pearlite as in Figure 20. This transformation is effected by the use of alloys. Hardness 550–750 HV (etched 4% picral, × 400)



Figure 22. Structure of cast iron water-quenched after 24 h at 770°C. Fully ferritic. Total carbon 3·15%, silicon 2·2%, manganese 0·67%, sulphur 0·11%, phosphorus 0·24% (etched 5% nital, × 400)

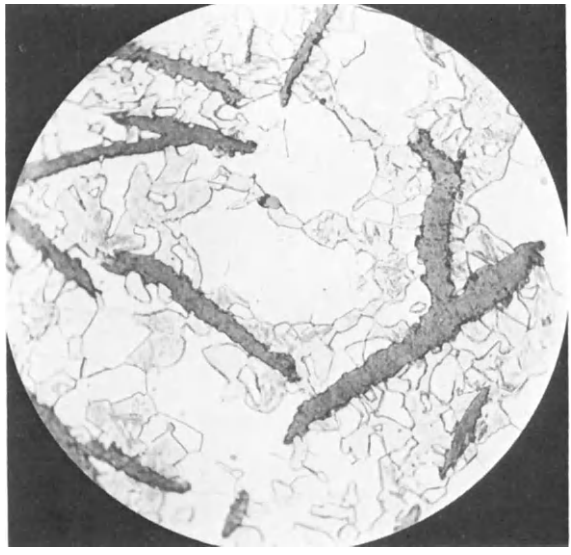


Figure 23. Sample of cast iron water-quenched after 23 h at 790°C. Some α to γ change, but predominantly ferritic. Cf. Figure 22 (etched 5% nital, × 400)

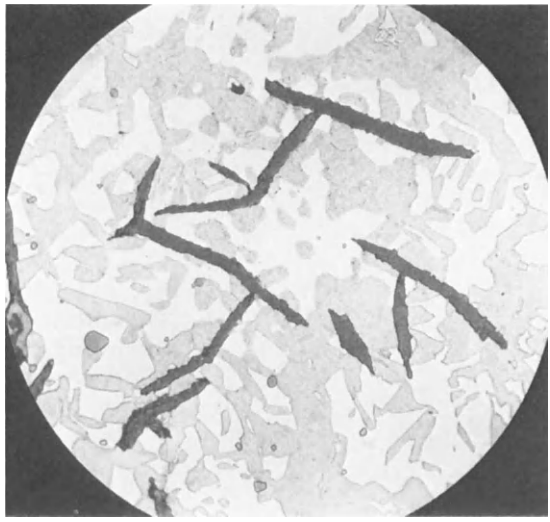


Figure 24. Sample of cast iron water-quenched after 24 h at 810°C. Considerable α to γ change, but some ferrite present (etched 5% nital, × 400)

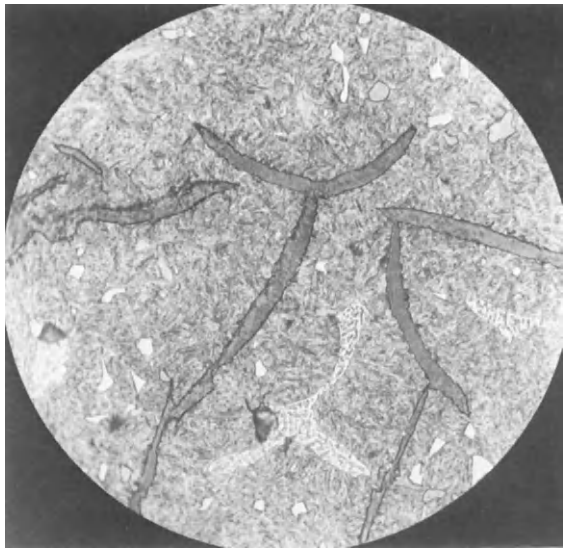


Figure 25. Sample of cast iron water-quenched after 24 h at 830°C. The α to γ change is complete. Fully martensitic. Cf. Figure 22 (etched 5% nital, × 400)

(4) Phosphide eutectic (melting point about 930°C) occurs in two distinct forms in cast irons with more than 0·06 per cent phosphorus. The pseudo-binary form is the normal form, consisting of ferrite and iron phosphide (Fe_3P). The true eutectic forms from the liquidus as austenite plus iron phosphide and on cooling the austenite transforms to ferrite and pearlite and with iron phosphide gives a bulk hardness of 420–600 HV (*Figure 9*).

It can, however, occur in the ternary form in mottled irons and in apparently grey irons (*Figures 16, 17, 18* and *19*) where it solidifies as austenite, iron carbide and iron phosphide, and at room temperature after transformation of the austenite it appears as a eutectic of ferrite, free carbide and iron phosphide. This form is likely to occur if chromium or vanadium are present in amounts over 0·1 per cent and has a bulk hardness of about 780–800 HV owing to the high hardness of the free carbide (800–1015 HV). If it occurs in apparently grey irons, it markedly increases tool wear in machining operations (*Figures 18* and *19*).

(5) Martensite is a fine acicular, slow etching structure, normally produced by very rapid cooling (quenching) of austenite through the critical temperature range, or by alloying. The structures in *Figures 14* and *21* are martensitic.

(6) *Figures 22–25* illustrate the effect of quenching after soaking at temperatures within the α to γ range of temperature (*Figure 7* and pp. 13–15). The change from α to γ is obviously not complete until a temperature of 30–40°C above the critical is reached. (*See* section on heat-treatment.)

(7) Acicular or bainitic transformation structures (rapid etching) which can be produced by isothermal quenching or alloying (*Figure 26*). These structures are often referred to as acicular ferrite and are softer and tougher than martensite, but harder and stronger than pearlite. A range of these acicular structures exists from the 'upper bainites' or acicular ferrites, to martensite, depending upon the transformation time, composition etc. (*See* sections on isothermal heat-treatment and acicular cast iron.)

(8) Austenite can be made stable at room temperature by the addition of alloys (nickel, manganese) which depress the critical temperature at which the γ to α change occurs (*Figure 27*). Although, however, the transformation may have been suppressed at room temperature, it may still take place at a lower temperature, dependent upon the amount of alloy present. *Figure 28* shows the structure of an iron which was austenitic at room temperature, but which became martensitic on exposure to a temperature of -60°C. (*See* section on austenitic cast iron.)

(9) Graphite can be produced in forms other than flake. The nodular form, illustrated in *Figures 29* and *30*, has been produced from the melt by the joint addition of magnesium and cerium. The nodular and aggregate forms shown in *Figures 32* and *33* have been produced by annealing a white iron. (*See* sections on nodular and malleable cast irons.)

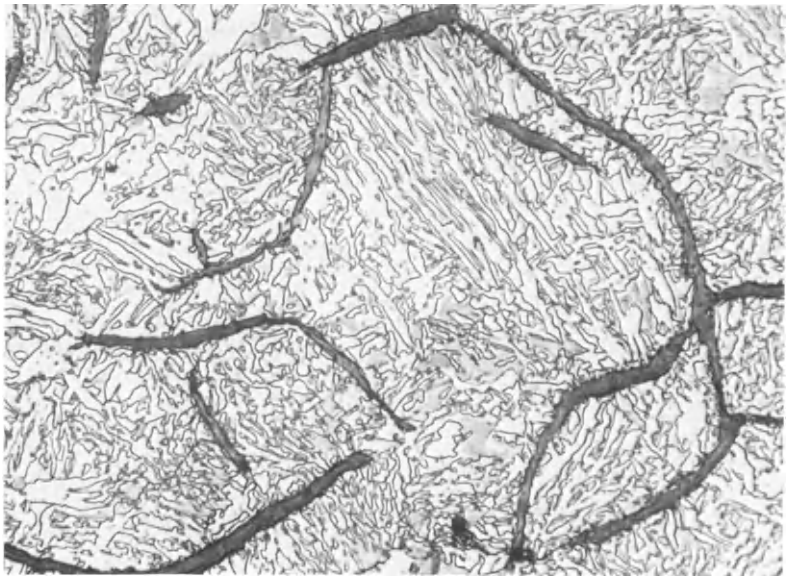


Figure 26. Acicular structure of iron of composition: total carbon 2·9%; silicon 1·67%; manganese 1·06%; sulphur 0·08%; phosphorus 0·048%; nickel 1·07%; molybdenum 0·83% (etched 4% picral, × 600)

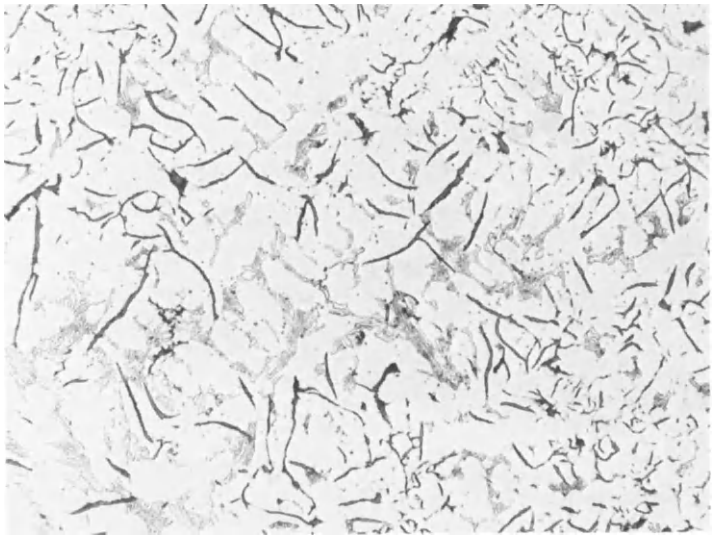


Figure 27. Austenitic cast iron (Ni-Resist) showing free carbides (etched 4% picral, × 100)

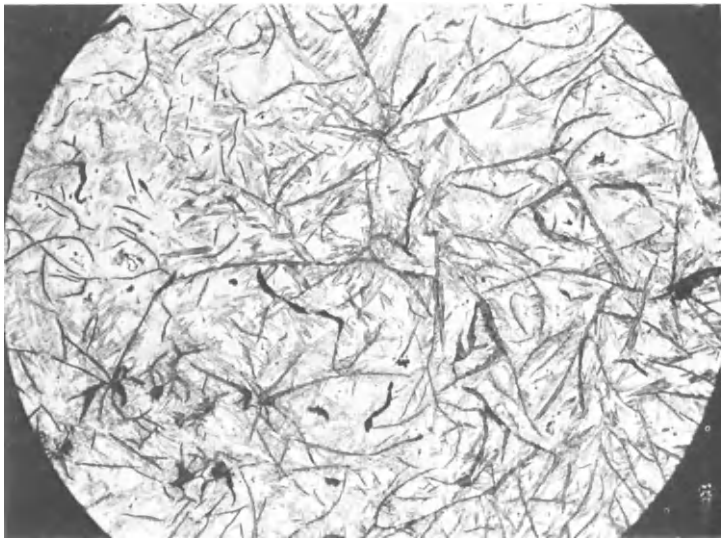


Figure 28. Austenitic cast iron showing martensite transformation as a result of exposure to a temperature of $-60^{\circ}C$ *(etched 4% picral, × 66)*

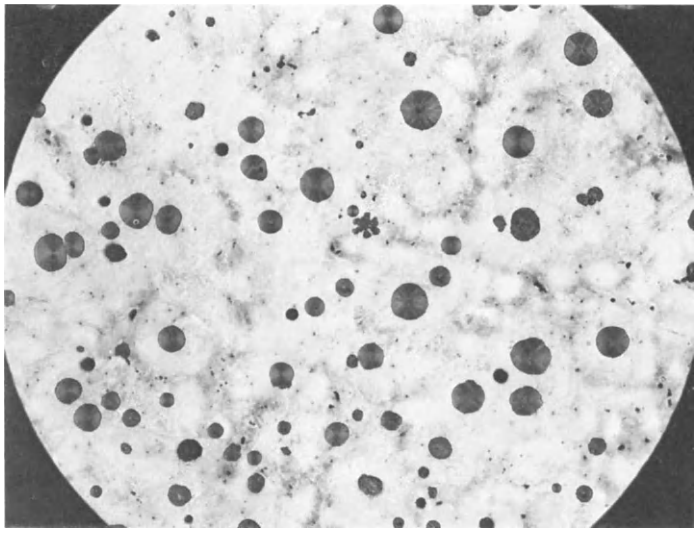


Figure 29. Nodular graphite is produced in the as-cast state by the joint addition of magnesium and cerium. In this case the matrix is pearlitic. Brinell 230–280. Tensile 35–45 tonf/in^2*. Elongation 2–6% (etched 4% picral, × 66)*

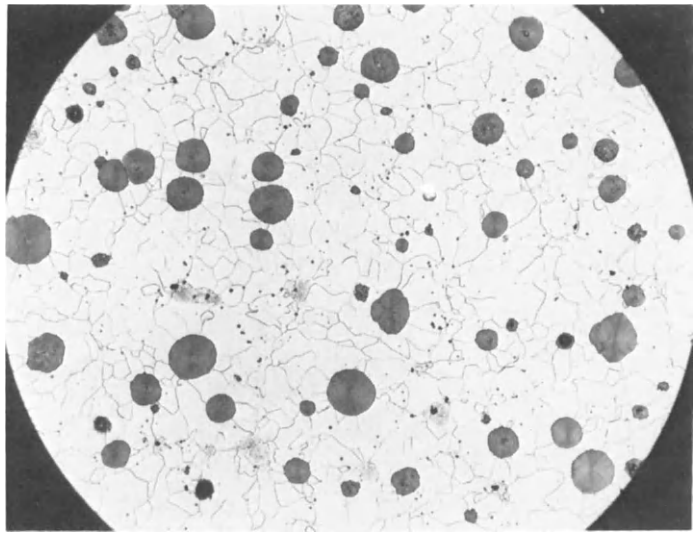


Figure 30. Nodular cast iron. By suitable heat treatment the pearlitic matrix can be annealed to the ferritic state, which is softer, more ductile and shock resistant than the pearlitic structure. Brinell 160–170. Tensile 24–27 tonf/in^2. Elongation 15–20% (etched 4% nital, × 66)

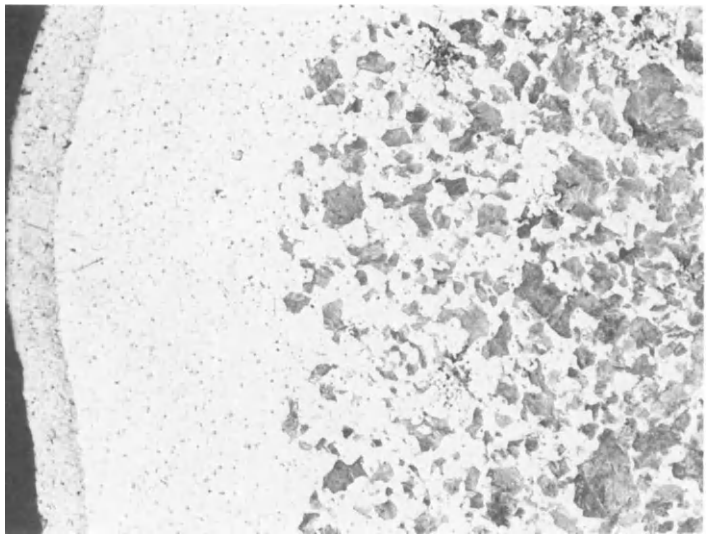


Figure 31. Whiteheart malleable cast iron. Produced by long annealing of white iron and characterised by a deep decarburised skin merging into a pearlitic centre, as shown in Figure 32 (etched 4% picral, × 40)

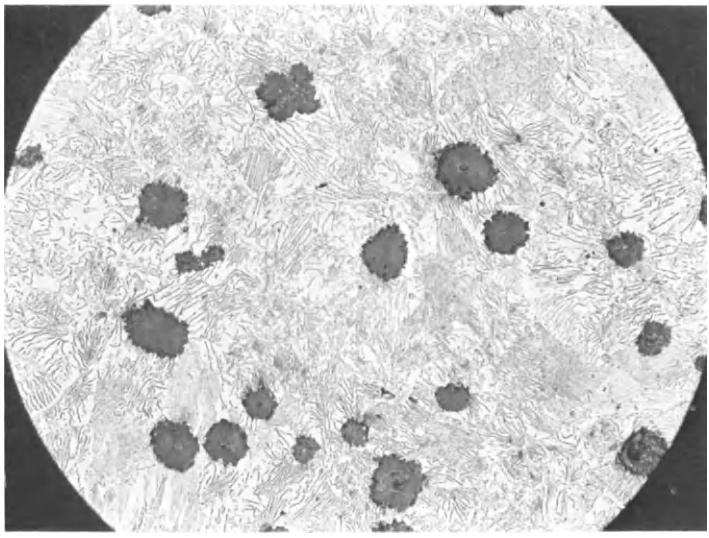


Figure 32. Pearlitic centre of whiteheart malleable with the graphite in the nodular form (etched 4% picral, × 66)

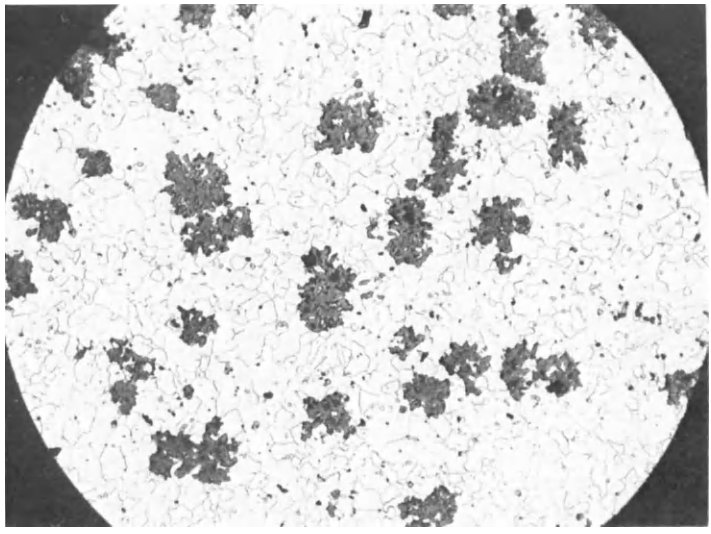


Figure 33. Blackheart malleable cast iron. Produced by long annealing of white iron in a neutral atmosphere. Characterised by aggregate graphite in a ferritic matrix in the centre of the section. As it is heat treated under non-oxidising conditions, there is only a very thin decarburised layer, with occasionally a rim of pearlite separating the annealed surface from the centre. Brinell hardness 109–149. Tensile 20–22 tonf/in^2. Elongation on 2 in: 10–15% (etched 5% nital, × 66)

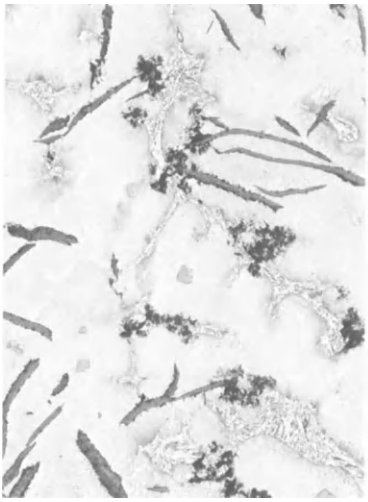


Figure 34. Widmanstätten graphite caused by presence of lead (etched 4% picral, × 66)



Figure 35. Widmanstätten graphite caused by presence of lead (etched 4% picral, × 400)

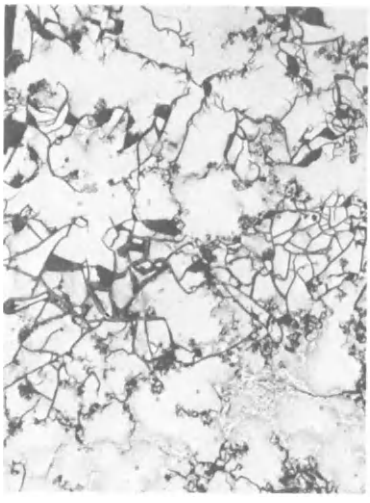


Figure 36. Mesh graphite caused by presence of tellurium (etched 4% picral, × 66)

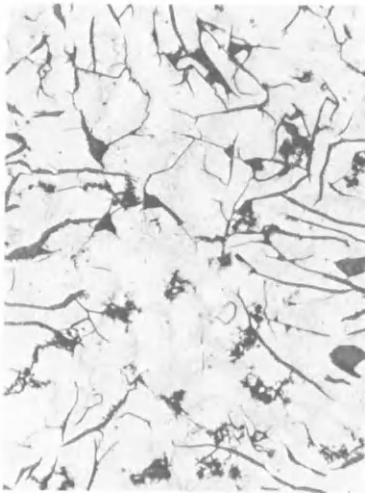


Figure 37. Mesh graphite caused by presence of tellurium (etched 4% picral, × 66)

(10) Windmanstätten graphite (*Figures 34* and *35*) may occur in grey iron in the presence of lead above 0·0007 per cent, particularly when hydrogen is present[1].

Mesh graphite (*Figures 36* and *37*) may also occur in the presence of tellurium and may be associated with excessive use of tellurium mould or core wash[2].

Both forms can cause catastrophic reduction in strength of the iron.

(11) The normal form of flake graphite can be modified to a compacted form where nitrogen contents are high, as in high steel cupola melts. The compacted form of graphite will increase the tensile strength, hardness and modulus of elasticity which in ingot moulds may increase the risk of major cracking.

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2

Mechanical, physical and electrical properties of cast iron

INTRODUCTION

The practical engineer is by force of circumstances often an empiricist—he finds a material satisfactory and then asks why afterwards. The bulk of the basic engineering materials was discovered in this way, but having discovered some of the virtues of a good material the engineer usually likes to put his perhaps somewhat fortuitous discovery on a firmer basis. In the field of cast iron, some of the earliest work on the strength of beams was carried out in England by Fairbairn, Hodgkinson and Tredgold in the period 1830–1850, and results showed immediately that the material had a more complicated response to stress than steel and the other engineering materials. An adequate body of working experience was established which was suitable for the conditions then existing, and although the anomalies were recognised and a good deal of work continued to be carried out, it was chiefly regarded as of academic interest. Since the beginning of this century, however, the amount of work carried out has been very greatly increased, and a fairly thorough knowledge of the properties of the material is now available. Unfortunately, however, much of the work is scattered, and it is difficult for the ordinary engineer to get a clear picture of the properties of the material without spending a great deal of time in examining the original literature. Chapter 2 of this work has been prepared to give in a concise form the data available on the mechanical and physical properties of the material.

In deciding to use cast iron for any given component, it is important that its fundamental properties should be understood. Although the structural constituents of steel can all be present in cast iron, there are two important constituents not normally present in steel which are responsible for the major characteristics of cast iron. These constituents are graphite and phosphorus. The presence of graphite, in particular, is characteristic of grey cast iron. The effect of graphite on the purely mechanical properties, as shown by the stress–strain relationship, is substantially that of a void, i.e. a steel matrix carrying a substantial number of voids distributed as graphite in cast iron would show very similar stress–strain characteristics. The effect of phosphorus is some-

what less fundamental, since there are many grades of castings produced in which the phosphorus content is purposely kept very low indeed, i.e. below 0·15 per cent. Nevertheless, the bulk of the castings produced in this country contain phosphorus ranging from 0·2 to 1·2 per cent, and as phosphorus forms a low melting point eutectic it is prone to segregation. At the same time, in the casting itself the phosphide eutectic tends to form a brittle network so that where shock-resistance and hydraulic soundness are of importance the phosphorus content is generally kept as low as possible.

The basic structural constituents of cast iron and steel are shown in *Table 4*.

Table 4

	Brinell hardness	*Ductility*
Ferrite	70–150	Ductile
Pearlite	175–330	Moderately ductile
Divorced pearlite	130–180	Ductile
Cementite	800–1 100	Brittle
Phosphide eutectic	400–600	Brittle
Graphite	–	–

It is often imagined that the presence of graphite in cast iron must mean that it will necessarily be porous even at appreciable thicknesses. This, however, is not the case, as the graphite flake in each eutectic cell is separated from those in the neighbouring cells, and a continuous network does not exist. Pressure tests with cast iron diaphragm as thin as 0·75 mm (0·03 in) have shown[1] that provided the wall thickness is greater than the actual size of the eutectic cell, absolute pressure-tightness is to be expected from a completely sound casting.

Thus, with a pig iron and various pearlitic engineering grey irons, at room temperature, a wall thickness of 3 mm (0·118 in) was impermeable to hydrogen up to 140 atmospheres pressure (2060 lb/in^2). Any porosity detected by the usual methods of pressure testing on castings of normal wall thickness is to be attributed to actual shrinkage cavities and channels and not to a continuous network of graphite.

In tension, graphite acts as a discontinuity in the matrix. It would be expected, therefore, that the strength of a cast iron would be inversely proportional to the amount of free graphite that is present, and this is substantially correct, although the character of the matrix itself also affects the strength. Unalloyed grey cast irons vary in strength from 6 to 20 ton/in^2 (93 to 309 N/mm^2), depending both upon the section size and upon the amount of graphite present. Since the strength depends considerably upon the total quantity of free graphite present, it would be expected that since graphite has a specific gravity of approximately 2·2, the stronger irons would be denser than the weaker irons. Reference to the section dealing with specific gravity confirms this assumption.

The effect of cooling rate and section size upon mechanical properties

The structure and properties of cast iron are probably more affected by the cooling rate from the solidification temperature than any other cast material. For grey cast irons there are two particularly important periods of cooling: the initial period of solidification in the region of 1135–1150°C, during which the graphite is formed, and the period covering the austenitic change point in the region of 650–720°C, which largely determines the character of the matrix. The size and distribution of graphite, and the presence of 'chill' caused by the formation of hard iron carbide are determined in the region of solidification, and for grey irons subsequent heat-treatment has very little effect upon the form of graphite.

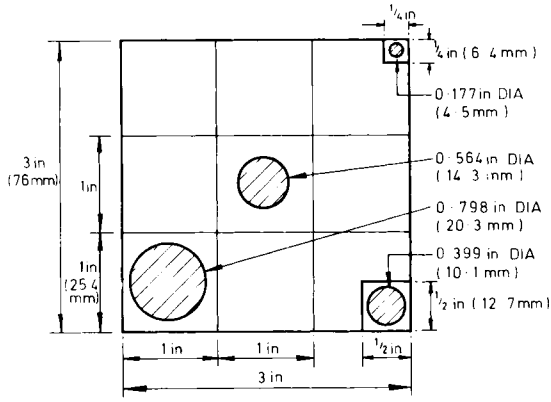


Figure 38. Cross section of a 3 in (76 mm) square bar

Since the structure and strength of a grey iron casting are mainly governed by the amount and form of the graphite, it is evident that the strength and structure depend not only upon composition, but also on the section of the casting which, in turn, governs the cooling rate. Thus, metal of the same composition cast from the same ladle may produce (a) a chilled, white, unmachinable cast iron, (b) a machinable grey casting with a chilled and unmachinable edge, (c) a strong, close-grained structure, or (d) a weak, open-grained structure, depending upon the section of the casting into which it is poured. For this reason, all important standards specify grey iron by reference to its strength when cast into test bars of fixed diameter and not by reference to the strength of the casting itself.

A good index of the rate of cooling of a casting is the volume/area ratio, since at a given temperature the amount of heat present is dependent upon the weight (or volume) of the casting, and all loss of heat takes place through the available surface area. Over a wide range of section sizes the cooling time of a casting varies with the square of the volume/area ratio, but the simple relationship of the cooling rate with the volume/area ratio, while far from exact, is nevertheless useful for comparing castings up to $1\frac{1}{2}$ in (38 mm) diameter of $\frac{3}{4}$ in (19 mm) section. In the simplest case, a flat plate of uniform

thickness, say $\frac{1}{4}$ in (6 mm), it is apparent that the rate of cooling of the extreme edges must be much greater than in the centre. It is a useful approximation to assume that the centre of a flat plate will cool at the same rate as a round bar whose diameter is twice the thickness of the plate, whereas the free edge of a flat plate will cool at the same rate as a bar whose diameter is equal to the thickness of the plate. This variation in cooling rate is, of course, well known in the foundry and machine shop where unmachinable 'chill' is always found first on thin edges.

Even in fairly compact castings, differences in cooling rate will affect the final strength. For example, *Figure 38* illustrates a cross-section of a 3 in (76 mm) square bar from which tensile specimens were taken at the corners and at the centre, and *Table 5* shows that even in this compact casting the mechanical strength and hardness are higher at the more rapidly cooled corners than in the more slowly cooled centre.

Table 5

Bar size: 3 in (76 mm) square × 16 in (406 mm)		Analysis: TC 3·35%, Si 1·15%, Mn 0·62%, S 0·04%, P 0·026%, CE value 3·79				
Position of specimen	*Centre*	*1 × 1 in (25·4 × 25·4 mm) Corner*	*½ × ½ in (12·7 × 12·7 mm) Corner*		*¼ × ¼ in (6·4 × 6·4 mm) Corner*	
Diameter of specimen, in	0·564	0·798	0·399	0·226	0·173	0·177
Diameter of specimen, mm	14·3	20·3	10·1	5·7	4·4	4·5
Tensile strength, ton/in^2	13·2	14·9	15·3	15·0	15·0	15·5
Tensile strength, N/mm^2	204	230	236	232	232	239
Brinell (3000/10, 750/5, 120/2)	171	182	193	194	194	199

More detailed reference will be made to this in Chapter 3.

Form of graphite

The foregoing has dealt mainly with grey cast iron in which the graphite is in the form of flakes or flake clusters, and most of the characteristic properties of grey iron depend on this graphite form.

The graphite, however, can be further varied in form, and in the malleable irons, whiteheart or blackheart, the graphite is in the form of nodules or aggregates. This structure is obtained by prolonged heat-treatment of white irons.

The nodular graphite structure, however, can be obtained in the as-cast state by the use of such additions as magnesium or cerium and magnesium. These irons are known as nodular or spheroidal graphite (SG) irons.

By modifying the form of the graphite in this manner, not only is the tensile strength increased, but a very high degree of ductility and shock resistance

Table 6 TENSILE TEST, BS 1452:1961

Nominal diameter of test bar as-cast in (mm)	*Main cross-sectional thickness of casting represented* in (mm)	*Minimum tensile strength* ton/in² (N/mm²)						
		Grade 10	*Grade 12*	*Grade 14*	*Grade 17*	*Grade 20*	*Grade 23*	*Grade 26*
0·6 (15)	Not exceeding $\frac{3}{8}$ (10)	11·0 (170)	13·0 (201)	16·0 (247)	19·0 (293)	22·0 (340)	25·0 (386)	28·0 (432)
0·875 (22)	Over $\frac{3}{8}$ and not exceeding $\frac{3}{4}$ (10–19)	10·5 (162)	12·5 (193)	15·0 (232)	18·0 (278)	21·0 (324)	24·0 (371)	27·0 (417)
1·2 (30)	Over $\frac{3}{4}$ and not exceeding $1\frac{1}{8}$ (19–29)	10·0 (154)	12·0 (185)	14·0 (216)	17·0 (263)	20·0 (309)	23·0 (355)	26·0 (402)
1·6 (41)	Over $1\frac{1}{8}$ and not exceeding $1\frac{5}{8}$ (29–41)	9·5 (147)	11·5 (178)	13·5 (208)	16·0 (247)	19·0 (293)	22·0 (340)	25·0 (386)
2·1 (53)	Over $1\frac{5}{8}$ (41)	9·0 (139)	11·0 (170)	13·0 (201)	15·0 (232)	18·0 (278)	21·0 (324)	24·0 (371)

can be achieved in the malleable and nodular irons, and their properties are referred to in the appropriate sections of this book.

Since the properties of cast iron are essentially dependent upon its microstructure, a brief section has been included (Chapter 1) in which the main structural types are indicated.

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THE TENSILE TEST

Standard specifications

The tensile test may be regarded as the standard test by which cast iron is specified. The tensile strength is the stress required to pull apart a test piece by an axially applied load. The test is almost always performed on a round, machined test piece in which the middle section of the length is reduced in diameter. The test piece is conveniently machined from independently cast bars or from broken transverse test bars and the British Standard Specification gives details of casting sizes appropriate to each size of cast bar. *Since the strength of cast iron is particularly dependent upon the section size into which it is cast, the strength of grey cast iron in BS 1452:1961 is defined in relation to five standard diameters of as-cast bars from which standard tensile specimens can be machined. Each diameter is regarded as representative of a range of section thicknesses, as shown in *Table 6*.

Each of the seven grades is denominated by the tensile strength required on the 1·2 in (30 mm) diameter bar. The specification for each grade, however, requires a higher tensile strength for bars of smaller diameter, and a lower tensile strength for bars of larger diameter. The variation in strength given in the British Standard Specification according to size of test bar, however, is less than has been found in practice, and merely indicates in an arbitrary manner the commercial difficulty of producing a given tensile strength in a given bar.

The properties determined by the tensile test

All the principal countries specify cast iron on the basis of standard diameter test bars, but it is common knowledge in the foundry industry that test bar results cannot be used to give the strength of the casting itself where section thicknesses and cooling rates may vary appreciably.

Everest has summarised this very adequately in a paper[1] in which he states:

(1) For all foundry alloys the separately cast test bar is essentially a check on metal quality only.

* *Speed of test* The rate of increase of stress on the test piece shall not exceed 10 ton/in^2/min (154 N/mm^2/min) after reaching 50% of the anticipated tensile strength (BS 1452:1961). Cf. Lloyd, H. K., *Metal Treatment and Dropforging*, **23**, July, 281–287 (1956)

(2) Strength relationships between the test bar and the casting vary with the nature of the alloy being cast.

(3) The control of castings, especially for vital services, demands the use of the separately cast test bar as a check on metal quality, but must be supported with check tests on the castings themselves. For repetition castings such tests can be conducted on a proportion of the castings produced, otherwise some suitable form of non-destructive testing must be applied.

It is important, therefore, to realise that national specifications merely define the metal in the ladle and do not give a direct indication of the strength of the casting itself. As explained earlier, the strength of the material in the casting is a function of the cooling rate of the particular portion under consideration. When, however, the strength of the metal cast under standard conditions is known, it becomes possible to estimate with some degree of accuracy the probable tensile strength to be found in a grey iron casting of a given thickness. If, however, the test bar were cast on to the casting itself, it would be impossible to relate the results obtained from such a test bar with those obtained from another test bar under standard conditions, and such a test would be almost valueless, since the test bar would not be representative of the casting itself, nor could it be related to other standards.

In the past it has been quite common practice to call for cast-on test bars, on the assumption that such a test bar must necessarily be representative of the casting itself. The foregoing remarks make it clear that this practice does not provide the necessary information and, apart from certain special cases, the use of the independently cast test bar has now almost entirely superseded the cast-on test bar.

The effect of section on strength

To interpret the results obtained from the standard test bar in terms of the strength of the casting itself, some relationship must be established between various section sizes and the standard bar. A statistical survey of a large number of industrially-cast grey iron test bars of different sizes has been made[2], and the results are shown diagrammatically in *Figure 39*. The curve shows the falling off in strength with increasing test bar size for the main British Standard grades of unalloyed grey cast iron. The line for each grade is really the centre of a zone of uncertainty amounting to ± 1–$1\frac{1}{2}$ ton/in^2 since there is a possible variation in each grade owing to differences in composition, pouring technique etc. A version of this curve is included in the BS 1452: 1961.

The total strain (recoverable and non-recoverable) of grey iron tensile test pieces at fracture usually varies between 0·5 per cent for high tensile iron and 0·75 per cent for low tensile iron.

The non-elastic strain varies between 0·2 per cent for high tensile irons and 0·6 per cent for low tensile irons (*see Table 73*).

British and foreign standard specifications for grey, nodular and malleable cast irons are given in Appendix.

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THE TRANSVERSE TEST

In the transverse test for cast iron, bars of specified length and diameter are supported at each end on rollers of suitable diameter, a load is applied to a point in the centre of the bar, and the bar tested to fracture (*Figure 40*).

It is customary also to determine the deflection which occurs before fracture so that an estimate of the ability of the material to deform is obtained, as well as its strength (*Table 7*).

The test is usually carried out on unmachined bars and the result may be

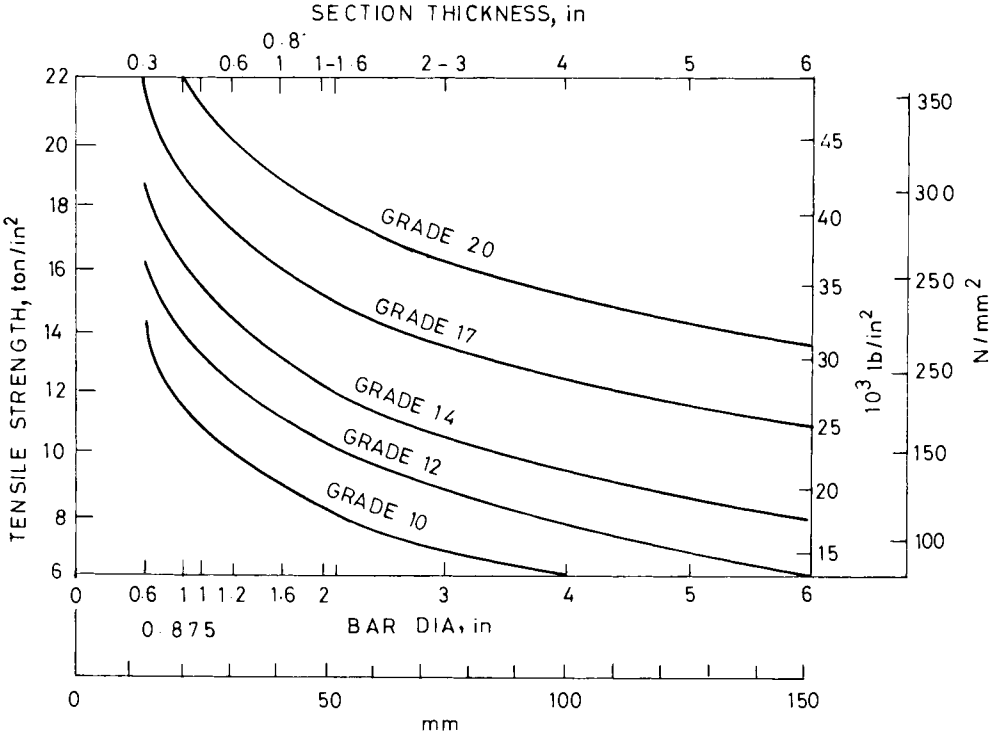


Figure 39. Effect of section on strength

Table 7 TRANSVERSE TEST, BS 1452:1961

Minimum breaking load
lb (kN)

Nominal diameter of test bar as-cast, in (mm)	0·6 (15)	0·875 (22)	1·2 (30)	1·6 (41)	2·1 (53)
Distance between support centres, in (mm)	9 (229)	12 (305)	18 (457)	18 (457)	24 (610)
Grade 10	430 (1·91)	975 (4·34)	1 640 (7·30)	3 750 (16·68)	6 100 (27·13)
12	485 (2·16)	1 100 (4·89)	1 850 (8·23)	4 250 (18·90)	6 950 (30·92)
14	565 (2·51)	1 255 (5·58)	2 060 (9·16)	4 750 (21·13)	7 830 (34·83)
17	645 (2·87)	1 440 (6·41)	2 380 (10·59)	5 390 (23·98)	8 680 (38·61)
20	725 (3·22)	1 625 (7·23)	2 700 (12·01)	6 150 (27·36)	9 970 (44·35)
23	805 (3·58)	1 815 (8·07)	3 030 (13·48)	6 900 (30·69)	11 250 (50·04)
26	885 (3·94)	1 995 (8·87)	3 350 (14·90)	7 660 (34·07)	12 540 (55·78)

Corresponding transverse rupture stress
ton/in^2 (N/mm^2)

Nominal diameter of test bar as-cast, in (mm)	0·6 (15)	0·875 (22)	1·2 (30)	1·6 (41)	2·1 (53)
Distance between support centres, in (mm)	9 (229)	12 (305)	18 (457)	18 (457)	24 (610)
Grade 10	20·5 (317)	19·9 (307)	19·3 (298)	18·7 (289)	18·0 (278)
12	23·1 (357)	22·4 (346)	21·8 (337)	21·2 (327)	20·5 (317)
14	26·9 (415)	25·6 (395)	24·3 (375)	23·7 (366)	23·1 (357)
17	30·7 (474)	29·4 (454)	28·1 (434)	26·9 (415)	25·6 (395)
20	34·4 (531)	33·2 (513)	31·9 (493)	30·7 (474)	29·4 (454)
23	38·2 (590)	37·0 (571)	35·7 (551)	34·4 (531)	33·2 (513)
26	42·0 (649)	40·7 (629)	39·5 (610)	38·2 (590)	37·0 (571)

Table 7 (continued)

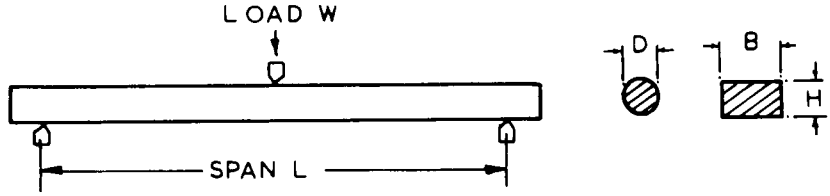
	Minimum deflection in (mm)				
Nominal diameter of test bar as-cast, in (mm)	0·6 (15)	0·875 (22)	1·2 (30)	1·6 (41)	2·1 (53)
Distance between support centres, in (mm)	9 (229)	12 (305)	18 (457)	18 (457)	24 (610)
Grade 10	[0·065] (1·65)	[0·09] (2·29)	[0·14] (3·56)	[0·11] (2·79)	[0·14] (3·56)
12	[0·07] (1·78)	[0·10] (2·54)	[0·15] (3·81)	[0·12] (3·05)	[0·15] (3·81)
14	[0·075] (1·91)	[0·105] (2·67)	[0·16] (4·06)	[0·13] (3·30)	[0·17] (4·32)
17	0·090 (2·29)	0·13 (3·30)	0·20 (5·08)	0·16 (4·06)	0·22 (5·59)
20	0·10 (2·54)	0·15 (3·81)	0·23 (5·84)	0·19 (4·83)	0·25 (6·35)
23	0·10 (2·54)	0·15 (3·81)	0·23 (5·84)	0·19 (4·83)	0·25 (6·35)
26	0·10 (2·54)	0·15 (3·81)	0·23 (5·84)	0·19 (4·83)	0·25 (6·35)

Note: The deflections in Grades 10, 12 and 14 were those commercially acceptable according to BS 1452:1948. They have been deleted from the revised specification BS 1452:1961, but have been included here for information and are shown in square brackets. The deflections in the higher grades have been raised from those in the original specification as an indication that the material has some improved impact resistance.

expressed either as the actual load to fracture for a given size of bar and span, or by the calculated skin stress at the tension side. The latter method should theoretically give the same skin stress for bars of the same strength material, but differing in diameter, in the same way that corrections for diameter do for the tensile test. This calculated figure is known as the *transverse rupture stress* and is obtained by the use of the formulae:

$$\text{TRS} = \frac{WL}{4Z}$$

where Z = modulus of section, W = load and L = span.



ROUND BARS,

$$\text{TRANSVERSE RUPTURE STRESS} = \frac{WL}{0{\cdot}3928D^3}$$

RECTANGULAR BARS,

$$\text{TRANSVERSE RUPTURE STRESS} = \frac{3WL}{2BH^2}$$

Figure 40. Transverse test

For round bars $Z = 0{\cdot}0982d^3$, where d = diameter of bar in inches. Therefore

$$\text{TRS} = \frac{WL}{0{\cdot}3928d^3}$$

For rectangular bars $Z = \frac{BH^2}{6}$, where B = breadth of section, H = height of section in inches. Therefore

$$\text{TRS} = \frac{3WL}{2BH^2}$$

[If W measured in N, L, B, H, d measured in mm, Then TRS measured in N/mm^2]

If cast iron behaved elastically, the transverse rupture stress would be the same as the tensile strength. Actually, however, the transverse rupture stress of grey iron as obtained by standard test bars varies from 1·6 to 2·1 times the tensile strength. This is because with a grey iron the deformation caused by the application of a load is at no time elastic. The formula upon which the transverse rupture stress is calculated is based on the assumption that the material of the test bar is elastic in its response to stress and that the neutral axis lies in the centroid of the section. With grey cast iron, owing to its non-elastic response to stress, the neutral axis is found nearer to the face in com-

pression so that the actual skin stresses in tension are lower than those calculated. The section on modulus of elasticity and stress–strain characteristics deals with this anomaly.

For standard test bars, the ratio between transverse and tensile stress is fairly constant for irons of given strength, as shown in *Figure 41*, but this

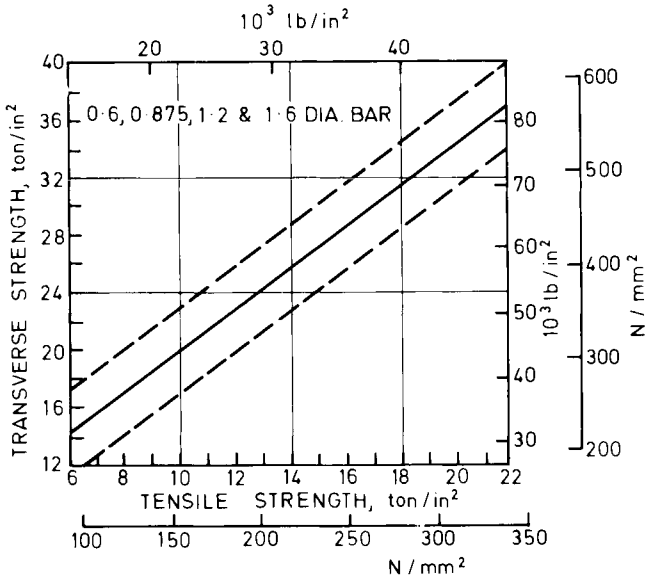


Figure 41. The relation between tensile strength and transverse rupture stress for bars varying from 0·6 in to 1·6 in in diameter

relationship only holds good for the test bar sections and cannot be safely extrapolated for beams of widely different size and section; this point is dealt with under the section dealing with beams and columns.

Deflection

Although the ability to obtain some measure of this deformability of cast iron is one reason for the original popularity of this test, the difficulty of obtaining really reproducible results with ordinary equipment is considerable, and in practice the deflection of uninoculated grey irons is more dependent upon the phosphorus content than on the tensile strength or the transverse rupture stress. The deflection at failure is dependent both on the modulus of elasticity and on the strength of the iron. The modulus of elasticity is lower for low strength than for high strength irons, and therefore for low strength irons with the same load, the deflection will be greater. At fracture, therefore, the difference in deflection between low and high strength irons

would be expected to be much less than the difference in the transverse rupture stress.

In practice, for irons of comparable phosphorus content there is practically no difference in deflection at fracture between irons of BS 1452, Grade 10 and Grade 20, but irons with phosphorus contents of 1 per cent or more will show only 50 to 80 per cent the deflection of those with less than 0·2 per cent phosphorus, and the test can therefore be regarded as a somewhat insensitive method of determining toughness.

For these reasons the test is now usually regarded as subsidiary to the tensile test. It is, however, useful as a control test.

THE COMPRESSION TEST

Compression tests are normally carried out on specimens with a length:diameter ratio of 2:1. It is important to quote the ratio since tests carried out with $\frac{L}{D}=1$ may be 10–12 per cent higher than $\frac{L}{D}=2$. An $\frac{L}{D}$ ratio of 3 appears to give results close to those carried out at $\frac{L}{D}=2$.

The compression strength of grey cast iron is approximately three to four times its tensile strength. Thus, for a Grade 10 iron, a compression strength of 35–40 ton/in^2 (541–618 N/mm^2) would be expected, and a Grade 20 iron would have a strength of 55–75 ton/in^2 (849–1158 N/mm^2).

There is also[1, 2, 3, 4] some relation between Brinell hardness and compression strength for grey irons ranging from 7·2 to 25 ton/in^2 (111 to 386 N/mm^2) tensile with hardnesses from 110 to 269. The ratio of compression strength in tons per square inch (N/mm^2) to Brinell number usually falls between 0·22 (3·4) and 0·26 (4·0) for uninoculated grey irons, ratios below 0·24 (3·7) occurring most frequently with hardnesses below 175, and ratios above 0·24 (3·7) for hardness above 175.

Grey cast iron actually fractures at its maximum compressive load, and in this respect differs from malleable iron, steel and other ductile materials which deform plastically. Hence, the effective load that can be carried by grey cast iron in compression is as high as that which can be carried by a steel with three to four times its tensile strength. In the case of malleable iron, the compression strength is approximately twice that of its tensile, but normally deformation becomes appreciable at loads above 12 ton/in^2 (185 N/mm^2). White iron has a compression strength of 95–120 ton/in^2 (1467–1853 N/mm^2) and appears also to conform quite well to the ratio of 0·25 (3·9) between the compression strength in tons per square inch (N/mm^2) and the diamond pyramid number[5].

The modulus of elasticity in compression of grey iron at low stresses is roughly 3–5 per cent higher than for tension. As the stress increases, the modulus tends to decrease, but to a lesser extent than for comparable tensile stresses. The stress–strain response of cast iron in compression is referred to in the appropriate section.

There appears to be little published information available on the modulus of elasticity in compression for malleable cast irons, but the figures in *Table 8* have been obtained for blackheart malleable and pearlitic malleable[6]:

Table 8

	0·1% Proof stress in compression ton/in² (N/mm²)		*Modulus of elasticity in compression* lb/in² × 10⁶ (GN/m²)		*Limit of proportionality (approx.)* ton/in² (N/mm²)
Blackheart malleable	11·1	(171)	23·1	(159)	7·5–8·0
	11·0	(170	24·0	(165)	(116–124)
Pearlitic malleable, 30 ton/in² tensile strength (463 N/mm²)	15·3	(236)	23·8	(164)	9·5–10·0
	15·1	(233)	22·5	(155)	(147–154)
Pearlitic malleable, 35 ton/in² tensile strength (541 N/mm²)	16·9	(261)	24·1	(166)	10·5–11·0
	16·8	(259)	24·4	(168)	(162–170)

The modulus of elasticity is markedly affected by the graphite content, whether in flake, aggregate or nodular form, and is lower in high graphite irons than in low graphite irons.

In whiteheart malleable cast irons, the modulus of elasticity in compression may vary with the degree of decarburisation, and figures up to 27×10^6 lb/in² (186 GN/m²) may be found.

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THE HARDNESS TEST

Although hardness testing is taken for granted at the present time, the property of hardness is an extremely complex one and standardisation of methods of testing were only introduced at the beginning of this century[1,2].

The two normally accepted methods of hardness testing for cast iron are the Brinell and the diamond pyramid. These two methods depend upon the measurement of an indentation made by applying a fixed load to a steel ball

in the case of the Brinell, and a diamond pyramid in the diamond pyramid test. The hardness is expressed as:

$$\frac{\text{Applied load, kg}}{\text{Contact area of recovered indent, mm}^2}$$

In the Brinell test the load (P) and the diameter of the ball (D) may be varied. In the diamond pyramid test the shape of the diamond is fixed, but the load may be varied.

It is important to remember that the Brinell hardness number (HB) of a given material is not independent of the test load, since in the same material the impressions made by the same size of ball under different loads are not geometrically similar. In general, the diameter of the impression should be within 0·25–0·5 on the ball diameter, the mean being 0·375D. It is important, therefore, for Brinell hardness tests, that the ratio $\frac{P}{D^2}$ should be in accordance with BS 240 (Part 1):1962 if comparable results are to be obtained. In grey iron practice it is usual to employ a 3000 kg load and 10 mm ball, or 750 kg load and 5 mm ball.

The diamond pyramid number (HV) in a uniform material is practically independent of the test load, as the impressions under varying loads are geometrically similar. The Vickers indenter is a polished sharp-pointed square pyramid of diamond with an included angle of 136° between opposite faces. This is the tangential angle of a Brinell spherical impression whose diameter is 0·375D, so that Vickers hardness numbers (HV) are substantially identical with Brinell numbers (HB) up to about 300 HB.

The Brinell test is rarely employed for hardnesses above 450*, since the ball tends to distort appreciably. The diamond pyramid number, however, may be regarded as a direct extension of the Brinell scale for hardnesses above 350–400, and is practically coincident with it for hardnesses below 350. Other methods, such as scleroscope or Rockwell, are most suitable for white and chilled irons. Owing to the presence of constituents of widely varying hardness, the diamond pyramid, even with 120 kg load, is only suitable for moderately fine-grained irons or for the purpose of exploring hardness in different regions of the same section. The Brinell is the most generally used test for hardnesses below 300–350 and, as it is unusual to use a ball diameter less than 5 mm, the area of the impression is relatively large and is therefore less affected by local variations in structure. The diamond pyramid number (BS 427 Part 1:1961) is the most suitable for hardnesses above 350. The accuracy of the Brinell test is dependent upon errors in the machine and errors in measurement of the impression. Variations of ±3–4 per cent are regarded as acceptable between individual impressions on a test block.

It cannot be emphasised too strongly that there is no clear relation between hardness and tensile strength for cast iron as there is for ordinary steel. A scatter curve reproduced from Mackenzie[3] in *Figure 42* of the relation between tensile and Brinell of a very large number of commercially-produced irons illustrates this point clearly.

* The use of a tungsten carbide ball extends the range to about 600 (BS 240: Part 1: 1962)

Barlow and Lorig[4] have shown that the ratio

$$\frac{\text{Tensile strength, lb/in}^2\ (\text{N/mm}^2)}{\text{Brinell hardness}}$$

varies between 120:1 (0·8:1) and 234:1 (1·6: 1) and that the scatter was greater in high strength (hypo-eutectic) irons; this has been confirmed by Jungbluth and Heller[5].

Collaud[6] has shown 194 hardness tests on 30 mm bars cast from a hot blast and cold blast cupola in relation to tensile strength and obtained the relation $HB = 125 + 5{\cdot}32T$ (or $0{\cdot}34T$), where T is the tensile strength expressed in tonf/in^2 (or N/mm^2). The scatter for irons between 10 and 24 tonf/in^2 (154 and 371 N/mm^2) tensile was about ± 15 points Brinell, but when compared with results from other sources the scatter is much wider.

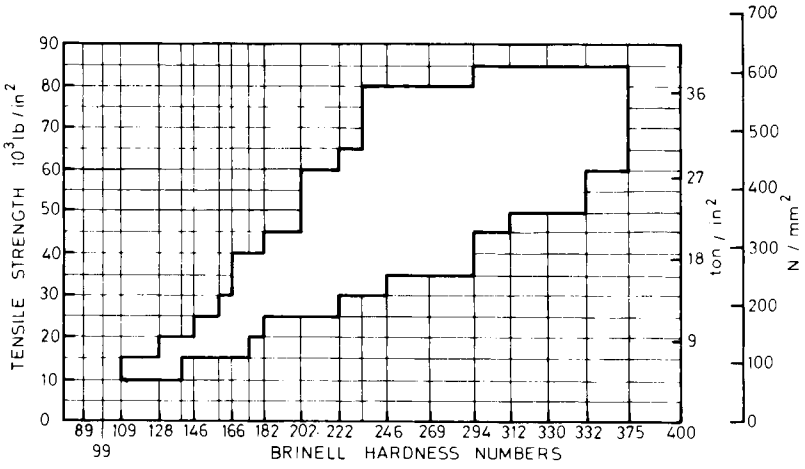


Figure 42. Relationship between tensile strength and Brinell hardness

Patterson[7] has proposed a basic hardness relation $HB = 100 + 6{\cdot}8T$ (or $0{\cdot}44T$), where T is the tensile strength expressed in tonf/in^2 (N/mm^2), and recognising that actually determined hardnesses may differ from that derived from the formula, utilises the relative hardness

$$\left(RH = \frac{\Delta}{100 + 6{\cdot}8T}\right) \quad \left(\text{or } \frac{\Delta}{100 + 0{\cdot}44T}\right)$$

where Δ is the determined hardness; the lower the relative hardness for a given tensile, the better the quality in terms of machinability and resilience.

These results confirm the absence of a clear relation between hardness and tensile strength.

In a heterogeneous material like cast iron, the hardness figure given by a Brinell impression is made up of the hardness of all the different structural constituents covered by the impression:

	HB	HV
Electrolytic iron[1]	70	
Silico-ferrite[9] (0·82% Si)	88	
Silico-ferrite[9] (2·28% Si)	124	
Silico-ferrite[9] (3·4% Si)	150	(180) 207 knoop[28]
Spheroidised pearlite[17]	130–180	
Normal pearlite (silico-ferrite + cementite)[10, 11, 17] depending on interlamellar distance	175–330	
Austenite (steel)[25]		239–450
Iron carbide Fe_3C[17, 20, 25]		800–1 080
Iron chromium carbide (<10% Cr)[20, 24] $(FeCr)_3C$		830–1 370
Iron chromium carbide (>13% Cr)[20, 24] $(FeCr)_7C_3$		1227–1475 (1800)
Tungsten carbide[23] WC[25]		1820–2470
WC_2[25]		3000–3400
Phosphide eutectic:		
pseudo binary[17, 20, 22, 25]		420–740 (av. 570)
ternary[20]		720–820
ternary[22]		810–835

The hardness of pseudo-binary phosphide eutectic is probably 530–600 HV. The higher value quoted above may include undisclosed ternary eutectic. The figure in any case is the approximate average of constituents of differing individual hardness.

The form and type of graphitic carbon also have a marked influence. In otherwise similar irons, an iron with coarse graphite is softer than one with fine graphite, and one with flake graphite is softer than one with nodular graphite. High graphite content lowers hardness and a 1·2 in (30 mm) bar with total carbon 3·7–3·9 per cent and silicon 1·2–1·5 per cent may give a hardness of 120 HV or less.

The degree of subdivision of the pearlite is also of importance[10] and may account for hardness variations between 175–180 HB at an interlamellar distance of 0·42 μm and 330 HB at an interlamellar distance of 0·24 μm[17]. Belaiew[27] quotes a maximum of 400 HB at interlamellar distance of 0·2 μm as in troostite.

In a coarse grained iron the small impression of a diamond indenter may be unduly affected by one structural constituent, and an accurate overall figure can only be obtained in these circumstances if the average of a large number of impressions is taken.

Phosphorus is one of the elements which produces a very large variation in hardness, irrespective of its effect on strength. Phosphorus is present in

cast iron as a low melting point eutectic with a hardness in the region of 420–700 HV. It frequently acts, however, as a centre for the deposition of associated carbides of iron, vanadium, chromium and molybdenum, and these carbides have hardnesses in the region of 800–1000 HV for Fe_3C, and up to 1475 HV for iron–chromium, iron–vanadium or iron–molybdenum carbides[25]. It has been stated that an increase of 1 per cent phosphorus raises the hardness of a casting by some 30 points Brinell[12]. The hardness of cast iron tends to increase as the following elements increase: phosphorus, manganese (above about 0·4 per cent), sulphur, chromium, molybdenum, vanadium. Manganese only increases hardness when present in an amount above that required to neutralise sulphur. The hardness tends to decrease as the following elements increase: graphitic carbon, silicon, nickel. Silicon and nickel only soften the iron by increasing the free graphite present in the structure. Their intrinsic effect on the matrix is to harden it, and in fully ferritic irons of the same structure, the lowest hardness will be obtained with low silicon or nickel content.

Any attempt to specify hardness together with other properties must be done with a full knowledge of the composition likely to be employed, the size of the casting to be tested, and the position at which the hardness test is taken.

Collaud[8] has shown a relationship between tensile (T), Brinell hardness (HB) and elastic modulus (E) expressed as $T = C \times HB \times E_o$ where the tensile and elastic modulus are expressed in similar terms (kg/mm^2, lb/in^2, N/mm^2 or tonf/in^2) and the value for E is that close to zero stress (E_o) which will be higher than that usually quoted (at 25 per cent maximum stress).

The constant C for grey iron[8,13] is usually between 9 and 10×10^{-6}.
The constant C for malleable iron[13] is usually between 15 and 16×10^{-6}.
The constant C for nodular cast iron[13] is usually between 16 and $16{\cdot}5 \times 10^{-6}$.

Some relationship between hardness and tensile strength can occur for closely defined conditions. For example, if a series of test bars of differing diameter are cast from the same ladle as the casting itself, the hardness of a tensile bar may be the same as the hardness of a portion of the casting with the same tensile strength as the original bar[14,15,16].

The variation of tensile strength and Brinell hardness with change of section has been examined by Collaud[8] who states that an unknown tensile (T_2) can be calculated from the hardness (HB_2) if the tensile (T_1) and the hardness (HB_1) of a test bar from the same cast are known, by using the relationship where T is expressed in kg/mm^2:

$$\frac{\log T_1 - \log T_2}{\log HB_1 - \log HB_2} = 1{\cdot}57$$

This relationship, however, is not always very exact and has the limitation that the hardness in the centre of a heavy section cannot be taken without destroying the castings, so that the tensile strength at this point cannot be determined except inferentially by means of a curve of the type shown in *Figure 39*.

The hardness test, however, can be usefully employed in conjunction with other properties, particularly in the production of repetition castings of controlled composition. Its inclusion in the specification with tensile strength, however, must be made with a full knowledge of the attendant factors which can affect the structure and properties of the casting.

Where hardness tests are used for acceptance tests, the presence of a decarburised layer may give artificially low readings, particularly in castings over 1 in (25 mm) in section. In such cases the surface layer should be ground off to a depth of about $\frac{1}{16}$ in (1·5 mm) for castings of about 1 in (25 mm) section, 0·1 in (2·5 mm) for castings of about 2 in (50 mm) section and $\frac{1}{8}$–$\frac{3}{16}$ in (3–5 mm) for castings of 3 in (75 mm) section of above, before the impression is made.

The following are typical hardness ranges for different classes of castings:

Ingot moulds	90–160 Brinell
Fully annealed grey iron engineering castings (the lowest Brinell can only be obtained with phosphorus less than 0·1 per cent and silicon below about 1·7 per cent. Thus, an iron having a chemical analysis: total carbon 3·45%, silicon 1·57%, manganese 0·53%, sulphur 0·133%, phosphorus 0·036%, copper 0·05% and remaining elements 0·02% or less—when annealed to the fully ferritic state gave a Brinell hardness of 92 in a section of approximately $1\frac{3}{8}$ in (35 mm)—ASTM graphite Type A, Size 4)	90–160 Brinell
Grey pearlitic irons, in sections $\frac{3}{4}$–$1\frac{1}{2}$ in (19–38 mm), unalloyed	180–240 Brinell
Grey pearlitic irons, in sections $\frac{1}{4}$–$\frac{3}{4}$ in (6–19 mm), unalloyed	190–250 Brinell
Light castings up to $\frac{1}{4}$ in (6 mm), high phosphoric, unannealed	215–250 Brinell
Light castings up to $\frac{1}{4}$ in (6 mm), high phosphoric (above 1·0% phosphorus) fully annealed	180–195 Brinell
Unalloyed white irons for abrasion-resistance	400–550 Brinell
Alloyed white irons for abrasion-resistance, chilled rolls etc.	550–750 Diamond pyramid

Hardnesses between 250 and 350 Brinell are used only for special applications as they generally coincide with the presence of chill or mottle, making machining difficult. The acicular and martensitic grey irons are said to be machinable up to about 330 Brinell because their hardness is produced by general structural change and not by the presence of free carbides with hardnesses in excess of 800 Brinell. Certain special alloyed irons are capable of developing surface hardnesses in excess of 900 diamond pyramid.

There is no clear relationship between hardness and wear-resistance. In abrasive wear the resistance of the metal is dependent upon either its initial hardness or the surface hardness developed by work-hardening, and the hardness of the abrasive.

In lubricated sliding wear there is no clear relationship between hardness and wear and a separate section deals with these conditions.

Where hardness tests are used for acceptance tests, the presence of a decarburised layer may give artificially low readings, particularly in castings over 1 in (25 mm) in section. In such cases the surface layer should be ground off to a depth of $\frac{1}{16}$ in (1·5 mm) for castings about 1 in (25 mm) section, 0·1 in (2·5 mm) for castings about 2 in (50 mm) section, and $\frac{1}{8}$–$\frac{3}{16}$ in (3–5 mm) for castings of 3 in (75 mm) section or above, before the impression is made.

Table 9 APPROXIMATE COMPARISON OF HARDNESS SCALES

Diamond pyramid number (BS 427:1961)	*Brinell number (BS 240:1962)*	*Shore scleroscope number*	*Rockwell C (BS 891:1962)*
50	50	—	—
100	95	18	—
150	145	24	—
200	195	30	—
250	245	36	22
300	295	42	30
350	338	47·5	35
400	380	53	40
450	420	58	44·5
500	460–470	62–67	48
550	500–520	65–70	50–51
600	535–560	70–74	52–56
650	565–600	72–78	56–57
700	595–645	74–82	57–61
750	630–695	76–86	59–62
800	—	78–90	60–64
850	—	80–93	62–64
900	—	—	63–67
950	—	—	64–68
Likely Variations: Below 400: ±10 Above 400: ±17		±5	±3

The figures in *Table 9* for diamond pyramid number, Brinell and Rockwell C are based on the conversions given in BS 860:1967. The Shore scleroscope figures are based on various sources[18, 19, 21, 26]. No conversion tables of this type are entirely satisfactory and the above table may be regarded only as an approximation. Knoop (HK) values are in general 25 points higher than HV values for the same specimen.

Hot hardness determinations have been made[29, 30] and their significance discussed in the sections on Creep and Furnace Rollers.

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MODULUS OF ELASTICITY OF CAST IRON

The modulus of elasticity at room temperature of the metallic matrix components of cast iron and steel, namely ferrite, iron carbide, pearlite and phosphide eutectic[1] is approximately constant at 30×10^6 lb/in^2 (207 GN/m^2). The presence of graphite as flakes or nodules, however, gives dispersed discontinuities which modify the stress/strain response, as shown by Thum[2] who perforated a mild steel sheet with holes and slots and demonstrated that such dispersed discontinuities modified the characteristic stress/strain curve of steel to something approaching that found in cast iron.

The modulus of elasticity of graphite-free white iron[3] (pearlite and free carbide) is in fact about 30×10^6 lb/in^2 (207 GN/m^2), but the presence of dispersed discontinuities (graphite) in grey, malleable or nodular cast iron lowers

the apparent elastic modulus below that of the matrix, the effect being more pronounced with flake graphite grey irons.

Relationship between stress and strain in cast iron

For most general engineering purposes, cast iron is not stressed above 25 per cent of its tensile strength and within this limit the following figures (*Table 10* and *Figure 43*) for elastic modulus may be employed for the purposes of calculation with the British Standard grades of grey cast iron. At a stress of 25 per cent of the tensile strength, the proof stress level of 0·1 per cent will not be exceeded and at a stress of 40 per cent of the tensile strength a permanent set of 0·01 per cent will not be exceeded for irons of 14–23 ton/in^2 (216–355 N/mm^2) tensile (BS 18: 1971).

The figures apply to stresses in tension up to 25 per cent of the tensile strength and to stresses in compression up to 75 per cent of the tensile strength.

Table 10

	BS 1452 Grade						
	10	*12*	*14*	*17*	*20*	*23*	*26*
Approximate elastic modulus, lb/in^2 × 10^6	10–15	11–16	13–17	14–19	16–21	18–22	18–22
GN/m^2	69–103	76–110	90–117	97–131	110–145	124–152	124–152

The range given for each grade covers the likely effect of graphite content —the lower figures applying to the higher graphite contents.

These figures will apply to most engineering requirements, but the stress/strain characteristics of cast iron are much more complicated than for steel, and the following notes summarise these differences, which may have considerable significance in some engineering applications.

The lateral response to stress is dealt with under the section headed 'Poisson's ratio'.

Stress/strain curve for steel

It is well established that steels having a variety of ferrite-pearlite structures, exhibit a modulus of elasticity more or less constant at a value of about 30×10^6 lb/in^2 (207 GN/m^2) and that heat treatment does not significantly modify this figure.

In fact, ferrite, pearlite, cementite and phosphide entectic all have moduli of elasticity[1, 3] of approximately 30×10^6 lb/in^2 (207 GN/m^2), so that the matrix structures of both grey and white irons also have the same modulus,

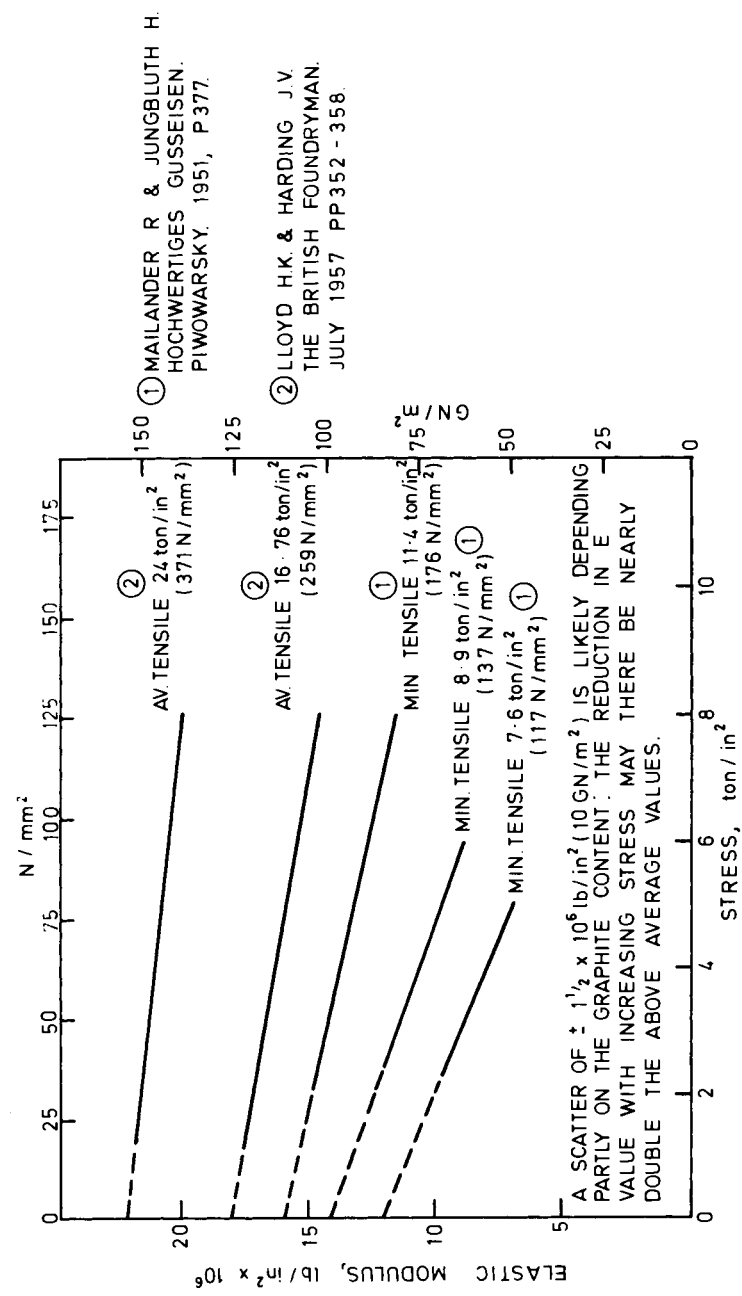


Figure 43. Elastic modulus of cast iron. Variation with tensile stress for irons of different strength (1 ton/in² = 15·2 in bar)

independent of composition. Interruption of the matrix by graphite or discontinuities, however, reduces the effective modulus by an amount dependent upon the form and quantities of discontinuities.

The characteristic stress/strain curve of mild steels is shown in *Figure 44*. In region 'a' the material follows Hookes Law, i.e. strain is directly proportional to the stress imposed and can be regarded as truly elastic.

When the stress reaches region 'b' sudden yielding occurs with little change in stress. In region 'c' the stress rises again with greatly increased strain until 'necking' occurs about point 'd', sufficiently to reduce the cross-sectional area of the specimen and the applied load falls as the specimen

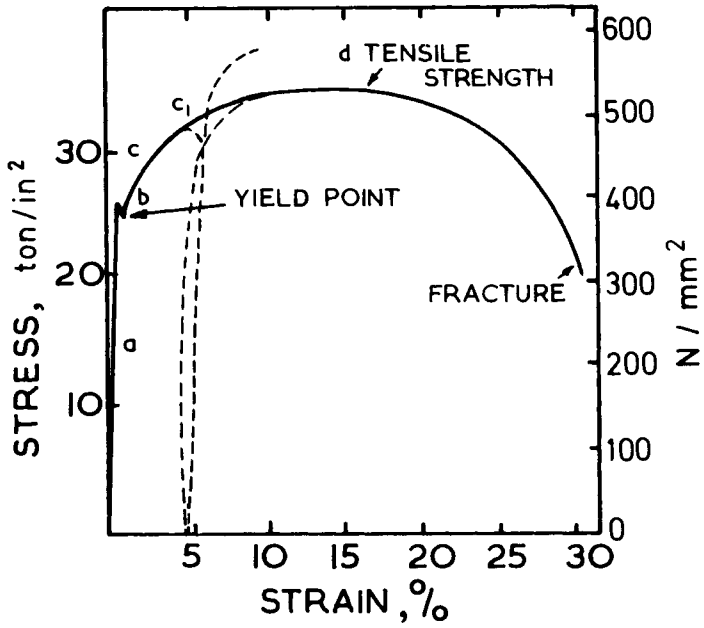


Figure 44. Typical stress/strain curve for carbon or low alloy steel

extends rapidly to final rupture, although the true stress, related to the reduced area of the specimen, continues to rise to the point of rupture.

The tensile strength is taken as the maximum load at point 'd' related to the original cross-sectional area.

The limit of proportionality occurs at about 0·9 of the yield point 'd' for rolled or normalised steel.

If the test bar is unloaded within the range of plastic deformation, e.g. at point 'c', it will exhibit a curve of recoverable strain for the unloading and reloading cycle up to or near the original load, after which plastic straining will be resumed. After a period of rest up to thirty days, or heat treatment at about 100°C for a much shorter period the material resumes elastic response at the original modulus of elasticity (30×10^6 lb/in^2 or 207 GN/m^2), but with a higher limit of proportionality[4].

This precise pattern of events does not occur with any cast irons. There

is no true yield point for grey, ferritic nodular, ferritic malleable or austenitic nodular cast irons, and significant necking occurs only with ferritic nodular iron and blackheart malleable irons just before rupture.

Stress/strain curve for cast iron

The presence of graphite in the flake, nodular or aggregate form modifies the stress distribution within the casting so that no sudden yield occurs,

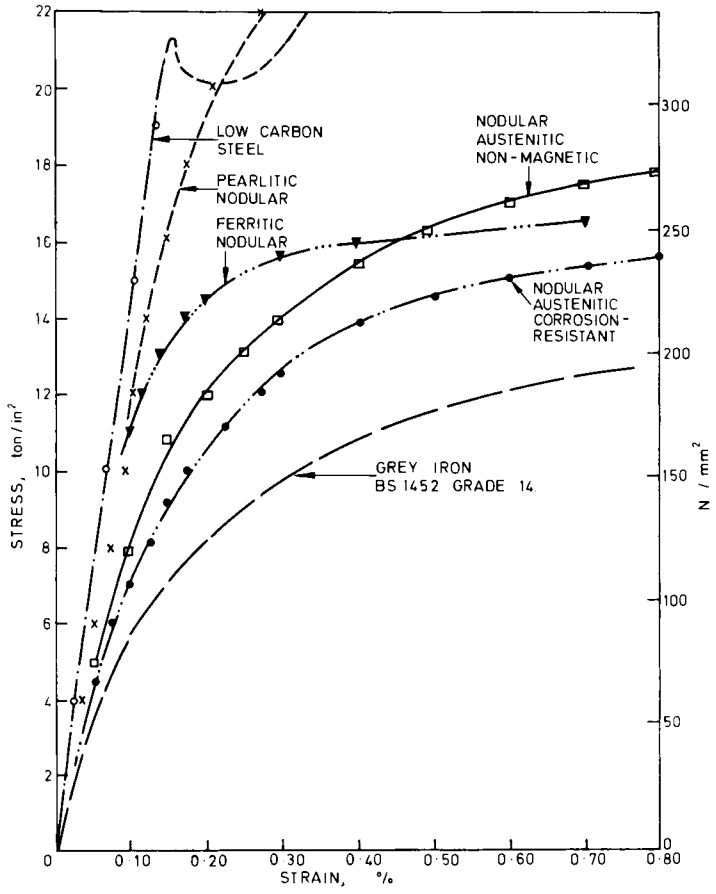


Figure 45. Typical stress/strain curves in tension for various cast irons and steels

but the curve of plastic deformation even in the ductile cast irons (e.g. ferritic or austenitic) merges into the 'elastic' portion. The use of the term 'yield point' is, therefore, inapplicable to cast irons.

Figure 45 shows stress/strain curves for mild steel and various types of cast iron which illustrate important differences in their response to stress.

GREY CAST IRON (FLAKE GRAPHITE)

In grey cast iron the matrix is broken up by graphite and when subjected to simple stress the matrix carries a complex stress system with regions of high local stress at the graphite flakes, so that the recoverable

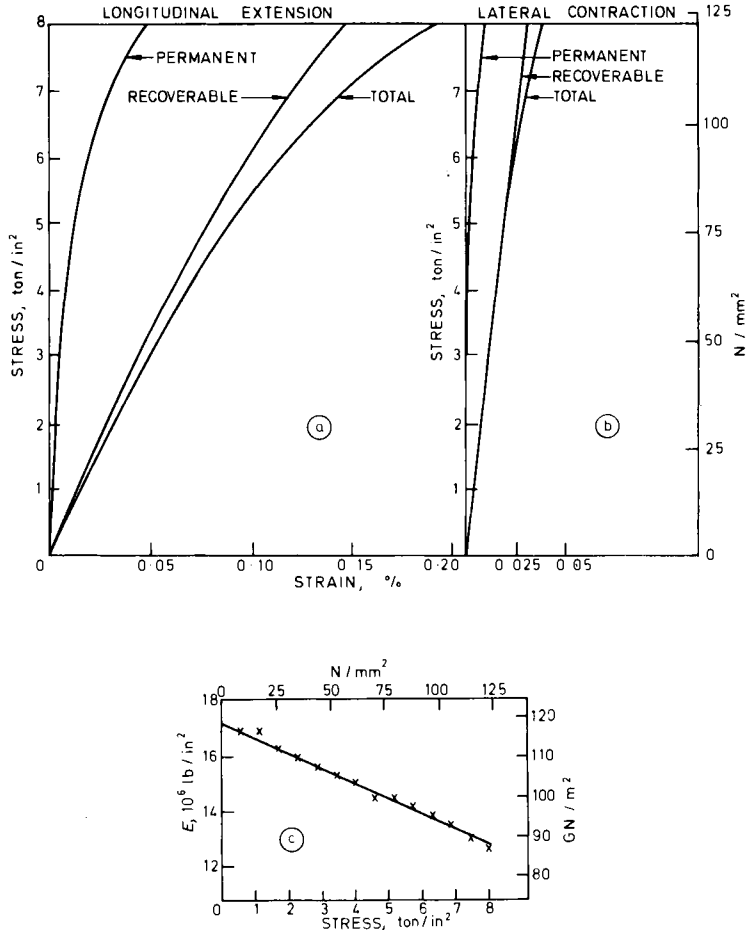


Figure 46. Stress/strain characteristics of a grey iron to BS 1452 Grade 17 tested on a 2·1 in bar: actual tensile 13·8 ton/in² (213 N/mm²) (a) Longitudinal stress/strain curve in tensile; (b) Lateral stress/strain curve in tensile; (c) Elastic modulus E (longitudinal) at different levels of stress

strain is not a straight line, but curved as shown in *Figure 45*. At no stress does grey cast iron show a truly elastic response. At the first application of every increment of load some plastic deformation occurs and the recoverable component of total strain does not follow a straight line, but a curve of recoverable strain.

The conventional method of plotting the stress/strain curve of grey iron

in tension is shown in *Figures 46a* and *b*, in which the plastic or permanent and the recoverable strains are shown together with the total strain.

The plastic or permanent strain is determined at several points by returning the load to zero, measuring the permanent set and plotting the curve of permanent strain; the curve of recoverable strain is the difference between the total and permanent strain.

The elastic modulus is obtained by plotting the slope of the curve of recoverable strain against stress as shown in *Figure 46c* and extrapolating to zero stress. The value of E at zero stress is known as the E_0 value, and in the following text the E value (secant) at a given stress is denoted by E_x.

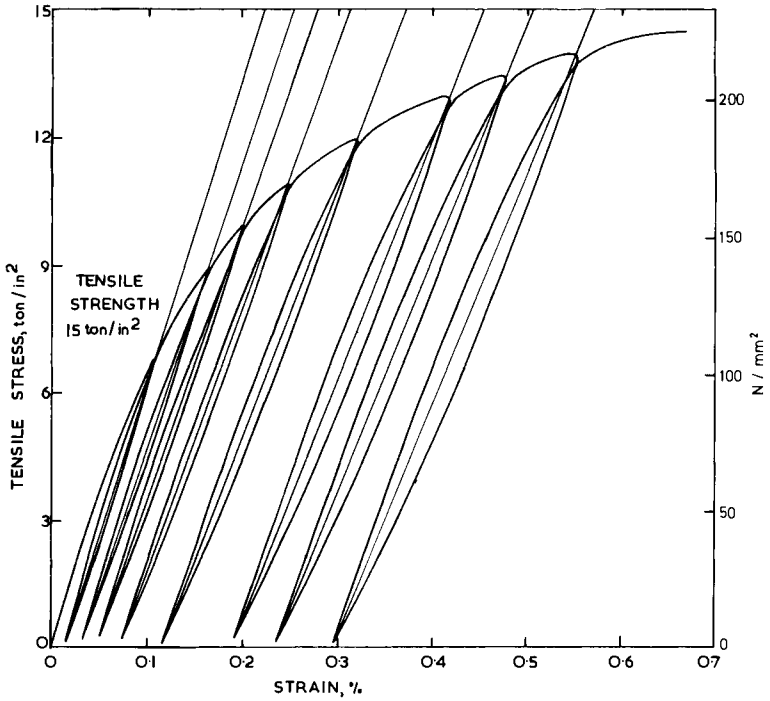


Figure 47(a). Tensile stress/strain curve of a flake graphite cast iron—as-cast condition

When the stress/strain curve for a grey iron is taken by applying successively increasing stresses and after each point returning to zero stress, curves of the type shown in *Figures 47a* and *b* are found[5]. This curve shows that not only is there plastic deformation at each increment of load, even the recoverable component is not a straight line, but that the strain curve follows one path while stress is decreasing and another while stress is increasing, thus forming a loop of recoverable strain and not the straight line required by Hooke's Law. This effect may also be found in other metals, such as cold worked steel (0·1–0·75 per cent carbon), cold worked brass etc., and has been discussed by many workers[4, 6].

In grey cast iron this effect occurs because the matrix is broken up by

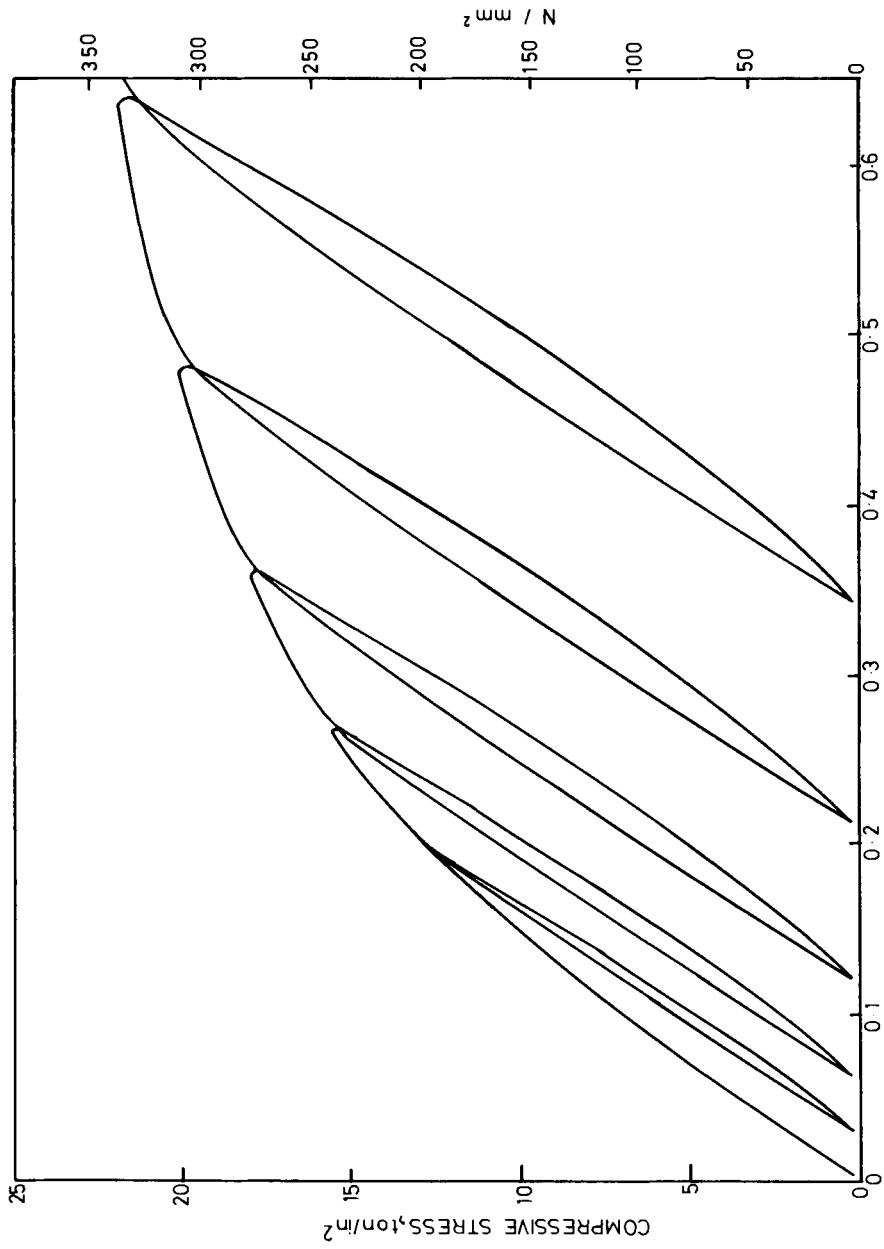


Figure 47(b). Compressive stress/strain curve of a flake graphite cast iron—as-cast condition

graphite and when it is subjected to simple stress, the matrix carries a complex stress system with regions of high local stress at the graphite flakes[7].

The cavities which contain the graphite flakes open up slightly under tension and local plastic deformations of the matrix occur even at low applied stresses, so that the total tensile strain is made up of four components[8, 9, 10]:

(1) Purely elastic deformation of the matrix.
(2) Plastic deformation of the matrix at points of high stress.
(3) Recoverable strain due to the opening of the graphite cavities.
(4) Permanent strain associated with the opening up of the cavities.

On unloading, the purely 'elastic' component is recoverable, but partial closure of the graphite cavities adds a further recoverable component so that the recoverable strain is not a straight line, but curved as shown in *Figure 47*.

If the specimen is stabilised by repeated application of a load equal to or higher than that to be ultimately applied, a curve of recoverable strain is obtained that will apply to any lower load (*Figure 48*). It will be seen also in *Figure 47* that the slope of the recoverable curve will be slightly reduced as the applied stress is increased. A similar pattern of curves occurs with compressive stress, but the change of slope with increasing stress is very small[11].

The value of the effective modulus of elasticity E_x obtained at any particular stress x will, therefore, depend upon the maximum value to which the test piece has previously been stressed. The higher the prestressing treatment, the lower will be the effective modulus.

However, the modulus gives only the relationship between stress and recoverable strain and although the modulus is reached by high levels of prestressing, it does ensure that little permanent strain subsequently occurs in service at stresses below this value.

These effects, however, are small under normal service conditions and can usually be ignored if the working stress does not exceed 0·25 of the tensile strength. Thus, for an iron of 14 ton/in^2 (216 N/mm^2) tensile *Figure 49*[12] shows the effect of prior loading on the modulus of elasticity E_x. From the curve, the effect of prestressing on the modulus of elasticity E_x at a working load of 4 ton/in^2 (62 N/mm^2) is shown below:

	Prior load	$E_0 \times 10^6$ lb/in^2 (GN/m^2)	$E_x \times 10^6$ lb/in^2 GN/m^2)
Initial loading	Nil	16·05 (111)	14·1 ($x=4$) (97)
After 5 loadings to 4 ton/in^2 (62 N/mm^2)	0·28	15·7 (108)	14·0 ($x=4$) (97)
After 20 loadings to 8 ton/in^2 (124 N/mm^2)	0·56	14·0 (97)	13·0 ($x=4$) (90)

Since it is unusual for castings to be stressed in service above about 0·25 of the tensile strength, it will be seen that for most purposes this change in modulus of elasticity can be ignored.

The modulus of elasticity may, therefore, be expressed in terms of the tangent to the stress/strain diagrams at or near zero stress, i.e. the E_0 value, or as the secant modulus E_x at a given stress x which is the mean slope of the curve of recoverable strain from this stress (*Figure 47*).

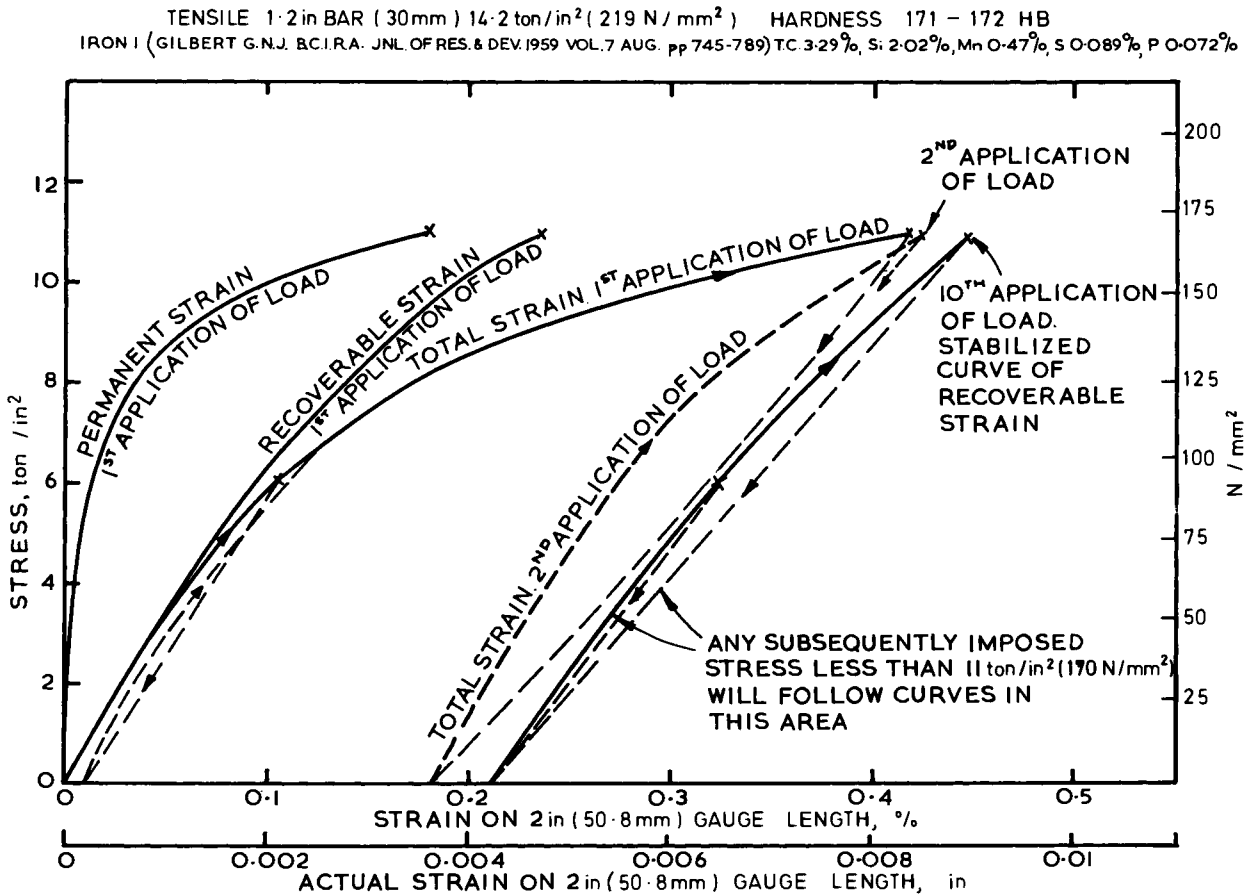


Figure 48. Typical stress/strain curves for the 1st application of stress of 11 ton/in² (170 N/mm²) and the stabilised stress/strain curve after 10 applications of a stress of 11 ton/in² (170 N/mm²) tensile

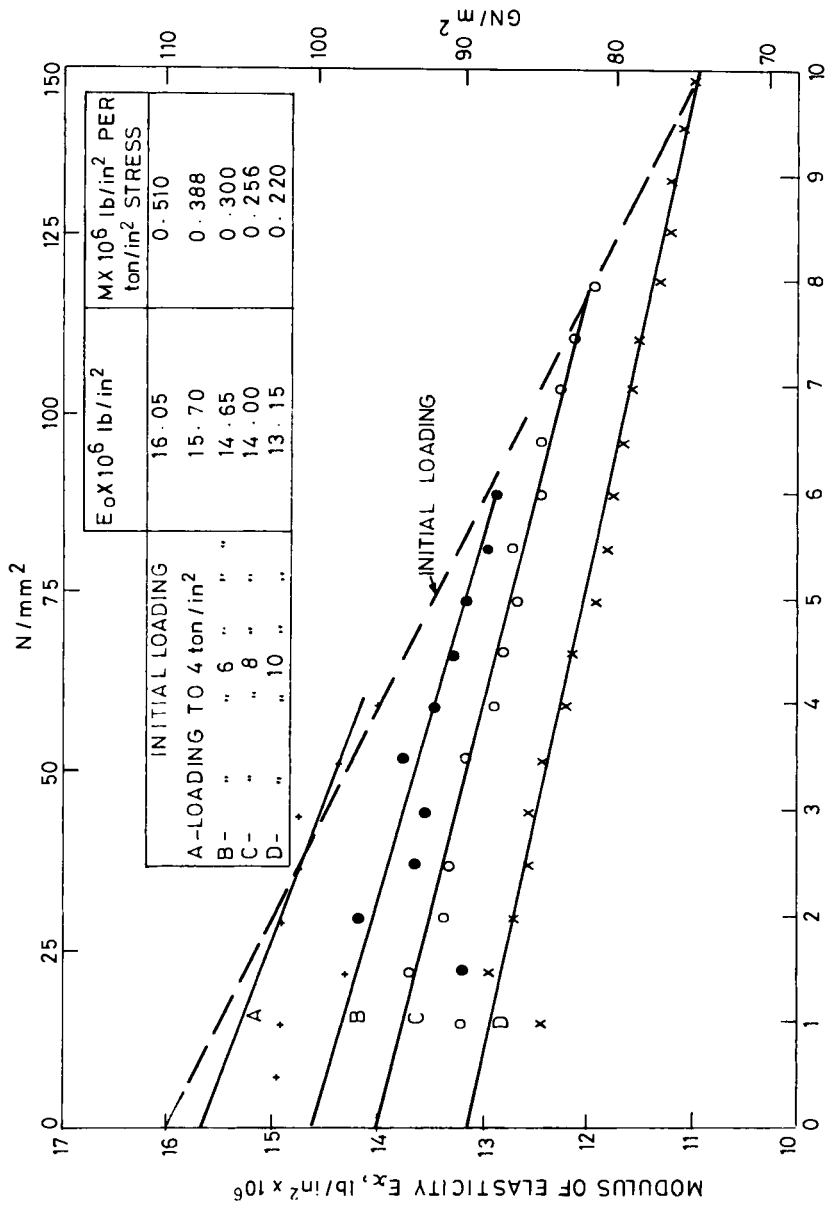


Figure 49. The variation of the modulus of elasticity E_x with stress for a test bar stabilised at a series of increasing stresses

Figure 43 indicates some determined values in tension of E_0 and E_x for various grades of iron which can be regarded as typical.

Dynamic testing

The modulus of elasticity can be determined dynamically by sonic vibration testing and gives slightly lower values than static tests on previously unstressed bars and higher values than static tests on previously stressed bars. The two values differ from each other by less than 10 per cent even for bars previously stressed to 60–75 per cent of the tensile strengths[11].

Factors influencing the modulus of elasticity

GRAPHITE

For irons of the same tensile strength, the modulus tends to fall as the graphite content rises[13].

FORM OF GRAPHITE

The effect of flake graphite is much greater than that of nodular or aggregate graphite, as shown in the following values for $\frac{E \text{ cast iron}}{E \text{ steel}}$:

Cast iron with nodular graphite	0·73–0·87
Malleable cast iron with nodular or aggregate graphite	0·75–0·9
As-cast pearlitic flake graphite iron depending on graphite content	0·4–0·7
Annealed ferritic flake graphite iron depending on graphite content	0·35–0·65

PHOSPHORUS

The phosphorus content has little effect on the E_0 value, but rising phosphorus tends to increase[14] the value of the secant modulus E_x, by decreasing the curvature of the stress/strain curve and markedly reduces elongation or strain at failure.

MATRIX STRUCTURE

For irons of the same graphite form and content, the E_0 value should not vary whether the matrix is ferritic or pearlitic. Annealing of a pearlitic iron to the ferritic state, however, causes deposition of graphite on to the existing

graphite flakes by decomposition of the carbide. The effect of this increase in graphite content is to lower the modulus of elasticity by 8 to 15 per cent[15, 16, 17].

The effect of graphite form (flake and undercooled), phosphorus content and matrix structure is illustrated in the following tests, in which irons of almost identical composition with high and low phosphorus contents (*Table 11*) were tested in the as-cast and annealed states[18].

Table 11 BASIC COMPOSITION

TC %	*Si* %	*Mn* %	*S* %	*P* %	*CEV*
3·05	1·95	0·57	—	0·25	3·77
2·99	1·91	0·49	—	0·65	3·70

These were either inoculated to produce flake graphite or treated with about 0·15 per cent titanium to produce undercooled graphite.

Table 12

	Tensile ton/in² (N/mm²)		E_0 lb/in² × 10⁶ (GN/m²)		*HB*		*Total strain at failure* %		*Eutectic cell count* No./in²	(No./cm²)
Phosphorus content	0·25	0·65	0·25	0·65	0·25	0·65	0·25	0·65		
NORMAL GRAPHITE:										
As-cast pearlitic	17·15 (265)	18·9 (292)	19·65 (135)	19·5 (134)	212	227	0·73	0·56	1260	(195)
Annealed ferritic	9·83 (152)	12·67 (196)	18·1 (125)	18·14 (125)	104	133	1·36	0·67	1260	(195)
UNDER-COOLED GRAPHITE:										
As-cast	13·84 (214)	14·95 (231)	21·45 (148)	20·64 (142)	222	231	0·332	0·273	601	(93)
Annealed	11·63 (180)	13·43 (207)	20·32 (140)	19·1 (132)	131	148	0·45	0·457	601	(93)

Normal graphite structures as-cast show higher tensiles, lower hardness, lower E_0 values, and higher total strain at failure than undercooled graphite structures (*Table 12*).

Total strain at failure (elastic and plastic) falls as the tensile strength rises. Thus, for Grade 10 iron the value is from 0·6 to 0·75 per cent, for Grade 17 about 0·58 per cent, and for Grades 20 and 23 about 0·50 per cent.

Higher phosphorus irons show higher tensiles, higher hardness, lower total strain at failure and little change in E_0 values.

Annealing to the ferritic state lowered tensile, hardness and E_0 values and

increased total strain to failure. The undercooled graphite structures when annealed showed higher tensile strengths than normal graphite.

SECTION THICKNESS AND EUTECTIC CELL SIZE

The E_0 value varies little with eutectic cell size or when the same metal is cast into light and heavy sections with fully pearlitic structures, although the change in the tensile strength may be considerable. As the section thickness increases, however, the curvature of the stress/strain curve increases and the E_x values will decrease. If increasing section thickness increases graphite content, the E_0 values will also decrease[18].

Collaud[5] has shown a relation between tensile (T), Brinell No. (HB) and elastic modulus (E_0) expressed as $T = C \times HB \times E_0$, where the tensile and E_0 values are expressed in similar terms (lb/in^2, ton/in^2, kg/mm^2, N/mm^2).

The constant C for pearlitic grey iron[15] is usually between 9 and 10×10^6
The constant C for malleable iron[15] is usually between 15 and 16×10^6
The constant C for nodular iron[15] is usually between 16 and $16{\cdot}5 \times 10^6$

COMPRESSION

In compression the strains are less affected by volume changes associated with the graphite flakes and the curve of recoverable strain is an approximate straight line at least up to 10–15 ton/in^2 (154–232 N/mm^2). A Grade 17

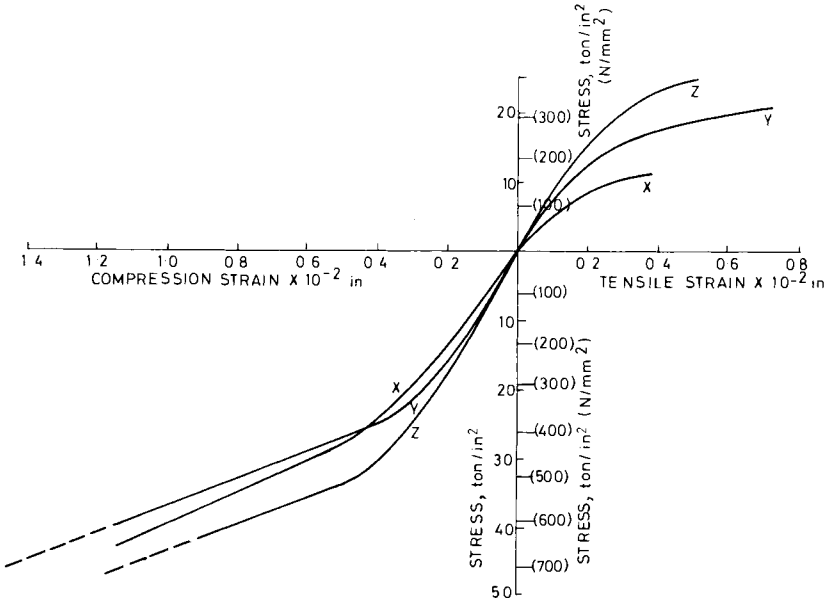


Figure 50. Stress/strain curves in tension and compression for materials X, Y, and Z (Table 13)

iron with an E_0 value of 16×10^6 lb/in^2 (110 GN/m^2) would have a value of $15{\cdot}5 \times 10^6$ lb/in^2 (107 GN/m^2) at 10 ton/in^2 (154 N/mm^2) compression. Prior stressing in compression does not affect appreciably the E value.

Three typical stress/strain curves in compression and tension are shown in *Figure 50* for the following irons:

Table 13

Material	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Cr* %	*CEV*	*Mean tensile strength 1·2 in bar (30 mm)* ton/in^2 (N/mm^2)		*HB*
X	3·14	2·81	0·64	0·077	1·26	—	—	4·5	10·2	(158)	198–226
Y	3·22	1·39	0·70	0·102	0·17	0·24	0·06	3·74	19·7	(304)	198–224
Z	2·50	2·03	0·55	0·039	0·06	1·17	0·06	3·2	24·3	(375)	213–236

A useful approximation for the elastic modulus of grey iron in tension and compression for stresses up to 0·25 of the tensile strength in tension and to 0·75 of the tensile strength in compression is quoted in *Table 10* at the beginning of this section.

Cyclic stressing in tension and compression has been fully discussed by Gilbert[12] and in the context of heat cracking by Dearden[19].

TRANSVERSE TEST

The E_0 values based upon the calculated transverse rupture stress appear to vary little (0·9–1·0) from those in tensile for irons varying from 11 to 20 ton/in^2 (170–309 N/mm^2) in tensile. The rate of falling off with increasing stress, however, is less than half of that in tensile[18, 20].

TORSION

The modulus of rigidity for ordinary cast irons of 12–14 ton/in^2 (185–216 N/mm^2) tensile is approximately $6{\cdot}9 \times 10^6$ lb/in^2 (48 GN/m^2) and for 18–20 ton/in^2 (278–309 N/mm^2) tensile irons, about $7{\cdot}9 \times 10^6$ lb/in^2 (54 GN/m^2). The modulus of rigidity is, therefore, approximately 0·38–0·43 times the modulus of elasticity in tension[18].

SHEAR

The modulus of elasticity in shear of grey cast iron ranges from 5 to $7{\cdot}5 \times 10^6$ lb/in^2 (34 to 52 GN/m^2) rising with tensile strength, e.g. an iron with 23 ton/in^2 (355 N/mm^2) tensile showed a modulus of elasticity in tension of 18×10^6 lb/in^2 (124 GN/m^2) and $7{\cdot}5 \times 10^6$ lb/in^2 (52 GN/m^2) in shear.

MALLEABLE IRONS

The elasticity modulus values in tension, compression and shear are shown in *Table 14*.

Table 14

	Malleable cast irons *Approximate modulus of elasticity* $lb/in^2 \times 10^6$ (GN/m^2)			
	Whiteheart		*Blackheart*[21, 22]	
In tension	27–28	(186–193)	24–27	(165–186)
In compression	26–27	(179–186)	22·5–25·0	(155–172)
In shear	9–10	(62–69)	12·5	(86)
In torsion	—		10·7	(74)

WHITE IRON

White iron [tensile 31–40 ton/in^2 (479–618 N/mm^2)] which is free from austenite may be regarded as an elastic material, whose modulus of elasticity[3] in tension and compression is 30×10^6 lb/in^2 (207 GN/m^2). When, owing to the presence of alloying elements, austenite is present, a very slight amount of plastic deformation may take place before fracture, and the modulus may drop slightly to 24–25 $\times 10^6$ lb/in^2 (165–172 GN/m^2). No figures for compression are available but are probably identical. Elastic strain at failure is about 0·23–0·27 per cent. Permanent strain at failure is about 0·15–0·36 per cent. Total strain at failure is about 0·4–0·6 per cent.

Dimensional stability of cast iron under stress

For normal engineering requirements, the plastic deformation on first application of stress of flake graphite grey iron and austenitic flake and nodular irons can be ignored. For large castings, where the maintenance of the higher dimensional accuracy is required—as in some machine tools—initial deformation can be virtually eliminated by prestressing for a suitable period of time to a stress higher than that to be subsequently imposed.

The effect of prestressing in tension on the stress/strain curve for grey iron has been studied by Gilbert[14] and the results are summarised in *Figure 48* which shows that prestressing a Grade 14 iron ten times to 10 ton/in^2 (154 N/mm^2) has produced a stabilised loop of recoverable strain for lower stresses and the same effect would be found for lower prestress loads.

Tottle[23] demonstrated in grey iron the occurrence of room temperature creep in bending, and Gut[24] has also studied this effect in bending of separate specimens and also a machine tool bed using cyclic and continuous preloading, and his results on the effect of prestressing are summarised in

Figure 51 for the complete casting, which was mounted centrally and loaded at both ends. Ten loading cycles were employed in which 1000 kg were applied ten times over 10 h, each cycle being of 30 minutes duration. The curve shows that the load for substantially elastic response has been increased from about 150 kg on first loading to about 750 kg after the cycles of prestressing were complete.

For castings requiring high dimensional accuracy, therefore, the advantage of prestressing above the normally applied loads before final machining may be significant.

NODULAR AUSTENITIC CAST IRON

Reference to *Figure 45* shows that the nodular austenitic irons exhibit a non-elastic response to initial stress similar to the grey irons and not the elastic response found with the ferritic and pearlitic nodular irons. Prestressing into the plastic range for a suitable period of time to a stress higher than that ultimately to be employed, however, will leave the material with an elastic response to stresses lower than those employed during prestressing (*Table 15*).

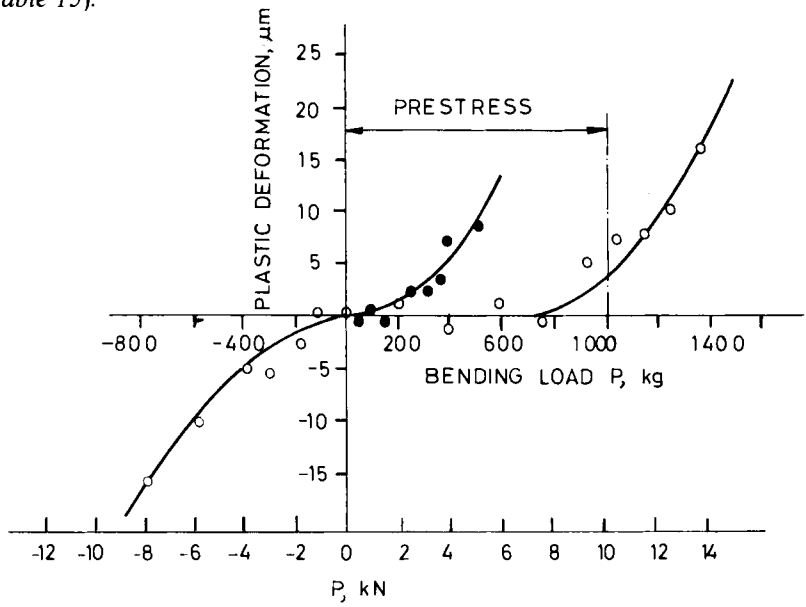


Figure 51. Test on machine tool bed. Influence of 10 cycles of load 0–1000 kg (0–9·81 kN) in 10 h

To illustrate this effect, samples were taken from an austenitic nodular iron casting [sections $1\frac{7}{8}$ in (48 mm)] of the following approximate composition:

Total carbon	3·35%
Silicon	2·4%
Manganese	5·0%
Phosphorus	0·07%
Nickel	12·3%

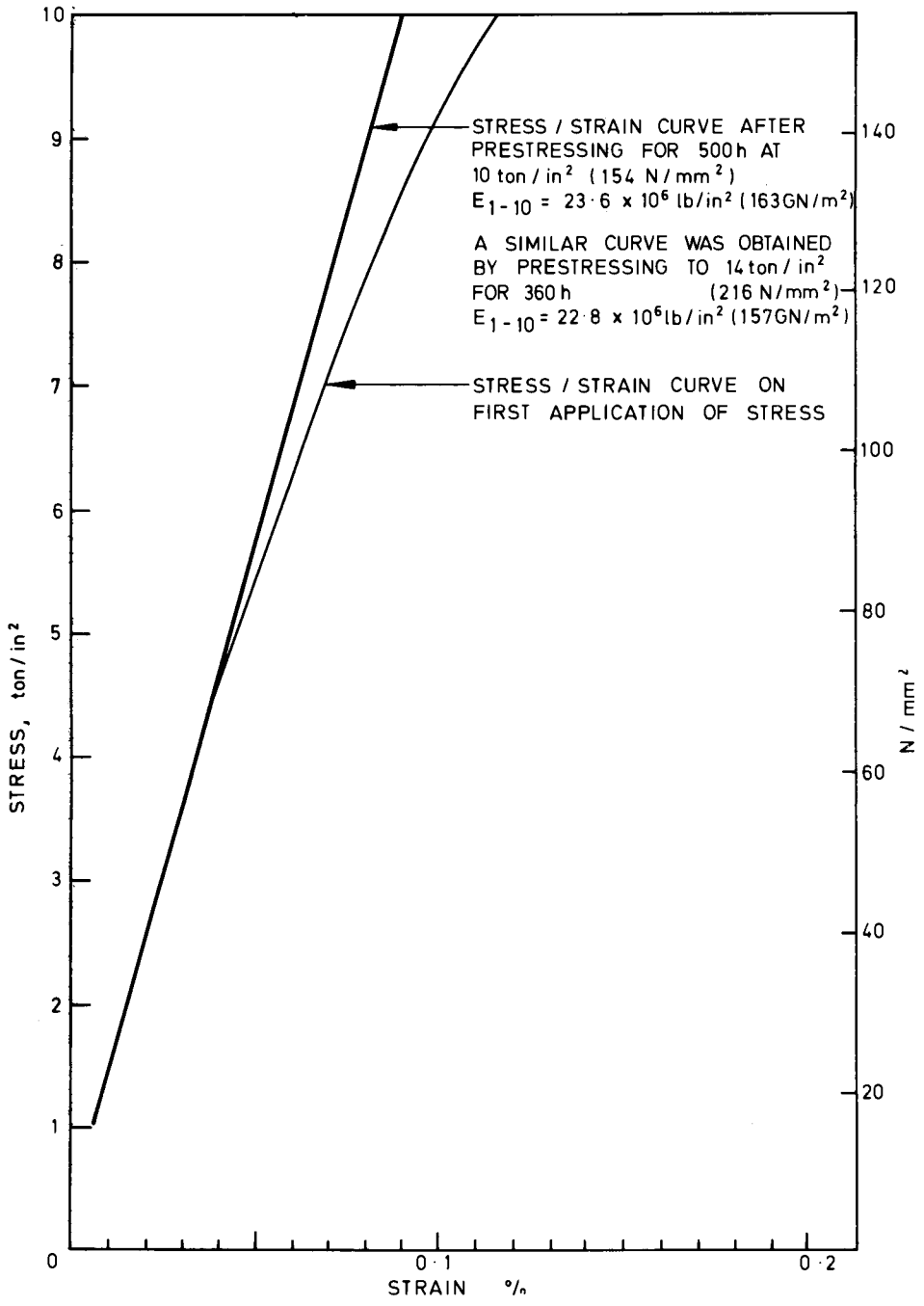


Figure 52. Effect of prestressing an austenitic nodular iron (Nodumag) on the stress/strain curve

Table 15 MECHANICAL PROPERTIES

	Tensile strength ton/in^2 (N/mm^2)	*Elongation* %	*0·2% proof stress* ton/in^2 (N/mm^2)	*E* lb/in^2 × 10^6 (GN/m^2)
0·875 in (22 mm) cast bar	26·5 (409)	14	15 (232)	21·3 (147)
Casting before prestressing	20·9 (323)	6	14·7 (227)	24·4 (168)

The stress/strain curve on a specimen cut from the casting is shown in *Figure 52* for the first application of stress.

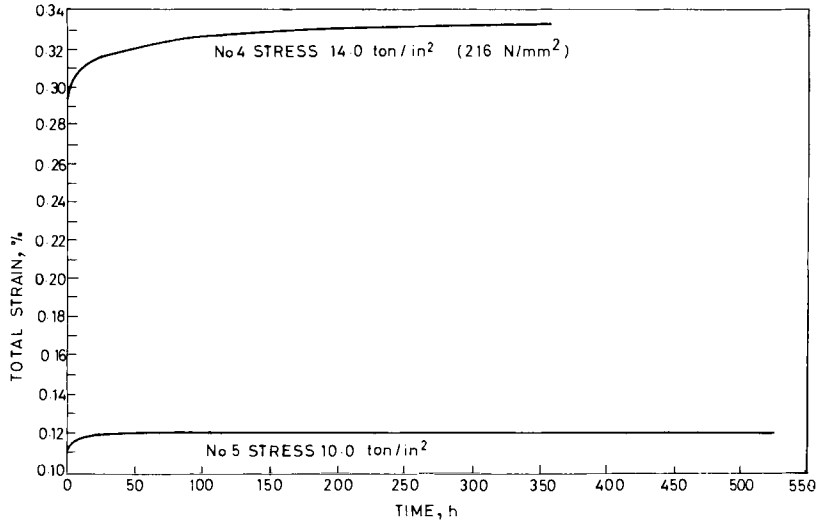


Figure 53. Creep tests at room temperature of test pieces cut from Nodumag casting section 1⅞ in (48 mm). Per cent total strain/time

Similar specimens were then loaded in a creep testing machine for periods up to 500 h at stresses of 14 and 10 ton/in^2 (216 and 154 N/mm^2) respectively and the total strains are recorded in *Figure 53*. At a stress of 10 ton/in^2 (154 N/mm^2) creep occurred up to about 50 h after which no further creep occurred up to 500 h. At 14 ton/in^2 (216 N/mm^2) creep was still continuing after 350 h. Both specimens were then removed and a further stress/strain diagram determined up to 10 ton/in^2 (154 N/mm^2) stress with load increments of 1 ton/in^2 (15 N/mm^2), the load being released after each increment down to a base load of 1 ton/in^2 (15 N/mm^2) (*Figure 52*). It will be seen that the stress for substantially elastic behaviour has been increased from about 4·5 ton/in^2 (69 N/mm^2) to at least 10 ton/in^2 (154 N/mm^2) with an *E* value of 23 × 10^6 lb/in^2 (159 GN/m^2) for both bars, with no significant hysteresis. This has been obtained by prestressing to give between 0·12 and 0·33 per cent permanent strain, so that the deformation to total failure is virtually unimpaired.

If a casting in this material were normally stressed up to 5 ton/in^2 (77 N/mm^2)—a factor of safety against rupture of 4—prestressing in the manner indicated would ensure that overloads up to 10 ton/in^2 (154 N/mm^2) would not produce significant deformation on return to normal loading.

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POISSON'S RATIO

When a specimen is subjected to tensile stress, the corresponding tensile strain or elongation is accompanied by a lateral strain across the section of the specimen which is shown by a reduction in the cross-section itself. The converse occurs in compression.

The ratio of elastic lateral strain across the bar to the elastic longitudinal strain is known as Poisson's ratio and in elastic materials the theoretical value is 0·25 and does not vary with stress. For steel a value of 0·3 is usually employed. Since cast iron does not show elastic response to stress, Poisson's ratio may vary with stress and grade of iron. It tends to fall as the tensile stress increases, but remains constant in compression until high stresses are imposed—more than 12 ton/in^2 (185N/mm^2)—when the value will rise.

Grey iron decreases in density under tensile stress owing to opening up of graphite cavities[1, 2] and this may also occur with graphite at certain orientations even in compression.

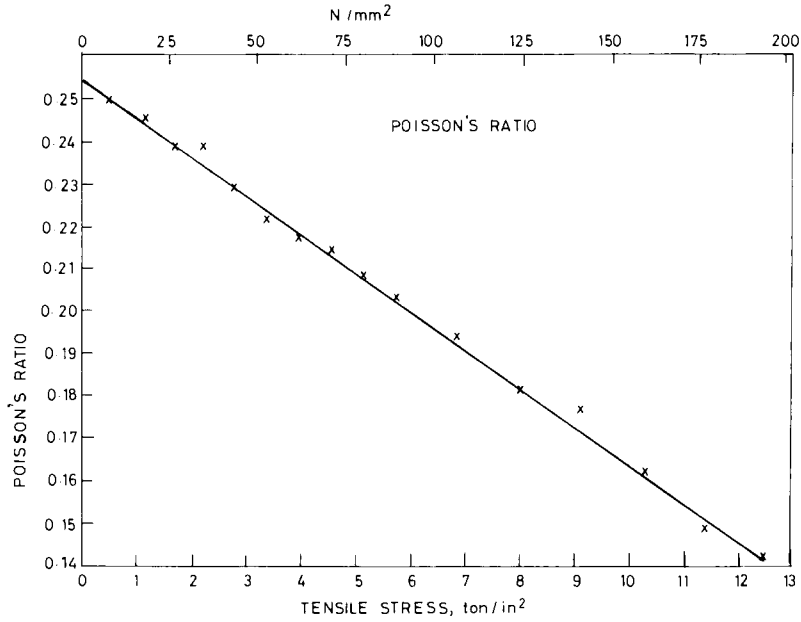


Figure 54(a)

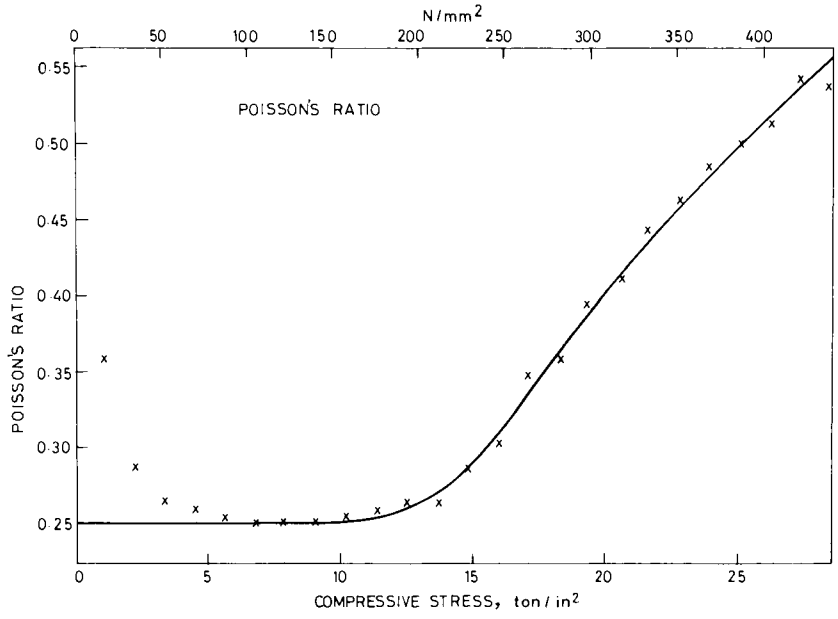


Figure 54(b)

Gilbert[2, 3, 4] has studied in detail the change of volume with stress and the variation of Poisson's ratio with stress for a typical engineering iron of 13·8 ton/in^2 (213 N/mm^2) tensile [1$\frac{3}{4}$ in (44 mm) bar] equivalent to BS 1452 Grade 14, and his results are summarised in *Figure 54a*, which shows that at zero stress a value of 0·25 is found, decreasing to about 0·22 at a tensile stress of 4 ton/in^2 (62 N/mm^2) which is the normal maximum likely to be used in practice. For tensile stresses the fall in Poisson's ratio varies linearly with stress.

For compressive stresses (*Figure 54b*) the value remains substantially constant at 0·25 until the stress exceeds about 12 ton/in^2 (185 N/mm^2), when the value rises sharply.

Poisson's ratio determined dynamically by sonic and ultrasonic methods gives results on as-cast or prestressed material 10–20 per cent lower than statically determined values.

A series of tests on very coarse graphite irons was made by Woolman[5], as shown in the following tables:

Table 16 (THE MECHANICAL PROPERTIES OF THESE IRONS ARE SHOWN IN TABLE 17)

Sample no.	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*ASTM graphite size*	*Matrix*
3	3·52	1·32	0·61	0·082	0·066	2–3	Pearlitic
12	3·82	1·49	0·43	0·086	0·076	1–3	Mainly pearlitic
4	3·84	1·01	0·66	0·065	0·21	1	Mainly pearlitic
14	4·12	2·07	1·03	0·051	0·036	1	Mainly pearlitic

Table 17

Sample no.	*Tensile strength* ton/in^2 (N/mm^2)		*Young's modulus (mean slope to 3 ton/in^2)* lb/in^2 × 10^6 (GN/m^2)		*Applied load* ton/in^2 (N/mm^2)		*Poisson's ratio at various temperatures*		
							Room temp.	450°C	650°C
3	10·3	(159)	13·1	(90)	1–4	(15–62)	0·21	0·14	0·25
12	7·7	(119)	10·5	(72)	1	(15)	0·25	0·14	0·18
					4	(62)	0·175	—	—
4	7·7	(119)	6·95	(48)	1	(15)	0·150	0·07	0·08
					4	(62)	0·125	—	—
14	5·7	(88)	5·2	(36)	1	(15)	0·12	0·05	0·09
					4	(62)	0·08	—	—

The figures were taken after repeated loadings. The two very coarse graphite irons (Samples 4 and 14) showed a marked fall in Poisson's ratio between the first, second and third loadings, and the values fall as the load increases. Woolman also gives values at 450 and 650°C.

Nodular iron in the as-cast, normalised and ferritic state shows values for Poisson's ratio within the elastic range of 0·26 to 0·29[6].

Malleable iron

Values of 0·17 have been quoted[7] for blackheart malleable cast iron, but there is doubt about this value and there is no reason to suppose that it differs significantly from the value of 0·26 to 0·29 found for ferritic nodular cast iron.

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LIMIT OF PROPORTIONALITY

Although the section dealing with elastic modulus makes it clear that there is no true limit of proportionality for any grey iron, nevertheless, an approximation to such a value can be usefully obtained by using the value for stress which produces a 0·01 per cent strain deviation from the tangent at the origin (E_o) of the stress/strain curve (*Figure 46*).

A useful series of tests was carried out by Lloyd and Harding[1] relating tensile strength, modulus of elasticity, limit of proportionality and permanent set, and the results are summarised in *Tables 18* and *19*.

Limit of proportionality is expressed as the stress corresponding to 0·01 per cent deviation in total strain from the tangent at the origin.

Work by Gilbert[2] basically confirms these figures and indicates that at a stress of 40 per cent of the tensile strength a permanent set (BS 18:1962) of 0·01 per cent will not be exceeded for irons from 14 to 23 ton/in^2 (216 to 355 N/mm^2) tensile.

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PROOF STRESS

This is defined as the stress which produces, while the load is still applied, a non-proportional extension equal to a specified percentage of the gauge length (BS 18:1962).

Table 18

Iron	Tensile strength: Mean of 4 tests, 1·0 (25 mm) bar ton/in²	(N/mm²)	Tensile strength: Calculated tensile, 1·2 in (30 mm) bar ton/in²	(N/mm²)	Analysis: TC %	CC %	Si %	Mn %	S %	P %	CE value	Modulus of elasticity lb/in² × 10⁶	(GN/m²)	Limit of proportionality: Stress for 0·01% strain deviation ton/in²	(N/mm²)	Percentage of tensile
A	14·1	(218)	12–14	(185–216)	3·40	2·85	2·29	0·73	0·09	0·04	4·3	16·8	(116)	4·3	(66)	30·7
B	16	(247)	16–17	(247–263)	3·22	2·57	1·99	0·68	0·12	0·18	3·94	18·6	(128)	4·4	(68)	27·3
C	16·8	(259)	14–16·5	(216–255)	3·38	2·74	2·00	0·65	0·11	0·28	4·14	18·1	(125)	4·9	(76)	29·1
D	19·2	(297)	22–24	(340–371)	2·99	2·21	1·36	0·70	0·11	0·08	3·47	21·0	(145)	7·2	(111)	36·6
E	22·2	(343)	22–23	(340–355)	2·95	2·10	1·92	1·02	0·11	0·16	3·65	22·0	(152)	7·0	(108)	31·3
F	24·0	(371)	22–25	(340–386)	2·98	2·18	1·39	0·92	0·09	0·06	3·46	21·8	(150)	7·9	(122)	31·7

Table 19 AVERAGE PERMANENT SET AT VARIOUS APPLIED STRESSES

Iron	Average tensile strength ton/in²	(N/mm²)	Average permanent set, % gauge length ton/in² (N/mm²): 2 (31)	4 (62)	6 (93)	8 (124)	10 (154)
A	14·1	(218)	0·001	0·006	0·014	0·029	0·061
B	16	(247)	0·002	0·005	0·012	0·022	0·044
C	16·8	(259)	0·001	0·003	0·008	0·017	0·034
D	19·2	(297)	0·002	0·005	0·010	0·017	0·024
E	22·2	(343)	0·001	0·0025	0·004	0·006	0·009
F	24·0	(371)	0·0005	0·001	0·003	0·005	0·008

Gilbert[1] has shown that for flake graphite cast irons tested at room temperature the proof stress bears a substantially constant relation to the tensile strength over the British Standard range of tensile strengths of 10–26 ton/in^2 (154–402 N/mm^2):

(1) The 0·01% proof stress = tensile strength × 0·28
(2) The 0·1% proof stress = tensile strength × 0·65

Thus, at a stress of one-quarter of the tensile strength, the 0·01 per cent proof stress level will not be exceeded and at stress of 0·6 of the tensile strength the 0·1 per cent proof stress will not be exceeded.

At a stress of 0·4 times the tensile strength, a permanent set of 0·01 per cent will not be exceeded for irons of 14–23 ton/in^2 (216–355 N/mm^2) tensile.

Lloyd and Harding[2] have carried out 0·1 per cent proof stress determinations at high temperatures on grey irons as shown in *Table 20*. It should be observed that the values above 300°C are above the 100 000 h rupture stress given in the section on heat resistance.

Table 20 PROOF STRESS OF CAST IRON (Rate of loading: 2 ton/in^2/min; 31 N/mm^2/min)

	Temperature					
	Room	*100°C*	*200°C*	*300°C*	*400°C*	*500°C*
Unalloyed:						
Tensile, ton/in^2	17	16·7	15·8	17·4	16·0	13·0
Tensile, N/mm^2	263	258	244	269	247	201
0·1% proof stress, ton/in^2	12·6	11·6	10·7	10·8	10·0	8·3
0·1% proof stress, N/mm^2	195	179	165	167	154	128
Brinell hardness	229	228	216	223	221	218
Alloyed acicular:						
Tensile, ton/in^2	21·6	19·0	20·0	21·4	20·9	16·0
Tensile, N/mm^2	334	293	309	331	323	247
0·1% proof stress, ton/in^2	17·6	15·0	18·0	16·6	19·0	10·4
0·1% proof stress, N/mm^2	272	232	278	256	293	161
Brinell hardness	339	333	323	329	345	311

	Composition							
	TC %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Cr* %	*Mo* %
Unalloyed	3·43	2·78	0·91	0·07	0·116	—	—	—
Alloyed acicular	3·40	2·16	1·32	0·08	0·1	1·77	0·44	0·8

Malleable irons[3]

The ratio of 0·2 per cent proof stress to tensile strength for the various types of malleable iron has been given as:

Whiteheart malleable = 0·55 times tensile strength
Blackheart malleable = 0·65 times tensile strength
Pearlitic blackheart = 0·63 times tensile strength

The 0·1 per cent proof stress can be taken as the 0·2 per cent value minus 0·5 ton/in^2 (8 N/mm^2).

Nodular or SG Irons[4] BS 2789: 1961

0·1% proof stress: Grades SNG24/17 to 37/2 (ferritic) = 0·65 × tensile strength
0·1% proof stress: Grades SNG42/2 and 47/2 (pearlitic) = 0·55 × tensile strength
0·2% proof stress: Grades SNG24/17 to 37/2 = 0·1% proof stress + 0·8–1·1 ton/in^2 (or +12–17 N/mm^2)
0·2% proof stress: Grades SNG42/2 and 47/2 = 0·1% proof stress + 2 ton/in^2 (or +31 N/mm^2)

Compression

Proof stress values in compression are given in the summarised data sections for grey, malleable and nodular or SG irons.

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SHEAR STRENGTH OF CAST IRON

This test is not widely quoted in British practice, although it is fairly widely used in Continental practice. In general, it can be taken that the shear strength of grey cast iron is from 1·1 to 1·6 times the tensile strength[1,2]. The ratio is lower for high strength irons than for low strength irons (*Table 21*).

Table 21

Tensile strength ton/in^2 (N/mm^2)	*Range of shear: tensile ratios*
15–20 (232–309)	1·30–1·10
10–15 (154–232)	1·45–1·30
7–10 (108–154)	1·70–1·45

Mazet[3] quotes the following figures:

Tensile strength, ton/in^2	22·8	22·2	19	15·9	12·7
Tensile strength, (N/mm^2)	(352)	(343)	(293)	(246)	(196)
Shear/tensile ratio	1·08	1·11	1·26	1·52	1·95

The above ranges are rather wide, since the published evidence is somewhat conflicting.

The modulus of elasticity in shear for grey cast iron is given[4] as $5{\cdot}0–7{\cdot}5 \times 10^6$ lb/in^2 (34–52 GN/m^2). The following figures have been given[5] for blackheart malleable iron in shear:

Shear strength	21·8 ton/in^2 (337 N/mm^2)
Yield point in shear	10·4 ton/in^2 (161 N/mm^2)

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TORSIONAL STRENGTH

As the shear strength of cast iron is greater than the tensile, a bar subjected to pure torsion fails in tension and tends to produce a helical shape of fracture. The ratio of torsional stress to tensile is approximately 1·11–1·45 for solid bars, and tends to be lower in high tensile than low tensile irons. The formula:

$$S = \frac{Tr}{J}$$

(where S = shearing stress; T = twisting moment; r = radius; and J = polar moment of inertia) for determining the ultimate torsion stress assumes cast iron to be elastic. This, of course, is not the case; hence, the intensity of stress from centre to surface does not increase uniformly. With hollow bars, where the wall thickness is small compared with the external diameter, the torsion:tensile ratio is practically unity[1].

Experimental work[2] has shown that the torsional strength of grey iron decreases as the diameter of solid cylindrical specimens increases.

The modulus of rigidity is from 0·35 to 0·4 of the modulus of elasticity in tension.

Blackheart malleable cast irons[3] with tensile strengths of 22·4–25·4 ton/in^2 (346–392 N/mm^2) have given figures for torsional rupture stress of 26 ton/in^2 (402 N/mm^2), giving a ratio of torsional stress to tensile stress of 1·02–1·16.

For blackheart malleable BS 310: 1958, Amendment No. 1, 1962, gives typical values for modulus of rigidity of $9–10 \times 10^6$ lb/in^2 (62–69 GN/m^2), and states that the 0·5 per cent proof stress or 0·5 per cent permanent set stress in torsion will be in the range 9–11 ton/in^2 (139–170 N/mm^2).

For nodular cast iron, the following figures (*Table 22*) have been given for modulus of rigidity:

$$\text{Modulus of rigidity } N = \frac{E}{2(1+v)}$$

where E = modulus of elasticity and v = Poisson's ratio

Table 22

	Tensile strength ton/in² (N/mm²)		*Elongation* %	*Modulus in tension* lb/in² (GN/m²)	*Torsion modulus* lb/in² (GN/m²)	*Modulus torsion / Modulus tension*
As-cast pearlitic	42·8	(661)	4	$25{\cdot}1 \times 10^6$ (173)	$9{\cdot}82 \times 10^6$ (68)	0·39
As-cast pearlitic	47·0	(726)	3	$25{\cdot}2 \times 10^6$ (174)	$9{\cdot}8 \times 10^6$ (68)	0·39
Normalised	67·3	(1039)	4	$24{\cdot}6 \times 10^6$ (170)	$9{\cdot}97 \times 10^6$ (69)	0·405
Ferritic	26·8	(414)	28	$24{\cdot}2 \times 10^6$ (167)	$9{\cdot}62 \times 10^6$ (66)	0·40
Ferritic	27·5	(425)	29	$25{\cdot}5 \times 10^6$ (176)	$9{\cdot}14 \times 10^6$ (63)	0·358

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IMPACT PROPERTIES

Some resistance to fracture by impact is exhibited by all materials and the range of such resistance extends, for example, from glass on the one hand to pure copper on the other*.

It is well known that there are very different degrees of shatter resistance in glass and yet the difference between the toughest glass and brittle glasses would not be apparent in the tests used to measure the impact resistance of steel.

Impact tests are therefore designed to give useful comparative figures for a particular range of materials. The Izod test is typical.

The hammer pendulum is arranged to strike a cantilever specimen at the point of maximum velocity [11·5 ft/s (3·5 m/s)] and the energy absorbed in fracture is measured from the loss of height of swing after impact. The Charpy machine uses the same principle, but the specimen is supported at both ends and struck in the middle.

The work required to fracture can be determined by measuring the area under the stress/strain curve at maximum load, either in tensile or slow bend tests, and although the values differ, these methods of test place different cast irons in the same order of merit as the Izod or the Charpy test[1, 2].

*Professor N. Goodman refers to the method of determining resistance to shock in his book *Mechanics Applied to Engineering*, Longmans Green, London, 382–383 (1926).

The following results[2] (*Table 23*) compare the energy to rupture in these three tests on four cast irons with two levels of phosphorus content. The tensile tests were taken on bars of 19·6 mm diameter and the bend and Charpy impact tests were carried out on bars of 20 mm diameter and 100 mm span (volume 31·4 cm^3).

Table 23

Iron		*Resilience: Area under stress/strain diagram divided by volume of specimen under test*				*Resilience Modulus of shock, ISOR946*	
		Tensile		*Slow bend*		*Impact*	
Tensile ton/in^2	*Phosphorus* %	ft.lb/in^3	(kg.cm/cm^3)	ft.lb/in^3	(kg.cm/cm^3)	ft.lb/in^3	(kg.cm/cm^3)
18·9	0·65	11·9	(10·1)	2·2	(1·89)	3·9	(3·3)
17·2	0·25	19·0	(16·2)	4·3	(3·64)	7·3	(6·2)
14·9	0·65	5·3	(4·5)	1·1	(0·93)	2·3	(1·97)
13·8	0·25	5·5	(4·6)	1·4	(1·19)	3·3	(2·82)

The energy to rupture in the tensile tests is greater than that in the bend or the impact tests as the whole volume of material is subject to the maximum applied stress and consequently the work done per unit volume is higher.

In the bend tests the energy to rupture is less than in the impact test, but the order of merit is the same in all three tests. In all tests the energy to rupture is less with the higher phosphorus material.

When a specimen fractures the total energy may be divided into two parts: (a) energy to initiate the crack; (b) energy to propagate the crack.

In brittle materials such as grey cast iron the energy to initiate is greater than the energy to propagate the crack.

In ductile materials the energy both to initiate and propagate a crack will be high.

Tough materials require considerable energy both to initiate and propagate a crack and the effect of a notch is to concentrate the stress and virtually to provide a ready-made crack so that the energy to propagate the crack assumes a greater proportion of the total energy to rupture.

In a brittle material, a notch reduces the energy to initiate a crack and as the energy to propagate the crack is low, the total energy absorbed will also be low.

Brittle materials can, therefore, best be distinguished from each other by the use of unnotched bars in which energy to initiate as well as to propagate a crack is included in the total energy measured.

The standard impact specimen for steels and ductile materials such as nodular cast iron is a notched bar 10 mm × 10 mm square, or a round notched test piece. The standard V notch is 2 mm deep, 45° included angle, 0·25 mm root radius (BS 131: Part 1, 1961). The impact resistance is measured in foot pounds, kilogramme metres or joules.

For grey cast iron the Izod test piece (BS 1349: 1947) uses a 0·798 in (20 mm) diameter unnotched test piece, 3 in (76 mm) long, with the standard 120 ft.lb

(163 J) Izod test machine (BS 131: Part 1, 1961) with a striking velocity of 11·5 ft/s (3·5 m/s). The impact resistance is measured in foot pounds.

The ASTM specimen (A327–68) uses a Charpy type beam impact unnotched specimen $1\frac{1}{8}$ in (29 mm) diameter tested on a 6 in (152 mm) span and the impact resistance is measured in foot pounds and for similar irons gives values about 2·5 times as high as BS 1349[3].

The International Standards Organisation draft recommendation (ISO R946) uses a Charpy type test with a cylindrical unnotched specimen 20 mm diameter × 120 mm long, tested on a 100 mm span. The striking energy E_d is 5 kgf.m (49 J) and striking velocity is 3·6–4·2 m/s. The impact resistance is measured in kgf.m (J).

The ISO recommendation also introduces the concept of impact modulus or modulus of shock (MC) for specimens not of standard size which is expressed as $\frac{E_r}{V}$ where E_r is the apparent energy to rupture expressed in this case as kgf.cm (J), and V is the volume of the test piece between the supports expressed in cubic centimetres (cm^3). It was found that within practical limits for the same material the impact modulus was independent of the diameter d of the machined specimen if the following conditions were observed:

(1) The specimens were proportional, length $= 6d$.
(2) Span $L = 5d$.
(3) The value of E_r/E_d is above 0·4.
(4) Specimen diameter is between 12 and 29 mm.
(5) Velocity of impact V and other conditions of test remain constant.

The modulus of shock (MC) is expressed as kgf.cm/cm^3 (J/m^3).

Energy losses in the impact test

When the hammer strikes the specimen, the initial kinetic energy in the hammer is dissipated into (a) energy to fracture the specimen, and (b) energy imparted to the flying fragments. Ideally, the method to determine the true energy to rupture would be to use hammers with varying energy content, but constant striking velocity, as suggested by Collaud[4], and to use the energy value that only just breaks the test piece, so that the energy of the flying fragments is as low as possible. Gilbert[1] has carried out tests using hammers of varying weight and constant striking velocity and showed that as the initial energy available was reduced, so the recorded impact value was reduced as the energy available for propelling the specimen was reduced. He also carried out tests with constant hammer capacity and varying striking velocity and showed that with constant initial energy and varying striking speeds, the recorded impact values were reduced as the striking speed decreased.

Küpfer[5] has examined the energy losses in the impact bending test and has shown that energy to fracture A_b = initial energy in hammer A_0 minus KA_r, where A_r is the residual energy in the hammer after rupture and K is a constant for a given test rig that is independent of the cast iron investigated and

can be determined experimentally. For 20 mm diameter test bars with test span of 100 mm and overall length 120 mm, the factor K is 1·022 for a hammer of 33·7 kgf.m (330 J) capacity and 1·14 for a hammer of 3·0 kgf.m (29 J) capacity.

While these considerations affect the precise interpretation of tests carried out under standardised conditions, it must be emphasised that impact test values merely serve to place differing materials in an order of merit with respect to shock resistance. The values cannot be used in precise design calculations and the values obtained by the grey iron tests cannot be compared with those for steel made with different specimen sizes, they can only be used as an order of merit among themselves.

Grey iron

In pearlitic irons, the single blow impact resistance of grey iron rises with tensile strength and will range from 6 to 12 ft.lb for 0·798 in diameter bars (IZOD specification BS 1349) or from 1·00 to 1·60 kgf.m for 20 mm bars to ISO R946 for irons ranging from 10 to 14 ton/in^2 (154 to 216 N/mm^2) tensile; 10 to 23 ft.lb IZOD or 1·5–2·8 kgf.m for ISO test for irons ranging from 14 to 20 ton/in^2 (216–309 N/mm^2) tensile; and 18 to 35 ft.lb IZOD or 2·2–3·0 kgf.m for ISO test for irons above 20 ton/in^2 (309 N/mm^2) tensile. Annealing a pearlitic iron to the ferritic[6] state increases the impact value markedly and lowers the tensile strength. Thus a pearlitic grey iron [20·9 ton/in^2 (323 N/mm^2) tensile] gave an impact value of 23 ft.lb (0·798 in bar, BS 1349) and when annealed [11·3 ton/in^2 (175 N/mm^2) tensile] gave an impact value of 85 ft.lb.

These figures, however, may be modified appreciably by the presence of phosphorus which progressively reduces the impact value until, when phosphorus content exceeds 0·7 per cent, even high tensile irons will show impact values[3] below 10 ft.lb.

The effect of section size from which the standard specimens are cut is less on impact resistance than on tensile strength[3].

Graphite in the undercooled form gives a lower strain to failure in the tensile test than random flake graphite, and also lowers the impact resistance.

Although the resistance to fracture under a single blow is progressively reduced as the phosphorus content of grey iron is increased above 0·2 per cent, with repeated blows of small impact energy (6–7 kgf.cm/cm^2 (0·6–0·7 J/cm^2) on a 20 mm diameter specimen), phosphorus contents of up to 1·4 per cent do not appear to affect the impact fatigue strength of cast iron based on an endurance of $5–6 \times 10^6$ blows[7, 8].

Malleable iron

WHITEHEART MALLEABLE

This is characterised by a very soft, deeply decarburised surface over a harder and stronger pearlitic core with nodules or aggregates of graphite distributed throughout. Because of the change in structure from edge to centre, this material does not lend itself to a standard impact test. Specimens cut from a casting give widely differing results, depending upon whether the decarburised face receives the blow or not. Similar unnotched 10 mm square bars cut from a malleable casting and tested in a 120 ft.lb Izod machine gave 34–40 ft.lb when struck so that the decarburised face was in tension, and 8–10 ft.lb when struck with the core face in tension. Standard 10 mm square notched Izod test pieces may give 2–3 ft.lb (3–4 J) whether the notch is in the core face or the cast face. Gilbert[1] has discussed the impact fracture of non-standard whiteheart malleable iron specimens, notched and unnotched, in which the decarburised face was parallel to the direction of the blow and has shown that at room temperature the optimum values are obtained when the decarburised layer is in the ductile range. The effect of temperature upon the ductile-to-brittle transition is discussed later in this and in a separate section (Low Temperature Properties).

The degree of malleability or shock resistance needs to be determined for each individual component and agreement must be reached between suppliers and purchasers.

BLACKHEART MALLEABLE

The heat treatment applied in the process does not produce a very deep decarburised skin and the structure is predominantly ferritic with aggregates or nodules of graphite distributed throughout. Tests using the standard 10 mm × 10 mm V-notched specimens (BS 131) have shown impact values from 7 to 14 ft.lb (9 to 19 J), depending upon graphite content. At some temperatures below room temperature a change to brittle behaviour will occur when in the test the impact value will fall to below 5 ft.lb (7 J) (BS 310:1958). No impact test, however, is included in the British Standard itself, and test requirements should be agreed between purchaser and supplier.

Nodular or SG Iron

In this material the graphite is nodular in form in the as-cast state and the tensile strength is considerably higher than that of grey iron and the impact resistance is much greater. Using the grey iron 0·798 in (20 mm) diameter unnotched Izod specimen, the specimen even in pearlitic grades may remain unbroken after the application of the maximum blow of 120 ft.lb (163 J).

When heat treated to the ferritic state, elongations from 10 to 30 per cent

may be obtained in the tensile test and the Charpy type impact tests on 10 mm × 10 mm square bars V-notched (BS 131, Part 2: 1972) will give values above 10 ft.lb (14 J) for irons to BS 2789:1961, Grades SNG 27/12 and 24/17.

Effect of temperature on the impact properties of cast iron

GREY IRON

With grey iron there is a steady fall in impact resistance of about 30 per cent at −100°C. The percentage drop at normal winter temperatures down to 20° of frost is small. There is no adverse effect on restoring to room temperature.

Rise in temperature from 0°C to 350°C reduces the impact resistance of grey irons by about 10 per cent, and the high silicon heat-resisting irons (5–7 per cent silicon) of the Silal type, which are very brittle at room temperature, show a marked rise in impact resistance and at 250°C may show greater impact resistance than normal grey irons[9].

The pearlitic grey irons appear to show a transition temperature between two levels of brittle failure which rises as the phosphorus content rises from about 0–50°C for phosphorus below 0·2 per cent, 100–150°C for 0·5 per cent phosphorus, and over 350°C for 0·77 per cent phosphorus[10]. Within normal ranges of atmospheric temperature, however, the effect can be disregarded.

MALLEABLE IRON AND NODULAR IRON

The impact resistance of steel, malleable iron and nodular or SG iron decreases rapidly over a well-defined temperature range (transition range) and marked embrittlement occurs at low temperatures. The transition from ductile to brittle failure is similar in character for all three materials when tested under notched impact at various temperatures, but the transition temperature may vary widely and may be as low as −100°C or at temperatures above 100°C, as shown in *Figure 55*.

The change from ductile to brittle failure in malleable and nodular cast iron is accompanied by a change in fracture appearance from dark (ductile) to bright crystalline (brittle) and the degree of embrittlement over the transition range can be judged by comparison of fracture appearance. Both appearances can be present in the same fracture if the speed of crack propagation has varied[10, 11].

The temperature of the transition depends upon chemical composition, microstructure and heat treatment and occurs at higher temperatures in a notched test than in an unnotched test. High speeds of loading favour brittle failure. For this reason, materials which are brittle in the notched impact test may have good elongations in the tensile test. Good elongation in tensile at room temperature is, therefore, no criterion of susceptibility to brittle failure under notched impact conditions.

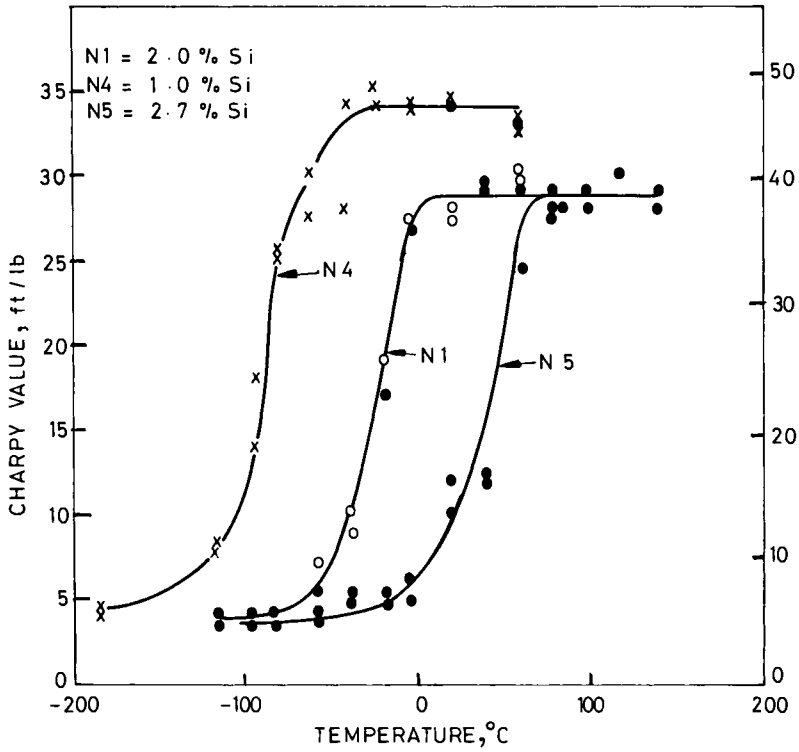


Figure 55. Effect of silicon on impact transition temperature of ferritic nodular iron

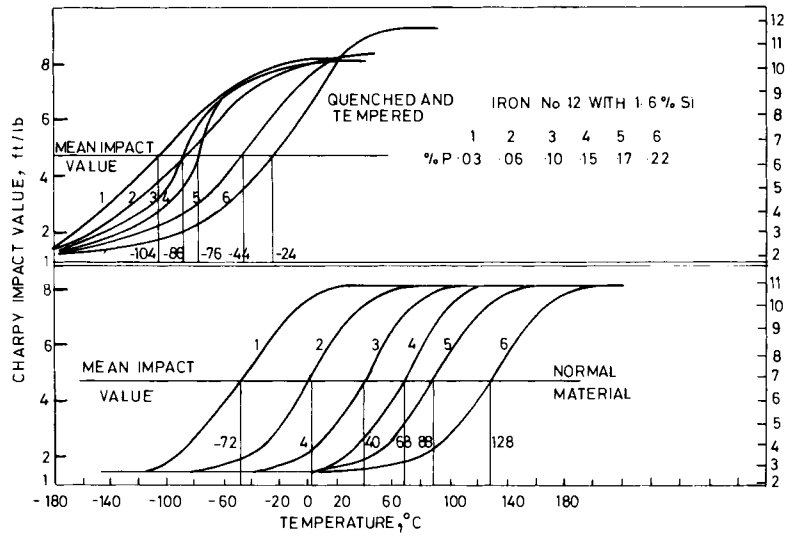


Figure 56. Effect of phosphorus on impact transition temperature of ferritic blackheart malleable iron

Pearlitic malleable and nodular irons have comparatively high transition temperatures in the notched impact test and do not develop the high impact resistance possible with ferritic materials. Ferritic nodular and malleable irons can be obtained with transition temperatures well below room temperature and are then suitable for low temperature service or for shock loading in the presence of notches[10,11]. For both materials the transition temperature rises with silicon content and for a given silicon content also rises as the phosphorus is increased. Higher proof stresses are also obtained by raising the silicon or phosphorus content, which will raise the transition temperature. Higher proof stress with a marked reduction in transition temperature, however, can be obtained by water quenching from 650°C and tempering and slow cooling from 350°C, and malleable irons given this treatment are not subject to the galvanising embrittlement which occurs as a direct result of water quenching from the galvanising bath at about 450°C[11,12] (*Figure 56*).

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FATIGUE RESISTANCE

Fatigue failure

Fatigue failure can be roughly defined as the failure which occurs by the repeated application of a load less than that required to cause failure by a single loading. Generally, failure of a ductile material in a tensile test occurs after considerable deformation, but a ductile material may fail in a brittle manner when subjected to alternating stresses. If a tensile stress well below that required to cause failure by a single application of load and an equal compressive stress are applied alternately to a specimen, then failure does not occur until the alternating load has been applied a number of times. As the stress is lowered, the number of applications of load to cause failure is increased, until a point is reached at which failure will not take place whatever the number of cycles. This stress is called the limiting fatigue stress, fatigue limit, or endurance limit. Fatigue results are usually plotted on a

fatigue curve in which the number of cycles of stress, N, required to cause failure, is plotted against the stress, S. It is usual to plot N on a logarithmic scale, and with this method of plotting the fatigue curve (or S–log N curve) usually resembles that shown in *Figure 57*, being represented by two straight lines[1]. The sloping line shows the cycles of stress to cause failure at stresses above the endurance limit, and the horizontal line gives the stress below which failure will not occur. With cast irons and steels, the 'knee' in the fatigue curve usually occurs between 1 and 10 million cycles of stress and most investigators have therefore assumed that if a specimen withstands 10 million cycles of stress it will withstand the stress indefinitely. However,

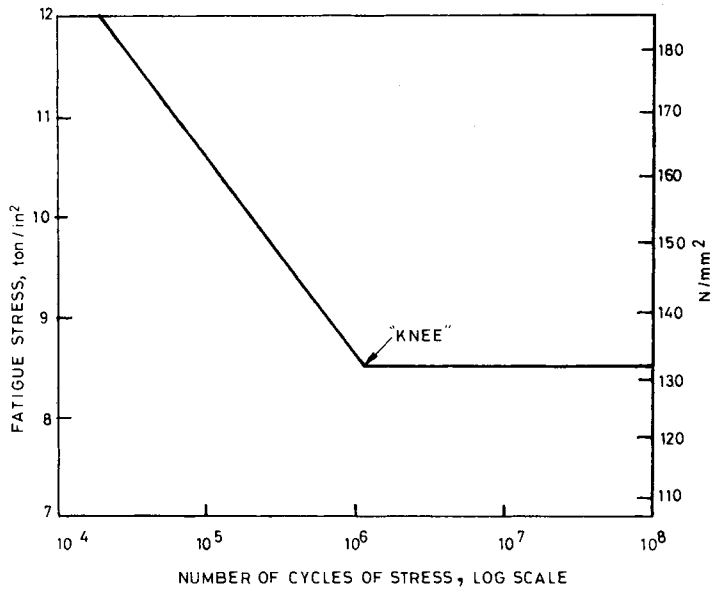


Figure 57. Typical fatigue curve for cast iron

since failures in cast irons have sometimes occurred after more than 10 million cycles, fatigue limits of specimens tested by BCIRA have been carried out on a basis of 20 million cycles.

The endurance ratio, $\frac{\text{stress at fatigue limit}}{\text{tensile strength}}$, of flake graphite cast irons of similar type and structure generally decreases as the tensile strength is increased by alloy additions or by heat treatment.

The endurance ratio for cast iron in reversed bending fatigue varies between 0·33 and 0·47. Lower endurance ratios appear to occur when tensile strengths are increased by alloy additions, without changing microstructure, and also when test specimens are cut from heavy sections[2] (*see also Table 25*). Heat treatment which raises the tensile strength does not raise the endurance limit in proportion to the increase in tensile strength and may also in some cases lower the endurance limit[3]. Acicular cast irons also have a lower

endurance ratio than pearlitic cast irons. High phosphorus irons (0·7–1·9 per cent phosphorus) appear to have slightly higher endurance ratios (0·51–0·58) than low phosphorus irons of the same tensile strength[4].

Notch sensitivity

Effect of surface finish and fillet radius (steel):

Moore and Kommers[5] have shown that fatigue specimens of steel roughly turned show fatigue strength of about 85 per cent of similar specimens that have been polished, the effect being more marked with hard than soft steels.

Fillet radius[6, 7]

Where a simple change in section is involved, without any fillet radius, the stress concentration effect is theoretically infinite. It is important to differentiate between the theoretical stress concentration factor based on the geometry of the component and the fatigue strength reduction factor based on actual testing, which may show variation owing to size effect where no theoretical difference exists.

In practice, the stress concentration factor would rarely exceed 10, as normal tools would produce a fillet of, say, 0·003–0·004 in (76–102 μm) in a nominally sharp corner. With a fillet radius of 0·006 25 in (159 μm), however, the theoretical stress concentration factor becomes 2·3, and with a fillet radius of 0·25 in (6350 μm) it becomes 1·5[7]. The strength reduction factor previously mentioned is usually less than the theoretical stress concentration factor for small components. As size increases, the stress reduction factor approaches the theoretical value. Thus, if design is based on the theoretical value, a margin of safety is inherently achieved.

Cast iron may be considered as an intrinsically strong material, weakened by the presence of graphite flakes which serve as notches or stress-raisers. A deliberately produced notch would not, therefore, be expected to affect seriously an already notched material. In general, the cast irons have low notch sensitivities. In some cases, notching has no effect on the endurance limit at all, while in others, usually with higher tensile strength, e.g. 19 ton/in^2 (293 N/mm^2) tensile, the effect is small (18–20 per cent reduction) compared with steels. It appears also that the flake graphite cast irons are relatively insensitive to notches, even in static tensile tests, and to surface finish in fatigue tests. However, cast irons are relatively insensitive only to small notches and as the notch radius increases above about $\frac{1}{4}$ in (6350 μm) there is little to choose between notch sensitivity of cast iron and wrought steel.

Low notch sensitivity, however, does not imply that the presence of a notch will not initiate a fracture if the basic section is too small; it can be interpreted to mean that where fatigue failure occurs at a notch, the mere provision of a fillet will be less effective in preventing a fatigue failure than would be the case with steel. The effect of a notch appears more significant in high strength materials.

Overstressing

Overstressing occurs when a specimen is subjected to a stress above the endurance limit. A small degree of overstress for a relatively small number of cycles may produce damage which may increase with the stress and the number of cycles.

Weak cast irons show a remarkable ability to absorb overstress, both in the notched and unnotched condition. This ability, however, is reduced by alloy additions which increase the tensile strength.

Understressing

Moore, Lyon and Inglis[8] and Kommers[9] have found large increases in fatigue limit after understressing on coarse flake graphite irons. These results have been only partially confirmed by work at BCIRA. A grey cast iron with graphite Type A, Size 4–5, tested as-cast [tensile 20·9 ton/in^2 (323 N/mm^2)] and annealed [tensile 11·3 ton/in^2 (175 N/mm^2)] showed no clear indication of increase in fatigue limit by understressing at 0·5 ton/in^2 (8 N/mm^2) below the fatigue limit in unnotched specimens, although a small rise was found with notched specimens and a similar result was found with coarse graphite (ASTM Type A, Size 3–4) pearlite iron [13·5 ton/in^2 (208 N/mm^2) tensile]. Pearlitic and ferritic nodular iron also showed no significant change.

The results of Moore and Kommers were obtained with four-point loading on coarse graphite irons, in which the whole of the specimen was uniformly stressed, whereas BCIRA tests were carried out on cantilever specimens and stressed locally. It may therefore be that only under uniformly stressed conditions is the fatigue limit affected by understressing.

In the present state of our knowledge it would be wise to disregard the likelihood of increase in fatigue limit by understressing either in flake graphite or nodular graphite cast irons.

Influence of surface rolling upon the fatigue strength of cast iron

It is well known that cold working the surface of steel shafts by rolling has a good effect on their fatigue properties. Experimental work carried out by the BCIRA[10] shows that the same effect is found in cast iron.

The fatigue limit (based on 20 million reversals) and the life at stresses above the fatigue limit of a flake graphite cast iron, are increased by surface rolling. The fatigue limit may be increased by 20 per cent in a pearlitic iron, and 100 per cent in a ferritic iron.

The fatigue limit of a pearlitic nodular cast iron may be increased 150 per cent, and may possibly be increased still further if higher pressures are applied. A ferritic nodular iron may show a 160 per cent increase in fatigue limit. The rolling pressure for maximum fatigue strength varies with the type of cast iron and the radius of the fillet or notch.

In Wöhler type notched fatigue test specimens, rolling pressures of

Table 24 FATIGUE STRENGTH OF CAST IRON

Type	*Minimum tensile strength 1·2 in (30 mm) bar* ton/in²	(N/mm)	*Unnotched* *Limiting fatigue stress* ton/in²	(N/mm²)	*Endurance ratio*	*Notched* *Limiting fatigue stress* ton/in²	(N/mm²)	*Endurance ratio*
GREY								
Low strength	8·93	(138)	4·15	(64)	0·47	4·15	(64)	0·47[2]
BS 1452 Grade 10	10	(154)	4–4·5	(62–69)	0·41–0·45	3·6–4·5	(56–69)	0·36–0·45[11]
BS 1452 Grade 12	12	(185)	5·4	(83)	0·45	4·3–6·3	(66–97)	0·34–0·44[11]
BS 1452 Grade 14	14	(216)	4·7–6·3	(73–97)	0·33–0·45	4·7–6·0	(73–93)	0·33–0·43[11]
BS 1452 Grade 17	17	(263)	7·7–8·5	(119–131)	0·45–0·46	7·0–8·5	(108–131)	0·41–0·45[11]
BS 1452 Grade 20	20	(309)	8–9·0	(124–139)	0·41–0·45	6·8–8·0	(105–124)	0·34–0·40[11]
BS 1452 Grade 23	23	(355)	9·5–9·8	(147–151)	0·44–0·45	8·0–8·4	(124–130)	0·36[11]
BS 1452 Grade 26	26	(402)	9·9–10·5	(153–162)	0·38–0·41	8·3–9·9	(128–153)	0·32–0·38[11]
Acicular	30·9	(477)	11·5	(178)	0·38	10·5	(162)	0·35[12]
	33·9	(524)	11·5	(178)	0·34	8·5	(131)	0·32[13]
Austenitic	14·1	(218)	5·36	(83)	0·38	3·13	(48)	0·22[14]
	11–15·6	(170–241)	4·0–8·0	(62–124)	0·36–0·5	—		—[25]
NODULAR	Standard							
BS 2789	test block							
Grade SNG 24/17	24	(371)	12·0	(185)	0·5	7·6	(117)	0·21[11]
Grade SNG 27/12	27	(417)	13·2	(204)	0·49	8·2	(127)	0·3[11]
Grade SNG 32/7	32	(494)	14·4	(222)	0·45	8·6	(133)	0·27[11]
Grade SNG 42/2	·42	(649)	17	(263)	0·41	10·2	(158)	0·24[11]
Grade SNG 47/2	47	(726)	18·8	(290)	0·40	11·3	(175)	0·24[11]

the order of 80 lb (356 N), applied for five complete rotations of a V-notch with a root radius of 0·25 mm appear to give optimum results. With increasing rolling pressure the fatigue limit at first increases and then decreases.

A most significant effect of surface rolling on the fatigue properties of cast iron is the very great increase in life obtained at stresses above the fatigue limit.

This may be of great importance when for some other reason than fatigue failure the life of a component may be short. In rolling mill practice, where machined or ground surfaces may wear out within one to five million stressed reversals, the roll neck stresses may be increased markedly above the limiting fatigue stress.

Surface rolling of fillets has been used to practical advantage to increase the fatigue limit of both cast iron and steel crankshafts.

Surface rolling, however, does not improve the fatigue properties in all circumstances. Under push-pull conditions surface rolling may reduce the fatigue value.

Typical notched and unnotched fatigue strengths of various cast irons are shown in *Table 24*.

Effect of as-cast section thickness

The above figures for grey iron are based on test pieces cut from 1·2 in (30 mm) as-cast test bars using cantilever specimens of 0·417 in (10·6 mm) diameter. The fatigue strength and endurance ratio is lower when fatigue test pieces are cut from heavier section castings of comparable tensile strength with running sections in excess of 1–1¼ in (25–32 mm) when endurance ratios of 0·34–0·38 may be expected[27].

The following results (*Table 25*) illustrate this important effect with standard size fatigue specimens [0·417 in (10·6 mm) diameter] cut from actual commercial castings such as diesel engine liners and pistons from BCIRA[11] and other tests[15] including low alloy irons.

These figures indicate that for heavy section castings [1¼ in (32 mm) or more] the ratio of fatigue strength to tensile is appreciably lower than for fatigue specimens cut from 1·2 in (30 mm) standard test bars, and endurance ratios of 0·34 to 0·38 may be expected rather than the higher value of 0·45 normally found with test specimens cut from 1·2 in (30 mm) test bars.

An interesting exception to this is shown in specimens 6 and 7 with phosphorus contents of 0·4–0·5 per cent, whose strain to failure in the tensile test is markedly lower than that determined for lower phosphorus iron, which would be expected to show higher impact resistance. High fatigue resistance is often associated with brittle materials, as shown subsequently for brittle white iron, and for low intensity impact fatigue by Piwowarski (*see* section on Impact Resistance).

Table 25

	Total carbon %	*Silicon* %	*Manganese* %	*Phosphorus* %	*Nickel* %	*Chromium* %	*Vanadium* %	*Molybdenum* %
1	3·2	1·68	0·6	0·14	—	—	0·15	—
2	3·18	1·51	0·62	0·18	—	—	0·14	—
3	3·22	1·41	0·51	0·11	—	—	0·07	—
4	2·98	1·13	0·9	0·34	—	—	0·23	—
5	3·25	0·94	0·98	0·36	—	—	0·37	—
6	3·25	1·11	0·66	0·43	—	—	0·2	—
7	3·11	1·45	0·51	0·50	—	—	0·07	—
8	3·15	1·26	0·68	0·15	1·39	—	—	0·36
9	2·97	1·04	0·54	0·22	—	—	—	—
10^{15}	3·05	1·4	0·91	0·11	1·54	0·3	0·49	—

	As-cast section in	(mm)	*HB*	*Tensile strength* ton/in²	(N/mm²)	*Strain to failure* %	*Fatigue limit* (ton/in²)	(N/mm²)	*Fatigue Tensile*
1	$1\frac{1}{2}$–$2\frac{1}{2}$	(38–64)	193	18·3	(283)	1–1·1	±7	(108)	0·38
2	$1\frac{1}{2}$–$2\frac{1}{2}$	(38–64)	193	22·2	(343)	1·1	±8·5	(131)	0·38
3	1·6 test bar	(41)	218	18·4	(284)	1–1·15	±7·0	(108)	0·38
4	2 –$2\frac{1}{2}$	(51–64)	149	11·4	(176)	—	±4·0	(62)	0·35
5	2 –$2\frac{1}{2}$	(51–64)	172	14·5	(224)	—	±5·5	(85)	0·38
6	$1\frac{1}{2}$–$2\frac{1}{2}$	(38–64)	203	16·7	(258)	0·57–0·60	±8·0	(124)	0·48
7	1·6 test bar	(41)	231	18·5	(286)	—	±8·0	(124)	0·43
8	$1\frac{3}{8}$	(35)	224	23·6	(264)	—	±8·0	(124)	0·34
9	$1\frac{3}{8}$	(35)	208	19·3	(298)	—	±7·5	(116)	0·39
10^{15}	$2\frac{1}{4}$	(57)	—	21	(324)	—	±7·15	(110)	0·34

Effect of size of actual fatigue specimens

The endurance ratio also appears to be lower with fatigue specimens machined to large diameters for both grey and nodular irons[16]. Thus Eagan[15] for as-cast grey iron [21·5 ton/in² (332 N/mm²) tensile] found endurance ratios of 0·35 on 1 in (25 mm) diameter fatigue specimens cut from $1\frac{7}{8}$ in (48 mm) thick castings, but this low endurance ratio could equally well be the result of the heavy section from which the specimen was cut.

Majors[17] showed a fall in fatigue strength for an annealed nodular iron [tensile 35 ton/in² (541 N/mm²) ferritic] as the fatigue specimen size increased.

Specimen diameter in (mm)	*Fatigue limit* ton/in² (N/mm²)	*Endurance ratio*
0·19 (4·8)	±16·1 (249)	0·46
0·32 (8·1)	±14·0 (216)	0·40
0·45 (11·4)	±13·5 (208)	0·385

Eagan[15] for nodular iron [30·7–38·2 ton/in^2 (474–590 N/mm^2) tensile] with specimens cut from as-cast sections of 2–2$\frac{3}{4}$ in (51–70 mm), tested three sizes of fatigue specimen:

Specimen diameter in (mm)	*Endurance ratio*
0·350 (8·9)	0·461
0·750 (19·1)	0·357
1·000 (25·4)	0·343

Pope[18] indicates a similar effect for steels in bending fatigue, but states that the effect is small in pulsating direct tension and that the effect is greater with steep stress gradients.

It appears, therefore, that in calculating permissible stresses for heavy cast iron sections, grey or nodular, a fatigue tensile ratio of not more than 0·34 should be employed in reversed bending, and not more than 0·25 in alternating direct tension and compression.

Effect of heat treatment

Oil quenching of grey and nodular cast irons from 870°C followed by tempering to 565°C may increase the tensile strength by 25–40 per cent, but the endurance limit may not be increased by more than 20–25 per cent, and in some cases may actually be reduced. The notch sensitivity is also increased. Before using heat treatment to increase the strength of a component subject to fatigue, the suitability of the material should be determined[3].

Effect of compressive stress on the fatigue range

Work by various investigators[8, 19, 20, 21, 24] has shown that the fatigue limit of cast iron in compression is markedly higher than in tension. The conventional Wöhler fatigue test stresses the specimen equally in tension and compression during each cycle. The diagram in *Figure 58*, based on work by Pomp[20] and Hempel[20, 21] using a pulsating machine instead of a rotating Wöhler machine, illustrates the change in safe stress for a particular cast iron when alternating stresses are applied over the possible ranges of stress through maximum tensile to the maximum compressive stress.

This curve shows quite clearly that as the alternating stresses become compressive in character, so the safe limiting stress range increased markedly.

In the diagram, which is a modified form of the Goodman diagram, the centre diagonal line roughly indicates the mean stress, and point 'a', where the mean stress is zero, shows the safe range of stress $\pm 3{\cdot}8$ ton/in^2 (59 N/mm^2) when tested in the conventional manner where the stresses alternate equally between tension and compression. At point 'b', where there is no compressive

stress and the mean stress is tensile [approximately $+3{\cdot}2\,\text{ton/in}^2$ ($49\,\text{N/mm}^2$)], the safe range of stress is $+6{\cdot}4\,\text{ton/in}^2$ ($99\,\text{N/mm}^2$) to zero. At point 'c', where there is no tensile stress and the mean stress is compressive [approximately $-9{\cdot}6\,\text{ton/in}^2$ or ($148\,\text{N/mm}^2$)], the safe range of stress is from zero to $-19{\cdot}6\,\text{ton/in}^2$ ($303\,\text{N/mm}^2$), and at point 'd', when the mean stress is $-21{\cdot}0\,\text{ton/in}^2$ ($324\,\text{N/mm}^2$) the safe range of stress is $-3{\cdot}2$ to $-39{\cdot}0\,\text{ton/in}^2$ (49 to $602\,\text{N/mm}^2$). At compressive stresses beyond this point, an increase in the effective section begins to occur owing to plastic deformation and the

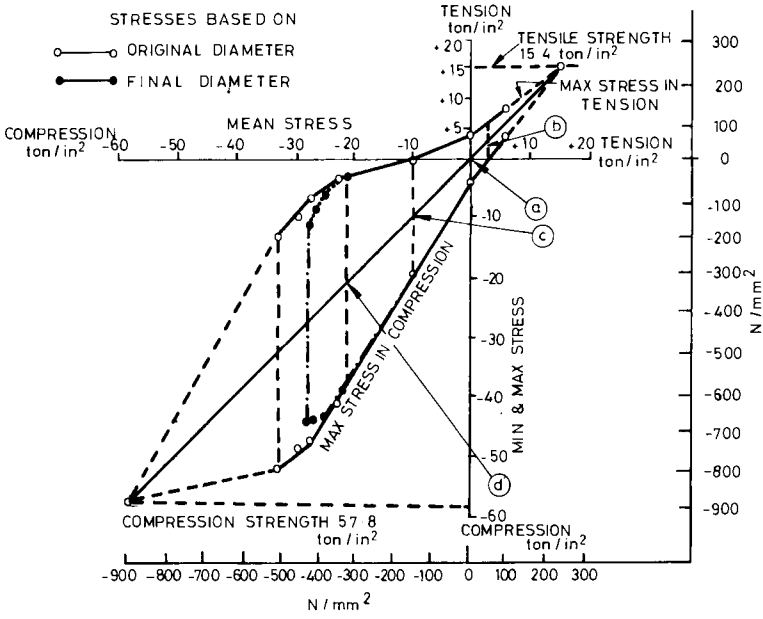


Figure 58

range of stress based on the actual diameter begins to fall off, although based on the original diameter the range of stress would appear to increase.

The curve is based on an iron with a tensile strength of $15{\cdot}4\,\text{ton/in}^2$ ($238\,\text{N/mm}^2$) and a compression strength of $57{\cdot}8\,\text{ton/in}^2$ ($893\,\text{N/mm}^2$). The form of the curve was similar for an iron of $12{\cdot}5\,\text{ton/in}^2$ ($193\,\text{N/mm}^2$) tensile. It should be noted in this example that the fatigue strength with zero mean stress is only $\pm 3{\cdot}8\,\text{ton/in}^2$ ($\pm 59\,\text{N/mm}^2$) (endurance ratio 0·25) which is substantially lower than would be expected from such an iron tested on a rotating Wöhler type test. A Wöhler test would be expected to give an endurance ratio of the order of 0·33 to 0·45.

Table 26

Ref. letter	*Tensile strength* ton/in²	(N/mm²)	*Brinell hardness*	*Fatigue limits* ton/in² (N/mm²)									
				20°C		*250°C*		*400°C*		*475°C*		*550°C*		*600°C*	
C	19·0	(293)	222	8·75	(135)	8·03	(124)	9·06	(140)	9·24	(143)	—	—	4·3	(66)
N	17·2	(266)	201	7·46	(115)	8·17	(126)	8·92	(138)	8·57	(132)	—	—	4·7	(73)
T	14·8	(229)	187	7·16	(111)	8·7	(134)	9·37	(145)	8·48	(131)	5·9	(91)	—	—

Table 27

Ref. letter	*TC* %	*GC* %	*CC* %	*CE value*	*Si* %	*P* %	*S* %	*Mn* %	*Ni* %	*Cr* %	*Tension* ton/in² (N/mm²)		*Compression* ton/in² (N/mm²)		*BH*
C	3·42	2·50	0·92	3·94	1·12	0·437	0·055	1·26	Nil	Nil	19·0	(293)	61	(942)	222
N	3·09	2·31	0·78	3·33	0·55	0·18	0·103	1·09	Nil	Nil	17·2	(266)	62	(957)	201
T	3·33	2·68	0·65	4·07	1·02	1·21	0·09	0·77	0·16	0·23	14·8	(229)	57	(880)	187

Effect of temperature

Tests by Fitzgeorge and Pope[22] (*Figure 59*) with three cast irons ranging from 14·8 to 19·0 ton/in^2 (229 to 293 N/mm^2) tensile gave the results presented in *Tables 26* and *27* with rotating beam type test machines.

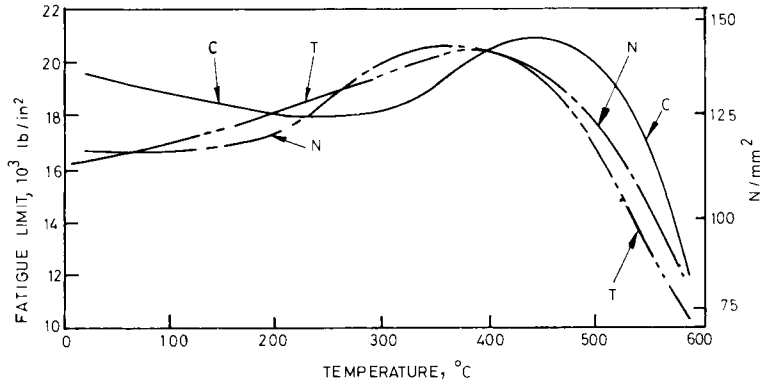


Figure 59. Variation in fatigue strength with temperature for cast irons C, N and T

Almost identical types of curve to those in *Figure 59* are shown for notched and unnotched ASTM Grade 20 iron by Collins and Smith's work[28] where the maximum is about 400–427°C.

Influence of shrink fits

It is well known that low fatigue resistance is characteristic of shrink fitted assemblies. Damage due to fretting is one factor influencing fatigue strength, but increasing the clamping load reduces at least initially the endurance limit of shafts[23].

Corrosion fatigue

Work at BCIRA on fatigue tests in demineralised water spray has shown reductions of fatigue limit from 11 to 33 per cent for grey and nodular irons[11]. Collins and Smith have reported a 50 per cent reduction in water for austenitic grey iron[14].

Malleable cast iron

The ratio of endurance limit to tensile strength appears to range from 0·43 for heat-treated pearlitic malleable, to 0·58 for ferritic malleable[26]. BS 309 (whiteheart) and BS 310 (blackheart) amendments, 1962, quote ratios of 0·4–0·6. The effect of a notch is to lower the endurance limit. The ratio of

unnotched to notched fatigue limit for blackheart malleable is given as 1·32 for ferritic blackheart, 1·70 for pearlitic blackheart, and 2·05 for heat-treated pearlitic blackheart[26].

With alternating direct tension and compression the ratio of endurance limit to tensile strength is of the order of 0·25[8, 20]. In the case of blackheart malleable cast iron the influence of increasing the ratio of compression to tensile stress in fatigue is markedly less than for grey iron[20, 16].

White cast iron

A series of tests carried out by the BCIRA was made on specimens chill-cast to the following compositions:

	Chill cast	*Sand cast*	
	(*1*)	(*2*)	(*3*)
	%	%	%
Total Carbon	3·33	3·2	2·2
Silicon	2·20	0·5	0·4
Manganese	0·60	0·44	0·34
Sulphur	0·097	0·025	0·02
Phosphorus	0·20	0·01	0·01
Nickel	0·02	1·1	1·1
Chromium	0·08	1·04	1·0
Copper	0·12	—	—
Molybdenum	—	0·27	0·25

The specimens were approximately 0·9 in (23 mm), tapering to 0·6 in (15 mm) in diameter as-cast, and specimen No. 1 was stress-relieved for 3 h at 500°C, followed by furnace cool. Specimens Nos. 2 and 3 were heat treated at 900°C for 4 h and air-cooled.

The specimens were surface ground to the final diameter of 0·417 in (10·6 mm) and were fatigue tested in a Wöhler rotating cantilever machine and the following results obtained:

	Chill cast	*Sand cast*	
	(*1*)	(*2*)	(*3*)
Tensile strength	28·8 ton/in^2	27·5	27·8
	(445 N/mm^2)	(425)	(429)
Hardness	570–585 DPN	460	410
Fatigue limit	±20·0 ton/in^2	±13	±15·5
	(309 N/mm^2)	(201)	(239)
Endurance ratio: $\frac{\text{Fatigue limit}}{\text{Tensile strength}}$	0·69	0·47	0·56

The endurance ratio of 0·69 is very high compared with grey cast irons. The irons with lower endurance ratio were substantially softer than the

chill-cast specimen. The life of specimens above the fatigue limit is very short, giving a more shallow curve than usual.

There appears to be no other published information on the fatigue strength of white cast iron.

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DAMPING CAPACITY

Damping capacity is that property which permits a material to absorb vibrational stresses.

Theoretically, a tuning fork made of truly elastic material should continue to vibrate indefinitely *in vacuo*. In practice, however, some energy is absorbed at each cycle of vibration within the body of the material itself and transformed into heat. The amount of energy absorbed per oscillation, expressed

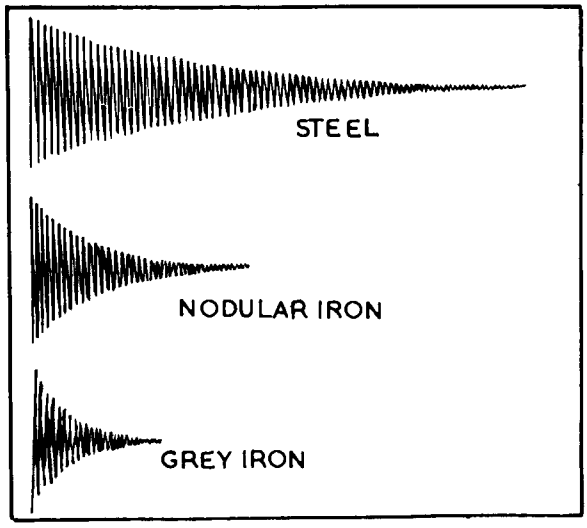


Figure 60. Damping of three materials

as a percentage of the initial energy, is a measure of the damping capacity of a material.

Cast iron has an exceptionally high damping capacity, particularly in its softer or weaker forms, the damping capacity is reduced as the tensile strength increases, and the clearest indication of the difference between soft grey iron, nodular iron and steel is shown in *Figure 60*[1].

Damping capacity has been studied in detail by many workers, notably by Föppl[2] and the curves in *Figure 61* are derived from torsional vibration tests carried out by a method similar to that devised by Föppl and Pertz in which a cylindrical bar held rigidly at each end has a symmetrical inertia

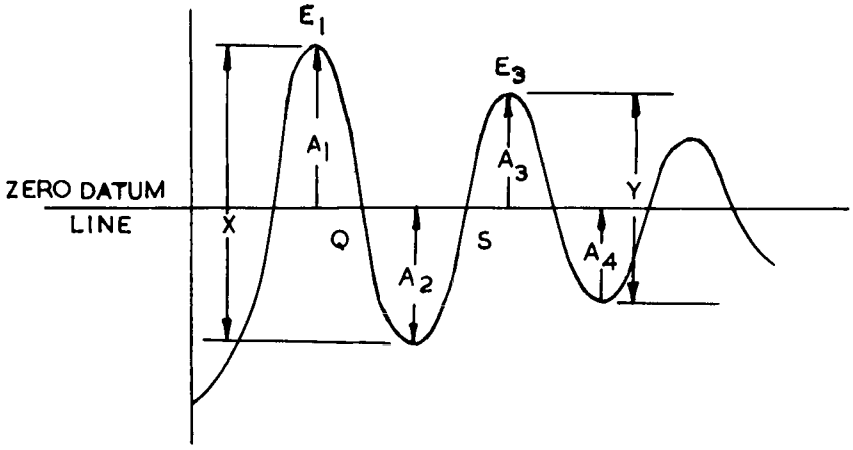


Figure 61. The track of a damped free torsional vibration

bar clamped rigidly to the centre. This inertia bar can be deflected to stress the test piece in torsion to any desired level of surface stress. If the inertia bar is released, it will continue to vibrate with steadily reducing amplitude and an autographic record of the vibrations gives results similar to those in *Figure 61*.

The ability of a material to damp out vibrations may be expressed in various ways:

If E = maximum stress energy stored in the specimen at the extreme position of the cycle,
ΔE = total energy dissipated per cycle, or energy required to keep the specimen in constant oscillation = $E_1 - E_3$,
A = amplitude measured from the mean line,

it can be shown that:

(1) $\dfrac{\Delta E}{E} = \dfrac{A_1^2 - A_3^2}{A_1^2}$ = damping ratio

(2) Expressed as a percentage loss of energy per cycle

$$\frac{\Delta E}{E} \times 100 = \frac{A_1^2 - A_3^2}{A_1^2} \times 100 \text{ per cent} = \text{damping capacity}$$

(3) As energy loss per unit of volume of specimen = in.lb/in^3 (J/m^3)

(4) As a percentage decrease in vibrational amplitude per cycle,

$$= \frac{A_1 - A_3}{A_1} \times 100 \text{ per cent}$$

A fairly full description of the torsional method of test and the calculation involved was given by Snelson[3], and reviews of the literature have also been published by Thomson[4] and Linacre[5].

In the torsional method of test for any given deflection, the stresses and strains vary from edge to centre of the specimen, and it is customary to relate the damping capacity with the shear stress at the surface.

In most metals and alloys, a high damping capacity is associated with low values of proof and ultimate tensile strength, modulus of elasticity and hardness[6].

The existence of damping capacity implies some non-linearity between stress and strain in the form of a hysteresis loop. This is a marked characteristic of the stress/strain properties of cast iron, and has been discussed in the section dealing with these properties. At very low strains the damping capacity of metals is, in general, independent of strain amplitude, but sensitive to frequency, depending upon the type of loading employed. At the higher strains associated with engineering applications, the damping capacity is usually greater and is strain dependent, but substantially independent of frequency[6].

Considerable effort has been made to relate damping capacity to fatigue endurance in other metals, and it appears that when the range of stress is small and within the endurance limit, the damping capacity generally decreases with time to some limiting value, presumably owing to work hardening in the region of active slip. However, it should be noted that in

steel damping may be increased slightly by cold work, particularly during the initial stages of a fatigue test involving a stress range above the endurance limit[6] and high levels of prestress appear to raise the damping of malleable iron. If the stress exceeds the endurance limit, damping invariably increases with the formation of cracks. Since the ratio of plastic to elastic strain increases with stress, the damping capacity would be expected to rise with the applied stress, and this is shown to be the case.

In grey cast irons the damping capacity is higher with low strength high carbon materials, and can reach high values with carbon above 3·5 per cent, as shown in *Tables 28* and *29* and *Figure 62*. Valuable work on the damping

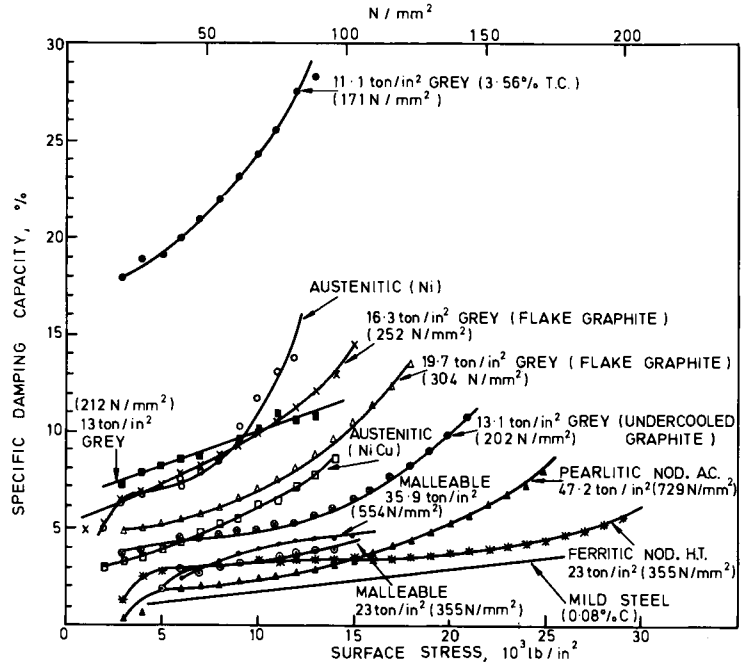


Figure 62. Damping capacity: comparisons of various irons 10 c/s torsional oscillation at room temperature from Table 28

capacity of cast iron has been carried out by Mme Plenard[7, 8] using the results from the decay of free vibrations and from resonance curves at very small, but variable, stresses, and she has shown that a very large range of results can be obtained on flake graphite irons, depending on the conditions under which the measurements were made. but that the values vary within much closer limits for nodular and malleable irons. Her conclusions, in order of decreasing influence, show that:

(1) Increasing carbon equivalent increases damping capacity.
(2) Increasing cooling rate, causing refinement of matrix and graphite structures, decreases damping capacity.

Table 28 (FOR TABLE 29) SPECIFIC DAMPING CAPACITY (%) OF GREY, NODULAR, AUSTENITIC AND BLACKHEART MALLEABLE CAST IRON[3, 6, 8]

	Structure	*Type graphite*	*Tensile* ton/in² (N/mm²)		*Brinell*	*Elongation* %	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Cu* %	*Other* %
GREY IRON														
F. 218	5–10% ferrite	A	11·1	(171)	162	—	3·56	2·42	0·54	0·03	0·13	—	—	—
F. 240	5% ferrite (uninoculated)	A & E	13·7	(212)	192	—	3·32	2·23	0·54	0·027	0·14	—	—	—
F. 247	10–20% ferrite (undercooled graphite)	D & E	13·1	(202)	189	—	3·27	2·19	0·56	0·027	0·14	—	—	Ti 0·35
F. 219	2% ferrite (inoculated)	A	16·3	(252)	196	—	3·27	2·23	0·53	0·029	0·14	—	—	—
F. 248	5% ferrite	A	19·7	(304)	219	—	3·14	2·00	0·56	0·027	0·14	0·7	—	Mo 0·43
	Pearlitic[8] (inoculated)	A	22·5	(347)	220	—	(2·9)	(1·6)	(0·9)	(0·08)	(0·08)	—	—	—
NODULAR														
F. 121 HT	Ferrite		25·9	(400)	130	27·3	3·66	1·80	0·41	0·012	0·025	0·76	—	—
F. 121 AC	As-cast pearlitic		47·2	(729)	244	5·3	3·66	1·80	0·41	0·012	0·035	0·76	—	—
AUSTENITIC IRON														
F. 127	21% nickel		16·6	(256)	125	9·0	2·53	1·88	1·05	0·025	0·025	20·7	—	Cr 1·91
F. 216	Ni Cu		17·6	(272)	152	3·0	2·55	1·94	1·25	0·039	0·028	15·2	7·3	Cr 2·02
BLACKHEART MALLEABLE														
2.1	Ferritic		23·0	(355)	145	16	2·2	1·4	0·35	0·12	0·06	—	—	—
1.11	Pearlitic		35·9	(554)	189	7	2·2	1·4	0·35	0·12	0·06	—	—	—
MILD STEEL	Ferritic[3]		33·2	(513)	(150)	22	0·08	0·06	0·34	0·019	0·002	—	—	Cr 0·01

Table 29 SPECIFIC DAMPING CAPACITY (%) OF CAST GROUND SAMPLES AS IN TABLE 28[3, 6, 8] TESTED AT 10 CPS (10 HZ) IN TORSIONAL OSCILLATOR AT ROOM TEMPERATURE

Surface shear stress in 1000 lb/in² (6·9 N/mm²)	1	2	3	4	5	6	7	8	9	10	11	12	13	14	15	16	17	18	19	20
SAMPLE REFERENCE:																				
F. 218 Grey 11·1 ton/in² (171 N/mm²)	—	—	18·0	18·8	19·3	20·0	21·0	22·0	23·2	24·4	25·6	27·6	28·4	—	—	—	—	—	—	—
F. 240 Grey 13·7 ton/in² (212)	—	—	7·3	7·85	8·3	8·6	8·7	8·55	9·45	10·0	10·4	10·65	10·7	—	—	—	—	—	—	—
F. 247 Grey 13·1 ton/in² (202) (Undercooled graphite)	—	—	3·6	3·8	4·0	4·55	4·55	4·60	4·70	4·85	5·25	5·3	5·65	6·0	6·5	7·0	7·6	8·3	9·0	9·85
F. 219 Grey 16·3 ton/in² (252)	4·85	5·1	6·4	6·8	7·2	7·7	8·2	8·8	9·35	10·0	10·6	11·3	12·1	13·0	14·5	—	—	—	—	—
F. 248 Grey 19·7 ton/in² (304)	—	—	4·9	5·0	5·2	5·4	5·7	6·1	6·5	7·0	7·5	8·1	8·85	9·6	10·5	11·4	12·4	13·5	—	—
Grey 22·5 ton/in² (347)	—	—	3·0	3·3	3·9	4·2	4·6	4·9	5·3	5·8	5·95	6·0	6·3	6·8	7·1	8·0	9·3	11·0	12·5	—
F. 121 HT Ferritic nodular 23 ton/in² (355)	—	—	1·25	2·52	2·7	—	—	3·0	3·2	3·35	—	3·35	3·36	3·38	3·42	3·45	3·51	3·56	3·64	3·73
F. 121 AC Pearlitic nodular 47·2 ton/in² (729)	—	—	0·2	0·6	1·8	1·9	2·0	2·1	2·25	2·4	2·5	2·7	2·9	3·15	3·4	3·7	4·05	4·4	4·8	5·25
F. 127 Austenitic (Nil)	—	5·05	6·4	6·9	—	7·1	7·9	8·9	10·2	11·75	13·1	13·9	—	—	—	—	—	—	—	—
F. 216 Austenitic (Ni Cu)	—	3·0	3·3	3·6	4·0	4·4	4·9	5·3	5·5	6·2	6·5	7·2	7·9	8·6	—	—	—	—	—	—
Blackheart Malleable 2·1 Ferritic 23·0 ton/in² (355)	—	—	—	—	1·9	2·9	2·7	3·0	3·2	3·4	3·6	3·7	3·9	4·0	—	—	—	—	—	—
Blackheart malleable 1·1 Pearlitic 35·9 ton/in² (554)	—	—	—	—	—	2·4	2·9	3·3	3·6	3·9	4·1	4·3	4·4	4·5	4·6	—	—	—	—	—
Mild Steel 0·08% C 33·2 ton/in² tensile (513)	—	—	—	—	1·0	—	—	—	—	1·9	—	—	—	—	2·0	—	—	—	—	2·5

(3) Ferritic (annealed) structures and hardened structures produced by quenching have higher damping capacities than as-cast material. In the hardened structures, the effect of martensite produces internal stress.

It seems probable that damping capacity can be estimated from a knowledge of the stress/strain properties by considering the ratio of the area enclosed by (a) the hysteresis loop (the work lost), and (b) the total area under the stress/strain curve. In general, the greater the curvature of the stress/strain curve, the greater will be the area of the hysteresis loop at a given stress, and it can therefore be concluded that as the curvature of the stress/strain curve increases, the damping capacity will increase.

Gilbert[9] has shown that for given irons, the stress/strain curvature increases as the section size increases, and the cooling rate decreases, confirming the conclusion (2) by Mme Plenard using direct measurement of damping capacity. Annealing has also been shown to decrease the slope of the stress/strain curve. Increasing phosphorus reduces the curvature of the stress/strain curve[10] and would, therefore, be expected to decrease damping capacity. This is confirmed qualitatively by Thum and Petri[11]. Undercooled graphite irons have steeper stress/strain curves than flake graphite irons and are shown in *Table 29* to have lower damping capacity. The same considerations apply to nodular and malleable cast irons. Austenitic flake graphite irons have damping capacities of the same order or slightly lower than inoculated flake graphite irons (*Tables 28* and *29*) indicating that the effect of flake graphite is greater than that of the matrix.

Lazan[12] has shown that the effect of the previous stress history upon damping capacity is of considerable importance, but is very complex.

Various determinations of the damping capacity of cast irons have been made and the values obtained vary appreciably depending upon the sources. *Tables 28* and *29* give results of tests carried out at 10 cps (10 Hz) in torsional oscillation at room temperature.

The high values for specification F218 are noteworthy and it will be seen from specimens F240 and F219 that inoculation has little effect on damping capacity, but that an iron (F247) with undercooled graphite is substantially lower than those with flake graphite of comparable strength and composition and is of the same order as an inoculated iron (Pearlitic, inoculated) with a much higher tensile strength.

The damping capacities of blackheart malleable and nodular iron are very similar and that of low carbon mild steel is of the same order as the values for pearlitic nodular cast iron. The values are summarised in *Figure 62*.

Although it is possible to express the damping capacity in fairly precise terms of energy or amplitude absorption, it is difficult to make use of the data in a quantitative manner. Harding[13] points out that the properties of ductility and shock resistance, which in most static applications are precautions against stresses getting out of control, are of less value under regulated dynamic stresses where failures are non-ductile. He considers that damping capacity plays an important part in the success of high duty cast iron crankshafts.

In riveted built-up steel structures, some degree of damping is auto-

matically included in the riveted joints. In welded structures this is not the case, and such structures tend to transmit vibrations more readily and to resonate more freely than riveted mild steel[1] or cast iron.

Moreover, the importance of pure structure weight in reducing the transmission of vibration is also not inconsiderable and may be of more significance in some cases than specific damping capacity.

Although no cases of actual failure through lack of damping capacity have come within the author's experience, one such case has been reported[13] and Yorgiadis[14] has attempted to show that the excitation stress required to produce fatigue failure in high duty cast iron [tensile 22·5 ton/in^2 (347 N/mm^2)] may be actually higher than for mild steel.

In normal engineering practice, however, the damping capacity of cast iron is most likely to be of value in promoting smooth running, and in the reduction of noise and chatter. For example, textile machinery beams carrying spindles rotating at 10 000–15 000 rev/min are usually made of cast iron, since with steel structures it has been found difficult to eliminate resonance, which interferes with the satisfactory working of the machines.

Cast iron gear boxes and transmission casings, particularly if low in strength and phosphorus content, are notably more silent than aluminium alloy or pressed steel, and the main gears of many marine engines are made with cast iron centres, with forged steel shrunk-on rings. The cast iron centres are favoured on account of their high damping capacity[15]. The high damping capacity of high carbon cast irons (TC $>3{\cdot}5$ per cent) has been shown to reduce the incidence of 'squeal' in disc brakes[16].

Wherever fretting, chatter, vibration or noise are troublesome, consideration should be given to the use of cast iron, even where the conventional mechanical properties of other materials would appear to make them more suitable.

Koenigsberger[17] has stated that damping inherent in a material is small compared with that of metal-to-metal rubbing surfaces, e.g. riveted joints.

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COEFFICIENT OF FRICTION

The coefficient of friction (μ) between two surfaces is the ratio of the force required to slide one surface over the other to the load pressing the two surfaces together:

$$\frac{F, \text{sliding force}}{N, \text{load}} = \mu, \quad \text{or } F = \mu N$$

The coefficient varies with the materials, the surface finish, the degree of lubrication, and so on; it is greater at rest than in motion and is theoretically independent of speed and load.

In practice, the friction between surfaces is a highly complex phenomenon and a considerable body of experimental work has been carried out in all parts of the world[1, 2].

Thus Coulomb's laws for dry friction state[3]:

(1) The friction is directly proportional to the normal pressure between the two surfaces.
(2) The friction is independent of the area of the surfaces in contact for any given normal pressure.
(3) The friction is independent of velocity of rubbing.
(4) The friction between two surfaces at rest is greater than when they are in motion.
(5) The friction depends upon the nature of the surfaces in contact.

The first two laws were based on tests at low pressures and are not necessarily entirely applicable in all circumstances; thus at high pressures, friction between metallic surfaces falls off and then rises to seizure as shown by Goodman[3]. Russian work[4] confirms this for cast iron (180 HB) against steel (170 HB to 650 HV) and showed coefficients of friction of 0·146–0·165 at 450 lb/in^2 (3·1 N/mm^2), falling to 0·113–0·126 at 85 lb/in^2 (0·6 N/mm^2) under varying rubbing speeds (not given).

The third law is only applicable between metallic surfaces at constant temperature; under normal conditions friction decreases with the velocity of rubbing[3]. Thus, in railway practice[5] when cast iron shoes operate on steel-tyred wheels, coefficients of friction for one level of force on a brake block vary markedly during a stop and fall from about 0·08 at 90 mph (145 km/h) to 0·06 after braking to 60 mph (97 km/h), and then rise to 0·18 after braking to 5 mph (8 km/h). Braking from low speeds of the order of 5 mph (8 km/h) may give coefficients of friction of the order of 0·3. Brake block pressures

vary from 50 lb/in^2 (0·3 N/mm^2) for wagons to 270 lb/in^2 (1·9 N/mm^2) for locomotives.

The third law is reasonably accurate for woven asbestos and moulded non-metallic composition brake shoe faces on steel or cast iron surfaces which give a fairly uniform coefficient of friction of the order of 0·3–0·4 throughout the stop with mean lining pressures of 10–100 lb/in^2 (0·07–0·7 N/mm^2) for drum brakes and 50–250 lb/in^2 (0·3–1·7 N/mm^2) for spot disc brakes and rubbing speeds up to 2600 ft/min (13·2 m/s). Brake fade may occur with temperatures of 250–450°C, depending on the lining employed[6].

Friction appears to depend upon the nature of the surface contact and may increase as the surface roughness increases. Thus, for two hardened steel surfaces, lubricated with mineral oil, and with varying degrees of surface finish, the following results were obtained[7]:

	Super finish	*Ground*	*Ground*	*Ground*
Surface roughness RMS	2	7	20	65
Coefficient of friction:				
Mineral oil	0·128	0·189	0·360	0·378
SAE 30	0·119	—	0·252	0·239
Oleic acid	0·099	0·163	0·195	0·238

The coefficient of friction has been stated to be independent of speed, but this is only the case at constant temperatures. If the temperature rises, the coefficient may fall with increasing speed until just before seizure. With lubricated surfaces, speed is very significant[8].

The following figures for coefficients of friction are in general use for smooth, dry engineering surfaces:

Cast iron on cast iron or brass	0·15–0·2
Cast iron on copper	0·18–0·2
Cast iron on steel	0·15–0·3

BS 153:1954 (Part 3a, girder bridges), quotes the following figures for expansion bearings under dead loads:

Steel on hard copper alloy bearings	0·15
Steel on cast iron or steel	0·25

Schmidt and Weiter[9] carried out tests with varying surface finishes and loads; they used slides of:

(1) Hardened steel, ground to 20 μinches (0·508 μm) RMS (approximately 17 CLA).
(2) Ground cast iron, 20 μinches (0·508 μm) RMS (approximately 17 CLA).
(3) Hand-scraped cast iron: 20 contact points/in^2 (31 000/m^2) according to generally accepted methods used in machine tool industry.

Dry surfaces were washed in CCl_4. Lubricated surfaces were washed in CCl_4 and oil applied with felt applicator.

Tests were taken on the above slides with sliders ground to 20 μinches (0·508 μm) RMS (approximately 17 μinches CLA), and coefficients of friction

were observed after varying times under contact before sliding commenced (cast iron slides: total carbon 3·0 per cent, silicon 1·8 per cent, nickel 0·9 per cent and manganese 0·7 per cent). In all cases the effect of time lapse between application of load and initiation of sliding is appreciable. The coefficient of friction under loads from 10 to 50 lb/in^2 (0·07 to 0·3 N/mm^2) rises as the stationary time increases to a fairly fixed level, after which no further increase occurs with stationary time. With cast iron, the rise is from 7·5 per cent to 20 per cent, and in most cases a stable state is reached within one minute of stationary load application, no further change occurring with longer applications of load.

The results after one minute at rest are given in *Table 30*, with figures for bearing bronze included for comparison.

Table 30 STATIC COEFFICIENT OF FRICTION AFTER 1 MIN AT REST

Slide material	*Ground cast iron*			*Scraped cast iron*			*Ground steel*		
Load, lb/in^2	10	30	50	10	30	50	10	30	50
Load, N/mm^2	0·07	0·21	0·34	0·07	0·21	0·34	0·07	0·21	0·34
Cast iron slider									
Dry	0·330	0·325	0·350	0·169	0·200	0·202	*	*	*
Lubricated	0·114	0·123	0·142	0·140	0·144	0·130	0·114	0·126	0·123
Bronze slider									
Dry	0·296	0·276	0·260	0·259	0·259	0·250	0·287	0·306	*
Lubricated	0·143	0·148	0·150	0·167	0·196	0·211	0·100	0·096	0·098

* Denotes increasing coefficient on successive tests.

It is interesting to note from these figures that the coefficient of friction for dry surfaces is lowest with a scraped cast iron surface.

Some tests were carried out by Duff[10] on two as-cast pearlitic nodular iron surfaces, milled and carefully cleaned [CLA about 63 (1·6 μm)] and placed in opposition so that the milling marks were directed in opposition and therefore did not interlock. The tests gave dry coefficient of friction of 0·14 which is slightly lower than the lower figure of 0·15 quoted earlier. This value would probably apply also to grey iron.

It was observed that the cleaned surface after the test carried a deposit of graphitic carbon (*see also* Burwell[7]) and at least under heavy pressures the coefficient of friction may be determined by the graphite film rather than the matrix structure, although the reserve of surface graphite is limited in a given surface.

The character of the two surfaces would appear to be of little importance under conditions of full hydrodynamic lubrication, since the load is carried on the oil film. The surfaces, however, are of great importance when starting from rest, in their ability to bed-in without seizure and in their compatibility with lubricants[11].

Cast iron machine tool slideways

The effect of surface finish, lubricant viscosity, load and sliding speed on cast iron slideways has been examined[8] and typical curves are shown in *Figure 63* which shows the effect of various surface finishes and sliding velocity with loads of about 20 lb/in^2 (0·14 N/mm^2). The coefficient of friction

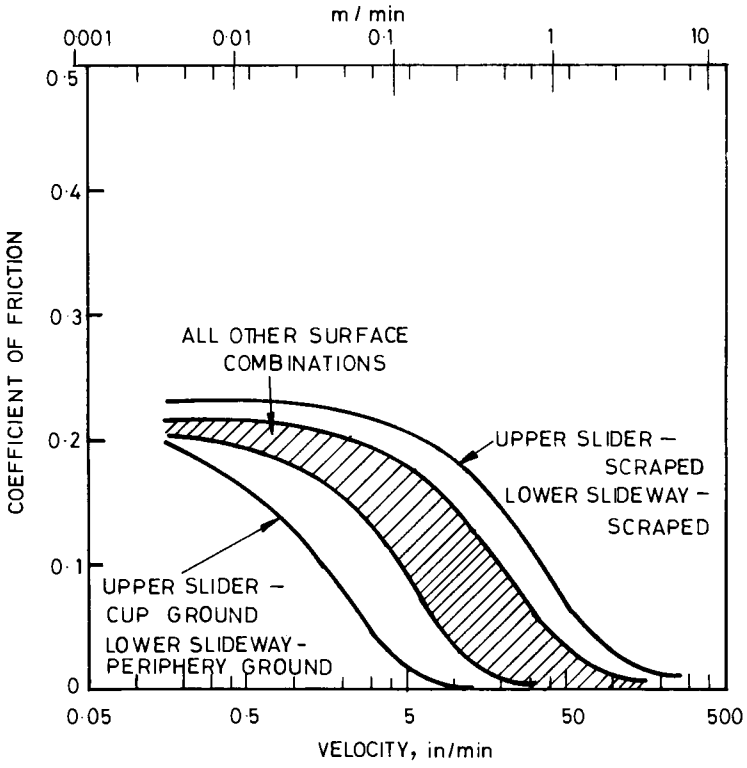


Figure 63

under these conditions was reduced as the viscosity of the lubricant increased. The magnitude of stick-slip decreased in general with velocity until full hydrodynamic lubrication occurred, when the friction was reduced to a very low level, as shown in *Figure 63*.

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DENSITY OF CAST IRON IN THE SOLID STATE

The term 'density' applies to the mass per unit volume of the material, e.g. g/cm^3, lb/in^3 or kg/m^3. The term 'specific gravity' is the relation of the density of the material to that of water at its maximum density (4°C) or at 15°C. The density of water at 4°C is 1 g/cm^3.

Since cast iron is usually composed structurally of ferrite, pearlite, cementite, graphite and phosphide eutectic, the specific gravity of the final material varies appreciably according to the relative amounts of these structural constituents. The densities of these components have been given as follows in g/cm^3:

Ferrite[1]	7·86	
Silico-ferrite, 3·5% Si[2]	7·67	
Pearlite[3]	7·78	
Cementite[4]	7·66	
Graphite[5]	2·25	
Phosphide eutectic[6]	7·32	
Austenite[7] (0·9% C)	7·84	
Martensite[7] (0·9% C)	7·63	α-martensite, 7·58 β-martensite, 7·8

The effect of graphite will be seen to be the most important. In general, also, the higher the carbon, the lower the tensile strength of the iron, and the specific gravity rises as the tensile strength and the hardness of the iron increases.

Carden[8] has studied the data available from various sources[9, 10] and has shown that for laboratory sized samples of cast iron the following expressions indicate the effect of composition on density:

$$\text{Density, lb/ft}^3 = 517 - (13{\cdot}95 \times \text{TC\%}) - (5{\cdot}67 \times \text{Si\%}) - (4{\cdot}43 \times \text{P\%})$$
$$\text{Density, g/cm}^3 = 8{\cdot}11 - (0{\cdot}223 \times \text{TC\%}) - (0{\cdot}091 \times \text{Si\%}) - (0{\cdot}071 \times \text{P\%})$$

Density decreases as carbon equivalent increases, while at a constant carbon equivalent value density increases as the carbon and phosphorus increase, but decreases as the silicon content increases.

The increase in density at constant carbon equivalent value which occurs as the carbon content increases is due to the fact that to maintain a constant

Table 31

	Laboratory samples g/cm³	*Bulk density on actual castings*						
		Actual weight	(kg)	*Section* in	(mm)	*CE value*	*Bulk density* lb/ft³	(g/cm³)
Heavy section low strength iron	6·8	½–1 ton	(508–1017)	14–20	(356–508)	3·84–4·26	430–435	(6·88–6·95)
High phosphorus iron 1¼ in (32 mm) section	7·06–7·14	0·83–1·3 lb	(0·38–0·59)	2	(51)	4·39	446	(7·14)
High phosphorus iron 5/32 in (4 mm) section	7·19–7·12	0·18–1·3 lb	(0·08–0·59)	¼–2	(6–51)	4·39–4·42	444–448	(7·12–7·19)
13–14 ton (201–216 N/mm²) tensile iron ¼–1¼ in (6–32 mm) section	7·2–7·3	17 lb	(7·7)	3¼	(83)	4·18	444	(7·1)
18 ton (278 N/mm²) tensile iron ¼–1¼ in (6–32 mm) section	7·25–7·4	81 lb	(36·7)	2¾	(70)	3·82	447	(7·15)
20–23 ton (309–355 N/mm²) tensile iron ¾–1¼ in (19–32 mm) section	7·4	852 lb	(386)	24 × 8	(610 × 203)	3·63	448	(7·18)
24–28 ton (371–432 N/mm²) tensile iron	7·45	—	—	—	—	—	—	—
White cast iron	7·58–7·73	—	—	—	—	—	—	—
Silal	6·9–7·2	—	—	—	—	—	—	—
Nicrosilal (austenitic)	7·2–7·4	—	—	—	—	—	—	—
Ni-Resist (austenitic)	7·5–7·6	—	—	—	—	—	—	—
Ni-Hard (white, martensitic)	7·6–7·8	—	—	—	—	—	—	—
Grain roll iron	7·46	—	—	—	—	—	—	—
Blackheart malleable	7·25–7·4	—	—	—	—	—	—	—
Whiteheart malleable	7·3–7·7	—	—	—	—	—	—	—
Nodular cast iron	7·1–7·2	—	—	—	—	—	449	(7·2)

Note: Ni-Resist and Ni-Hard are trade names registered by International Nickel Ltd.
1 g/cm³ = 62·428 lb/ft³.

carbon equivalent value with increasing carbon content involves lowering the silicon by three times the rise in carbon content.

Manganese and sulphur variations have no effect on density.

Bulk density

The foregoing deals with the density of small samples, but bulk density of entire castings is normally lower than that obtained from small samples, owing to the effect of porosity.

The determination of true bulk density of entire castings is a matter of considerable difficulty, but *Table 31* includes values for small laboratory samples as well as some accurately determined bulk densities.

Table 32 gives conversion figures for the various units of density.

Table 32

g/cm^3	6·6	6·7	6·8	6·9	7·0	7·1	7·2	7·3	7·4
lb/ft^3	412	419	425	430	437	443	449	455	462
lb/in^3	0·238	0·242	0·246	0·249	0·252	0·256	0·26	0·264	0·268

The following approximate values for bulk densities of BS grades of grey iron are quoted in BS 1452:1961:

BS Grade	*10*	*12*	*14*	*17*	*20*	*23*	*26*
lb/ft^3	430–443	435–450	437–450	440–455	455–460	450–460	450–460
g/cm^3	6·9–7·1	7·0–7·2	7·0–7·2	7·1–7·3	7·3–7·4	7·2–7·4	7·2–7·4

Full annealing of a pearlitic grey iron reduces the density by 0·1 to 0·2 g/cm^3 owing to deposition of graphite caused by the breakdown of pearlite[11].

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DENSITY OF MOLTEN IRON AND STEEL

For general purposes, molten cast iron is usually assumed to have a density of 0·24 lb/in^3 (6·65 g/cm^3), i.e. specific gravity 6·65. However, the determination of densities of molten metals at temperatures above 1200°C offers considerable practical difficulties. Probably one of the earliest determinations was carried out by Mallet[1], who obtained a density of 6·65 g/cm^3 for molten cast iron whose density in the solid was 7·17 g/cm^3. Mallet, however, noted the anomaly that the cold cast iron would float on the molten cast iron.

Sauerwald, Allendorf and Landschutz[2] quote a figure of 7·05 g/cm^3 for molten white iron, and 6·92 g/cm^3 for molten grey iron at 1200°C. Benedicks, Berlin and Phragmen[3], using a method by which the height of a column of liquid metal was compared with that of a column of mercury supporting the same gas pressure, give a figure of 6·92 ± 0·07 g/cm^3 at 1550°C for 0·1 per cent carbon steel. Later work by Berlin[4] gives the following figures for the density of steel (*Table 33*).

Table 33

Temperature °C	*Steel*	
	Carbon-free g/cm^3	*1·2% Carbon* g/cm^3
1700	6·9	6·7
1650	7·0	6·8
1600	7·2	6·9
1550	7·3	7·0
mp	7·4	—

Desch[5], using sinkers of alundum and tungsten powder, obtained the results for steel and cast iron shown in *Table 34*.

These figures were confirmed by Jolly[6], using specially prepared sinkers of steel and cast iron, who obtained the results shown in *Table 35*. The composition of the molten cast iron used for this test was:

	%
Total Carbon	3·12
Silicon	1·96
Manganese	0·98
Sulphur	0·077
Phosphorus	0·8

Table 34

Carbon % *after melting*	*Temperature* °C	*Density* g/cm^3
0·03	1 530	6·79
0·04	1 545	7·05
0·04	1 560	6·97
0·04	1 550	6·95
0·04	1 545	6·97
0·04	1 550	7·06
3·38	1 300	7·26
3·54	1 340	7·23
3·65	1 320	7·31
3·65	1 310	7·28
3·12	1 465	7·17
3·38	1 405	7·23
3·34	1 345	7·32
3·14	1 350	7·23

Table 35

Sinkers	*Density of cold specimen* g/cm^3	*Result*	*Approx. metal temperatures* °C
Cast iron specimen			
(1)	7·10	Floated	1 440
(2)	7·16	Floated	1 440
Steel specimens	7·52	Sank	1 400
	7·50	Sank	1 395
	7·27	Sank	—
	7·12	Just floated	1 385
	6·98	Floated	1 375

At the end of the test the molten metal was cast into bars of different diameters, and the following densities were determined (*Table 36*).

Table 36

Diameter of bar in	*Density of specimen* g/cm^3
1·6	7·207
1·2	7·213
0·875	7·217

These results indicate that the density of this sample of molten iron was between 7·16 and 7·27 g/cm^3, or about 0·26 lb/in^3.

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SPECIFIC HEAT AND TOTAL HEAT

The mean apparent specific heat of pure iron is given by Awbery and Challoner[1] (*Table 37*).

Table 37

	Range of temperature				
	50–100°C	*50–700°C*	*50–950°C*	*650–700°C*	*900–950°C*
Specific heat					
cal/g.°C	0·112	0·147	0·166	0·198	0·226
J/kg.°C	469	615	695	829	946

Pattison and Willows[2] give the specific heat of γ-iron as 0·161 cal/g.°C (674 J/kg.°C) and of solid δ-iron as 0·144 cal/g.°C (603 J/kg.°C). They also quote a figure of 3·7 cal/g (15 491 J/kg) for A_4 heat of transformation.

The specific heat of cementite (Fe_3C) is given by Umino[3] as 0·149 cal/g.°C (624 J/kg.°C) at 150°C, and 0·220 cal/g.°C (921 J/kg.°C) at 850°C.

The specific heat of graphite[4] is given as 0·17 cal/g.°C (712 J/kg.°C) at 20°C, 0·254 cal/g.°C (1063 J/kg.°C) at 138°C, 0·445 cal/g.°C (1863 J/kg.°C) at 642°C and 0·454 cal/g.°C (1901 J/kg.°C) at 896°C.

Umino[3] and Egorenkov[5] have given figures for the specific heat of various irons at different temperatures. Compositions and mean specific heats are given in *Tables 38* and *39*.

Table 38

	TC %	*Si* %	*Mn* %	*S* %	*P* %	*Cu* %
(a) (Pig iron)	4.22	1·48	0·73	0·03	0·12	0·22
(b) (Pig iron)	4·31	1·11	0·53	0·2	0·12	0·21
(c)	3·71	1·50	0·63	0·069	0·147	—
(d)	3·72	1·41	0·88	0·078	0·540	—
(e)	3·61	2·02	0·80	0·08	0·88	—
(f)	2·8	1·48	0·73	0·023	0·12	—

The figures in *Table 39* indicate that the specific heat for high phosphorus iron is appreciably lower than for low phosphorus iron.

Roeser[6], measuring the heat losses from a hot metal car, arrived at a specific heat of 0·23 cal/g.°C (963 J/kg.°C) for molten pig iron at 1350°C.

Table 39

Temperature range °C	*Mean specific heat, 0°–t°C* cal/g.°C (J/kg.°C)					
	(*a*)	(*b*)	(*c*)	(*d*)	(*e*)	(*f*)
0–100	0·131 (548)	—	—	—	—	—
0–200	0·134 (561)	—	0·110 (561)	0·09 (377)	0·069 (289)	—
0–300	0·137 (574)	—	0·118 (494)	0·104 (435)	0·090 (377)	0·133 (557)
0–400	0·140 (586)	—	0·121 (507)	0·111 (465)	0·101 (423)	0·135 (565)
0–500	0·142 (595)	—	0·123 (515)	0·115 (481)	0·107 (448)	0·140 (586)
0–600	0·148 (620	—	0·128 (536)	0·120 (502)	0·112 (469)	0·145 (607)
0–700	0·154 (645)	—	0·144 (603)	0·132 (553)	0·122 (511)	0·153 (641)
0–800	0·168 (703)	—	0·159 (666)	0·152 (636)	0·145 (607)	0·105 (440)
0–900	0·172 (720)	—	0·162 (678)	0·156 (653)	0·151 (632)	0·170 (712)
0–1000	0·175 (733)	0·176 (737)	0·161 (674)	0·156 (653)	0·151 (632)	0·172 (720)
0–1100	0·178 (745)	0·179 (749)	0·160 (670)	0·155 (649)	0·150 (628)	0·175 (733)
0–1200	0·219 (917)	0·22 (921)	0·208 (871)	0·203 (850)	0·199 (833)	0·216 (904)
0–1250	—	0·22 (921)	—	—	—	—
0–1300	0·218 (913)	—	0·203 (850)	0·198 (829)	0·194 (812)	0·217 (909)
0–1350	—	—	0·201 (842)	0·195 (816)	0·191 (800)	—

Note: Umino's results[3]: (a) and (b); Egorenkov's results[5]: (c) to (f)

Table 40 TRUE SPECIFIC HEAT OF WHITE AND GREY IRON 0°C–1000°C [cal/g.°C (J/kg.°C)]

Temperature	*0°C*	*100°C*	*200°C*	*300°C*	*400°C*	*500°C*	*600°C*
White iron[7], 3·15% C	0·13 (544)	0·135 (565)	0·15 (628)	0·155 (649)	0·17 (712)	0·175 (733)	0·185 (775)
Grey iron[7], 3·36% C	0·13 (544)	0·14 (586)	0·15 (628)	0·16 (670)	0·17 (712)	0·19 (795)	0·21 (879)
Grey iron[3], 4·22% C	—	0·13 (544)	0·139 (582)	0·148 (620)	0·149 (624)	0·172 (720)	0·187 (783)
Temperature	*700°C*	*730°C*	*750°C*	*800°C*	*850°C*	*900°C*	*1000°C*
White iron[7], 3·15% C	0·21 (879)	0·23 (963)	0·17 (712)	0·17 (712)	—	0·195 (816)	0·23 (963)
Grey iron[7], 3·36% C	0·28 (1172)	0·33 (1382)	0·25 (1047)	0·22 (921)	0·21 (879)	0·21 (879)	0·25 (1047)
Grey iron[3], 4·22% C	0·207 (867)	—	—	0·194 (812)	—	0·198 (829)	0·202 (846)

The true specific heats for a white iron containing 3·15 per cent carbon[7], a grey iron with 3·36 per cent carbon and a grey iron with 4·22 per cent carbon[3] (iron 'a' in *Tables 38* and *39*) are shown in *Table 40*.

The mean apparent specific heat over a very short temperature range has been determined for certain steels by Awbery and Challoner[1], and one of these steels (total carbon, 0·485 per cent; silicon, 1·98 per cent; manganese, 0·9 per cent; sulphur, 0·047 per cent; phosphorus, 0·44 per cent; nickel, 0·156 per cent; copper 0·637 per cent) gave the figures shown in *Figure 64*, based

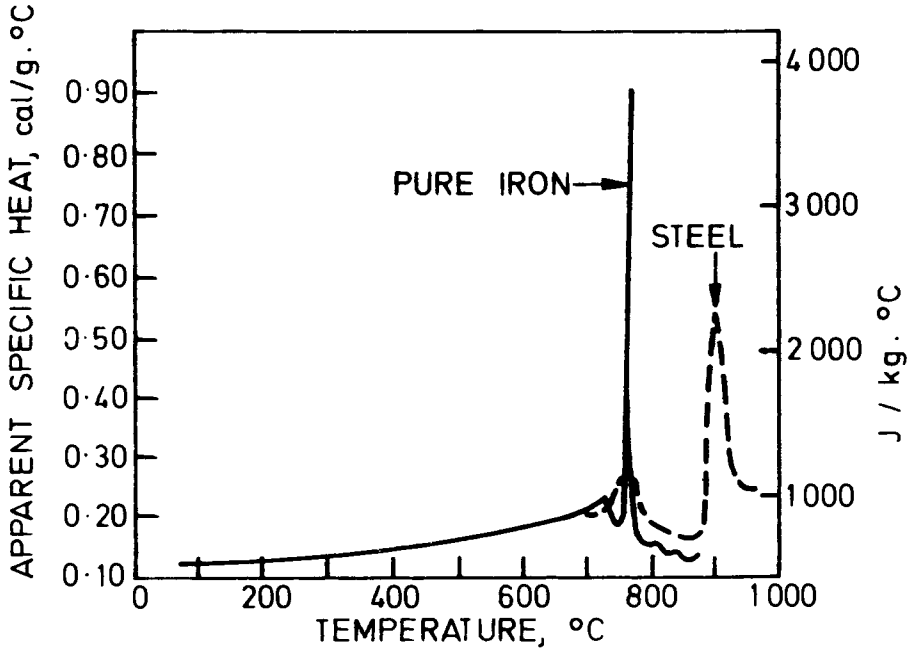


Figure 64. Mean apparent specific heat curves for pure iron and steel with 1·98% Si and 0·485% C

on determinations over ranges of approximately 10 degrees Celsius. A corresponding curve for pure iron is also shown in *Figure 64*. The work of Egorenkov shows similar curves for the cast irons previously quoted.

Total heat

A curve of the total heat requirement for pure iron and for two of the cast irons referred to above is given in *Figure 65* as determined by Egorenkov[5] and Awbery and Challoner[1].

Egorenkov's total heat figures are shown in *Table 41*.

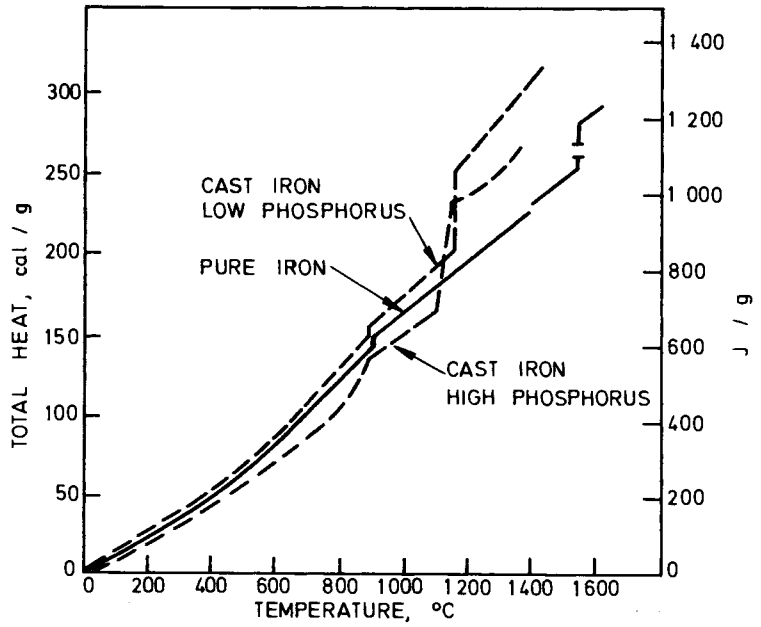


Figure 65. Total heat of two cast irons compared with pure iron

Table 41

Temperature range °C	*Total heat* cal/g (J/g)			*Pure iron from* 50°C[8]
	(c)[5]	(e)[5]	(f)[5]	
0–200	22·0 (92)	13·8 (58)	—	17·6 (74)
0–300	35·3 (148)	27·1 (113)	40·0 (167)	30·4 (127)
0–400	48·5 (203)	40·3 (169)	54·0 (226)	44·1 (185)
0–500	61·7 (258)	53·5 (224)	70·0 (293)	59·2 (248)
0–600	76·6 (321)	67·5 (283)	87·0 (364)	76·2 (319)
0–700	100·7 (422)	85·7 (359)	107 (448)	95·3 (399)
0–800	127·3 (533)	116·1 (486)	132 (553)	117·8 (493)
0–900	145·4 (609)	135·8 (569)	153 (641)	135·3 (566)
0–1000	161·3 (675)	151 (632)	172 (720)	153·3 (642)
0–1100	176·5 (739)	165·5 (693)	192 (804)	167·4 (701)
0–1150	243·0 (1017)	231·5 (969)	—	174·7 (731)
0–1200	250·0 (1047)	238·5 (999)	260 (1089)	182·3 (763)
0–1300	264·3 (1107)	251·8 (1054)	282 (1181)	198 (829)
0–1350	271·5 (1137)	258·0 (1080)	—	—

Malleable irons

Direct calorimetric determinations of mean specific heat of malleable iron between room temperature and various higher temperatures[9] are given in *Table 42.*

Table 42

Temperature range °C	*Mean specific heat* cal/g. °C(J/kg. °C)
20–100	0·122 (511)
20–200	0·125 (523)
20–300	0·128 (536)
20–400	0·133 (557)
20–500	0·139 (582)
20–600	0·146 (611)
20–700	0·159 (666)

Note: Specific heat values in terms of calories per gram per °C, and Btu per pound per °F are numerically identical. To convert calories per gram into Btu per pound multiply by 1·8.

The Swiss standard[10] gives a figure for the specific heat of whiteheart malleable iron of 0·11–0·12 cal/g. °C (461–502 J/kg. °C) for a temperature range of 10–100°C.

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LATENT HEAT OF FUSION

There is some discrepancy in the various figures that are available:

	cal/g	(kJ/kg)
Pure iron[1,2]	65	(272)
	65·5	(274)
High carbon cast iron[3] (TC 4·22%; Si 1·48%; Mn 0·73%; S 0·03%; P 0·12%)	47	(197)
High carbon white iron[4] (TC 4·31%; Si 1·11%; Mn 0·53%; S 0·02%; P 0·12%; Cu 0·21%)	46·6	(195)
High carbon white iron[5] (TC 4·35%)	59	(247)
For grey iron (TC 2·5–3·5%; Si 1·5–2·5%) a probable range is	50–55	(209–230)
Whiteheart and blackheart malleable iron[6]	23	(96)

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SURFACE TENSION OF CAST IRON

The surface tension of cast iron has been measured by various methods [1,2,3] and values from 800 to 2100 dyn/cm (0·8 to 2·1 N/m) have been recorded for irons of varying composition. Figures from 1000–1200 dyn/cm (1·0–1·2 N/m) are characteristic of normal grey irons. BCIRA work has given a figure of 845 dyn/cm (0·845 N/m) at 1560°C for an iron with TC 3·75, Si 2·07 and P 0·019 per cent.

CARBON

As the carbon content increases, the surface tension decreases in the hypoeutectic range, and increases in the hypereutectic range.

The eutectic value (4·0 per cent C) was 1070–1120 dyn/cm (1·07–1·12 N/m) at 1400–1500°C, and at 2·0 per cent carbon the value ranged from 1170 to 1200 dyn/cm (1·17 to 1·20 N/m).

SILICON

Silicon has little effect on surface tension, each 1 per cent addition lowering the value by about 12 dyn/cm (12 mN/m).

MANGANESE

Manganese has no significant effect below 1·5 per cent.

PHOSPHORUS

Phosphorus lowers the value by about 30 dyn/cm (30 mN/m) for each 1 per cent.

SULPHUR

Sulphur reduces surface tension markedly from about 1200 dyn/cm (1·2 N/m) at 0·04 per cent sulphur to about 800 dyn/cm (0·80 N/m) at 0·2 per cent sulphur for temperatures between 1300 and 1400°C.

MAGNESIUM AND CERIUM

Additions of cerium and magnesium apparently increase the surface tension of cast iron and nodular irons usually have values between 1300 and 1400 dyn/cm (1·3 and 1·4 N/m). However, if their effects are studied at constant sulphur and oxygen contents it is found that both magnesium and cerium have a marked effect in decreasing the surface tension of cast iron.

TEMPERATURE

Surface tension increases with temperature, e.g. from 1100 dyn/cm (1·10 N/m) at 1300°C to 1170 dyn/cm (1·17 N/m) at 1400°C, but falls again above 1500°C.

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VISCOSITY OF MOLTEN IRON

Some values for the viscosity of iron–carbon alloys at various temperatures are given by Gilbert Cavalier and Louis-Didier Lucas[1]. Their results are given in *Table 43*, relating to variation in carbon per cent weight and temperature.

A summary of the methods and results employed for iron–carbon alloys and the effect of temperature, carbon, phosphorus, manganese silicon and sulphur has been given by Linacre[2] and for iron–carbon alloys and the effect of temperature, carbon and sulphur by Barfield and Kitchener[3].

Fuller[4] has summarised the published data on properties of liquid metals and it appears that:

(1) Increasing carbon content produces a pronounced minimum in the isothermal viscosity curves around 2 per cent by weight.
(2) Increasing silicon content decreases viscosity.
(3) Increasing manganese content decreases viscosity, but is dependent on carbon content and the effect is non-linear.
(4) Increasing phosphorus content decreases viscosity.
(5) Increasing sulphur content increases viscosity.

Table 43

Carbon wt.%	*Liquidus temperature* °C	*Eutectic temperature*† °C	η							*Activation energy of viscous flow Q* (kcal/mole)	(kJ/mole)
			melting point	1200°C	1300°C	1400°C	1500°C	1600°C	1700°C		
0·0	1536		5·03	—	—	—	5·32*	4·58	4·00	9·9	(41)
0·36	1504	—	5·04	—	—	—	5·06*	4·29	3·69	11·0	(46)
0·45	1497	—	5·07	—	—	—	5·02	4·27	3·66	11·2	(47)
0·52	1492	—	5·13	—	—	—	4·99	4·23	3·62	11·3	(47)
0·95	1469	—	4·94	—	—	—	4·72	4·08	3·58	9·6	(40)
2·07	1378	—	5·64	—	—	5·41	4·47	3·79	3·25	11·1	(46)
2·47	1343	—	5·64	—	—	5·16	4·48	3·94	3·52	8·4	(35)
3·18	1275	—	7·18	—	6·75	5·47	4·55	3·86	3·33	10·9	(46)
3·96	1182	1147	10·53	10·19	7·90	6·31	5·18	4·32	3·69	11·8	(49)
4·02	1180	1146·9	10·86	10·26	7·91	6·29	5·15	4·30	3·66	11·9	(50)
4·12	1170	1148	11·07	10·09	7·85	6·27	5·12	4.29	3·66	11·9	(50)
4·23	1155	1147	11·38	10·05	7·84	6·20	5·08	4·25	3·66	11·8	(49)
4·42	—	1147·3	—	10·25	7·95	6·31	5·20	2·35	3·72	11·9	(50)
4·45	—	1148·4	—	9·98	7·80	6·25	5·14	4·31	3·69	11·5	(48)
4·76	—	1147	—	—	—	6·43	5·38 .	4·58	3·98	10·5	(44)

η = viscosity; units in centipoise [= mNs/m^2] [= 0·001 Pa s]

* = undercooled

† = eutectic determined for 4·92% carbon at 1147·5°C (average of seven values)

Viscosity affects the flow of iron through tapholes and a curve relating cast iron flow to diameter of taphole under a ferrostatic head of about 11 in (279 mm) is given in *The Cupola and Its Operation* (American Foundrymen's Society, 1954, p. 68) and indicates a rate of about 24 tons per hour through a taphole of 1 in (25 mm) diameter.

The rate of flow will vary directly with the taphole area under a given head and directly with the square root of the head for a given area. Viscosity also affects fluidity or the flow of metal in moulds and has been discussed by Evans[5] who showed that fluidity of cast iron was directly related to a modified carbon equivalent value $\left(\text{CEF} = \text{TC} + \frac{\text{Si}}{3} + \frac{\text{P}}{2}\right)$.

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THERMAL CONDUCTIVITY

The thermal conductivity of cast iron is partly dependent upon its structure. Figures for the main structural constituents of cast iron at room temperature are given in *Table 44*. The thermal conductivity of graphite along the basal plane is shown in *Figure 66*.

Table 44

	Thermal conductivity cal/cm/cm²/s/deg C | (W/m.°C)		
	0–100°C	*500°C*	*1000°C*
Graphite along C-axis	approx. 0·2 (84)	—	—
Graphite along basal plane[4, 5] *(see Figure 66)*	0·7–1·0 (293–419)	0·2–0·3 (84–126)	0·1–0·15 (42–63)
Ferrite[1, 4]	0·17–0·19 (71–80)	0·1 (42)	0·07 (29)
Pearlite[2, 3, 6]	0·12 (50)	0·105 (44)	—
Cementite[2, 3]	0·017 (7)	—	—

Pearlite shows a marked reduction in thermal conductivity of about 0·01 cal/cm/cm²/s/deg C (4 W/m.°C) at the Curie point (220°C). There is a very slight fall with rise of temperature of about 0·0017 (0·7) per 100°C both below and above this point up to the critical temperature[6].

The thermal conductivity of these constituents falls with rising temperature and for graphite along the basal plane is of the order of 0·7 (293) to 1·0 (419) at room temperature and 0·15 (63) to 0·28 (117) at 900°C (SI units in brackets).

Some austenitic steels and alloyed cast irons may show a rise in conductivity with temperature[7] and a 0·015 per cent C, 2 per cent Si iron alloy has shown the same effect[8].

The thermal conductivity of graphite varies with the crystallographic plane and is highest parallel to the basal plane. A graphite flake may be regarded as a curved basal section of a graphite crystal and its thermal conductivity is greatest along its length. The thermal conductivity in a direction perpendicular to the basal plane is about one-fifth of the above figures and is probably rather higher than for ferrite[9].

It would be expected from the above figures, therefore, that a high carbon ferritic iron would have a higher thermal conductivity than a low carbon pearlitic iron, and that the presence of free cementite, as in mottled or chilled iron, would lower the thermal conductivity.

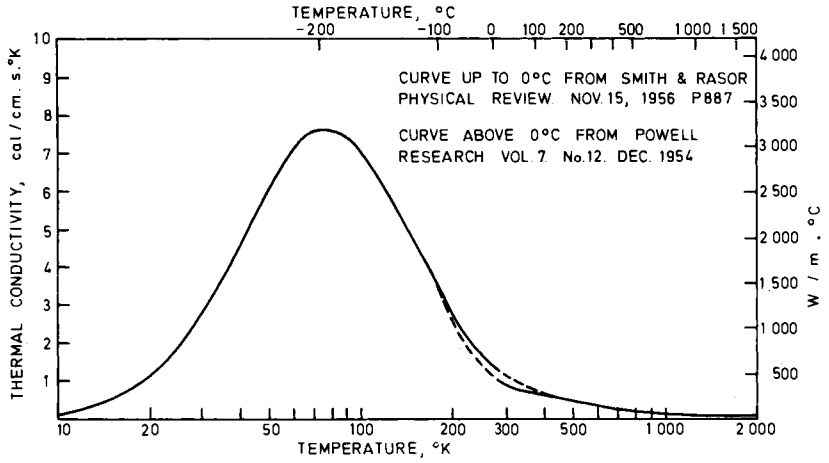


Figure 66. Thermal conductivity of natural graphite along the basal plane

These assumptions are largely borne out in practice, but are subject to interferences both by alloying constituents and the physical form adopted by the graphite. The most comprehensive series of papers on this subject has been published by Donaldson[3,10,11] and Söhnchen[12], and these, together with other published information and work carried out by BCIRA, show clearly that of the alloying constituents in the matrix, silicon is the most significant.

CARBON

Increasing carbon content may decrease the thermal conductivity because of the consequent increase in the amounts of one or more of the low conductivity micro-constituents, i.e. pearlite and cementite. An increase in the graphite content of otherwise similar irons, however, increases the thermal conductivity[12].

SILICON

According to Donaldson[11] and Söhnchen[12], silicon lowers the conductivity of cast iron. The general effect is shown in *Figure 67*.

MANGANESE

Increasing manganese decreases thermal conductivity by about 3·3 per cent for an addition of about 1·5 per cent manganese[11].

PHOSPHORUS

Phosphorus gives a slight decrease in thermal conductivity, equivalent to a fall of about 6 per cent for a 1 per cent rise in phosphorus content[3].

ALUMINIUM

Aluminium has a pronounced influence in lowering the thermal conductivity of grey cast iron; according to Donaldson[11], an alloy of total carbon 2·7 per cent, silicon 0·96 per cent, manganese 0·58 per cent, aluminium 7 per cent, and chromium 0·95 per cent, has a thermal conductivity of 0·079 cal/cm/cm^2/s/deg C (33 W/m.°C) at 100°C, and 0·072 (30 W/m.°C) at 400°C.

CHROMIUM

Data on the effect of chromium are conflicting, but Donaldson[10] shows a rise in conductivity of 7 per cent for 0·5 per cent chromium additions. Söhnchen,[12] however, states that a decrease occurs since chromium stabilises low conductivity iron carbide, and his figures show a drop in conductivity of nearly 15 per cent for an addition of 0·5 per cent chromium.

COPPER

According to Lorig and Schnee[13], additions of copper up to 2 per cent lower the thermal conductivity slightly, whereas greater additions up to 3 per cent have little or no increased effect. Donaldson[11] states that the effect of copper is about half that of silicon. Maréchal and Listray[14] indicate that copper reduces the thermal conductivity slightly, but the effect is not very significant.

NICKEL

An addition of 1 per cent of nickel may be assumed to lower the thermal conductivity by about 5–6 per cent. However, if silicon is simultaneously decreased on the addition of nickel, conductivity may be increased[10].

MOLYBDENUM

Molybdenum, according to Donaldson[11], tends to increase the thermal conductivity by about 7 per cent for an addition of 0·58 per cent molybdenum.

TUNGSTEN

Donaldson[10,11] reports that tungsten increases the thermal conductivity of cast iron by about 7 per cent for an addition of approximately 0·5 per cent tungsten.

VANADIUM

Donaldson[10] shows no change in thermal conductivity with an addition of vanadium of 0·12 per cent.

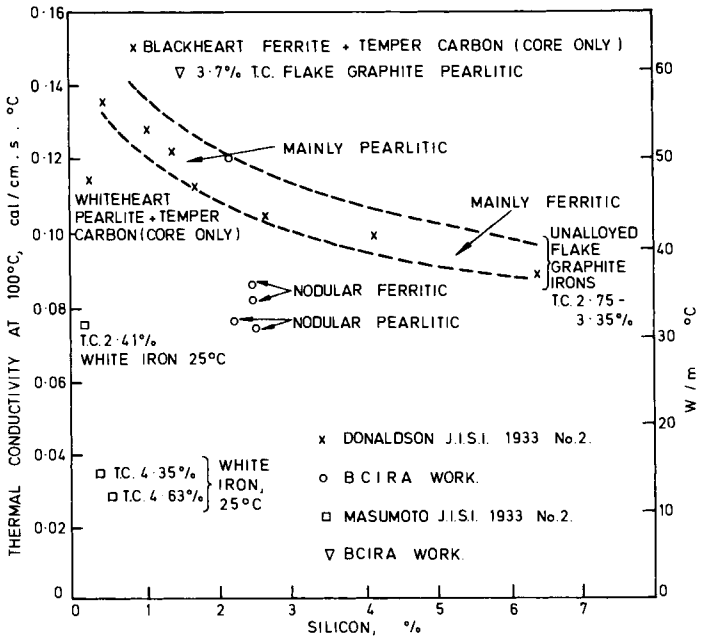


Figure 67. Thermal conductivity of cast iron at 100°C

The curves in *Figure 67* illustrate the dependence of thermal conductivity upon the silicon content in ordinary grey iron. Additions of practically all other alloying elements reduce the thermal conductivity, unless they have a secondary effect upon structure; for example, by the conversion of cementite into graphite and pearlite, or graphite and ferrite. It should be observed that although the addition of silicon lowers the thermal conductivity, one of the effects of increasing silicon content is to increase the amounts of both graphite and ferrite, both of which have relatively high thermal conductivities.

Of the structural constituents normally present in cast iron, graphite has the highest thermal conductivity. It is to be expected, therefore, that its effect will be not merely dependent upon the amount present, but also upon the form in which it is present. Thus, for irons in which the graphite is present in random flake form, the thermal conductivity increases as the graphite

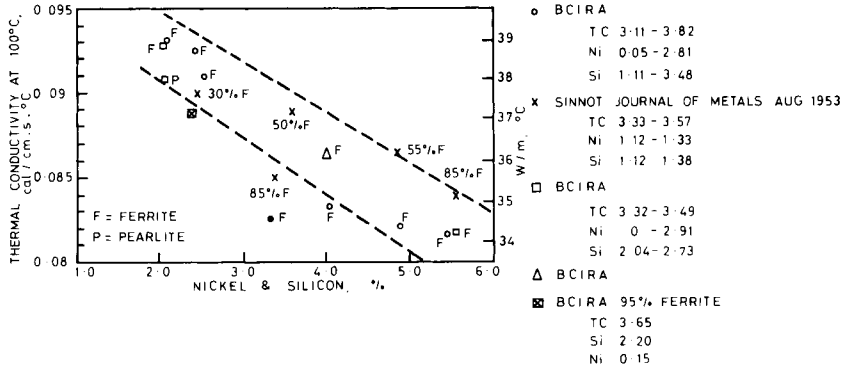


Figure 68. Thermal conductivity of ferritic nodular iron

increases. If, however, the graphite is distributed in nodules or in the compacted form of flake graphite associated with higher nitrogen contents, then for the same silicon content the thermal conductivity appears to be substantially lower than for flake graphite irons, presumably because the nodular form offers less free path than the flake form. This is illustrated in *Figure 68.*

Since the amount of graphite present in nodular iron is a comparatively small percentage of the total volume, its effect is likely to be less significant, in spite of its high conductivity, than the effect of matrix composition and differences between pearlitic and ferritic irons.

The effect of annealing a pearlitic flake iron is shown in *Table 45* and nodular ferritic iron has a slightly higher conductivity than nodular pearlitic iron. Donaldson shows ferritic blackheart malleable (core only) to have a markedly higher thermal conductivity than whiteheart malleable, but other results do not confirm this. The effect of the graphite content is very marked and *Table 45* shows a 20 per cent increase in conductivity when the carbon is raised from 3·18 per cent to about 4 per cent. The higher thermal conductivity of high carbon irons makes them particularly suitable for conditions involving thermal shock, such as ingot moulds brakedrums etc.

Table 45 THERMAL CONDUCTIVITY OF VARIOUS CAST IRONS AND STEEL

Material	*Ref.*	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*Mean temp.* °C	*K* cal/cm/cm²/s/deg C	(W/m °C)	*Mean temp.* °C	*K* cal/cm/cm²/s/deg C	(W/m °C)
High carbon	16	4·1	2·44	0·092	0·068	0·052	100	0·142	(59)	400	0·112	(47)
High carbon	16	3.73	1·66	1·12	0·059	0·087	100	0·140	(59)	400	0·110	(46)
High carbon*	16	3.98	1·32	0·66	0·074	0·079	100	0·136	(57)	400	0·101	(42)
High carbon	16	3·58	1·52	0·53	0·101	0·06	100	0·133	(56)	400	0·112	(47)
Normal grey												
Pearlitic†	16	3·18	2·08	0·60	0·112	0·097	100	0·117	(49)	400	0·096	(40)
Annealed ferritic†	16	3·18	2·08	0·60	0·112	0·097	100	0·127	(53)	400	0·104	(44)
Low Silicon												
Perlit	5	3·35	0·65	—	—	—	90	0·136	(57)	430	0·112	(47)
	17	3·09	0·55	1·09	0·103	0·18	100	0·155	(65)	400	0·108	(45)
White iron	2	2·41	0·12	—	—	—	25	0·076	(32)	—	—	—
Blackheart malleable‡	3	2·36	1·03	—	—	—	100	0·151	(63)	400	0·139	(58)
Ferritic	17	2·51	1·01	0·56	0·183	0·047	100	0·117	(49)	400	0·111	(46)
Pearlitic	17	2·51	1·06	1·16	0·159	0·045	100	0·104	(44)	400	0·099	(41)
Whiteheart malleable§	3	2·8	0·39	—	—	—	100	0·114	(48)	400	0·105	(44)
Pure iron	18	—	—	—	—	—	100	0·165	(69)	400	0·116	(49)
Steel	7	0·23					100	0·122	(51)	400	0· 02	(43)
	7	0·4					100	0·115	(48)	400	0·098	(41)
	7	1·2					100	0·108	(45)	400	0·089	(37)

* 0·16 per cent chromium; 0·31 per cent molybdenum.

† 0·14 per cent chromium.

Note: 1 kcal/m². min = 0·3687 Btu/ft². min = 0·006 98 W/cm².

‡BS 310: 1958 Amendment 1 1962 gives 0·12–0·15 cal/cm.s.°C. Falls with increasing temperature.

§ BS 309: 1958 Amendment 2 1962 gives 0·11–0·15 cal/cm.s.°C increases with total carbon, decreases with temperature.

Table 46 THERMAL CONDUCTIVITY OF VARIOUS ALLOY IRONS

Material	*Ref.*	*TC* %	*Si* %	*Ni* %	*Cr* %	*Cu* %	*Average temp.* °C	*K* cal/cm/cm²/s/deg C	(W/m °C)	*Average temp.* °C	*K* cal/cm/cm²/s/deg C	(W/m °C)
Silal	3	2·75	6·49	—	—	—	100	0·088	(37)	430	0·082	(34)
Nicrosilal	3	1·81	6·42	18·7	2·0	—	80	0·071	(30)	430	0·063	(26)
Nickel—chromium cast iron	6	2·8	2·5	1·7	0·5	—	100	0·101	(42)	400	0·092	(39)
Austenitic cast iron	6	2·4	1·8	13·7	3·4	6·4	100	0·081	(34)	400	0·075	(31)
30% chromium		1·7	1·5	—	30·0	—	87	0·050	(21)	453	0·055	(23)
Acicular cast iron	16	3·3	1·06	3·95	0·57	Mo 1·08	100	0·087	(36)	400	0·077	(32)
Nodular pearlitic	17	3·22	2·44	1·35	Mg 0·056	—	100	0·0742	(31)	400	0·0718	(30)
Nodular ferritic	17	3·52	2·05	0·05	Mg 0·066	—	100	0·0929	(39)	400	0·0911	(38)
Nodular austenitic	17	2·95	1·85	20·7	Mg 0·12	—	100	0·0455	(19)	400	0·0437	(18)

White irons

White irons[2] consisting primarily of carbide and pearlite, show very much lower thermal conductivities than grey irons, the thermal conductivity falling rapidly as the carbon and free carbide content increases (*Figure 67*).

Austenitic cast iron

The thermal conductivities of these irons are 20–30 per cent below that of ordinary grey iron.

Nodular graphite irons

In irons in which nodular graphite is formed by the addition of magnesium or cerium and magnesium, with or without nickel, the thermal conductivities are 20–30 per cent lower than those obtained for grey irons of a similar content. This is undoubtedly associated with the form of the graphite, which offers a much less effective free path through the body of the metal. Nickel-free nodular irons may be 8–10 per cent higher in conductivity than nickel-bearing irons, and the thermal conductivity of ferritic nodular irons is 10–15 per cent higher than for pearlitic nodular irons. For mainly ferritic nodular irons, the thermal conductivity varies roughly inversely as the sum of the nickel and silicon content (*Figure 68*).

Effect of temperature

The thermal conductivity (cal/cm/cm^2/s/deg C) of grey irons over a wide range of composition falls by 0·0035 to 0·0045 (1·5 to 1·9 W/m.°C) for each 100 deg C rise between about 100°C and 450°C. For austenitic cast irons of the Ni-Resist and Nicrosilal types, the fall is approximately 0·002 (0·8 W/m.°C) for each 100 deg C rise between 100°C and 450°C.

The figures given in *Tables 45* and *46* are fairly typical for various cast irons. The figures for steel have been included for comparison purposes.

Effect of stress on thermal conductivity

When a cast iron is stressed in tension, the spaces occupied by the graphite increase in volume. On unloading only partial recovery occurs and the spaces occupied by the graphite have a greater volume than before stressing. The effect of prestressing a grey iron up to 90 per cent of its tensile strength has been found to have no effect on the subsequent thermal conductivity of the iron[15].

Thermal diffusivity[7] may be significant where locally applied heat is dispersed to the surrounding material and is expressed as:

$$\text{Thermal diffusivity } (D) = \frac{\text{Thermal conductivity}}{\text{Density} \times \text{Specific heat}}$$

For pearlitic grey iron,

$$D = \text{approx.}\ \frac{0{\cdot}120\,\text{cal/cm/cm}^2\text{/s/°C}}{7{\cdot}3\,\text{g/cm}^3 \times 0{\cdot}14\,\text{cal/g.°C}} = 0{\cdot}12\,\text{cm}^2\text{/s}$$

or

$$\frac{50\,\text{W/m.°C}}{7300\,\text{kg/m}^3 \times 586\,\text{J/kg.°C}} = 0{\cdot}000\,012\,\text{m}^2\text{/s}$$

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THERMAL EXPANSION

The rate of linear expansion of cast iron increases with rising temperature, so that in quoting coefficients of thermal expansion it is important to give either the range of temperature over which the tests were taken, or the mean temperature if the range is narrow.

At a temperature between 200 and 300°C (the lower Curie point) the rate of expansion or contraction of grey cast iron undergoes a change. At this point the iron carbide in the metallic matrix loses or gains its magnetic properties on heating or cooling respectively (*see* section on Magnetic Properties), and above this point there is an increase in rate of expansion.

The effect upon the reversible thermal expansion of the normal elements carbon, silicon, manganese, sulphur and phosphorus in the amounts usually found in grey iron is very small. Driesen[1] states that in a substantially pure iron–carbon series of alloys, raising the carbon from 0·06 per cent to 3·66 per cent lowers the coefficient of thermal expansion over the range 0 to 100°C from 11·7 to $8{\cdot}6 \times 10^{-6}$/°C, but the effect of carbon appears to be much less marked than this in the case of normal grey irons, and there is no indication

Table 47 THERMAL EXPANSION OF CAST IRON

		Composition					*Linear coefficient of thermal expansion,* $\times 10^{-6}/°C$									
	Ref.	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*0–100°C*	*0–200°C*	*0–300°C*	*0–400°C*	*0–500°C*	*0–600°C*	*0–200°C*	*0–300°C*	*0–400°C*	*0–500°C*
Electrolytic iron	2	—	—	—	—	—	12·0	12·6	13·3	13·8	14·3	14·7	—	—	—	—
Ingot mould (pearlitic)	3	3·77	1·81	0·43	0·096	0·037	—	12·0	—	12·5	—	12·8	—	—	—	—
Ingot mould (pearlitic)	3	3·49	2·07	0·73	0·039	0·042	—	12·5	—	12·5	—	12·8	—	—	—	—
Ingot mould (pearlitic)	3	3·47	2·45	0·78	0·083	0·029	—	11·9	—	12·5	—	13·3	—	—	—	—
Grey iron (low carbon)	4	3·12	2·0	0·93	0·151	0·255	10·6	11·3	11·9	—	—	—	11·7	14·2	15·6	14·3
Grey iron (high carbon)	4	3·66	1·44	0·85	0·129	0·29	10·4	11·1	11·7	12·7	12·7	—	—	—	—	—
Heat resisting iron (Silal)	5	2·25	5·8	—	—	—	—	11·3	11·6	12·0	12·0	12·9	—	—	—	—
High silicon iron (Duriron)	2	0·41	14·7	0·75	0·015	0·023	12·2	12·9	13·7	14·6	15·4	16·3	—	—	—	—

in later figures that the reversible coefficient of expansion will vary significantly within the normal range of composition run by a single foundry*.

Typical figures from the literature are given in *Table 47*. Figures for three high phosphoric irons are given in *Table 48*. The range of composition[6] was:

Total carbon	3·2–3·4 per cent
Silicon	2·8–3·0 per cent
Manganese	0·4–0·6 per cent
Sulphur	0·1–0·12 per cent
Phosphorus	0·8–1·0 per cent

Table 48

Linear coefficient of thermal expansion, $\times 10^{-6}$/°C						
20–120°C	*20–190°C*	*20–285°C*	*20–310°C*	*20–400°C*	*20–480°C*	*20–575°C*
10·9[a]	11·7	12·2	12·2	12·8	13·1	13·2
10·9[b]	11·7	12·0	12·0	12·4	12·7	13·1
11.2[c]	11·8	12·1	12·2	12·8	13·1	13·3

Note: a, b, c. Structure at commencement: (a) and (b) mainly ferritic, (c) mainly pearlitic.

Table 49

Test bar no.	*TC* %	*Si* %	*Mn* %	*S* %	*P* %
Melt 1 No. 1	3·06	2·57	0·45	0·116	0·42
Melt 1 No. 2	—	—	—	—	0·67
Melt 1 No. 3	—	—	—	—	0·92
Melt 1 No. 4	—	—	—	—	1·08
Melt 1 No. 5	2·97	2·55	0·42	0·112	1·18
Melt 2 No. 1	3·15	0·98	0·44	0·131	0·43
Melt 2 No. 2	—	1.57	—	—	—
Melt 2 No. 3	—	2·00	—	—	—
Melt 2 No. 4	—	2·30	—	—	—
Melt 2 No. 5	2·86	2·62	0·45	0·124	0·40
Melt 3 No. 1	3·05	1·43	0·41	0·126	0·38
Melt 3 No. 2	—	—	—	—	0·69
Melt 3 No. 3	—	—	—	—	0·95
Melt 3 No. 4	—	—	—	—	1·23
Melt 3 No. 5	2·82	1·41	0·30	—	1·01
Melt 4 No. 1	2·95	0·96	0·54	0·124	1·10
Melt 4 No. 2	—	1·50	—	—	—
Melt 4 No. 3	—	1·99	—	—	—
Melt 4 No. 4	—	2·35	—	—	—
Melt 4 No. 5	2·81	2·77	0·43	—	1·03

Figures covering a range of compositions for temperatures above the lower Curie point are given by Palmer[7]:

Melt No. 1 Silicon 2·5 per cent with increasing phosphorus content.
Melt No. 2 Phosphorus 0·4 per cent with increasing silicon content.

* The thermal coefficients of expansion, per °C, of graphite have been given as: Basal plane, $6{\cdot}6 \times 10^{-6}$; *C* axis, $26{\cdot}0 \times 10^{-6}$ (*see* Ref. 13)

Melt No. 3 Silicon 1·4 per cent with increasing phosphorus content.
Melt No. 4 Phosphorus 1·2 per cent with increasing silicon content.

The chemical analyses obtained on the test bars cast from these melts are given in *Table 49*.

The value of the thermal expansion coefficient over a fixed range of temperature, both on heating and on cooling, was calculated for each specimen tested. The results are given in *Table 50*.

Table 50

Specimen no.	*Linear coefficient of thermal expansion,* $\times 10^{-6}$/°C	
	Heating (253 to 664°C)	*Cooling (664 to 253°C)*
Melt 1 No. 1	14·22	13·96
Melt 1 No. 2	14·20	13·70
Melt 1 No. 3	14·09	12·97
Melt 1 No. 4	14·36	13·07
Melt 1 No. 5	13·81	13·04
Melt 2 No. 2	14·58	14·24
Melt 2 No. 3	14·33	13·84
Melt 2 No. 4	14·09	13·71
Melt 2 No. 5	14·46	13·64
Melt 3 No. 1	13·86	13·87
Melt 3 No. 2	13·97	13·89
Melt 3 No. 3	14·46	14·06
Melt 3 No. 4	14·58	13·41
Melt 3 No. 5	14·46	13·64
Melt 4 No. 1	14.58	13.39
Melt 4 No. 2	14·09	12·69
Melt 4 No. 3	14·34	13·46
Melt 4 No. 4	14·55	13·51
Melt 4 No. 5	14·01	13·43

Thermal expansion at low temperatures

The linear coefficient of expansion of a 4 per cent carbon cast iron (other elements unknown) has been given[4] as $8{\cdot}5 \times 10^{-6}$/°C for a temperature range of -191 to $+16$°C.

Low expansion grey cast iron

Austenitic cast iron with a range of coefficients of thermal expansion can be obtained by the addition of substantial quantities of nickel[8, 9]. Coefficients of thermal expansion from 10 to 93°C of 3·0 to $4{\cdot}0 \times 10^{-6}$/°C can be obtained with iron containing 34 per cent to 36 per cent nickel. A typical range of composition is quoted as total carbon 2·4 per cent max.; silicon, 1·0–2·0 per cent; manganese, 0·5 per cent max.; nickel, 34–36 per cent; chromium, 0·1 per cent max.; sulphur, less than 0·12 per cent; phosphorus, less than 0·2 per

cent. The coefficient of expansion increases markedly with temperature rise, and the mean figure from 10 to 205°C within this nickel range is from 5 to 6×10^{-6}/°C.

Austenitic cast iron

Austenitic cast irons in the range of 14–22 per cent nickel have a thermal expansion of approximately 19×10^{-6}/°C (20–200°C)[10]. This marked increase must be borne in mind wherever austenitic materials are employed as constituent parts of a structure which includes other materials with different coefficients of expansion, and which is subjected to an appreciable range of temperature.

Malleable cast iron

The coefficient of thermal expansion of whiteheart and blackheart at temperatures up to about 400°C is approximately the same as that for grey iron, namely, 10 to 12×10^{-6}/°C.

White iron

The coefficient of thermal expansion from room temperature to 276°C is approximately 10 to 11×10^{-6}/°C and from 276 to 684°C from 15·9 to $16{\cdot}4 \times 10^{-6}$/°C. A marked increase in expansion occurs at the Curie point (200–300°C).

Temperatures above 400–450°C

When taken to temperatures of about 450°C and above, the thermal expansion of unalloyed grey cast iron ceases to be strictly reversible, owing to various effects, including graphitisation in pearlitic irons and also oxidation. A slight increase may occur just below the critical transformation temperature, owing to sub-critical graphitisation, and a marked increase occurs at about 960°C, but at 1000–1050°C under non-oxidising conditions, a very marked contraction occurs with phosphorus contents of 0·9 per cent and above, caused by solution of graphite by the molten phosphide eutectic. On cooling from above the critical temperature, a marked expansion occurs at the critical temperature itself. The critical transformation temperature and the amount of expansion increases with silicon content. The final dimensions of the specimen are made up of the sum of the various reversible and irreversible dimension changes, and in low phosphorus irons tends to increase with increasing silicon[7], a single heating to 1050°C in air causing permanent increases in length of up to 1 per cent with silicon contents of about

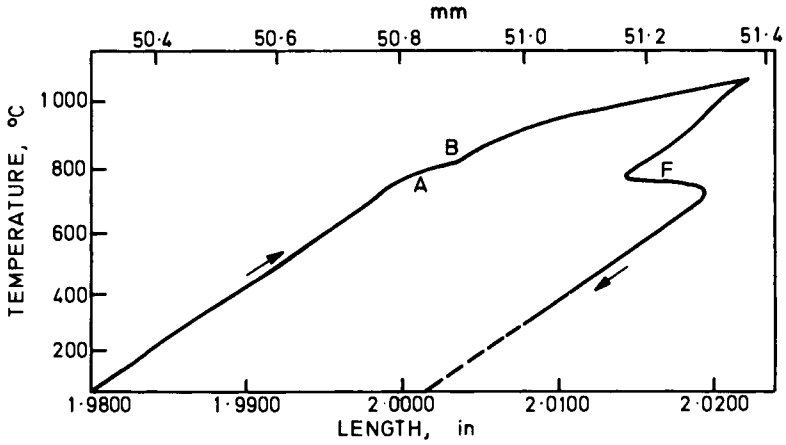


Figure 69. TC 2·86%, Si 2·62%, Mn 0·45%, S 0·124%, P 0·4%

2·5 per cent and phosphorus of 0·4 per cent. With 1·1 per cent phosphorus and 2·35 per cent silicon, a slight permanent reduction in length may occur under non-oxidising conditions. Typical dilatometric curves for 0·4 per cent and 1·1 per cent phosphorus are shown in *Figures 69* and *70*. Unstressed growth of low silicon grey cast iron (total carbon 3·29 per cent; silicon, 1·27 per cent; manganese, 0·28 per cent; sulphur 0·12 per cent; phosphorus, 0·72 per cent) has been given as 6×10^{-8} in/in.h at 450°C, and 313×10^{-6} in/in.h at 538°C[11]. This growth would tend to be higher with higher silicon contents[12].

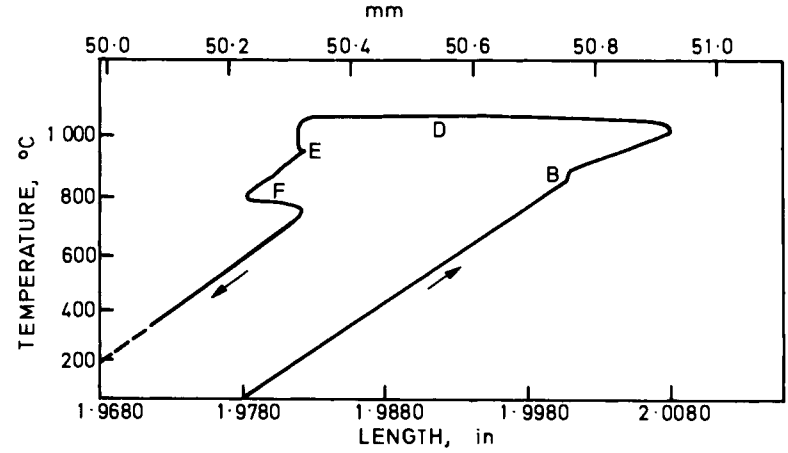


Figure 70. TC 2·81%, Si 2·77%, Mn 0·43%, S 0·124%, P 1·1%

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ELECTRICAL RESISTIVITY[1, 2]

In general, the electrical resistivity of cast iron ranges between 25 and 80 microhms/cm^3 (0·25 and 0·80 $\mu\Omega$.m). The various elements which may be present in cast iron all have their individual effects. *Table 51* gives some idea of the actual resistivity of various materials.

Table 51

Material	*Electrical resistivity* microhms/cm^3	($\mu\Omega$.m)	*Remarks*
Electrolytic copper	1·59	(0·0159)	Best metallic conductor
Pure iron	9·8	(0·098)	
Nickel	11·8	(0·118)	
Chromium	13·8	(0·138)	
Graphite[3] (basal plane)	30·0	(0·300)	
Graphite[3] (*C*-axis)	1×10^4	(100·0)	
Silicon	$6{\cdot}0 \times 10^4$	(600·0)	
Sulphur	4×10^{15}	(4×10^{13})	Insulator
Cast iron	20–80	(0·20–0·80)	

Carbon and silicon

These elements have by far the greatest influence on the electrical resistivity of cast iron. The lower the carbon and the silicon, the lower the resistivity of the iron itself. When carbon is present as graphite, coarse flake graphite structures give the highest resistivity, with a lowering of the resistivity as the flakes become finer. This is illustrated by some tests (*Table 52*) carried out with an iron of the following composition:

	%
Total carbon	3·80
Silicon	2·50
Manganese	0·58
Sulphur	0·014
Phosphorus	0·039

Table 52

Section size in (mm)	*Graphite size*	*Electrical resistivity* microhms/cm³ (μΩ.m)
3·0 (76)	Coarse flake	103·6 (1·036)
2·1 (53)	Medium-coarse flake	94·6 (0·946)
1·2 (30)	Medium flake	91·4 (0·914)
0·875 (22)	Fine-medium flake	84·4 (0·844)
0·6 (15)	Fine flake	77·4 (0·774)

Ferritic nodular graphite structures give the lowest resistivity of all (*Table 53*).

Table 53

TC %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Cu* %	*Mg* %	*Ce* %	*Electrical resistivity,* microhms/cm³	(μΩ.m)
3·60	2·33	0·84	0·011	0·042	—	2·20	—	0·045	59·0	(0·590)
3·77	2·96	0·67	0·010	0·037	—	1·70	—	0·066	65·4	(0·654)
4·08	2·85	0·64	0·012	0·034	—	—	—	0·064	66·5	(0·665)
3·25	1·46	0·24	0·012	0·021	0·67	—	0·039	—	40·4	(0·404)
3·04	2·40	0·25	0·014	0·022	0·72	—	0·055	—	50·8	(0·508)

Silicon increases the resistivity (1) by increasing the amount of graphite in the structure, and (2) because silico-ferrite has a higher electrical resistivity than pure ferrite. This is illustrated in *Table 54* for grey iron.

Table 54

TC %	*Si* %	*Mn* %	*S* %	*P* %	*Electrical resistivity,* microhms/cm³	(μΩ.m)
3·01	1·43	0·53	—	—	53·7	(0·537)
3·05	1·65	0·53	0·032	0·029	61·5	(0·615)
3·07	1·89	0·64	0·035	0·032	68·3	(0·683)
3·05	1·92	0·72	0·030	0·045	71·3	
3·03	2·29	0·62	0·036	0·041	75·9	(0·759)
3·08	2·41	0·52	0·011	0·037	76·5	(0·765)
3·05	3·08	0·73	0·021	0·072	90·4	(0·904)
2·71	5·6	1·03	0·072	(0·1)	116·6	(1·166)
3·16	5·35	1·53	0·045	0·105	142·5	(1·425)

Manganese and sulphur

These elements do not affect the resistivity of ordinary grey cast irons in which the sulphur is balanced by the manganese.

Phosphorus

Up to 0·2 per cent phosphorus has an almost negligible effect upon electrical resistivity, but above 0·2 per cent the phosphorus has an effect on the graphite structure which produces an increase in resistivity. The effect, however, as shown in *Table 55* is not great.

Table 55

TC %	*Si* %	*Mn* %	*S* %	*P* %	*Electrical resistivity*, microhms/cm³	($\mu\Omega$.m)
3·05	2·36	0·77	0·021	0·11	76·0	(0·760)
3·05	2·37	0·77	0·015	1·13	85·7	(0·857)

Alloying elements

Of the alloying elements, nickel and aluminium only have any appreciable effects, both raising the resistivity with increasing additions. This is shown in *Tables 56* and *57*.

Table 56

TC %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Electrical resistivity* microhms/cm³	($\mu\Omega$.m)
3·52	1·63	0·64	0·041	0·050	Nil	86·4	(0·864)
—	—	—	—	—	0·47	79·4	(0·794)
—	—	—	—	—	0·86	99·1	(0·991)
—	—	—	—	—	1·22	103·3	(1·033)
—	—	—	—	—	1·45	105·8	(1·058)
—	—	—	—	—	1·81	103·2	(1·032)

Table 57

TC %	*Si* %	*Mn* %	*S* %	*P* %	*Al* %	*Electrical restivity* microhms/cm³	($\mu\Omega$.m)
2·95	0·90	0·64	0·072	0·041	0·06	44·1	(0·441)
2·95	1·01	—	—	—	0·22	48·7	(0·487)
2·99	0·94	—	—	—	0·66	55·6	(0·556)
3·00	0·91	—	—	—	1·18	63·5	(0·635)
3·00	0·91	—	—	—	2·30	89·0	(0·890)
2·99	0·93	—	—	—	3·34	93·7	(0·937)

Copper, molybdenum, chromium and vanadium have very little effect beyond a slight tendency to lower resistivity. The principal effect of these elements is mainly the result of any change they may produce in the structure of the iron and for this reason rate of cooling and pouring temperature must also be taken into consideration.

Austenitic irons

An austenitic iron, for example, of composition:

	%
Total carbon	2·35
Silicon	2·00
Manganese	0·72
Sulphur	0·068
Phosphorus	0·051
Nickel	16·5
Chromium	1·62
Copper	7·48

had a resistivity of 129·2 microhms/cm^3 (1·292 $\mu\Omega$.m). All cast irons of similar high alloy content give resistivity values of the same order or higher.

Nicrosilal, a typical composition for which is:

	%
Total carbon	1·75–2·0
Silicon	4·5–5·0
Manganese	0·6–1·0
Sulphur	below 0·1
Phosphorus	below 0·1
Nickel	18·0–20·0
Chromium	2·0–4·0

has an electrical resistivity of the order of 120–160 microhms/cm^3 (1·2–1·6 $\mu\Omega$.m).

Martensitic irons

These give resistivity values between those of normal grey irons and the austenitic irons.

Pearlitic irons

When the matrix is substantially pearlitic, the lowest resistivity is obtained when it is in the coarse lamellar form; conversely, refining the pearlite appears to raise the electrical resistivity of the iron.

Malleable cast iron

Both whiteheart and blackheart have specific resistivities[2,4] ranging from 24 to 37 microhms/cm^3 (0·24 to 0·37 $\mu\Omega$.m). This relatively low resistivity is partly the result of the nodular and aggregate form of the graphite, and is partly due to the low silicon content of the matrix.

Resistivity figures for specific irons are also given in the next section on magnetic properties.

Temperature coefficient of electrical resistance is given by:

$$R_t = R_0(1 + xt)$$

where R_t = resistance at t°C, R_0 = resistance at 0°C, and x = temperature coefficient.

There is little published information on this subject, but work by Maréchal and Listray[5] gives values between 0·00088 and 0·0011/°C for irons of the following approximate composition:

	%
Total carbon	3·0–3·4
Silicon	1·9–2·1
Manganese	0·5
Sulphur	0·1
Phosphorus	0·7–0·4

An average value of 0·001 may, therefore, be accepted as reliable.

Effect of stress on electrical resistivity

When a cast iron is stressed in tension the spaces occupied by the graphite increases in volume. On unloading only partial recovery occurs and the spaces occupied by the graphite have a greater volume than before stressing.

With grey irons with tensile strengths of 13–21 ton/in^2 (201–324 N/mm^2) a stress up to 80 per cent of the tensile strength increases the resistivity by between 1·6 and 0·9 per cent. The weakest iron exhibits the greatest increase in resistivity[6].

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MAGNETIC PROPERTIES

Most of the work on the magnetic properties of cast iron prior to 1950 (e.g. Partridge[1]) did not take full notice of the effect of microstructure and the following data are taken mainly from the work of Hillman[2] which includes a fuller range of irons and photomicrographs than are here shown.

Although the magnetic properties of cast iron do not approach those of the permanent magnet steels on the one hand, or the silicon steels on the other, cast iron is widely employed where its magnetic properties may be of significance, and the following points may be noted:

(1) Iron castings can be made in intricate shapes where steel forgings would be out of the question, and steel castings would offer some difficulty.
(2) The magnetic properties of cast iron are less affected by tensile and compressive stresses than are those of steel.
(3) Where cast iron has been permanently magnetised, the temperature coefficient of loss in magnetism is much smaller than that of magnet steels, and shock losses are also very small.
(4) Non-magnetic cast irons play an important part as structural members in electrical machinery and reduce magnetic leakages because of their low permeability.

Definitions of the various magnetic terms employed

A magnetic field is a region in space in which a magnetic state exists, and the lines of direction of this field form closed paths which are termed lines of force or lines of flux.

The magnetomotive force is the total magnetising influence acting on any part of a magnetic circuit. The unit of measurement is the gilbert.

The magnetic intensity of the *magnetising force* at any point in a magnetic circuit is defined as the magnetomotive force per unit length of path of the magnetic flux. It is denoted by H and the unit used is the oersted, which is equivalent to one gilbert per centimetre.

The magnetic induction or *flux density* is the flux per unit area normal to the direction of flux. It is denoted by B and the unit used is the gauss, which is equivalent to one maxwell per square centimetre.

The magnetic permeability is that property of an isotropic medium which determines, under specified conditions, the magnitude of the relation between magnetic induction and magnetising force in the medium. It is denoted by μ and expressed as $\frac{B}{H}=\mu$.

The remanent magnetism or *remanence* is the magnetic induction which remains in a magnetic circuit after the removal of an applied magnetic force. The extent to which an opposing magnetic force must be applied to remove this remanence is a measure of the strength with which the material retains its magnetism and is known as the coercive force.

When a sample of iron is subjected to an alternating magnetisation, the relation between B and H is different for increasing and decreasing values of the magnetic intensity. This phenomenon is due to irreversible processes which result in energy dissipation, producing heat. This energy loss is termed the hysteresis loss, and is expressed either in ergs per cubic centimetre per cycle, or in watts loss per pound of metal at a standard frequency of alternation and magnetic induction.

Influence of structure

Ferrite gives a low hysteresis loss and high permeability, whereas pearlite gives a high hysteresis loss and low permeability. The presence of free cemetite causes low values of magnetic induction, permeability and remanent magnetism, together with increased coercive force and hysteresis loss.

Irons with an austenitic structure are non-magnetic, and the presence of martensite renders the iron feebly magnetic.

As a general rule, the magnetic properties of cast iron largely depend upon the nature of its structure, and the influence of alloy additions on these properties of cast iron in either the as-cast or annealed state is an indirect one, depending on how these alloys affect the structure of the finished material.

Influence of grain size

As the grain size in the ferritic state increases, the hysteresis loss decreases.

Influence of temperature

A rise in temperature causes a considerable increase in permeability up to the critical point. When this point is reached the induction falls practically to zero and the iron becomes non-magnetic.

Curie point

The Curie point is the temperature at which a substance ceases to be ferromagnetic and becomes paramagnetic. For pure iron, this occurs at the α–γ transformation temperature of 770°C and falls slightly with increasing silicon content, so that with 5 per cent silicon the Curie point is approximately 730°C[3]. For silicon-free cementite (Fe_3C) the Curie point[4] is in the region of 205–220°C. Cementite in cast iron is substantially free from silicon up to 4·8 per cent silicon in the overall composition. The change is progressive with rise in temperature.

Demagnetisation of cast iron can be effected by heating to 550°C, holding for two hours and slow cooling to 100°C.

Effect of carbon

A study of the magnetic properties of iron–carbon alloys has been made by Crangle and Sucksmith[4].

Effect of graphite

According to Hillman, graphite influences magnetic properties by its amount and form. High permeability is favoured by the nodular or temper carbon form compared with the flake form. The presence of coarse graphite, as opposed to fine graphite, appears to lower the remanent magnetism in both the ferritic and pearlitic states. There is not enough evidence to comment on the effect of graphite size on the coercive force in either the ferritic or pearlitic states. As the graphite becomes finer there is a tendency for the hysteresis loss to increase, and this appears to be more marked in the case of pearlitic matrices. The effect of graphite size on the maximum permeability in the pearlitic state is indeterminate, although the presence of fine graphite tends to lower the value in the completely ferritic condition.

Effect of graphite shape

According to Hillman the graphite shape has little effect on the coercive force or remanent magnetism in the pearlitic or ferritic states. Nodular graphite appears to show slightly increased hysteresis loss in the pearlitic state, but in the ferritic state the hysteresis loss is considerably reduced. It is possible to obtain high permeability and low hysteresis loss when the graphite is in an aggregate or nodular form in a completely ferritic matrix. The combined carbon must be as low as possible (pearlite absent); free cementite must be absent and the phosphorus content should also be below 0·1 per cent.

Effect of matrix

According to Hillman, the highest magnetic induction and permeability are found in irons which have been annealed to the ferritic condition. In the case of a low phosphorus flake graphite iron, for example, the maximum permeability may be increased by a factor of 4 on annealing. Both the remanent magnetism and coercive force tend to show lower values in the annealed state.

The advantages of the more continuous matrix produced when the graphite is in nodular form may be lost if alloying elements are present. This is particularly noticeable in the case of nodular cast irons, which may contain up to 2 per cent of nickel. Nodular irons made without the use of nickel have a lower hysteresis loss, and a higher maximum permeability than any grey flake graphite iron.

Decrop, in an excellent study[5], has noted that the transition from coarse flake type A to fine graphite type D, accompanied by total disappearance of

pearlite, noticeably increases the values for induction, remanent induction and coercive force. Higher induction values are obtained with the nodular cast irons with ferritic or pearlitic matrices, the difference in matrix appearing to be less important than the compactness of the graphite; this is particularly noticeable by comparison with ferritic flake graphite irons. The highest initial or maximum permeability is obtained with ferritic matrices and nodular or temper carbon forms, and the graphite form is the predominating influence.

The presence of high percentages of alloying elements in sufficient amounts to produce an austenitic matrix renders the iron non-magnetic and gives a maximum permeability of the order of 2 or less.

Effect of alloying elements

MANGANESE

Manganese in general is prejudicial to high magnetic induction and permeability, both in as-cast and annealed states. In the annealed state it increases the coercive force and hysteresis loss, but decreases the remanent magnetism.

SULPHUR

Sulphur, when present as manganese sulphide, has little effect on the magnetic properties.

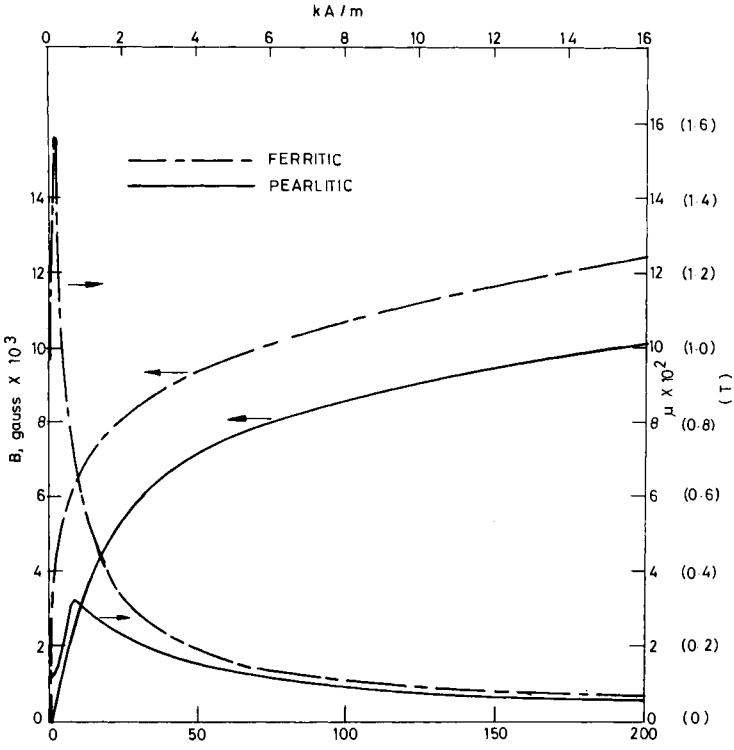


Figure 71

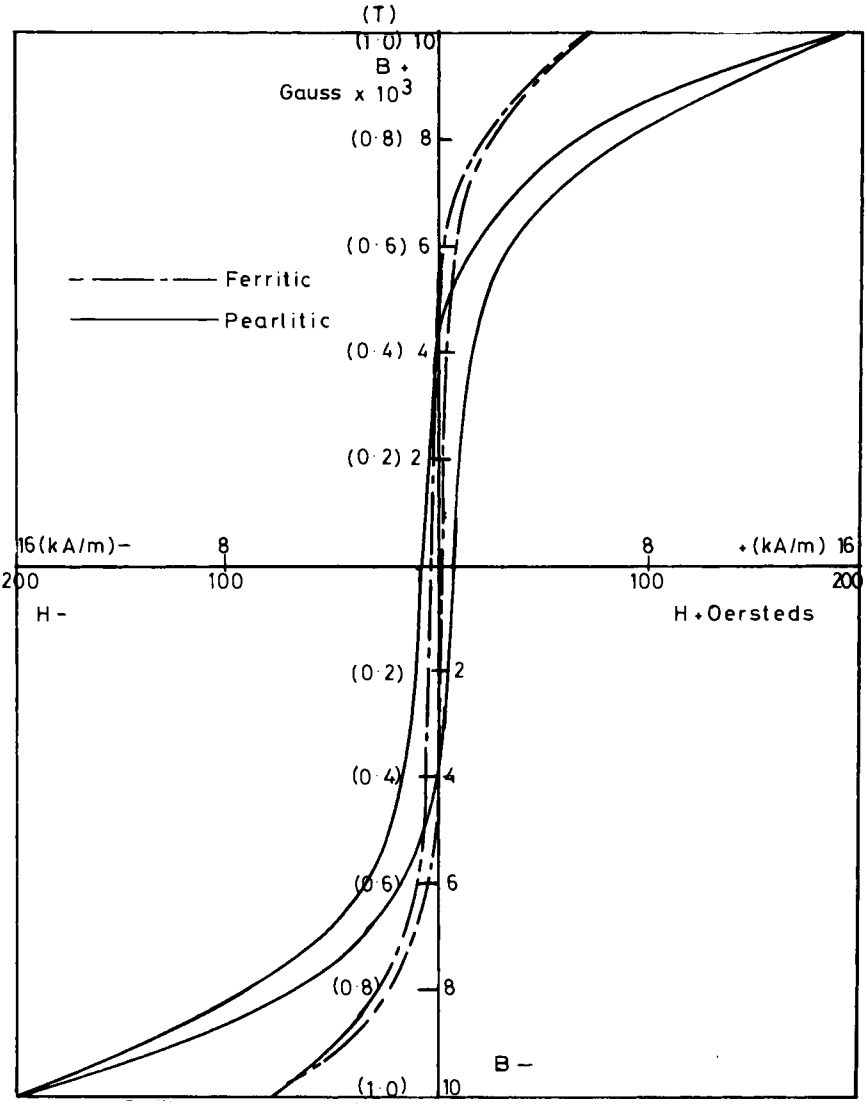


Figure 72

SILICON

Silicon appears to have very little effect on the magnetic properties of cast iron in the pearlitic or ferritic states. Decrop[5] shows that Silal with 6 per cent silicon and fully ferritic, differs little in magnetic properties from normal grey iron.

A study of the magnetic properties of the iron–silicon alloys has been made by Guggenheimer and others[3].

Table 58 MAGNETIC AND ELECTRICAL PROPERTIES: GREY IRON, LOW PHOSPHORUS (SPECIMEN NO. 1) SEE FIGURES 71 AND 72

Graphite: Coarse flake, ASTM size 3A

Chemical analysis, as-cast:

	%
Total carbon	3·12
Silicon	2·22
Manganese	0·67
Sulphur	0·067
Phosphorus	0·13
Nickel	<0·03
Chromium	0·04

	Pearlitic (as-cast)	*Ferritic (annealed)*	
Combined carbon (%)	0·70	0·06	
Remanent magnetism (gauss)	4130 (0·413 T)	4350	(0·435 T)
Coercive force (oersted)	7 (557 A/m)	2·5	(199 A/m)
Hysteresis loss	26 960 (2696J/m^3) = 8·41 (3·8 W/kg)	6960 2·24	erg/cm^3/cycle (696 J/m^3) for $B = 10\,000$ gauss (1T) watt/lb (1·0 W/kg) for $B = 10\,000$ gauss (1T) at an alternating frequency of 50 c/s (50 Hz)
Value of H (oersted) for $B = 10\,000$ (1T)	200 (15·9 kA/m)	74	(5·9 kA/m)
Maximum permeability μ max.	315 (396 μH/m)	1560	(1960 μH/m)
Value of H (oersted) for μ max.	8 (637 A/m)	2·5	(199 A/m)
Specific resistance (ohm/cm^3)	73×10^{-6} (0·73 $\mu\Omega$.m)	71×10^{-6}	(0·71 $\mu\Omega$.m)

Table 59 MAGNETIC AND ELECTRICAL PROPERTIES: GREY IRON, HIGH PHOSPHORUS (SPECIMEN NO. 2)

Graphite: Medium flake, ASTM size 5A/B
Chemical analysis, as-cast:

	%
Total carbon	3·30
Silicon	2·04
Manganese	0·52
Sulphur	0·065
Phosphorus	1·03
Nickel	0·34
Chromium	0·25

	Pearlitic (as-cast)		*Ferritic (annealed)*	
Combined carbon (%)	0·77		0·11	
Remanent magnetism (gauss)	4920	(0·492 T)	4390	(0·439 T)
Coercive force (oersted)	9	(716 A/m)	3·5	(279 A/m)
Hysteresis loss	27 290	(2729 J/m^3)	11 930	erg/cm^3/cycle (1193 J/m^3) for $B = 10\,000$ gauss (1 T)
	= 8·35	(3·8 W/kg)	3·88	watt/lb (1·8 W/kg) for $B = 10\,000$ gauss (1 T) at an alternating frequency of 50 c/s (50 Hz)
Value of H (oersted) for $B = 10\,000$ (1 T)	109	(8·7 kA/m)	101	(8·0 kA/m)
Maximum permeability μ max.	281	(353 μH/m)	760	(955 μH/m)
Value of H (oersted) for μ max.	13	(1035 A/m)	4	(318 A/m)
Specific resistance (ohm/cm^3)	77×10^{-6}	(0·77 $\mu\Omega$.m)	75×10^{-6}	(0·75 $\mu\Omega$.m)

PHOSPHORUS

The presence of phosphide eutectic in a ferritic matrix with a phosphorus content of the order of 1 per cent increases the hysteresis loss by a factor of 2 and reduces the maximum permeability to about half of the values obtained in the absence of phosphorus. The effect of phosphorus on other factors is indeterminate and in the presence of a pearlitic matrix its effect is completely masked.

NICKEL

Nickel decreases the permeability and increases the hysteresis loss. If present in sufficient amounts to produce austenite, the iron becomes non-magnetic.

CHROMIUM

Chromium reduces magnetic induction, permeability and remanent magnetism, and increases the coercive force and hysteresis loss.

COPPER

Copper reduces permeability and increases hysteresis loss.

Some typical figures are given, together with the compositions and microstructures of the castings concerned, in *Tables 58* to *65*.

Table 60 MAGNETIC AND ELECTRICAL PROPERTIES: GREY IRON, VERY COARSE, LOW PHOSPHORUS. INGOT MOULD IRON (SPECIMEN NO. 3)

Graphite: Coarse flake, ASTM size 1A
Chemical analysis:

	%
Total carbon	4·17
Silicon	2·05
Manganese	1·20
Sulphur	0·08
Phosphorus	0·07
Nickel	< 0·03
Chromium	trace

	Pearlitic	
Combined carbon (%)	0·86	
Remanent magnetism (gauss)	2790	(0·279 T)
Coercive force (oersted)	8	(637 A/m)
Hysteresis loss	17 420	erg/cm^3/cycle (1742 J/m^3) for $B = 10\,000$ gauss (IT)
	= 5·64	watt/lb (2·6 W/kg) for $B =$ 10 000 gauss (IT) at an alternating frequency of 50 c/s (50 Hz)
Value of H (oersted) for $B = 10\,000$ (IT)	216	(17·2 kA/m)
Maximum permeability μ max.	245	(309 μH/m)
Value of H (oersted) for μ max.	10	(796 A/m)
Specific resistance (ohm/cm^3)	101×10^{-6}	(1·01 $\mu\Omega$.m)

Table 61 MAGNETIC AND ELECTRICAL PROPERTIES: GREY IRON, LOW SILICON HOT MOULD IRONS, (A) PEARLITIC, AND (B) FERRITIC (ANNEALED) (SPECIMENS NOS. 4A AND 4B)

Graphite: (a) medium flake, ASTM size 3A; (b) medium coarse flake, ASTM size 3/4A
Chemical analysis:

	(*a*) *Pearlitic* %	(*b*) *Ferritic* %
Total carbon	3·34	3·34 (2·52)
Silicon	0·83	0·91
Manganese	0·33	0·20
Sulphur	0·038	0·021
Phosphorus	0·048	0·023
Nickel	0·05	<0·03
Chromium	<0·01	<0·02

	(*a*) *Pearlitic*		(*b*) *Ferritic*	
Combined carbon (%)	0·88		n.d.	
Remanent magnetism (gauss)	5215	(0·5215 T)	6185	(0·6185 T)
Coercive force (oersted)	8	(637 A/m)	2·5	(199 A/m)
Hysteresis loss	26 450	(2645 J/m^3)	9380	erg/cm^3/cycle (938 J/m^3) for $B = 10\,000$ (1T) gauss
	=8·07	(3·7 W/kg)	2·96	watt/lb (1·3 W/kg) for $B = 10\,000$ gauss (1T) at an alternating frequency of 50 c/s (50 Hz)
Value of H (oersted) for $B = 10\,000$ (IT)	78	(6·2 kA/m)	55	(4·4 kA/m)
Maximum permeability μ max.	318	(400 μH/m)	1355	(1703 μH/m)
Value of H (oersted) for μ max.	14	(1114 A/m)	3·0	(239 A/m)
Specific resistance (ohm/cm^3)	42×10^{-6}	(0·42 $\mu\Omega$.m)	37×10^{-6}	(0·37 $\mu\Omega$.m)

Table 62 MAGNETIC AND ELECTRICAL PROPERTIES: WHITEHEART MALLEABLE IRONS (SPECIMEN NOS. 5A AND 5B)

Graphite: Aggregate

Chemical analysis:

	(*a*) *Pearlitic centre ferritic rim* %	(*b*) *Mainly ferritic (long anneal)* %
Total carbon	1·01	0·46
Silicon	0·66	0·66
Manganese	0·25	0·25
Sulphur	0·166	0·166
Phosphorus	0·055	0·055
Nickel	0·03	0·03
Chromium	<0·01	<0·01

	(*a*) *Pearlitic*		(*b*) *Ferritic*	
Combined carbon (%)	0·62		0·28	
Remanent magnetism (gauss)	7460	(0·746 T)	7370	(0·737 T)
Coercive force (oersted)	4·5	(358 A/m)	2·5	(199 A/m)
Hysteresis loss	14 910	(1491 J/m^3)	8420	erg/cm^3/cycle (842 J/m^3) for $B = 10\,000$ (1T) gauss
	=4·52	(2·1 W/kg)	2·58	watt/lb (1·2 W/kg) for $B = 10\,000$ gauss (1T) at an alternating frequency of 50 c/s (50 Hz)
Value of H (oersted) for $B = 10\,000$ (IT)	20	(1592 A/m)	13	(1035 A/m)
Maximum permeability μ max.	730	(917 μH/m)	1455	(1828 μH/m)
Value of H (oersted) for μ max.	7	(557 A/m)	3·5	(279 A/m)
Specific resistance (ohm/cm^3)	26×10^{-6}	(0·26 $\mu\Omega$.m)	24×10^{-6}	(0·24 $\mu\Omega$.m)

Table 63 MAGNETIC AND ELECTRICAL PROPERTIES: BLACKHEART MALLEABLE IRON. COMMERCIALLY PRODUCED MATERIAL: DECARBURISED AREA $\frac{1}{32}-\frac{3}{32}$ IN RING SAMPLE TESTED (SPECIMEN NO. 6)

Graphite: Aggregate, mainly ferritic matrix

Chemical analysis:

	%
Total carbon	1·26
Silicon	0·83
Manganese	0·26
Sulphur	0·05
Phosphorus	0·066
Nickel	0·03
Chromium	0·02

Combined carbon (%)	0·25	
Remanent magnetism (gauss)	6190	(0·619 T)
Coercive force (oersted)	1·5	(119 A/m)
Hysteresis loss	4940	erg/cm^3/cycle (494 J/m^3) for $B = 10\,000$ gauss (IT) = 1·52 watt/lb (0·7 W/kg) for $B = 10\,000$ gauss (IT) at an alternating frequency of 50 c/s (50 Hz)
Value of H (oersted) for $B = 10\,000$ (IT)	9·6	(764 A/m)
Maximum permeability μ max.	2120	(2664 μH/m)
Value of H (oersted) for μ max.	2·2	(175 A/m)
Specific resistance (ohm/cm^3)	$29{\cdot}5 \times 10^{-6}$	(0·295 $\mu\Omega$.m)

Table 64 MAGNETIC AND ELECTRICAL PROPERTIES: NODULAR GRAPHITE CAST IRONS, (A) PEARLITIC, (B) FERRITIC, (C) FERRITIC, NICKEL-FREE (SPECIMEN NOS. 7A, 7B AND 7C)

Nickel is normally employed as one of the alloying elements in the production of this iron, and has a marked influence on the magnetic properties. Figures for a nickel-free nodular iron are included for comparison.

Chemical analysis:

	(*a*) *Pearlitic*	(*b*) *Ferritic*	(*c*) *Ferritic nickel-free*
	%	%	%
Total carbon	2·90	2·84	3·64
Silicon	2·61	2·61	1·41
Manganese	0·72	0·72	0·30
Sulphur	0·013	0·013	0·019
Phosphorus	0·044	0·044	0·017
Nickel	2·18	2·23	<0·03
Magnesium	0·13	0·13	0·028

	(*a*) *Pearlitic*		(*b*) *Ferritic*		(*c*) *Ferritic nickel-free*	
Combined carbon (%)	0·72		0·19		0·06	
Remanent magnetism (gauss)	5400	(0·54 T)	4640	(0·464 T)	6065	(0·6065 T)
Coercive force (oersted)	11·5	(915 A/m)	2·5	(199 A/m)	1·5	(119 A/m)
Hysteresis loss	30 390	(3039 J/m^3)	7000	(700 J/m^3)	4495	erg/cm^3/cycle (450 J/m^3) for B = 10 000 gauss (IT)
	= 10·11	(4·6 W/kg)	2·33	(1·1 W/kg)	1·43	(0·6 W/kg) watt/lb for B = 10 000 gauss (IT) at an alternating frequency of 50 c/s (50 Hz)
Value of H (oersted) for B = 10 000 (IT)	63	(5013 A/m)	27	(2149 A/m)	8	(637 A/m)
Maximum permeability μ max.	290	(364 μH/m)	1150	(1445 μH/m)	2060	(2589 μH/m)
Value of H (oersted) for μ max.	15	(1194 A/m)	4	(318 A/m)	3	(239 A/m)
Specific resistance (ohm/cm^3)	71×10^{-6}	(0·71 $\mu\Omega$.m)	58×10^{-6}	(0·58 $\mu\Omega$.m)	39×10^{-6}	(0·39 $\mu\Omega$.m)

Table 65 MAGNETIC PROPERTIES: NON-MAGNETIC AUSTENITIC IRONS (SPECIMEN NOS. 8A AND 8B)

Graphite: (a) medium flake, ASTM size 5/6E (Ni-Resist); and (b) 4/5A and E (Nomag)
Chemical analysis:

	(*a*) *Ni-Resist corrosion resisting type*	(*b*) *Nomag type*
	%	%
Total carbon	2·72	2·5–3·0
Silicon	2·41	1·5–3·0
Manganese	1·09	4·0–7·0
Sulphur	0·02	0·1
Phosphorus	0·04	0·1–1·0
Nickel	15·5	11–12
Chromium	1·81	—
Copper	5·17	—
Aluminium	—	1·0

These materials are almost completely non-magnetic.

	(*a*) *Ni-Resist*		(*b*) *Nomag*	
Combined carbon (%)	0·51			
Maximum magnetic permeability	2·1	(2·64 μH/m)	1·03–1·1	(1·29–1·38 μH/m)
Specific resistance (ohm/cm^3)	133×10^{-6}	(1·33 $\mu\Omega$.m)	150×10^{-6}	(1·50 $\mu\Omega$.m)
Resistance temperature coefficient per deg C (0–100°C)			0·00045	

A considerable range of irons and structures has also been examined by Decrop[5].

Note: The names Ni-Resist and Nomag are trade names, registered respectively by International Nickel Ltd. and Ferranti Ltd.

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NON-MAGNETIC CAST IRONS*

Non-magnetic irons are almost invariably austenitic and differ mainly in the alloying elements employed to stabilise the austenite. They may be divided into the following groups:

(1) Nickel alloys with good mechanical, high temperature properties and corrosion resistance.
(2) Nickel, copper, manganese alloys with or without aluminium. These alloys are cheaper than group 1, but may have poorer corrosion and high temperature resistance.
(3) Manganese alloys with or without aluminium or copper. The cheapest of all the groups, but requires high carbon to avoid promotion of free carbides. This lowers mechanical properties and corrosion resistance is also lowered.

The addition of manganese and copper may be used to reduce the amount of nickel required and hence the cost of the finished alloy.

The electrical resistivity of these alloys is high and the mechanical properties are largely dependent upon the carbon content and the form of the graphite. Increasing the carbon improves machinability, but lowers mechanical properties. Altering the form of the graphite to the nodular form markedly improves the mechanical properties. The presence of chromium tends to produce carbides and to increase the value of μ.

It is important that the alloy content (nickel, manganese, copper) is adequate, particularly with thick sections, otherwise martensite may form as a result of slow cooling, and the non-magnetic properties will be impaired.

Table 66 illustrates the properties of several varieties of non-magnetic iron, and Tyskzo[1] suggests that the manganese, copper and aluminium irons with a high carbon content and manganese up to 12 per cent can be used for thin walled castings, and with lower carbon (2·8 to 3·2 per cent) and manganese (12 to 16 per cent) for heavy section castings.

Tyskzo makes the interesting observation that when the aluminium content exceeds about 20 per cent, alpha iron loses its ferro-magnetic properties,

* *Note:* The non-magnetic irons normally employed in British practice are both austenitic and details are given in *Table 65* in the preceding section

Table 66

Type	*C* %	*Si* %	*Mn* %	*Ni* %	*Cu* %	*Cr* %	*Al* %	*Tensile* ton/in^2	(N/mm^2)
Nickel/nickel/copper or nickel/copper/chromium (austenitic)	2·8–3·4	1·5–2·5	0·5–1·5	14–22	Up to 8·5	Up to 2·0	—	9·5–19·0	(147–293)
Nickel/manganese or nickel/copper/manganese (austenitic)	3·2–3·6	1·5–2·5	5–10	5–10	Up to 5·0	—	—	9–14	(139–216)
Manganese/copper (austenitic)	3·3–3·8	1·5–2·5	7–12	—	Up to 4·0	—	—	9·0–12·5	(139–193)
Manganese/copper/aluminium (austenitic)	2·8–3·7	3·5–4·5	7–16	—	1·5–3·5	—	2·0–4·5	9·0–16·5	(139–255)
Manganese/aluminium/silicon (austenitic)	3·0–3·6	4·0–5·0	7–15	—	—	—	1·0–5·0	6·5–15·0	(100–232)
Aluminium/iron (ferritic)	1·0–2·0	Up to 5·0	Up to 1·5	—	Up to 3·0	—	22–34	6·5–10·0	(100–154)

Type	*Hardness* HB	*Permeability*	(μH/m)	*Resistivity* ohm/cm^2/cm	($\mu\Omega$.m)
Nickel/nickel/copper or nickel/copper/chromium (austenitic)	130–190	1·1–3·0	(1·4–3·8)	130–170 × 10^{-6}	(1·3–1·7)
Nickel/manganese or nickel/copper/manganese (austenitic)	140–190	1·1–3·0	(1·4–3·8)	130–150 × 10^{-6}	(1·3–1·5)
Manganese/copper (austenitic)	230–280	1·1–3·0	(1·4–3·8)	130–160 × 10^{-6}	(1·3–1·6)
Manganese/copper/aluminium (austenitic)	150–220	1·1–3·0	(1·4–3·8)	150–200 × 10^{-6}	(1·5–2·0)
Manganese/aluminium/silicon (austenitic)	170–300	1·1–3·0	(1·4–3·8)	160–200 × 10^{-6}	(1·6–2·0)
Aluminium/iron (ferritic)	170–200	1·005–1·008	(1·263–1·267)	150–200 × 10^{-6}	(1·5–2·0)

and *Table 66* shows an aluminium iron alloy which is ferritic and which has very satisfactory non-magnetic properties although it is weak mechanically. He claims that it can be produced successfully in the foundry, but implies the use of special techniques.

REFERENCE

1. TYSKZO, Z., *Fonderie*, No. 140, 411–420 (Sept. 1957)

3

General properties of commercial cast irons

STRENGTH, COMPOSITION AND STRUCTURE OF GREY CAST IRONS

Figures 73–75 give a simple first approximation of the relationship between strength, composition and structure[1, 2].

The chief difficulty in using these diagrams is that of interpreting the results in terms of foundry practice, since it is quite impossible to make any allowance for section size except on the basis of standard sections such as round bars or flat plates, and it is necessary to interpret the results likely to be obtained in actual castings in terms of standard conditions. Schneidewind and McElwee[3] use the term 'equivalent diameter' to express the cooling rate of a

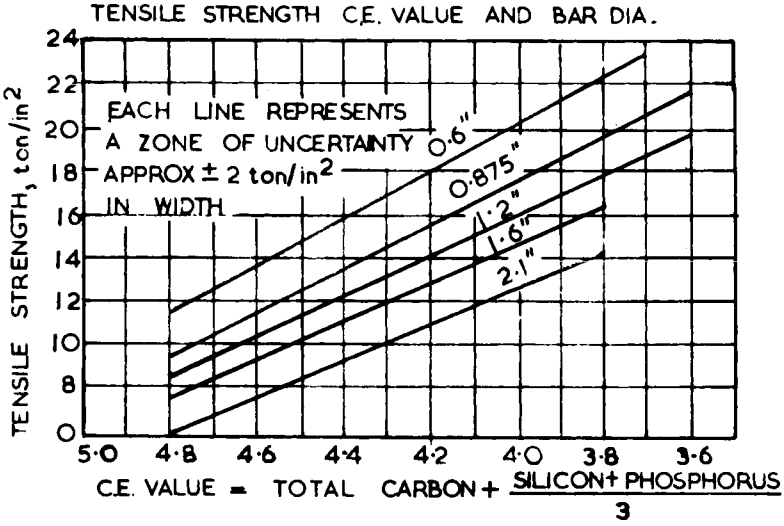


Figure 73. Tensile strength, CE value and bar diameter relationships

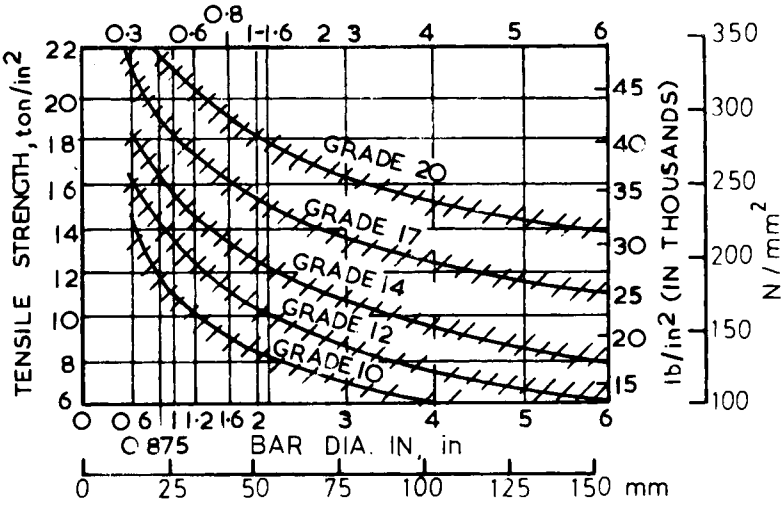


Figure 74. Variation of strength of cast iron with bar diameter

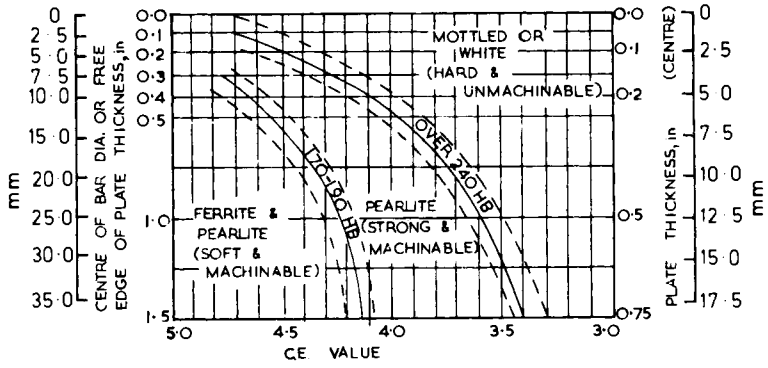


Figure 75. Diagram relating section size, CE value and structure

given casting section, and this is a convenient method of expressing cooling rates for castings.

Cooling rate

The rate of cooling of a casting increases as the ratio of volume to area, V/A, decreases. This ratio must be calculated in consistent units, either metric or English. The effect of this ratio on cooling rate can be very simply demonstrated by *Figure 76*, which illustrates the effect of position in a casting upon the cooling surface of a unit of fixed size. In *Figure 76*, a unit $2 \times \frac{1}{8} \times \frac{1}{4}$ in ($\frac{1}{16}$ in^3) is considered, both in the centre and at the extreme edge of a plate $12 \times 12 \times$

$\frac{1}{4}$in. In the centre the unit has two effective cooling surfaces totally $\frac{1}{2}$in^2 in area, giving a volume to area ratio of 0·0625/0·5 = 0·125. At the edge there are three effective cooling surfaces totalling 1·0 in^2 in area and giving a volume

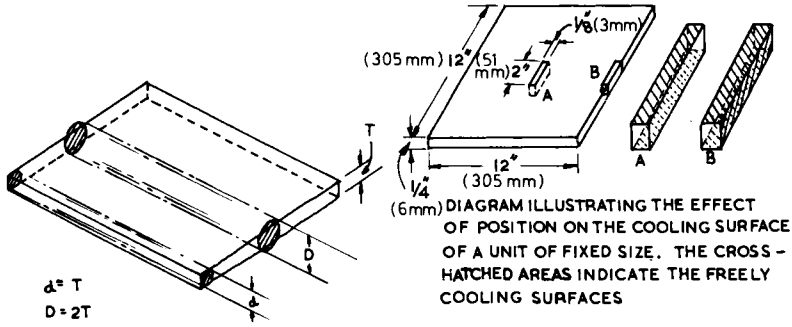


Figure 76

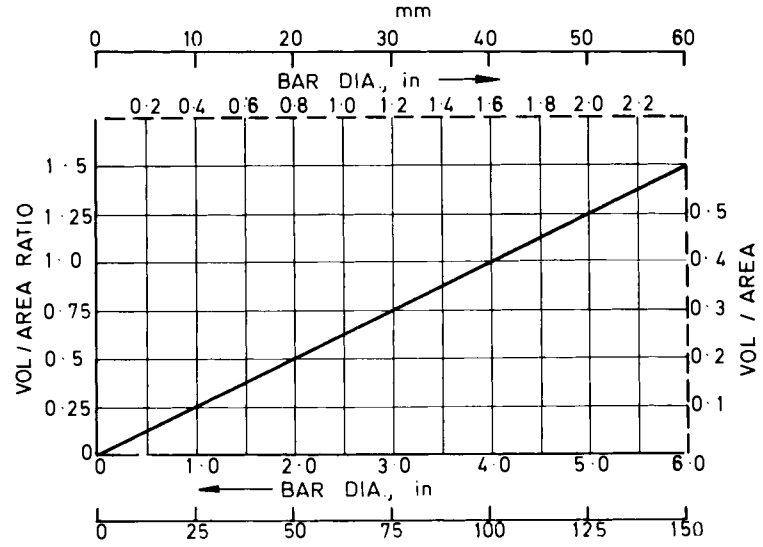


Figure 77. Volume/area ratios for the centres of long bars of various diameters

to area ratio of 0·0625/1·0 = 0·0625. Thus, from these figures the rate of cooling over a given temperature range should be twice as great at the edge as in the centre. If it is now desired to find the sizes of test bar which theoretically will cool at the same rate as the centre or edge of the plate under discussion (*Figure 76*) we must choose bars having similar values of V/A. These values

can be obtained from *Figure 77*, which gives values of volume to area ratio for the centres of long bars of various diameters. We can see that for $V/A = 0{\cdot}125$, and $V/A = 0{\cdot}0625$, we require test bars of $\frac{1}{2}$in (12·7 mm) diameter and $\frac{1}{4}$in (6·4 mm) diameter, respectively.

Equivalent diameter

The equivalent diameter of a given section may, therefore, be defined as the diameter of a long cylindrical bar whose centre section will cool at the same rate as the casting section in question when cast under standard conditions. It is important to refer the equivalent diameter to the centre portion of a long bar, since the free end of a cylindrical bar is subject to a correction similar to that applied at the edge of a flat plate. For ordinary test bar dimensions, the equivalent diameter of a free end of a cylindrical bar can be taken as three-quarters the full diameter of the bar.

All this has been discussed previously by many writers and assumes that at the solidification point:

(1) The total amount of heat to be dispersed is proportional to the weight or volume of the casting.
(2) The rate of heat transfer through the moulding material is constant.
(3) All the heat is dissipated through the surface area of the casting.
(4) The cooling rate over the casting section in question is constant.

The cooling rate of castings in sand moulds

The cooling rate of a casting in a sand mould is, therefore, mainly dependent upon its section and, more precisely, upon its volume/area ratio. If cooling conditions were uniform, the cooling rate from a given temperature would be expected to be directly proportional to the weight or volume of the casting, and the heat is dissipated through the surface area. This, however, does not apply in sand moulds since the loss of heat from the casting raises the temperature of the surrounding sand and the conditions for free heat flow no longer apply[4]. Assumption (4) is also an approximation, since the skin of a casting cools more rapidly than the centre (*Figure 38*).

Chvorinov[5] has stated that the freezing time of a casting varies as the square of the volume/area ratio. In an attempt to assess the validity of this statement, data have been drawn from the literature where precise cooling curves have been published and from some BCIRA data, and the cooling interval has been taken from the eutectic to the eutectoid arrest (about 1150–720°C). It has not been possible to separate green sand, dry sand and oil sand moulds, as the information is not always clear, but their effects will be most marked in thin section castings. The sources are quite varied, but the curve shows general agreement with Chvorinov's rule from very small section castings up to 20 in diameter bars. These data are summarised in *Figure 78*.

The influence of type of sand could be more marked over the short period

of time involved in freezing than over the more prolonged period to the eutectoid arrest.

The tests were taken from the centres of the bars or castings and it is important to remember that the volume/area ratio alters near the edge of a plate

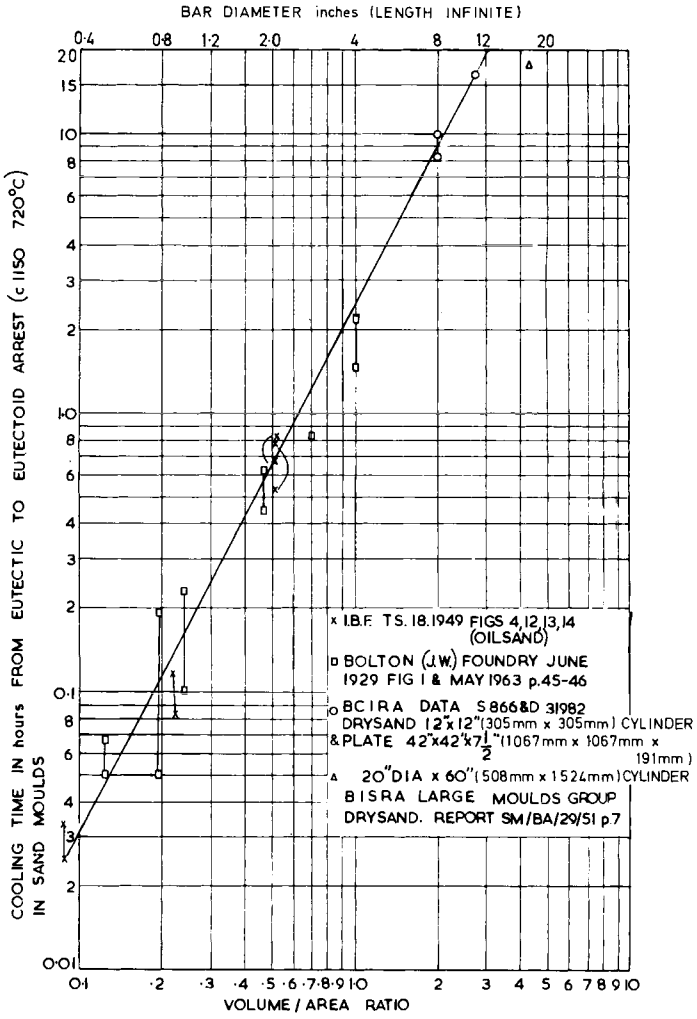


Figure 78. The time of cooling (h) from eutectic to eutectoid arrest of grey cast irons in relation to the volume/area ratio of cylindrical bars and flat plate in sand

or the end of a bar, so that the cooling rate at these regions is more rapid than in the centre. This effect is well known and is reflected in the mechanical properties of a test bar which are lower in the centre than at the end or the edge (*Figure 38*).

The effect may be quite significant in test bars of different as-cast lengths and it should be noted that whereas until the 1956 revision of BS 1452 the

as-cast length of the standard test bar was not less than 21 in (533 mm) for the 1·2 in (30 mm) bar to accommodate the transverse test, in the 1956 and subsequent revisions, the tensile test became mandatory and the as-cast length for the tensile test was reduced to 9 in (229 mm) overall length.

The theoretical volume/area ratio for a 1·2 in (30 mm) bar of infinite length is 0·3, for the centre of a 21 in (533 mm) long bar is 0·292, and for a 9 in (229 mm) bar is 0·281. It would, therefore, be expected that the tensile strength of otherwise similar material would be slightly higher on specimens taken from the centre of a 9 in (229 mm) long as-cast bar than from 21 in (533 mm) long bars. The effect is unlikely to raise the probable tensile strength by more than 0·25 ton/in^2 (4 N/mm^2).

Cast iron in the British Standards is evaluated on the basis of test bars of varying diameters, and BS 1452/1961 gives a useful general relation of test bar diameter to the main cross sectional thickness of the casting, as shown in *Table 67*.

Table 67

Main cross-sectional thickness of casting		*Diameter as-cast*		*Approximate minimum overall length*	
in	(mm)	in	(mm)	in	(mm)
Not exceeding $\frac{3}{8}$	(9·5)	0·6	(15)	5	(127)
Over $\frac{3}{8}$, not exceeding $\frac{3}{4}$	(9·5–19)	0.875	(22)	7	(178)
Over $\frac{3}{4}$, not exceeding $1\frac{1}{8}$	(19–29)	1·2	(30)	9	(229)
Over $1\frac{1}{8}$, not exceeding $1\frac{5}{8}$	(29–41)	1·6	(41)	11	(279)
Over $1\frac{5}{8}$	(41)	2·1	(53)	13	(330)

For more detailed, but still approximate, evaluation it is convenient to assess the cooling rate of a section in terms of these equivalent test bar diameters, and from the foregoing the following simple approximations in relative cooling rates given in *Table 68*, may be employed for section thicknesses up to about 1·6 in (41 mm).

Table 68

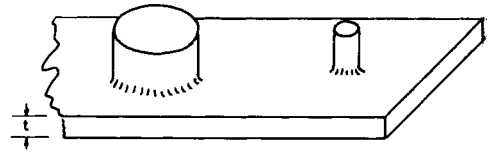
	Centre	*Free edge*	*Free corner*
PLATE			
Actual thickness, in (mm)	1 (25)	1 (25)	1 (25)
Equivalent diameter, in (mm)	2 (51)	1 (25)	0·6–0·8 (15–20)
	Centre	*Free end*	
BAR OR BOSSES			
Actual diameter, in (mm)	1 (25)	1 (25)	
Equivalent diameter, in (mm)	1 (25)	0·75 (19)	

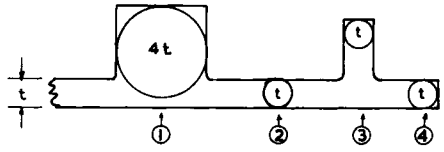
Thus, if the above figures are taken as inches (mm), the centre of a flat plate 1 in (25 mm) in thickness will cool at about the same rate as the centre section of a cylindrical bar 2 in (51 mm) in diameter, while the free corner of the 1 in (25 mm) plate will cool at about the same rate as the centre section of a cylindrical bar 0·6–0·8 in (15–20 mm) in diameter.

The effect of adjacent portions of the casting is not readily predictable on a useful quantitative basis, but the use of inscribed circles, as described below, may help in assessing relative cooling rates of castings at the drawing board stage, and the equivalent diameters can then be used with *Table 70* and *Figures 73–75* to determine chilling tendencies and likely mechanical properties.

THE USE OF INSCRIBED CIRCLES[14]

Where an inscribed circle is tangential to three major sides of a section, as in (1), (3) and (4) in *Figure 79*, the cooling rate can be assumed to be equal to, or faster than, that of a bar of the same diameter, and the possibility of hard spots, depending on composition and cooling rate, must be considered.





INSCRIBED CIRCLES IN SECTIONAL DRAWING

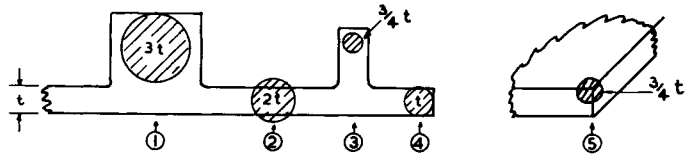


Figure 79. Cooling rate relative to cylindrical bars of 'equivalent diameter'

Thus, in the above example in which a plate of thickness t has two bosses, one of diameter t and another of diameter $4t$:

(a) The cooling rate in the centre of the plate (point 2) would be equivalent to that of the centre of a test bar $2t$ in diameter (*see also Figure 79*).
(b) The cooling rate at the edge of the plate (point 4) would be equivalent to that of a test bar t in diameter (*see also Figure 79*).
(c) The cooling rate at the tip of the smaller boss (point 3) and the corner (point 5) would be equivalent to that of a test bar $\frac{3}{4}$ in (19 mm) in diameter.
(d) The cooling rate at the tip of the larger boss (point 1) would be equivalent to that of a test bar $3t$ in diameter.

Thus, the risk of obtaining hard spots would be greatest at points 3, 4 and 5, and it would be necessary to specify an iron that would not chill at these points.

EFFECTS OF CORES

Internal cores, if smaller in cross section than the metal wall thickness, can be assumed without much error to behave as solid metal. If the minimum cross-sectional dimension of the core is more than twelve times the wall thickness of the castings the special effect of the core can usually be ignored. As a very rough guide *Table 69* may be used to assess the effect of cores.

Table 69 MINIMUM CORE DIMENSIONS RELATED TO WALL THICKNESS

Minimum core dimension between metal walls	*Less than metal wall thickness*	*Equal to or twice wall thickness*	*2 or 4 times wall thickness*	*4 to 12 times wall thickness*	*12 or more times wall thickness*
Correction to 'equivalent diameter' of casting thickness	Core taken as solid metal	× 2	× 1·5	× 1·2–1·5	× 1

Composition and strength

THE CARBON EQUIVALENT VALUE

The carbon equivalent value:

$$CE = \text{Total carbon}\% + \frac{Si\% + P\%}{3}$$

is an index which combines the effect of silicon and phosphorus upon the eutectic of iron and carbon, and has been briefly discussed in the section on constitution and microstructure. The carbon equivalent value therefore deter-

mines how close a given composition of iron is to the theoretical iron–carbon eutectic (CE = 4·3) and therefore how much free graphite is likely to be present for a given cooling rate.

Since the mechanical properties of a cast iron are largely controlled by the graphite content, the carbon equivalent value can provide a useful index of the probable strength in a given section.

Composition and structure

The carbon, silicon and phosphorus contents also greatly affect the tendency to chill and the carbon equivalent value is a useful guide to chilling tendency or hardness of a given section.

The permanent effect of silicon as opposed to its transient inoculating effect in reducing chill formation is to increase the difference between the graphite–austenite and carbide–austenite eutectics. The graphite–austenite eutectic has a wider possible freezing range with the higher silicon contents[6] than indicated in *Figures 1* and *7*.

The carbon equivalent value, however, is only an approximate guide to these properties, since graphite form and matrix structures can be very markedly affected by pouring and melting temperatures, alloying or residual elements, inoculation and so on. Nevertheless, it is a valuable index for commercially produced materials, provided its limitations are understood. It is suitable for irons with total carbon between 3·2 and 3·5 per cent. Total carbon above 3·5 per cent gives lower tensile results and heavily inoculated irons give higher tensile results (*Figure 73*).

Increase in phosphorus content of a grey iron from 0·2 to 1·0%P increases the tensile strength in 1·2 in (30 mm) bars by 1–2 ton/in^2 (15–31 N/mm^2) at a constant carbon equivalent value of about 4·1. The increase in tensile strength is accompanied by lower total strain to failure, i.e. increased brittleness as the phosphorus content increases. Ingot mould irons (CE 3·8–4·6) cast into 1·2 in (30 mm) bars, are lower in tensile strength by 2–4 ton/in^2 (31–62 N/mm^2) than predicted by the CE tensile curve, mainly because of the low phosphorus content (< 0·08 per cent).[7]

The curves of strength versus carbon equivalent values were originally based on test bars industrially produced before 1949 in which phosphorus values in Great Britain were predominantly above 0·5 per cent. The curves were stated to give the best correlation with total carbon between 3·2 and 3·5 per cent and carbons above 3·5 per cent would give lower results[1]. Much of present-day production is below 0·5 per cent phosphorus and work at BCIRA has shown that for irons in which silicon plus phosphorus is below 2·5 per cent, the total carbon value gives a relationship with tensile strength in 1·2 in (30 mm) bars which includes total carbons above 3·5 per cent such as ingot mould irons and down to 2·9 per cent total carbon.

This relationship is shown in *Figure 80* with 90 per cent confidence limits of ±2·5 ton/in^2 (39 N/mm^2) for 1·2 in (30 mm) bars. Bars of 0·875 in (22 mm) diameter appear to fall predominantly within the upper half of the band, and 1·6 in (41 mm) bars in the lower half.

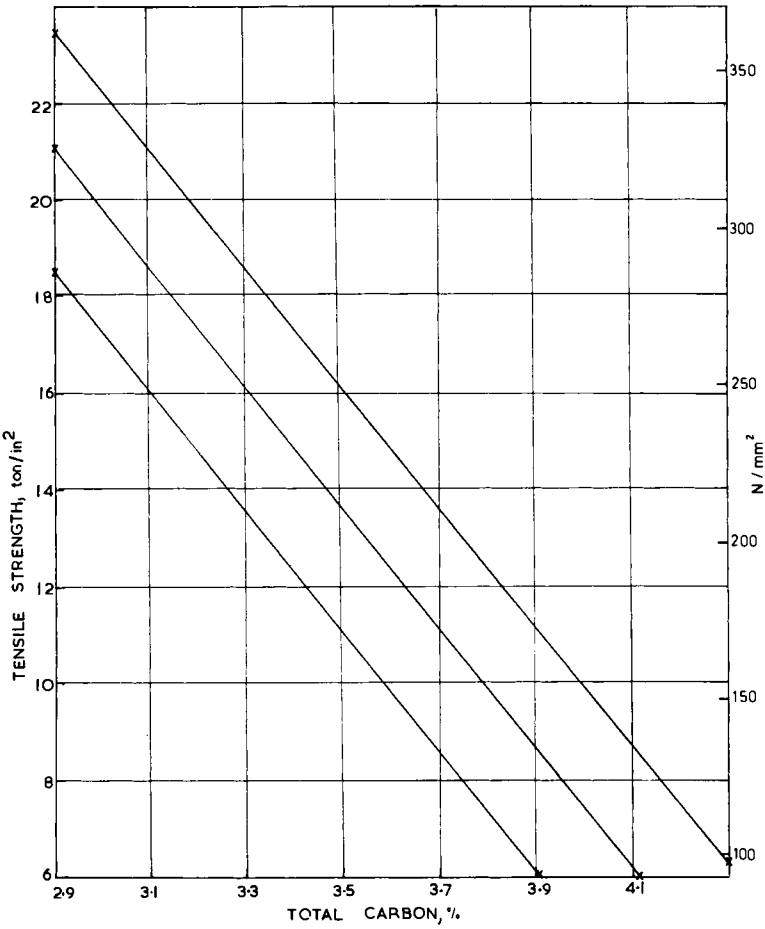


Figure 80. Tensile strength (ton/in²) versus total carbon (Si + P less than 2·5%) for 1·2 in bars of uninoculated and unalloyed grey iron (Donald, BCIRA Blantyre)

Inoculated irons appear to fall in the upper half of the band or slightly above it.

The section on tensile testing explains the relationship between the properties of a test bar and those found in the casting itself.

Although appreciable variations occur in properties of apparently similar irons, it is possible to give a reasonable assessment of the likely properties of a given metal, when poured into castings or test bars, from the analysis.

The structure and mechanical properties of a cast iron of given composition depend primarily upon the cooling rate after pouring into the mould; it has already been shown how to express the cooling rate of a given section in terms of a cast test bar section. Since the equivalent diameter of a given casting section is the diameter of a long cylindrical bar whose centre section will cool at the same time as the casting section, it may therefore be expected to have a similar structure and similar mechanical properties.

The standard bases of comparison are:

(1) Tensile strength.
(2) Analysis, as defined by the carbon equivalent value. The carbon equivalent value may be less precise in assessment of tensile strength than the total carbon value, but has the advantage that some prediction of structure is possible in different section thicknesses (*Figure 75*).
(3) Structure, defined by machinability:
  (a) Soft machinable (ferritic).
  (b) Strong machinable (pearlitic).
  (c) Hard or unmachinable (containing free carbide).

These can be determined by reference to *Figures 73* and *75*.

The relationship between analysis as expressed by the carbon equivalent value (CE) and the tensile strength likely to be found in various diameter test bars, is shown in *Figure 73*. Thus, an iron with a CE value of 4·0 would be expected to give a tensile of 17–18 ton/in^2 (263–278 N/mm^2) in an 0·875 in (22 mm) diameter as-cast bar. The slopes of these curves in *Figure 73* have been modified slightly from the earlier version, following work at BCIRA which included a more satisfactory range of phosphorus contents in commercially produced test bars from 0·03–1·4 per cent phosphorus.

Since, for a given iron, the strength falls with section thickness, and a cylindrical test bar cools more rapidly than a flat plate, *Figure 74* shows the falling off of strength with section expressed both as bar diameter and plate or section thickness. Zuithoff[8] shows curves in good agreement with *Figure 74*. Thus an iron giving 14 ton/in^2 (216 N/mm^2) in a 1·2 in (30 mm) diameter as-cast test bar (BS 1452), Grade 14, would give 11–13 ton/in^2 (170–201 N/mm^2) in a 2·1 in (53 mm) diameter bar or a 1·0–1·6 in (25–41 mm) section, or 18 ton/in^2 (278 N/mm^2) in a 0·6 in (15 mm) diameter bar, or a 0·3 in (8 mm) section. Thinner sections than this, however, would be liable to chill, or to be unmachinable in any grade higher than Grade 14.

Figure 75, based on Sipp[9], indicates the chilling tendency of various irons. The diagram is divided into three sections: on the left are the soft irons, in the centre are the strong machinable irons, and on the right are the hard unmachinable irons. Thus, an iron with a carbon equivalent value of 4·1 (Grade 14) would be pearlitic and machinable in bar diameters down to 0·4–0·5 in (10–13 mm) diameter, or in sections 0·2–0·3 in (5–7 mm) thick, but it might become hard and unmachinable in thinner sections. Sections of $\frac{1}{4}$ in

Table 70

Tensile strength, ton/in^2 (N/mm^2) on 1·2 in (30 mm) diameter bar (corresponding to BS 1452, Grade No.:)	10 (154)	12 (185)	14 (216)	17 (263)	20 (309)
Approximate carbon equivalent value	4·5–5·55	4·3–4·35	4·1–4·2	3·85–3·9	3·55–3·6
Minimum thickness for chill-free edge, in (mm)	0·1–0·2 (2·5–5)	0·2–0·3 (5–8)	0·3–0·4 (8–10)	0·5–0·65 (13–17)	0·8–1·1 (20–28)

(6 mm) or less in thickness require analyses giving carbon equivalent values of 4·4 or more if they are to remain soft and grey.

Table 70 summarises the data on tensile strength, carbon equivalent value and chilling tendency for unalloyed irons contained in *Figures 73–75*.

EFFECT OF ALLOYS

Alloys are added to cast iron to confer special properties, but their use should only be specified where they are known to provide beneficial effects. The elements usually employed for alloying purposes are nickel, copper, chromium, molybdenum, vanadium and titanium.

Although it is usually only the mechanical or chemical properties that require modification, all these elements have an effect on casting properties and structure which must be considered. For example, while silicon is the element normally used to control chill, the alloying elements mentioned also affect chill, and where additions of more than about 0·2 per cent are made, this effect must be taken into account. The suggested allowances for the normal alloying elements in relation to silicon are given in *Table 71*.

Table 71

Graphitisers					*Carbide stabilisers*			
C %	*P* %	*Ni* %	*Cu* %	*Al* %	*Mn* %	*Mo* %	*Cr* %	*V* %
+3·0	+1·0	+0·3	+0·3	+0·5	−0·25	−0·35	−1·2	−1 to −3·0

Thus, 1 per cent of carbon is roughly equivalent in graphitising power to 3 per cent of silicon, 1 per cent of nickel is roughly equivalent to 0·3 per cent of silicon, 1 per cent of chromium roughly cancels the effect of 1·2 per cent of silicon and if the appropriate correction is made to the CE values, the effect on tensile and chill can be approximately determined from *Figures 73–75* at least for chromium.

The effect of alloy additions on tensile strength can only be given approximately, but the figures in *Table 72* are based on the observations of Crosby[10], and Schneidewind and McElwee[3]. Chromium above 0·5 per cent is likely to form massive carbides, and molybdenum above 0·6 per cent in the presence of nickel may give acicular structures.

Table 72 EFFECT OF ALLOY ADDITIONS ON TENSILE STRENGTH

	Mo	*Cr*	*V*	*Ni*	*Cu*	*Mn*
Up to approximately this % of:	1·0	0·5–1·0	0·35	3·0	1·5	0·5–1·0
Approximate % increase in tensile strength for each 1% alloy	40–45	30–20	30–80	5–10	8–10	0–10

CARBON SATURATION DIAGRAM

Fairly simple methods have been given in this section for estimating the tensile strength and structure in varying sections obtainable for given analyses of cast iron.

A comprehensive diagram has, however, been developed by Collaud[11] based on extensive work which covers a wider range of properties than can

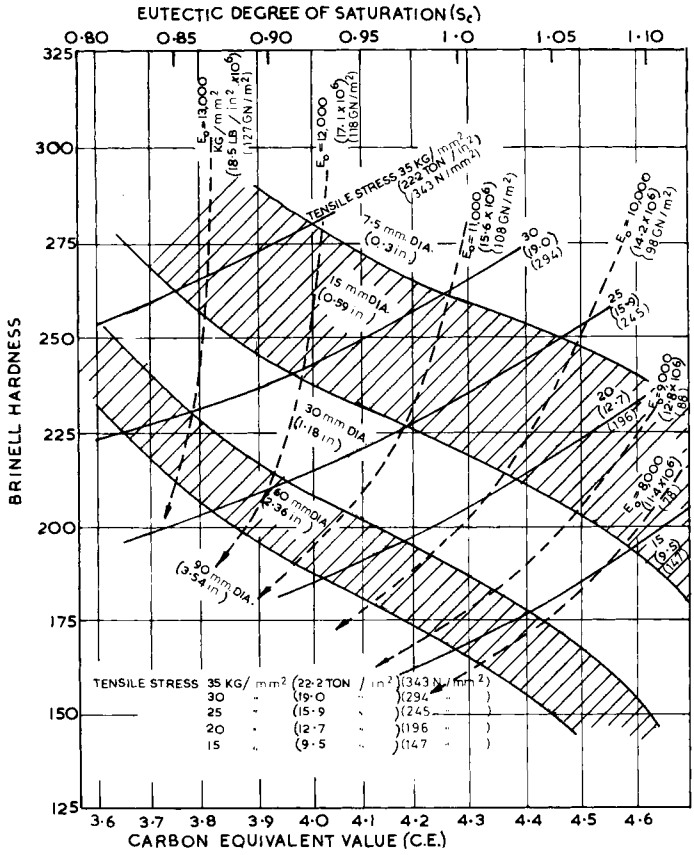


Figure 81. Relationship between the tensile strength, Brinell hardness, modulus of elasticity at zero stress, composition and bar diameter (Collaud).

$$\textit{Carbon equivalent value } (CE) = \textit{total carbon } \% + \frac{\textit{silicon } \% + \textit{phosphorus } \%}{3}$$

$$\textit{Carbon saturation value } (S_c) = \frac{\textit{total carbon}}{4{\cdot}23 - \frac{\textit{silicon } \%}{3{\cdot}2} - 0{\cdot}275\ \textit{phosphorus } \%}$$

Table 73 SUMMARISED DATA ON GREY CAST IRONS[13]

The figures summarised below for given grades of iron can be regarded as typical, but individual irons of similar tensile strength may differ appreciably from these figures for other properties owing to variations in production methods, analysis, graphite size, form and content, matrix structure etc., and the figures must be regarded as a guide and not as absolute.

Properties	*Grade BS 1452:1961*						
	10	*12*	*14*	*17*	*20*	*23*	*26*
Min. tensile strength (1·2 in bar), ton/in^2	10	12	14	17	20	23	26
(30 mm bar) N/mm^2	154	185	216	263	309	355	402
Proof stress (0·01%), ton/in^2	2·8	3·3	3·9	4·7	5·5	6·3	7·2
Proof stress (0·01%), N/mm^2	43	51	60	73	85	97	111
Proof stress (0·1%), ton/in^2	6·5	7·8	9·1	11·1	13·0	15·0	16·9
Proof stress (0·1%), N/mm^2	100	120	141	171	201	232	261
Total strain at failure <0·4% P, %	0·75	0·7	0·65	0·58	0·50	0·50	0·50
Elastic strain at failure, %	0·15	0·17	0·18	0·20	0·22	0·25	0·28
Plastic strain at failure, %	0·60	0·53	0·47	0·38	0·28	0·25	0·22
Total strain at failure 0·4–1·0% P, %	0·60	0·5	0·40	—	—	—	—
Elastic strain at failure, %	0·15	0·17	0·18	—	—	—	—
Plastic strain at failure, %	0·45	0·33	0·22	—	—	—	—
Notched tensile 45° V radius, 0·01 in, ton/in^2	8	9·6	11·2	13·6	16	18·4	20·8
0·25 mm, N/mm^2	124	148	173	210	247	284	321
Compressive strength, ton/in^2	40	44·8	49·6	56·7	63·8	70·9	78
Compressive strength, N/mm^2	618	692	766	876	985	1095	1205
Proof stress (0·01%), ton/in^2	5·6	6·6	7·8	9·1	11·0	12·6	14·4
Proof stress (0·01%), N/mm^2	86	102	120	141	170	195	222
Proof stress (0·1%), ton/in^2	13·0	15·6	18·2	22·2	26·0	30·0	33·8
Proof stress (0·1%), N/mm^2	201	241	281	343	402	463	522
Shear strength, ton/in^2	11·5	13·8	16·1	19·6	23·0	26·4	29·9
Shear strength, N/mm^2	178	213	249	303	355	408	462
Torsional strength, ton/in^2	11·5	13·8	16·1	19·6	23·0	26·4	29·9
Torsional strength, N/mm^2	178	213	249	303	355	408	462
Modulus of rigidity, lb/in^2	6×10^6	$6{\cdot}5 \times 10^6$	$6{\cdot}9 \times 10^6$	$7{\cdot}5 \times 10^6$	$8{\cdot}2 \times 10^6$	$8{\cdot}2 \times 10^6$	$8{\cdot}4 \times 10^6$
Modulus of rigidity, GN/m^2	41	45	48	52	57	57	58

Modulus of elasticity,							
tension, lb/in^2	15×10^6	$16{\cdot}2 \times 10^6$	$17{\cdot}4 \times 10^6$	$18{\cdot}8 \times 10^6$	$20{\cdot}0 \times 10^6$	$20{\cdot}5 \times 10^6$	$21{\cdot}0 \times 10^6$
tension, GN/m^2	103	112	120	130	138	141	145
compression, lb/in^2	15×10^6	$16{\cdot}2 \times 10^6$	$17{\cdot}4 \times 10^6$	$18{\cdot}8 \times 10^6$	$20{\cdot}0 \times 10^6$	$20{\cdot}5 \times 10^6$	$21{\cdot}0 \times 10^6$
compression, GN/m^2	103	112	120	130	138	141	145
Poisson's ratio	0·26	0·26	0·26	0·26	0·26	0·26	0·26
Fatigue limit (Wöhler):							
Unnotched, ton/in^2	±4·5	±5·4	±6·3	±7·7	±9·0	±9·8	±9·9
Unnotched, N/mm^2	69	83	97	119	139	151	153
V-notched, ton/in^2	±4·5	±5·3	±6·0	±7·0	±8·0	±8·4	±8·3
V-notched, N/mm^2	69	82	93	108	124	130	128
Hardness HB <0·4% P increases about 3 points HB for each 0·1% P	134–164	149–182	162–199	182–223	201–247	226–276	250–305
Thermal expansion, −100°C to 20°C, 10^{-6}/°C	10·0	10	10	10	10	10	10
Thermal expansion, 20°C to 400°C, 10^{-6}/°C	12·5	12·5	12·5	12·5	12·5	12·5	12·5
Thermal conductivity, at 100°C, cal/cm^2/sec/°C/cm	0·115	0·113	0·110	0·106	0·103	0·099	0·095
Thermal conductivity, at 100°C, W/m.°C	48	47	46	44	43	41	40
Thermal conductivity, at 500°C, cal/cm^2/sec/°C/cm	0·115	0·113	0·110	0·106	0·103	0·099	0·095
Thermal conductivity, at 500°C, W/m.°C	48	47	46	44	43	41	40
Specific gravity	7·05	7·10	7·15	7·20	7·25	7·30	7·30
Magnetic and electrical properties, μ max.	250–300	250–300	250–300	250–300	250–300	250–300	250–300
Magnetic and electrical properties, μH/m	314–377	314–377	314–377	314–377	314–377	314–377	314–377
Resistivity, electrical, ohm/cm^3	80×10^{-6}	78×10^{-6}	76×10^{-6}	73×10^{-6}	70×10^{-6}	67×10^{-6}	64×10^{-6}
Resistivity, electrical, $\mu\Omega$.m	0·80	0·78	0·76	0·73	0·70	0·67	0·64
Design Stresses							
Direct tension, ton/in^2 max.	2·5	3·0	3·5	4·2	5·0	5·7	6·5
Direct tension, N/mm^2 max.	39	46	54	65	77	88	100
Direct compression, ton/in^2 max.	10·4	12·5	14·6	17·8	20·8	24·0	27·0
Direct compression, N/mm^2 max.	161	193	225	275	321	371	417
Fatigue, ton/in^2 max.	±1·5	±1·8	±2·1	±2·6	±3·0	±3·3	±3·3
Fatigue, N/mm^2 max.	23	28	32	40	46	51	51

be assessed by the methods given above. The diagram is based on the carbon saturation figure (S_c), where

$$S_c = \frac{\text{Total carbon}}{4{\cdot}23 - \frac{\text{Silicon}}{3{\cdot}2} - 0{\cdot}275\ \text{Phosphorus}}$$

Although a conversion from S_c to CE values cannot be absolute owing to the differing constants employed, a sufficiently close comparison can be made, and the diagram shown in *Figure 81* has been modified to include CE values as well as the original S_c values. There is a substantial measure of agreement between the two sets of curves.

A detailed review of the factors affecting the mechanical properties of flake graphite irons has been given by Gilbert [12, 13].

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LOW ALLOY CAST IRONS

The most usual alloy additions of nickel, copper, chromium, molybdenum and vanadium are made to confer strength, wear resistance, corrosion resistance or heat resistance. High strengths of up to 22 ton/in^2 (340 N/mm^2) with good machinability can be obtained in grey, unalloyed irons by the addition of inoculants such as ferrosilicon or calcium silicide to irons with carbon equivalent values that would normally solidify mottled or white. These methods usually require the use of high steel charges, and with satisfactory technical control are economical both in cost of raw materials and alloys. The use of alloys alone is in some cases preferable, but the use of inoculants is widely employed in conjunction with alloy additions to give further advantages.

All alloy additions, however, are expensive and should therefore only be specified where they have a definite known function to perform. In most industrially used castings the use of these alloying elements is unnecessary, provided the properties of the unalloyed irons are fully understood and correctly

employed, but alloys can be of great value where properly used and can confer properties not obtainable in unalloyed irons. The following details may serve as a guide to their proper employment.

NICKEL

Nickel[1, 2] is soluble in iron in all percentages, it lowers the γ to α transformation and lowers the carbon content of the iron–carbon eutectic by about 0·06 per cent for each 1 per cent of nickel[3]. This is not significant for low alloy

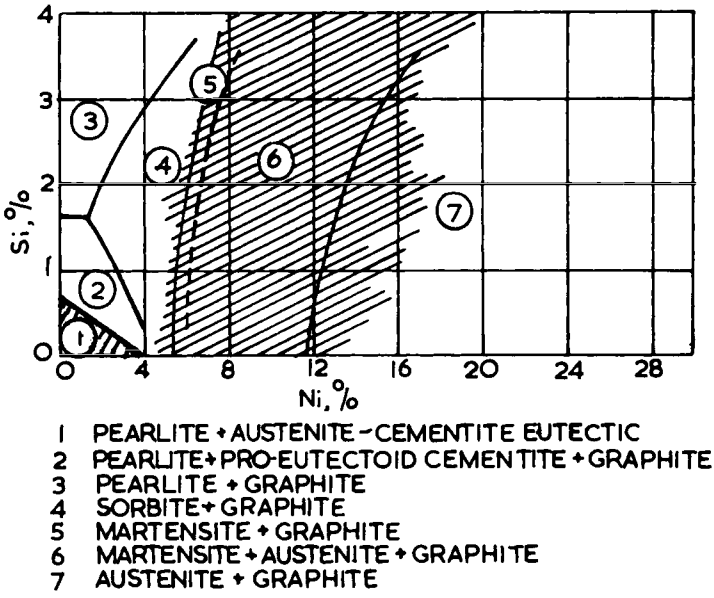


Figure 82. Constitution of nickel–silicon–iron alloys with 3·5% carbon. Based on as-cast sample 3 in × 1½ in × 1 in which is roughly equivalent in cooling rate to the centre of a 1 in diameter test bar or the centre of an extended plate about ½ in in thickness

contents, but becomes important with high nickel contents. Nickel also appears to lower the melting point of cast iron by about 5°C for 1 per cent nickel and slightly improves the fluidity[4]. The graphitising effect is approximately 0·3 that of silicon. It has little effect on the cell count in as-cast material.

The effect of nickel on microstructure of 3·5 per cent carbon irons with varying silicon contents[3] is indicated in *Figure 82.*

COPPER

Copper[2, 6, 7] has a limit of solubility in cast iron of 3·0 to 3·5 per cent and at these percentages can be detected microscopically as a separate constituent

containing 96 per cent copper and 4 per cent iron[6]. Each addition of 1 per cent of nickel increases the solubility of copper[8] by about 0·4 per cent. Copper lowers the carbon content of the iron–carbon eutectic by about 0·075 per cent for each 1 per cent copper. It increases the hardenability and lowers the γ to α change point by about 6–10°C for each 1 per cent copper[8] and refines the pearlite. It lowers the solidification temperature of cast iron by about 2°C for each 1 per cent addition[4]. The graphitising value of copper is about 0·2 to 0·35 than of silicon[6].

NICKEL AND COPPER

Nickel and Copper[1, 2, 6, 7] are both graphitisers and tend to keep castings grey and free from chill. They are normally used in amounts ranging from 0·3 to 1·5 per cent. Their effect on strength in normal sections is to increase tensile by 8–10 per cent for each 1 per cent addition (*see Table 72*), but their addition permits the use of lower total carbon and lower silicon than would normally be employed, without incurring chill or hard castings, and for this reason higher tensile strengths can be obtained. They increase basic hardness by about 10–20 points Brinell for each 1 per cent addition, by forming a solid solution harder than the alloy-free iron, by keeping the iron pearlitic, and also by refining the pearlite itself. They do not form free carbides, so that while their effect on wear resistance is not great apart from their effect in suppressing the formation of free ferrite, the risk of producing carbide-containing irons of low machinability is greatly reduced.

They are particularly valuable in reducing section sensitivity, i.e. in producing strong and dense iron in the centre of a heavy section without increasing the liability to chill in thin sections[1, 7].

An addition of 0·5 to 1 per cent copper can effect a considerable improvement in resistance to corrosion by dilute sulphuric acids, by hydrochloric acid, seawater, sulphur-bearing fuel oils and to atmospheric corrosion, particularly in the presence of sulphurous fumes[6].

Nickel and copper also markedly increase hardenability by increasing the depth of hardening for a given quench speed as a result of their effect on the transformation rate at the γ and α change point[7]. This applies to bulk hardening, but for surface hardening the effect may be less marked, since the depth and time of heat penetration is of particular importance.

Nickel and copper are of particular value when used in conjunction with chromium and molybdenum, where they offset the chilling tendency of the latter two elements and therefore enable the maximum strength to be obtained without unduly affecting machinability in varying sections (reference may also be made to the section on acicular irons).

Nickel and copper do not increase pearlite stability at temperatures above 450°C.

CHROMIUM, MOLYBDENUM AND VANADIUM

These elements[2] are carbide stabilisers. They increase the tendency of a casting to chill, and to form free carbides. They increase the strength of a grey casting quite markedly by 3 to 4 per cent for each 0·1 per cent addition (*see Table 72*). Chromium and molybdenum are used in amounts from 0·3 to 1·0 per cent, and occasionally up to 1·5 to 2·0 per cent. They also increase the hardness (a) by reducing the tendency to form free ferrite, (b) by refining the pearlite and (c) by the formation of free carbides.

Where an increase in tensile strength only is required, these elements should be used with basic phosphorus figures of less than 0·1–0·15 per cent, since they tend to form complex carbide phosphides which lower machinability, and by lowering the effective alloy concentration in the matrix, reduce the strengthening effect of the addition. At the same time, the low melting point of the phosphide complex increases the risk of segregation in heavy or complex sections.

Considerable data on nickel, copper and molybdenum and very full bibliographies have been published[6, 7, 9].

Chromium raises the carbon content of the iron–carbon eutectic by about 0·06 per cent for each 1 per cent addition of chromium, up to about 9 per cent chromium, and raises the melting point by about 1–1·5°C for each 1 per cent addition of chromium. It has a marked tendency to form free carbides, particularly in conjunction with phosphide eutectic[10, 11]. Chromium raises the α to γ transformation temperature and increases hardenability by decreasing the critical cooling rate[11]. In the absence of mottle it increases the tensile strength by about 3–4 per cent, and the hardness by about 5–7 points Brinell for each 0·1 per cent addition of chromium.

Chromium alone is of particular value up to 0·5 per cent in preventing the formation of free ferrite in critical sections where, under conditions of marginally lubricated sliding wear, it is important that free ferrite should be absent. It is sometimes used in low phosphorus automobile cylinder blocks where a fully pearlitic structure is desirable and the iron must be machinable. It also markedly increases the pearlite stability under heat, and in amounts up to about 1 per cent, depending upon the section, can produce strong, dense irons with a marked resistance to growth by pearlite breakdown at temperatures from 450 to 650°C. The machinability of these irons will decrease as the chromium increases or the carbide content increases, either as a result of section thickness or the presence of phosphorus.

Molybdenum[2] is a useful single element addition for increasing the tensile strength of low phosphorus irons (0·15 per cent phosphorus max.). Up to about 0·8 per cent it increases the tensile strength by 3·5–4·0 per cent for each 0·1 per cent addition, and in low phosphorus irons appears less likely than chromium to produce massive carbides which adversely affect machinability. Even with phosphorus contents of 0·35 per cent an appreciable increase in strength is obtained. For example, an iron with total carbon 3·5 per cent, silicon 2·0–2·08 per cent, manganese 0·6 per cent, phosphorus 0·25–0·36 per cent, to which 0·8 per cent molybdenum was added, gave an increase in tensile strength equivalent to 2–3 per cent and a rise in hardness of 2–4 points Brinell

for each 0·1 per cent molybdenum addition in test bars 0·6–2·1 in diameter. The impact strength was also increased. In very heavy sections, however, even with phosphorus contents of 0·15 per cent, there is a very marked tendency to segregate[10, 12] and phosphorus contents should therefore be as much below 0·1 per cent as possible. Above 0·6–0·8 per cent molybdenum, in conjunction with nickel or copper, acicular microstructures may be formed, associated with very high tensile strengths and high machinable hardness of about 320 HB, and these irons are dealt with in the section on acicular iron. Molybdenum greatly increases hardenability by reducing the γ to α transformation rates.

Molybdenum is also a very powerful pearlite stabiliser and confers great resistance to growth and creep within the region 350–650°C.

Vanadium[2] increases the carbon content of the iron–carbon eutectic by about 0·1 per cent for each 1 per cent addition. It is employed in amounts from 0·15 to 0·35 per cent and is particularly useful for wear resistance, notably in diesel liners in conjunction with 0·2–0·3 per cent phosphorus, where it reduces solubility in sulphuric acid and produces a dispersed phosphide/carbide complex with a hardness of 600–1000 HV, which appears valuable under the marginal conditions of lubrication that occur in heavy marine diesel liners with running sections of 2–4 in (51–102 mm) in thickness. From 0·37 per cent up to 0·61 per cent it raises the pearlite/austenite change point by about 9°C on heating, but does not affect the temperature on cooling[13]. It appears to be less prone than chromium to produce very massive carbides. It reduces electrical resistivity by about 1 microhm/cm^3 (0·01 $\mu\Omega$.m) for each 0·1 per cent addition of vanadium, and increases the tensile strength, particularly in low phosphorus irons (below 0·1 per cent phosphorus) by about 3 to 8 per cent for each 0·1 per cent addition, and it increases the hardness (by refining the pearlite) by about 8–10 points Brinell for each 0·1 per cent addition. Vanadium also increases pearlite stability and increases resistance to growth and creep within the temperature range 350–650°C. Its effect on pearlite stability at high temperatures is probably rather less than that of chromium or molybdenum (see also p. 300).

ALUMINIUM

In amounts of 1–5·5 per cent, aluminium confers marked oxidation resistance on grey irons[14]. In amounts of 0·01–0·03 per cent it is also used to control chill and annealability in malleable iron. Aluminium, however, is best avoided in grey iron foundries, as in quantities from 0·01 to 0·1 per cent it can very greatly increase the risk of pinholing[15] through hydrogen absorption from the mould.

Aluminium raises the pearlite change point by about 16°C for a 2 per cent addition, and 96°C for a 6·2 per cent addition, and raises the freezing point by about 16°C for each 1 per cent addition up to 5·75 per cent aluminium. Additions of 0·08–0·1 per cent can be used to eliminate the effect of nitrogen in producing compacted graphite in ingot moulds.

ANTIMONY

Antimony may be present in pig iron up to about 0·02 per cent; it may occur in larger amounts in castings as a result of the use of vitreous enamelled scrap, which may introduce 0·3–0·5 per cent, since antimony persists on re-melting. It tends to stabilise pearlite, lowers the tensile strength, transverse strength, deflection and impact strength, and raises hardness. 0·1 per cent antimony may lower the impact strength by 30–50 per cent. It may markedly increase the tendency of light castings to crack when present in quantities of more than 0·1 per cent[16]. Russian work basically confirms these conclusions and does not confirm previously reported advantages of antimony additions to cast iron for bearing purposes[17].

ARSENIC

Arsenic is present in some pig irons in amounts up to 0·1 per cent. Its effect on normal grey irons in the amounts usually present is not regarded as significant, but in nodular irons it markedly increases pearlite stability and tensile strength at high temperatures[18, 19] and in the absence of cerium interferes with the formation of nodular graphite structures.

BISMUTH

Bismuth is not normally present in pig irons, but can be introduced in scrap, and can promote the formation of Widmanstätten graphite (*Figures 34* and *35*), causing a marked reduction in tensile strength. It is, however, very rarely present in sufficient quantity to affect grey irons seriously. It acts, however, as a subversive element (like lead, titanium and antimony) to magnesium in the production of nodular iron[18]. It is also used in amounts up to 0·002 per cent to control chill and annealability of malleable iron.

BORON

Boron may be present in pig iron up to 0·002 per cent and occasionally up to 0·006 per cent. Boron is a carbide stabiliser and is widely used in amounts up to 0·001 per cent to control annealability of malleable iron. It is not normally employed in grey irons, but can be introduced accidentally from vitreous enamel[16] or from borax-bonded refractories, and may interfere with chill control methods.

LEAD

Lead[20, 21] may be present in pig irons in very small amounts and occasionally above 0·003 per cent, but may be introduced accidentally in scrap or with

scrap copper additions. Lead is an element that should be avoided in all cast irons as even in amounts of 0·0007 per cent it may produce Widmanstätten graphite (*Figures 34* and *35*) which is generally associated with lead contents above 0·001 per cent, particularly in the presence of aluminium and hydrogen[22]. Widmanstätten graphite is most likely to occur in section thicknesses above $\frac{1}{2}$–1 in (13–25 mm), and is usually associated with a drastic loss of strength. Mesh graphite (*Figures 36* and *37*) can also occur with lead associated with aluminium[21] and with tellurium[20] and drastically reduces the strength of the iron. Both forms of graphite may reduce tensile strength below 6 ton/in^2 (93 N/mm^2).

TELLURIUM

Tellurium is a most potent chill producer and may be used as a core or mould wash in local spots to promote soundness. It must not, however, be used extensively in a mould as an amount of 0·0007 per cent will produce Widmanstätten or mesh graphite and a grave loss of tensile strength[20]. The weakening effect of lead and tellurium is most pronounced in castings of section above $\frac{1}{2}$ in (13 mm).

TIN

Tin is a powerful pearlite stabiliser[23] even for irons with a carbon equivalent value of 4·7 and may be used where fully pearlitic structures are essential in amounts up to 0·15 per cent. In grey irons up to this amount, by eliminating free ferrite, it may raise the tensile strength by about 7 per cent, the hardness by 30 points Brinell and the modulus of elasticity by 4–5 per cent. There is no significant loss of impact strength in amounts below 0·15 per cent[24]. The effect on tensile strength and hardness is most marked in hypereutectic irons and is primarily caused by elimination of free ferrite. Tin does not promote the formation of free cementite and for this reason may be preferable to chromium in preventing ferrite formation for conditions of lubricated and sliding wear. In nodular iron tin has no adverse effect on nodule formation, but increases tensile and lowers elongation by stabilising pearlite. In nodular iron the pearlite with 0·085 per cent tin is stable at 500°C for long periods of time. The stability of tin-induced pearlite is an advantage where growth at temperatures above 450°C must be avoided. Tin should be absent where ferritising annealing is to be employed. A full summary of the effect of tin has been given by Robins[25].

TITANIUM

Titanium is occasionally employed up to 0·3 per cent to assist in the production of undercooled graphite structures[26], which are to be preferred in such alloys as Silal. It is present in many phosphoric pig irons in amounts up to

0·2 per cent. Titanium (0·015–0·025 per cent) will usually eliminate the effect of nitrogen in producing fissure defects in sand castings[27], or compacted graphite in ingot moulds.

MAGNESIUM AND CERIUM

Antimony, lead, titanium and bismuth interfere with nodule formation by magnesium alone in nodular iron. However, this effect can be neutralised by addition of cerium as mischmetall. (*See* section on nodular iron.)

The use of alloys in castings of various thicknesses

THIN CASTINGS UP TO $\frac{1}{2}$ in (13 mm) SECTION

There is little difficulty in obtaining tensile strengths up to 14–16 ton/in^2 (216–247 N/mm^2) in unalloyed iron of such sections, and over 20 ton/in^2 (309 N/mm^2) by inoculation with ferrosilicon or calcium silicide. If strengths of about 16–17 ton/in^2 (247–263 N/mm^2) are required without inoculation, 0·5 per cent nickel or copper[1, 6, 7] will permit a reduction in silicon of about 0·2 per cent or in carbon of about 0·1 per cent from a given level without increasing the risk of chill in thin sections. With 0·5–1·0 per cent nickel or copper, 1·8–2·2 per cent silicon, 3·0–3·2 per cent total carbon, tensiles of 17–19 ton/in^2 (263–293 N/mm^2) should be obtained in these sections. For higher tensile strengths without unduly increasing chill, combined additions of chromium and nickel in the ratio of 0·3 per cent chromium to 1·0 per cent nickel or copper may be used up to about 1·5 per cent nickel, to give tensiles up to about 22 ton/in^2 (340 N/mm^2) with carbon and silicon in the range quoted above. Where strengths above 20–22 ton/in^2 (309–340 N/mm^2) are required, additions of 0·3–0·6 per cent molybdenum, either alone or with 1 per cent nickel or copper, will produce pearlitic irons of high strength in this region. Tensiles above 24 ton/in^2 (371 N/mm^2) can only be reliably attained in grey irons with acicular structures, using molybdenum 0·8–1·5 per cent, and nickel 1·0–2·5 per cent, depending on section.

Inoculation with 0·2–0·3 per cent silicon gives marked freedom from chill when the initial silicon content is lowered by the same amount. Lowering the total carbon markedly increases the strength and the modulus of elasticity in relation to the tensile strength, both in unalloyed and alloyed cast irons. As a result of the increase in modulus of elasticity, complex castings with carbon contents below 3·0 per cent are particularly prone to carry substantial internal casting stresses and may require stress-relieving heat treatment. Total carbon contents below 3·0 per cent require greater care in the foundry and good technical control.

Table 74a MECHANICAL PROPERTIES OF TEST BARS CUT FROM HEAVY SECTION CASTINGS: COMPOSITION

	Specimen number and location	*Actual wall thickness* in	(mm)	*TC* %	*Si* %	*Mn* %	*P* %	*Cr* %	*Cu* %	*Ni* %	*Mo* %	*V* %
Plates for chemical vessels, 8 ft (2·4 m) dia., $3\frac{3}{4}$–5 in (95–127 mm) thick. Test pieces trepanned from centre	1	$3\frac{3}{4}$	(95)	3·41	2·38	0·74	0·69	—	—	0·01	—	—
	2	5	(127)	3·18	0·77	0·58	0·26	<0·04	—	1·01	<0·01	—
	3	5	(127)	2·98	1·09	0·6	0·1	<0·04	—	0·98	<0·04	—
Marine diesel cylinders, 24–33 in (610–838 mm) outside dia. × 2–4 in (51–102 mm), as-cast thickness. Test pieces cut from wall	4	$2\frac{1}{2}$	(64)	2·98	1·13	0·90	0·34	0·05	—	0·07	—	0·23
	5	$2\frac{1}{2}$	(64)	3·45	1·0	1·03	0·12	0·2	—	0·05	0·02	0·01
	6	$2\frac{1}{2}$	(64)	3·52	0·79	0·85	0·12	0·06	—	0·03	0·02	0·29
	7			3·25	0·94	0·98	0·36	0·05	—	0·07	—	0·37
	8	$2\frac{1}{2}$	(64)	3·12	0·72	0·95	0·16	0·1	—	0·11	0·01	trace
	9			3·20	1·68	0·66	0·14	—	—	—	—	0·15
	10	$2\frac{1}{2}$	(64)	3·03	1·24	0·82	0·08	0·04	—	0·02	0·31	0·02
Paper drying cylinder 16 ft (4·9 m) long × 10 ft (3 m) dia. × 4 in (102 mm) thick. Test pieces cut from wall	11	4	(102)	2·9	1·2	(0·7)	<0·1	<0·1	—	2·5	0·7	—
Hot mould propeller 86 in (2·2 m) dia.	Tip	0·8–1·0	(20–25)	3·46	0·75	1·0	0·174	—	—	—	—	—
	Blade	1·5–2·5	(38–64)	3·46	0·75	1·0	0·174	—	—	—	—	—
	Boss	3–4	(76–102)	3·46	0·75	1·0	0·174	—	—	—	—	—

Table 74b MECHANICAL PROPERTIES OF TEST BARS CUT FROM HEAVY SECTION CASTINGS

	Specimen number and location	*Tensile strength* ton/in²	(N/mm²)	*E* lb/in² × 10⁶	(GN/m²)	*Brinell Hardness No.*	*Transverse rupture stress* ton/in²	(N/mm²)	*Izod impact* ft. lb	(J)	*Fatigue stress* ton/in²	(N/mm²)
Plates for chemical vessels, 8 ft (2·4 m) dia., 3¾–5 in (95–127 mm) thick. Test pieces trepanned from centre	1	6–6·9	(93–107)	13·8	(95)	130	—	—	—	—	—	—
	2	15·9	(246)	19·9	(137)	201	—	—	—	—	—	—
	3	21·4	(331)	20	(138)	174	—	—	—	—	—	—
Marine diesel cylinders, 24–33 in (610–838 mm) outside dia. × 2–4 in (51–102 mm), as-cast thickness. Test pieces cut from wall	4	11·4	(176)	—	—	149	—		—	—	±4·0	(62)
	5	11·8	(182)	—	—	180	—	—	—	—	—	—
	6	13·3	(205)	—	—	156	—	—	—	—	—	—
	7	14·5	(224)	—	—	172	—	—	—	—	±5·5	(85)
	8	13·4	(207)	—	—	174	—	—	—	—	—	—
	9	18·4	(284)	22	(152)	193	—	—	—	—	±7·0	(108)
	10	18·4	(284)	—	—	186	—	—	—	—	—	—
Paper drying cylinder 16 ft (4·9 m) long × 10 ft (3 m) dia. × 4 in (102 mm) thick. Test pieces cut from wall	11	24·6	(380	—	—	—	—	—	—	—	—	—
Hot mould propeller 86 in (2·2 m) dia.	Tip	15·1–14·2	(233–219)	—	—	172–200	29·6	(457)	15–17	(20–23)	—	—
	Blade	13–11	(201–170)	14·0	(97)	163–170	28·8	(445)	19–26	(26–35)	±5·5	(85)
	Boss	10·2	(158)	—	—	163	—	—	13–17	(18–23)	±5·0	(77)

CASTINGS FROM $\frac{1}{2}$ in TO $1\frac{1}{2}$ in (13 mm TO 38 mm) SECTION

Unalloyed irons can give tensiles up to about 14–16 ton/in^2 (216–247 N/mm^2), using carbon equivalents of 3·7–3·85 and total carbon figures up to about 3·2 per cent. Above this range of tensile strengths, similar alloy additions can be made to those in the previous section, but the silicon should be lowered to about 1·6–1·8 per cent for the heavier end of the range.

CASTINGS FROM $1\frac{1}{2}$ in TO 6 in (38 mm TO 152 mm) THICKNESS

Unalloyed iron will not normally give tensiles above 10–14 ton/in^2 (154–216 N/mm^2) in the centres of very heavy sections, even by lowering the total carbon to the region of 2·8–2·9 per cent, with silicon 1·0–1·3 per cent. Tensiles of over 15 ton/in^2 (232 N/mm^2) in sections of 2–6 in (51–152 mm) definitely require the use of alloys. Additions of molybdenum or chromium up to about 0·5 per cent will give a marked increase in tensile strength.

In the heaviest sections, however, additions of chromium or molybdenum may give rise to centre carbide segregation with phosphorus above 0·15 per cent. This may be of no significance where surface machining only is required, but may create difficulties if deep sawing or milling operations are required. Nickel or copper alone are probably the most suitable alloying elements to use in castings requiring reasonable strength combined with low section sensitivity, and freedom from carbide segregation, particularly necessary where machining operations are carried out through the thickest sections.

Low total carbon figures will increase the tensile strength and also the modulus of elasticity, but may reduce the resistance to wear and scuffing under conditions of marginal lubrication.

Tensile strength above 22 ton/in^2 (340 N/mm^2) is attainable in 4 in (102 mm) thick castings with nickel or copper (over 1·5 per cent) and molybdenum (over 0·4 per cent). (*Table 74*, item 3.)

Table 74 gives some values that have been recorded from actual castings. The relation between test bar values and the actual strength in the casting has not been satisfactorily determined and appears to be much more affected by the total carbon figure than in castings below 2 in (51 mm) thickness.

The importance of knowing the precise properties required is illustrated in examples (1), (2) and (3) in *Table 74*. The lowest strength plate (1) actually gave over thirty years service under conditions where the higher strength alloy castings cracked in 2–5 years. Under these particular conditions a low modulus of elasticity was more important than a high tensile strength.

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MALLEABLE CAST IRON

This important engineering material differs from grey cast iron in that when cast it is hard and brittle, with a white fracture, and only becomes malleable and ductile after prolonged heat treatment under decarburising conditions in the case of whiteheart malleable, and neutral conditions in the case of blackheart malleable. As it is essential that the as-cast structure should contain no primary graphite, the silicon and carbon contents must be related to the section thickness of the casting.

These terms, whiteheart and blackheart, are based upon the appearance of the final fracture of the two materials, and reflect the basic differences in their structure, as indicated in Chapter 1.

Whiteheart malleable cast iron

This material is normally melted in the cupola and in the hard as-cast condition has an approximate composition of:

	%
Total carbon	3·0–3·7
Silicon	0·4–0·9
Manganese	0·2–0·4
Sulphur	0·3 max.
Phosphorus	0·1 max.

The chemical composition will vary from one foundry to another and according to the section of the castings, and is of less significance than the final structure and mechanical properties obtained after annealing.

The hard castings are annealed in a decarburising atmosphere which is produced, in some cases, by packing them in iron cans with blended iron ores and heating them up to 950–1050°C for a total cycle lasting up to about 7 days. A very considerable tonnage is also annealed in furnaces without ore packing, in which decarburisation is effected by controlling the composition of the surrounding atmosphere. Controlled oxidation of the carbon in the castings occurs, giving a carbon gradient from edge to centre; the process produces a material with a deep decarburised surface, which is fully ferritic close to the surface, but becomes pearlitic further below the surface and gradually merges into a pearlitic centre containing graphite nodules or temper carbon.

The variation in structure between edge and centre makes it impossible to get a satisfactory assessment of the mechanical properties by cutting test bars from the centre, and the acceptance tests are always based on annealed, as-cast (unmachined) tensile and bend test bars or by tests to destruction on the actual castings.

British Standard Specification 309:1958 requires the minimum properties in Imperial units given in *Table 75*. The specification was revised in BS 309:1972 using SI units with a conversion chart to Imperial units. Both SI and Imperial units are given in *Table 75*.

The chemical composition of the metal of the castings is considered to be of less importance than the mechanical properties of the malleablised castings.

The hardness of both Grades W.22/4 and W.24/8 does not normally exceed 200 Brinell in the core, but in heavy sections with fine pearlite may rise to 248 Brinell, which is the maximum permitted by the 1958 specification. In BS 309:1972 the maximum hardness for the corresponding grades W.410/4 and W.340/3 is given as 229.

In both the 1958 and 1972 specification 'The diameter of the test bar is intended to be representative of the important sectional thickness of the casting' and in the 1958 specification the section thickness of the casting is indicated in *Tables 1* and *2*. In the 1972 revision *Tables 1* and *2* do not give the casting section. *Table 75* in this work summarises both specifications. Comparable specifications for other countries are given in the Appendix.

Table 75 WHITEHEART MALLEABLE CAST IRON BS 309:1958, MINIMUM FIGURES

Grade	Sectional thickness represented in	L (mm)	Diameter of bar in	Ultimate tensile strength ton/in²	(kg/mm²)	Elongation	Bend test* Bar size in	Angle of bend	Mandrel dia. in	(mm)	Yield point† ton/in²	(kg/mm²)
W.22/4	Over $\frac{5}{16}$	(Over 7·9)	0·564	22	(34·6)	4% on 2·0 in 4% on 1·7 in	8 × 1 × 0·475	45°	2·00	(50·8)	13	(20·5)
	$\frac{5}{16}$ and under	(7·9 and under)	0·357	20	(31·5)	6% on 1·25 in 6% on 1·0 in	7 × 1 × 0·25	45°	1·75	(44·4)	12	(18·9)
W.24/8	Over $\frac{5}{16}$	(Over 7·9)	0·564	24	(37·8)	8% on 2·0 in 8% on 1·7 in	8 × 1 × 0·475	90°	2·00	(50·8)	14	(22·1)
	$\frac{5}{16}$ and under	(7·9 and under)	0·357	22	(34·6)	10% on 1·25 in 10% on 1·0 in	7 × 1 × 0·25	90°	1·75	(44·4)	13	(20·5)

*See clause 6a(ii).
†Optional (see clause 6a(iii).

BS 309:1972, SI Units, minimum figures

Grade	Diameter of bar mm	Tensile Strength N/mm²	ton/in²	0·5% proof stress N/mm²	ton/in²	0·2% proof stress N/mm²	ton/in²	Elongation %	Hardness, HB max.
W.410/4 (1958 Spec. W.24/8)	9	350	22·9	190	12·1	170	11	10	229
	12	390	25·3	230	14·9	210	13·7	6	229
	15	410	26·7	250	16·3	220	14·4	4	229
W.340/3 (1958 Spec. W.22/4)	9	270	17·5	—	—	—	—	7	229
	12	310	20·2	—	—	—	—	4	229
	15	340	22·3	—	—	—	—	3	229

There is an important section of the industry producing almost fully decarburised castings whose thickness does not exceed $\frac{1}{8}$ to $\frac{3}{16}$ in (3–5 mm). Typical of this class of work is that produced for pipe fittings and for the cycle trade. The importance of satisfactory decarburisation in this instance is that the fittings can be brazed to the steel tube at temperatures above the critical (740°C) without risk of re-solution of carbon with the possibility of embrittlement.

Although the centre of a whiteheart casting is normally pearlitic, the presence of the decarburised ferritic skin ensures freedom from brittleness and also, as this skin is very soft (about 95–110 Brinell), it can absorb plastically a considerable amount of energy from a local blow.

If, however, the decarburised skin is machined away or penetrated by deep threading, some of the shock resistance is lost.

Although for these reasons impact tests are not normally carried out on whiteheart malleable iron, the following figures are of interest. Whiteheart castings $\frac{5}{8}$ in (16 mm) in thickness were made, which in the standard annealed as-cast test bar [0·575 in (14·6 mm) actual diameter] gave a maximum tensile strength of 24·6 ton/in^2 (380 N/mm^2) and an elongation of 10 per cent on 2 in (51 mm). This material was machined to give the standard 10 mm square Izod test-piece, so that one face included the cast decarburised face, while the opposite face was in the core of the original bar. These specimens, when notched, gave Izod figures of 2–3 ft.lb (3–4 J), whether the notch was in the core face or the cast face. Similar specimens unnotched gave 30–40 ft.lb (41–54 J) when the blow was on the decarburised face, and 8–10 ft.lb (11–14 J) when the blow was on the core face. When using the standard Izod 0·45 in (11·4 mm) diameter round bar (as-cast and annealed) the Izod values were: notched 7 ft.lb (9 J), and unnotched 22–32·5 ft.lb (30–44 J).

These figures are taken from a single series of tests, and while not necessarily typical, give some indication of the shock resistance of the material.

Whiteheart malleable is very widely employed where ductility and shock resistance are called for, and is very freely machinable when compared with grey irons and steel, but is less freely machinable than blackheart malleable.

Whiteheart malleable shows a marked increase in shock resistance at temperatures about 100°C and is widely employed for conveyor chains in heat-treatment furnaces at temperatures up to 450°C. The stress to rupture in 100 000 h at 350°C is about 9 ton/in^2 (139 N/mm^2) for fully decarburised material to 15 ton/in^2 (232 N/mm^2) for material with room temperature strength of 30 ton/in^2 (463 N/mm^2). (*See* sections on heat resistance.) The shock resistance falls with falling temperature. It is widely used for fittings which require to be galvanised and does not appear to suffer from galvanising embrittlement.

Although the lower sulphur and higher manganese whiteheart malleable castings can be fusion welded, the ductility of the casting will be greatly impaired unless they are re-annealed. Weldability can be improved by restricting silicon to 0·45 per cent and sulphur to 0·12 per cent. They can, however, be brazed or bronze welded, and in fully decarburised sections a very shock resistant assembly can be produced. The data summarised in *Table 76* have been derived from various sources. The figures given can be considered as

Table 76 WHITEHEART MALLEABLE CAST IRON[1, 2, 3, 4, 5]

	*Imperial units BS 309:1958**	*SI units BS 309:1972*
Tensile strength, ton/in^2	22–24	270–410 N/mm^2
Yield, ton/in^2	12–14	(205–230 N/mm^2)
Limit of proportionality, ton/in^2	(7–7·6)	(108–118 N/mm^2)
0·1% proof stress	(11·6–12·7)	(180–200 N/mm^2)
0·2% proof stress (0·55 × tensile strength)	(12·1–13·2)	(149–226 N/mm^2)
0·5% proof stress	12·8–13·9	190–250 N/mm^2
Elongation %, min.	4–8	3–10%
Shear strength, ton/in^2: 0·9 × tensile strength	(20–22)	(242–370 N/mm^2)
Poisson's ratio	0·26	0·26
Cold bend, 8 × 1 × 0·375 in	45–90°	—
Brinell hardness, HB	120–248	229 max.
Fatigue strength/tensile strength ratio	0·40–0·6	0·45
Modulus of elasticity in tension and compression, lb/in^2	24–27 × 10^6	175·8 kN/mm^2
Modulus of elasticity in shear, lb/in^2	(9–10 × 10^6)	(62–68 kN/mm^2)
Modulus of rigidity, lb/in^2: 0·4 × modulus of elasticity	10–12 × 10^6	70·3 kN/mm^2
Specific gravity	7·3–7·7	7·3–7·7
Density, lb/in^3	0·264–0·278	—
Specific electrical resistance at 20°C, $\mu\Omega$.cm	27–37	24–26$\mu\Omega$.cm
Temperature coefficient of resistance (10–100°C)	0·0045/°C	—
Coefficient of thermal expansion −100–400°C	10–12·5 × 10^6	10–12·5 × 10^6
Specific heat (10–100°C) cal/g × °C	0·11	0·46 J/g.°C
Thermal conductivity (10–100°C) cal/cm^2/sec/°C/cm	0·10–0·128	41·9–45·2 W/m.°C
Latent heat of fusion, cal/g	(23)	(23 cal/g)
Melting point of unannealed white iron	(1 200°C)	(1 200°C)
Melting range of annealed malleable iron	(1 200–1 450°C)	(1 200–1 450°C)
Magnetic properties for fully decarburised whiteheart		
oersted (*H*)	25 50	100
gauss (*B*)	11 500 12 500	13 500
Design stresses: Tension, direct, ton/in^2	(4·7–5·7)	(73–88 N/mm^2)
Compression, direct, ton/in^2	(6·9–8·2)	(106–127 N/mm^2)
Fatigue, unnotched, ton/in^2	(±3·0–3·6)	(±46–55 N/mm^2)
Fatigue, notched, ton/in^2	(±1·8–2·0)	(±25·8–31 N/mm^2)

*Values in parentheses do not appear in the standards.

typical, but individual irons may differ appreciably and the figures must be regarded as a guide only.

Blackheart malleable cast iron

This material is produced with a lower total carbon content and a slightly higher silicon content than whiteheart. A typical range of as-cast analysis is shown below.

	%
Total carbon	2·2–3·0
Silicon	0·7–1·2
Manganese	0·2–0·5
Sulphur	0·05–0·16
Phosphorus	0·12 max.

The hard castings are then packed in an inert material, such as crushed slag or sand, and annealed at 850–950°C for periods up to 7 days, or heated in unpacked furnaces with controlled atmosphere. First stage annealing takes place up to 70 h at 850–950°C followed by controlled slow cooling down to about 650°C. Special grades are also produced with carbon contents below 1·8 per cent and silicon contents up to 2·0 per cent, and these require much shorter annealing cycles than the lower silicon varieties.

Because the heat treatment is carried out mainly under neutral conditions, the degree of decarburisation is very much less than whiteheart malleable iron, and by controlled cooling through the critical temperature practically the whole of the carbon is deposited as aggregate graphite, leaving a matrix almost completely ferritic in the centre and edge. There may be some residual pearlite in between the decarburised edge and fully ferritic centre.

There are three standard grades of this material in British practice (*Table* 77), specified under BS 310:1958 and BS 310:1972 (SI units).

Table 77 BLACKHEART MALLEABLE CAST IRON BS 310:1958, MINIMUM FIGURES. THE HARDNESS OF ALL THESE GRADES NORMALLY FALLS BETWEEN 110 AND 149 MAX. BRINELL, THE LATTER FIGURE IS THE HIGHEST PERMITTED BY BS 310:1958.

Grade	*Sectional thickness represented*	*Dia. of bar* in	*Tensile strength* ton/in²	(kg/mm²)	*Elongation % on 2 in or 1·7 in*	*Bend test*, degrees bend round 2 in dia. 8 × 1 × 0·375 in*	*Yield point†* ton/in²	(kg/mm²)
B18/6‡	All sizes	0·564	18	(28·4)	6	30	11	(17·3)
B20/10	All sizes	0·564	20	(31·5)	10	90	12	(18·9)
B22/14	All sizes	0·564	22	(34·6)	14	135	13	(20·5)

**See* clause 4(a).
†Optional.
‡*See* foreword of BS 310: 1958.

BS 310:1972 SI UNITS, MINIMUM FIGURES. IT MAY BE EXPECTED THAT THE CASTINGS WILL NOT HAVE A HARDNESS GREATER THAN 149 HB. THIS VALUE IS FOR INFORMATION ONLY AND IS NOT A MANDATORY REQUIREMENT OF THE STANDARD.

Grade	*Diameter of test bar* mm	*Tensile strength* N/mm²	ton/in²	*0·5% proof stress* N/mm²	ton/in²	*Elongation* %	*0·2% proof stress** N/mm²	ton/in²
B340/12	15 mm as-cast or	340	22·0	200	12·9	12	190	12·3
B310/10	not less than 12 mm	310	20·1	190	12·3	10	180	11·6
B290/6	machined	290	18·8	170	11·0	6	—	—

*For information only.

Table 78 TYPICAL TEST BAR PROPERTIES OF FERRITIC BLACKHEART MALLEABLE CAST IRON[3, 4, 5, 6, 7]

	*Imperial units BS 310: 1958**	*SI units BS 310: 1972*
Mechanical properties in tension		
Ultimate stress (0·564 in diameter bar) min.	18–22 ton/in^2	290–340 N/mm^2
Yield point (divider method)	11–13 ton/in^2	(170–200 N/mm^2)
Elongation on 2 in, min.	6–14%	6–12%
Proportional limit	(8·4–10·4 ton/in^2)	
Modulus of elasticity—tension and compression	21–24·5 × 10 lb/in^2	168·9 kN/mm^2
Brinell hardness number	110–149	
Poisson's ratio	0·26	0·26
Limit of proportionality (0·47 × tensile)	(8·4–10·4 ton/in^2)	130–160 N/mm^2
0·1% proof stress, ton/in^2	(11–14)	(170–216 N/mm^2)
0·2% proof stress, ton/in^2 (0·65 × tensile)	(11·7–14·3)	180–220 N/mm^2
0·5% proof stress, ton/in^2	(12·1–14·7)	186–226 N/mm^2
Bend test (1 in × $\frac{3}{8}$ in round 2 in mandrel) room temperature	30°–135°	
Mechanical properties in compression		
Ultimate stress (0·5 in diameter × 1 in)	(No definite failure)	
Proportional limit	(9·7–11·8 ton/in^2)	150–180 N/mm^2
Modulus of elasticity	(23–24·5 × 10^6 lb/in^2)	168·9 kN/mm^2
Limit of proportionality	(9·7–11·8)	150–180 N/mm^2
0·1% proof stress	(12·1–14·7 ton/in^2)	187–227 N/mm^2
0·2% proof stress	(12·2–14·8 ton/in^2)	188–230 N/mm^2
0·5% proof stress	(12·3–14·9 ton/in^2)	190–230 N/mm^2
Mechanical properties in shear (approx. 0·9 times the tensile)		
Ultimate stress	16·2–19·8 ton/in^2	250–306 N/mm^2
Modulus of elasticity in shear	12–13 × 10^6 lb/in^2	83–90 kN/mm^2
Mechanical properties in torsion		
Modulus of rupture	16·2–19·8 ton/in^2	250–306 N/mm^2
Modulus of rigidity	9–10 × 10^6 lb/in^2	67·6 kN/mm^2
Other properties		
Fatigue limit (0·45 of tensile strength)	8–10 ton/in^2	130–155 N/mm^2
Impact value (Izod 10 mm^2 notched)	7–14 ft lb	
Specific gravity	7·25–7·40	7·25–7·4
Specific heat at 20–100°C	0·122 cal/g.°C	0·5–0·69 J/g.°C
Heat conductivity (K) at 100°C	0·117–0·151 cal/cm^2/sec/°C/cm thickness	45·6–49 W/m.°C
at 400°C	0·111–0·139 cal/cm^2/sec/°C/cm thickness	49 W/m.°C

Table 78 (*continued*)

	*Imperial units BS 310: 1958**	*SI units BS 310: 1972*
Electromagnetic properties		
Magnetic induction ($H=100$)	14 000–15 200 gauss	
Max. permeability (μ)	1500–2140	2000
Intensity for max. permeability (H for μ max.)	3 oersted	
Remanence ($B=12\,000$)	5250 gauss	0·5 T
Coercive force ($B=12\,000$)	2·38–2·45 oersted	120 A/m
Hysteresis loss ($B=12\,000$)	6900–8350 erg/cm^3/cycle	0·6 μJ/mm^3 for $B=1$ T
Electrical resistivity	27–34 microhm/cm^3 at 20°C	30 $\mu\Omega$.cm
Maximum design stresses		
Direct tension, ton/in^2	(6·3–7·7)	(97·5–119 N/mm^2)
Direct compression, ton/in^2	(7·3–8·8)	(112–136 N/mm^2)
Fatigue, unnotched, $\pm$ton/in^2	(2·7–3·3)	(42–51 N/mm^2)
Fatigue, notched, $\pm$ton/in^2	(1·6–2·0)	(25–31 N/mm^2)

* Values in parentheses do not appear in the standards.

Grade B 18/6 material has been included chiefly to cover metal generally produced in a cupola furnace, intended particularly for applications in which pressure tightness is more important than a high degree of strength or ductility. The relatively high carbon content of cupola metal appears to reduce the solidification shrinkage found with low carbon irons.

The test bars are sand-cast to shape, as for whiteheart, annealed with the appropriate batch of castings and tested without machining the gauged areas [0·564 in (14·3 mm) diameter].

Because of its lower hardness, blackheart malleable is very freely machinable and is very widely employed for automobile components where shock resistance and machinability are required.

The mechanical properties of blackheart malleable iron, shown in *Table 78*, are markedly affected by the method of manufacture and the composition; for example, each additional 0·1 per cent of silicon increases the yield and ultimate stress by not less than 0·6 ton/in^2 (9 N/mm^2), and each 0·01 per cent of phosphorus increases the yield and ultimate stress by about 0·22–0·27 ton/in^2 (3–4 N/mm^2). The elongation and Brinell hardness are not affected significantly. The avoidance of primary graphite in heavy sections limits the amount of silicon that may be present in a given section, and increasing both silicon and phosphorus raises the brittle–ductile transition temperature. (*See* section on impact.)

It has been shown, however, that if the phosphorus content is below 0·07 per cent and the graphite is in a suitable form, brittle–ductile transition temperatures as low as −45°C may be recorded in sections up to 3 in (76 mm) in diameter[8]. Susceptibility to galvanising embrittlement increases markedly as the phosphorus content exceeds 0·07 per cent unless quenching from 650°C

prior to galvanising is carried out. Water quenching from 650°C followed by tempering for 1 h at 350°C and slow cooling, will lower the brittle-transition temperature for the mean impact value of an iron containing 0·12 per cent phosphorus from +72 to −22°C. This treatment also raises the yield point, tensile strength and elongation.

PROPERTIES AT HIGH TEMPERATURES

Some short term tensile properties of a blackheart malleable iron are given in *Table 79*. Kanter and Guarnieri[9] carried out short term and creep tests on cupola-produced blackheart malleable which are summarised in *Table 80*. The material, as-cast, contained total carbon 3·16 per cent, silicon 0·91

Table 79 SHORT-TIME HIGH TEMPERATURE TENSILE TESTS OF BLACKHEART MALLEABLE CAST IRON (*see also* Chapter 4—Hear Resistance)

Composition:	%
Total carbon	2·2
Silicon	1·05
Manganese	0·35
Sulphur	0·08
Phosphorus	0·07

Temperature °C	*Ultimate tensile strength* ton/in²	(N/mm²)	*Yield stress* ton/in²	(N/mm²)	*Elongation* %	*Reduction of area* %
15	24·0	(371)	15·7	(242)	16	23
100	23·5	(363)	16·5	(255)	14	22
200	23·0	(355)	17·5	(270)	12	20·5
300	22·5	(347)	18·5	(286)	10·5	19
400	22·7	(351)	19·7	(304)	10	18
500	20·5	(317)	19·0	(293)	11	20
600	11·5	(178)	11·0	(170)	17	33
700	7·2	(111)	6·0	(93)	23	41
800	4·2	(65)	4·0	(62)	16	44

Table 80

		Short term tests					*Total creep per 10 000 h*			
Temperature		*Tensile strength*		*0·2% set stress*		*Elongation % on 2 in (51 mm)*	*0·10%*		*1·0%*	
°F	C	ton/in²	(N/mm²)	ton/in²	(N/mm²)		ton/in²	(N/mm²)	ton/in²	(N/mm²)
80	26·7	18·5	(286)	13·6	(210)	9·0			–	–
500	260	17·4	(269)	12·0	(185)	6·0				
600	316	17·2	(266)	11·1	(171)	5·5	9·4	(145)	12·5	(193)
700	371	--		--	--		6·7	(103)	8·9	(137)
800	427	14·8	(229)	10·2	(158)	5·5	3·5	(54)	5·5	(85)
1000	538	7·2	(111)	6·4	(99)	10·0	0·45	(7)	1·1	(17)

per cent, manganese 0·47 per cent, sulphur 0·092 per cent, phosphorus 0·17 per cent. Annealed, the total carbon was 2·78 per cent and the structure was almost fully ferritic.

At 372°C, after a small initial growth up to 200 h amounting to about 0·000 15 in/in, a sustained stress of 4·48 ton/in^2 (69 N/mm^2) showed no further creep up to 2400 h.

Kanter and Guarnieri concluded that stresses up to $2\frac{1}{2}$ ton/in^2 (39 N/mm^2) are justified at working temperatures of 410°C, and that permanent growth up to 480°C is small and is attributable to graphitisation of residual pearlite. (*See also* section on heat resistance.)

Ferritic blackheart can be cyanide hardened to 600–900 HV to 0·000 25 in (6μm) depth which increases fatigue strength from $\pm$11·5 ton/in^2 (178 N/mm^2) to 15·2 ton/in^2 (235 N/mm^2).

Pearlitic blackheart

This is a stronger, less ductile form of malleable iron produced either (a) by alloying normal blackheart compositions with carbide stabilisers (e.g. manganese), or (b) by modifying the cooling cycle through the critical range so as to retain a predominantly pearlitic matrix, or (c) by a suitable reheating treatment, e.g. air or oil quenching and tempering of the normal blackheart product. Although in general the structures produced are usually pearlitic, improved physical properties can be obtained if the pearlite in the matrix is spheroidised. Pearlitic blackheart malleable iron is characterised by higher tensile strength and yield or proof stress, lower elongation and higher hardness. Pearlitic malleable irons are covered by BS 3333:1961 and BS 3333:1972 (SI units) and typical properties are shown in *Table 81*. Specifications for varying grades of pearlitic blackheart, however, have been issued by Sweden and the USA and are summarised in the section on standard specifications (Appendix). The material can be oil-quenched and tempered in suitable sections to give hardnesses of 48–52 R_c.

The composition of pearlitic blackheart malleable, however, markedly affects its hardenability, and iron low in silicon and maganese may show a very poor response to hardening treatment (*see* section on heat treatment).

Hardenability curves for many pearlitic blackheart irons have been published[10] and are referred to in the section on heat treatment.

Selective hardening

This may be carried out by induction or flame processes or by lead or salt bath treatment; by suitable quenching and tempering, hardnesses from 400 to 600 DPN may be obtained.

Table 81 TYPICAL PROPERTIES OF PEARLITIC BLACKHEART MALLEABLE [3, 5, 6] TO BS 3333. THE DATA SUMMARISED BELOW CAN BE REGARDED AS TYPICAL, BUT INDIVIDUAL IRONS MAY DIFFER APPRECIABLY OWING TO DIFFERENCES IN PRODUCTION METHODS ETC., AND THE FIGURES MUST THEREFORE BE REGARDED AS A GUIDE AND NOT AS ABSOLUTE.

	*BS 3333:1961 Grades P28/6 & P33/4 imperial units**	*BS 3333:1972 Grades P440/7–P690/2 SI units*
Mechanical properties in tension, shear and torsion		
Tensile strength, ton/in^2	28–33	440–690 N/mm^2
Limit of proportionality (0·005% proof stress)	10·3–12·4	—
0·1% proof stress	17·1–20·6	—
0·2% proof stress (0·60–0·62 × tensile)	17·6–21·3	250–520 N/mm^2
0·5% proof stress	18–22·5	270–540 N/mm^2
Elongation % on 2 in, min.	6–4	2–7
Modulus of elasticity, lb/in^2, tension and compression	25×10^6	172·4 kN/mm^2
Shear strength, ton/in^2 (0·9 × tensile strength)	25·2–29·7	(395–620 N/mm^2)
Torsional strength, ton/in^2 (0·9 × tensile strength)	25·2–29·7	(395–620 N/mm^2)
Modulus of rigidity, lb/in^2	10×10^6	68·9 kN/mm^2
Poisson's ratio	0·26	0·26
Brinell hardness	143–229	149–285
Properties in compression		
Limit of proportionality, ton/in^2	14·7–17·4	(227–268 N/mm^2)
0·2% proof stress, ton/in^2	18·6–22·1	(287–341 N/mm^2)
0·5% proof stress, ton/in^2	18·8–22·8	(290–353 N/mm^2)
Fatigue (Wöhler) unnotched, ± ton/in^2 0·45–0·6 tensile	12·6–14·4	±98–153 N/mm^2
notched, 0·6 × tensile	7·1–8·6	±265–415 N/mm^2
Coefficient of thermal expansion		
20–100°C	10×10^{-6}/°C	10–12·5 × 10^{-6}/°C
20–200°C	11×10^{-6}/°C	
20–400°C	$12{\cdot}5 \times 10^{-6}$/°C	
Thermal conductivity		
100°C/cal/cm^3 sec °C	0·100–0·108	38·5–45·2 W/m.°C
400°C/cal/cm^3 sec °C	0·100–0·092	
Specific gravity	7·3	7·3
Design stresses, maximum		
Direct tension, ton/in^2	(7·7–9·3)	(118–144 N/mm^2)
Direct compression	(11·0–13·0)	(170–201 N/mm^2)
Fatigue, unnotched, ± ton/in^2	(4·2–4·8)	(±68·9–74 N/mm^2)
Fatigue, notched, ± ton/in^2	(2·4–2·9)	(±37–44·7 N/mm^2)

Table 81 (*continued*)

	*BS 3333:1961 Grades P28/6 & P33/4 imperial units**	*BS 3333:1972 Grades P440/7–P690/2 SI units*

Strength at elevated temperatures

	Room temp.	100 °C	200 °C	300 °C	400 °C	450 °C	500 °C	550 °C	600 °C	650 °C
Ultimate tensile stress, ton/in^2	34	33	32	31	30	33	31	24	17	12
N/mm^2	525	510	494	479	463	510	479	371	263	185
Yield stress, ton/in^2	22	21	19	18	20	23	18	13	11	8
N/mm^2	340	324	293	278	309	355	278	201	170	124
Elongation, %	6	6	6	6½	7	7	9	12	15	18

BS 333: 1961, Grades P28/6, P33/4, unalloyed
Stress to rupture in 100 000 h at 350°C: 9–13 ton/in^2
(*See* section on heat resistance, p. 273)

Electromagnetic properties (*Figure 83*)[11, 12]

Magnetic induction: $H=$ 10	3000–5000 gauss	$H=796$ A/m 0·3–0·5 T
$H=$ 50	10 000–11 500 gauss	$H=3979$ A/m 1·0–1·15 T
$H=100$	12 000–13 000 gauss	$H=7958$ A/m 1·2–1·3 T
Max. permeability ($\mu=B/H$)	350–660 440–829 μH/m	
H for μ max.	10–15 oersted 796–1194 A/m	

Mechanical properties after heat treatment

Normalised, 780–800°C		
Ultimate tensile stress	40–44 ton/in^2	618–680 N/mm^2
Yield stress	28–32 ton/in^2	432–494 N/mm^2
Elongation	3%–4%	
Brinell hardness no.	229–302	
Oil hardened and tempered		
Ultimate tensile stress	45–50 ton/in^2	695–772 N/mm^2
Yield stress	30–35 ton/in^2	463–541 N/mm^2
Elongation	1%–2%	
Brinell hardness no.	223–401	

*Values in parentneses do not appear in the standards.

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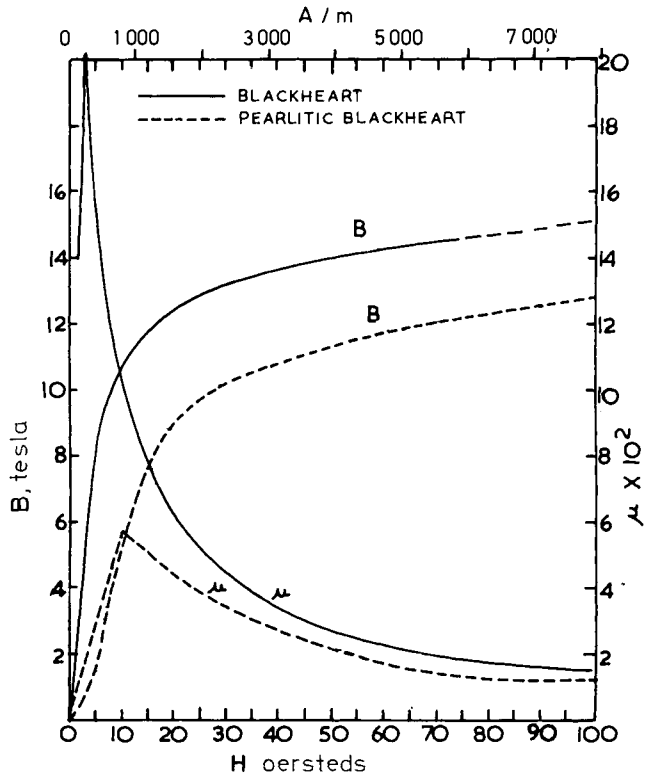


Figure 83. Magnetisation curve for blackheart and pearlitic blackheart malleable cast iron

NODULAR CAST IRON

R. Barton, B.Sc.*

The development of a process for the direct production of free graphite in the nodular form was first introduced by H. Morrogh of the BCIRA in July 1946, using cerium as the nodularising addition. Subsequent work carried out by International Nickel Ltd. produced similar results when using magnesium instead of cerium.

Since the introduction of nodular iron there has been a continuous increase in its production and its fields of application. There have also been numerous production methods devised which have been claimed to result in the formation of graphite nodules, but at the present time magnesium, or magnesium

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alloy, is invariably employed to effect the desired graphite form during solidification.

FLAKE GRAPHITE CAST IRON

The microstructure of grey cast iron is characterised by graphite flakes, devoid of strength, dispersed throughout a steel-like matrix. The flakes act as effective stress raisers in the metal and explain the relatively low strength [10–20 ton/in^2 (154–309 N/mm^2) tensile strength] of cast iron compared with its steel-like matrix [30–60 ton/in^2 (463–927 N/mm^2)]. Although by control of chemical composition and the use of inoculation techniques a refinement of the graphite and matrix structures can be obtained, the highest strength of an as-cast flake graphite cast iron is of the order of 28–30 ton/in^2 (432–463 N/mm^2), in the acicular type of cast iron.

MALLEABLE CAST IRON

Malleable cast iron is initially made from material possessing a chemical composition such that solidification proceeds according to the metastable iron/iron–carbide system rather than the iron–graphite system. This produces castings that are extremely hard and brittle. However, by holding the casting at an elevated temperature (950–1000°C), breakdown of the iron carbide (cementite) occurs, producing graphite in the form of aggregates which, owing to their configuration, have a less pronounced effect upon the matrix than normal flake graphite. The result is a material which possesses a good measure of strength combined with ductility. Thus, the properties of the steel-like matrix are further realised.

NODULAR CAST IRON

Whereas the production of graphite aggregates or spheroids in malleable iron is only accomplished by heat treatment, nodular graphite can readily be produced in the as-cast state by cerium and/or magnesium. Since the magnesium addition has proved the more adaptable and economic of the two processes, it is general to use this element as the basis for the production of most nodular iron. However, work by Morrogh has shown that the presence of minute amounts of certain deleterious elements (e.g. titanium, lead, bismuth and arsenic) can result in inferior structures, but that these effects can be overcome by the addition of cerium to the extent of 0·005 to 0·01 per cent. Thus, unless pure charge materials can be guaranteed, it is the recommended practice to use a combined addition of magnesium and cerium for the production of a good nodular graphite structure.

Foundry practice in the production of nodular cast iron

PATTERNMAKERS' CONTRACTION

As with other cast irons, it is impossible to quote a particular patternmaker's allowance for nodular iron which will satisfy all types of casting, the actual value being a function of casting design and mould conditions. However, the following figures may be accepted for general guidance when first changing over to the production of nodular iron:

(1) Where the casting is grey and predominantly pearlitic in the as-cast state, the contraction allowance should be 0·1–0·15 in per ft (1 : 120–1 : 80).
(2) When the casting is pearlitic as-cast and ultimately heat treated to the ferritic condition, or alternatively when it has a ferritic matrix in the as-cast condition, the patternmakers' contraction is reduced, owing to graphitisation, to 0·05–0·1 in per ft (1 : 240–1 : 120).
(3) When thin sectioned castings are involved and carbide is formed in the as-cast condition, subsequent annealing will give a marked expansion, and the overall patternmakers' allowance may therefore vary from a contraction of 0·05 in per ft (1 : 240) to an expansion of 0·1 in per ft (1 : 120), the actual value depending upon the amount of carbide in the as-cast state.

FLUIDITY

The fluidity of nodular iron at a particular temperature is similar to that of a flake graphite cast iron of equivalent composition. The factors affecting the fluidity at a fixed temperature are the carbon, silicon and phosphorus contents, the first element having the greatest effect. Thus, as most nodular irons have a high carbon content (3·4–4·0 per cent) their fluidity is generally higher than that of high duty flake graphite castings where the carbon is usually held between 3·0 and 3·2 per cent.

SOLIDIFICATION SHRINKAGE

Because of its mode of solidification, nodular iron normally requires more extensive feed metal than grey cast iron and care should be taken in design to avoid extreme and rapid changes in section, and to avoid heavy metal sections in parts of the casting where access to feed metal is impossible. All general recommendations for casting design apply to nodular iron and it is essential that the co-operation of the foundry be invited at the earliest opportunity when designing components to be made in nodular iron.

HEAT TREATMENT

In the as-cast state the microstructure of nodular iron may range from a completely white structure where thin sections are involved, to that of nodular graphite in a completely ferritic matrix. By virtue of the fact that, like grey irons, nodular irons have their own 'carbon bank' available, heat treatments can be applied to produce a wide range of matrix structures. Thus, by heating to a temperature above the critical, graphitic carbon will dissolve in the austenite until equilibrium conditions are attained. By air-cooling from this temperature the austenite undergoes the pearlitic transformation, thus maintaining an amount of carbon dispersed in the matrix. On the other hand, a very slow cooling through the temperature range will allow depletion of the carbon from the matrix to occur, and instead secondary graphite is deposited around the graphite nodules. The heat treatments normally applied on a commercial scale may be summarised as follows:

(1) Heating to a temperature of 850–900°C, the length of soaking time depending upon the amount of free carbide present. Air-cooling from this temperature will provide a predominantly pearlitic matrix except in heavy sections.
(2) Removal of carbides and austenitising as in (1). Oil-quenching from 850–900°C will produce a martensitic matrix which may then be tempered to produce the required range of mechanical properties. This treatment is limited to castings of relatively thin section unless alloying elements are present.
(3) Removal of carbides and austenitising as in (1). Cooling to 680–700°C and usually holding at this temperature for a period of time in order to effect complete graphitisation with the production of a ferritic matrix. The need for holding and the time involved will vary with basic composition and section size, but is usually of the order of 8–12 h. The material is finally cooled to about 600°C before removal from the furnace.
(4) The air-cooled material from (1), possessing a pearlitic matrix, may be annealed at a sub-critical temperature (700–720°C), giving a completely ferritic matrix as in (3). The time of holding at the sub-critical temperature is dependent upon composition and section size, but 8–12 h should be adequate.

The sub-critical annealing of an as-cast structure is not recommended, as inferior impact properties are obtained owing to the non-homogenisation of the as-cast structure at this temperature.

STRESS-RELIEVING

Any castings that are complex and likely to contain residual stresses, either in the as-cast condition or after normalising, may be stress-relieved by heating to a temperature of 450–550°C. Although no marked structural change occurs during this treatment, there is a tendency for softening to occur when the higher temperatures are used.

Table 82 PHYSICAL PROPERTIES OF NODULAR CAST IRON

Specific gravity: 7·1–7·2

Coefficient of thermal expansion:
20–600°C: $13–14 \times 10^{-6}$ in/in.°C
0–100°C: $10·5–11·8 \times 10^{-6}$ in/in.°C

Thermal conductivity[1]:
For a given composition the change from flake graphite to nodular graphite results in a fall in thermal conductivity equal to about 0·025–0·03 cal/cgs/°C. Variations in silicon and nickel content are also likely to affect the conductivity values, the general trend being shown in *Figure 84*. For maximum conductivity, nickel should be absent and the silicon held to as low a level as possible.

Typical values for pearlitic and ferritic matrix structures obtained at increasing temperatures are as follows:

	Thermal conductivity cal/cm²/cm/sec/°C (W/m.°C)			
Matrix	100°C	200°C	300°C	400°C
Pearlitic (2·4% Si, 1·35% Ni)	0·0742 (31·1)	0·0734 (30·7)	0·0725 (30·4)	0·0718 (30·1)
Ferritic (2·6% Si, 1·43% Ni)	0·0833 (34·9)	0·0831 (34·8)	0·0829 (34·7)	0·0827 (34·6)

Electrical resistance:
Electrical resistance measurements on nodular iron at room temperature have given the following results:

Pearlitic (2% nickel)	71×10^{-6} ohm/cm³ (0·71 μΩ.m)
Ferritic annealed (2% nickel)	58×10^{-6} ohm/cm³ (0·58 μΩ.m)
Ferritic annealed (nickel-free)	39×10^{-6} ohm/cm³ (0·39 μΩ.m)

Magnetic properties[2, 3]:
Changing the graphite shape from flake to the nodular form results in an increase in magnetic permeability. Further improvement and a lowering of the hysteresis loss is obtained with a ferritic matrix. For use as a soft magnetic material, the matrix must be ferritic, nickel should be absent and the available results also suggest that high silicon contents are beneficial.

Pearlitic Matrix (2% nickel):

Remanent magnetism	5400 gauss (0·54 T)
Coercive force	11·5 oersted (915 A/m)
Hysteresis loss	30 390 erg/cc/cycle (3039 J/m³) (for $B = 10\,000$ gausses) ($B = 1·0$ T)
Maximum permeability (μmax.)	290 (364 μH/m)
Value of H for μmax.	15 oersted (1194 A/m)

Ferritic Matrix:

	2·6% Si, 2·23% Ni	2·6% Si, Nickel-free
Remanent magnetism	4640 gauss (0·464 T)	6065 gauss (0·6065 T)
Coercive force	2·5 oersted (199 A/m)	1·5 oersted (119 A/m)

Table 82 (*continued*)

Hysteresis loss	7000 erg/cm^3. cycle, $B = 10\,000$ (700 J/m^3)	3395 erg/cm^3, cycle (340 J/m^3)
Max. permeability (μ max.)	1150 (1445 μH/m)	2060 (2589 μH/m)
Value of H for μ max.	4 (318 A/m)	3 (238 A/m)
	2·1% Si, <0·2% Ni	2·8% Si, <0·2% Ni
Remanent magnetism	5600 gauss (0·56 T)	5590 gauss (0·559 T)
Coercive force	1·83 oersted (146 A/m)	1·44 oersted (115 A/m)
Max. permeability (μ max.)	1700 (2136 μH/m)	2810 (2739 μH/m)

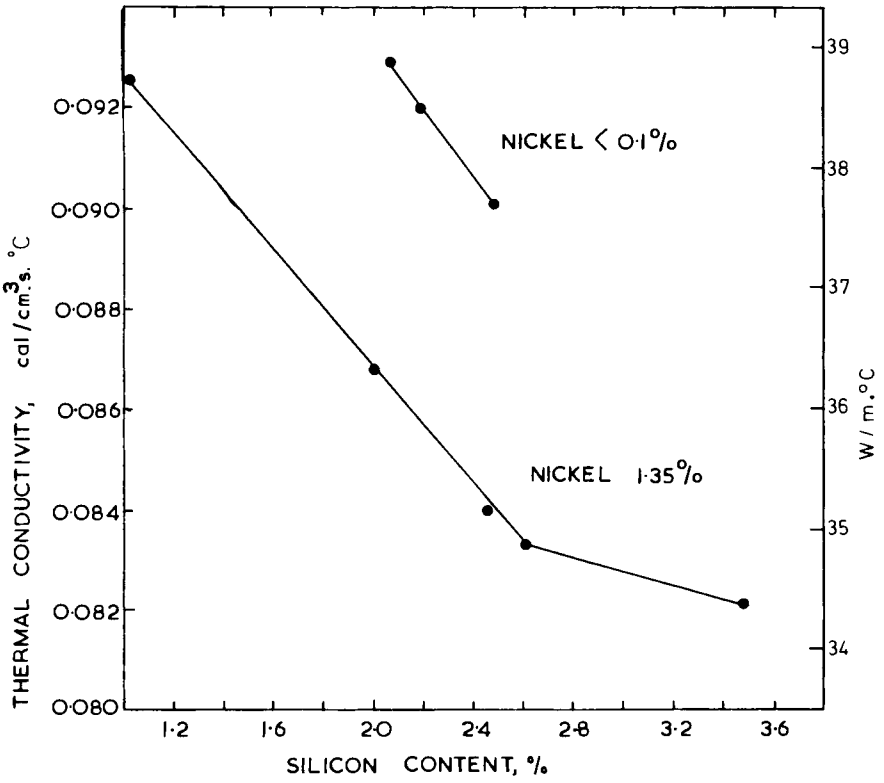


Figure 84. Thermal conductivity effect of silicon and nickel in ferritic matrix

Mechanical properties for nodular cast iron (*see Tables 183a, b, c, d* and *e* on pp. 477–481)

Unlike grey cast iron, nodular iron behaves elastically over a considerable stress range, in both tension and compression[4]. The effect of the graphite

is to reduce the useful cross-sectional area, decreasing the modulus of elasticity to $24–26 \times 10^6$ lb/in^2 (165–179 GN/m^2), compared with a value of 30×10^6 lb/in^2 (207 GN/m^2) for steel. In addition, nodular iron does not have a yield point, the deviation of the stress/strain curve from the elastic portion being gradual but continuous. Gilbert considers that at stresses above the

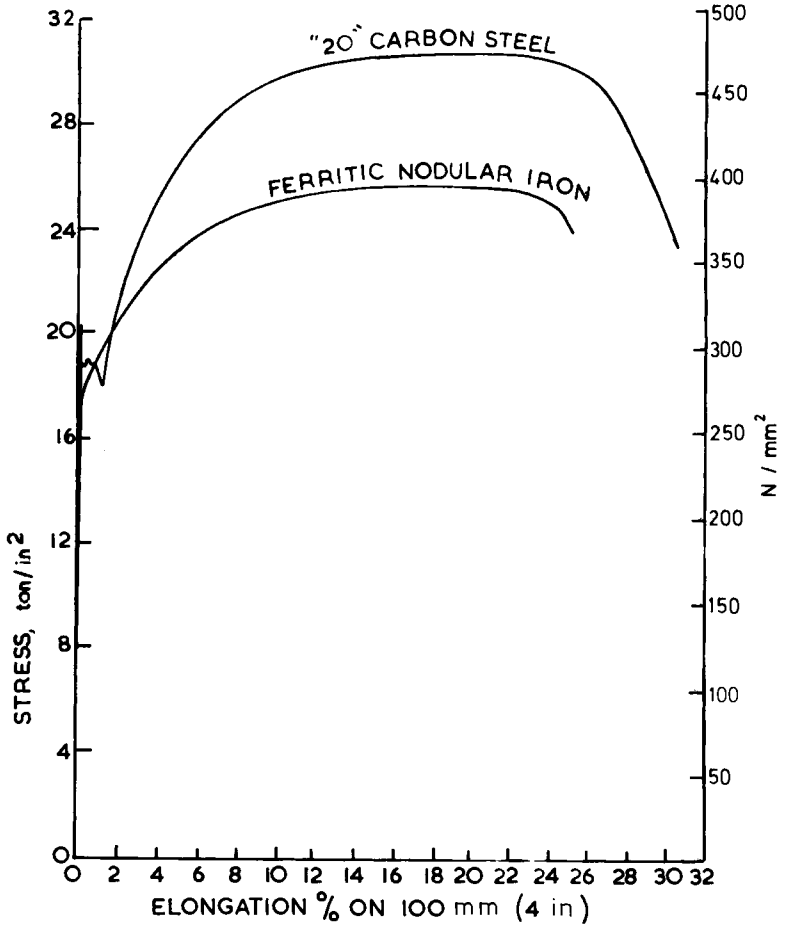


Figure 85. Stress/extension curves for a '20' carbon steel in the as-rolled condition and a 1·9% silicon ferritic iron

elastic limit the graphite nodules do not deform as though they are part of the matrix, voids forming in the longitudinal direction under tensile stress, giving an overall increase in volume which is not compensated for by compaction in the lateral direction. The complete stress/elongation curve for a low carbon steel and a ferritic nodular iron are shown in *Figure 85*. These show that although the final elongation of the steel is appreciably higher than the nodular iron, the difference is gained by local elongation or necking after the fracture has been initiated.

A wide range of mechanical properties can be produced in nodular iron by varying the matrix condition, the stronger and harder materials being associated with a pearlitic or quenched and tempered structure, and the lower strengths and high elongations with the ferritic matrix. Intermediate levels of strength and elongation values can be obtained by controlling the ratio of pearlite to ferrite in the matrix, typical values being illustrated in *Figure 86.*

BS 2789:1961, covering iron castings with spheroidal or nodular graphite, contains six grades with different levels of tensile strength and elongation, the minimum mechanical properties specified together with the type of associated matrix structure being shown in *Table 83.*

The different types of matrix structure are obtained either by control of the heat treatment cycle or the composition of the material. Considerably higher strengths, 50–55 ton/in^2 (772–849 N/mm^2) can readily be obtained by normalising or employing a quench and temper heat treatment, possibly in conjunction with the addition of alloying elements—copper, nickel, manganese and molybdenum.

An increase in proof stress and tensile strength can be attained in the ferritic condition by the addition of elements such as silicon, nickel and manganese, but always at the expensive of impact ductility, especially with manganese and silicon.

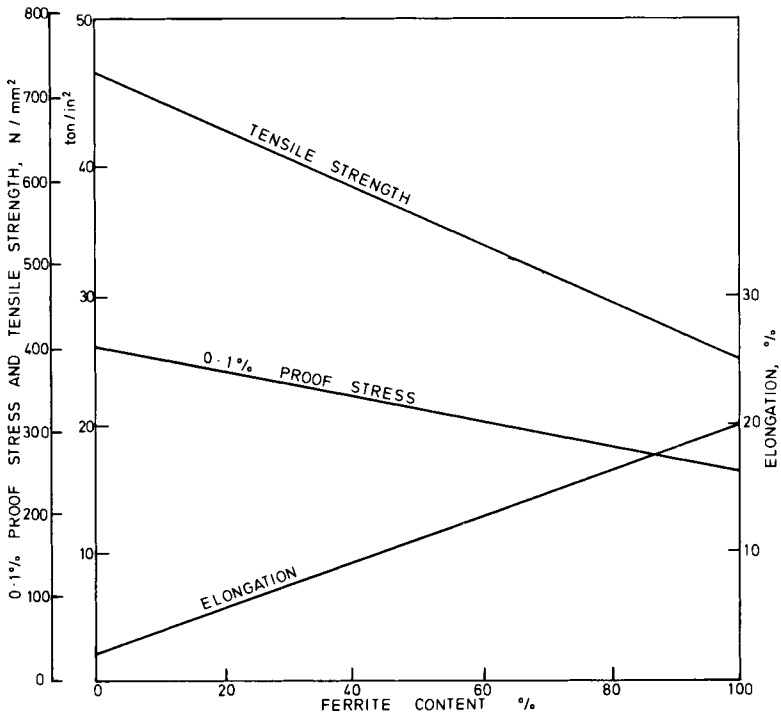


Figure 86. Effect of ferrite content of matrix on tensile properties

Table 83 BS 2789:1961

Property	Grade					
	SNG 24/17	*SNG 27/12*	*SNG 32/7*	*SNG 37/2*	*SNG 42/2*	*SNG 47/2*
0·1% proof stress, ton/in²*	14	17	20	22	23	26
0·2% proof stress, ton/in²*	15	18	21	23	25	27
0·5% proof stress, ton/in²*	15	18	22	25	28	30
Tensile strength, ton/in²	24	27	32	37	42	47
Elongation, %	17	12	7	2	2	2
Brinell hardness*	130–170	150–180	190–220	220–250	220–260	250–300
Modulus of elasticity, $\times 10^6$ lb/in²*	24–26	24–26	24–26	24–26	24–26	24–26
Fatigue limit (unnotched), ton/in²	12	13	14	16	17	19
Fatigue limit (notched), ton/in²	8	8	9	9	10	11
Matrix type*	Ferrite	Ferrite	Ferrite/ Pearlite	Pearlite/ Ferrite	Pearlite	Pearlite

BS 2789:1973

Property	Grade					
	370/17	*420/12*	*500/7*	*600/3*	*700/2*	*800/2*
0·2% proof stress, N/mm²	230	250	310	350	400	460
Tensile strength, N/mm²	370	420	500	600	700	800
Elongation, % min.	17	12	7	3	2	2
HB†	<179	<201	170–241	192–269	229–302	248–352
Matrix type†	Ferrite	Ferrite	Pearlite and ferrite	Pearlite and ferrite	Pearlite	Pearlite or tempered structure

*Not included in specification—for information only.
†For information only.

For any given matrix structure the tensile properties determined at stresses up to at least the 0·2 per cent proof stress level are not affected by the use of very heavy or large cast sections (*Figure 87*). The ultimate properties are, however, affected when large sections are employed. In the case of the pearlitic high strength nodular irons, early failure along the stress/strain curve results in a decrease in tensile stress, whereas in the ductile ferritic grades of nodular iron having stress/strain curves which become nearly horizontal before failure, early failure is shown mainly by a decrease in elongation.

HARDNESS

The hardness of nodular iron varies according to its matrix type and composition. With a fully pearlitic matrix, values up to 300–310 HB are obtained and for a ferritic matrix the lowest general hardness level is about 150 HB.

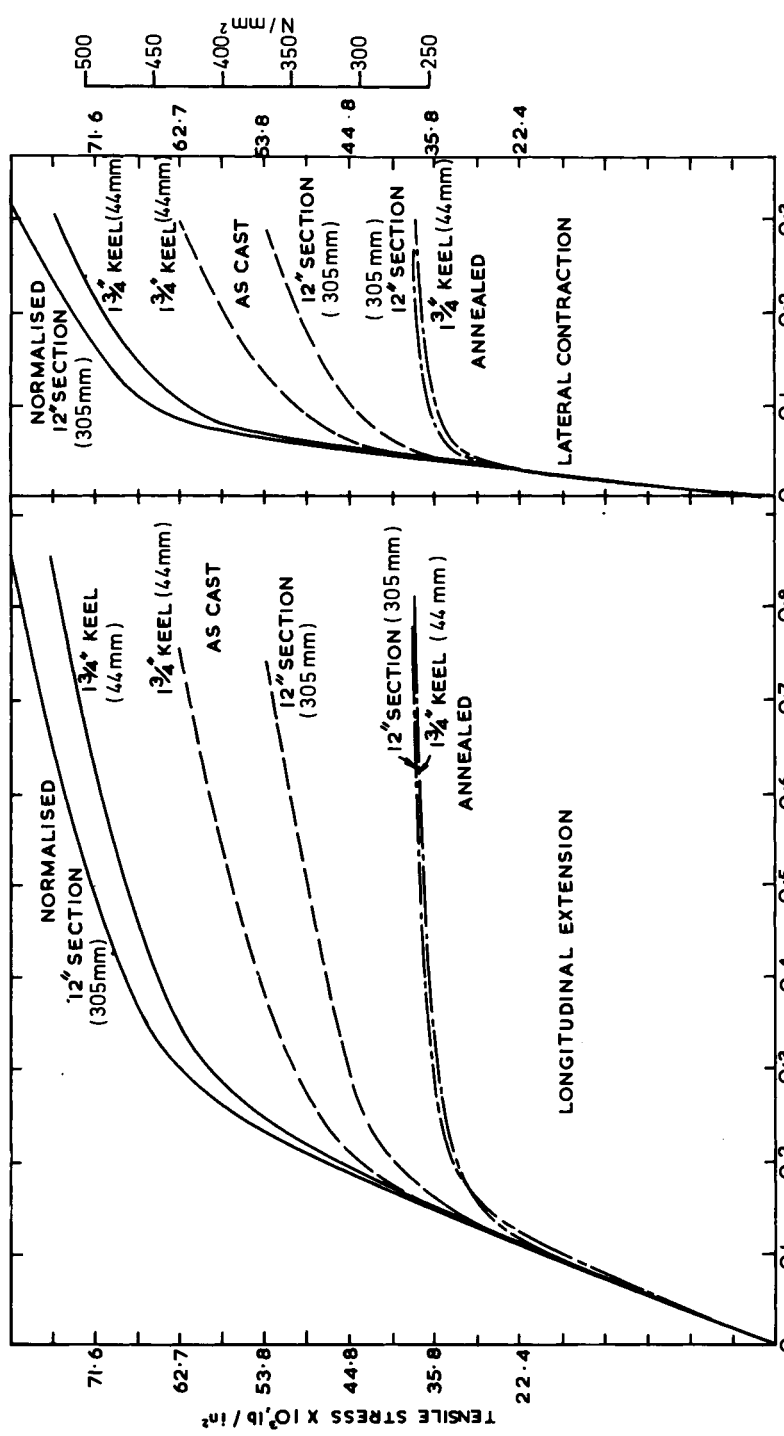


Figure 87. Comparison of stress/total strain curves

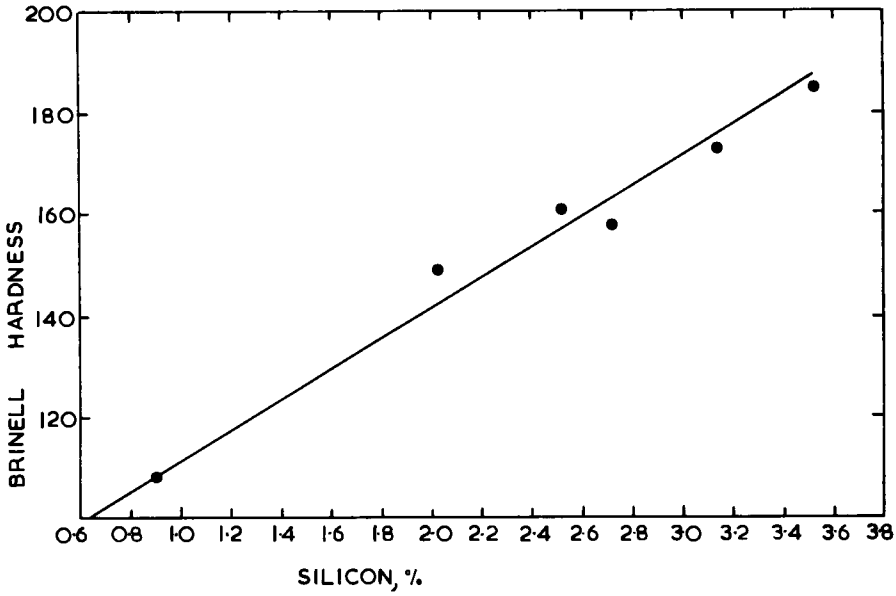


Figure 88. Relationship between silicon content and hardness in ferritic nodular iron

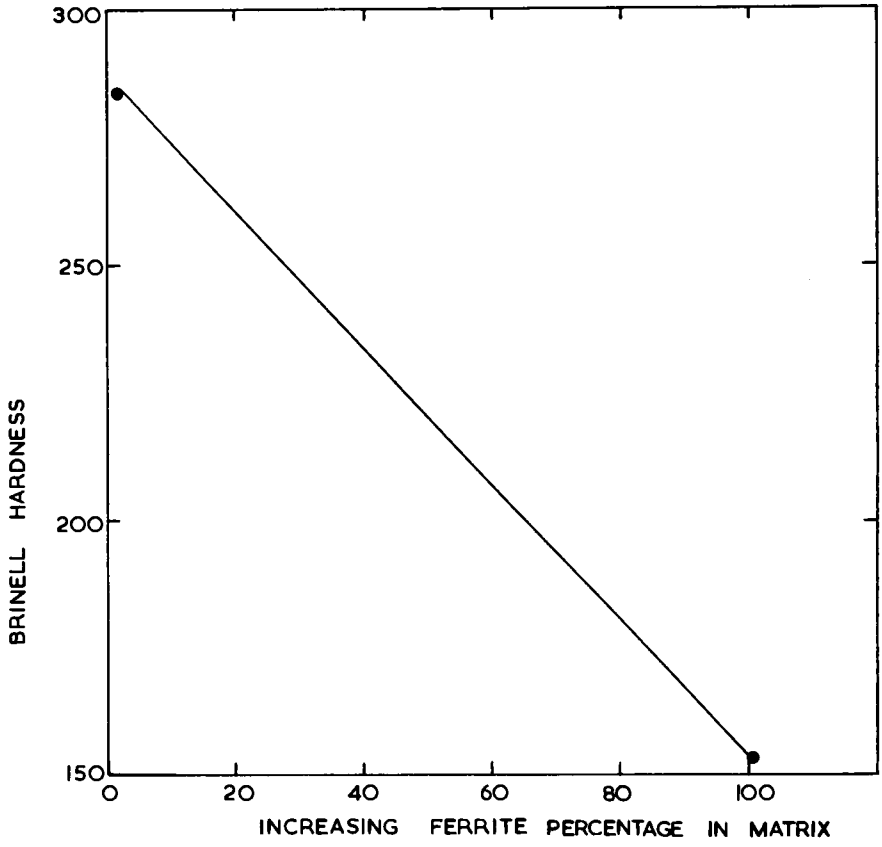


Figure 89. General relationship between hardness value and ferrite content of matrix

Increasing the amount of silicon, manganese or alloying elements such as copper, nickel and molybdenum increases the base hardness level owing to strengthening of the ferrite or to a refinement of the pearlite. The typical effect of silicon in a ferritic nodular iron is shown in *Figure 88* and that of the matrix structure in *Figure 89*.

FATIGUE LIMIT

The fatigue limit of nodular iron generally increases with increasing tensile strength, the endurance ratio varying according to the type of heat treatment carried out, and the matrix structure involved. *Table 84* shows results which have been reported by BCIRA employing the Wöhler type fatigue test.

In the ferritic condition, increasing silicon increases the unnotched fatigue limit and the dynamic stress concentration factor, but decreases the endurance ratio. The tensile strength of the as-cast pearlitic materials is reduced by increasing the silicon content, with little effect on fatigue limit, the endurance ratio thus being slightly higher at the upper silicon levels. Normalising to produce a refined pearlite structure gives increased tensile strength and fatigue limit, the endurance ratio, however, being lower than for the as-cast pearlitic matrix. Similarly, a quench and temper treatment to provide high tensile strengths gives fatigue strengths of the order of 22 ton/in^2 (340 N/mm^2) with a reduced endurance ratio compared with the lower strength as-cast and ferritic nodular irons.

IMPACT PROPERTIES

Like other non-austenitic ferrous materials, nodular iron exhibits the change from ductile to brittle failure as the temperature of testing is lowered. This effect is more apparent where the speed of loading is high and the specimen contains a notch, as in the commonly used notched impact test. Only the ferritic nodular irons have sufficient ductility to provide a ductile type of failure at and below room temperature with a notched impact test piece. Pearlitic and predominantly pearlitic materials have transition temperatures well above room temperature when notched test pieces are employed. The ferritic nodular irons must always be employed if shock loading conditions are to apply and using the standard 10 mm square notched Charpy test piece, values of 10–14 ft.lb (14–29 J) are obtained in the fully ductile condition. Although metallurgical factors have, in general, little effect upon this maximum level, they do markedly alter the temperature at which the change from ductile to brittle failure occurs. Some of these factors are:

(1) Metal Composition: Raising the level of elements such as silicon, manganese, phosphorus and nickel raises the impact transition temperature.
(2) Heat Treatment: Correct heat treatment procedure must be adopted if maximum ductility is to be obtained from the matrix. The cycle must

Table 84

Matrix structure	*Silicon content* %	*Tensile strength*		*Fatigue limit*				*Endurance ratio unnotched*	*Dynamic stress concentration factor*
				Unnotched		Notched			
		ton/in^2	(N/mm^2)	ton/in^2	(N/mm^2)	ton/in^2	(N/mm^2)		
Ferritic (heat-treated)	1·19	23·0	(355)	11·5	(178)	9·0	(139)	0·50	1·28
	1·87	26·9	(415)	13·5	(208)	8·5	(131)	0·50	1·59
	2·62	30·6	(473)	12·5	(193)	7·5	(116)	0·41	1·67
	3·21	35·0	(541)	14·5	(224)	8·0	(124)	0·41	1·81
Pearlitic (as-cast)	1·80	48·0	(741)	18·0	(278)	11·5	(178)	0·38	1·57
	2·47	42·7	(659)	17·0	(263)	8·0	(124)	0·40	2·13
	2·82	40·5	(625)	18·5	(286)	10·5	(162)	0·44	1·76
Pearlitic (heat-treated)	2·06	67·3	(1039)	22·0	(340)	13·5	(208)	0·37	1·63
Quenched and tempered (600°C)	1·96	60·1	(928)	22·0	(340)	13·5	(208)	0·37	1·63
Quenched and tempered (550°C)	1·96	66·7	(1030)	22·0	(340)	12·6	(195)	0·33	1·76

include a soaking at a temperature above the critical temperature range (880–920°C) to avoid a ferrite sub-grain boundary structure, and a slow enough cooling rate through the critical temperature range or a sub-critical hold, to ensure a fully ferritic structure.

The above factors are interrelated, having a cumulative effect upon the transition temperature, and must be considered by the producer when meeting the fully ductile grade SNG 24/17 of BS 2789:1961.

By the correct choice of composition and heat treatment, completely ductile failures can be obtained down to about −20°C with mean impact values in the transition range down to −50°C. Although the maximum notched impact value for nodular iron, 10–14 ft.lb (14–19 J), is far less than for some of the cast or wrought ferritic steels, the transition temperatures are lower, making the nodular iron less prone to failure in a brittle manner.

ELEVATED TEMPERATURE PROPERTIES

Short-time tensile tests show very little decrease in strength at temperatures up to about 400°C, although Palmer[5] has shown that the type of base material employed can provide some variation in the results when pearlitic structures are employed. Typical values for pearlitic and ferritic materials are shown in *Table 85*.

Although the pearlitic iron had the highest creep resistance at 426°C, its structural instability at the higher temperature of 538°C gave poor long-term load-bearing properties, being inferior to the ferritic matrix (*Table 86*).

The creep resistance of an as-cast pearlitic nodular iron at 400°C has been evaluated by Towers[7], the results being shown in *Figure 90*. These show that the stresses to produce 0·1, 0·2 and 0·5 per cent strain in 1000 h are respectively 9·5, 11·7, and 15·0 ton/in^2 (147, 181 and 232 N/mm^2) and compare with 5·5 and 6·8 ton/in^2 (85 and 105 N/mm^2) for 0·1 and 0·2 per cent strain respectively when employing grey cast iron[8].

TEMPER EMBRITTLEMENT

Nodular iron may become embrittled[9] with certain metal compositions when quenched from a temperature of 400–650°C if it has been previously slowly cooled from 650°C. This embrittlement or raising of the transition temperature is at a maximum when the quenching temperature is 450–500°C, but can be removed by subsequent quenching from 650°C. The susceptibility of nodular iron to temper embrittlement is increased by additions of silicon and phosphorus, the latter being more dangerous, but the addition of molybdenum to a susceptible composition inhibits embrittlement.

In addition, a temper embrittlement similar to that found in quenched and tempered steels can also be produced in similarly treated nodular irons. This embrittlement occurs on slow cooling from the tempering temperature, but can be suppressed by using a water quench treatment. The effect is only

Table 85

Temperature of test °C	*Pearlitic matrix*					*Ferritic matrix*				
	0.1% proof stress		*Tensile strength*		*Elongation*	*0·1% proof stress*		*Tensile strength*		*Elongation*
	ton/in²	(N/mm²)	ton/in²	(N/mm²)	%	ton/in²	(N/mm²)	ton/in²	(N/mm²)	%
Room	33·2	(513)	51·8	(800)	3	24·1	(372)	34·0	(525)	15
200	27·7	(428)	49·6	(766)	5	20·0	(309)	31·1	(480)	12
400	26·3	(406)	43·8	(676)	10	16·0	(247)	28·9	(446)	14
500	17·6	(272)	34·8	(537)	12	12·3	(190)	20·5	(317)	17
575	9·3	(144)	22·2	(343)	14	—	—	—	—	—
600	7·7	(119)	19·0	(293)	15	—	—	—	—	—
650	—	—	—	—	—	3·9	(60)	7·7	(119)	45

Table 86

Matrix structure	*Temperature of test* °C	*Limiting creep stress (for 0·0001%/h min. rate)*		*Limiting rupture stress* ton/in² (N/mm²)		
		ton/in²	(N/mm²)	10 h	100 h	1000 h
Pearlitic (normalised) tensile strength 59·6 ton/in² (920 N/mm²)	426	8·48	(131)	29·0 (448)	23·2 (358)	18·75 (290)
	538	0·78	(12)	10·6 (164)	7·6 (117)	4·06 (63)
	649	0·21	(3)	3·17 (49)	1·83 (28)	1·07 (17)
Ferritic, tensile strength 29·2 ton/in² (451 N/mm²)	426	6·25	(97)	16·5 (255)	13·8 (213)	11·2 (173)
	538	1·79	(28)	5·8 (90)	4·45 (69)	3·35 (52)
	649	0·25	(4)	2·23 (34)	1·47 (23)	0·98 (15)

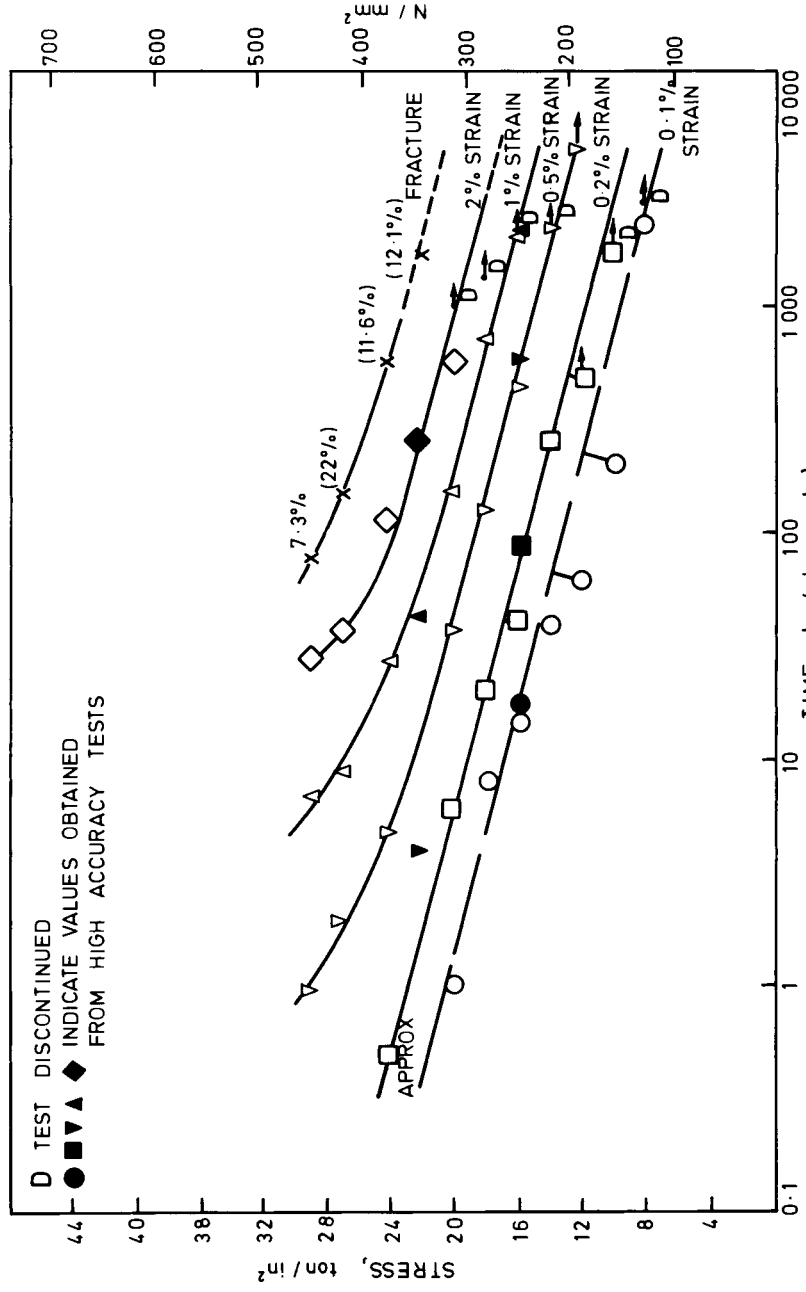


Figure 90. Creep tests at 400°C pearlitic nodular cast iron (Towers)

associated with irons of relatively high silicon and phosphorus contents, 0·5 and 0·1 per cent respectively. However, even with the quenching of a susceptible composition, the impact transition temperature is higher, because of the greater amounts of silicon and phosphorus, and there is thus a double advantage in using low silicon and phosphorus contents.

LOW TEMPERATURE TENSILE PROPERTIES[10]

At low temperatures the tensile strength of pearlitic nodular irons decreases rapidly to a value of about 36 ton/in^2 (556 N/mm^2) and the elongation before failure is reduced to zero. With ferritic nodular iron, however, the tensile strength rises progressively as the temperature falls to −269°C, although there is a marked reduction in elongation at these temperatures as shown in *Table 87*.

Table 87

	Tests at 20°C			*Tests at −269°C*		
	Tensile strength ton/in^2	(N/mm^2)	*Elongation* %	*Tensile strength* ton/in^2	(N/mm^2)	*Elongation* %
Pearlitic nodular iron (as-cast)	48·3	(746)	5	36·3	(561)	—
Ferritic nodular iron (annealed)	25·9	(400)	27	42·4	(655)	1·4

General properties of nodular cast iron

DAMPING CAPACITY

Damping capacity of nodular iron is intermediate between that of flake graphite cast iron and steel.

HEAT RESISTANCE

When flake graphite cast iron is exposed to elevated temperatures, growth of the material occurs by a breakdown of the combined carbon and by oxide penetration down the graphite flakes. The growth from the first cause is limited, but the latter is continuous with time. In the case of nodular iron, heating also produces breakdown of the combined carbon, but the spheroidal shape of the graphite provides a much lower surface area available for oxidation to occur, with a consequent reduced rate of growth.

Tests have shown that over an exposure period of 6 years, no growth occurs at 350°C and even at 400°C growth (less than 0·02 per cent) can be prevented either by prior annealing to the ferritic condition, or alloying the nodular iron with small amounts of pearlite stabilisers such as chromium,

molybdenum and tin. At higher temperatures, higher growth rates must be anticipated and typical values are given in Chapter 4.

WEAR RESISTANCE

The complexity of the factors causing wear, and the many variables in the conditions found in different applications, limit the value and dependability of laboratory test results. A summary of some recent Russian research showed that in the case of 'dry' conditions, wear increases directly with increasing ferrite content of the matrix, and conversely there is increasing wear resistance with more pearlitic structures. In the latter case, there appears to be little or no difference between as-cast structures and those obtained after normalising, although wear falls sharply after normalising nodular iron containing approximately 20 per cent of ferrite in the as-cast state. It was said that nodular iron having a substantially pearlitic matrix (ferrite up to 8 per cent) had a greater wear resistance than grey irons.

When considering lubricated conditions, nodular iron gives a satisfactory running-in performance at least comparable with a steel or cast iron.

FLAME-HARDENING

Although flake graphite cast iron can be flame-hardened, the higher tensile strengths and toughness associated with nodular iron would be expected to make the material more amenable to the hardening techniques. This has been borne out in practice and by correct choice of materials and methods hardnesses up to Rockwell 64 may be expected.

The structure of nodular iron may vary from pearlitic, through a range containing varying amounts of pearlite and ferrite, to a completely ferritic matrix, the type depending upon chemical composition, section size and heat treatment. As the carbon content of the pearlite goes into solution much more readily than do graphite nodules when the iron is heated to a temperature above the critical, it has been found that high hardness levels are much more readily acquired by the predominantly pearlitic materials.

Tests carried out on 1·875 in (48 mm) diameter pearlitic test bars hardened at temperatures ranging from 775 to 880°C have produced the hardness levels shown in *Table 88*.

Table 88

Hardening temperature °C	*Hardness 0·015 in (0·38 mm) from surface* VPN	*Depth of martensite zone* in	(mm)
775	698	0·078	(1·98)
800	690	0·073	(1·85)
850	707	0·085	(2·16)
880	669*	0·090	(2·29)

*This increases to 730 at a distance of 0·070 in (1·78 mm) from the surface.

Hardness tests carried out on ferritic nodular iron bars, 1·625 in (41 mm) in diameter did not produce a completely martensitic zone and this resulted in the relatively low hardness levels given in *Table 89*.

Table 89

Temperature of treatment °C	*Hardness at edge* BHN
825	255
850	283
880	294

Higher hardening temperatures and longer heating times or double heat treatment will influence the amount of carbon in solution and will probably produce higher hardness levels in the ferritic irons.

The most satisfactory type of iron to use for single pass flame or induction hardening is, therefore, the pearlitic type, free from carbides. It is also preferable to stress-relieve the castings at a temperature between 850 and 875°C before flame hardening.

CORROSION RESISTANCE

When attempting to evaluate the performance of nodular iron in corrosive liquids or atmospheres, the general rate of corrosion may be taken as similar to that of flake graphite cast iron. By alloying with nickel, an austenitic matrix may be obtained which greatly improves the corrosion resistance to acids, industrial waters and so on.

MACHINING OF NODULAR IRON

When comparing the machinability of nodular iron with other materials, and especially with flake graphite cast iron, the microstructure and the mechanical properties of the materials must be taken into consideration. By various heat treatments nodular iron may be produced with pearlitic, pearlitic–ferritic, and ferritic matrices, the machinability being lower with the former irons than with the latter type. This trend is, of course, usual in that improved machinability is generally associated with low mechanical properties. Experience shows that as-cast predominantly pearlitic nodular irons have approximately the same machinability as cold-drawn mild steel, but in the annealed ferritic state their machinability is at least as good as that of grey irons of comparable hardness.

The type of chip produced on machining nodular iron varies with the type of matrix, being much more ductile in the case of the ferritic matrix and resembling that produced when machining steel. This does not normally produce difficulties, except when deep holes are being machined and there is a risk of clogging the drill.

Table 90

	Turning		*Drilling*		*Tapping*		*Planing*		*Milling*	
	High-speed steel tools ft/min (m/s)									
Ferritic	50–150	(0·25–0·76)	80–130	(0·41–0·66)	20–30	(0·10–0·15)	40–100	(0·20–0·51)	50–125	(0·25–0·64)
Pearlitic–ferritic	40–90	(0·20–0·46)	50–100	(0·25–0·51)	15–20	(0·08–0·10)	30–75	(0·15–0·38)	35–65	(0·18–0·33)
Pearlitic	40–90	(0·20–0·46)	50–70	(0·25–0·36)	15–20	(0·08–0·10)	30–75	(0·15–0·38)	25–65	(0·13–0·33)
	Cemented carbide tools ft/min (m/s)									
Ferritic	175–400	(0·89–2·03)							200–400	(1·02–2·03)
Pearlitic–ferritic	100–300	(0·51–1·52)							175–350	(0·89–1·78)
Pearlitic	100–300	(0·51–1·52)							175–350	(0·89–1·78)

Table 90, due to International Nickel Ltd., gives machining speeds for three types of nodular iron.

Welding of nodular cast iron

The superior strength and ductility possessed by nodular iron, compared with normal flake graphite iron, make the material more suitable for welding, both as a means of reclamation of castings and of joining two or more castings together. However, the presence of a high total carbon content produces a risk of obtaining both carbides and martensite during the welding process and steps must be taken to reduce the amount of these phases.

ARC WELDING (MANUAL)

The 55 per cent nickel-iron electrode is often recommended for both reclamation and fabrication work because the strength of the weld metal is comparable in strength and toughness with that of the base iron. Although some castings of simple shape can be welded without preheat, it is generally necessary to preheat the ferritic grade of nodular iron to 120–180°C and the pearlitic grade to 200–300°C if cracking in the heat-affected zone is to be avoided. A post-weld heat treatment, full annealing for the ferritic grade and normalising (air-cool from 850–900°C) is essential if the maximum mechanical properties are to be realised from the heat-affected zone. Repair work should, wherever possible, be carried out before the usual foundry heat treatment.

The mechanical properties of the welded joints produced using these electrodes are not normally as good as those of the base metal, even after post-weld heat treatment, and this applies especially to the impact properties of the fully ferritic materials. For example, butt welds have been reported to be approximately 66 per cent of the base metal strength for a ferritic iron, and have approximately 33 per cent base metal ductility in the as-welded and annealed conditions[11].

METAL INERT GAS (MIG) FINE WIRE WELDING PROCESS

This process[12] can be used economically to repair and join nodular iron castings and the lower heat input involved often permits the ferritic grades to be welded without preheat. This may also apply to the pearlitic grades if high welding stresses are not anticipated. Post-weld heat treatment may not be necessary, but this omission depends upon the complexity of the casting design and also on the level of properties required after welding. Excellent as-welded tensile strength can be obtained[13], and with the ferritic grades the elongation values recorded are normally well in excess of those obtained from welded and fully ferritised specimens joined with nickel-iron stick electrodes.

GAS FUSION WELDING

This method normally employs nodular iron filler rods containing higher than usual levels of magnesium and cerium to offset losses during welding, and is principally employed for the reclamation of defective castings where the weld must have mechanical properties and corrosion resistance matching as closely as possible those of the casting.

Although, because of the higher overall thermal input of this type of welding, preheat is not essential, the presence of eutectic carbide in the weld necessitates the use of post-weld heat treatment if it is to be readily machinable and also if the maximum properties, especially ductility, are to be realised. Normalising or annealing treatments are employed appropriate to the final condition of the casting.

With good welding techniques, strong welds can be obtained with pearlitic nodular irons which usually have tensile properties comparable with those of the base iron. The tensile strengths of post-weld annealed irons are also comparable with those of the base iron and although the ductility of the weld zone is impaired, the elongation values obtained are similar to those recorded on annealed irons following manual metal arc welding with nickel-iron electrodes.

POWDER WELDING

This is a technique[14] which is capable of producing joint strengths approaching or equivalent to those of the base iron without fusion of the base iron, thereby avoiding carbide formation which is prevalent in manual arc and gas fusion welding. Preheat temperatures up to 250°C are beneficial as this avoids the formation of martensite during cooling. If this, and careful cooling of the casting after welding, are exercised, there is no need to carry out post-weld heat treatments other than for stress relief. Joint strengths of welded pearlitic nodular irons are comparable to those obtained by manual arc welding using nickel-iron electrodes. With ferritic irons the tensile strength of the welded area is equivalent to that of the base iron, but the ductility is impaired somewhat as with arc welding.

Applications of nodular cast iron

Since its introduction in 1947, nodular cast iron has found numerous applications throughout all fields of industry, but these uses are increasing as the properties and possibilities of the material are more fully understood. Nodular irons should not be regarded as being a universal replacement for any other particular type of material; its application and design should be judged from its characteristic properties:

(1) Its excellent fluidity, enabling it to be cast into intricate shapes.
(2) Its nodular graphite structure, which can be produced in an almost

unlimited range of section sizes with very little falling off in mechanical properties.

(3) Its ability to be heat treated to give a range of mechanical properties which vary from a tensile strength of 60 ton/in^2 (927 N/mm^2) and very little shock resistance, to a highly shock resistant material having tensile strengths of about 25 ton/in^2 (386 N/mm^2).

(4) Its good wear resisting properties, which can be further enhanced by a surface hardening treatment.

(5) Its corrosion resisting properties, superior to those of low carbon cast steels.

(6) Its resistance to growth and scaling at elevated temperatures, which is much superior to that of flake graphite cast iron.

Cost of nodular iron

The earlier methods of production of nodular iron, although still widely used, involved the addition of a nickel–magnesium alloy, and in certain instances the addition of up to 1·5 per cent nickel. This increased the cost of the material considerably, but present-day treatment methods are obviating the necessity to add nickel and so treatment costs show a downward trend.

At present the cost of nodular iron generally exceeds that of malleable iron, and in some cases is equal to that of cast steel. This high cost is due to:

(1) The use of good quality raw materials with low phosphorus contents.

(2) The nodularising treatment. This will vary considerably, depending upon the type of melting furnace used and the method used for effecting the magnesium addition.

(3) The low casting yield obtained. This is generally only of the order of 50 per cent compared with the 80 per cent that may be anticipated in the production of flake graphite cast iron.

(4) The cost of heat treatment. Although the time cycle is short compared with malleable iron, it is found that the cost of heating the castings up to treatment temperature is a major part of the overall cost of the heat treatment process.

(5) The cost of control required to ensure that the correct structure is obtained in castings.

Austenitic nodular iron

By the use of alloy additions, principally nickel, and to a lesser extent chromium and manganese, a range of nodular irons having austenitic matrix structures may be produced. In addition to the excellent corrosion and heat resistance properties which are associated with the austenitic matrix, much higher tensile strengths and elongations are obtainable than with flake graphite austenitic irons. The alloy element content is adjusted to suit the service conditions and the appropriate British Standard, 3468:1962, specifies four basic types of composition, as follows:

Table 91

Type: AUS	Analysis range % TC	Si	Mn	P	Ni	Cr	Tensile strength ton/in²	(N/mm²)	Elongation %	HB (max.)	V-notch impact ft.lb	(J)
202A	3·0 max.	0·60–1·5	0·70–1·5	0·080 max.	18·0–22·0	1·0–2·50	24	(371)	8	201	—	—
202B	3·0 max.	1·0–2·80	0·70–1·5	0·080 max.	18·0–22·0	2·0–3·50	24	(371)	6	255	—	—
203	3·0 max.	1·0–2·80	1·80–2·40	0·080 max.	21·0–24·0	0·5 max.	24	(371)	20	170	20	(27)
204	3·0 max.	4·5–5·5	1·0–1·5	0·080 max.	18·0–22·0	1·0–2·5	24	(371)	10	230	—	—
205	2·6 max.	1·5–2·8	0·5 max.	0·080 max.	28·0–32·0	2·5–3·5	24	(371)	7	201	—	—

AUS Type 202 is normally selected for general corrosion resistant applications and Types 204 and 205 where maximum heat resistance is required. The latter has a lower thermal expansion and should be selected where severe thermal cycling is involved.

The high ductility of Type 203 makes it particularly suitable for applications involving a combination of corrosion resistance and toughness, especially at low temperatures (down to −100°C) where the drop in notched impact value associated with the ductile-to-brittle transition of the ferritic nodular irons is not encountered. By increasing the manganese content of Type AUS 203 to about 4 per cent, good notch ductility can be assured to temperatures as low as −196°C (International Nickel Ltd., material designated D2M).

All of these irons have low magnetic permeabilities, but where this property is important the AUS Type 203 or the high manganese Nodumag* materials should be selected.

*Trade Mark, Ferranti Ltd.

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AUSTENITIC CAST IRON

The principal properties governing the use of austenitic cast iron are:

(1) Corrosion resistance (Ni-Resist type)*.
(2) Low expansion (Minovar type)*.
(3) Non-magnetic properties (Nomag type)*.
(4) High temperature resistance (Nicrosilal type)*.

In this group of irons the γ to α transformation is completely suppressed at room temperature by the addition of nickel, copper or manganese, or combinations of all three elements[1]. These elements tend to move the time/temperature transformation curves (*see* section on heat treatment, *Figure 115*) to the right, producing first of all martensitic structures and then, with sufficient addition of alloy, the austenitic structure remains stable down to almost any desired temperature. The Nehrenburg formula generalises the effect of the elements on the martensitic change point (*Ms*). *See also* p. 220.

$$Ms\ (^{\circ}C) = 500 - 300\ \text{carbon} - 33\ \text{manganese} - 22\ \text{chromium} - 17\ \text{nickel} - 11\ \text{silicon} - 11\ \text{molvbdenum}$$

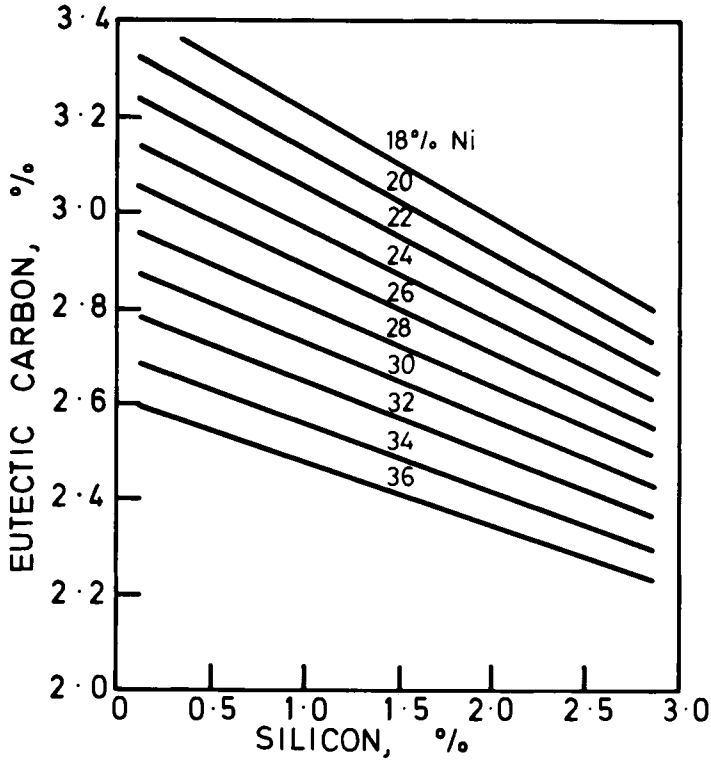
where the elements are the percentages present in solution in the austenite. If, therefore, the alloy content is adjusted to depress the *Ms* point below room temperature, the matrix will remain austenitic. Intermediate amounts of alloy addition may produce acicular or martensitic structures at room temperature. These structures are dealt with in other sections of this book. The carbon content of austenite with 12 per cent nickel appears to be about 0·6 per cent, and for 24 per cent nickel about 0·2 per cent[2]. The eutectic carbon content decreases with increasing nickel and silicon contents (*Figure 91*). The influence of silicon decreases with increasing nickel content. The following formula has been put forward[3]:

$$\text{Carbon Equivalent} = 4{\cdot}3\% - 0{\cdot}33\ (\%\ \text{Si}) - 0{\cdot}047\ (\%\ \text{Ni}) + 0{\cdot}0055\ (\%\ \text{Ni}) \times (\%\ \text{Si}) \pm 0{\cdot}10\%$$

It is important to bear in mind that austenitic structures produced by alloying are fully stable over a wide range of temperatures, provided the alloy content is adequate, but an iron that contains just sufficient nickel, copper or manganese to make it austenitic at room temperature may transform to martensite at lower temperatures; castings required for use at sub-zero

*These are trade names: (1) and (2) International Nickel Ltd.; (3) Ferranti Ltd.; (4) BCIRA

temperatures must have a sufficiently high alloy content to prevent transformation at their lowest working temperature, otherwise they will become brittle and remain so when the temperature is raised to normal. Cold work may initiate the austenite–martensite change at room temperature even with nickel up to 25 per cent if the manganese is below 0·6–0·7 per cent, and may increase the machining difficulties.



$$CE = 4 \cdot 30\% - 0 \cdot 33\,(\%\,Si) - 0 \cdot 047\,(\%\,Ni) + 0 \cdot 0055\,(\%\,Ni)\,(\%\,Si) \pm 0 \cdot 16\%$$

Figure 91. Eutectic carbon content with increasing silicon and nickel contents[3]

Corrosion-resisting austenitic iron (Ni-Resist)

Much of the information in this section is taken from a paper by Barton[4].

In the more common types of corrosion-resisting irons the copper content varies between 5·5 and 7·5 per cent, but in some instances, where freedom from copper is essential (e.g. in the handling of certain foodstuffs), no copper is used and the nickel content increased. The composition of the main types of flake graphite austenitic cast iron to BS 3468:1962 is given in *Table 92*.

Where the presence of copper offers advantages in corrosion resistance, AUS 101 A and B are recommended.

Table 92 AUSTENITIC FLAKE GRAPHITE CAST IRONS BS 3468:1962

Grade	*AUS 101*				*AUS 102*				*AUS 104*		*AUS 105*	
	A		*B*		*A*		*B*					
Total carbon	—	3·0	—	3·0	—	3·0	—	3·0	1·6	2·2	—	2·60
Silicon	1	2·8	1	2·8	1·0	2·8	1·0	2·8	4·5	5·5	1·0	2·0
Manganese	1	1·5	1	1·5	1·0	1·5	1·0	1·5	1·0	1·5	0·4	0·8
Nickel	13·5	17·5	13·5	17·5	18·0	22·0	18·0	22·0	18·0	22·0	28·0	32·0
Copper	5·5	7·5	5·5	7·5	—	0·5	—	0·5	—	0·5	—	0·50
Chromium	1·0	2·5	2·0	3·5	1·0	2·5	2·0	3·5	1·8	4·5	2·5	3·5
Tensile, minimum, ton/in^2 (N/mm^2)	9	(139)	12	(185)	9	(139)	12	(185)	12	(185)	11	(170)
Hardness, max.	212		248		212		248		248		212	
Elongation, %, min.	2		—		2		—		2		—	

For handling caustics, food etc., where copper contamination cannot be tolerated, AUS 102 and 104 are recommended.

The higher chromium levels are said to give better machinability[6].

AUS 105 has a coefficient of expansion of about $6–12{\cdot}4 \times 10^{-6}$/°C. (*See also* section on low expansion alloys.) The coefficient of thermal expansion of Grades 101–104 is about 18×10^{-6}/°C.

SILICON

As Ni-Resist has an austenitic matrix, it is less susceptible to silicon changes than the normal unalloyed grey iron. However, the silicon content should be adjusted in conjunction with the carbon content to give an iron machinable in thin sections and yet dense in the heavier sections. International Nickel Ltd. publishes the following (*Table 93*) as a guide in choosing the correct silicon content:

Table 93

Carbon content	*Section thickness*	*Silicon range*
2·6–2·9%	Under $\frac{1}{2}$ in (13 mm)	2·5–2·25%
	$\frac{1}{2}$–1 in (13–25 mm)	2·25–1·75%
	1–2 in (25–51 mm)	1·75–1·5%
	Over $\frac{1}{2}$ in (13 mm)	Under 1·5%
2·9–3·1%	Under $\frac{1}{2}$ in (13 mm)	2·25–2·00%
	$\frac{1}{2}$–1 in (13–25 mm)	2·00–1·5%
	1–2 in (25–51 mm)	1·5–1·2%
	Over 2 in (51 mm)	Under 1·2%

The corrosion resistance of these alloys is dealt with in the section on corrosion resistance. The importance of an adequate alloy content to ensure stability at low temperatures is illustrated in *Table 94* taken from data by Gilbert[5].

Table 94

TC %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Cr* %	*Cu* %	*Temperature of phase change* °C
2·95	2·64	0·60	0·025	0·035	13·93	1·06	6·36	−10
2·91	2·44	0·57	0·022	0·031	15·55	1·17	6·17	−50
2·65	2·43	0·48	0·016	0·051	15·52	1·41	8·65	−88
2·58	2·36	0·98	0·122	0·080	15·55	1·61	8·02	−120
2·42	1·85	0·98	0·086	0·062	15·50	1·72	7·06	−175

For use at sub-zero temperatures, therefore, it is desirable that the manganese is not less than 0·9 per cent, nickel not less than 15·5 per cent, and copper not less than 7 per cent. In a copper-free alloy the nickel should be not less than 22 per cent. Although the Nehrenburg formula indicates that chromium has a greater effect than nickel, in actual practice anything above about 1 per cent chromium forms carbides and is not in solution in the austenite.

PHOSPHORUS

Although phosphorus is usually kept below 0·4 per cent to avoid shrinkage and brittleness, for bearings and cylinder liners the wear resistance is increased by a phosphorus content up to 1 per cent, which gives a continuous network of steadite. Phosphorus up to 1 per cent has no adverse effect on corrosion resistance.

Lead, tin and zinc are deleterious and may easily be present in non-ferrous scrap. Lead, in particular, in quantities as small as 0·0007 per cent will produce an unfavourable form of graphite known as Widmanstätten graphite, which materially reduces the strength of the alloy.

MICROSTRUCTURE

The normal structure for a Ni-Resist consists of flake graphite distributed throughout an austenitic matrix, small amounts of carbide and phosphide also being present. If a high strength iron is required it is an advantage to inoculate the molten metal with about 0·5 per cent silicon, added as ferrosilicon, which results in the production of graphite in random form.

Where maximum resistance to heat and corrosion is required, the graphite should be in the undercooled state. This may be obtained by the addition of ferrosilicon–titanium or nickel–titanium (29 per cent titanium, 5 per cent aluminium, balance nickel) to the ladle. Alternatively, a pig iron having a high titanium content may be used.

The microstructure of a Ni-Resist where the austenite→martensite transformation has occurred is shown in *Figure 28*. Although in this particular case it is due to a low temperature treatment, similar structures are observed where there is an insufficient nickel or nickel plus copper content to provide complete stabilisation of the austenite.

Table 95 PHYSICAL PROPERTIES OF NI-RESIST[7]

Specific heat:	0·11–0·12 cal/g (at 20°C) (461–502 J/kg.°C)
Specific gravity:	7·36
Thermal conductivity:	0·095 cal/cm^2/cm/s/°C (40 W/m.°C)
Thermal expansion:	The values quoted for Grades AUS 101 and 102 are 18·7–19·3 × 10^{-6}/°C; Grade AUS 105 12·4 × 10^{-6}/°C; 35% Ni 5·0 × 10^{-6}/°C
Electrical resistivity:	140–170 microhm.cm (1·4–1·7 $\mu\Omega$.m)
Magnetic response:	AUS 101 A and B and 102 A non-magnetic, AUS 102 B and 105 magnetic

The figures given below are those quoted by Gilbert[5]:

Composition	*Sample No. 1* %	*Sample No. 2* %	*Sample No. 3* %
Total carbon	2·95	2·91	2·65
Silicon	2·64	2·44	2·43
Manganese	0·60	0·57	0·48
Sulphur	0·025	0·022	0·016
Phosphorus	0·035	0·031	0·051
Nickel	13·93	15·55	15·52
Chromium	1·06	1·17	1·41
Copper	6·36	6·17	8·65
Coefficient of thermal expansion, × 10^{-6}, 0–200°C	18·2	18·0	17·7
Immersed liquid O_2, tested 0–200°C	12·4	13·3	14·5

The maximum value is found at nickel contents of 15–20 per cent. Further increases in nickel content cause a decrease in thermal expansion, until at 35 per cent nickel a material with a very low value is obtained. *See Figure 92.*

Electrical resistivity:	130 microhm/cm^3 (1·3 $\mu\Omega$.m)
Temperature coefficient of resistance:	0·000 318 between 20°C and 430°C

Magnetic response[8]:

Composition	%
Total carbon	2·72
Silicon	2·41
Manganese	1·09
Sulphur	0·020
Phosphorus	0·042
Nickel	15·5
Chromium	1·81
Copper	5·17
Maximum permeability	2·1

MECHANICAL PROPERTIES OF NI-RESIST

Transverse strength Austenitic irons have a lower transverse strength and higher deflections than the pearlitic type of grey iron.

Modulus of rupture, ton/in^2	24–28 (371–432 N/mm^2)
Deflection, in	0·3–0·6 (7·6–15·2 mm)

The above figures are for an iron with a total carbon content of 2·44 per cent and a relatively high chromium content of 3·30 per cent.

Compressive strength[6] Compressive strength = 44·5–53 ton/in^2 (687–819 N/mm^2)

Torsional strength[6] Torsional strength = 15·5–18 ton/in^2 (239–278 N/mm^2)

Elastic modulus[6] Elastic modulus = 12–14 × 10^6 lb/in^2 (83–97 GN/m^2)

The elastic modulus values are taken at 25 per cent of tensile strength.

Tensile strength This value ranges from 9·0–15·5 ton/in^2 (139–239 N/mm^2) when tested at room temperature, the actual figure being dependent upon the composition and microstructure. Eash[10], in determining the effect of ladle inoculation on austenitic cast iron, showed that low strengths usually exist in irons having a dendritic graphite and carbide arrangement, whereas high quality materials possess a more random distribution of these constituents.

Gilbert[5] carried out a series of tensile tests at sub-atmospheric temperatures:

Composition:	%
Total carbon	2·95
Silicon	2.64
Manganese	0·60
Sulphur	0·025
Phosphorus	0·035
Nickel	13·93
Chromium	1·06
Copper	6·36

Testing temperature	*Tensile strength*	
°C	ton/in^2	(N/mm^2)
16	9·1	(141)
−4	8·3	(128)
−20	9·7	(150)
−45	12·5	(193)
−55	12·6	(195)
−83	16·3	(252)
−100	17·2	(266)

The structural change occurred at approximately −10°C, increasing amounts of martensite being observed with further lowering in temperature.

Tapsell, Becker and Conway[9] carried out a series of short-time high temperature tensile tests on a material of the following composition:

	%
Total carbon	2·44
Silicon	1·11
Manganese	0·76

	%
Sulphur	0·057
Phosphorus	0·26
Nickel	16·56
Chromium	3·3
Copper	7·3

Testing temperature °C	*Tensile strength* ton/in²	(N/mm²)
air	11·3	(175)
450	10·2	(158)
538	9·6	(148)

Compared with Nicrosilal, Ni-Resist shows a lower rate of decrease in strength with temperature, but inferior elongation values.

Hardness The usual range of Brinell hardness numbers quoted for Ni-Resist is 120–248, the actual figure being dependent upon the composition, mainly the carbon and chromium contents.

The hardness of the chromium-free material is about 100 on the Brinell scale, this being increased by about 20 points for each 1 per cent chromium added.

Hardness at sub-atmospheric temperatures is dependent upon the temperature at which the structural change occurs for each individual material. Figures are given by Gilbert[5] for an iron of composition:

	%
Total carbon	2·91
Silicon	2·44
Manganese	0·57
Sulphur	0·022
Phosphorus	0·031
Nickel	15·55
Chromium	1·17
Copper	6·17

Temperature °C	*BHN (tested at 16°C after low temp. treatment)*
16	86
−3	<90
−21	<90
−43	95
−59	122
−75	157
−114	195
−141	219

The temperature at which the structural change occurred was approximately −50°C. This change, austenite→martensite, is irreversible unless the material is reheated to the stable austenite range. The hardness values at room temperature are unusually low.

Impact value On account of variations in test pieces and procedures used there is not a great amount of useful information available on resistance to shock. Unnotched Izod tests (BS 1349) have shown a ratio of austenitic to pearlitic iron of approximately 8:1 in shock strength.

The unnotched Charpy impact value, using a $\frac{5}{8}$ in (16 mm) diameter specimen and a 3 in (76 mm) span, is approximately 16 ft.lb (22 J).

When tested at low temperatures[5] there is a considerable rise in Charpy impact value with decrease in temperature as the phase change austenite →martensite is approached. At temperatures below the phase change the impact strength falls off rapidly.

Table 96 shows the effect of low temperatures on impact strength using a material of the following composition:

	%
Total carbon	2·91
Silicon	2·44
Manganese	0·57
Sulphur	0·022
Phosphorus	0·031
Nickel	15·55
Chromium	1·17
Copper	6·17

Table 96

Temperature °C	*Tested at temperature* ft.lb	(J)	*Tested at 16°C after low temp. treatment* ft.lb	(J)
16	5·7	(8)	—	—
−3	7·2	(10)	—	—
−21	8·9	(12)	6·1	(8)
−43	8·4	(11)	4·8	(7)
−59	7·4	(10)	6·0	(8)
−75	6·0	(8)	4·0	(5)
−114	4·5	(6)	3·2	(4)
−141	3·1	(4)	2·8	(4)

The temperature at which structural change occurred was approximately −50°C.

Creep Tapsell, Becker and Conway[9] carried out creep tests (*Table 97*) on an iron of the following composition:

	%
Total carbon	2·44
Silicon	1·11
Manganese	0·76
Sulphur	0·057
Phosphorus	0·26
Nickel	16·56
Chromium	3·3
Copper	7·3

Table 97

Temperature °C	*Stress applied* ton/in²	(N/mm²)	*Duration of test days*	*Final strain % on 2 in (51 mm)*	*Rate of strain, minimum, or at 40 days* in/in/h
450	8	(124)	42 U	1·04	0·000 001 8
„	6	(93)	40 U	0·57	0·000 000 9
„	4	(62)	40 U	0·32	0·000 000 4
„	3	(46)	44 U	0·13	0·000 000 3
„	Growth test		56	0·06	0·000 000 2
538	6	(93)	6 B	1·5	0·000 009 5
„	4	(62)	42 U	0·44	0·000 001 0
„	3	(46)	40 U	0·19	0·000 000 2
„	3	(46)	40 U	0·32	0·000 000 5
„	2	(31)	40 U	0·11	0·000 000 1
„	Growth test		41	0·01	0·000 000 0

B = Broken; U = Unbroken after 40 days.

The growth of this iron at 450°C exceeds the growth at 538°C, and thus the stress log rate of strain/h for the tests at 450°C crosses that of the test at 538°C. At about 3 ton/in² (46 N/mm²), the rate of strain at both temperatures is about the same, but at lower stresses the rate of strain at 538°C is lower than that at 450°C. This may be due to the more rapid graphitisation and the earlier attainment of a stable state at higher temperatures.

Table 98

	Composition %	*Testing temperature* °C	*Stress for a creep rate of 0·1% in 1000 h* lb/in²	(N/mm²)
Total carbon	2·44			
Silicon	1·11			
Manganese	0·76	450	14 000	(97)
Sulphur	0·06			
Phosphorus	0·26			
Nickel	16·56	540	8 900	(61)
Chromium	3·30			
Copper	1·30			

Table 99 FATIGUE STRENGTH[12]

	Composition %	*Properties*		
Total carbon	2·38			
Silicon	2·05	Tensile strength, ton/in²	14·8	(229 N/mm²)
Manganese	0·88	Elongation	4%	
Sulphur	0·072			
Phosphorus	0·06	Fatigue limit, unnotched, ton/in²	±7·0	(±108 N/mm²)
Nickel	17·45			
Chromium	1·5	Fatigue limit, V-notched, ton/in²	±6·0	(±93 N/mm²)
Copper	7·05			

Creep data[11] published by the American Society for Testing Materials and fatigue strength data are presented in *Tables 98* and *99*, respectively.

HEAT RESISTANCE

A comprehensive series of growth tests has been carried out by Hallett[13] on a range of heat-resisting cast irons, and the following results were obtained for Ni-Resist:

Table 100

						750°C, 300 h		*850°C, 300 h*		*950°C, 300 h*		*1050°C, 240 h*	
TC %	*Si* %	*P* %	*Ni* %	*Cu* %	*Cr* %	*Scale* g/dm²	*Growth* %	*Scale* g/dm²	*Growth* %	*Scale* g/dm²	*Growth* %	*Scale* g/dm²	*Growth* %
2·74	1·93	0·04	14·62	7·16	2·11	4	0·4	9	0·3	16	0·75	30–32	1·7
2·85	1·85	—	14·38	7·36	4·84	3	0·2	6	0·4	9	Nil	16–20	0·7

The growth figures were based on a 240 h exposure.

In these tests it was found that in its resistance to scaling Ni-Resist was inferior to Silal and Nicrosilal at temperatures up to 850°C. The higher chromium content, however, improved the scaling resistance by amounts ranging from one-third to three-quarters.

Tests were also carried out to check the above figures against those obtained under service conditions. Small base-plate castings were heated up to 1020°C. After 2 h at this temperature they were then cooled to 600°C and the cycle was then repeated. The high chromium Ni-Resist samples were severely scaled, cracked and distorted after 119 cycles.

For uses in steam service the maximum service temperature should be approximately 320°C. With higher temperatures, growth over long periods of time becomes apparent and the material should be used only after trials have been carried out.

MACHINABILITY

The major difficulty in machining Ni-Resist appears to be the tendency of the material to work-harden. Light rubbing cuts should therefore be avoided

Table 101

Type of tool	*Speed*		*Feed/rev*	
	ft/min	(m/min)	in	(μm)
Carbide turning	80–200	(24–61)	to 0·015	(381)
High speed steel:				
turning	30	(9)	0·015	(381)
drilling	35	(11)	0·002–0·005	(51–127)

and in order to produce a smooth finish in conjunction with good tool life a cutting lubricant should be used.

International Nickel Ltd. recommend the cutting speeds and feeds presented in *Table 101*.

With hardness figures in excess of 160, Ni-Resist becomes increasingly difficult to machine and the speeds should be accordingly reduced.

CORROSION RESISTANCE

(*See* section on corrosion)

USES OF NI-RESIST

The exceptional properties which Ni-Resist possesses over a normal unalloyed grey cast iron are:

(1) Its resistance to attack against corrosive media.
(2) Its good heat-resisting properties.
(3) The non-magnetic property of the austenitic matrix.

The uses to which Ni-Resist castings are put may be sub-divided according to which property is required, and typical examples are given below.

Non-magnetic castings Generator and covers, alternator castings, magnetic separator drum ends and pulleys.

Engine, pump and boiler castings Pump bodies and impellers, rotors and gears, superheater tube spacers, valve seatings and exhaust manifolds.

Furnace parts and heat-resisting castings The maximum temperature at which Ni-Resist castings are usually employed is 800°C, the higher chromium content being used for these high temperatures. In service, where there is appreciable sulphur in the atmospheres, the ceiling temperature is about 550°C. Typical uses are: firebars, stoker links, firedoors, annealing cradles and electric hot plates.

Corrosion-resisting castings Handling plant for the manufacture of chemicals and food, tar stills, sewage plant pipe and many applications in the petroleum industry.

Bearings and cylinder liners Ni-Resist is used quite extensively for cylinder liners where intermittent conditions are involved or where conditions of service are severe. As the material is not very susceptible to bedding-in, it is desirable that running-in should be carefully done and that full bearing loads should not be applied until a substantial running-in period has been allowed.

A high graphite content is desirable and a suitable composition is:

	%
Total carbon	2·7–3·0
Silicon	2·0–2·2
Manganese	0·8–1·0
Phosphorus	0·3–1·0
Nickel	12·0–16·0
Chromium	1·5–2·0
Copper	5·5–7·5

Low expansion alloy[14]

Although the normal austenitic steels and irons have a high coefficient of thermal expansion (approximately 18×10^{-6}/°C), the group of high nickel steels known as Invar* have the characteristic property of low thermal expan-

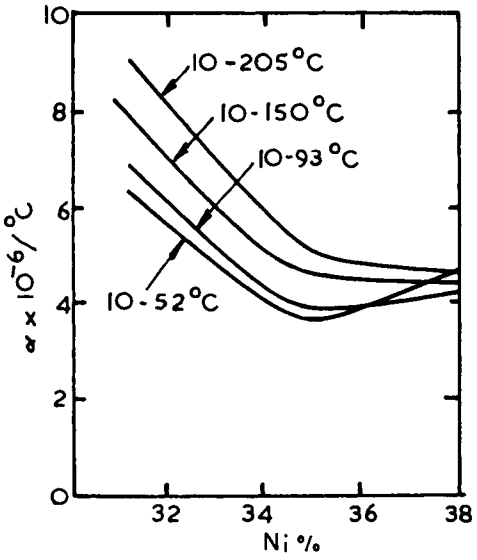


Figure 92. Coefficient of expansion (α) of high nickel cast iron

sion ($0{\cdot}4 \times 10^{-6}$/°C) within the temperature range 0–200°C. A similar group of alloys is available in cast iron and a range of coefficients of expansion (4–8×10^{-6}/°C) is available by varying the nickel content (*Figure 92*). The properties of the alloy with the lowest coefficient of thermal expansion (Minovar*) are given in *Table 102*.

* Trade name registered by International Nickel Ltd.

Table 102 PROPERTIES OF 35–36% AUSTENITIC NICKEL IRON LOW EXPANSION TYPE

Tensile	20–25 000 lb/in^2 (138–172 N/mm^2)
Compression	80–100 000 lb/in^2 (552–689 N/mm^2)
Torsion	30–35 000 lb/in^2 (207–241 N/mm^2)
Torsion modulus	4 500 000 lb/in^2 (31 GN/m^2)
Modulus at 25% tensile	10 500 000 lb/in^2 (72 GN/m^2)
Transverse	1800–2000 lb/in^2 (1·2 in) (30 mm : 46 mm) 12–14 N/mm^2
Deflection	0·6–0·9 in (15–23 mm)
Endurance limit	9900 lb/in^2 (68 N/mm^2)
Brinell	100–125
Impact	150 ft.lb (203 J) (compared with 25–35 (34–47 J) for grey iron)
Specific gravity	7·6
Density	0·275 lb/in^3 (7·6 g/cm^3)
Melting point	1230°C
Thermal conductivity	0·094 cal/cm^2/cm/s/°C (39 W/m°C)
Specific resistance	100–170 microhm/cm^3 (1–1·7 $\mu\Omega$.m)

Non-magnetic irons

The normal austenitic irons given in *Table 92* are almost non-magnetic, but by substituting manganese for some of the nickel a still lower magnetic permeability is obtainable. The properties of this material, Nomag*, are given in the section on magnetic properties. The martensite change point for a non-magnetic iron of this type, with an analysis of 3·56 per cent total carbon, 2·22 per cent silicon, 5·2 per cent manganese, 0·025 per cent sulphur, 0·026 per cent phosphorus, and 10·4 per cent nickel, was −166°C. It has the following properties: tensile strength, 9–14 ton/in^2 (139–216 N/mm^2); Brinell 140–190; magnetic permeability, 1·03–1·11 (1·29–1·39 μH/m); resistivity 130–150 ohm/cm^3 (1·3–1·5 $\mu\Omega$.m). A nickel-free variety[15] is said to give: tensile 9–14 ton/in^2 (139–216 N/mm^2); Brinell 230–280; magnetic permeability 1·1–1·3 (1·38–1·63 μH/m); specific resistance 130–180 ohm/cm^3 (1·3–1·8 $\mu\Omega$.m) (total carbon, 3·3–3·8 per cent; manganese, 7–12 per cent; copper up to 4 per cent).

Nicrosilal

HIGH TEMPERATURE RESISTANCE

Details of the creep and oxidation resistance of the Ni-Resist irons have already been given in this section, but a further range of austenitic alloys with a very high degree of oxidation resistance above 700°C was developed by the BCIRA, under the name of Nicrosilal (BS 3468:1962, AUS 104).

The approximate range of composition of Nicrosilal is: total carbon 1·6–2·2 per cent; silicon 4·0–4·5 per cent; manganese 0·6–1·2 per cent; nickel 18·0–22·0 per cent; chromium 2·0–4·0 per cent; phosphorus 0·1–0·5 per cent. Details

*Trade name registered by Ferranti Ltd.

of the high-temperature properties, strength and oxidation resistance are given in the section on heat resistance.

The mechanical and physical properties at room temperature are given in *Table 103.*

Table 103 PROPERTIES OF NICROSILAL

Tensile strength	11–18 ton/in^2 (170–278 N/mm^2)
Elongation	Up to 3% on 2 in (51 mm) gauge length
Brinell hardness number	140–190
Transverse rupture stress [0·875 in (22 mm) dia. as-cast, 12 in (305 mm) centres]	25–29·5 ton/in^2 (386–456 N/mm^2)
Deflection	0·49–1·0 in (12–25 mm)
Specific gravity	7·2–7·4
Coefficient of thermal expansion	18×10^{-6}/°C
Thermal conductivity at:	
100°C	0·063 cal/cm^2/cm/s/°C (26 W/m°C)
430°C	0·070 cal/cm^2/cm/s°C (29 W/m°C)
Electrical resistivity	160 microhm/cm^3 (1·6 $\mu\Omega$.m)
Magnetic permeability	1·1–1·4 (1·3–1·76 μH/m)

The microstructure is normally austenitic, with free chromium-bearing carbides where chromium exceeds about 1·8 per cent. For the best mechanical properties and oxidation resistance the graphite should be fine and preferably undercooled. If the silicon exceeds 5 per cent, a further phase, silico ferrite, may be formed, which is brittle. With silicon above 5 per cent, chromium below 1·8 per cent, and nickel below 17 per cent, heating within the range

Table 104 PROPERTIES OF NICROSILAL WITH DIFFERING PERCENTAGES OF CHROMIUM (BASED ON 1·25 IN (32 MM) DIAMETER BARS)

	Soft (2% *Cr*)	*Hard* (5% *Cr*)
Tensile strength, ton/in^2 (N/mm^2)	14·0 (216)	20·4 (315)
Elongation, on 2 in (51 mm), %	2·0	Nil
Modulus of rupture in transverse, ton/in^2 (N/mm^2)	36·0 (556)	45·8 (707)
Deflection, in (mm)	0·47 (12)	0·47 (12)
Brinell hardness number	143	190

Short-term tensile properties at high temperatures

Temp. °C	*Yield point tensile strength* ton/in^2 (N/mm^2)				*Elongation on 2 in (51 mm)* %		*Reduction of area* %	
	Soft	*Hard*	*Soft*	*Hard*	*Soft*	*Hard*	*Soft*	*Hard*
20	10 (154)	14·0 (216)	10·9 (168)	18·0 (278)	1·2	1·0	1·8	1·8
200	9·4 (145)	15·0 (232)	10·7 (165)	17·6 (272)	2·0	1·2	4·5	3·6
500	9·2 (142)	15·0 (232)	10·0 (154)	17·2 (266)	1·5	2·8	3·6	5·0
650	— —	10·6 (164)	— —	13·3 (205)	—	2·0	—	3·0
800	4·0 (62)	5·0 (77)	4·0 (62)	7·8 (120)	1·5	6·0	2·0	5·2

400–700°C will cause phase changes making the casting magnetic and hard causing changes that will lead to cracking. Within the correct composition range, however, Nicrosilal has excellent mechanical properties, both at normal and high temperatures, and very high oxidation resistance up to 900–1000°C. With the nickel, manganese and chromium figures at the upper limits, the austenite/martensite transformation temperature[12] will be below −150°C.

The soft variety (*Table 104*) is readily machinable with high speed steel tools, but the hard variety requires carbide tools.

The corrosion rate of both varieties in 1 per cent and 5 per cent sulphuric acid is found to be approximately one-hundredth that of grey iron, and in salt spray approximately one-fifth.

Both varieties are readily weldable with welding rods of the same material. Under test, broken tensile specimens electric arc welded gave the following results:

	Soft		*Hard*	
	ton/in²	(N/mm²)	ton/in²	(N/mm²)
Original tensile strength	12·0	(185)	17·6	(272)
After welding	11·5	(178)	18·4	(284)

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ACICULAR CAST IRON

This is a special group of cast irons which are alloyed with nickel or copper and molybdenum so that at normal cooling rates the transformation from γ to α iron is partly suppressed and an intermediate range of matrix structures is produced[1, 2, 3]. The material is characterised by high strength [23–30 ton/in² (355–463 N/mm²)] and high machinable hardness up to 320 Brinell.

The acicular irons usually contain more than 1·2 per cent nickel and 0·8 per cent molybdenum, give tensile strengths of 26–32 ton/in² (402–494 N/

mm^2), hardnesses from 255 to 320 Brinell, and impact figures (BS 1349) on the 0·798 in (20 mm) dia. unnotched bar of 30–45 ft/lb (41–61 J).

Since the acicular structure is dependent upon the effect of alloys upon the critical transformation rate, the amount of alloy required depends upon the section thickness (*see Figures 93* and *94*). Molybdenum is not usually less than 0·8 per cent and is preferably above 1 per cent, and the nickel content may vary with the section, from about 1·0 per cent to 1·4 per cent for sections below 1–1½ in (25–38 mm), up to 2·5 per cent to 3·0 per cent nickel for sections of 2–4 in (51–102 mm)^2.

Typical properties for a commercially produced acicular iron cylinder 11½ in (292 mm) outside diameter, 20 in (508 mm) long, and 1½ in (38 mm) section are given[4] in *Table 105*.

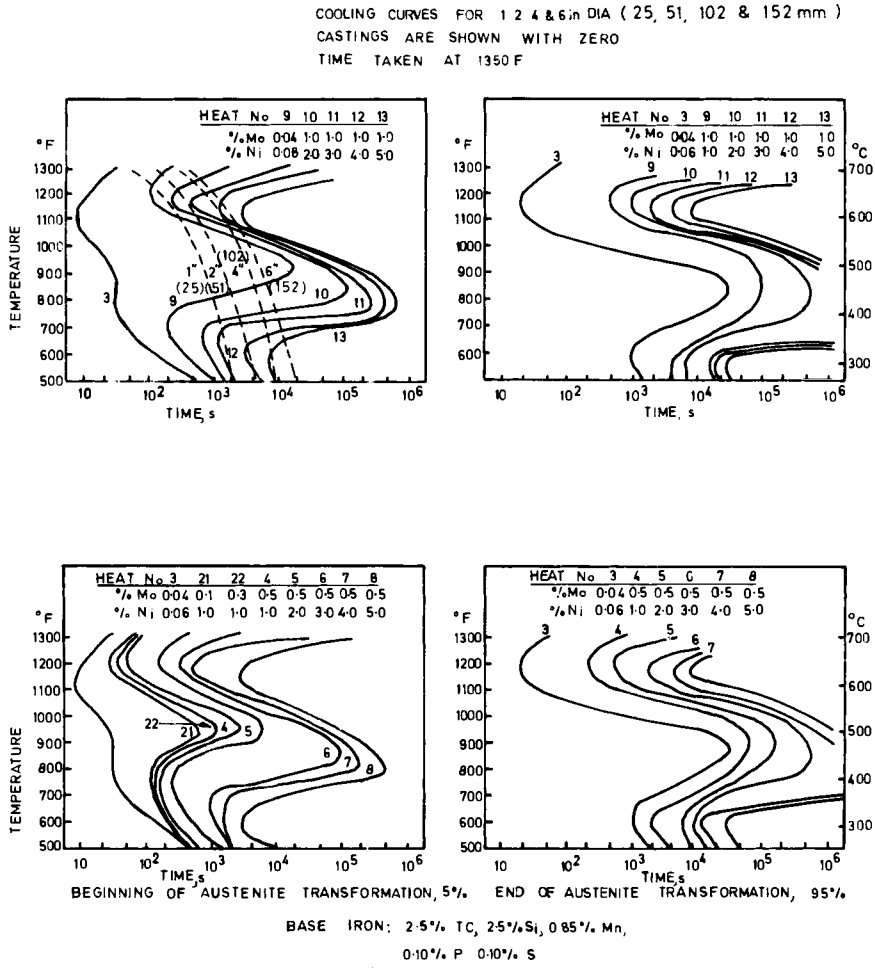


Figure 93. Time–transformation curves for Ni–Mo cast irons

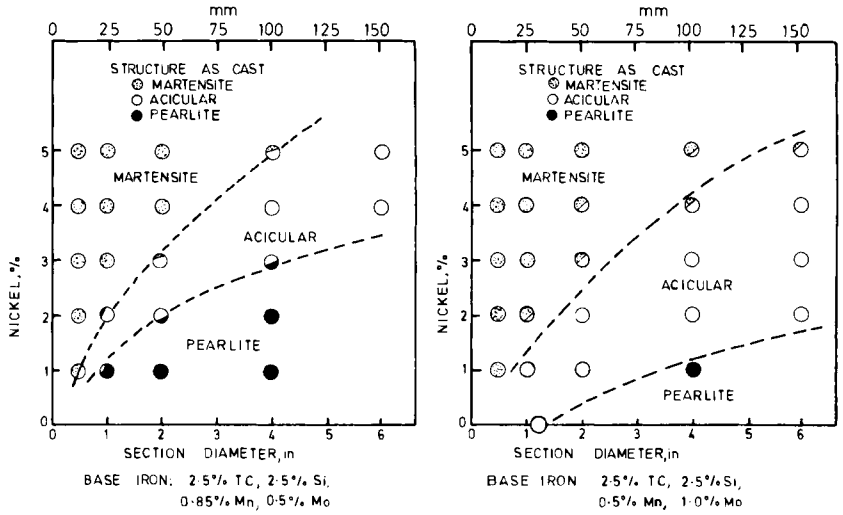


Figure 94. Effect of section size on structure of Ni–Mo cast irons

Table 105 PROPERTIES OF ACICULAR CAST IRON[4]

Composition (%):

TC	*CC*	*GC*	*Si*	*Mn*	*S*	*P*	*Ni*	*Mo*	*Cu*
2·72	0·58	2·14	1·78	0·74	0·039	0·048	1·87	1·57	0·10

Mechanical properties:

Tensile ton/in^2 (N/mm^2)	*Hardness Brinell*	*Impact* ft.lb (J)	*Shear* ton/in^2 (N/mm^2)	*Compression* ton/in^2 (N/mm^2)	*Fatigue unnotched* ton/in^2 (N/mm^2)
30·3* (468)	256	44 (60)	27·5 (425)	86 (1328)	±11·5 (178)

Modulus of elasticity, lb/in^2 × 10^6 (GN/m^2):

Tensile	*Torsion*	*Compression*
18·1 (125)	8·06 (56)	20·0 (138)

Electrical resistivity: 66·13 microhm/cm^3 (0·66 $\mu\Omega$.m)

Coefficient of thermal expansion, per °C × 10^{-6}:

20–100°C	20–200°C	20–300°C	20–400°C
14·4	14·5	14·8	15·2

Density: 7·32 g/cc, 0·264 lb/in^3

* The elongation in tensile on 3 in (76 mm) gauge of a 0·798 in (20 mm) diameter bar was 0·85%.

Acicular irons have proved of value when high strength in heavy sections is required, and where wear resistance is called for, as in die blocks for hot or cold forging and pressing.

They are more readily machinable at high levels of hardness (250–320 HB) than unalloyed irons, owing to the absence of free cementite.

If satisfactory structures and toughness are to be obtained, the phosphorus contents should be below 0·1 per cent.

Copper can be interchanged for nickel up to about 3 per cent[5, 6, 7].

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MARTENSITIC CAST IRON

This term is applied in this section only to irons which are martensitic as a result of alloy additions, and not to unalloyed irons made martensitic by quench and temper treatments.

As discussed in the section on heat treatment, the addition of elements such as nickel, copper and molybdenum, markedly affects the critical cooling rate through the austenite/ferrite change point. Partly because of its greater solubility, nickel is more effective than copper, and if sufficient nickel is added the change may be completely arrested at normal rates of cooling, to produce austenite (*see* the section on austenitic cast iron), or with lower percentages of nickel and given rates of cooling, various intermediate structures may be produced. The effect is illustrated in *Figure 82* in the section on the effect of alloying elements in normal grey irons. Use is made of this in the commercial production of martensitic cast irons which are available in two distinct forms—grey and white.

Martensitic grey irons

In these irons the silicon is sufficiently high for the normal formation of graphite to occur on solidification, but sufficient nickel has been added to slow up the austenite/ferrite transformation rate, and so for a given cooling rate (or section thickness) to produce a martensitic matrix instead of a pearlitic one[1, 2].

Martensitic grey irons are of particular value where hardnesses from 300 to 450 are required, but some degree of machinability is also desirable. The

presence of graphite lowers the effective hardness of the martensite, and in the absence of free carbide the material can be machined at suitable feeds and speeds. The amount of nickel required will vary to some extent with the section thickness, but the data given in *Table 106* are for an actual melt cast into 1·2 in (30 mm) diameter sand-cast bars. Chromium from 0·5–0·8 per cent appears to be desirable. Heat treatment for 5 hours at 230°C will remove most retained austenite.

Where preliminary machinability is important, an iron containing 6·5 per cent nickel and 3·75 per cent manganese will be austenitic when cast in normal sections, giving a tensile of 12·8 ton/in^2 (198 N/mm^2) and a hardness of 160 Brinell. After machining it can be heat treated at 600°C for 15 hours to give a tensile strength of approximately 20 ton/in^2 (309 N/mm^2) and a hardness of 410–420 Brinell.

Table 106

Composition					*Properties*		
TC %	*Si* %	*Mn* %	*Ni* %	*Cr* %	*Tensile strength* ton/in^2	(N/mm^2)	*BHN*
3·3	1·05	0·71	4·95	0·49	19	(293)	404

Martensitic white irons

In these irons the silicon is kept very low, so that at solidification iron carbide is formed instead of graphite, but again nickel is added for precisely the same reason as in the case of the grey martensitic irons, i.e. to ensure the formation of martensite instead of pearlite. Since nickel, however, is also a graphitiser of about one-third the effectiveness of silicon, chromium is also added as a carbide stabiliser to offset the graphitising effect of nickel, so that the final structure is one of iron/chromium carbides with martensite.

The martensitic white irons based on nickel additions were developed in the laboratories of International Nickel Ltd. under the name of Ni-Hard. Ordinary white irons have Brinell hardnesses ranging from 350 to 550 since they consist structurally of iron carbide with a hardness of 900 to 1200, and pearlite with a hardness of 220 to 300. The alloyed martensitic white irons can develop hardnesses from 500 to 780 DPN because the carbide hardness remains the same or a little higher, but martensite and some retained austenite with a hardness of 600 to 700 DPN takes the place of pearlite.

The amount of nickel required varies with the section thickness (or cooling rate) and the chromium is increased with the nickel content to offset its graphitising effect. Maximum hardness is obtained with high carbon contents which increase the amount of free carbide.

Maximum toughness is obtained with some sacrifice of hardness by means of a low total carbon, which reduces the amount of free carbide.

Table 107 gives the generally accepted range of composition[3, 5].

Table 107

Section in (mm)	*Properties*		*TC* %	*Si* %	*Mn* %	*P* %	*Ni* %	*Cr* %
	Maximum hardness:					up to		
$\frac{1}{4}$–$\frac{1}{2}$ (6–13)	Sand-	Chill-	3·3–3·6	0·8	0·3–0·5	0·4	2·5	1·5
$\frac{1}{2}$–2 (13–51)	cast	cast	3·3–3·6	0·6	„	„	3·2	1·7
2–4 (51–102)	550–650	600–725	3·3–3·6	0·5	„	„	4·5	2·0
	Maximum toughness:							
$\frac{1}{4}$–$\frac{1}{2}$ (6–13)	Sand-	Chill-	2·7–3·2	0·5	„	„	3·0	1·5
$\frac{1}{2}$–2 (13–51)	cast	cast	2·7–3·2	0·5	„	„	4·0	1·7
2–4 (51–102)	525–625	575–675	2·7–3·2	0·5	„	„	5·0	2·0

The patternmaker's shrinkage is $\frac{5}{32}$ in to $\frac{1}{4}$ in per ft (1:750–1:48).

The typical mechanical properties are compared with those of unalloyed pearlitic white irons[4, 5, 6] in *Table 108*.

Stress relief at 200–275°C for 4–48 h is recommended and increases both the strength and impact resistance. The effect of heat treatment and cooling rate of casting on the microstructure and hardness is complex. Such heat treatment, however, reduces residual austenite and increases tempered martensite and thereby increases fatigue resistance[7, 8].

HEAT RESISTANCE OF NI-HARD TYPE IRONS

The chromium content of these irons makes them extremely stable at high temperatures, and they are stated to be capable of operating for 14 000–28 000 h at 480–540°C with a drop in hardness from 660 to 620 DPN.[5]

Martensitic white irons can also be obtained without the use of nickel by the addition of molybdenum and chromium which also have a marked effect on the austenite transformation rate[9]. The typical composition range is:

	%
Total carbon	3·0–3·5
Silicon	0·3–0·6
Manganese	0·6–0·9
Sulphur	0·06 max.
Phosphorus	0·1 max.
Nickel	low
Chromium	15·0–18·0
Molybdenum	2·75–3·25

This gives complex carbides of high hardness in a matrix of martensite and retained austenite. The retained austenite is unstable, and while it contributes some toughness to the casting, it is said to transform to hard martensite under abrasion.

Mechanical properties of 1·2 in (30 mm) sand-cast bars are given in *Table 109*.

Table 108

	High carbon 3·0–3·5% (hard)		*Low carbon 2·75–2·9% (tough)*	
	Unalloyed pearlitic	*Alloyed 2·5–4·75% Ni martensitic*	*Unalloyed pearlitic*	*Alloyed 2·5–4·75% Ni martensitic*
Tensile strength, ton/in^2 (N/mm^2)	16–18 (247–278)	16–17 (247–263)	18–30 (278–463)	21–36 (324–556)
Modulus of elasticity, lb/in^2 × 10^6 (GN/m^2)	30 (207)	24–26 (165–179)	30 (207)	24–26 (165–179)
Hardness (DPN)	450–600	590–800	400–550	550–650
Transverse rupture stress, ton/in^2 (N/mm^2)	28–35 (432–541)	31–53 (479–819)	33–47 (510–726)	35–55 (541–849)
Transverse deflection:				
1·2 in bar, 12 in span, in	0·05–0·1	0·08–0·11	0·07–0·11	0·1–0·12
30 mm bar, 305 mm span, mm	1·27–2·54	2·03–2·79	1·78–2·79	2·54–3·05
Specific gravity	7·0–7·8	7·6–7·8	7·6–7·8	7·6–7·8
Coefficient of expansion:				
0–430°C	$10–11 \times 10^{-6}$	$8–9 \times 10^{-6}$	$10–11 \times 10^{-6}$	$8–9 \times 10^{-6}$
0–95°C	—	$12·2–14·2 \times 10^{-6}$	—	$12·2–14·2 \times 10^{-6}$
276–649°C	$15·9–16·4 \times 10^{-6}$	—	$15·9–16·4 \times 10^{-6}$	—

Table 109

Composition %	*Properties*	*As-cast*	*Heat-treated**
Total carbon 3·37	Transverse strength, 18 in span, lb	2235	2900
	Transverse strength (457 mm) span, N	9942	12 900
Chromium 15·0	Transverse rupture stress, ton/in^2	26·4	34·5
	Transverse rupture stress, N/mm^2	408	533
Molybdenum 2·95	Deflection, in	0·125	0·137
	Deflection, mm	3·18	3·48
	Diamond pyramid hardness no.	519	803

* Heat treated at 1040–1070°C for 1 h per in section, air-cooled and stress-relieved at 200–240°C. This may be particularly necessary with heavy section castings which have developed pearlitic structures as a result of slow cooling in the mould.

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WHITE CAST IRON

White cast irons, usually because of their low silicon content, have solidified without the formation of graphite and normally consist of free cementite and pearlite, as described in the section on constitution and structure.

They are characterised by a completely white fracture and high hardness (400–600 DPN). The hardness figures are the result of the carbide hardness (900–1200 DNP) and pearlite hardness (200–300 DPN).

The high hardness makes them very suitable for abrasion-resistant castings and the hardness can be still further increased by alloying with nickel and chromium so that the austenite transforms to martensite instead of pearlite, giving an overall hardness of 540–800 DPN, depending upon whether the material is sand- or chill-cast (*see* section on martensitic cast irons). Since the hardness is particularly dependent upon the amount of carbide present, the high total carbon white irons are harder than the low total carbon irons, but are weaker and somewhat more brittle. Additions of 1·0 to 1·5 per cent nickel and 1·5 to 2·0 per cent manganese in high carbon chilled iron may, by refining the pearlite, raise hardness from about 550 to 700 DPN.

White cast iron would be expected to be free from the effect on stress re-

sponse caused by graphite discontinuities and should behave in a substantially elastic manner. It would therefore be expected that the skin stress or transverse rupture stress in bending would be similar to the maximum tensile stress. Figures by de Longe[1], however, show transverse rupture stresses [as-cast, 1·2 in diameter by 12 in span (30 mm × 305 mm)] approximately 1·75 to 2·0 times the tensile strength for both alloyed and martensitic white irons, and other figures by Grant[2] show similar ratios for unalloyed white irons with varying manganese and phosphorus contents. Grant states that tensile strength is very sensitive to section size, and Williams[3] has pointed out that the influence of structure is also very marked in materials of the same composition and section. It is evident, therefore, that it is undesirable to deduce the tensile strength from the transverse rupture stress.

Stress-relief can be carried out at 530–600°C provided uniform heating and cooling is assured. The material can be hot-formed at temperatures between 750 and 1000°C.

The figures[2, 3, 4] given in *Table 110* are probably representative of pearlitic unalloyed white cast irons cast into 1·2 in (30 mm) diameter bars (total carbon, 2·5–3·5 per cent; silicon, 0·4–0·67 per cent; manganese, 0·3–0·96 per cent; phosphorus, 0·1–0·6 per cent).

TABLE 110 PROPERTIES OF WHITE CAST IRON

	Sand-cast	*Chill-cast*
Hardness (DPN)	410–520	435–550
Tensile strength, ton/in^2 (N/mm^2)	15–30	18–30
	(232–463)	(278–463)
Modulus of elasticity, lb/in^2 (GN/m^2)	30×10^6	30×10^6
	(207)	(207)
Transverse rupture stress, 1·2 in bar,		
12 in span, ton/in^2	24–40	33–47
(30 mm bar, 305 mm span), N/mm^2	371–618	510–726
Specific gravity	7·6–7·8	7·6–7·8
Impact strength (BS 1349)		
0·798 in dia. machined from 0·875 in as-cast dia., ft.lb	6–11	—
(20 mm from 22 mm), J	8–15	—
Compression strength, ton/in^2 (N/mm^2)	98–112	—
	(1514–1730)	—
Thermal conductivity, cal/cm^3/s/°C (W/m.°C)	0·035–0·076	—
	(15–32)	—
Thermal expansion coefficient:		
0–276°C	$10–11 \times 10^{-6}$	—
276–684°C	$15·9–16·4 \times 10^{-6}$	—

The stress/strain properties of three white irons with varying carbon contents are shown in *Table 111*. The test bars were sand-cast to shape and ground, and were heat treated at 550°C for 1 h and furnace cooled to relieve residual stresses.

The modulus of elasticity is fairly constant at a value of 30×10^6 lb/in^2 (207 GN/m^2). It may therefore be inferred that this value applies to eutectic carbide which is in agreement with that found by Laszlo and Nolle[5] who

Table 111

	TC %	*Si* %	*Mn* %	*S* %	*P* %
(1)	2·57	1·03	0·21	0·10	0·07
(2)	2·22	1·12	0·22	0·10	0·05
(3)	1·77	0·96	0·22	0·10	0·07

	Tensile strength ton/in²	(N/mm²)	*Limit of proportionality, 0·005–0·01% proof stress* ton/in²	(N/mm²)	*Strain to failure* %	*Permanent strain at failure* %	*Modulus of elasticity* 10^6 lb/in²	(GN/m²)
(1)	33·9	(524)	22·5–25·0	(347–386)	0·447	0·194	29·4	(203)
	36·6	(565)	22·5–25·0	(347–386)	0·575	0·248	30·4	(210)
(2)	38·7	(598)	26·3–28·8	(406–599)	0·570	0·279	29·9	(206)
	34·9	(539)	26·3–28·8	(406–599)	0·400	0·147	30·6	(211)
(3)	35·4	(547)	26·3–28·8	(406–599)	0·630	0·361	29·7	(205)
	31·9	(493)	26·3–28·8	(406–599)	0·380	0·141	30·8	(212)

found that ferrite, pearlite and cementite had the same value (*see also* section on stress/strain).

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HIGH SILICON CAST IRON[1, 2]

High silicon cast iron or acid-resisting cast iron, is the name given to those irons containing between 14 and 18 per cent silicon, which have been developed especially to withstand acid media. Their discovery is normally credited to Jouve, a French metallurgist, in 1908, and since then a considerable amount of work has been carried out to determine the properties and methods of production[3]. High silicon cast iron is covered by BS 1591: 1949, which contains details of composition and heat treatment required.

The normal range of composition quoted is: total carbon not less than 0·35 per cent, and not more than 1 per cent; silicon not less than 14·25 per cent and not more than 15·25 per cent; manganese 1·0 per cent (not more than), manganese above 0·4 per cent may discolour acid; sulphur not more than 0·1 per cent; and phosphorus not more than 1·0 per cent. Silicon 17–18 per cent may be required to give some resistance to hydrochloric acid at certain concentrations and temperatures, but results in more fragile than normal grades.

The corrosion resistance is dealt with in the section on corrosion.

Constitution

The constitution of the 14–15 per cent silicon alloys has, for a considerable time, been known to consist of a solid solution, but above this the structure becomes complex, and variations were quoted in the literature on the composition and nature of phases present.

Work carried out during 1939–1945[2] evolved the constitutional diagram shown in *Figure 95*. A solid solution (α) exists at room temperature up to 15·2 per cent silicon, but at high silicon contents a new phase (ε) appears.

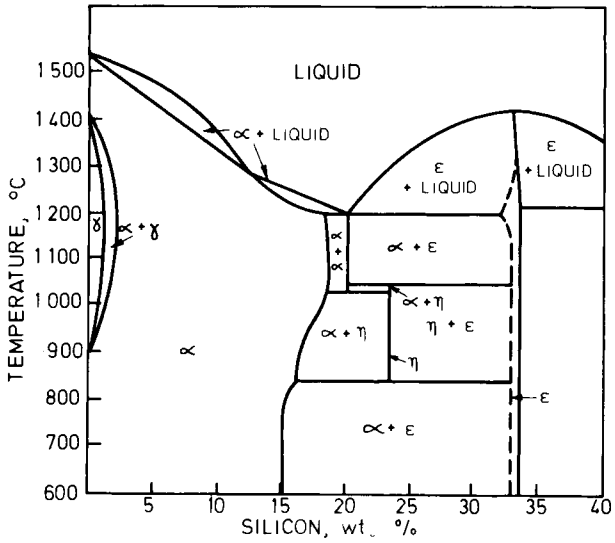


Figure 95. Iron–silicon equilibrium diagram

At temperatures between 825 and 1030°C, the η phase is stable and although apparently not stable at room temperature, the phase changes are so sluggish that under normal rates of cooling the η phase appears in alloys containing between 15·2 and 30 per cent silicon.

The effect of carbon on the iron–silicon equilibrium diagram does not appear to alter the distribution of the phases present, but does reduce slightly the freezing range of the system and also tends to reduce the silicon value of the eutectic. The majority of the carbon present exists as flake graphite, although a carbide is sometimes found in the as-cast microstructure, especially if rapid cooling has occurred immediately after solidification.

Effect of elements

CARBON

The control of this element is most important in the production of high silicon iron, as too small an amount will produce excessive shrinkage and unsound

castings, while too high a carbon content causes the deposition of kish graphite before and during solidification, again producing defective castings. The optimum carbon content of the 14–16 per cent silicon irons is about 0·5 per cent, and with a higher silicon content the value should be reduced to about 0·3 per cent. The fact that the kish or excess graphite is readily thrown out of solution as the melt cools assists in obtaining the correct carbon content, and this fact is used to advantage in cupola melting.

SILICON

A high level of silicon is required to give maximum corrosion resistance, especially to hydrochloric acid, but an excessive amount causes the appearance of the η phase in the microstructure, which is stated to increase considerably the brittleness of the material. In view of this, the silicon content is normally 14·24 per cent, although the British Standard for the material does provide for a range of 14·25 to 15·25 per cent.

MANGANESE

This is usually present in commercial castings in amounts varying from 0·4 to 0·9 per cent and although no separate phases are produced, there is reason to believe that a high manganese content encourages the formation of the carbide phase. Manganese above 0·4 per cent is said to discolour the acid.

SULPHUR

The sulphur content will depend on the type of furnace and the quality of the raw materials used.

PHOSPHORUS

Phosphorus appears as an iron phosphide complex and is usually associated with any iron carbide which is present. Unlike the carbide, however, iron phosphide is not removed by annealing. The presence of large amounts of phósphorus increases the brittleness of the iron and when melting is carried out in furnaces other than the cupola, the value is usually kept below 0·3 per cent, and preferably below 0·1 per cent. However, in cupola melting, owing to the difficulty of attaining a high metal temperature, the presence of phosphorus in amounts of 0·7–1·0 per cent assists in providing the necessary fluidity and may reduce carbon pick-up.

OTHER ELEMENTS

The only other element which is added to high silicon iron is molybdenum and it has been claimed that the addition of 3·5 per cent of this element markedly increases the corrosion resistance to hydrochloric acid.

Microstructure

With silicon contents of up to 15·2 per cent, the microstructure of high silicon cast iron is a solid solution, the carbon being dispersed throughout the matrix as graphite. The graphite may be present in the form of fine flake, coarse flake and/or in acicular or rod-like form.

With higher silicon contents both η and ε phases can be present and can be identified under polarised light as shown by Hurst and Riley[4].

Properties	
Specific gravity	7·0 (428 lb/ft^3)
Specific heat	0·130 (544 J/kg °C)
Coefficient of expansion	12–15·9 × 10^{-6}/°C
Electrical resistivity	50 microhm/cm^3 (0·50 $\mu\Omega$.m)
Tensile strength	6–10 ton/in^2 (93–154 N/mm^2)
Transverse strength	18 ton/in^2 (278 N/mm^2)
Modulus of elasticity	18·0 × 10^6 lb/in^2 (124 GN/m^2)
Brinell hardness no.	450–520
Impact strength (BS 1349: 1947)	4–6 ft.lb (5–8 J)

Impact strength

The main disadvantage of this material is its excessive brittleness and this creates difficulties in handling and in the foundry.

Corrosion resistance[5]

The corrosion resistance of high silicon cast iron depends on the formation of a silica-like oxide film early in the service life of the casting. The iron atoms are leached out of the silico-ferrite lattice and the remaining silicon atoms are oxidised to form a protective silicon–oxygen film which inhibits further corrosion.

High silicon cast iron has good resistance to both sulphuric and nitric acids at all concentrations and temperatures, and also to mixtures of sulphuric and nitric acids.

Its usefulness in handling hydrochloric acid is somewhat restricted and where the corrosion rates are to be held below 5 mils/year (127 μm/year) the

temperature must not exceed 80°C for a 1 per cent acid, and 0°C for a 40 per cent acid. Where great resistance to hydrochloric acid is required, the silicon content must be increased to 18 per cent, or an addition of 3·5 per cent molybdenum made, this being the usual American practice.

High silicon iron has no resistance to hydrofluoric acid, but is resistant to phosphoric acid at room temperature. It is resistant to organic acids at all temperatures and concentrations.

Its behaviour in alkalis is inferior to that of ordinary grey iron.

Production

High silicon iron can be made in the cupola, or in electric or reverberatory furnaces or oil-fired rotary furnaces. However, the cupola is very little used at present, owing to:

(1) The very high carbon pick-up which occurs during melting.
(2) The difficulty of obtaining adequate metal temperature.

Details of cupola melting and moulding practice have been given by Marsden[6].

Electric melting eliminates the risk of hydrogen pick-up from products of combustion and is normally employed.

High silicon irons tend to cause growth of acid linings and in cupolas the linings must be kept very clean between melts to avoid bridging.

The usual charge materials are steel scrap, ferrosilicon, hematite, refined iron and scrap returns, a typical charge for a rotary furnace being:

Hematite (Si 2·0 per cent)	2·0 per cent
Ferrosilicon (90 per cent)	9·5 per cent
Steel scrap	38·5 per cent
High silicon iron scrap	50·0 per cent

These materials must be stored under cover to avoid rusting and hydrogen pick-up.

High silicon iron castings are extremely prone to gas unsoundness owing primarily to the evolution of hydrogen during solidification. It therefore becomes necessary to eliminate all possible sources of hydrogen pick-up by:

(1) Use of clean and dry raw materials of low hydrogen content.
(2) Avoidance of excess metal temperatures.
(3) Absence of damp ladles, high moisture content sand and damp cores.
(4) Drying of moulds and cores must be very thorough and skin-drying is usually ineffective.
(5) Denseners may be required at junctions.

The usual practice in this country is to melt the charge materials first and to produce a high silicon pig of required composition. Subsequent remelting of this pig produces a melt with a low gas content, giving sound castings.

A paper published in Russia in 1938[7] showed the importance in their methods of manufacture of using ferrosilicon with low aluminium content.

When this was done little trouble was apparently found in obtaining sound castings, even with no remelting. The use of low aluminium content ferrosilicon was also found by Krupps[8] to be advantageous. Aluminium and calcium should be below 2 per cent and as much lower as possible[7]. This is quite interesting in view of the recent work carried out at BCIRA on the effect of aluminium on the formation of pinholes in grey and nodular cast irons, which attributed this to high hydrogen pick-up.

The shrinkage of high silicon cast iron is much higher than that of normal grey cast iron and a patternmaker's concentration allowance of $\frac{3}{16}$ in per foot (1:64) is made. The high shrinkage value and the relatively low strength of the material necessitate the use of mould materials which break down readily after solidification of the metal. Correct design of the casting is most important in order to avoid the production of high internal stresses with consequent cracking.

Cores must be capable of easy removal before annealing, and steel chaplets should not be used.

Heat treatment

High silicon cast iron is a low strength, brittle material and it is very susceptible to cracking owing to a build-up of internal stresses associated with differential cooling rates of various sections. These stresses may be removed by stripping the castings from the moulds before a temperature of 800°C is reached and placing them in an annealing furnace at 800–850°C. The castings are allowed to remain in the furnace for 4–5 h, after which they are cooled to room temperature over a period of about 30 h.

Machining

Owing to the high hardness of these irons, machining with carbide tipped tools is preferable. In any case, machining is usually confined to flanges and is normally carried out by grinding methods.

The material has excellent wear resistance against hard wet abrasives.

High silicon iron can be gas- or arc-welded (BS 1591 : 1949) with high silicon iron rods. The castings must be carefully preheated and maintained at a temperature of not less than 700°C throughout the whole of the welding operation by means of a gas flame. A slightly oxidising flame is required with a normal cast iron flux. A flux of equal parts of powdered ferrosilicon, ferromanganese and borax has been used. When the weld is completed the casting should be maintained at 700–800°C for about 1 h, and then slowly cooled.

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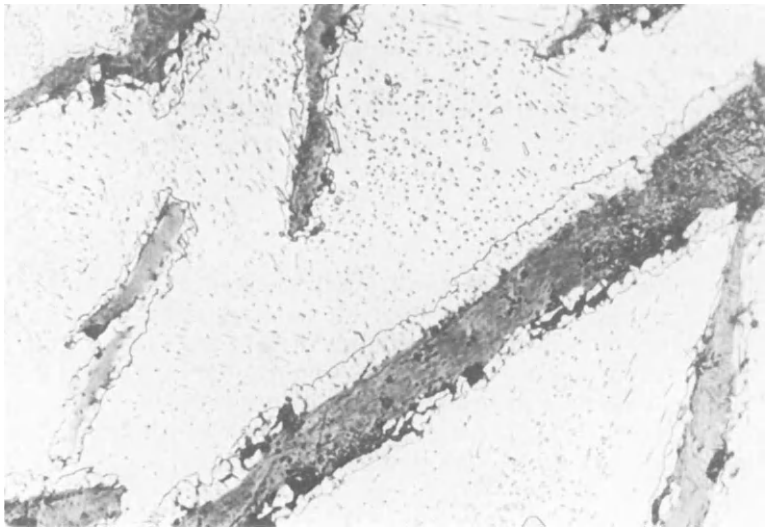


Figure A. Ingot mould iron. The spheroidised cementite indicates that it had been held below the α to γ transformation point (see Figure 6) and oxidation of the metal around the graphite flakes had occurred. The temperature then rose above that at which carbon and iron oxide can co-exist in equilibrium (Figure B) but still below the α to γ transformation point and the iron was reduced to free iron (etched 4% picral, ×600, reduced to six-tenths on reproduction)

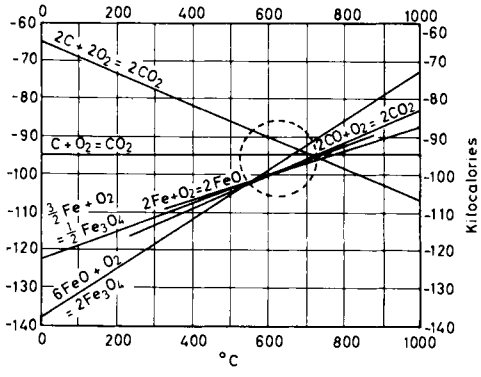


Figure B. The standard free energy of formation of iron oxides, CO and CO_2 as a function of temperature

4

Special properties affecting service

HEAT RESISTANCE

A brief consideration of the heat resistance requirements of castings will help to determine the type of iron that can be most usefully employed for a particular application (*Table 112*).

Failure in service may occur as a result of scale, growth and mechanical failure. A very full survey of the work on growth and scaling has been given by Hughes[1].

Scaling

Scaling is caused by oxidation or other chemical attack in which the surface of the part is progressively destroyed.

The scale on pure iron consists of three layers: an outer one of Fe_2O_3; an intermediate one of Fe_3O_4; and an inner one of FeO. Scale on cast iron is more adherent than with steel and hinders free access of the atmosphere to the metal below, so that attack is reduced unless the scale is cracked or removed mechanically. Scale on cast iron is usually continuous and adherent at least up to 800°C.

Oxidation of the metallic matrix around the graphite flakes may occur preferentially to graphite oxidation at temperatures above 400°C and below 650–700°C. Above 650–700°C carbon and iron oxide will not exist together in equilibrium[2,3] and oxide so formed at a lower temperature may be reduced by the graphite if the temperature rises above 650–700°C, yielding a layer of ferrite around the graphite flakes[4] (see *Figures A* and *B*, p. 252).

EFFECT OF ATMOSPHERE

At temperatures over 800°C the rate of surface oxidation for ferrous metals is greatest in air. Where products of combustion provide the ambient

Table 112 HEAT-RESISTING CAST IRON

Type	*Suitable service conditions*	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Cr* %	*Mo* %	*Room temperature properties*		*Coefficient of expansion* per °C	*Stress to rupture in 10 000 h* ton/in² (N/mm²)		
										Tensile strength ton/in² (N/mm²)	*Brinell hardness*		*350°C*	*500°C*	*700°C*
Hematite, high carbon unalloyed	Resistant to thermal shock, not resistant to growth or scaling above 450–500°C	3·6–4·0	1·4–2·0	0·4–0·6	0·05	0·05	—	—	—	6–10 (93–154)	120–140	$9–11 \times 10^{-6}$	4–8 (62–124)	(1·0) (15)	—
High phosphorus grey iron	Cheap to produce; moderately scale-resistant up to about 600°C. More rigid at high temperatures than low phosphorus materials. Suitable for firebars	3·0–3·3	2·2–3·0	0·4–0·7	<0·12	0·9–1·4	—	—	—	10–12 (154–185)	200–240	13×10^{-6} up to 650°C	10–15 (154–232)	(3·0) (46)	(0·1–0·2) (2–3)
Low alloy grey iron	Suitable for temperatures up to 400–700°C where dimensional stability will be greater than unalloyed irons, and some improvement in oxidation resistance will occur	3·0–3·4	1·6–2·8	0·4–1·0	<0·12	0·1–0·4	—	0·5–2·0	Up to 0·5	19–25 (293–386)	200–260 Grey 350–450 White	$11–12 \times 10^{-6}$ air temp. $13–14 \times 10^{-6}$ 0–650°C	15–20 (232–309)	3·0–4·0 (46–62)	(0·1–0·2) (2–3)
Silal	Cupola-produced; the cheapest oxidation and growth resistant iron. Suitable for temperatures up to 800°C	Up to 2·3	5·5–7·0	0·5–0·8	0·06	0·1–0·3	—	—	—	9–17 (139–263) (0·875 in dia. bar)	220–255	13×10^{-6} 0–500°C	—	2·0 (31)	(0·1–0·2) (2–3)
	Crucible or H.F. furnace: low carbon; highly growth and oxidation resistant up to 850–900°C	1·8–2·0	5·5–7·0	0·5–0·8	0·06	0·1–0·3	—	—	—						
Nicrosilal (austenitic)	Suitable for temperatures between 650 and 900°C	1·5–2·0	4·5–5·0	0·6–1·0	0·1	<0·1	18·0–23·0	2·0–2·4	—	9–16 (139–247)	150–200	18×10^{-6}	(7–12) (108–185)	(4–6) (62–93)	<1·0 (<15)
Ni-Resist (low Si) (austenitic)	Very resistant up to 800–850°C and stable between 500 and 600°C	2·4–3·1	1·0–2·0	0·8–1·4	<0·1	0·1–1·0	12·0–17·0	2·6–3·6	Copper % 5·5–8·0	8–16 (124–247)	120–170	18×10^{-6}	(7–12) (108–185)	(4–6) (62–93)	<1·0 (<15)
Chromium irons (ferritic)	Scale resistant up to 950–1000°C. Almost completely scale resistant up to 1050°C	2·0–2·8	1·4–1·8	1·0–1·6	0·1	<0·1	—	14·0–17·0	—	20–35 (309–541)	250–480	$9–10 \times 10^{-6}$ Room temp.			
		1·0–1·3	1·0–1·3	0·7–1·0	—	—	—	30·0–33·0	—	24–35 (371–541)		13×10^{-6} 0–1000°C	20–25 (309–386)	8–10 (124–154)	(2–3) (31–46)

atmosphere, steam is more aggressive[5] than CO_2, and scaling will be more pronounced with high hydrogen-containing fuels such as oil or town gas than with coke or anthracite. Scaling can be minimised by the use of atmospheres containing hydrogen or carbon monoxide, but above 800°C can only be prevented where the reducing gas appreciably exceeds the amount of water vapour or carbon dioxide. Carbon deposition can occur where free hydrocarbons such as methane or free carbon monoxide (13 per cent) are present and deposition on graphite flakes can cause severe growth and distortion at temperatures of 350–530°C and probably above.

Sulphur-bearing atmospheres are more aggressive than non-sulphur-bearing, but the high chromium irons appear to be highly resistant to sulphur-bearing atmospheres.

Scaling can be minimised by the addition of alloy elements which in themselves reduce the susceptibility of the iron to oxidation and scaling.

It is usual to assess the heat resistance of cast irons in terms of their resistance when exposed to air at a given temperature for a given period of time, and a wide range of literature is available on this subject. It is important to remember that very few service conditions actually conform to these experimental conditions, and the results may be very markedly modified by, for example, the presence of sulphur, the presence of reducing gases, the conditions occurring during heating up, cooling or stand-by such as moisture, condensation, the presence of ash, clinker, fluxes, and so on. Nevertheless, in general, the standard oxidation tests can be used to indicate an order of merit. Different investigators have used different methods of measuring the degree of attack, e.g. g/cm^2 or oz/ft^2 or depth of penetration of scale. The figures quoted later in this section are, however, based on investigations in which it has been possible to make approximate conversion of units to a standard form, g/dm^2 (kg/m^2) of scale.

Breakdown by scaling does not normally become important until temperatures in excess of 600–700°C are reached. Above 750°C, unalloyed irons are attacked with rapidly increasing speed as the temperature increases. The three principal alloy elements employed to improve oxidation resistance in this range are silicon, chromium and aluminium. Some resistance to oxidation above 750°C is conferred by silicon contents above 3·0 per cent, chromium contents above 1 per cent, or aluminium contents above 0·9 per cent. Solely from the point of view of scaling, low contents of chromium (below 1·5 per cent) are of limited value; silicon from 3·8–6·5 per cent, however, effects a very marked improvement, and aluminium from 1–7 per cent also appears to offer very substantial advantages provided the founding difficulties are overcome.

The 6–7 per cent silicon, 6·8 per cent aluminium, 15–30 per cent chromium irons, and the austenitic irons, are particularly suitable for temperatures up to 900°C, subject to strength requirements. At higher temperatures the 24–35 per cent chromium irons are most suitable. The high aluminium irons and high chromium irons are resistant to sulphurous atmospheres.

Growth

Substantial dimensional changes can occur, causing severe distortion by non-uniform growth, damage to the surrounding portions of the furnace or other functional disadvantages. Growth may be caused by:

(1) Structural changes in the casting itself without oxidation, notably by breakdown of combined carbon in pearlite. Graphitisation of 1 per cent combined carbon[1] should theoretically involve an increase in volume of 2·04 per cent. Actual growth is always greater.
(2) Deep internal oxidation.
(3) Carbon deposition on graphite flakes in atmospheres[6] containing carbon monoxide in the temperature range 350–550°C.

Thermal cycling through the critical α to γ temperature range generally results in catastrophic growth[1]. Increasing silicon raises the critical temperature and if added in sufficient quantity, as in Silal, raises the α to γ change point above 850–900°C (*Figure 6*). Provided the working temperature is below the critical, the material does not scale rapidly.

Growth due to carbide or pearlite breakdown below the α to γ change can be controlled by (a) increasing the stability of the combined carbon either by the use of low silicon (e.g. below 1·5 per cent), which gives a marginal improvement for temperatures up to 550°C[1], or by the addition of carbide stabilising elements, notably chromium (0·5–2·0 per cent); or (b) by preventing the formation of combined carbon by high silicon additions.

The unstressed growth of a pearlitic iron (TC 3·29 per cent, Si 1·27 per cent, Mn 0·28 per cent, S 0·12 per cent, P 0·72 per cent) has been given as 6×10^{-8} in/in/h at 450°C in air, and $3{\cdot}3 \times 10^{-5}$ in/in/h at 538°C in air[9, 10]. Phosphorus tends to stabilise pearlite and the rate of growth would be higher with a low phosphorus iron.

Growth by pearlite breakdown can be eliminated by the use of ferritic irons. Grey irons can be annealed to the ferritic state by heat treatment for 1–4 h just below the critical point (*Figure 6*). Annealing above the critical may accelerate deep oxidation at temperatures below the critical.

Nodular or SG irons show much less growth up to 500°C than grey irons and the ferritic varieties are almost free from growth (*Table 113*). Furnace door frames too large for easy annealing can be produced in the predominantly ferritic state as-cast by selection of raw materials and will show low growth at least up to 500°C.

Growth by deep oxidation can only be controlled by the addition of alloy elements such as silicon and chromium which inhibit oxidation of the matrix.

Silicon below 1·6 per cent ensures that a considerable degree of stability is conferred on any pearlite that is present. Silicon from 2·0–3·5 per cent does not entirely prevent the formation of pearlite, but makes it less stable, without conferring any important oxidation resistance. In the region up to 500–550°C, however, the ordinary high phosphorus, high silicon irons (1·0–1·4 per cent phosphorus, 2·8–3·3 per cent silicon) frequently give excellent service, and are cheap to produce.

Table 113 EFFECT ON CAST IRON OF PROLONGED EXPOSURE IN AIR TO TEMPERATURES UP TO 500°C AND SUPERHEATED STEAM AT 427°C. GILBERT[20] HAS CARRIED OUT LONG-TERM TESTS IN AIR AND SUPERHEATED STEAM ON A WIDE RANGE OF CAST IRONS AND THE RESULTS ARE SUMMARISED BELOW

								Nodular iron		*Cupola blackheart malleable*
					Grey iron			*Pearlitic*	*Ferritic*	
	Initial tensile strength			ton/in² (N/mm²)	Less than 12·9 (199)	15·5 to 18·0 (239–278)	18·0 or more (278)	46·9–48·9 (724–755)	28·1 (434)	16·6–18·4 (256–284)
	°C	°F	*Duration*							
In air	400	752	64 weeks	Growth, %	Nil	Nil	Nil	Nil	Nil	Nil
	400	752	64 weeks	Scaling, g/dm² (kg/m²)	0·25 (0·025)	0·2 (0·02)	0·19 (0·019)	0·12 (0·012)	0·12 (0·012)	0·16 (0·016)
				Loss of strength*, %	Nil	Nil	Nil	Nil	Nil	Nil
	450	842	64 weeks	Growth, %	0·32 (0·30–0·33)	0·21 (0·05–0·31)	0·13 (0·02–0·26)	0·02–0·03	0·01	0·01
				Scaling, g/dm² (kg/m²)	0·45 (0·045)	0·37 (0·037)	0·35 (0·035)	0·23 (0·023)	0·21 (0·021)	0·25 (0·025)
	450	842	64 weeks	Loss of strength*, %	13·4 (−10·3 to −16·5)	7·0 (+7·3 to −13·3)	3·6 (+8·9 to −12·1)	2·5 (0·7–4·3)	7·0	0–1·5
				Loss of hardness*, %	16·0 to 19·0	Nil to 26·8	Nil to 30·4	No change	No change	No change (+8%)
	500	932	64 weeks	Growth, %	1·03 (0·98–1·07)	0·6 (0·3–0·78)	0·55 (0·27–0·83)	0·19 (0·12–0·26)	0·01	Nil
				Scaling, g/dm² (kg/m²)	1·59 (0·159)	0·96 (0·096)	0·76 (0·076)	0·4 (0·04)	0·42 (0·042)	0·61 (0·061)
	500	932	64 weeks	Loss of strength*, %	42·1 (40·2–44·1)	30·5 (−12·4 to −43·4)	25·3 (−12·4 to −36·2)	16·0 (13·0–19·0)	9·0	7·6
				Loss of hardness*, %	36–45	15·6 to 26·4	16·6 to 44·0	11·2–16·1	No change	No change (+10%)
Super-heated steam	427	800	49 weeks	Growth, %	0·079 (0·025–0·134)	0·0017 (−0·0017 to +0·01)	0·0033 (−0·0017 to +0·01)	−0·01	−0·02	−0·01
				Scaling, g/dm² (kg/m²)	0·532 (0·0532)	0·086 (0·0086)	0·052 (0·0052)	0·055 (0·0055)	0·09 (0·009)	0·071 (0·0071)
				Loss of strength*, %	3·3 gain (−4·0 to +10·7)	3·75 loss (−11·26 to +11·93)	0·63 low (−13·4 to +9·81)	1·3–13·0	0–0·4	0–6·1
				Change of hardness*	No change	No change	No change	No change	No change	No change

Changes of less than 7·0% in tensile strength and 0·02% in growth are unlikely to be significant.
* At room temperature after exposure.

In the region from 500–750°C, small percentages of chromium, from 0·5–2·0 per cent, may reduce growth to very low figures[1], and provide an economical method of improving the performance of a component. The straight addition of chromium in these quantities confers a rather small but appreciable increase in the resistance to oxidation. Silicon above 3·8 per cent prevents growth by raising the critical temperature and by restricting the formation of pearlite; at the same time it also improves the specific resistance to oxidation, provided the graphite is in the undercooled form.

Mechanical failure

Mechanical failure may be caused solely by thermal shock, by differential growth, or by sheer inability to carry the required load.

Failure by thermal shock can be partly countered in grey irons by using a very high graphite content, which provides a cushion for suddenly applied stresses by lowering the modulus of elasticity, or by the use of irons carrying a ductile constituent, such as ferrite or austenite, according to the temperature range. Breakage by distortion can be eliminated by the use of irons that show low growth, but purely mechanical failure can be avoided only by a knowledge of the load that a particular material will carry for prolonged periods of time at a given temperature. An indication of this is given by the creep resistance, which is dealt with later.

Reference to the tables given later in this section will show that as the temperature rises above 400°C, so the short-term strength of all metals and alloys is progressively decreased[7]. Above about 350°C, however, most alloys under prolonged loading show some degree of plastic yield, even to quite small loads, and it is customary, therefore, to assess their strength on the basis of the stress to rupture in 100 000 h, or the stress to produce 0·1 per cent total strain in 10 000, 30 000 or 100 000 h.

It should, however, be borne in mind that in actual practice these stresses can frequently be very markedly exceeded, where appreciable distortion is permissible, and cases can be quoted where, for example, 30 per cent chromium irons have been used to give a useful industrial life at temperatures of 1050°C, although prolonged exposure to these temperatures will cause slow progressive distortion, even under the weight of the component itself. Creep resistance is dealt with later in this section.

It will be convenient to consider irons which are commercially useful over different ranges of temperature under oxidising conditions.

TEMPERATURES UP TO 450°C

In this range of temperatures, direct oxidation attack is very small, and failure by growth is also unlikely to occur. Any unalloyed cast iron is, therefore, suitable for these conditions, subject to the mechanical requirements. An

exception to this generalisation is where the castings are subjected to superheated steam[8]; under these conditions it is important that:

(a) The carbon equivalent value is kept below 4·0.
(b) The iron is pearlitic.

Grey, nodular and malleable cast irons These irons with silicon contents up to 2 per cent, and carbon up to 3·6 per cent show no measurable growth in air after 11 000 h and less than 0·078 per cent after 6 years at 350 and 400°C, and pearlitic irons with carbon equivalent values of less than 3·95 (BS 1452, Grade 17 or above) show less than 0·1 per cent growth[8, 9, 10] after 8750 h in superheated steam at 750°F (400°C). Such irons are, therefore, entirely suitable for use in superheated steam up to 662°F (350°C). Additions of 0·5 per cent or more of chromium or molybdenum are particularly useful in this range to reduce growth and increase resistance to creep.

Work by Gilbert[9, 10] confirms that for irons with CE values below 3·95 exposed to superheated steam at 427°C (800°F) for 8230 h, growth is less than 0·01 per cent, and no loss of strength or hardness, when tested at room temperature, occurs. Exposure to air up to 10 800 h showed no loss of strength or hardness at 400°C. At 450°C, little loss of strength or hardness was shown with grey irons of CE less than 3·9 and containing over 0·5 per cent molybdenum or chromium, with or without nickel, after 10 800 h.

Malleable and ferritic nodular irons are free from growth[9, 10] and hardness changes up to 500°C after exposures up to 6 years on return to normal temperature.

Chromium contents above 0·5 per cent confer almost complete stability of dimensions, tensile strength and hardness up to 450°C on return to room temperature, and even at 500°C growth is unlikely to exceed 0·3 per cent, and loss of tensile strength at room temperature after exposure is unlikely to exceed 15 per cent after 64 weeks exposure.

The work summarised in *Table 113*[9] yields the following conclusions:

(i) Growth, scaling, loss of tensile strength and loss of hardness are less at all temperatures above 400°C with irons above 15·5 ton/in^2 (239 N/mm^2) than irons of less than 12·9 ton/in^2 (199 N/mm^2).
(ii) Unalloyed pearlitic grey irons up to 2·0 per cent silicon and more than 15·5 ton/in^2 (239 N/mm^2) tensile show no significant change in dimensions or properties after 64 weeks at 400°C or 49 weeks in superheated steam at 427°C (800°F). Pearlitic irons with carbon equivalent values between 4·2 and 4·3 and tensiles below 13 ton/in^2 (201 N/mm^2) show less than 0·08 per cent growth and no significant change in properties under these conditions, but one high carbon iron with a tensile strength of less than 10 ton/in^2 (154 N/mm^2) showed detectable scaling in superheated steam at 427°C. Burgess and Barlow[8] have also shown that irons of low strength or which have been annealed to the ferritic state show marked growth in superheated steam at 400°C (750°F).
(iii) An alloyed iron containing more than 0·55 per cent chromium showed negligible growth and loss of tensile strength and no loss of hardness after 64 weeks at 450°C or 49 weeks in superheated steam at 427°C

(800°F). This iron also showed the lowest loss of strength (12·4 per cent) and hardness (15·6 per cent) of all the pearlitic grey irons tested after 64 weeks at 500°C in air.

(iv) Manganese contents of about 1·5 per cent also appeared to confer marked stability in growth, hardness and tensile up to 450°C.

(v) Arsenic and tin appear to confer no useful improvement to the heat resistance of grey irons. Nodular irons containing 0·06 per cent tin, however, showed no appreciable change in hardness, tensile strength or elongation after 64 weeks at 500°C.

(vi) The stability of blackheart malleable iron and ferritic nodular iron is more noteworthy, both in air up to 500°C and in superheated steam up to 427°C (800°F).

Malleable iron is very widely employed for the use of conveyor belts and chains up to 450°C, and tests justify tensile loadings up to 2·5 ton/in^2 (39 N/mm^2) at temperatures up to 413°C under conditions of continuous loading[11].

TEMPERATURES FROM 400 TO 550°C

In this range, the rate of oxidation attack is also extremely low, even at the upper end, but failure by growth is quite likely to occur. In many cases, where replacement costs justify the procedure, unalloyed irons are entirely suitable and the ordinary high phosphoric grey irons will frequently give a very useful life, even in the upper regions of this range.

Unstressed growth at about 530–550°C in unalloyed iron may be of the order of $3–4 \times 10^{-6}$ in/in/h, and the normal short-term tensile strength of any iron is unlikely to exceed 7–10 ton/in^2 (108–154 M/mm^2). There is rarely any justification for using high alloy irons within this range of temperature unless extreme dimensional accuracy is called for. The addition of 0·5–1·5 per cent of chromium should serve to reduce growth to negligible proportions[1].

Table 113 summarises the effect on grey iron, nodular iron and malleable iron of prolonged exposure to air at 500°C and superheated steam at 427°C.

TEMPERATURES FROM 550 TO 750°C

This range of temperature is frequently a difficult one to cover, since the rate of oxidation, although slow, is becoming appreciable. At the same time, at the lower end of the scale, pearlite breakdown occurs fairly rapidly, and the upper end of the scale extends into the austenite change point for many irons*.

Within this range the unalloyed irons may still give useful service where replacement costs are not a serious consideration, and where the maintenance of extreme dimensional accuracy is not required. For example, large annea-

* Chapter 1: Constitution and structure of cast iron

ling pots [6 ft long × 4 ft diameter × 2 in wall thickness (1·8 m × 1·2 m × 5 cm)] used for softening wire at temperatures of 750–780°C, have given lives of two to five years when made in iron containing: total carbon, 2·8–3·3 per cent; silicon, 2·9–3·2 per cent; manganese 0·5–0·7 per cent; sulphur 0·09 per cent; and phosphorus 1·0–1·4 per cent[12]. In this case, close dimensional tolerances were not required since the furnace construction allowed for some growth, and the temperature was below the critical α to γ temperature. Where dimensional accuracy is important, the addition of 0·5–1·5 per cent of chromium will materially reduce the rate and the amount of growth. It will not, however, appreciably affect the amount of oxidation. For example, the addition of 0·5 per cent of chromium may reduce the growth of a 6 in (152 mm) bar after approximately 300 h at 750°C from 0·012 in to 0·002 in, although about the same amount of scale will be produced in both cases. Silal (5·5–7·0 per cent silicon) is particularly suitable in this region.

At 750°C the rate of attack in air is becoming appreciable with the unalloyed and low alloyed irons, and is likely to be of the order of 10–20 g/dm^2 (1–2 kg/m^2) for a 300 h exposure; this figure is unlikely to vary appreciably over the normal range of unalloyed irons up to about 3·5 per cent silicon, or 2 per cent chromium. Increasing the silicon content to 3·8 per cent to 5 per cent will reduce the oxidation loss to approximately one-half to one-tenth, and by raising the α to γ transformation, the growth will be reduced to very small amounts provided the graphite is in the undercooled form.

Of the high alloy irons, the austenitic grey and nodular irons are particularly suitable, where shock resistance and dimensional stability are important, at temperatures above 700°C. Resistance to oxidation is extremely high. In the temperature range 350–500°C flake graphite austenitic irons may show some reduction in length due to oxidation of graphite. In the temperature range 650–800°C the austenitic grey or nodular irons suffer little deterioration in mechanical properties. The high silicon austenitic irons (BS 3468:1962) AUS 104 and 204, particularly with high chromium and nickel contents, have much better scaling properties than the other irons in this range as the scale, on cooling, is adherent and no sub-surface oxidation occurs to produce growth[1].

In the range 400–650°C prolonged exposure may produce slight growth and embrittlement in the high silicon varieties (AUS 104 and 204), but the other varieties show little growth or deterioration in mechanical properties. This does not detract from the value of the high silicon varieties above 650°C if the period of time during heating and cooling in the lower temperature range is small in relation to total exposure.

TEMPERATURES FROM 750 TO 950°C

Within this range the unalloyed irons have very limited application, at any rate under heavily oxidising conditions. Both the oxidation rate and the loss of strength with time are extremely rapid. There are, nevertheless, certain

applications where unalloyed irons can be employed, particularly where, for any reason, rate of breakdown is rapid even with alloyed irons, and the cost of replacement becomes more important than the initial life. Conditions of thermal shock, in which a casting may be operating for a period within this temperature range, and then be subjected to sudden cooling, may justify the use of a high carbon hematite iron in spite of low resistance to oxidation rather than an alloy iron, which is less resistant to the conditions of thermal shock.

In general, however, this range of temperature requires the use of alloys, and the straight addition of chromium is rarely adequate to fulfil the requirements, unless the quantity is such that the casting becomes white and unmachinable; even under these conditions, loss by oxidation is fairly rapid. Alloys which are more particularly suitable in this range are Nicrosilal, BS 3468:1962 AUS 104 and 204, Silal and 15–35 per cent chromium iron. Nicrosilal, containing up to 5 per cent of chromium, is a particularly useful austenitic iron in this temperature region. It is very highly resistant to oxidation, shows very low growth and, as it is austenitic, shows marked resistance to mechanical and thermal shock, particularly during the heating-up stages. It is less satisfactory than Ni-Resist in the temperature range up to 700°C unless the nickel content is very high, but above this temperature is superior to Ni-Resist. Silal with 6 per cent silicon is a satisfactory alloy up to 800–850°C, provided that the carbon content is kept low (preferably below 2·0 per cent). Its tendency to brittleness when cold does not persist above red heat, where its shock resistance is substantially the same as any other non-austenitic heat-resisting iron.

A good quality Silal with carbon content below 2·2 per cent can readily be produced from the cupola using 50 per cent charges of steel and silvery pig (14 per cent silicon). If the titanium content is kept above 0·1 per cent (up to about 0·3 per cent) and CO_2 or any dry inert gas is bubbled through for about one minute before pouring, very satisfactory structures are obtainable, provided phosphorus is below 0·2 per cent.

Aluminium cast irons containing more than 1–1·5 per cent aluminium, and particularly those containing 6–7 per cent aluminium, are also extremely satisfactory in laboratory tests, especially in the upper range of temperature. They are, however, brittle, like Silal, but where they are not subjected to mechanical or thermal shock, their resistance is extremely satisfactory[13]. The introduction of aluminium-bearing scrap into the foundry is open to objection and, together with founding difficulties, may account for the small industrial use that has been made of aluminium for heat resistance.

At the upper end of this range, the 15–30 per cent chromium irons are particularly suitable[1], as they are almost completely unaffected by oxidising conditions, and their growth is negligible. These irons are materially stronger than the aluminium and silicon irons, but are less resistant to shock than the austenitic irons.

TEMPERATURES ABOVE 950°C

For temperatures above 950°C, the most suitable material is the 30 per cent chromium iron*. This shows very marked resistance to oxidation, even up to temperatures of 1050°C. Nicrosilal (4–5 per cent chromium) may be used up to about 1000°C. At these temperatures, the resistance to load is very small in both cases†, as would be expected from the figures for creep strength. Nevertheless, where loading is very light and some distortion is permissible, these materials have a very useful life at temperatures of this order.

Scaling of cast iron

The figures given in *Table 114* are approximate only, and have been taken from various authorities. The units in which they were originally expressed have been converted to a uniform basis of g/dm^2 (kg/m^2).

The figures for silicon up to 3·8 per cent are based on Williams' tests[13] in which the specimens were at temperature for the times stated and were air-cooled thirteen times during the course of the test. There is a tendency for the higher phosphorus irons to be more resistant than the lower phosphorus irons, but the effect is not large enough to be reported separately. The figures for 5·7 per silicon iron (Silal) are extrapolated (240 h to 300 h) from Hallett's tests[5] which were carried out without intermediate cooling.

From the above it appears that increasing silicon to 3·0–3·8 per cent gives a marked improvement in oxidation resistance, and an addition of 2 per cent chromium increases the resistance still further. Low carbon Silal has a much greater oxidation resistance at all levels of temperature and growth is likely to be one-tenth to one-half of the other materials.

White and Elsea[14] state that a Silal type iron with total carbon 2·2 per cent; silicon, 6·2 per cent; copper, 1·32 per cent; and chromium, 0·49 per cent was extremely resistant to growth and oxidation even at 900°C.

In the case of the figures by Williams, the total time of 300 h is that at which the castings were actually at temperature. The castings were actually cooled to air temperature approximately 13 times during the course of the test, and the time of cooling and heating up is not included.

In the case of Hallett[15] the specimens were exposed continuously to the temperature without air cooling. This probably accounts for the tendency for Williams' figures to come higher than those of Hallett. The general agreement, however, is reasonably satisfactory for this type of test, and serves clearly to indicate the order of merit of the various types of iron.

In general, an unalloyed iron containing 3·9 per cent of silicon had better general heat resisting properties than 2 per cent silicon irons containing up to 2 per cent chromium. The introduction of 2 per cent chromium reduces scaling up to 850°C at all levels of silicon up to 3·9 per cent by about one

* Annealing pots made of a 30 per cent chromium iron undergoing 10 cycles of 100 h at 1000–1050°C and stacked 6 ft (1·8 m) high have given satisfactory service[12]

† For 30 per cent chromium irons the 0·1 per cent total creep strength in 1000 h at 1000°C probably does not exceed 0·3 ton/in^2 (5 N/mm^2) in tension

Table 114 HEAT RESISTANCE OF ALLOY CAST IRONS AT TEMPERATURES OF 750–1050°C. LOSS OF WEIGHT AFTER SHOT BLASTING EXPRESSED AS G/DM² (KG/M²) AFTER EXPOSURE AT TEMPERATURE FOR TIMES INDICATED

TC %	Si %	Cr %	P %	750°C 300 h Loss of weight g/dm² (kg/m²)	800°C 500 h Loss of weight g/dm² (kg/m²)	800°C 500 h Growth %	850°C 300 h Loss of weight g/dm² (kg/m²)	850°C 300 h Growth %	950°C 300 h Loss of weight g/dm² (kg/m²)	950°C 300 h Growth %	1050°C 200 h Loss of weight g/dm² (kg/m²)
2·91–3·15	2·0 (1·9–2·17)	0	0·13–1·3	19* (1·9)	66 (6·6)	1·57	39·0 (3·9)	1·17	134 (13·4)	3·0	–
2·91–3·15	2·0 (1·9–2·17)	2 (1·90–2·26)	0·13–1·3	–	31·5 (3·15)	0·86	25·4 (2·54)	0·84	101 (10·1)	2·8	–
2·67–3·05	3·0 (2·87–3·18)	0	0·07–1·42	–	21·6 (2·16)	0·42	29·6 (2·96)	1·24	81·6 (8·16)	2·8	–
2·67–3·05	3·0 (2·87–3·18)	2 (1·89–2·29)	0·07–1·42	–	8·6 (0·86)	0·76	14·5 (1·45)	0·87	22·1 (2·21)	2·0	–
2·8–2·95	3·8 (3·6–3·95)	0	0·11–1·43	–	16 (1·6)	0·22	13·9 (1·39)	0·32	53 (5·3)	1·7	–
2·8–2·95	3·8 (3·6–3·95)	2 (1·85–2·16)	0·11–1·43	–	7·6 (0·76)	0·39	7·9 (0·79)	0·44	39 (3·9)	1·7	–
1·89	5·7	–	0·16	0·8 (0·08)	–	–	3 (0·3)	–	7 (0·7)	–	24–32 (2·4–3·2)

Aluminium Irons with Chromium (240 h)

TC %	Si %	P %	Al %	Cr %							
2·94	3·27	1·21	0·94	0·04	–	8·0 (0·8)	0·46	8·3 (0·83)	0·26	78 (7·8)	2·3	–
2·97	3·16	1·08	1·03	0·83	–	7·7 (0·77)	0·2	9·6 (0·96)	0·29	116 (11·6)	3·08	–
2·83	1·87	0·25	2·08	0·59	–	– –	–	80·5 (8·05)	3·6	– –	–	–
2·77	2·02	0·23	3·75	0·71	–	– –	–	5·2 (0·52)	0·71	– –	–	–
3·06	2·7	1·31	3·95	0·02	–	3·6 (0·36)	0·29	2·1 (0·21)	0·06	90 (9·0)	3·62	–
2·81	2·74	1·36	3·71	0·70	–	1·0 (0·10)	0·01	1·5 (0·15)	0·02	87 (8·7)	2·13	–
2·72	2·74	1·36	4·89	1·99	–	1·1 (0·11)	0·31	1·1 (0·11)	0·02	80 (8·0)	2·13	–
3·0	1·6	0·2	5·9–6·8	0·8	<1 (<0·1)	–	–	–	–	–	–	–

Ni-Resist

TC %	Si %	P %	Ni %	Cu %	Cr %							
2·8	2·54	0·04	15·05	7·37	1·41	4·4 (0·44)	–	–	–	–	–	–
2·74	1·93	0·04	14·62	7·16	2·11	4 (0·4)	–	–	9 (0·9)	–	16 (1·6)	–	30–32 (3–3·2)
2·85	1·85	–	14·38	7·36	4·84	3 (0·3)	–	–	6 (0·6)	–	9 (0·9)	–	16–20 (1·6–2·0)

Nicrosilal

TC %	Si %	Mn %	P %	Ni %	Cr %							
1·78	4·63	0·91	0·1	22·46	2·52	1·3 (0·13)	–	–	3·5 (0·35)	–	–	–	–
2·1	4·9	0·89	0·05	22·79	4·63	<0·5 (<0·005)	–	–	0·5–1 (0·05–0·1)	–	1·3 (0·13)	–	14–16 (1·4–1·6)

High Chromium Irons

TC %	Si %	Mn %	Ni %	Cr %							
2·14	1·54	1·56	0·2	16·6	–	–	–	<1 (<0·1)	–	1·2 (0·12)	–	3–4 (0·3–0·4)
1·28	1·17	0·75	0·24	33·03	–	–	–	–	–	–	–	<1 (<0·1)
1·12	1·78	0·57	9·36	29·62	–	–	–		–	–	–	<2 (<0·2)

* Scaling at or below 500 C is unlikely to exceed 1·0 g/dm² (0·1 kg/m²) after 10 000 h. Growth <1·0 per cent.

Table 115 SHORT TERM MECHANICAL PROPERTIES OF DIFFERENT MATERIALS AT TEMPERATURES UP TO 700°C

	TC %	*Si* %	*Mn* %	*P* %	*S* %	*Cr* %	*Mo* %	*Ni* %	*Mg* %		*20°C*	*200°C*	*300°C*	*400°C*	*500°C*	*600°C*	*700°C*
Pearlitic grey iron	3·06	3·29	0·66	0·26	0·12	—	—	—	—	Tensile ton/in²	21·5	20·2	20·8	19·2	12·9	4·9	—
										N/mm²	332	312	321	297	199	76	
	3·42	2·80	0·69	0·61	0·12	—	—	—	—	Tensile ton/in²	11·8	11·5	11·9	11·5	7·4	3·3	—
										N/mm²	182	178	184	178	114	51	
Malleable iron	2·86	0·68	0·43	0·074	0·19	0·05	—	—	—	Tensile ton/in²	26·1*	26·4	20·7	20·3	10·6	8·0	—
										N/mm²	403	408	320	314	164	124	
		0·71								0·2% Proof stress	16·3	16·4	13·3	10·8	6·9	3·6	—
											252	253	205	167	107	56	
Molybdenum malleable iron	3·20	0·43	0·29	0·068	0·21	0·03	0·60	—	—	Tensile ton/in²	22*	25	20·9	19·3	14·8	9·6	—
										N/mm²	340	386	323	298	229	148	
										0·2% Proof stress	20	17·4	15·4	15·3	9·9	5·5	—
											309	269	238	236	153	85	
Nodular iron	3·35	2·06	0·35	0·066	0·002	—	—	0·48	0·06	Tensile ton/in²	34·2*	31	32·4	27·5	22·6	10·3	4·3
										N/mm²	528	479	500	425	349	159	66
										0·2% Proof stress	28	26·2	27·5	22·6	13·5	4·9	1·7
											432	405	425	349	208	76	26
Silal	1·8–2·0	5·5–7·0	0·60	0·1–0·2	<0·1	—	—	—	—	Tensile ton/in²	15–18	—	—	—	18	13·8	5·2
										N/mm²	232–278	—	—	—	278	213	80
Nicrosilal	1·7–2·0	4·5–5·0	0·6–1·0	0·1	<0·12	2·0–4·0	—	18–20	—	Tensile ton/in²	16–17	—	—	—	12·4	12·0	10·8
										N/mm²	247–263				192	185	167

* Elongation: nodular iron, 1%–2%; malleable iron, 2%–3%; Mo malleable iron, 0·5%–1%.

half. This improvement, of course, is more significant in the lower silicon and phosphorus irons. Very significant improvements are obtained by the addition of 3 per cent aluminium, for temperatures at least up to 850°C. It seems desirable for the low aluminium irons to contain more than 2 per cent silicon, and probably more than 2·5 per cent silicon, for the best results, and the presence of 0·7–1·0 per cent chromium is also an advantage. One per cent aluminium with 3 per cent silicon and 0·8 per cent chromium also appears to have useful properties. The service of aluminium irons, as distinct from laboratory tests, has not yet been confirmed, and some caution should be exercised in their use, particularly in view of the risk of introducing aluminium through back scrap into other castings, which may introduce pinholing defects.

Mechanical properties at high temperatures

Up to temperatures of 300–400°C, cast irons of Grade 14 and above are capable of carrying sustained loads up to 2 ton/in^2 (31 N/mm^2) tensile without more than 0·1 per cent total creep strain in 30 000 h. As the temperature rises, however, the sustained load which can be carried without distortion falls very rapidly. This is not always apparent from short-term tensile tests as, for example, in *Table 121* where short-term tensile figures at 500°C of 7·4 and 12·9 ton/in^2 (114 and 199 N/mm^2) are given for grey irons. When tested under sustained loads, however, such irons would show some creep after 1000 h at 1 ton/in^2 (15 N/mm^2).

It is customary, therefore, to express the high temperature strength of iron or steel in terms of:

(1) Stress to produce rupture in 100 000, 30 000 or 10 000 h.
(2) Stress to produce 0·1 per cent total creep strains in 100 000, 30 000 or 10 000 h.

SHORT-TERM TESTS

The published work in this field for cast iron is limited, but the information is summarised in *Table 115*.

Another typical short-term tensile [1 ton/in^2/min loading rate][16] shows the results presented in *Table 116* for a pearlitic iron of composition: total carbon, 3·05 per cent; silicon, 2·13 per cent; manganese, 0·66 per cent;

Table 116

Temperature	*150°C*	*200°C*	*400°C*	*500°C*	*600°C*	*700°C*	*After 100 h at 700°C before test, ferritic*
Tensile							
ton/in^2	19	19·8	20·4	16·0	11·0	6·6	3·0
N/mm^2	293	306	315	247	170	102	46

sulphur, 0·065 per cent. These figures differ appreciably from those given in *Table 115* for a different grade of iron.

Additions of vanadium or chromium may increase the hot strength under these conditions by 25 per cent to 30 per cent.

LONG-TERM CREEP DATA

Long-term creep data are given at the end of this section.

HARDNESS AT HIGH TEMPERATURES

A steady reduction in hardness occurs as the temperature rises (*Table 117*), and the following short-term figures have been obtained for two irons using 10 mm ball, 3000 kg load, load applied for 30 s, and the impression measured at the temperatures of test[17].

Table 117

BHN	*15°C*	*100°C*	*200°C*	*300°C*	*400°C*	*500°C*	*600°C*	*700°C*
1	252	241	231	233	225	194	139	—
2	211	201	191	186	174	163	100	27

EFFECT OF TEMPERATURE ON FATIGUE STRENGTH

Curves showing the effect of temperature on the fatigue strength of three different irons were produced by Fitzgeorge and Pope[18] and are reproduced in *Figure 59* on page 98. It will be observed that these values for reversed bending are actually higher than for 0·1 per cent total creep per 100 000 h for temperatures above 300°C under steady load.

COMPRESSIVE STRENGTH OF CAST IRON AT HIGH TEMPERATURES

Tests were carried out by Ingberg[19] and Sale[20] on cast iron cylindrical columns of the following analysis:

	%
Total carbon	3·52
Silicon	1·58
Manganese	0·37
Sulphur	0·115
Phosphorus	0·72
Carbon equivalent value	4·19

Tests at room temperature of this iron gave the results shown in *Table 118.*

Table 118

	Maximum compressive stress	*Transverse modulus of rupture*	*Limit of proportionality*	*Elastic modulus*
ton/in^2	39	27·9	17·9	—
lb/in^2	87 000	60 000	40 000	$16{\cdot}6 \times 10^6$
N/mm^2	602	431	276	(114 GN/m^2)

From the analysis, the tensile of a 1·2 in (30 mm) diameter as-cast bar would be expected to be 13–14 ton/in^2 (201–216 N/mm^2) at room temperature.

The test bar was a hollow, round specimen with an outside diameter of $1\frac{1}{2}$ in (38 mm) and an inside diameter of $\frac{1}{2}$ in (13 mm).

The tests were taken on specimens $9\frac{1}{4}$ in (235 mm) long, of which a 6 in (152 mm) length was maintained at temperature. *Figure 96* shows the change in properties with rising temperatures. The yield point was selected as the point at which there was a deviation of 0·001 in strain from the initial modulus line.

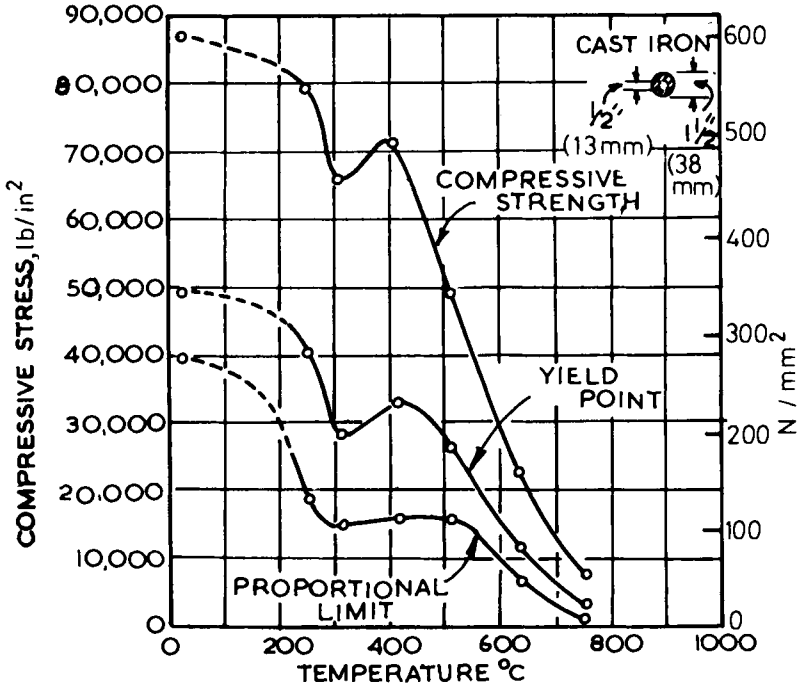


Figure 96. Proportional limit, yield point and compressive strength of cast iron tested in compression at constant temperatures

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STRESS RESISTANCE AT HIGH TEMPERATURES

Creep and stress to rupture

At any temperature, deformation of cast iron and other materials is dependent upon time as well as stress, and a specimen at a sufficiently high stress may continue to creep or deform plastically and permanently the longer the time of stress application. This is indicated in the section on stress/strain relationship. Tapsell[1] states '... creep may be defined as the deformation of a material occurring with time under and due to an externally applied load. As defined, creep is not meant to include dimensional changes with time due to internal forces consequent on structural or constitutional changes.' This definition does not mention temperature, but in most engineering materials creep is not appreciable under normal working stress except at relatively high temperatures, although the temperature level at which creep of significant magnitude occurs varies with the class of material[2] although with cast iron measurable creep can occur at room temperature with sufficiently high stress and may not reach a limiting value until periods up to 50–200 h have been reached (*see* section on stress/strain).

The standard form of creep curve of strain[2] plotted against time is shown in *Figure 97*. If a stress below the proportional limit is applied at room temperature to a steel, deformation will occur immediately on application of load and will remain constant within normal measurable limits regardless of the

length of time and the strain-time characteristics will be represented by the line OAB in *Figure 97.*

At some higher temperature an elastic deformation OC plus possibly a plastic deformation CD will occur immediately on application of load and the deformation will increase with time. This deformation usually follows a

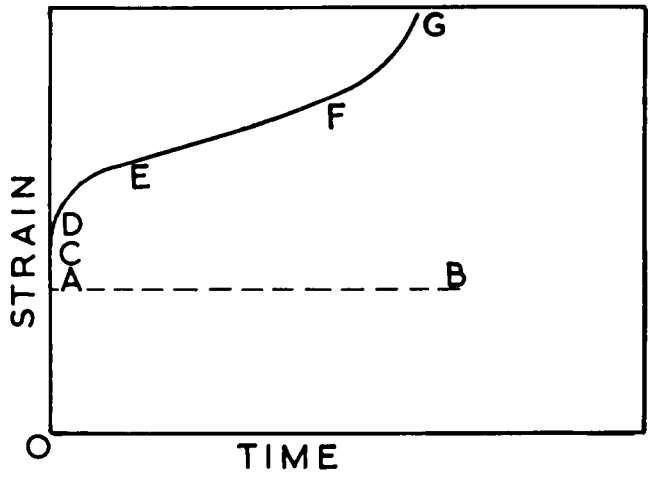


Figure 97. Classical creep curve of strain plotted against time

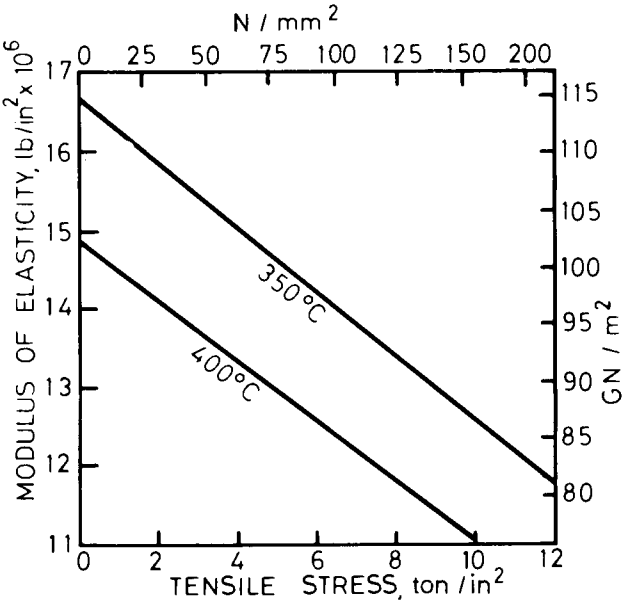


Figure 98. Variation of modulus of elasticity with tensile stress at 350 and 400°C of unalloyed grey cast iron. Tensile strength 16·6 ton/in² (256 N/mm²)

curve of the type CG. This is the generally accepted classical creep curve and is usually considered to be divided into three parts:

CE—The primary stage—a period of decreasing creep rate.
EF—The secondary stage where the creep rate is constant although in practice it is frequently only a point of inflexion.
FG—The tertiary stage, a period of increasing creep rate which ends in fracture.

In this country it has been agreed that a standard creep curve should show the total plastic or permanent strain, excluding elastic or recoverable strain not proportional to stress, plotted against time. For this reason with cast iron the modulus of elasticity must be determined at each test temperature and stress. The variation of modulus of elasticity with stress is shown in *Figure 98* at 350 and 400°C. This enables the recoverable strain to be determined at any creep test stress which can then be subtracted from the total strain recorded, giving the total permanent strain.[3].

In some earlier work it was common to report creep results in terms of designated creep rate at a given stress, e.g. 0·1 per cent limiting creep stress per 1000 h. These figures, however, were based on secondary creep rates, and it is now generally accepted that creep results should be based on total creep strain and not on creep rate.

Earlier results based on creep rate must be interpreted with caution, but stress-to-rupture values are valid.

Curves for varying percentages of total creep strain in tensile are shown expressed as percentage total plastic strain plotted against time in hours on a logarithmic scale for grey cast iron in *Figure 99*[5].

Stress to rupture

If the applied stress is sufficiently high at the temperature of test for the tertiary stage of creep to develop, rupture will ultimately occur if the stress is applied for sufficient time. Larson and Miller[4] have developed a relation which evaluates the relative effects of time and temperature which can be adapted to rupture and creep behaviour: $T(20+\log t)=$ constant, where $T=$ temperature in °K and $t=$ time in hours, which is based upon rate-process theories. The constant 20 may not be universally applicable, but appears to apply to most metals with sufficient accuracy.

Using this relationship it is possible to make use of short-time tests to determine long-time properties, and a series of curves is given for various cast irons which makes it possible to predict the stress to rupture in times up to 100 000 h for a range of temperatures. These curves are based on tests of duration up to 30 000 h and include short-time tests and the results are remarkably consistent from various workers where the materials under test are comparable.

This method of plotting appears to be valid also for total creep strain where information is available.

The curves for grey irons and alloyed grey irons in *Figure 99*[5] are plotted

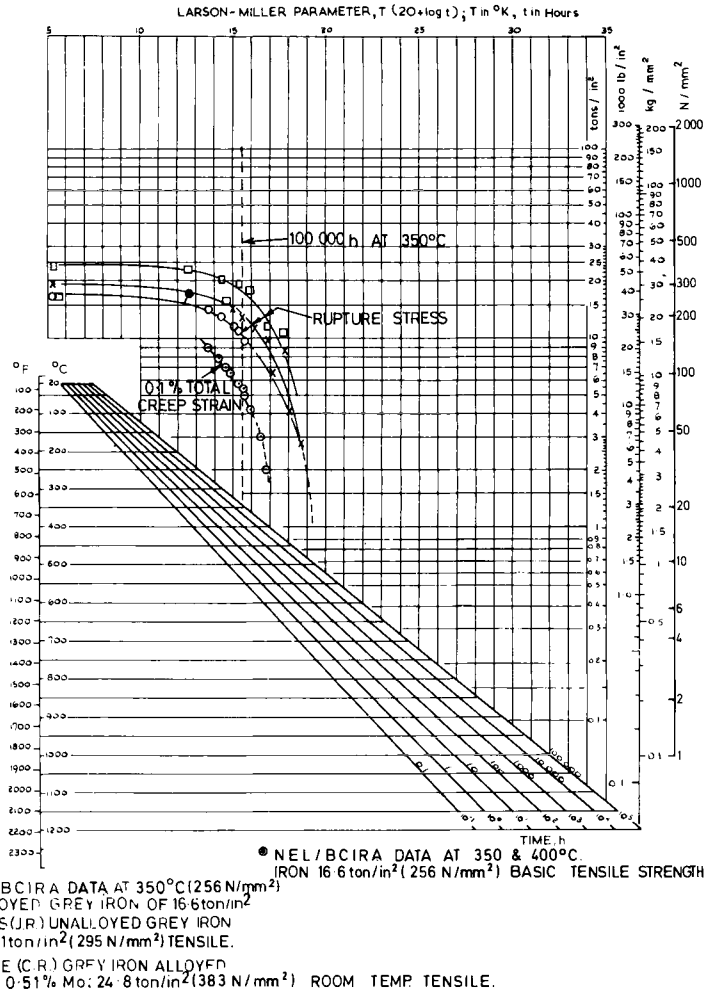


*Figure 99. Rupture strength, grey iron unalloyed 16–19 ton/in² (247–293 N/mm²) tensile; alloyed 0·51% molybdenum 24·8 ton/in² (383 N/mm²) tensile. N.E.L./BCIRA data from 1, 2 and 3 in diameter bars (25, 51, 76 mm). **Tottle** data from 0·875 in (22 mm) diameter*

by the method devised by the National Engineering Laboratory in which time, temperature and stress are included on one curve to the Larson–Miller parameter. Short-time tensile tests up to 600°C in which duration of test is assumed to be 0·1 h, fall consistently with little scatter on the curves for rupture tests up to 400°C and with rupture times up to 30 000 h. Therefore, short-time tensile tests at varying temperatures can be used to give a preliminary assessment of the suitability of materials for long-term stressed conditions at various temperatures.

Similar curves for nodular, malleable, austenitic irons were published by Angus[5] together with the associated data and references.

Permissible maximum stress

Figures for maximum permissible stress in service can be based upon (a) a safety factor against rupture, or (b) a safety factor against permanent distortion under load where dimensional tolerances are imperative.

A factor of safety should take care of scatter in experimentally determined values and of accidental loads over and above the nominal working load.

It is suggested that one-third of the stress to rupture in 100 000 h at 350°C is a suitable criterion for the maximum permissible working stress. In the case of grey and nodular iron for which evidence is available, this stress is below that required to produce 0·1 per cent total creep strain in 100 000 h at 350°C and should therefore provide a reasonable factor of safety against distortion or rupture.

One-fifth of the stress in 100 000 h at 350°C would provide a more conservative factor of safety.

The value of 350°C was used in relation to steam engineering but values up to 450°C are quoted in *Tables 120–123*; above this temperature oxidation becomes significant.

The following values (*Table 119*) are suggested on the basis of these criteria, based on BCIRA work[5, 6].

Table 119

Material	*Stress to rupture in 100 000 h at 350°C* ton/in²	(N/mm²)	*Maximum recommended stress at 350°C* ton/in²	(N/mm²)
Factor of safety			1/5	1/3
Grey iron to BS 1452:1961				
Grade 17–20	9–10·5	(139–162)	2 (31)	3–3·5 (46–54)
Grades 23 and 26	14·5	(224)		4·8 (74)
Nodular or SG iron to BS 2789:1961:				
Ferritic Grades SNG 24/17	12·4	(192)	2·4 (37)	4 (62)
SNG 27/12	12·4	(192)	2·4 (37)	4 (62)
Pearlitic Grade SNG 42/2	22·3	(344)	4·4 (68)	7 (108)
Blackheart malleable iron to				
BS 310:1958 Ferritic B22/14	8·9	(137)	1·8 (28)	3 (46)
BS3333:1961 Pearlitic P28/6	13·2	(204)	2·8 (43)	4·4 (68)
P33/4	13·2	(204)	2·8 (43)	4·4 (68)
Whiteheart malleable iron to BS 309:1958, Grades 22/4 and 24/8	8·9	(137)	1·8 (28)	3 (46)
High chromium cast irons ferritic				
24% Cr:1% TC:1% Si	20–25	(308–386)	4·5 (62)	7–8 (108–124)
34% Cr:2·1% TC:1% Si	20–25	(308–386)	4·5 (62)	7–8 (108–124)
Austenitic cast irons BS 3468:1962				
Flake graphite Ni Si Cu Cr	(AUS 101) 15:1·6:6:2 (Mn 1–1·5)			2·6 (40)
Ni Si Cr	(AUS 102) 20:2:2 (Mn 1–1·5)			2·8 (43)
Ni Si Cr	(AUS 102) 20:2:3 (Mn 1–1·5)			3·6 (56)
Ni Si Cr	(AUS 104) 20:5:3 (Mn 1–1·5)			3·7 (57)
Ni Si Cr	(AUS 105) 30:1·5:3 (Mn 0·4–0·8)			3·6 (56)
Ni Cr	30:3 (Si 1–2; Mn 0·4–0·8)			3·3 (51)
Nodular Properties Ni Cr	(AUS 202) 20:2 (Si 1–2·8, Mn 1·2)			4·7 (73)
Ni Si	(AUS 203) 22·5:2 (Cr >0·5, Mn 2·1)			4·2 (65)
Ni Si Cr	(AUS 204) 20:5:2 (Mn 1·2)			5·0 (77)
Ni Si Cr	(AUS 205) 30:2:3 (Mn up to 0·5)			4·7 (73)

Table 120 DESIGN STRESSES FOR GREY CAST IRONS[6] — IMPERIAL UNITS AND SI UNITS. RECOMMENDED MAXIMUM TENSILE DESIGN STRESSES IN THE TEMPERATURE RANGE −150°C TO 450°C

Grey iron castings to BS 1452:1961	*Maximum tensile design stress at stated temperature (°C)*																
	−150	−100	−50	−25	0	20	50	100	150	200	250	300	350	375	400	425	450
Tensile strength, 10 ton/in^2	1·3	1·9	2·5	2·5	2·5	2·5	2·4	2·3	2·2	2·1	2·0	1·9	—	—	—	—	—
154 N/mm^2	20	29	39	39	39	39	37	36	34	32	31	29	—	—	—	—	—
Tensile strength, 12 ton/in^2	1·5	2·3	3·0	3·0	3·0	3·0	2·9	2·8	2·7	2·6	2·5	2·4	—	—	—	—	—
185 N/mm^2	23	36	46	46	46	46	45	43	42	40	39	37	—	—	—	—	—
Tensile strength, 14 ton/in^2	1·8	2·7	3·5	3·5	3·5	3·5	3·4	3·3	3·2	3·1	3·0	2·9	3·2	2·5	1·9	1·4	1·1
216 N/mm^2	28	42	54	54	54	54	53	51	49	48	46	45	49	39	29	22	17
Tensile strength, 17 ton/in^2	2·2	3·2	4·3	4·3	4·3	4·3	4·2	4·1	4·0	3·9	3·8	3·7	3·3	2·7	2·0	1·5	1·2
263 N/mm^2	34	49	66	66	66	66	65	63	62	60	59	57	51	42	31	23	19
Tensile strength, 20 ton/in^2	2·5	3·8	5·0	5·0	5·0	5·0	4·9	4·8	4·7	4·6	4·5	4·4	3·5	2·8	2·1	1·6	1·2
309 N/mm^2	39	59	77	77	77	77	76	74	73	71	69	68	54	43	32	25	19
Tensile strength, 23 ton/in^2	2·9	4·4	5·8	5·8	5·8	5·8	5·7	5·6	5·5	5·4	5·3	5·2	4·8	4·3	3·3	2·7	2·0
355 N/mm^2	45	68	90	90	90	90	88	86	85	83	82	80	74	66	51	42	31
Tensile strength, 26 ton/in^2	3·3	4·9	6·5	6·5	6·5	6·5	6·4	6·3	6·2	6·1	6·0	5·9	4·8	4·3	3·3	2·7	2·0
402 N/mm^2	51	76	100	100	100	100	99	97	96	94	93	91	74	66	51	42	31
BASIS FOR DESIGN STRESS	Equal to design stress at 20°C to −50°C and then decreasing to half this value at −150°C						Based on 0·384 × 0·1 per cent proof stress value						Based on one-third stress to produce rupture in 100 000 h[5]				

Table 121 DESIGN STRESSES FOR MALLEABLE CAST IRONS[6] — IMPERIAL UNITS AND SI UNITS. RECOMMENDED MAXIMUM TENSILE DESIGN STRESSES IN THE TEMPERATURE RANGE −75°C TO 450°C

Malleable iron castings	*Maximum tensile design stress ton/in² and N/mm² at stated temperature (°C)*															
	−75	−50	−25	0	20	50	100	150	200	250	300	350	375	400	425	450
BLACKHEART MALLEABLE IRONS TO BS 310:1958																
B 18 6. ton in²	6·3	6·3	6·3	6·3	6·3	6·0	5·8	5·6	5·4	5·2	5·0	3·0	2·4	1·9	1·5	1·2
N mm²	97	97	97	97	97	93	90	86	83	80	77	46	37	29	23	19
B 20 10. ton in²	7·0	7·0	7·0	7·0	7·0	6·8	6·6	6·4	6·2	5·9	5·7	3·0	2·4	1·9	1·5	1·2
N mm²	108	108	108	108	108	105	102	99	96	91	88	46	37	29	23	19
B 22 14. ton in²	7·7	7·7	7·7	7·7	7·7	7·5	7·3	7·1	6·9	6·7	6·5	3·0	2·4	1·9	1·5	1·2
N mm²	119	119	119	119	119	116	113	110	107	103	100	46	37	29	23	19
WHITEHEART MALLEABLE IRON TO BS 309:1958																
W 22/4 (Over $\frac{5}{16}$ in. i.e. 8 mm). ton/in²	—	5·2	5·2	5·2	5·2	5·0	4·8	4·5	4·3	4·1	3·9	3·0	2·4	1·9	1·5	1·2
N/mm²		80	80	80	80	77	74	69	66	63	60	46	37	29	23	19
W 22/4 ($\frac{5}{16}$ in. i.e. 8 mm and under). ton/in²	—	4·7	4·7	4·7	4·7	4·5	4·3	4·1	3·8	3·6	3·4	3·0	2·4	1·9	1·5	1·2
N/mm²		73	73	73	73	69	66	63	59	56	53	46	37	29	23	19
W 24/8 (Over $\frac{5}{16}$ in. i.e. 8 mm). ton/in²	—	5·7	5·7	5·7	5·7	5·5	5·3	5·0	4·8	4·6	4·4	3·0	2·4	1·9	1·5	1·2
N/mm²		88	88	88	88	85	82	77	74	71	68	46	37	29	23	19
W 24/8 ($\frac{5}{16}$ in. i.e. 8 mm and under). ton/in²	—	5·2	5·2	5·2	5·2	5·0	4·8	4·5	4·3	4·1	3·9	3·0	2·4	1·9	1·5	1·2
N/mm²		80	80	80	80	77	74	69	66	63	60	46	37	29	23	19
PEARLITIC MALLEABLE IRON TO BS 3333:1961																
Tensile strength 28·5 ton/in² (cooled in still air or as-annealed)	—	—	—	—	7·8	7·6	7·4	7·2	6·9	6·7	6·5	4·4	3·5	2·6	2·0	1·5
440 N/mm²					120	117	114	111	107	103	100	68	54	40	31	23
Tensile strength 28·5 ton/in² (air blast quenched)		—	—	—	7·9	7·7	7·4	7·2	7·0	6·7	6·5	4·4	3·5	2·6	2·0	1·5
440 N/mm³					122	119	114	111	108	103	100	68	54	40	31	23
Tensile strength 33·0 ton/in² (cooled in still air or as-annealed)	—	—	—	—	8·7	8·4	8·1	7·8	7·6	7·3	7·0	4·4	3·5	2·6	2·0	1·5
510 N/mm²					134	130	125	120	117	113	108	68	54	40	31	23
Tensile strength 33·0 ton/in² (air blast quenched)	—	—	—	—	9·3	9·0	8·7	8·4	8·1	7·9	7·6	4·4	3·5	2·6	2·0	1·5
510 N/mm²					144	139	134	130	125	122	117	68	54	40	31	23
Tensile strength 35·0 ton/in² (cooled in still air or as-annealed)	—	—	—	—	9·0	8·8	8·5	8·2	8·9	7·6	7·4	4·4	3·5	2·6	2·0	1·5
541 N/mm²					139	136	131	127	137	117	113	68	54	40	31	23
Tensile strength 35·0 ton/in² (air blast quenched)	—	—	—	—	9·9	9·7	9·4	9·1	8·8	8·5	8·3	4·4	3·5	2·6	2·0	1·5
541 N/mm²					153	150	145	141	136	131	128	68	54	40	31	23
Tensile strength 36·9 ton/in² (air blast quenched)	—	—	—	—	10·6	10·2	9·9	9·6	9·2	8·9	8·5	5·8	4·7	3·4	2·5	1·8
570 N/mm²					164	158	153	148	142	137	131	90	73	53	39	28
Tensile strength 36·9 ton/in² (oil quenched)	—	11·7	11·7	11·7	11·7	11·4	11·1	10·7	10·4	10·1	9·7	5·8	4·7	3·4	2·5	1·8
570 N/mm²		181	181	181	181	176	171	165	161	156	150	90	73	53	39	28
Tensile strength 44·7 ton/in² (oil quenched)		—	—	—	15·3	14·9	14·6	14·2	13·9	13·6	13·2	5·8	4·7	3·4	2·5	1·8
690 N/mm²					236	230	225	219	215	210	204	90	73	53	39	28
BASIS FOR DESIGN STRESS	Equal to design stress at 20 C for recommended irons at the temperatures shown. Not recommended shown by ' '				Based on 0·56 of the 0·1 per cent proof stress value for blackheart malleable iron and 0·45 of the 0·1 per cent proof stress value in other malleable irons							Based on one-third stress to produce rupture in 100 000 h[5]				

Table 122 DESIGN STRESSES FOR SPHEROIDAL OR NODULAR GRAPHITE IRONS[6]—IMPERIAL UNITS AND SI UNITS. RECOMMENDED MAXIMUM TENSILE DESIGN STRESSES IN THE TEMPERATURE RANGE −75 TO 450°C

Irons with spheroidal or nodular graphite to BS 2789:1961	*Maximum tensile design stress at stated temperature (°C)*															
	−75	−50	−25	0	20	50	100	150	200	250	300	350	375	400	425	450
SNG 24/17, ton/in²	8·1	8·1	8·1	8·1	8·1	7·8	7·5	7·2	6·9	6·7	6·4	4·1	3·2	2·5	1·9	1·4
N/mm²	125	125	125	125	125	120	116	111	107	103	99	63	49	39	29	22
SNG 27/12, ton/in²	—	8·9	8·9	8·9	8·9	8·6	8·3	8·1	7·8	7·5	7·3	4·1	3·2	2·5	1·9	1·4
N/mm²		137	137	137	137	133	128	125	120	116	113	63	49	39	29	22
SNG 32/7, ton/in²	—	—	9·4	9·4	9·4	9·1	8·9	8·7	8·5	8·2	8·0	4·6	3·6	2·7	2·0	1·5
N/mm²			145	145	145	141	137	134	131	127	124	71	56	42	31	23
SNG 37/2, ton/in²	—	—	—	—	10·0	9·8	9·5	9·3	9·1	8·9	8·6	5·7	4·4	3·3	2·3	1·7
N/mm²					154	151	147	144	141	137	133	88	68	51	36	26
SNG 42/2, ton/in²	—	—	—	—	10·4	10·2	9·9	9·7	9·5	9·3	9·0	7·4	5·7	4·2	2·8	2·1
N/mm²					161	158	153	150	147	144	139	114	88	65	43	32
SNG 47/2, ton/in²	—	—	—	—	11·6	11·4	11·2	10·9	10·7	10·5	10·3	7·4	5·7	4·2	2·8	2·1
N/mm²					179	176	173	168	165	162	159	114	88	65	43	32
BASIS FOR DESIGN STRESS	Equal to design stress at 20°C for recommended irons at the temperatures shown. Not recommended shown by '—'					Based on 0·56 and 0·52 of 0·1 per cent proof stress value for Grades SNG 24/17 and SNG 27/12 respectively and a factor of 0·45 for the other Grades						Based on one-third stress to produce rupture in 100 000 h[5]				

Table 123 DESIGN STRESSES FOR AUSTENITIC CAST IRONS AT HIGH TEMPERATURES[6]—IMPERIAL UNITS AND SI UNITS. RECOMMENDED MAXIMUM TENSILE DESIGN STRESSES IN THE TEMPERATURE RANGE 20°C TO 675°C

Grade BS 3468:1962		*Maximum tensile design stress ton in^{-2} and N mm^{-2} at stated temperature (°C)*																	
		20	50	100	150	200	250	300	350	400	450	500	525	550	575	600	625	650	675
FLAKE OR LAMELLAR GRAPHITE IRONS																			
L–Ni Cu Cr 15 6 2, AUS 101A,	ton/in^2	3·5	3·4	3·2	3·1	3·0	2·8	2·7	2·6	2·4	2·3	—	—	—	—	—	—	—	—
	N/mm^2	54	53	49	48	46	43	42	40	37	36								
L–Ni Cr 20 2 (1% Cr), AUS 102A,	ton/in^2	3·2	3·1	2·9	2·8	2·7	2·5	2·4	2·3	2·1	2·0	—	—	—	—	—	—	—	—
	N/mm^2	49	48	45	43	42	39	37	36	32	31								
L–Ni Cr 20 2 (2% Cr), AUS 102A,	ton/in^2	3·7	3·6	3·5	3·3	3·2	3·1	2·9	2·8	2·7	2·5	—	—	—	—	—	—	—	—
	N/mm^2	57	56	54	51	49	48	45	43	42	39								
L–Ni Cr 20 3 (3% Cr), AUS 102B,	ton/in^2	4·5	4·4	4·2	4·1	4·0	3·8	3·7	3·6	3·4	3·3	—	—	—	—	—	—	—	—
	N/mm^2	69	68	65	63	62	59	57	56	53	51								
L–Ni Si Cr 20 5 3, AUS 104,	ton/in^2	4·7	4·5	4·4	4·3	4·1	4·0	3·9	3·7	3·6	3·5	—	—	—	—	—	—	—	—
	N/mm^2	73	69	68	66	63	62	60	57	56	54								
L–Ni Cr 30 3, AUS 105,	ton/in^2	4·3	4·1	4·0	3·9	3·7	3·6	3·5	3·3	3·2	3·1	—	—	—	—	—	—	—	—
	N/mm^2	66	63	62	60	57	56	54	51	49	48								
NODULAR OR SPHEROIDAL GRAPHITE IRONS																			
S–Ni Cr 20 2 (1% Cr), AUS 202A,	ton/in^2	5·2	5·0	4·9	4·8	4·6	4·5	4·4	4·2	4·1	4·0	—	—	—	—	—	—	—	—
	N/mm^2	80	77	74	74	71	69	68	65	63	62								
S–Ni Cr 20 2 (2% Cr), AUS 202A,	ton/in^2	5·6	5·5	5·4	5·2	5·1	4·0	4·8	4·7	4·5	4·4	—	—	—	—	—	—	—	—
	N/mm^2	86	85	83	80	79	62	74	73	69	68								
S–Ni Cr 20 3 (3% Cr), AUS 202B,	ton/in^2	5·9	5·8	5·7	5·5	5·4	5·3	5·1	5·0	4·9	4·7	3·2	2·6	2·1	1·6	1·3	1·1	0·8	0·7
	N/mm^2	91	90	88	85	83	82	79	77	76	73	49	40	32	25	20	17	12	11
S–Ni Cr 20 3 (2% Cr, 1% Mo),	ton/in^2	No Results. Take as for S–Ni Cr 20 3										3·9	3·1	2·5	2·0	1·6	1·3	1·1	0·9
	N/mm^2											60	48	39	31	25	20	17	14
S–Ni Si Cr 20 5 2, AUS 204,	ton/in^2	5·9	5·8	5·6	5·5	5·4	5·2	5·1	5·0	4·8	4·7	2·3	1·9	1·5	1·2	0·9	0·7	0·6	0·5
	N/mm^2	91	90	86	85	83	80	79	77	74	73	36	29	23	19	14	11	9	8
S–Ni 22, AUS 203,	ton/in^2	5·2	5·0	4·9	4·8	4·6	4·5	4·4	4·2	4·1	4·0	2·3	1·9	1·5	1·2	0·9	0·7	0·6	0·5
	N/mm^2	80	77	76	74	71	69	68	65	63	62	36	29	23	19	14	11	9	8
S–Ni Mn 23 4,	ton/in^2	5·0	4·9	4·8	4·6	4·5	4·4	4·3	4·1	4·0	3·8	—	—	—	—	—	—	—	—
	N/mm^2	77	76	74	71	69	68	66	63	62	59								
S–Ni Cr 30 3, AUS 205,	ton/in^2	5·7	5·5	5·4	5·3	5·1	5·0	4·9	4·7	4·6	4·5	3·2	2·6	2·1	1·6	1·3	1·1	0·8	0·7
	N/mm^2		85	83	82	79	77	76	73	71	69	49	40	32	25	20	17	12	11
S–Ni Si Cr 30 5 5,	ton/in^2	No Results. Take as for S–Ni Si Cr 20 5 2										2·3	1·9	1·5	1·2	0·9	0·7	0·6	0·5
	N/mm^2											36	29	23	19	14	11	9	8
BASIS FOR DESIGN STRESS		Based on 0·45 of the 0·1 per cent proof stress value										Based on one-third stress to produce rupture in 100 000 h[5]							

Design stresses at elevated and sub-zero temperatures

The previous tables[6] (*Tables 120–123*) give the estimated maximum design stress for sub-zero temperatures from −75 to +300°C based on 0·1 per cent proof stress for grey, nodular and malleable cast irons, and from 350 to 450°C, based on one-third of the stress to produce rupture in 100 000 h.

For austenitic cast irons the maximum design stress is based on 0·45 of the 0·1 per cent proof stress from 20 to 450°C, and on one-third stress to produce rupture in 100 000 h from 500 to 675°C.

The 0·1 per cent proof stress and stress to rupture in 100 000 h at the various temperatures can be derived from these figures.

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LOW TEMPERATURE PROPERTIES

Probably the most comprehensive series of tests upon the effect of low temperatures was made by Gilbert[1, 2], and most of the following data are derived from this work. For experimental reasons the impact tests were carried out on non-standard specimens [$\frac{5}{8}$ in (16 mm) and $\frac{3}{8}$ in (10 mm) diameter × 3 in (76 mm) span Charpy], but the order of change in impact strength with temperature can be regarded as typical.

Grey cast iron

Tests were made on fifteen irons cast in green sand in bars 0·875 in (22 mm) and 1·2 in (30 mm) in diameter, of which the following tests are typical. Compositions are shown in *Table 124* and properties in *Table 125*.

It will be seen that irons 10 and 14 varied little in tensile down to −100°C, but that the other materials showed increases up to 14 per cent of the room temperature strength.

A further series of tests by Gilbert[3] on flake and undercooled graphite irons is shown in *Table 126*.

In general, the impact strength fell progressively as the temperature dropped, the 2·13 per cent nickel iron showing the greatest falling off below −40°C.

A particularly interesting feature of this investigation was the indication of a transition temperature between two levels of brittle failure in flake

Table 124

No.	*ASTM graphite size*	*Matrix*	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Tensile strength* ton/in² (N/mm²)	*Brinell hardness*	*Impact (Charpy)* $\frac{5}{8}$ in dia. × 3 in (16 mm × 76 mm) ft.lb (J)
1	A4	Pearlite	3·20	1·90	0·69	0·03	0·039	–	15·2 (235)	207	8·3 (11)
10	A4–5	Pearlite	3·12	1·92	0·90	0·057	0·63	–	16·3 (252)	255	3·6 (5)
11	A5	Pearlite	3·13	2·31	0·69	0·141	0·67	–	17·4 (269)	244	4·6 (6)
14	A4	Fine pearlite	2·89	1·68	0·54	0·063	0·041	2·13	21·7 (335)	252	7·4 (10)
15	C3	Pearlite 30% ferrite	3·79	3·06	0·63	0·02	0·037	–	6·9 (107)	119	5·0 (7)

Table 125

No.	*Property*	*Temperature* °C						
		Room	*−20*	*−40*	*−60*	*−80*	*−100*	*−120*
1	Tensile, ton/in²	15·2	17·0	17·8	17·9	17·8	—	18·4
	N/mm²	235	263	275	276	275		
1	Impact, ft.lb	8·3	6·8	7·9	7·2	6·2	5·3	5·4
	J	11	9	11	10	8	7	7
10	Tensile, ton/in²	16·3	15·4	15·8	16·4	—	17·0	—
	N/mm²	252	238	244	253		263	
10	Impact, ft.lb	3·6	3·0	3·1	3·1	3·2	3·0	—
	J	5	4	4	4	4	4	
11	Tensile, ton/in²	17·4	18·4	18·4	18·8	18·3	—	19·3
	N/mm²	269	284	284	290	283		298
11	Impact, ft.lb	4·6	4·1	3·4	3·7	3·3	3·3	3·2
	J	6	6	5	5	4	4	4
14	Tensile, ton/in²	21·7	21·3	20·8	23·4	21·7	21·6	—
	N/mm²	335	329	321	361	335	334	
14	Impact, ft.lb	7·4	7·2	6·5	5·5	4·7	4·3	3·5
	J	10	10	9	7	6	6	5
15	Tensile, ton/in²	6·9	—	6·7	7·6	—	7·9	7·8
	N/mm²	107	—	103	117	—	122	120
15	Impact, ft.lb	5·0	4·8	4·9	3·9	3·3	3·1	3·0
	J	7	7	7	5	4	4	4

Table 126

	TC %	*Si* %	*Mn* %	*S* %	*P* %	*Ti* %	*Tensile strength* 20°C ton/in² (N/mm²)	*Tensile strength* −269°C ton/in² (N/mm²)
Pearlitic flake graphite	3·25	2·08	0·53	0·019	0·029	—	16·0 (247)	15·1 (233)
Undercooled graphite mainly ferritic	3·34	1·91	0·54	0·018	0·026	0·13	13·2 (204)	21·5 (332)

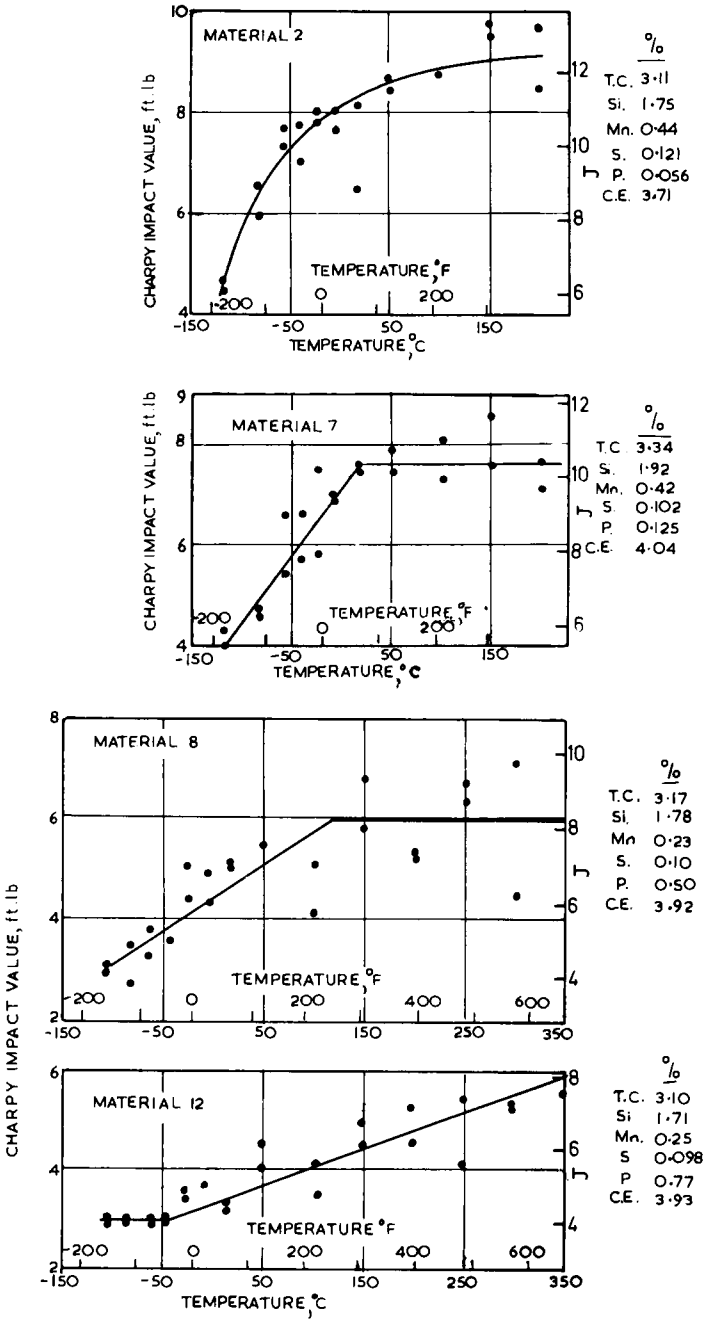


Figure 100. Impact tests at various temperatures showing transition temperatures on the flake graphite cast irons. Specimens $\frac{5}{8}$ in (16 mm) diameter tested on 3 in (76 mm) span

graphite cast irons, comparable with the change from ductile to brittle failure in steels and nodular irons. In this case, however, the transition occurred over a wide range of temperatures which appeared to increase with the phosphorus content[2] (*see Figure 100*).

FATIGUE STRENGTH OF GREY IRONS AT LOW TEMPERATURES

Boone and Wishart[4] determined the endurance limit of various metals, including cast iron, between 80°F (27°C) and −50°F (−46°C), using both notched and unnotched specimens (*Table 127*). The compositions of the materials are shown in *Table 128*.

Table 127 ENDURANCE LIMITS AND STRESS CONCENTRATION FACTORS FOR TWO GRADES OF IRON

Specimen temperature °C	*Endurance limit Unnotched specimen* lb/in² (N/mm²)		*Endurance limit Notched specimen* lb/in² (N/mm²)		*Stress concentration factor: Endurance limit, unnotched / Endurance limit, notched*	
	Grey cast iron	*Meehanite cast iron*	*Grey cast iron*	*Meehanite cast iron*	*Grey cast iron*	*Meehanite cast iron*
27	9000 (62)	21 000 (145)	10 000 (69)	20 000 (138)	0·90	1·05
10	9000 (62)	—	10 000 (69)	—	0·90	—
−20	9500 (66)	23 000 (159)	9500 (66)	22 000 (152)	1·00	1·05
−40	11 500 (79)	—	13 500 (93)	—	0·85	—

Endurance limit based on 10×10^6 cycles.

Table 128

	TC %	*Si* %	*Mn* %	*S* %	*P* %
Grey cast iron*	3·25	1·1	0·60	0·096	0·46
Meehanite cast iron	3·02	1·26	0·90	0·080	0·15

* Annealed 30 min at 1040°F (560°C) and cooled in the furnace.

Although the analysis and heat treatment would not be expected to yield a ferritic material, it seems probable that the grey iron was ferritic with a tensile strength not exceeding 10 ton/in² (154 N/mm²). The Meehanite cast iron would probably be pearlitic with a tensile strength above 20 ton/in² (309 N/mm²).

Low temperature properties of acicular, martensitic and austenitic grey irons

The matrix structure of these irons depends upon the lowering of the martensitic transformation by the use of alloys, and if the temperature falls sufficiently, the austenitic matrix may become irreversibly martensitic. *Tables 129* and *130* give the composition and phase change temperatures of some of these irons, together with the changes in mechanical properties at low temperatures.

Table 129

Type	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Cr* %	*Mo* %	*Cu* %	*Temperature of phase change* °C
Acicular	3·17	2·11	0·5	0·03	0·029	1·59	—	0·75	—	−48
Martensitic grey	2·81	1·81	0·52	0·05	0·046	4·87	—	—	—	−26
Austenitic Ni-Resist (a)	2·95	2·64	0·6	0·025	0·035	13·93	1·06	—	6·36	−10
(b)	2·65	2·43	0·48	0·016	0·051	15·52	1·41	—	8·65	−88

Table 130

Properties	*Temperature* °C						
	Room	*−20*	*−40*	*−60*	*−80*	*−100*	*−120*
ACICULAR							
Tensile, ton/in²	21·4	22·6	23·5	24·7	24·6	25·2	—
N/mm²	331	349	363	381	380	389	
Impact ($\frac{5}{8}$ in dia.)*, ft.lb	16–18	15·5	16·7	14·3	10·6	10·4	8·5
J	22–24	21	23	19	14	14	12
Brinell	229	235	229	266	272	271	280
MARTENSITIC							
Tensile, ton/in²	21·9	22·3	—	22·9	22·8	22·2	—
N/mm²	338	344	—	354	352	343	
Impact ($\frac{5}{8}$ in dia.)*, ft.lb	7–10	10·4	8·6	7·5	7·8	5·5	5·0
J	9–14	14	12	10	11	7	7
Brinell	309	325	329	359	380	395	404
AUSTENITIC (a)							
Tensile, ton/in²	9·1	9·7	12·1	12·6	16·3	17·2	—
N/mm²	141	150	187	195	252	266	—
Impact ($\frac{5}{8}$ in dia.)*, ft.lb	9·2	8·5	5·2	4·3	3·6	2·9	3·0
J	12	12	7	6	5	4	4
Brinell	89	93	118	128	179	191	197
AUSTENITIC (b)							
Tensile, ton/in²	9·8	—	10·2	10·8	12·7	14·8	17·5
N/mm²	151	—	158	167	196	229	270
Impact ($\frac{5}{8}$ in dia.)*, ft.lb	16·0	22·9	22·6	28·5	28·2	19·2	15·4
J	22	31	31	39	38	26	21
Brinell	97	103	102	108	105	143	184

* Specimen unnotched, 3 in span.

The impact strength increased very markedly as the temperature fell to the transformation point, but fell sharply at lower temperatures.

These changes may be regarded as typical of such irons.

These results show that with acicular and martensitic irons the tensile strength and hardness increase as the temperature falls and the impact strength falls steadily as the temperature falls, but when the temperature drops below the transformation temperature a sharper fall in impact strength occurs.

The first of the two austenitic irons chosen (a), owing to low alloy content, had a transition temperature not much below room temperature (−10°C), and shows a progressive rise in tensile and hardness as the transformation proceeds, while the impact strength falls off rapidly. The second (b), with a higher alloy content, had a much lower transformation temperature (−88°C), and showed a slow rise in tensile down to the transformation point and a sharp rise after transformation. The hardness showed little change until transformation occurred and a very sharp rise afterwards. The low temperature change in Ni-Resist alloys is also dealt with in the section on austenitic cast iron.

Malleable cast iron[5]

The tensile strength of ferritic blackheart, quenched and tempered pearlitic blackheart and balanced manganese/sulphur whiteheart malleable iron increases progressively with decreasing temperature at a rate of about 5 ton/in^2 (77 N/mm^2) per 100°C fall in temperature. The rate of increase in tensile for excess sulphur whiteheart malleable is considerably lower. Proof stress values in all malleable irons also increase with decreasing temperature.

Elongation of ferritic blackheart and whiteheart with balanced manganese/sulphur is virtually unchanged down to −100°C, but excess sulphur whiteheart with lamellar pearlite has a ductile-to-brittle transition close to room temperature.

The elastic modulus increases for all malleable iron by about $0{\cdot}5 \times 10^6$ lb/in^2 (3 GN/m^2) per 100°C fall in temperature.

The impact strength of blackheart malleable has been fully investigated and the information is summarised by Gilbert[1]. The material shows a marked fall in impact strength at low temperature, the transition temperature from ductile to brittle failure depending upon various factors such as composition, heat treatment, presence of notches, and so on.

Simmons, Rosenthal and Lorig[6], in an investigation on the mechanical properties of malleable cast iron, included results of impact tests from room temperature to −60°C. The compositions of the irons tested are given in *Table 131*.

The sub-zero tests were carried out using Charpy key-hole notched specimens of single and double width, unnotched specimens, wedge-shaped specimens and tensile impact specimens. The notched-bar toughness of all the irons decreased gradually at sub-zero temperatures. Only in iron B was this increasing brittleness shown by a sudden decrease in the double width Charpy

test bar values, similar values in all other irons being approximately twice the values obtained from the single width bars. Unnotched specimen values also decreased with temperature, although the loss was slight.

Impact values given by the curves for A and B materials are quoted in *Table 132* for the notched and unnotched specimens, so that the magnitude of changes involved may be appreciated.

Table 131

Iron	*TC* %	*Si* %	*Mn* %	*S* %	*P* %
A	2·30	1·05	0·302	0·084	0·162
B	2·33	0·98	0·340	0·078	0·168

Table 132 CHARPY IMPACT STRENGTH OF MALLEABLE IRONS AT DECREASING TEMPERATURES (FT.LB AND J)

Temperature °C	*Single width notched bar*		*Double width notched bar*		*Unnotched bar*	
	A	*B*	*A*	*B*	*A*	*B*
24	8·3 (11)	6·9 (9)	15·6 (21)	11·8 (16)	90 (122)	81 (110)
0	7·3 (9)	6·3 (9)	11·8 (16)	10·8 (15)	90 (122)	71 (96)
−18	6·3 (9)	5·2 (7)	10·1 (14)	8·1 (11)	90 (122)	71 (96)
−45	3·9 (5)	3·3 (4)	8·6 (12)	3·7* (5)	71 (96)	63 (85)

* Sudden decrease at approximately −38°C.

This subject is also dealt with in the section on impact strength.

Nodular cast irons

Ferritic nodular irons from 1·5–2 per cent silicon (P < 0·022 per cent) increase steadily in tensile strength from about 25 to 27 ton/in^2 (386 to 417 N/mm^2) to about 41 ton/in^2 (633 N/mm^2), as the temperature falls to −269°C (*Table 133*). The elongation at room temperature (27–33 per cent) remains roughly constant down to −140 to −150°C and then falls off very sharply to about 2 per cent at −200°C. The 0·1 per cent proof stress rises steadily to −196°C[7].

With silicon 2·5–2·8 per cent, the tensile strength rises up to −196°C with an inflexion at −120 to −150°C; the 0·1 per cent proof stress rises steadily, but the elongation falls off sharply as the temperature falls below −150°C.

IMPACT

The impact resistance decreases rapidly over a well-defined temperature range—the transition temperature (*see* section on impact testing)—and

Table 133

Specimen No.	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Mg* %
1	3·75	1·63–1·71	<0·01	<0·015	0·016	0·68–0·73	0·07
2	3·69–3·81	1·53–1·66	0·37	<0·017	<0·021	0·81	0·07
3	3·46–3·69	2·07	0·37	0·015	0·02	0·8	0·07

		Room temp.	*−60*	*−90*	*−120*	*−150*	*−196*	*−269*
1	Tensile, ton/in²	24·1	26·9	28·4	29·8	31·2	35·1	40·8
	N/mm²	372	415	439	460	482	542	630
	0·1% proof, ton/in²	14·1	16·8	18·3	21·4	23·8	31·7	—
	N/mm²	218	259	283	331	368	490	—
	Elongation, %	32	32	35	27	2·6	4	0·5
2	Tensile, ton/in²	24·5	28·2	29·6	31	32·3	35·1	40·5
	N/mm²	378	436	457	479	499	542	625
	0·1% proof, ton/in²	14·3	17·8	19·3	21·9	24·4	32·5	—
	N/mm²	221	275	298	338	377	502	—
	Elongation, %	28·5	25	26	27	27	2·5	0·5
3	Tensile, ton/in²	26·7	30·4	32·3	33·5	34·6	37·9	42·5
	N/mm²	412	470	499	517	534	585	656
	0·1% proof, ton/in²	17·0	20·5	21	—	—	33·5	—
	N/mm²	263	317	324	—	—	517	—
	Elongation, %	27·5	26	25	25	20·5	—	0·5

marked embrittlement occurs at low temperatures, particularly with higher silicon or phosphorus content irons. The change from ductile to brittle fracture occurs at a higher temperature in impact than in tensile testing. If the transition temperature is taken as the mean value in the transition range (fracture 50 per cent ductile), the unnotched transition temperature will be about 50°C lower than the notched temperatures.

The notched impact transition temperatures for the above irons would be −50°C to −20°C, i.e. about 100–120°C higher than the ductile-brittle transition in tensile testing.

Pearlitic as-cast nodular irons with silicon 1·5–2·8 per cent, in general show a fairly steady fall in tensile strength from about 42–45 ton/in² (649–695 N/mm²) at room temperature, to about 35 ton/in² (541 N/mm²) at −100°C with an inflexion at about −100°C followed by a rapid fall from about 35 ton/in² (541 N/mm²) to about 25 ton/in² (386 N/mm²) at −269°C. Elongation falls fairly steadily from 6–12 per cent at room temperature to about 1 per cent at −100°C.

Pearlitic normalised nodular irons (900°C for 4 h—air-cooled) showed a less marked, fairly steady fall in tensile from 55 to 65 ton/in² (849–1004 N/mm²) tensile at room temperature to 45–52 ton/in² (695–803 N/mm²) tensile at −269°C without this inflexion shown by as-cast material. A steady fall in elongation occurred from 2–6 per cent at room temperature to 1–2 per cent at −100°C and about zero at −200°C.

The ductility and high strength of the ferritic nodular irons with silicon

of 2 per cent and less, indicates their suitability for use down to −140°C under conditions not involving impact. Where impact is involved, temperatures of −20 to −50°C are about the lower limit with silicon of 2 per cent or less.

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WEAR RESISTANCE

Three clearly defined types of wear must be considered: abrasive wear, dry sliding wear and lubricated sliding wear. Polishing is a separate category, since it involves surface flow rather than metal removal.

Abrasive wear

Avery has classified the many aspects of abrasive wear[1] and divided them into three main groups: (1) gouging abrasion, (2) grinding abrasion, (3) scratching abrasion; to which a fourth may be added—impact abrasion.

Abrasion resistance is almost entirely dependent upon the hardness of the material[2]. Twelve per cent manganese steel is an apparent exception to this rule. Its initial hardness is about 190–230 HB, but it is effective only under conditions of abrasion or impact that provide sufficient force to work harden the surface; a surface skin, not more than 0·02 in thick is then produced with hardness of about 600–700 HV[3], but this work-hardened surface under severe impact or abrasion can transform to martensite with a hardness of about 900 HV[2, 9]. Conditions of wear that cannot produce local work hardening do not as a rule show this material to advantage and fully hard metal may be more suitable. It is, however, extremely resistant to gouging abrasion and gross impact.

The abrasion resistance of a surface or micro constituent therefore depends not only on its initial hardness, but also on its capacity for work hardening[2, 4].

Grinding abrasion requires that the crushing force exceeds the strength of the abrasive particles being ground, and can be termed 'high stress abrasion'. In a ball mill or impeller grinder, considerable impact may also be involved and the stresses which crush the abrasive may also fragment the hard constituent in an alloy, such as carbides in white iron or high chromium cast iron.

The type of scratching or erosive abrasion which occurs in a ploughshare may be termed 'low stress abrasion' as impact is a small component. Where impact stresses are low, however, the difference between high and low stress abrasion may be artificial, since if a material or micro constituent is penetrated, stresses of the order of its hardness are required. Cutting by a single abrasive particle can be regarded as extreme negative rake machining[5].

In a two-phase structure such as white iron, the hardness of the carbides may be of great significance, since plain iron carbide (Fe_3C) has a hardness of the order of 1000 HV, which is much the same as quartz. With high chromium iron (chromium more than 10 per cent) the carbide hardness is of the order of 1400 HV[6] (*Table 135*). The matrix hardness, however, is also of significance. If some ductility is present, slight yielding may avoid fracture of the carbides, but if the matrix hardness is very much lower than that of the carbides, preferential wear can occur and in wet grinding even of soft abrasives such as wood, cavitation erosion of the matrix may occur. Corrosion resistant materials such as 25–35 per cent chromium irons may well show up to advantage under these conditions and fully martensitic Ni-Hards are better than those with retained austenite.

The hardness of minerals is usually based on Mohs' scale, where ten minerals are classified so that each mineral can be scratched by the next one up in the scale and can scratch the next one below it. This arbitrary scale is consistent with diamond pyramid hardness tests which can be carried out without surface fracture even with sapphire, indicating that some plastic deformation can occur[7]. The following figures give the diamond pyramid hardnesses associated with Mohs' scale[7, 8]:

	Mohs' scale	*HV*
Talc	1	2–3
Gypsum	2	36–76
Calcite	3	109–172
Fluorite	4	190–250
Apatite	5	566–850
Orthoclase	6	714–795
Quartz	7	1103–1260
Topaz or Garnet	8	1200–1648
Corundum	9	2060–2720
Diamond	10	8000–10 000

The hardness values normally associated with the micro constituents of cast iron are given in the section on hardness. In general[7], a surface of hardness *Hs* will be scratched by a point of hardness *Hp* if $Hp \geqslant 1{\cdot}2\ Hs$. The hardness of silicon carbide has been given as 3000 HV[2].

The significance of the relative hardness of the abrasive and the metal has been discussed by Tabor[7] and Richardson[2]. Tabor has observed that true scratching[2] ceases when the tool is less than about 20 per cent harder than the abrasive and wear resistance becomes very great when the metal hardness exceeds the abrasive hardness by 20 per cent when the metal should scratch the abrasive. These conclusions are confirmed by Richardson.

White or chilled irons are available in the unalloyed state with hardnesses of 450–700 DPN, and in the alloyed state up to 600–900 HV. As pointed out in the section on hardness, cast iron is a heterogeneous material, and in the white and chilled state normally contains pearlite with a hardness of 200–300, and cementite[8] with a hardness of 800–1100. The combined presence of these two constituents makes it difficult to obtain hardnesses higher than 550 for an unalloyed iron. By the addition of nickel and chromium, it is possible to produce irons in which the pearlite constituent, with a maximum hardness of about 350, is substituted by martensite, which has a hardness of 700 DPN*, so that the structure consists of two hard constituents, instead of a hard and a soft constituent (*see* section on constitution and structure). Unless sufficient alloy is added to obtain this structural change, very little improvement in the abrasive wear resistance is obtained by the addition of smaller quantities of these elements.

Dry abrasive wear occurs typically in shot-blasting equipment, ploughshares, coke or mineral chutes, high speed pulverising equipment, and similar situations. Two clear-cut sets of conditions arise: in the first case, the abrasive material strikes the surface at a high velocity, so the surface stresses are very high; in the second case, abrasion is caused by sliding movement of the abrasive material under comparatively low pressures. However, as regards surface stress, if a material is penetrated or worn, stresses of the order of its hardness are required whether the abrasion is 'light' or 'heavy'. The different modes of wear may arise from the extent to which the abraded surface fractures on contact and this seems to result from the scale effect, whereby brittle solids shatter when critical dimensions of plastically strained zones are exceeded. Where design and cost allow, the very hard alloyed irons are particularly suitable for conditions of impact abrasion where the energy of impact is insufficient to cause major fractures; the unalloyed white irons, by virtue of their lower costs, are particularly suitable for the less arduous sliding wear duties. The importance of the cost of replacement or continuity of service, however, may make the use of the harder alloyed irons economical, even under less arduous conditions.

In certain cases, abrasive wear is combined with corrosion. Even unalloyed white irons tend to resist corrosion better than ordinary grey iron, but under very severe conditions the use of corrosion-resisting hard irons may be justified, in spite of their high cost. A notable example of this is the high chromium iron containing up to 33 per cent chromium. This material has proved particularly valuable in such applications as brick-making dies and clay-working machinery, where the combined effect of erosion by the moving clay, and corrosion by the attendant moisture makes the conditions particularly arduous. The 30 per cent chromium irons have hardnesses rarely exceeding 350 DPN, but in spite of the low apparent hardness, the wear resistance of these materials under these conditions is frequently much more satisfactory than that of the conventional nickel-chromium irons, with hardnesses of 600 and above (*see Table 135*).

High silicon iron (14–15 per cent silicon) has also a limited application

* Aubrion (*Fonderie*, 1959, June, pp. 266–271) shows figures for water-quenched cast iron martensite of 786–817 DPN (50 g)

Table 134

Type	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Cr* %	*Section* in (mm)	*Diamond pyramid no.*
Unalloyed	3·2–3·8	0·4–0·8	0·3–0·7	0·1	0·2	—	—	—	400–650
	3·0–3·4	0·9–1·3	0·6–1·0	0·1	0·2	—	—	—	400–550
Ni-Hard*	2·7–3·2	0·5	0·3–0·5	0·1	0·4	3·0	1·5	$\frac{1}{4}$–$\frac{1}{2}$ (6–13)	525–625 sand-cast
						5·0	2·0	2–4 (51–102)	575–675 chill-cast
	3·3–3·6	0·5	0·3–0·5	0·1	0·4	2·5–3·0	1·5	$\frac{1}{4}$–$\frac{1}{2}$ (6–13)	525–650 sand-cast
		0·8	0·3–0·5	0·1	0·4	4·5	2·0	2–4 (51–102)	600–725 chill-cast
30% chromium	2·5–2·9	0·33–0·65	0·6–0·8	0·1	0·1	—	28–33	—	350–450 (a)
12–18% chromium 2–4% molybdenum	3·0–4·0	0·4–1·0	0·5–0·9	0·06	0·1	—	12–18	—	600–950 (b)
Grey martensitic	3·0–3·3	1·0–1·3	0·7–1·0	0·1	0·1	4·8–5·2	0·5–0·8	—	390–420 as-cast
	3·0–3·3	1·0–1·3	3·5–3·9	0·1	0·1	6·0–6·5	—	—	160 as-cast, 400–450 heat treated 15 h at 600°C
High silicon iron	0·5–0·65	14·5–15·5	0·3–0·9	0·05	Up to 1·0	—	—	—	450–520

* Trade name registered by International Nickel Ltd. The shock resistance of these castings is markedly increased by stress relieving heat treatment at 200–230 C for 4 h plus one hour per inch of section.

(a) These irons are normally ferritic with high hardness carbides. They can be made martensitic by additions of 1·8–4·0 per cent molybdenum, depending on section, or by air-quenching from 1050 C, giving hardness up to 850 DPN.

(b) For castings above 1·5 in (38 mm) thick, slow heating to 950–1070 C followed by air-cooling will increase the hardness by removing pearlite or bainite and give a matrix structure of martensite and retained austenite (Climax Molybdenum alloy 15:3).

in such circumstances, particularly where its characteristic brittleness is no disadvantage.

Occasionally, hard wear-resisting surfaces require a certain degree of machining, and these requirements may be met by using a martensitic grey iron containing approximately 4–6·5 per cent nickel.

One grade has a hardness of about 400 DPN as-cast, and can be machined with difficulty with special tools. With a higher nickel content the iron is austenitic as-cast and is freely machinable. It can then be heat treated (15 h at 600°C) to give hardnesses of 450–520 DPN.

Table 134 gives particulars of the various grades of abrasion-resisting iron referred to in this section.

Carbide hardness of white irons for abrasion resistance, pearlitic and martensitic

White irons normally consist of free carbides in a matrix which may contain ferrite, pearlite, austenite and martensite. The carbides themselves in unalloyed irons consist of iron carbide, Fe_3C. If chromium is added up to about 10 per cent, some of the chromium goes into the carbide and forms a mixed carbide $(FeCr)_3C$. These forms of iron and chromium carbide are orthorhombic and appear needle-like and coarse. They are not significantly different in hardness from the iron carbide and both have hardnesses of the order of 800–1200 DPN. This type of carbide has been designated Type C1 by Norman[6].

If the chromium content exceeds about 10 per cent, a further type of carbide $(CrFe)_7C_3$ appears, and this is designated Type C2 by Norman. This carbide is trigonal in form and tends to be equidimensional rather than needle type[9]. It has a hardness of 1300–1800 DPN[6, 9]. The difference in the character of these carbides is of particular significance under conditions of mineral abrasion, since the C1 type has about the same hardness as quartz, whereas the C2 type is appreciably harder than quartz.

Above about 30 per cent chromium the predominating carbide is $(CrFe)_4C$, which has a cubic structure[9].

The structure of the matrix, as well as the hardness of the carbides, affects the gross hardness of a casting when tested by the conventional methods of diamond pyramid with loads up to 120 kg. The normal matrix structure of an unalloyed white iron is pearlitic, and as the hardness of the pearlite rarely exceeds 300–330 DPN, the overall hardness of the casting rarely exceeds 650 DPN, and may be as low as 400, in spite of the high hardness of the carbides. Differences in hardness in the unalloyed irons occur both as a result of cooling speed and the carbon content, and the higher the carbon content, the higher the finished hardness of the casting, owing to the increased amount of free carbide. If the iron is eutectic with respect to carbon, it will have the finest distribution of carbides, while a hypereutectic composition will result in coarse carbides with poorer strength and ductility, but under some circumstances, better abrasion resistance. The eutectic range for 12 per cent chromium is between 3·5 and 3·7 per cent carbon. For a 30 per cent chromium

iron it is between 2·4 and 2·8 per cent carbon[6]. Rapidly cooled sections of pearlitic white iron containing up to 3·7 per cent carbon may show hardnesses up to 650 DPN, whereas pearlitic white irons with carbon contents below 3·3 per cent may show overall hardnesses from 450–550. Chilling of a white iron tends to increase its hardness.

Since the softest constituent in an unalloyed white iron is pearlite, the overall hardness can be increased if, by alloying or heat treating, the pearlitic areas can be made martensitic. Theoretically this can be carried out by heat treatment in which the iron is taken up above the critical temperature and oil- or water-quenched. Normal unalloyed white irons, however, are not very responsive to this treatment and cracking may occur with complex castings.

The method which is usually adopted to obtain a martensitic matrix is to alloy with nickel and chromium so as to reduce the austenite transformation rate and permit the retention of martensite during the normal process of cooling. This method forms the basis of the well-known Ni-Hard* cast irons in which 2·5–5·0 per cent nickel is added with 1·5–2·0 per cent chromium[10]. This is sufficient to retain martensite and some austenite at the normal rates of cooling through the critical temperature, so that the final structure consists of iron-chromium carbides of the C1 type in a matrix of martensite. The final overall hardness of such castings ranges from 650 to 780 DPN. These alloying elements, however, usually leave a certain amount of retained austenite, together with the martensite, and this, together with the lower hardness carbides, limits the maximum hardness that can be obtained to about 750–800 DPN.

Further increases in hardness can be obtained by using much higher chromium contents[6, 11].

With chromium above 10 per cent, the C2 type of carbide is likely to predominate. The matrix structure up to 10 per cent chromium is normally pearlitic, but in the region of 10–15 per cent chromium the matrix becomes increasingly ferritic, depending upon the chromium and the silicon contents; this is accompanied by a drop in hardness.

Although the carbides are of the C2 type, the wear resistance is nevertheless markedly affected by the character of the matrix. If, for example, as a result of the presence of nickel or manganese the matrix contains retained austenite, the hardness will be reduced. The austenite, however, may be unstable and break down under load to form martensite. In these circumstances the surface of the casting will work harden markedly, and for certain purposes this has an advantage in that the body of the material is relatively tough and a renewable hard skin is constantly being produced as a result of the abrasive conditions. This process, however, is not always an advantage, since the volume change caused by the breakdown of austenite to martensite induces very heavy compressive stresses in the surface skin and may ultimately lead to spalling, and on occasions to complete breakdown of the body of the casting[6]. Where these conditions exist it may be an advantage to ensure that the whole of the austenite has broken down to martensite. This can be effected:

(1) By refrigeration followed by a stress relieving treatment at about 200°C.

* Trade name registered by International Nickel Ltd.

(2) By tempering to 260–450°C for up to 48 h followed by air-cooling followed by holding for 4 h at 275°C, which transforms the retained austenite to fine martensite and bainite.

(3) By re-heating the castings to suitable austenitising temperatures, followed by mild quenching. This procedure is well adapted to the 12–30 per cent chromium irons, since their type C2 carbides are discontinuous and permit the use of heat-treatment procedures similar to those used for tool steels. For molybdenum-free irons an austenitising temperature of 1150–1200°C followed by air-cooling may suffice. For iron with 2·5 per cent molybdenum a temperature of 950–1100°C may serve[11].

(4) By using selectively those alloying elements which suppress pearlite, but have little tendency to stabilise austenite or lower its martensite transformation temperature. This method may be used in conjunction with the others. Molybdenum appears to be particularly suitable for this purpose. High chromium martensitic cast irons, produced either by heat treatment or by additions of molybdenum, are capable of giving hardnesses in excess of 800 DPN. Heat treatment of the molybdenum-containing irons appears to reduce retained austenite to a marked degree, and therefore to reduce the risk of surface hardening which may cause spalling or disintegration under certain conditions of use[11].

Low temperature heat treatment of white irons to relieve transformation stresses appears to increase the shock resistance significantly without reducing wear resistance or hardness. Holding in the range 200–230°C for 4 h, plus

Table 135 MICRO-HARDNESS TESTS ON CHROMIUM CARBIDES IN CAST IRON BARS OF VARYING CHROMIUM CONTENT[9]

	C %	*Cr* %	*Diamond pyramid no.*	*No. of tests*	*Load* g	*Probable carbide type*
1	4·41	0·92	1092	10	200	
2	4·45	1·96	1154	8	200	
3	4·57	2·60	1372	8	200	
4	4·57	2·60	1349	8	200	
5	4·24	4·57	831	8	200	
6	4·24	4·57	843	8	100	$(FeCr)_3C$ Ortho-rhombic
7	4·24	4·57	852	7	200	
8	5·00	5·58	1092	6	200	
9	5·00	5·58	1198	8	200	
10	4·84	8·25	1151	9	200	
11	4·84	8·25	1113	8	100	
12	4·50	13·55	1450	8	100	
13	4·50	13·55	1432	8	200	
14	5·42	22·0	1470	4	200	$(FeCr)_7C_3$
15	5·42	22·0	1468	4	100	Trigonal
16	5·42	22·0	1227	4	200	
17	5·42	22·0	1475	4	100	
18	3·07	31·5	1140	8		$(FeCr_4)C$ Cubic

one hour for each inch of section thickness will increase toughness and strength without reducing hardness, and should be carried out with all Ni-Hard castings. Slow cooling from the stress relieving temperature is not necessary[12].

Cavitation erosion may occur in wet grinding plates if much retained austenite is present and heat treatment to transform to martensite is desirable. Manganese below 0·5 per cent is normally recommended to avoid retained austenite which may be difficult to transform even with long-term heat treatment.

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Dry sliding wear

This form of wear takes place more particularly where it is imperative to absorb energy between two surfaces, as in brakes and brakedrums and clutch plates.

The surface may be mated either to metal shoes or tyres as in railway brakes, or to some form of non-metallic surface such as an asbestos composition, and by the dissipation of heat it absorbs energy to slow down or stop a moving part, or alternatively, to start it up without shock. No liquid lubrication is permissible and the qualities required of the surface are:

(1) Good thermal conductivity.
(2) Ability to resist galling.
(3) Good frictional properties.
(4) Reasonable strength.
(5) Freedom from surface volume changes caused by phase transformation.
(6) Low elastic modulus to accommodate thermal stress.

Cast iron fulfils all these requirements better than any other material. For example, copper has a higher thermal conductivity, but tears and galls easily. Steel has a high strength, but high elastic modulus and tends to gall. Graphite resists galling, but is too weak and has a low coefficient of friction. Cast iron, on the other hand, has good thermal conductivity, does not gall readily, has a reasonable coefficient of friction, and adequate strength and a low elastic modulus, and is widely used in place of steel. Components of this type rarely fail because of the direct stresses imposed upon them.

Failure may take place as a result of:

(1) Wearing or scoring.
(2) Heat checking caused by continuous running at temperatures below the critical.
(3) Heat checking by sudden application of load, producing surface temperatures above the critical α to γ change point.

Where failure by either form of heat checking is not important, resistance to wearing and scoring can best be obtained by utilising a pearlitic grey cast iron with a hardness between 200 and 240 Brinell.

Failure by continuous running at temperatures below the critical (approximately 760°C) occurs by deep oxidation penetrating down the graphite flakes, and is frequently accompanied by a structural change in which the pearlite is spheroidised. This change is accompanied by a slight reduction in hardness.

This sort of failure is primarily a function of the design of the component, and is particularly liable to occur in brakedrums which are used in vehicles operating in crowded cities, where braking is not particularly fierce, but takes place sufficiently frequently to build up the temperature of the drum, and the rate of heat dissipation is too low to keep the temperature of the whole of the braking component below the temperature at which this deterioration can occur.

While this type of deterioration is primarily a function of the design of the brakedrum, some improvement can be effected by ensuring that the structural stability of the iron is high, so that the hardness does not drop rapidly at the temperature attained.

Additions of chromium and molybdenum increase the stability of the pearlite and will increase the strength of the matrix. Their use, however, is not invariably advantageous and in lower carbon irons (below 3·3 per cent) they may increase the liability to cracking under thermal shock and lower the thermal conductivity. Additions of molybdenum and chromium are most likely to be of value in increasing the mechanical strength of high carbon irons (3·4 per cent total carbon and more).

Typical compositions which would meet these requirements are:

Total carbon, %	3·2–3·65
Silicon, %	1·7–2·1
Manganese, %	0·5–0·7
Chromium or Molybdenum, % (usually with more than 3·4% total carbon)	0·3–0·5

Sulphur, %	0·12
Phosphorus, %	0·15 max.
Brinell hardness	200–230

Failure by heat checking under sudden application of load is in a different category, since no amount of heat dissipation from an external surface can completely prevent steep gradients from the braking surface rising above the critical temperature if the metal itself has too low an effective thermal conductivity. Under very severe conditions it is possible for surface melting and for temperatures in excess of 950°C[1, 2, 3] to occur for depths of up to 0·003 in (0·076 mm), while the body of the drum is comparatively cool. Subsurface porosity will markedly lower thermal conductivity and increase the likelihood of heat checking.

Under these conditions the overheated surface may be quenched on release of the load by the cooler metal in the body of the drum, and cause hard spots of martensite to form, which may be no more than 0·001 in in depth. The volume change caused by the martensite and the very severe thermal stresses which occur may cause heat checking on the surface, and in extreme cases the casting itself may fracture.

The actual area of a brakedrum in contact with the brake pads is usually about 60 per cent of the total swept area of the drum or about one-eighth to one-twelfth in the case of disc brakes, and is very important from the point of view of preserving the braking surface. Since surface temperatures developed in contact are very high and extremely local in extent, a short period of free radiation may reduce maximum temperatures markedly. Thus the drum or disc will heat up more uniformly and the risk of thermal shock will be reduced.

Where the ratio of contact area to swept area is high, the development of local high temperatures is particularly likely to occur and this will increase the tendency to surface breakdown.

In most cases the actual mechanical failure of the drum is caused by the violent thermal stresses imposed, rather than by the mechanical stresses imposed in the braking system itself. For service conditions of this type it is probably better to regard the braking surface as expendable, and to renew it frequently, rather than to attempt to attain an indefinite life.

It has been found by experience that cast iron carrying a very high graphite content is particularly suitable for such conditions, even though the mechanical strength of the iron itself has been reduced. The presence of a large amount of graphite, however, reduces the likelihood of the formation of very steep temperature gradients, partly because of its own high intrinsic thermal conductivity, which permits the rapid dissipation of heat from the braking surface, but also by providing a constantly renewable film of graphite which, by spreading over the mating surfaces, reduces the tendency to local welding and seizing, thereby minimising the risk of formation of local hot spots.

The tendency for the cast iron swept surface of a brakedrum or disc to develop heat cracks is primarily dependent on the rate at which energy is fed into it[3, 4]. This can be expressed in terms of horse power per square inch (watts per square metre) of swept area and in a vehicle undergoing uniform

deceleration this value which is a maximum (HP max. or watt max.) at the onset of braking is $PaV/550gA$ (or PaV/A), when P = weight of vehicle in pounds (kg), a = deceleration in ft/s^2 (m/s^2), V = initial velocity in ft/s (m/s), $g = 32{\cdot}2$ ft/s^2, A = swept area in square inches (m^2), 1 HP = 550 ft.lb/s (= 745·7 watts). 1 HP/in^2 = 1·155 837 MW/m^2.

With ordinary brakedrum irons, surface deterioration will occur if HP max. (W max.) exceeds 0·7–0·9 HP/in^2 (0·8–1·0 MW/m^2). With disc brakes, deterioration may commence with 1·1–1·3 HP max./in^2 (1·3–1·5 MW/m^2) of swept area. The value for disc brakes appears to be higher than for drum brakes because of the higher ratio of swept area to brake pad area. These figures are based on commercial experience and the areas involved are the total contact areas. Brake shoe pressures, however, are not uniform over the entire area and the true maximum figures are probably much higher. (*See also* section on coefficient of friction.)

Grey irons with total carbon 3·4 per cent and above appear to resist heat checking better than those with lower total carbon, but will wear more rapidly.

The deterioration of cast iron brake surfaces of the automobile type has been dealt with by Angus, Lamb and Scholes[3].

Cast iron containing from 3·4 to 4 per cent of carbon has been used for heavy brakedrums and segments operating under severe conditions of this type, the high graphite content serving to provide a renewable anti-welding coating between the two surfaces, to provide channels of high thermal conductivity from the braking surfaces, and to increase the resistance to thermal shock. The presence of so much graphite, however, does materially reduce the tensile strength and also the hardness, so that the innate resistance of such materials to scoring and wear is likely to be lower than that of harder materials with lower carbon contents. Molybdenum up to 0·7 per cent may be employed to increase the matrix strength and to reduce the rate of solution of carbide into austenite and thereby retard the effect of local short-term high temperature.

The form in which the graphite is present in sliding surfaces is of importance. The very finely divided undercooled form not merely offers the lowest resistance to wear, but is also more susceptible to thermal shock and provides the least effective means of conducting suddenly-applied heat from the surface. The so-called random flake type of distribution, with a phosphorus content below 0·2 per cent, is unquestionably the most suitable compromise from the point of view of resistance to thermal shock, mechanical strength and resistance to wear.

For the most severe service, where violent braking loads are applied, the use of ferritic nodular cast iron has proved particularly useful. It is more resistant to thermal shock than normal grey irons and its mechanical strength combined with ductility makes it capable of resisting the combined centrifugal and shoe pressure loads which exist in brakedrums. For this type of service the pearlitic nodular irons appear to be much less effective as their resistance to thermal shock is much lower. For severe conditions braking surfaces should be regarded as expendable.

RAILWAY BRAKES

Railway brakes operate normally with cast iron blocks against the steel tyre of the wheel. The coefficient of friction at commencement of application may vary from about 0·3 at 5 mph (8 km/h) to 0·08 at 90 mph (145 km/h). During a stop from 90 mph (145 km/h), however, it may vary from 0·08 at 90 mph (145 km/h) to about 0·18 at 5 mph (8 km/h)[5].

The energy input per square inch of brake block can be expressed as HP per square inch (W/m^2)=

$$\frac{\text{Surface speed of tread (ft/min)} \times \text{brake block resistance (lb/in}^2\text{)}}{33\,000}$$

$$\left(\text{or } \frac{m}{s} \times \frac{N}{m^2} = W/m^2\right)$$

and in the case of a train travelling at 75 mph (121 km/h), with a brake block pressure of 150 lb/in^2 (1·03 N/mm^2) and a mean coefficient of friction between block and tyre of 0·13, the horsepower per square inch (W/m^2) of brake block =

$$\frac{(75 \times 88) \times (150 \times 0{\cdot}13)}{33\,000} \quad \left[\text{or } \frac{(120{\cdot}7005 \times 1000) \times (1034214 \times 0{\cdot}13)}{3600}\right]$$

$$= 3{\cdot}9 \text{ HP/in}^2 \text{ (or } 4{\cdot}5 \text{ MW/m}^2\text{)}$$

This is a very high value compared with normal automobile braking practice and the brake block becomes very hot, and molten and red hot particles may be sprayed off. The blocks are expendable and require regular renewal.

Wear of the block has been shown to increase rapidly with graphitic carbon contents of more than 2·2 per cent, but is also greatly influenced by phosphorus content, and falls as the phosphorus content is increased[6].

British practice has employed blocks with a composition range of:

Total carbon, %	3·1–3·4
Graphitic carbon, %	2·5–2·8
Combined carbon, %	0·50–0·7
Silicon, %	1·4–1·8
Manganese, %	0·5–0·8
Sulphur, %	0·08–0·13
Phosphorus, %	0·8–1·3
Carbon equivalent value	3·8–4·43

but American practice uses phosphorus up to 0·57 per cent. Under British conditions of service the composition quoted above has provided a satisfactory compromise between tyre and block wear, the former being more important than the latter.

The use of phosphorus contents of 3 per cent has been stated markedly to reduce sparking and the risk of undercarriage fires. Massive phosphide increases the risk of thermal cracking which must be allowed for in design.[7]

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Fretting corrosion

This is a form of pitting and wear which can occur between two dry components, nominally fixed to each other, but which may move slightly in relation to each other under vibration. It is characterised by the exudation of red iron oxide (known frequently as 'cocoa') at the junction of the two components, and appears to be caused by local welding of high spots, which under conditions of continual vibration cause minute particles to be torn from the surface. These particles are immediately oxidised to form the characteristic red oxide powder exudation typical of the defect.

This is a particularly difficult form of wear to deal with, and frequently causes severe pitting and distortion of the mating surfaces. Some protection against this form of wear can be obtained by phosphating one of the surfaces. Fretting corrosion may initiate fatigue cracks[2] and may also initiate scoring during subsequent sliding, as discussed in the following section.

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Lubricated sliding wear

This occurs with two clear types of bearing surface:

(1) Slides, cylinders and pistons in which movement is reciprocating.
(2) Bearings in which the movement is essentially continuous.

SLIDES, CYLINDERS AND PISTONS

Cast iron is an extremely good bearing metal, its uses range from textile machinery spindles running under low load at about 10 000 rpm, down to heavy planing machine guides operating at a few strokes per minute. Under conditions of this sort, hardness is not an entirely reliable index of wear resist-

ance. It is well known that under conditions of reciprocating sliding lubricated wear, in which a hardened steel surface is mated to an ordinary grey iron surface with a hardness not exceeding 240 Brinell, the greatest wear may occur on the hardened steel face where some abrasive conditions exist.

The presence of graphite is said to contribute to the value of the material as a bearing metal. This is probably correct, particularly in the initial stages of service, when the presence of graphite may well act to avoid seizing and scoring during the initial running-in period. It is well established that surfaces that have been carefully run-in without scoring show very marked increases in subsequent life. The work of Bowden and Tabor[1], and others, has shown that this effect is partly one of removing surface irregularities by deformation and actual abrasion, and provided this process can be carried out without the production of gross scoring, the final surface will show a much more satisfactory life. The presence of graphite among other materials is known to facilitate this effect. The amount of graphite, however, on the surface of a cast iron sliding component is very small, and there is little doubt that under normal conditions of usage the amount that is available is exhausted in a very short time, and its effect is therefore likely to be confined to the initial running-in period. The removal of graphite, however, leaves the surface covered with a minute series of voids, and these undoubtedly act as local reservoirs for oil, and also serve as points of termination for local damage by acting as receptacles for abrasive material which would otherwise be rolled backwards and forwards between the sliding surfaces. Precisely how important this effect is in producing the well-known value of cast iron in this connection cannot be clearly stated, but it is certain that it is considerable.

The composition of the iron itself, however, both chemical and structural, must also be taken into consideration. The graphite may be present in random flake form, or in the very fine undercooled pattern. All the evidence goes to show that the latter type gives poorer wear resistance[2, 3] than the former under conditions of boundary lubrication. The matrix may be fully pearlitic, or consist of a mixture of ferrite and pearlite. The evidence shows that a fully pearlitic matrix is normally preferable to one containing mainly free ferrite. Phosphide eutectic may be present as a structural constituent, occupying an appreciable amount of the surface area. It has been shown that irons containing a high phosphorus content, e.g. about 0·7 per cent, are more resistant to marginally lubricated wear than those containing low phosphorus contents[4, 5, 6]. With iron containing 0·7 per cent phosphorus or more, the presence of 5–10 per cent free ferrite associated with moderately coarse random flake graphite, is not detrimental and is probably beneficial. Satisfactory lubrication is, however, more important than microstructure, which becomes significant under conditions of limited or marginal lubrication.

Surface hardness is only one of the factors affecting cylinder bore wear. Thus Boegehold[7] quotes the case of a cylinder block in which, following welding, one cylinder was fully ferritic, while the others were fully pearlitic, but the wear was the same in all cases. Peschanskii[8] quotes air-cooled cylinders surface-hardened to 50 Rc which wore more in 100 h than as-cast cylinders (with 0·2–0·3 per cent chromium and 0·5 per cent phosphorus) with hardness 207–220 HB wore in 500 h. Lees[9] also showed that in 500 mm diesel cylinders

no correlation was found between wear and hardness between 155 and 205 HB.

For special purposes, irons may be produced containing disseminated free carbide in a matrix of pearlite. These irons may be produced by the use of alloys such as chromium and molybdenum, after subjection to a long heat treatment to produce the cementite in the required form. Such structures are particularly suitable for very high duty work, such as aeroplane cylinder liners. In heavy diesel cylinders, however, a dispersed carbide/phosphide is often obtained by the use of vanadium 0·1–0·2 per cent, together with phosphorus contents of 0·1–0·3 per cent. The use of vanadium in this manner has been shown to give very marked reduction in cylinder bore wear in marine and diesel engines[10]. The reason for the improvement is obscure and may be due to the presence of hard, dispersed iron vanadium carbides (HV 1500)[11] in the microstructure, but it may also be due to the presence of vanadium in the carbide lamellae of the pearlite, thus reducing their solubility in sulphuric acid[10] (see also p. 180). The surface finish of marine diesel liners above 600 mm diameter may be deliberately coarse (120–200 μinch CLA) to facilitate running-in.

For many years, in engineering machine tool practice, it has been customary to operate with plain cast iron slides, which in some cases have been cast against chills in the foundry to densen the structure. This, as a rule, rarely increases the hardness beyond 220 Brinell, but does undoubtedly tend to reduce the coarseness of the structure at the point of chilling. Local densening may actually reduce the surface hardness by as much as 10–15 points, by producing a very fine undercooled form of graphite which is associated with soft ferrite. Where densening is employed, therefore, it is desirable to allow sufficient machining allowance to remove the surface carrying the undercooled graphite.

In recent years there has been a tendency to utilise flame hardening as a means of increasing the surface hardness of sliding surfaces. It is often assumed that because this process produces surface hardnesses of 350–500 Brinell, the wear resistance of necessity must also be increased. This, however, must not be taken for granted, as the resistance of a surface to lubricated wear does not always bear a direct relation to its general hardness[8].

It may well be that the effect of flame hardening on wear resistance is greater where iron with a phosphorus content of 0·4 per cent or below is used, and where abrasive conditions may arise. There is no reason to suppose that for surfaces with low specific loads, and free from abrasion it has any beneficial effects for iron containing phosphorus of 0·5 per cent or more.

From work carried out on actual machine tools, the following provisions appear to give the best insurance against gross failure[6]:

(1) Hardness, 180–240 Brinell.
(2) Phosphorus, 0·7–1·4 per cent.
(3) Where for reasons of foundry technique a low phosphorus iron is employed, the matrix should be fully pearlitic. Where high phosphorus iron is employed, the presence of 5–10 per cent of ferrite in a pearlitic matrix is not detrimental.

(4) Medium to coarse random graphite (ASTM Type A, Size 3–5) and total carbon about 3·2–3·3 per cent.
(5) If slideways are densened, sufficient machining allowance should be made to ensure removing all undercooled graphite.
(6) Where conditions are likely to be abrasive, hardened surfaces may offer advantages.
(7) Resistance to lubricated wear is favoured by high carbon, low silicon and high phosphorus.

In general, it appears that for reciprocating lubricated sliding wear, whether in cylinder liners or slideways, the most favourable results are obtained with fully pearlitic irons of high carbon content (e.g. 3·4 per cent) with moderately coarse random flake graphite and with a dispersed hard constituent such as phosphide eutectic or free cementite.

It is possible, however, to over-estimate the importance of structure on lubricated wear resistance. Where lubrication is complete, the structure is of little importance. There are, for example, many applications where undercooled graphite structures, with some ferrite, are normally employed with complete success, e.g. in centrifugally-produced piston rings.

The use of high phosphorus irons, also, is frequently ruled out in complicated castings of the automobile cylinder block type because of difficulty in producing hydraulically sound castings and satisfactory service is still obtained with low phosphorus irons, provided lubrication is adequate. Where conditions of poor or marginal lubrication occur, the presence of free ferrite, undercooled graphite or the absence of phosphide and carbide are detrimental.

Fretting wear and corrosion

Scoring and grooving on lubricated slides is frequently experienced in machinery where speeds and loads are low, and lubrication appears to be effective. There is little doubt that much of this damage is caused during the period when the machine is idle, and not during actual service. It is common experience of machine tool manufacturers that attempts to reproduce heavy scoring by continuous operation very rarely succeed. Damage appears to commence when the slideways are at rest, but are subject to constant vibration. This is particularly likely to occur during transit, for example, by rail or by diesel-engined cargo ship.

The trouble probably originates by the local rubbing of high spots upon each other, which are thereby denuded of oil, so that welding can take place between them. As soon, therefore, as the machine is put into operation the welded parts tear apart and commence the destruction of the surface.

There is no complete answer to this particular problem, but resistance to this form of wear appears to be greater with high phosphorus irons than with low phosphorus irons[12]. At the same time, application of phosphate coatings to one of the sliding surfaces also appears to minimise the risk of damage of this sort, partly by providing an anti-welding layer and partly by forming an additional oil reservoir. The effect is very similar to fretting corrosion,

which is referred to in the section on dry sliding wear. Large graphite size (e.g. ASTM C2) does not necessarily give poor resistance to fretting[12].

Bearings

Although cast iron has been widely used as a bearing metal from the earliest times, it is still frequently not recognised as such in the engineering industry. Nevertheless, its continual employment in quite critical conditions, from automobile distributor bearings to marine propeller shaft bearings, indicates that its use as a bearing material should always be seriously considered in comparison with the conventional white metals and bronzes.

The chief advantages to be gained by the use of cast iron are: (1) cheapness, (2) load-carrying capacity, (3) intrinsic oil-holding capacity. Its chief disadvantage is lack of ability to yield plastically to accommodate imperfect shaft alignment.

The value of cast iron as a bearing material depends essentially upon its microstructure, which normally consists of a pearlitic matrix with a hardness of 180–300 Brinell, in which are dispersed flakes of graphite which themselves act as a temporary lubricant, usually sufficient to cover the running-in period. Graphite is normally removed quite rapidly in the early stages of use, leaving a series of minute cavities throughout the surface, which then act as oil reservoirs, and which will also accommodate a limited amount of dust and debris without damage to the shaft. In addition, the matrix may also carry phosphide eutectic, dependent on the phosphorus content of the original material. Phosphide eutectic has a hardness of the order of 400 Brinell and, in the case of medium phosphorus irons, is dispersed fairly uniformly, but when the phosphorus rises to 0·7 per cent or above, it tends to form a complete network. Thus in the phosphorus-containing irons, a satisfactory combination of a hard load-bearing network is provided in a softer but fairly rigid matrix, together with uniformly dispersed oil-carrying cavities. Provided, therefore, the initial finish of the mating surfaces is reasonably good, a steel shaft will bed on to the bearing surface and give a bearing with a low coefficient of friction and high wear resistance, in addition to a high load-carrying capacity, which is particularly suitable for journal bearings, in which flexibility of the shaft is not required.

LUBRICATION AND BEARING LOADING

Although the metal and microstructure of the bearing is of great significance in relation to the mating surface, particularly in starting from rest, the viscosity, quality and availability of the lubricant at the interface of the two surfaces is of overriding importance, since the maintenance of a continuous film is essential to prevent local metal to metal contact which will lead to local high temperature, local welding and ultimate seizure.

The criterion ZN/P takes account of the lubricant, the sliding speed and the unit pressure between the surfaces. Z is the dynamic viscosity at the working temperature expressed in centipoise and falls as the temperature rises. N is commonly expressed in revolutions per minute, but may be expressed

in terms of surface speed in ft/s (m/s), and P is commonly expressed in lb/in^2 (N/mm^2). These units are not strictly consistent, but as N/P can be regarded as a constant for a given set of conditions, a value for ZN/P for these conditions can be given which is the boundary between full hydrodynamic lubrication and the region in which lubrication may become unstable or discontinuous.

When expressed in consistent units the coefficient of friction $\mu = f(ZN/P)$ under conditions of hydrodynamic lubrication[15,16,17] and falls slightly as ZN/P decreases with rising temperature, until hydrodynamic lubrication begins to break down and with metal to metal contact it rises catastrophically. Thus with full hydrodynamic lubrication the fluid coefficient of friction will be of the order of 0·01 to 0·001; as the lubrication becomes 'boundary' in character, μ will range from 0·01 to 0·10 and as boundary lubrication gives way to dry friction μ may range from 0·1 to 0·4 and lead to seizure.

For ordinary prepared surfaces ZN/P should not fall below a value of 10 or below 1 for very smooth surfaces [where Z is expressed in centipoise at the temperature of operation, N is expressed in revolutions per minute, and P in pounds per square inch (N/mm^2)], if hydrodynamic lubrication is to be ensured.

If N is expressed as ft/s (N_1) instead of rpm, the value ZN_1/P should be not less than 0·087 for ordinary prepared surfaces.

The $P \times V$ value (where P = working pressure on the bearing expressed in lb/in^2, and V = surface velocity expressed in ft/s) is an old criterion which is still used as a rough guide when lubrication conditions are poor and sliding velocity is low and no other factor seems appropriate. This value was based on the fact that the loss of power in a bearing (i.e. the heat developed) varies as the frictional loss and the velocity, thus PV is expressed in lb/in^2 and ft/s, or ft.lb/s/in^2 of bearing surface[17]. The PV factor is also referred to as the rubbing factor.

The PV factor would only assess the severity of the thermal operating conditions if the coefficient of friction were constant, which is obviously not generally correct. The spread of PV values is extremely wide and for rotating bearings with cast iron (HB less than 240) and steel or cast iron shafts, PV values from 600 to 4300 have been employed. Under these conditions, the maximum bearing pressures given in the literature range from 150 to 500 lb/in^2 (1·03 to 3·44 N/mm^2), but values up to 1000 lb/in^2 (6·89 N/mm^2) have been successfully employed where surface finish and lubrication are suitable.

The main purpose of a lubricant is to separate the sliding surfaces by a load-carrying film with a low shear strength and the lubricant may be oil, grease or a solid such as graphite or molybdenum disulphide. Extreme pressure lubricants usually contain sulphur, chlorine or phosphorus, which appear to operate by providing an anti-welding film between the mating surfaces.

The design of rotating or flat faced bearings should be such that the lubricant can form a separating wedge between the mating surfaces. In rotating bearings this is performed by the motion of the shaft, which in full hydrodynamic lubrication is virtually rotating in a fluid film and oil access must be to the point of lowest oil pressure.

In thrust bearings in the Michell type, the wedge of lubricant is formed by the tilting of the thrust pads. Even in flat-faced segmental non-tilting thrust bearings there is evidence that a thermal wedge is formed which increases the oil pressure well in excess of the oil supply pressure[18], and such bearings with radial oil grooves appear capable of bearing loads up to 75 per cent of those carried by tilting pad bearings, provided the exit oil pressure at the oil grooves can be maintained.

It might appear, therefore, that provided a satisfactory oil wedge can be assured, the material and microstructure of the bearing surface is of little significance. Bearings, however, normally start from rest and under some load when lubrication is dependent upon residual lubricant held between the two surfaces and under these conditions the microstructure and oil-carrying capacity of the surface is of great importance.

Cast iron has a significant oil-carrying capacity and the value of phosphated or allied coatings lies partly in their oil-carrying capacity as well as their specific anti-welding characteristics. For journal bearings the ratio of length to diameter is normally about 1:1 and lower ratios (e.g. 1:3) which may be required for design reasons reduce the factors of safety by increasing side leakage. If the bearing length exceeds $2d$, insufficient oil flow may cause overheating[19]. Diametrical clearances of 0·001–0·0015 inch per inch diameter (0·025–0·038 mm per 25 mm) are normal for continuous running, but for intermittent impact loading lower clearances may be used.

RELATIVE HARDNESS OF BEARING SURFACES

It has been regarded as a result of practical experience that reciprocating sliding surfaces of similar material should preferably differ in hardness by 10–20 per cent and according to Opitz[23] it does not matter whether the moving or the fixed surface is the harder. It is also customary in bearings with steel journals (HB 150–170) to use tin- or lead-base white metals (HB 17–30) and to use the harder copper tin alloys (HB 60–100) with work-hardened or heat treated steel shafts (more than 250 HB).

Cast iron bearings with steel shafts These have been widely employed, for example:

(1) In the textile industry for spindle speeds up to 15 000 rpm with very light bearing pressures of the order of 0·25 lb/in^2 (1·7 mN/mm^2) and surface speeds of 25 ft/s (7·6 m/s). With PV values of 6–13 a life of 10 years is normal.

(2) For long precision plain bearings in automobile distributors run at speeds of about 2000 rpm, with side loads of the order of 10 lb/in^2 (69 mN/mm^2) (PV=40–50) using steel spindles, HV 600, and surface finish of 4–32 μinches (0·1–0·8 μm) CLA. The cast iron bearing may be pearlitic, 0·1 per cent phosphorus iron (HB 230), or ferritic 0·6 per cent phosphorus iron (HB 120–130) with a surface finish of 15–30 μinches (0·4–0·8 μm) CLA and usually lasts the life of the vehicle.

(3) Ordinary journal bearings in high or low phosphorus grey pearlitic cast iron (HB 180–240) for use with steel shafts (HB 150–170). Loads of 100–500 lb/in^2 (0·69–3·44 N/mm^2) and surface speeds of 2–5 ft/s (0·6–1·5 m/s) (PV 500–1000) are not uncommon.

Experimental bearings with $5\frac{1}{2}$ in (140 mm) diameter steel journals (HB 155) with surface finish 20 μinches (0·5 μm) CLA running in pearlitic high and low phosphorus cast iron bearings (HB 196–246) with surface finish of 150 μinches (3·8 μm) CLA and diameter clearance of 1·1 thou./in diameter, have operated for seventeen hours with bearing pressures of 400 lb/in^2 (2·76 N/mm^2) and surface speeds of 144 ft/s (43·9 m/s) (PV 57 600). The surface finish of the cast iron bearings was much coarser than is usually considered advisable.

(4) Road roller bearings of annealed ferritic high phosphorus cast iron (HB 140–150) with surface finish 10–25 μinches CLA and steel shafts (HB 207) with bearing loads of 160 lb/in^2 (1·10 N/mm^2), and reversing surface speeds of 0·12–0·61 ft/s (37–186 mm/s) (PV 19–96) have given 4–7 years service with greasing at 8-hourly intervals.

(5) Abrasive lubricated wear occurs in excavators, dredgers, track-wheels etc., and Russian work[20] has indicated that bearings of austenitic high manganese cast iron, as-cast, can operate successfully against hardened steel shafts (HV 500) with lubricant containing 10 per cent of sand by weight and under bearing loads from 35–200 lb/in^2 (0·24–1·38 N/mm^2) (PV values up to 300). The cast iron bearings had a composition of total carbon, 3·5 per cent; silicon, 3·6 per cent; manganese, 10·5 per cent; and sulphur and phosphorus below 0·14 per cent (HB probably 140–280). The success of this alloy is attributed to the change from austenite to martensite under high local stress.

Cast iron versus cast iron bearing surfaces have been successfully employed for:

(1) Heavy lens-type thrust bearings on impulse water turbines using castor oil lubrication forced through a central hole and distributed by two radial grooves. The bearings were $6\frac{5}{8}$ in (168 mm) diameter (D) with a spherical radius of $2D$ carrying a load of 15 tons (900 lb/in^2) (6·2 N/mm^2) at 150 rpm. The maximum linear rubbing speed was 4·3 ft/s at the circumference, with a maximum PV of 4300. If the room temperature (70°F) viscosity of castor oil is taken as 725 centipoise, the ZN/P value when starting from rest is 120 and at 140°F (viscosity 79 centipoise), the value is about 13, both of which are above the value of 10 given earlier. The surface finish was of the order of 20 μinches (0·5 μm) CLA and bearings lasted 15–20 years.

Both cast iron bearing surfaces were of high phosphorus grey iron (total carbon 3·0–3·4 per cent; silicon 2·1–2·5 per cent; phosphorus 1·4 per cent) and were fully pearlitic (HB 180–240). Low phosphorus pearlitic cast iron bearings with total carbon 3·22 per cent, silicon 1·4 per cent and phosphorus 0·24 per cent (HB 160–190) gave much poorer life.

(2) Slides and guideways of grey iron (HB 180–200) as in machine tools

with loads up to 80 lb/in^2 (0·55 N/mm^2) and reversing linear speeds up to 1·3 ft/s (0·4 m/s) (PV 1040) or trunk-type cast iron pistons in medium speed diesel engines with peak side loads of 29 lb/in^2 (0·20 N/mm^2) and mean sliding speeds of 22·3 ft/s (6·8 m/s) (PV 650). Such surfaces should be fully pearlitic with random type graphite ASTM Type A, Size 3–5.

(3) Cam and tappets as in automobile and internal combustion engine practice (*see* section on high local loadings) that may be white iron (HV 480–550) which operate intermittently at speeds of 8–40 ft/s (2·4–12 m/s), and surface loadings of 80 000–160 000 lb/in^2, i.e. 552–1103 N/mm^2 (56–112 kg/mm^2). Above a contact stress of $H/3$ kg/mm^2 (H = hardness of the softer of the two components) wear changes from mild to severe under lubricated conditions, as shown by Burwell and Strang[21]. This is confirmed in normal automobile practice where chilled cams and tappets, HV 430–550 kg/mm^2 ($H/3 = 143$–183) normally operate at about $S_{max.}$ 55–85 kg/mm^2 [78 000–121 000 lb/in^2 (538–834 N/mm^2)] although values up to 126 kg/mm^2 [180 000 lb/in^2 (1241 N/mm^2)] may be used for racing vehicles with a short life. Normal automobile practice, therefore, gives a factor of safety of about 2 compared with the maximum value of $H/3$ kg/mm^2 given by Burwell and Strang[21]. The PV values are very high indeed. Failure by flaking often occurs by fatigue cracks parallel to and just below the surface at the region of maximum Hertzian stress; cadmium plating of the cast iron surface appears of considerable value under conditions of high local loadings as in worm gears[22].

(4) Cast iron shafts or drums of 2 in (51 mm) diameter or more are also used in rotating driers and kilns running at relatively low speeds and loads on white metal bearings.

Bearings for rotating steel shafts Since normal steel shafts rarely exceed 210–220 HB, the hardness and surface finish of both shaft and bearing is of considerable importance and if low phosphorus iron bearings (less than 0·4 per cent phosphorus) are employed, the material should be fully pearlitic with random flake graphite (ASTM Type A, Size 3–5) with hardness preferably not exceeding 180 HB.

An advantage may be gained by the use of high phosphorus irons (0·7 per cent) which have been annealed to give a ferritic structure (HB 120–160) since they will bed in more readily than the pearlitic variety and be less affected by surface finish. Annealing of low phosphorus cast iron bearing surfaces is dangerous.

The surface finish, however, is of great importance in all cast iron bearings and in general should be from 15 to 30 μinches (0·38 to 0·76 μm) CLA for both surfaces. The surface finish of the steel is more important than that of the cast iron and is preferably not more than 15 μinches (0·38 μm) CLA. It also appears preferable to operate with cast iron with a high graphite content, with total carbon content of 3·2–3·5 per cent or more.

Since heavy machining cuts distort or tear the surface to a depth up to 0·1 in[13], the capacity of this layer to absorb further cold work during running-

in is diminished and may influence its ability to bed in and its subsequent resistance to wear and seizure. It is essential, therefore, that finished cuts or grinding should be adequate to remove this heavily distorted layer and light enough not to distort further the surface to a significant depth.

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Polishing[1, 2, 3]

Polishing involves metal flow[2] rather than metal removal and the importance of oxide films or oxide or other powders in producing polishing of a surface instead of scratching or tearing appears to be partly a function of the relative melting points of the oxide or polishing powder and the surface to be polished[1]. The hardness of the powder at room temperature is of little importance.

Thus camphor (mp 178°C) will polish Woods metal (HB 11–12, mp 68°C), but not tin (HB 5, mp 232°C)[3].

Lead oxide (mp 880°C) will polish speculum metal (65 per cent copper; 35 per cent tin; HB 460, mp 720°C), but not nickel (HB 90, mp 1452°C).

Zinc oxide (mp 1975°C) is softer than quartz (HV *c.* 1100, mp 1710°C), but polishes it readily.

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ABRASIVE WEAR RESISTANCE AND MECHANICAL PROPERTIES

The resistance of a surface to sliding wear by another surface under conditions of full hydrodynamic lubrication is virtually independent of its structure, as the two surfaces do not come into contact.

When the operation is started up, however, a condition of marginal lubrication may arise before the oil film has been fully established and abrasive or seizing wear may occur at regions of metallic contact. In these circumstances cast iron is known to operate well and is widely used for piston rings and cylinder liners and other bearing applications.

It is often assumed that wear resistance increases with hardness, as indeed it often does, but there are exceptions and Oberle[1] has indicated that the combination of low elastic modulus which spreads the load, with high Brinell hardness to resist the applied load should result in maximum resistance to surface deterioration by unlubricated sliding wear, as indicated by the relation:

$$\frac{\text{Brinell hardness} \times 10^6}{\text{Elastic modulus (lb/in}^2\text{)}}$$

which he terms the 'Modell'.

It is well established that random flake graphite cast iron is superior to iron with undercooled graphite and that a pearlitic matrix is preferable to a ferritic one, and this service experience is clearly reflected in the Modell values given in *Table 136*.

It is also well established that high phosphorus irons (0·6 per cent and above) are preferable to low phosphorus irons where founding difficulties can be overcome. This is also reflected in the Modell values, but to a less marked extent than the effect of pearlitic and ferritic matrices.

The known poor performance of soft steel sliding surfaces is reflected in the lowest Modell value quoted.

There are, of course, other significant factors in 'lubricated' wear resistance besides Modell value. Thus, soft white or copper-bearing metal bearings have a low Modell value (0·7–2·5), but their melting point is much lower and the ease with which they can conform to misalignment and with which a welded junction can shear makes them particularly valuable as bearing metals.

In cast iron, the effect of graphite as a temporary running-in lubricant and the effect of the graphite cavities in retaining oil are additional advantages.

The Modell value is most likely to be significant for the comparison of

Table 136

	Tensile ton/in²	(N/mm²)	E_0 lb/in² × 10⁶	(GN/m²)	*HB*	*Modell*
White iron[2]	35	(541)	30	(207)	500	16·7
GREY IRON FLAKE GRAPHITE						
Low phosphorus (<0·3%)						
BS 1452 Grade 10[3]	10	(154)	15	(103)	134–164	8·9–11·0
Grade 17[4] pearlitic	17	(263)	18·8	(130)	182–223	9·7–11·9
Grade 17[4] annealed ferritic	12·67	(196)	18·14	(125)	104	5·7
GREY IRON FLAKE GRAPHITE						
High phosphorus (0·65–0·9%)						
BS 1452 Grade 10[3]	10	(154)	15	(103)	146–176	9·7–11·7
Grade 17[4] pearlitic	18·9	(292)	19·5	(134)	227	11·7
Grade 17[4] annealed ferritic	12·7	(196)	18·1	(125)	133	7·4
GREY IRON UNDERCOOLED GRAPHITE[4]						
Low phosphorus, 0·25% pearlitic	13·84	(214)	21·45	(148)	222	10·4
Low phosphorus, 0·25% annealed ferritic	11·63	(180)	20·3	(140)	131	6·5
High phosphorus, 0·65% pearlitic	14·95	(231)	20·64	(142)	231	11·2
High phosphorus, 0·65% annealed ferritic	13·4	(207)	19·1	(132)	148	7·8
Mild steel[1]	34	(525)	30	(207)	150	5

similar types of material and for cast iron and steel the following figures can be used:

Wear resistance:	Good	Poor	Bad
Modell:	9–12·5	7–8	Below 7

Annealed high phosphoric irons (0·7–1·0 per cent phosphorus) are known to give good bearing service in spite of their low Modell value of 7·4.

There is another significant difference between cast iron and steel. With steel the modulus of elasticity remains constant with stress up to the yield point. With cast iron the modulus of elasticity falls as the stress rises and this therefore means that the Modell value increases as the stress increases.

Gilbert's results[5] show this clearly for flake graphite cast irons with tensile about 16 ton/in² (247 N/mm²), but ranging from 0·06 to 1·0 per cent phosphorus, when E_0 values are compared with E values at 9 ton/in² (139 N/mm²).

The figures in *Table 137* demonstrate clearly the rise in Modell value with applied stress.

With steel, in the absence of work hardening the Modell value will remain constant at about 5.

Table 137

Iron no.	P %	HB	$E_o \times 10^6$ zero stress lb/in²	(GN/m²)	Modell at E_o	E_9 $Stress_9$ (139 N/mm²) lb/in²	(GN/m²)	Modell at E_9
1	0·06	178	$19·2 \times 10^6$	(132)	9·3	$13·3 \times 10^6$	(92)	13·4
2	0·26	206	$19·8 \times 10^6$	(137)	11·4	$14·7 \times 10^6$	(101)	14·0
3	0·45	202	$20·1 \times 10^6$	(139)	10·1	$15·4 \times 10^6$	(106)	12·1
4	0·64	209	$19·7 \times 10^6$	(136)	10·6	$15·6 \times 10^6$	(108)	13·4
5	0·84	220	$20·4 \times 10^6$	(141)	11·8	$16·3 \times 10^6$	(112)	13·5
6	1·0	237	$19·9 \times 10^6$	(137)	11·9	$16·3 \times 10^6$	(112)	14·5

If the Modell value is regarded as an index of surface resistance under conditions of marginal lubrication, the rise in the Modell value with stress for cast iron may be another reason for its usefulness as a bearing material.

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THE CORROSION OF CAST IRON

By R. I. Higgins, B.Sc. and H. H. Collins, F.R.I.C.*

Introduction

The aim of this section is to provide a brief outline of the corrosion behaviour of cast iron. The presentation is essentially factual, since several standard works (*see* p. 329) are available which deal very thoroughly with the detailed mechanism of the various corrosion processes and their rate controlling factors.

Corrosion may be defined as a chemical or electro-chemical reaction of a metal to its environment. When it occurs, the main points of interest to the engineer are the distribution of the corrosion damage (i.e. whether the attack is uniform, of the pitting type, or inter-crystalline etc.), and the rate at which it occurs. When a metal corrodes uniformly, a most useful method of expressing the corrosion rate is in terms of depth of penetration below the original metal surface per unit of time. In this survey, corrosion rates will be expressed in terms of mil/year, where 1 mil = 0·001 in (25·4 μm). When a metal is subject to pitting attack it is obviously the maximum pit depth/

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unit time that is of interest. Unless stated otherwise, corrosion rates quoted apply only when the liquids concerned are at rest. Increases in liquid velocity may produce appreciable increases in corrosion rate and this effect must be considered when selecting a construction material. Because of the large number of variables that may affect corrosion rates under industrial conditions (e.g. impurities, temperature fluctuations, degree of aeration, velocity, abrasion, film formation etc.), it would be unreasonable to expect the corrosion rates quoted in this chapter to be duplicated exactly in service. Nevertheless, they do provide a valuable guide to the rate of deterioration which may be anticipated under industrial conditions.

From the definition of corrosion given above, it is obvious that 'corrosion resistance' or 'corrodibility' are not specific properties of a metal in the same way as tensile strength or density. The rate of corrosion of cast iron depends not only on the metal, but also on its environment. In addition to such factors as chemical composition, temperature and movement of a corrosive liquid, the term 'environment' includes any films and corrosion residues or products which may form or accumulate on the metal surface.

Unalloyed cast irons

INFLUENCE OF THE METAL

Under the type of corrosive conditions where unalloyed cast iron has a practical application, the most important requirement is that the casting should be sound. As long as this condition is achieved, variations in chemical composition or microstructure will produce no marked alteration in corrosion rate. When a phosphoric grey iron corrodes in a substantially neutral and quiescent water, there remains behind on the metal surface a three-dimensional array of graphite particles, and the relatively corrosion-resistant phosphide eutectic cells. If the iron is pearlitic, these phases will tend to be bound together by the carbonaceous residue from the iron carbide lamellae in the pearlite, and depending on environmental conditions this skeleton of the original microstructure will be more or less plugged with rust. Such differences of corrosion rates as have been observed (within the range of unalloyed cast irons) are due to variations in the protective nature of the film or layer which forms in this manner. Chilled or white cast irons produce a dense, highly tenacious carbonaceous surface residue when they corrode, and this explains their slight superiority in certain circumstances over grey iron of similar composition. As far as grey irons are concerned, a dense pearlitic matrix is probably the most important requirement in encouraging the formation of a protective layer of corrosion products on the metal surface.

Providing the casting is sound and the conditions of exposure reasonably uniform, cast iron is characterised by comparatively uniform corrosion. This is undoubtedly bound up with the existence of several finely divided phases in the microstructure. Pitting attack can occur on sound cast iron (e.g. pump impellers, marine propellers, cylinder liners), but when it does it is indicative of a non-uniform environment or condition of exposure.

Acidity In general, unalloyed grey or white cast irons possess no useful resistance to dilute mineral acids. In very dilute acids the presence of air, or other oxidising agents, such as ferric salts, appreciably increases the corrosion rate. If corrosion rates are to be held below 10 mil/year (254 μm/year) in moderately aerated solutions, it is unwise to exceed a total acid concentration of 0·001 *N*, irrespective of the acid concerned.

(1) Mineral acids: Unalloyed cast iron possesses no useful resistance to hydrochloric acid at any concentration or temperature. Dilute sulphuric, nitric and phosphoric acids are also very aggressive, corrosion rates amounting to inches per year in some cases. However, owing to the insolubility of surface films of ferrous sulphate in strong sulphuric acid, concentrations of this acid greater than 65 per cent can generally be handled by cast iron at room temperature, corrosion rates being less than 5 mil/year (127 μm/year). Free SO_3 (oleum) does not increase the corrosion rate, but is liable to produce spontaneous cracking of the cast iron. Any influence which militates against the formation or retention of the insoluble sulphate film might lead to excessive corrosion rates, even in acid concentrations greater than 65 per cent. High liquid velocity, or abrasion of the metal surface, fall into this category; so also do high temperatures, the maximum useful temperature [for corrosion rates around 5–50 mil/year (127–1270 μm/year)] being 100°C for 95–100 per cent acid, 75°C for 75–95 per cent acid, and about 30°C for 65–75 per cent acid, as shown in *Figure 101*.

Concentrated sulphuric acid is strongly hygroscopic and if exposed to air is self-diluting. Under such conditions, extremely high corrosion rates can result from attack by the diluted acid. In practice, they can generally be avoided by taking adequate precautions to prevent the free access of air to the acid.

While nitric acid in concentrations of more than 40 per cent passivates steel, the phenomenon is too unreliable to permit cast irons to be used with confidence, even for strong nitric acids. However, in mixed nitrating acids containing more than 60 per cent nitric and 10 per cent sulphuric acids, the corrosion rate of cast iron is low, around 5 mil/year (127 μm/year), up to a temperature of 110°C. There appears to be some possibility that mixed acids can cause cracking of cast iron in a similar manner to oleum.

Pure phosphoric acid will attack cast iron at all concentrations and temperatures, but crude concentrated phosphoric acid containing inhibiting impurities, such as arsenic, may be handled satisfactorily by cast iron.

(2) Organic acids: Dilute solutions of organic acids such as acetic, oxalic, citric, lactic, butyric etc., attack cast iron at uneconomical rates, especially if well aerated. Temperature and velocity are also accelerating factors. Corrosion by the organic acids present in petroleum or coal tar is not generally serious, being between 2 and 10 mil/year (51 and 254 μm/year) at temperatures around 375°C.

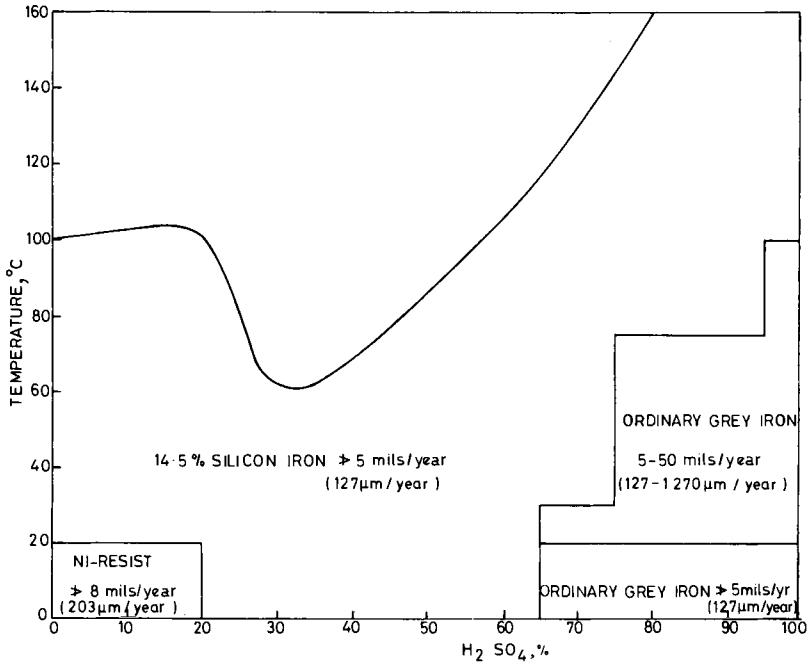


Figure 101. Corrosion rates of cast iron in sulphuric acid

Alkalis Cast iron is not corroded by dilute alkali solutions at any temperature, but hot solutions exceeding about 30 per cent concentrations will attack it, with an accompanying evolution of hydrogen to form soluble salts. Broadly speaking, if corrosion rates are to be held below 8 mil/year (203 μm/year), the temperature should not exceed 80°C for concentrations up to 70 per cent (*Figure 102*). Corrosion rates may range as high as 50–100 mil/year (1270–2540 μm/year) in boiling solutions of more than 50 per cent concentration. Molten caustic soda (650°C) may attack cast iron initially at rates around 800 mil/year (20 320 μm/year), but it is probable that this figure decreases somewhat after a few days. In spite of these high corrosion rates, caustic concentration and fusion pots are made from cast iron, since the material is cheap and the thick wall necessary to give the pot mechanical strength also confers a long life on the pot. It is of the utmost importance to ensure that concentration and fusion pots are cast in sound metal, as any unsoundness, particularly at the bottom of the pan, will lead to pitting attack and premature failure.

Salt solutions Pure water does react slowly with cast iron to evolve hydrogen, but the rate of the reaction is so low that it is of no practical consequence. Dissolved salts may, however, bring about a marked increase in the corrosion rate, especially if they hydrolyse to form an acid solution. Those salts which hydrolyse to form an alkaline solution retard the corrosion rate. In all the salt solutions which can be handled by unalloyed cast iron, any corrosion which takes place is almost entirely dependent on the availability of oxygen

to the corroding metal; that is to say, the corrosion is of the oxygen absorption type rather than the hydrogen evolution type. Factors that increase the supply of dissolved oxygen to the corroding metal (e.g. increased oxygen content of the solution, high velocity or agitation of the liquid) may therefore be expected to increase corrosion rates. The increase in corrosion rate brought about by increasing the degree of aeration of a salt which hydrolyses to give

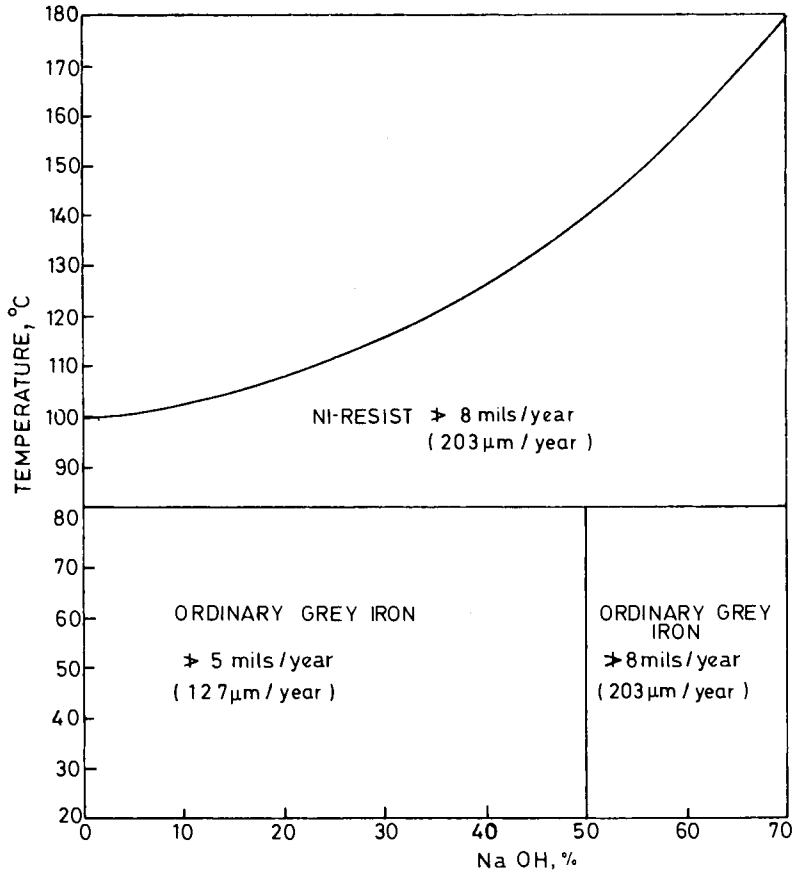


Figure 102. Corrosion rates of cast iron in sodium hydroxide

a dilute acid solution is particularly marked. Oxidising salts in dilute acid solutions are also powerful stimulators of corrosion. They can be much more dangerous than dissolved air because of their greater solubility.

Salts like aluminium chloride and sulphate and, to a lesser degree, calcium chloride, hydrolyse to give acid solutions which are corrosive to cast iron, especially if well aerated. Oxidising salts such as the soluble ferric, cupric and mercuric compounds, which also hydrolyse to give an acid solution, are particularly corrosive. For example, at room temperature a 0·1 per cent solu-

tion of aluminium sulphate corrodes cast iron at about 5 mil/year (127 μm/year), whereas under similar conditions the corrosion rate in a 0·1 per cent solution of copper sulphate is about 30 mil/year (762 μm/year).

Other oxidising salts like chromates, nitrites, nitrates and permanganates, may eliminate corrosion of cast iron in conditions not far from neutrality, but in strongly acid solutions they too can accelerate the corrosion rate. The presence of chlorides in the solution tends to offset the corrosion inhibiting effect of these oxidising agents.

The sulphates and chlorides of the alkali metals give neutral solutions which, if well aerated, might give corrosion rates around 1–5 mil/year (25–127 μm/year) under quiescent conditions, and possibly 30 mil/year (762 μm/year) at extremely high velocities.

Ammonium salts are more corrosive than those of the alkali metals, the difference increasing with increasing concentration and temperature. However, the presence of free ammonia in solution tends to reduce the corrosion rate in ammonium salt solutions. If sufficient free ammonia is present, as in the case of gas-works liquor, corrosion is almost completely inhibited. It should be noted, however, that if relatively thin films of gas-works liquor become exposed to the atmosphere, the ammonia will evaporate, and since atmospheric oxygen has ready access to the remaining salt solution, it will be quite corrosive to cast iron. This condition, for example, could arise in a leaky valve.

Natural and industrial waters Cast iron has a useful resistance to many natural waters, especially under conditions where carbonaceous corrosion residues can become plugged with rust or other insoluble corrosion products.

The rate of corrosion of an iron by sea-water is principally dependent on the rate of flow of the water. In general, quiescent sea-water will corrode iron at rates not greater than 0·005 in/year (127 μm/year) [25 mg/dm^2/day (2·5 g/m^2.d)], but the rate will increase with the velocity of flow up to 30 ft/s when a roughly constant rate of 0·03–0·05 in/year (762–1270 μm/year) is attained. *Table 138* gives some comparative data for the corrosion rates of the various classes of cast iron exposed to quiescent sea-water. The values given in *Table 138* are for essentially unpolluted sea-water. If the sea-water is polluted, for example by industrial effluents, then its corrosivity will be changed in a way dependent on the contaminant.

The corrosive action of fresh waters on cast iron depends largely on their tendency to deposit a protective calcium carbonate scale on the metal surface. This tendency can be measured experimentally and expressed as the so-called 'Langelier Index'. This index, which is defined as the difference in pH of a water before and after coming to equilibrium with marble (calcium carbonate) dust, is a measure of the excess of carbon dioxide in solution over that required to hold calcium bicarbonate in solution. If the index has a negative value (pH value increases on equilibrating with marble dust) the water will have no tendency to deposit a protective calcium carbonate scale. Many soft peaty or moorland waters fall into this category. If the index has a positive value there will be a tendency to precipitate a protective calcium carbonate scale. 'Hard' waters generally, though not invariably, have a positive

Table 138 COMPARATIVE CORROSION RATES FOR CAST IRONS EXPOSED TO SEA-WATER in units of mg/dm^2/day (g/m^2.d)

Source:	Hudson[1]	Kress[2]	Paris and de la Bruniere[3]	Speidel and Wittmoser[4]	LaQue[5]
Location:	Emsworth UK	Cuxhaven Germany	Laboratory	Laboratory*	Harbor Island, USA
Period of test:	2 years	6 months	380 days	220 days	3 years
Steel	—	44 (4·4)	24 (2·4)	18 (1·8)	—
Grey iron	12 (1·2)	17 (1·7)	17 (1·7)	6 (0·6)	0·032 in/year (813 μm/year)†
White iron	—	—	—	9 (0·9)	—
Malleable iron	16 (1·6)	—	—	9 (0·9)	—
Pearlitic nodular graphite iron	—	10 (1·0)	—	14 (1·4)	0·11 in/year (279 μm/year)†
Ferritic nodular graphite iron	—	21 (2·1)	16 (1·6)	5 (0·5)	—

* Artificial sea-water.
† Rate of pitting (100 mg/dm^2/day = 0·020 in/year = 508 μm/year).

Langelier Index, and are therefore not appreciably corrosive to cast iron. Hard waters softened by the soda lime process usually have a zero or slightly positive Langelier Index, but if softened by a base exchange process (which does not remove any carbon dioxide) they may acquire a negative Langelier Index. In such a case softening will increase the corrosivity of the water. The presence of other anions in solution, notably chlorides, decreases the protective properties of the corrosion products formed in waters having a positive Langelier Index.

Mine waters are likely to be appreciably corrosive to unalloyed cast iron, partly because of their acidity, but mainly because of their content of ferric salts.

The reaction of diluted effluent from factories on cast iron is determined by the specific reactions of the dissolved salts. Corrosion rates are likely to be higher in acid than in alkaline effluents. Chlorides in slightly acid waters are effective in stimulating corrosion.

Atmospheric corrosion Atmospheric corrosion rates are determined by relative humidity and pollution. At relative humidities greater than 70 per cent, corrosion proceeds even though there is no visible moisture film on the metal surface. This means that atmospheric corrosion can occur most of the year in the United Kingdom. Sulphur dioxide is probably the main stimulator of atmospheric corrosion in the United Kingdom. Industrial atmospheres are therefore more corrosive than rural atmospheres; marine atmospheres containing airborne chlorides are intermediate between the two. The corrosion rate of cast iron in the heaviest polluted general urban atmosphere is unlikely to exceed 5 mil/year (127 μm/year), and in rural districts it will

Table 139 SUMMARY OF ATMOSPHERIC CORROSION RATE DATA IN UNITS OF $mg/dm^2/day$ ($g/m^2.d$)

Environment	*Rural*		*Urban*			*Industrial*				*Marine*	
Source	ASTM[6]	Roll[7]	Friend[8]	Nekrytyi[9]	Roll[7]	ASTM[6]	Dearden[10]	Friend[11,12]	Roll[7]	LaQue[5]	ASTM[6]
Wrought iron	—	—	25 (2·5)	—	—	—	—	15–19 (1·5–1·9)	—	—	—
Steel	10 (1·0)	—	—	12 (1·2)	—	—	34 (3·4)	24–32 (2·4–3·2)	—	36 (3·6)	27 (2·7)
Grey iron	—	—	14–21 (1·4–2·1)	—	—	—	32 (3·2)	11–12 (1·1–1·2)	—	6 (0·6)	—
White iron	—	—	1–3 (0·1–0·3)	—	—	—	13 (1·3)	—	—	—	—
Malleable iron (0·1% S)	6 (0·6)	—	—	21 (2·1)	—	10 (1·0)	—	—	33 (3·3)	—	9 (0·9)
Malleable iron (0·1% S)	7 (0·7)	33 (3·3)	—	—	49 – (4·9)	19 (1·9)	—	—	56 (5·6)	—	12 (1·2)
Pearlitic malleable iron	5 (0·5)	—	—	—	—	11 (1·1)	—	—	—	—	10 (1·0)
Pearlitic nodular graphite iron	6 (0·6)	—	—	—	—	13 (1·3)	—	—	—	9 (0·9)	10 (1·0)
Ferritic nodular graphite iron	9 (0·9)	—	—	—	—	12 (1·2)	—	—	—	—	16 (1·6)

be nearer to 1 mil/year (25 μm/year). It is probable that the corrosion rates decrease with increasing time to exposure. Because of the thickness of most iron castings exposed to the atmosphere, these corrosion rates do not constitute a serious problem, even where the cast iron is unpainted. A summary of some corrosion data quoted by workers in this field is given in *Table 139*.

Soil corrosion This aspect of the corrosion of cast iron is extremely complex. Much research has been carried out on the corrosion of cast iron in various soils, but it still remains difficult to use this information to make a quantitative prediction of the life of a cast iron pipe line. However, there are several general features which are worthy of note.

Well-drained open soils are not corrosive to cast iron. By and large, the heavy water-logged clay soils present the most corrosive conditions in the United Kingdom, particularly when the pipe line runs through pockets of clay separated by reasonably well aerated soils. Sulphate-reducing bacteria can accelerate corrosion in anaerobic soils. In soils where corrosion is most serious, corrosion takes the form of pitting attack, although this may not be immediately obvious on opening up a trench for pipe inspection. It is clear that under these conditions the maximum pit depth rather than the average corrosion rate is the important factor. Pitting attack is the most serious type of corrosion encountered and under these conditions an assessment of the distribution of pit depths, and in particular the maximum pit depths, should be used as a criterion of corrosion rather than the average corrosion rate. Pitting attack may be stimulated by patchy soil contact on the pipe, a fact which emphasises the desirability of consolidating the back-fill uniformly round the pipe. It is important to realise that all pipe bursts are not necessarily due solely to corrosion. There is evidence to suggest that some are due to ground movement arising from the swelling and contraction of soils with seasonal variation in rainfall. The use of flexible joints or couplings is said to reduce failures caused by ground movement.

Typical pitting rates for various soils are probably of little value when considering any particular problem, but as an indication of the enormous range of observed corrosion rates, a maximum pitting rate of 2 mil/year (51 μm/year) may be quoted for one well-aerated loam soil compared with 40 mil/year (1016 μm/year) for a particular water-logged clay soil. The electrical conductivity of the soil provides a convenient and useful index of soil corrosivity, particularly when it can be correlated within one area with the history of corrosion leaks.

A method of predicting with some certainty the extent of corrosion damage which may be expected in a particular soil has been developed by Booth *et al.* at the National Physical Laboratory. This involves measurements of soil resistivity, pH, redox potential and water content. The soil would be expected to be aggressive if the resistivity was <2000 ohm.cm (<20 Ω.m) and redox potential, a measurement of sulphate-reducing bacteria activity, was $<0{\cdot}400$ (corrected to pH 7). Borderline cases were resolved by consideration of the water content which was considered aggressive at 720 per cent w/w.

The most important piece of work on the corrosion of iron and steel by soils is that carried out under the auspices of the National Bureau of Stan-

dards[13]. In this work, several cast irons and steels of differing analyses and structures were buried for periods of up to $17\frac{1}{2}$ years in 125 test sites distributed throughout the United States. The cast materials buried included sand cast and metal spun pipes and malleable iron elbows. The sand cast pipes presumably had a pearlitic, flake graphite structure, while the metal spun pipes were presumably ferritic with fine flake graphite.

Figure 103 illustrates the average corrosion rates found in this investigation for soils of different resistivities. Although no statistically significant relationship between soil resistivity can be observed in the data obtained by the National Bureau of Standards (Romanoff) in their extensive investigations into the corrosion of ferrous metals in the soil, it is evident from *Figure 103* that at resistivities of <1000 ohm.cm (<10 Ω.m) the amount of corrosion obtained was much higher than it was when the soil resistivity was >1000

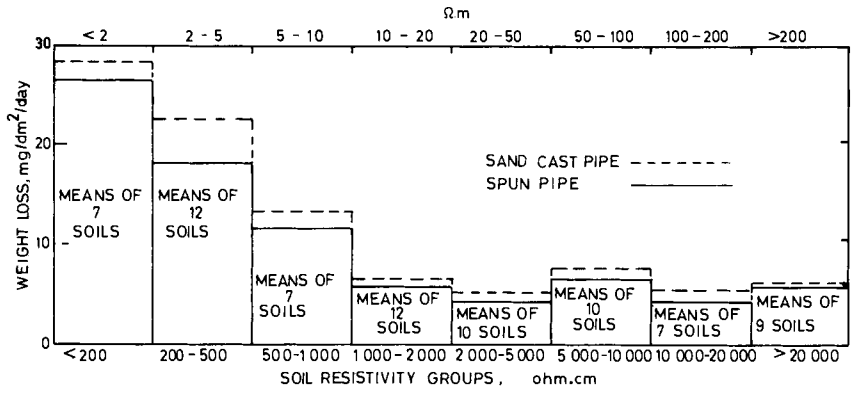


Figure 103. Correlation between resistivity of a soil and its corrosivity (after Romanoff)

ohm.cm (>10 Ω.m). It must be emphasised that *Figure 103* represents mean corrosion rates, but in fact the rates were subject to appreciable variation from one soil to another within the same resistivity class and even from one sample to another within the same soil. It will be noticed that in general the pearlitic irons have a slightly lower corrosion resistance than the ferritic irons.

In addition to tests on ferritic and pearlitic pipe specimens, tests were included on malleable iron elbows and some results of these tests are presented in *Table 140* (comparison of the values for grey iron with those in *Figure 103* will indicate the difficulty of quoting 'typical' figures for soil corrosion rates).

A statistical examination of the National Bureau of Standards (Fuller, BCIRA) results indicates that taking into account the greater wall thickness of grey iron and nodular iron pipe, the grey iron pipe would have a life about 10 per cent greater than steel pipe, and nodular iron pipe life up to 3 times as great as steel or grey iron.

Table 140 NBS[13] DATA FOR COMPARATIVE CORROSION RATES OF STEEL, GREY AND MALLEABLE CAST IRON AFTER 10 YEARS BURIAL

Soil resistivity, ohm.cm (Ω.m)	60 (0·6)	263 (2·63)	408 (4·08)	1270 (12·70)	11 400 (114·0)	13 700 (137·0)
Corrosion rates; mg/dm^2/day (g/m^2.d):						
Steel	6·8 (0·68)	15·8 (1·58)	14·1 (1·41)	9·7 (0·97)	3·2 (0·32)	6·2 (0·62)
Malleable iron	16·6 (1·66)	14·0 (1·40)	11·7 (1·17)	0·9 (0·09)	0·9 (0·09)	5·9 (0·59)
Grey iron	8·9 (0·89)	14·0 (1·40)	26·2 (2·62)	12·6 (1·26)	3·7 (0·37)	9·4 (0·94)

Low alloy cast irons

Low alloy additions to cast iron do not produce any spectacular effect on its inherent resistance to corrosion. Laboratory investigations have indicated that they may have a slight effect under conditions where corrosion rates are too high to be of commercial value (notably in dilute acids), but there are very few cases on record where low alloy additions have given a useful increase in corrosion resistance in the field. One way in which they might exert an effect is by producing a type of pearlite which forms stronger or more adherent and less permeable corrosion residues.

On the other hand, copper has a direct chemical effect, in that it precipitates from solution sulphide ions which accelerate the corrosion of cast iron in acid solutions. It is not necessary for these sulphide ions to be originally present in the corrosive liquid; they can and do originate from the cast iron under acid conditions.

Where previous experience has shown that unalloyed cast iron is not altogether unsuitable for a specific application, that is to say, when its corrosion rate is almost within the acceptable limits, low alloy additions might give a worthwhile increase in the working life of the casting. In this connection, additions of 3–5 per cent nickel will improve resistance to alkalis, both molten and in solution. Additions of 0·5–1·0 per cent copper might be useful in increasing resistance to slightly acid waters or to sulphur-containing crude oils. Additions of chromium up to 1 per cent, or molybdenum up to 0·5 per cent are sometimes employed in conjunction with low alloy additions of copper or nickel, and it is possible that their use encourages the retention of a protective layer of corrosion products on the metal surface.

High alloy cast irons

The highly alloyed, corrosion-resisting cast irons available to the engineer are the austenitic irons Ni-Resist and Nicrosilal, high silicon cast iron, and high chromium cast iron. These irons are frequently used to avoid contamination of a liquid with iron, rather than with the sole object of preventing the deterioration of the casting. The corrosion behaviour of these alloys is discussed below.

NI-RESIST

There are two main classes of Ni-Resist, the copper-containing and the copper-free. The former contains about 14 per cent nickel and 7 per cent copper, while the latter contains 20–22 per cent nickel. Both types may contain up to 2 per cent chromium in order to increase the hardness and stiffness of the material. The precise significance of the chromium content, as far as corrosion resistance is concerned, does not appear to have been investigated very thoroughly, but it is probable that it has only a minor effect and that the main explanation of the corrosion resistance of Ni-Resist lies in its high nickel

content. Although Ni-Resist is in no sense rustless, it has, for many liquids or solutions, a corrosion resistance as good as, or better than, phosphor bronze, in addition to being a more castable and, especially in the case of the copper-containing variety, a cheaper alloy.

Mineral acids One of the most interesting characteristics of Ni-Resist is its resistance to cold dilute sulphuric acid. At concentrations up to 20 per cent and temperatures up to 20°C, corrosion rates range from 0–8 mil/year (0–203 μm/year) under quiescent conditions (*Figure 101*). Increases in temperature or velocity, and probably aeration, can also bring about a marked increase in corrosion rate in this region of concentration. For example, the corrosion rate of Ni-Resist in 5 per cent sulphuric acid at 90°C is around 500 mil/year (12 700 μm/year). At concentrations between 20 and 60 per cent and temperatures up to 30°C, the corrosion rate may range up to 60 mil/year (1524 μm/year). The behaviour of Ni-Resist in sulphuric acid at concentrations greater than 65 per cent at temperatures up to 100°C does not appear to have been investigated satisfactorily, but it is doubtful whether Ni-Resist will have any marked industrial advantage over ordinary grey iron in this general region. It has been claimed that increasing the silicon content of Ni-Resist up to a maximum of 8 per cent (e.g. using a Nicrosilal*, a typical composition of which is nickel 20 per cent, chromium 2 per cent, silicon 6 per cent) increases the resistance to sulphuric acid.

The superiority of Ni-Resist over unalloyed grey cast iron in cold concentrated hydrochloric acid is only of academic interest, the corrosion rate of 370 mil/year (9398 μm/year) still being too high to be of commercial value. However, Ni-Resist may be of value in handling cold dilute solutions of hydrochloric acid to which the air has no access. Corrosion rates at rest are less than 8 mil/year (203 μm/year) for the 1 per cent acid, up to 20°C, and less than 20 mil/year (508 μm/year) for concentrations between 1 per cent and 20 per cent at temperatures up to 20°C (*Figure 104*). It has been claimed that the resistance of Ni-Resist to hydrochloric acid can be slightly improved by omitting the chromium content. Increases in temperature, velocity and aeration can markedly increase the corrosion rate of Ni-Resist in dilute hydrochloric acid.

Nitric acid of all concentrations is corrosive to Ni-Resist, which corrodes in this medium at rates not far removed from those for ordinary grey iron.

The corrosion rate of Ni-Resist in 50 per cent phosphoric acid at 20°C is around 5 mil/year (127 μm/year).

Organic acids Ni-Resist is moderately useful for non-critical parts working in some dilute organic acids such as formic, acetic, oxalic, and so on. For example, the corrosion rate in cold acetic acid up to 25 per cent concentration, and containing some dissolved air, is around 20–30 mil/year (508–762 μm/year). Corrosion rates in fatty acid stills at temperatures of 230°C generally range from about 10–30 mil/year (254–762 μm/year), compared with 200–800 mil/year (5080–20 320 μm/year) for ordinary grey iron.

* An austenitic cast iron developed and patented by The British Cast Iron Research Association, primarily as a heat-resisting iron

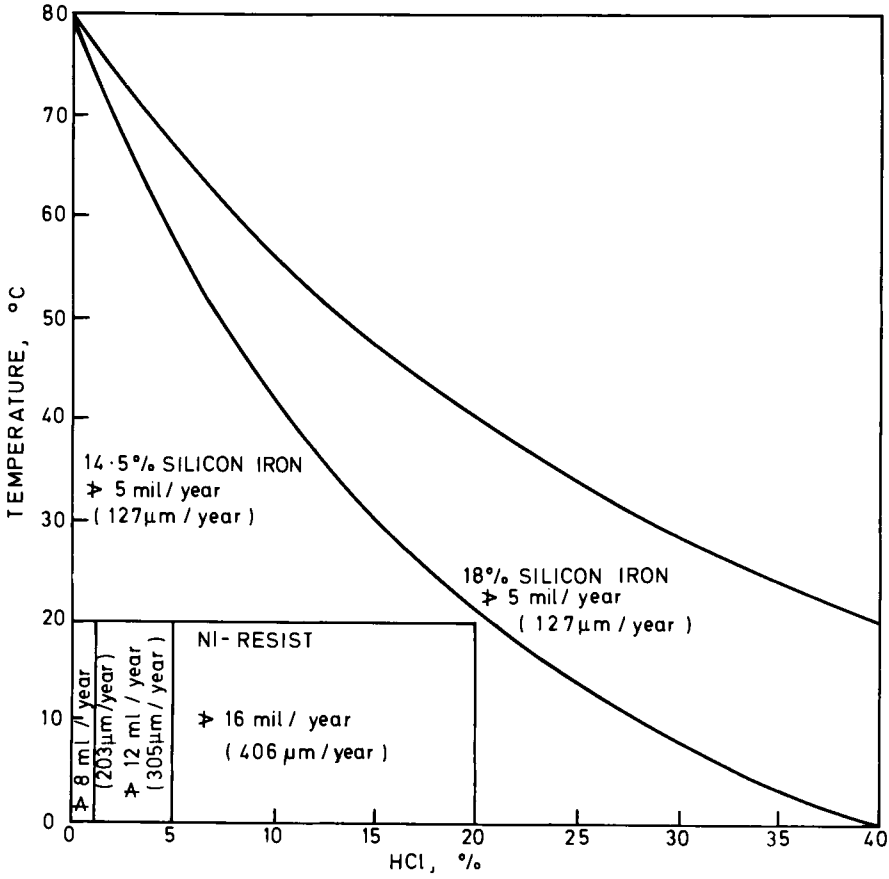


Figure 104. Corrosion rates of cast iron in hydrochloric acid

Alkalis Ni-Resist has a much better resistance to alkalis than unalloyed cast iron, particularly at concentrations greater than 30 per cent and temperatures greater than 80°C, a fact which leads to its considerable application where contamination of the alkali must be avoided or where critical parts are involved. Published data are somewhat conflicting, but it is probably safe to assume that the corrosion rate will not normally exceed 8 mil/year (203 μm/year) at concentrations between 30 and 70 per cent for all temperatures from 80°C to the boiling point (*Figure 101*). In molten caustic soda at 670°C initial corrosion rates are 840 and 260 mil/year (21 336 and 8604 μm/year) for grey iron and Ni-Resist respectively, but it is likely that both these rates decrease appreciably after the first few days' exposure.

Salt solutions Ni-Resist shows a useful improvement over ordinary grey iron in its corrosion resistance to solutions of ammonium salts, and those salts

like aluminium sulphate which hydrolyse to give dilute acid solutions. In cold 5 per cent aluminium sulphate the corrosion rates for ordinary grey iron and Ni-Resist are respectively 40 mil/year (1016 μm/year) and 16 mil/year (406 μm/year), while those for cold 5 per cent ammonium chlorides are 50 mil/year (1270 μm/year) and 3 mil/year (76 μm/year). The corrosion rate of Ni-Resist in many cold neutral or faintly alkaline solutions is less than 5 mil/year (127 μm/year). Roughly speaking, it is usually about five times more resistant than ordinary cast iron under such conditions, but this ratio should not be used as a firm design factor. It is cited merely as a broad indication of one aspect of the corrosion behaviour of Ni-Resist.

Compared with ordinary grey iron, Ni-Resist shows reduced rates of corrosion in solutions of ferric, cupric, stannic and mercuric salts, but even so the corrosion rates are still too high to be of much commercial interest. Dilute solutions of salts like citrates and thiocyanates, which form soluble complexes with iron and are therefore very corrosive to ordinary grey iron, as well as the hypochlorites found in bleaching liquors, can sometimes be handled by Ni-Resist where non-critical parts are involved.

Natural and industrial waters The increased resistance of Ni-Resist to saline and slightly acid solutions is of practical value for critical parts like valves, and for components subject to high velocity or abrasion (e.g. pumps) in the handling of many types of natural water (including sea-water), industrial effluents and dilute acid wash waters used in the chemical industry. However, some mine waters may be appreciably corrosive to Ni-Resist, owing to their content of ferric salts, coupled with acidity.

HIGH SILICON CAST IRON

High silicon cast irons contain 14·5 per cent or more of silicon. In certain circumstances, up to 3·5 per cent molybdenum has been incorporated in this type of iron. These alloys are hard, brittle and, except for grinding, unmachinable. Their high hardness enables them to withstand the combined effects of corrosion and abrasion, but their low strength and brittle nature make them especially susceptible to thermal shock, and appropriate steps must be taken in plants dealing with hot liquids to see that steep temperature gradients do not occur. The engineering properties of the high silicon cast irons lie closer to stoneware than to the usual metals or alloys, and because of this their use is generally only justified by corrosion rates of less than 10 mil/year (254 μm/year).

As a class, these alloys depend on a silica-like oxide film for their corrosion resistance. This film, which is really a corrosion product, is formed fairly soon after the metal is placed in service. Broadly speaking, the iron atoms are leached out of the silico-ferrite lattice, and the surface network of silicon atoms which remains is oxidised to form a protective silicon–oxygen film which effectively arrests further corrosion in many media. Halide and sulphite ions, particularly in acid solutions, are inimical to the development of this protective film, whilst hydrofluoric acid, of course, dissolves it.

Mineral acids High silicon cast iron has a good resistance to sulphuric and nitric acids at all concentrations and temperatures. Attack by boiling sulphuric acid reaches a maximum at approximately 30 per cent concentration with corrosion rates around 10 mil/year (254 μm/year); at 50 per cent concentration the corrosion rate is around 2–3 mil/year (51–76 μm/year) and at higher concentrations less than 1 mil/year (25 μm/year) (*Figure 101*). Although the corrosion rate is very low in oleum, there is a danger of spontaneous cracking. Boiling nitric acid has a maximum corrosion rate of 40 mil/year (1016 μm/year) at about 3 per cent concentration, but this corrosion rate decreases fairly uniformly to around 4 mil/year (102 μm/year) at 50 per cent concentration, and becomes even smaller at higher concentrations.

High silicon cast iron has an excellent resistance to mixed acids (sulphuric and nitric) at all concentrations and temperatures.

It is difficult to summarise quantitatively the corrosion rates of high silicon iron in phosphoric acid. As a general guide, however, it can be noted that high silicon cast iron has a satisfactory resistance [corrosion rate around 5 mil/year (127 μm/year)] to all concentrations of pure phosphoric acid at room temperature. Check corrosion tests should be carried out before this type of iron is used for concentrated solutions at or near their boiling points. Crude phosphoric acid may contain sufficient fluoride to be appreciably corrosive to high silicon iron, and although the general use of this metal for handling such liquids cannot be recommended, check corrosion tests might show that it has an adequate resistance in specific applications.

Compared with its excellent resistance to sulphuric and nitric acids, high silicon cast iron has only a limited range of usefulness in handling hydrochloric acid, particularly if ferric chloride is present. Very broadly speaking, if corrosion rates are to be held below 5 mil/year (127 μm/year), the maximum temperature for the 1 per cent acid should not exceed 80°C, decreasing to 0°C for the 40 per cent acid. This range of application of high silicon iron can be extended by increasing the silicon content to 18 per cent or by adding 3 per cent molybdenum to the standard iron (*Figure 104*). The former measure is characteristic of British practice; the latter is employed in the USA. Again, very broadly speaking, in the case of the 18 per cent silicon iron, the maximum temperature, for corrosion rates to be 5 mil/year (127 μm/year) or less, is 80°C at 1 per cent concentration, decreasing to 20°C for the 40 per cent acid. It is quite possible that the high silicon irons, with or without the molybdenum addition, will give satisfactory service at temperatures and concentrations outside the ranges mentioned above, if ferric chloride is neither originally present nor forms as a corrosion product. Appropriate plant tests, properly interpreted, might be necessary in order to decide whether high silicon irons are suitable for some applications involving hydrochloric acid.

High silicon cast iron has no useful resistance to hydrofluoric or sulphurous acids.

Organic acids High silicon cast iron is highly resistant to solutions of organic acids such as formic, acetic, lactic, tannic, citric etc., at any concentration or temperature.

Alkalis The high silicon cast irons are inferior to ordinary grey iron in their resistance to alkali attack. For example, the corrosion rate in boiling 20 per cent caustic soda or 50 per cent caustic soda at 80°C is around 50 mil/year (1270 μm/year). Fused alkalis attack high silicon irons at a very rapid rate.

Salt solutions High silicon cast iron is of value in handling some acid salt solutions; for neutral salt solutions other materials are generally to be preferred for mechanical reasons, while in alkaline solutions, high silicon cast iron offers no advantage from the corrosion point of view and may, in fact, be inferior to ordinary grey iron.

The corrosion rate of high silicon cast iron in acid solutions reflects its behaviour in solutions of the parent acid. Thus, salts like aluminium sulphate which hydrolyse to give sulphuric acid in solution, can be handled satisfactorily by high silicon iron. Correspondingly, salts which readily hydrolyse to give sulphurous acid or acids of the halogen family (e.g. sodium bisulphite and ferric chloride) are generally too corrosive to be compatible with high silicon cast iron. In certain cases, the resistance of high silicon cast iron to acid halide solutions (and halide solutions containing free halide, such as brine containing a little free chlorine) can be usefully improved by increasing the silicon content up to 18 per cent or by adding 3 per cent molybdenum, as mentioned earlier. In such cases, check corrosion tests are recommended before the material is introduced into service.

Natural and industrial waters Mine waters, in general, are resisted satisfactorily by high silicon iron, but this material is not usually recommended for application in mines on account of its comparatively low strength and brittle nature.

As far as corrosion resistance is concerned, high silicon irons are suitable for handling a wide variety of acid wastes, effluents, and wash waters, but appropriate arrangements must be made to give pipes, bends etc., adequate mechanical support.

HIGH CHROMIUM CAST IRON

In this type of iron all the carbon exists in the form of chromium carbide, i.e. there is no free graphite. In order to develop an outstanding degree of corrosion resistance, sufficient chromium must be present to combine with all the carbon and still leave more than 13 per cent chromium in solid solution in the ferritic matrix. For a wide variety of purposes the mechanical and corrosion requirements are best satisfied by an iron having a chemical composition with about 30 per cent chromium and 1·2 per cent carbon. Most commercial alloys contain 1–3 per cent silicon, which improves their usefulness at high temperatures by elevating the ferrite–austenite transition temperature; it also prevents excessive air-hardening of the ferritic matrix and refines the structure of an alloy of the above composition. It is likely that the presence of 1–3 per cent silicon has little, if any, influence on the corrosion resistance of these cast irons.

Although the corrosion behaviour of the high chromium cast irons has not been extensively discussed in the literature, the general principles involved are well understood. This type of iron owes its corrosion resistance to an invisible surface film of 'chromium oxide' and its behaviour in a corrosive medium depends on whether this film is formed or repaired more rapidly than it is broken down. Since it is an oxide film, oxidising agents encourage its formation and repair. Nitric acid is an example of a chemical oxidising agent which can maintain the oxide film, and dissolved oxygen (from the air) can also have the same effect. On the other hand, reducing conditions, absence or scarcity of dissolved oxygen such as might arise under a surface deposit of solid material, the presence of chloride ions, and sudden temperature changes, encourage the breakdown of the film and can lead to high corrosion rates. High chromium cast iron has a good wear and abrasion resistance and is not susceptible to inter-granular corrosion as are some stainless steels.

Mineral acids One of the most useful corrosion characteristics of high chromium cast iron is its ability to resist nitric acid, particularly in dilute solution. Corrosion rates in this medium are less than 5 mil/year (127 μm/year) at all temperatures up to the boiling point for concentrations up to 70 per cent. At room temperature the corrosion rate is also less than 5 mil/year (127 μm/year) for 70–95 per cent acid, but fuming acid may attack at rates up to 50 mil/year (1270 μm/year). Boiling concentrated nitric acid is very corrosive to high chromium cast iron. It will be observed, therefore, that high chromium cast iron is complementary to high silicon iron as a construction material for handling nitric acid, the former being more suitable for the weaker acids, and the latter for the stronger acids.

No reliable information is available on the behaviour of high chromium cast iron in sulphuric acid. It is possible that the addition of about 3 per cent molybdenum may produce a serviceable alloy, but it seems that the whole subject requires more careful investigation. However, when nitric as well as sulphuric acid is present, as in the case of mixed nitrating acids, high chromium cast iron has a useful application. It is difficult to give precise data, but as a rough guide it can be assumed that if the nitric to sulphuric acid ratio is greater than 2:3, corrosion rates of high chromium cast iron will be in the region of 5 mil/year (127 μm/year) or less for all temperatures up to 70°C. This generalisation may not apply for mixed acids containing more than 80 per cent water.

It is extremely doubtful whether the high chromium cast irons have any really useful resistance to hydrochloric acid at any concentration or temperature. It has been claimed, however, that a 0·5 per cent solution of hydrochloric acid at 20°C gives rise to a corrosion rate around 5 mil/year (127 μm/year). It has also been claimed that this type of iron is compatible with bright dip solutions (mixed nitric, sulphuric and hydrochloric acids) used in brass manufacturing plants; but, doubtless, much depends on the exact composition of the solution.

The high chromium cast irons have a useful resistance to all concentrations of sulphurous acid at temperatures up to 80°C. Data on corrosion rates are somewhat conflicting, but they are unlikely to exceed 50 mil/year (1270 μm/

year) and may well be less. Corrosion rates in phosphoric acid are around 5 mil/year (127 μm/year) or less for concentrations up to 60 per cent at all temperatures up to boiling point. Eighty-five per cent acid at 80°C corrodes at about 5 mil/year (127 μm/year).

Organic acids High chromium cast iron has a good resistance to acetic acid at all temperatures and concentrations, the corrosion rate being less than 5 mil/year (127 μm/year). On the other hand, formic acid, doubtless because of its reducing character, is more aggressive. Even so, a 10 per cent solution at 20°C and the 100 per cent acid at all temperatures up to boiling point, both give corrosion rates of less than 5 mil/year (127 μm/year). A boiling 1·5 per cent solution of lactic acid and a 28 per cent solution of citric acid at 100°C both corrode at less than 5 mil/year (127 μm/year).

Alkalis High chromium cast iron has no outstanding resistance to alkalis; indeed, the few data available suggest that it is inferior to unalloyed grey iron in this field.

Salt solutions High chromium cast iron has a good resistance to the sulphite liquors used in the paper-making industry. For example, the corrosion rate in a liquor of density 1·04 is stated to be less than 5 mil/year (127 μm/year) at the boiling point. Corrosion rates in hydrochloric bleaching liquors at room temperature may range from 5 to 50 mil/year (127 to 1270 μm/year).

Some salts which hydrolyse to give acid solutions are compatible with high chromium iron. For example, 63 per cent calcium chloride at 100°C and zinc chloride solution (density 1·8) at 110°C both give a corrosion rate less than 5 mil/year (127 μm/year). Aluminium sulphate appears to be more aggressive, although cold solutions up to 5 per cent concentration can probably be handled satisfactorily.

Saturated ammonium sulphate at 100°C, boiling 50 per cent ammonium sulphate, boiling 50 per cent ammonium nitrate and saturated ammonium nitrate at 120°C all give corrosion rates of less than 5 mil/year (127 μm/year).

Natural and industrial waters In the absence of solid deposits or crevice effects, high chromium cast iron is practically immune to corrosion by aerated sea-water and most mine waters. High chromium cast iron is, in fact, the most suitable type of cast iron for pumping acid mine waters.

Many dilute acid effluents and wash waters could, undoubtedly, be handled by high chromium cast iron in accordance with the data already given for mineral acids. Applications which involve mechanical abrasion accompanied by mildly corrosive liquids can sometimes produce rapid deterioration of a casting. High chromium cast iron may find a useful application in such cases, owing to its capacity for work-hardening, coupled with an excellent corrosion resistance against the liquid phase concerned. Wet sand pumps and brick-making dies provide suitable examples.

Atmospheric corrosion It is claimed that high chromium cast irons containing about 34 per cent chromium and 1·1 per cent carbon do not rust when

exposed to the atmosphere in the as-cast state (i.e. with the casting skin intact). A similar alloy containing 2·3 per cent carbon is stated to be only slightly inferior in this respect, but on increasing the carbon to 3·1 per cent, a marked rust film develops. However, in the machined and polished state, all these alloys were found to be perfectly resistant, i.e. stainless.

CAVITATION EROSION

Cavitation erosion can occur wherever the speed of motion of a surface through a liquid can cause bubble formation by local pressure drop. Two effects appear significant: bubble collapse and surface scouring.

Bubble collapse is most likely to be of significance with high speed impellers such as dynamometers and surface scouring in cast iron propellers. An excellent general summary of these effects has been given by Godfrey[14] and Higgins[15] has dealt with cast iron.

Bubble collapse—implosion—is capable of producing surface pressures of the order of 100 ton/in^2 (1544 N/mm^2) and can certainly produce actual surface deformation of the metal[16]. The distortion has been confirmed at BCIRA by annealing a pitted as-cast bronze dynamometer impeller, which produced twinning in the α constituent consistent with previous cold work at the surface. There appears to be a direct relationship between cavitation erosion and hardness, erosion being less with high hardness materials[14].

In slower speed impellers, such as large diameter cast iron marine propellers, the scouring effect of high speed fluid—bubble flow can remove surface corrosion products and produce self-perpetuating local anodic spots in the cathodic top area, thus leading to pitting attack.

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PROTECTIVE SURFACE COATINGS

Cast iron is normally regarded as reasonably resistant to atmospheric and natural corrosion, when compared with steel, but it is customary to supply some kind of surface finish both for increased protection as well as for decorative purposes.

Painting

The application of paint to as-cast or machined surfaces requires the normal precautions, namely, that the surfaces should be free from grease and loose scale or rust. In particular, where a good surface finish is required, it is desirable for the casting to be thoroughly dry before paint is applied, since the presence of graphite on the surface of the iron makes it possible for appreciable quantities of moisture to be absorbed in the surface. This may cause imperfect adhesion and flaking after the casting has left the works. Provided normal precautions are taken, however, paint surfaces with a very high standard of finish and durability can be obtained.

Bower-barffing

This process is very widely used for the protection of boilers, cooking pots etc., and depends upon the formation of a substantial thickness of magnetic

oxide of iron on the surface, by heat treatment of the casting in a suitable atmosphere. Provided the coating is continuous and undamaged on subsequent transport, it is claimed to offer a material increase in resistance to corrosion.

Galvanising

This process involves immersing the thoroughly-cleaned casting in a bath of molten zinc and producing a uniform layer of zinc approximately 2 oz per square foot of exposed area (610 g/m^2), about 0·0034 in thick (86 μm). It is widely employed for the protection of both grey and malleable cast irons which are to be sited in exposed positions, or in water pipes. The process is particularly valuable since it offers substantial protection even if the coating is damaged.

The requirements for successful galvanising are primarily that castings should be free from all surface and sub-surface defects, and cleaned prior to galvanising by shotblasting, pickling, or both. It is particularly important that this preliminary cleaning is carried out thoroughly to avoid the risk of the formation of bare spots. There is no difficulty in covering iron in which the graphite is moderately coarse, since the zinc coating bridges the graphite flakes. Grey irons in which the graphite is exceptionally coarse, however, tend to form excessive amounts of dross in the zinc bath. Galvanising of whiteheart or blackheart malleable irons calls for the same precautions.

Plating

Cast iron can be electroplated with all the normal elements: copper, nickel, chromium, zinc, and so on. It is important that the casting should be reasonably close-grained and free from surface and sub-surface defects.

If a reasonable degree of corrosion resistance is to be expected from plating, then it is essential that a satisfactory thickness of plating is applied, and where the casting is ultimately to be chromium-plated, an undercoating of at least 0·0005–0·0008 in (13–20 μm) of nickel, or copper and nickel, should be applied before chromium plating. If the casting is reasonably close-grained, a very high standard of polish can be obtained, providing the electroplated coatings are sufficiently thick. Cast iron is capable also of carrying a substantial thickness of hard chromium plate, which is particularly suitable for wearing parts of internal cylinder bores where lubrication is adequate.

Metal spraying

Cast iron forms a suitable base for most forms of metal spraying, and is regularly employed with coatings of aluminium or zinc, where further protection from corrosive influences is required, and the normal processes of galvanising are inadequate.

Vitreous enamelling

This process is particularly suitable where full protection from corrosion is required, and a good appearance must be maintained. A vitreous enamel coating is essentially a layer of glass which is fused on to the surface of the iron. It is very widely employed in household equipment: for example, baths, cookers, stove pipes etc., and in a much harder variety, is also used in the coating of chemical vessels, or vessels employed for the preparation of food. The process lends itself to both a wide variety of decorative and corrosion-resisting finishes.

MACHINABILITY

The mechanism of cutting

The cutting process starts by applying a force to the material to be cut over a localised area too small to support the force applied. The cutting edge must therefore be sharp, i.e. have a minimum thickness. Once the cut has commenced, however, it continues by a wedging and shearing process ahead of the tool tip as well as by the application of high intensity load at the tip.

A razor blade is intended to cut a hair which requires about 5 grams force to cut with an edge with a radius of curvature about 5×10^{-5} mm and a wedge angle of 7–10°. The metal of the blade, stainless or straight carbon steel, has a hardness of about 740–820 HV.

The cutting edge of a tool for cutting metal need not be so sharp. A carbide tipped tool can be diamond honed to an edge radius of 5×10^{-6} in (127 nm), but is frequently edge honed to a radius of 0·001–0·003 in (25–76 μm) to increase its resistance to impact loading. It must also be supported by having a greater wedge angle (to prevent the edge from breaking up) to continue the cut by promoting a wedge or shear effect ahead of the tip and to conduct heat away from the tip.

Included wedge angles for cutting tools vary from 7 to 10° for razor blades, 20° for wood chisels, 60° for cutting soft metals, to over 90° with negative rake cutting tools.

The permanence of the extreme cutting edge is mainly dependent upon:

(1) Its hardness relative to the material to be cut. If the tool is less than 1·2 times as hard as the metal to be cut it will cease to cut a chip from a flat surface[1]. This hardness may range from 400 to 700 HV with knives, saws and scissors for cutting bread, wood or fabric, where ease of resharpening is important, from 740 to 800 HV for carbon steel chisels for wood, or cutting tools for copper or brass, from 800 to 900 HV for high speed steels, and from 1050 to 1850 HV for cemented carbide, used for cutting wood or metals, although the individual carbide grains may be over 2000 HV.

(2) The temperature at the tool tip, which is dependent upon cutting speed

or horse power per unit of edge length, the thermal conductivity of the metal being cut, and the wedge angle or the cross-section of the tool behind the cutting edge which governs heat conduction from the edge through the tool.

(3) The hardness of the metal of the tool tip at the working temperature and its resistance to softening at the temperature attained.

(4) The abrasiveness of the metal being cut, which is not directly related to the Brinell hardness since copper and 12 per cent manganese steel have an exceptional dulling effect on the cutting edge[2] and cast irons with small amounts of free carbide produce more tool wear than irons of the same Brinell hardness without free carbides.

The work required to separate the chip from the metal at the extreme cutting edge is small in comparison with the total work expended in friction between the chip and the tool, and in shear[3]. The coefficient of friction between the chip and the tool appears to govern the character of the chip with ductile materials and is, of course, affected by lubrication.

A discontinuous chip from a ductile material is favoured by a high coefficient of friction, a continuous chip is favoured by a low coefficient of friction, and a continuous chip with a built-up edge on the tool by a high coefficient of friction. With brittle materials a discontinuous chip will form whatever the coefficient of friction.

The character of the chip or swarf, continuous or discontinuous, affects its disposal from the cutting edge and is of importance in maintaining continuous cutting. Most of the heat developed in cutting is removed in the swarf.

In operations such as filing, sawing or broaching, the cut for each tooth must be limited as the swarf must be contained in the space between each cutting edge before its discharge at the end of the stroke.

Cast iron is a heterogeneous material, and the proportion in which the various constituents are present materially affects the manner in which it machines. The Brinell hardness is a satisfactory criterion of machinability only in structurally similar materials, since the Brinell figure is made up of the complex sum of the quantities and hardnesses of the individual constituents existing under the impression. The following constituents are commonly present in cast iron and their form and quantity will affect the machinability.

(1) GRAPHITE

The hardness of the graphite can be disregarded. Its prime effect appears to be to provide discontinuities which facilitate chip breakage, and its ability to act as an anti-welding lubricant is probably one of the important factors reducing tool wear. In grey iron, the form of the graphite appears to be less significant than the quantity, but there is some indication that coarse graphite machines more readily than undercooled graphite, but produces a less highly finished surface. Increasing the percentage of graphite reduces the power required, particularly in ferritic irons.

(2) FERRITE

In cast iron, ferrite may range from 100 to 150 Brinell, and since most cast iron ferrites are high in silicon, their ductility is reduced and the machinability correspondingly increased. Irons containing free ferrite are more freely machinable than pearlitic irons of otherwise similar structure. Field and Stansbury[4] found that increasing the amount of ferrite, for example, by annealing, improved the machinability to a very marked extent.

(3) PEARLITE

Most engineering irons are substantially pearlitic in character, in order to achieve reasonable machinability with good wear resistance and strength. Pearlite may be either coarse or fine: for instance, in whiteheart malleable irons, very coarse pearlite is present in which the hardness is of the order of 170–180. By rapid cooling, alloying or heat treatment, a very fine pearlite can be obtained, in which the hardness may be as high as 330. Increasing the amount of pearlite decreases the machinability, and coarse pearlite is more readily machinable than fine pearlite[5].

(4) PHOSPHIDE EUTECTIC

This constituent is clearly present in the microstructure of all cast irons containing more than about 0·06 per cent of phosphorus, and at about 0·7 per cent it forms an almost continuous network. Its hardness ranges from about 420 to 560 DPN in the pseudo-binary form, and from 760 to 870 DPN in the ternary form. In the pseudo-binary form its effect on machinability does not become marked, particularly in ferritic materials, until the phosphorus content exceeds about 0·4 per cent. Some work carried out by Loria[6] using carbide tools with cutting speeds of 185, 315 and 512 ft/min (0·94, 1·60 and 2·60 m/s), at a nominal depth of $\frac{1}{16}$ in (1·6 mm) and a feed of 0·0102 in (259 μm) per revolution, showed that within the speed range, phosphorus up to 0·5 per cent had no detrimental effect on tool life and, indeed, actually appeared to improve machinability when compared with phosphorus contents below 0·35 per cent. This effect was also noted by Field and Stansbury[4]. Above 0·5 per cent of phosphorus, machinability may fall somewhat, but even up to 1·4 per cent phosphorus, castings are quite freely machinable provided the ternary form is absent. Ternary phosphide eutectic, however, contains free carbide and may materially reduce machinability by increasing tool wear.

(5) CARBIDES

The presence of free carbides with hardnesses of 800 HV and above, even in the smallest amounts, markedly reduces machinability and decreases tool life. This is particularly so if the carbides are present in fairly massive form.

Field and Stansbury, however, have shown that with carbide tipped tools, dispersed carbides do not affect tool life at a cutting speed of 200 ft/min (1·02 m/s), but seriously reduce tool life at 250 and 1000 ft/min (1·27 and 5·08 m/s).

(6) AUSTENITE

This structure is only present in cast irons that have been highly alloyed with nickel, copper and manganese. The austenite itself may have a hardness ranging from 120 to 160, but the machinability is usually very much better than is obtained with austenitic steels. However, the secondary effect due to the low thermal conductivity of austenitic irons may affect the maximum speed of cut.

(7) OTHER MATRIX STRUCTURES

Sorbitic, acicular or martensitic structures may be obtained by heat treatment or alloying, with hardnesses ranging from 280 to 420, and even in these forms, provided free carbide is absent, machining operations can be carried out much more readily than with irons of the same hardness in which free carbide is present.

Malleable irons

In general, blackheart malleable irons consisting primarily of ferrite and temper carbon, are more freely machinable than grey irons. Whiteheart malleable iron, which is usually pearlitic in the centre, is also very freely machinable, but its machinability is more difficult to assess directly, in view of the presence of a deep decarburised surface layer.

Nodular irons

As-cast nodular irons have approximately the same machinability as a cold-drawn mild steel, but in the annealed ferritic state their machinability is at least as good as that of grey irons of comparable hardness. There appears to be little correlation between nodule size and machinability[7].

Summarising the foregoing, among the ordinary run of castings the most important factors producing free machinability are:

(1) High carbon content.
(2) Absence of free carbide, whether dispersed or in the more massive forms associated with chill, mottle or segregation.
(3) The amount of free ferrite present.

Where a very high degree of machinability is called for and strength and resistance to wear are not important, annealing of grey or nodular cast irons

to the ferritic state will improve the machinability sufficiently, in some cases, to make possible increases in machining speed for the same tool wear, of the order of from 2 to 10 times. Le Grand[8] has published a summary of the effect of various factors on machinability, based on wide industrial experience.

Grey and malleable cast irons are regarded as freely machinable materials, but the concept of machinability is actually a most difficult one to define.

Machinability must be related at least to the following interdependent factors:

(1) Power required for a given cut.
(2) Tool angle.
(3) Cutting speed.
(4) Tool life.

There are, however, very many other factors to be taken into consideration, and the following notes merely serve to indicate very briefly some of the factors involved.

Cutting force[2]

This is the force required to remove a chip of given area (depth × feed). The reference area is usually 0·001 in^2 (0·65 mm^2). The force varies with the rake angle and cutting speed. With positive rake the cutting force increases with speed, and with negative rake the cutting force decreases with speed (*Figure 105*)[9]. The cutting force itself, however, can be resolved into at least three components: tangential, radial and feed thrust.

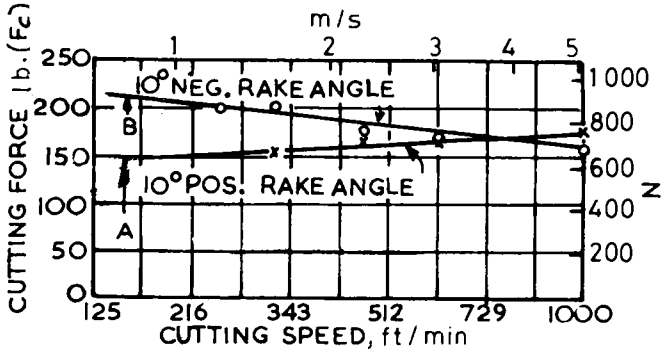


Figure 105. Curves of cutting force versus cutting speed from data by Merchant and Zlatin. Material: SAE 4340, 210 Brinell. Feed: 0·0011 in/revolution. Tool: sintered carbide 3° clearance angle

Tool angles

These affect cutting speeds, feeds, tool life and power. The rake and clearance angles in *Table 141* refer to the cutting edge in relation to the work surface (*Figure 106*).

Table 141

Material	*Hardness* HB	*Cutting angles* (degrees)								*Cutting speeds* ft/min	
		Working rake		*Clearance*		*Working rake*		*Clearance*		*(Depth 0·09–0·375 in [2·29–9·53 mm] Feed 0·01–0·03 in [0·25–0·76 mm])*	
		HSS	*Carbide*	*HSS*	*Carbide*	*HSS*	*Carbide*	*HSS*	*Carbide*	*HSS*	*Carbide*
Malleable cast iron											
Blackheart	120–150	0–8	−7 — +8	4–8	4–8	6–14	−7 — +10	4–8	4–6	50–180	200–800
Pearlitic	170–250	0–8	−7 — +6	4–8	4–8	6–8	−7 — +6	4–8	4–6	40–100	125–320[1]
Grey cast iron	130–170	0–8	−7 — +4	4–8	4–8	6–14	−7 — +10	8–10	4–7	80–160	200–800
	170–200	0–8	−7 — +3	4–8	4–8	6–14	−7 — +10	8–10	4–7	60–160	200–300
	200–240	0–3	−7 — +3	3–6	0–8	4–6	−7 — +6	3–7	3–7	30–100	150–250
White and chilled iron	HV 350–560	0	0	3	2	0	0	3	2	—	25–150

Note: The working maximum rake is defined as the greatest angle between the face of the tool and a plane at right angles to the main surface of the work (BS 1886:1952).

Based on the above figures, the following ratios of cutting speeds may be used for other operations[18]: plunge cutting 0·5; milling, face, 0·5–1·0; milling, form, 0·2; screw cutting 0·1–0·2; tapping 0·1–0·4.

The cutting speeds are based on a tool life of one hour. The maximum figures are based on Field and Stansbury[4] for carbide milling. Other figures are drawn from various sources[10, 18, 19] and can only be regarded as indicative.

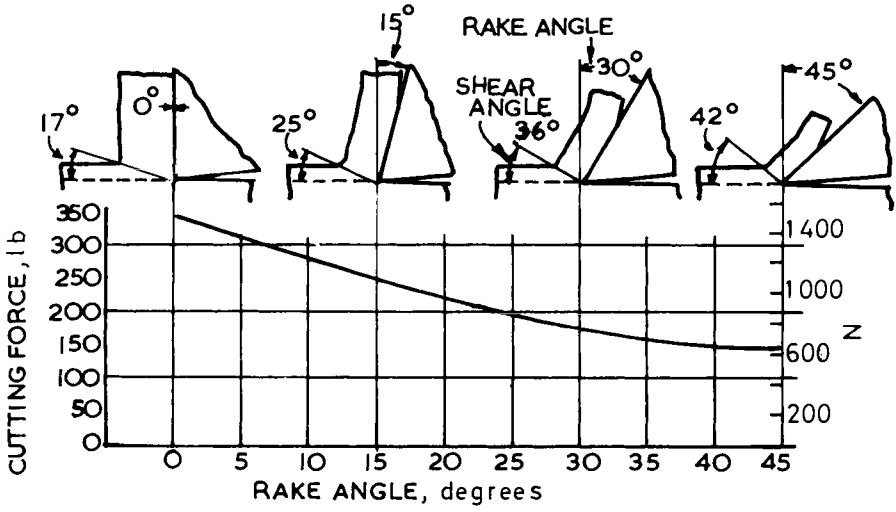


Figure 106. Relationship between cutting force and positive rake angle, typical for cutting speeds employed with high speed steel cutters. Material: SAE 1020 steel. Tool: HSS (18–4–1) 5° clearance. Depth of cut: 0·005 in. Cutting speed: 5.2 in/min. Cutting fluid CCl_4

Tool life

This is a fundamental consideration, and formulae relating various factors include the required tool life (usually in minutes) as a major factor. It is related to cutting speed by the formula: $VT^{\eta} = C$, where V = cutting speed in ft/min, T = tool life in min, C is a constant, and $\eta = \frac{1}{12}$ for cast iron. In general, tool wear is quite rapid in the early stages of the life of a tool, and then the edge settles down to a cutting stage that is maintained for a considerable time. Finally, the edge deteriorates rapidly and no longer cuts satisfactorily.

Cutting speed

This is not merely directly related to tool life, but markedly affects the power required to remove a given amount of metal in a given time. The effect of cutting speed on the cutting force required by positive and negative rake tools has already been referred to (*Figure 105*) and the reason for the reduction in force which occurs with negative rake tools as the speed increases is directly related to the manner for chip formation, as shown by Ernst[9] in *Figure 107*. At the speeds normally employed, however, negative rake milling does require more power than positive rake milling, but can operate at higher speeds, with improved finish and less heat distortion in the work, as the bulk of the heat is removed in the chip.

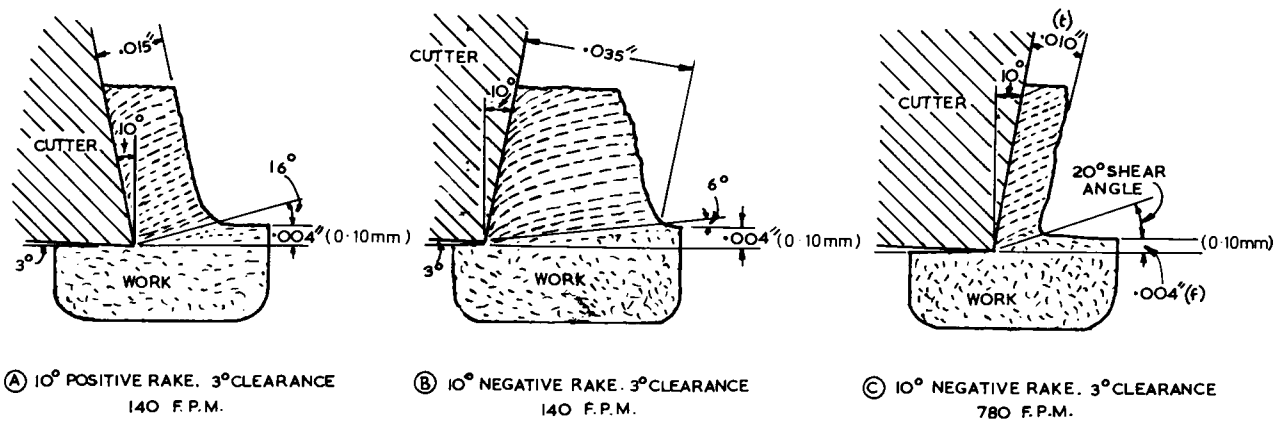


Figure 107. Effect of rake angle and cutting speed on chip formation and shear angle

In practice, two extreme combinations are:

(1) Slow speed, deep cut and heavy feed.
(2) High speed and fine feed.

The former is chosen when roughing steel or cast iron with high speed steel tools, and the latter when using carbide tipped tools.

In general, tool wear is quite rapid in the early stages of the life of the tool and then the edge settles down to a cutting state that is maintained for a considerable time. Finally, the edge deteriorates rapidly and no longer cuts satisfactorily.

The effect of nose radius and approach angle for single point tools can considerably modify cutting speeds.

The data on tool angles and cutting speeds (*Table 141*) are included for completeness and cover normal practice. For more detailed information reference must be made to the authorities quoted.

In castings where the skin properties vary markedly from the sub-surface properties, e.g. very heavy grey iron castings, whiteheart malleable and annealed nodular iron, it may be necessary to alter the cutting procedure after the skin has been removed.

Surface finish

The quality of the surface finish is affected by feed and depth of cut, microstructure, cutting speed and the nose radius of the turning tool[10]. The effect of feed and depth of cut is shown in *Table 142* using a HSS tool with back rake 8°, side rake 14°, relief angle 6°, end cutting edge angle 10° and side cutting edge angle 15°.

Table 142

Feed in/rev (mm/rev)	0·005 (0·127)	0·01 (0·254)	0·015 (0·381)	0·02 (0·508)	*Depth of cut*
Surface roughness RMS	90	175	275	390	0·125 in (3·18 mm)
Surface roughness RMS	90	120	220	325	0·052 in (1·32 mm)
Surface roughness RMS	40	100	200	310	0·010 in (0·25 mm)

The nose radius of a turning tool very markedly affects the final surface finish, as shown in *Table 143*: tool HSS, speed 75 ft/min (0·38 m/s); depth of cut $\frac{1}{16}$ in (1·59 mm); speed 0·02 in/rev (0·508 mm/rev).

The effect of microstructure and cutting speed on surface finish is shown in *Figure 108*[10].

Surface finish is also affected by the size of the graphite flakes (*see* section on surface finish, p. 351). It is also very markedly affected by both preliminary and final machining. Carden and Lamb[11] have shown that heavy machining

Table 143

Nose radius, in	0	$\frac{1}{16}$ (1·6 mm)	$\frac{1}{8}$ (3 mm)	$\frac{3}{16}$ (4·8 mm)	$\frac{1}{4}$ (6·4 mm)
Surface roughness μRMS	500	280	110	100	95
Surface roughness μCLA (approx.)	430	240	95	85	80

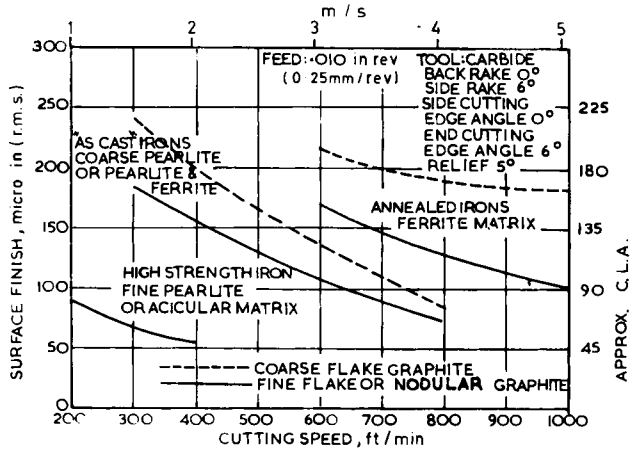


Figure 108. The influence of microstructure and cutting speed on surface finish. Finer graphite, a harder matrix, and higher cutting speed individually and collectively improve the instrumentally determined surface finish

cuts can tear and distort the surface to a depth of about 0·01 in (0·25 mm) or more and unless the finishing cut removes the distorted layer, an appearance of open grain will be left on the finished surface, which may be mistaken for dispersed porosity. With nodular iron, in which the graphite nodules are surrounded by a ferrite layer, the ferrite may be flowed over the nodules and may form asperities that can be destructive of bearings or plastic or leather washers and sealing rings[11, 12].

Drilling of grey cast iron

Although drills are commonly ground to a point angle of 118° irrespective of the material drilled, research[13] by the Production Engineering Research Association has shown that the point angle giving the longest life varied considerably with the material drilled. In tests on cast iron, for example $\frac{5}{8}$ in (16 mm) diameter holes were drilled at a tip speed of 154 ft/min (0·78 m/s) (940 rpm) and a feed of 0·010 in/rev (0·25 mm/rev). Under these severe conditions drills with a point angle of 120° failed after a few holes, whereas 60° point angle drills had an average life of more than 600 holes. Although random variations in machining conditions may prevent gains of this magnitude being achieved in production operations, substantial improvements in drill performance have been reported by firms who have applied PERA recom-

mendations for point shape. As the most suitable point shape for drilling cast iron depends partly on the drilling conditions and partly on the iron drilled, various recommendations covering a range of conditions have been made by PERA.

Le Grand[8] gives an optimum tip cutting speed of 85 ft/min (0·43 m/s) with 0·007 in per revolution (0·18 mm per rev) feed with a compound point angle of 100° for the point and with the lips chamfered to 60° included angle, and on $\frac{1}{2}$ in (13 mm) drills quotes increases of 250 per cent in metal removal per re-grind, as compared with 118° point angle drills.

CARBIDE TIPPED DRILLS AND REAMERS

Le Grand[8] quotes tip cutting speeds of 138 to 260 ft/min (0·70 to 1·32 m/s) at feeds of 0·006–0·007 in per revolution (152–178 μm per rev) for drill diameters from $\frac{1}{4}$ in to $\frac{9}{16}$ in (6 to 14 mm). Reaming is carried out at 86–118 ft/min (0·44–0·60 m/s).

The determination of machinability

The determination of machinability is not a simple matter, and indices have been used based on drilling tests, single point lathe tests, milling etc. A great deal of valuable experience has been gained by these methods.

Machinability can be expressed in a variety of ways:

(1) Load required on a standard drill to maintain a given feed.
(2) Amount of metal removed for a given amount of tool wear.
(3) Amount of metal removed for a given cut depth and tool thrust.
(4) Tangential force required to cut a chip of given area (depth × feed).
(5) Power to remove 1 in^3 of metal per minute (16·4 cm^3/min).

The methods employed by various investigators have largely depended upon the problems involved.

Table 144

Material	*Tensile strength* ton/in^2	(N/mm^2)	*Brinell hardness*	*Dalcher rating* lb	(N)
Grey iron	11·8	(182)	(190–210)	38–41	(169–182)
	14	(216)		38–41	(169–182)
	35·5	(548)	285	67	(298)
Austenitic grey iron	11	(170)	131	38	(169)
Mild steel (SAE 1040) (0·4% carbon, 0·055% sulphur max.)	35	(541)	152	43	(191)
Mild steel (SAE 3435) (0·35% carbon, 3% nickel, 0·8% chromium)	42	(649)	197	60	(267)

The following are the main methods that have been employed:

(1) The Dalcher drilling test, which rates the machinability of a material by the thrust in pounds required to maintain a given speed and feed, with a standard drill. A low Dalcher figure indicates good machinability. Some typical figures[14] are given in *Table 144.*

(2) A widely employed method is to measure the amount of metal removed by a standard tool at different cutting speeds, before the wear land on the tool exceeds 0·03 in (0·76 mm). This method has been quite extensively used in milling tests, and a considerable amount of data has been obtained from its use, particularly with relation to cast iron. Using this method, Field and Stansbury[4] have carried out the most detailed investigation on the effect of microstructure, and drew the following conclusions:

   (a) Tool life increases as the pearlite/graphite distribution becomes coarser.
   (b) The presence of 5 per cent of free carbide in a given structure decreases tool life considerably.
   (c) The presence of 5 per cent of phosphide eutectic (steadite) decreases tool life slightly at high speeds, but increases life at low speeds.
   (d) Ferrite/graphite structures improve tool life up to 50 times over pearlite/graphite structure at low speeds, or increase the cutting speeds for the same tool life (*Figures 108* and *109*).
   (e) The form of the graphite has no appreciable effect on tool life.
   (f) The difference in power required to mill pearlite/graphite cast irons ranging in hardness from 172 to 240 Brinell, including those containing 5 per cent carbide or steadite, is small in relation to the increase in power required as the tool becomes dull.

(3) A further method of rating is based on the assumption that an iron with superior machinability will cut with a heavier feed than one of poorer machinability when a fixed pressure is used to produce tool travel. The constant pressure test ratings appear to agree with tool life results. By this method, using a single point tool, a cutting speed of 32 ft/min (0·16 m/s), and a tool thrust of 81 lb (360 N) for $\frac{1}{8}$ in (3 mm) deep turning, with a standard cold drawn mild steel* as the arbitrary standard of 100, Loria[7] obtained the machinability indices of various materials shown in *Table 145* which also includes the results from other sources.

(4) The tangential force index, as described by Schlesinger[2] measures the tangential force in pounds exerted on the tool to produce a chip or cut of 0·001 in^2 (0·645 mm^2) produced by a single point tool of standardised pattern. The equipment required is a simple dynamometer carrying the tool and fitted to a normal workshop lathe or shaper. The value does not vary between cutting speeds of 30–500 ft/min (0·15–2·54 m/s), but is reduced at speeds of the order of 1000 ft/min (5·08 m/s).

* SAE Standard B.1112, Screw Stock: carbon 0·13%; manganese 0·07–1·0%; phosphorus 0·07–0·12%; sulphur 0·16–0·23%. Tensile strength 30·3 ton/in^2, Brinell hardness 137

For constant feed, the index falls as the area of cut increases. A drop of 20–25 per cent occurs with a chip area of 0·01 in^2. Figures obtained

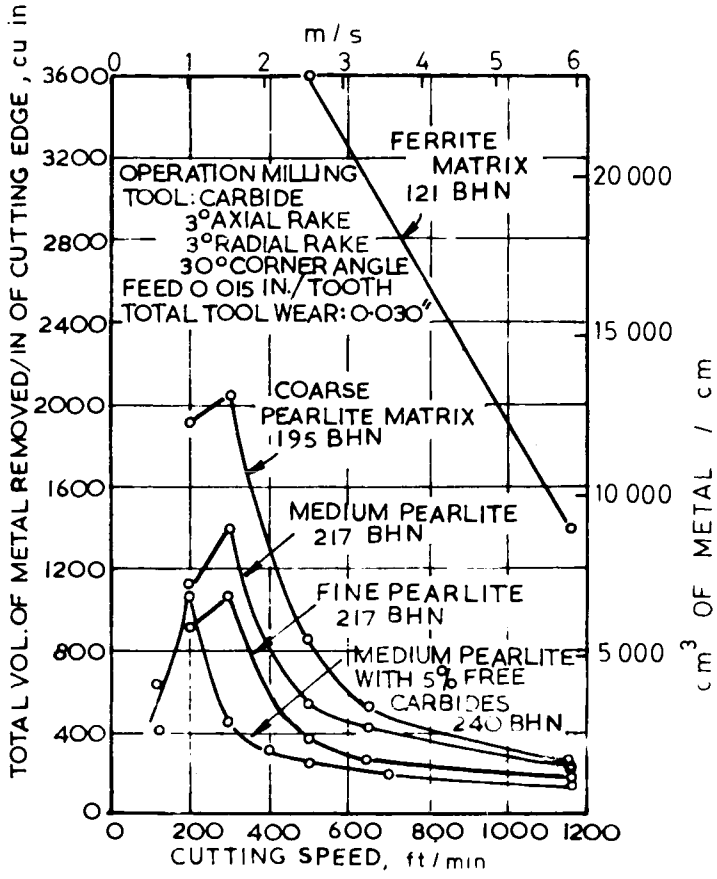


Figure 109. Tool life curves in milling for a series of grey irons with different matrix structures

by Schlesinger are included in *Table 145*. From this index, the power required at the tip of a tool for a given cut can be obtained.

A = area of cut in in^2 (depth × feed)
P = machinability index, i.e. lb/in^2 of cut
S = cutting speed in ft/min
$A \times P$ = pressure in pounds required at a tool tip
Therefore, horsepower (W) required at tool tip

$$= \frac{(A \times P) \text{ lb} \times S \text{ ft}}{33\,000}$$

Table 145 MACHINABILITY RATINGS[2, 4, 7, 15] BASED ON:
(i) Tangential force, lb/in^2, chip area (feed × depth); (ii) Constant pressure index.

Material	*Tensile strength*		*Brinell hardness*	*Tangential force index*			*Constant pressure index*
	ton/in^2	(N/mm^2)		lb/0·001 in^2 of cut	lb/in^2 of cut	N/mm^2 of cut	
Grey iron, soft	(8–10)	(124–154)	<170	(125)	(125 000)	862	172
Grey iron, medium	14–16	(216–247)	179–190	145	145 000	1000	138–150
Grey iron, hard	17	(263)	210–230	(170)	(170 000)	1172	124–127
Blackheart malleable	22–23	(340–355)	128–143	(90–100)	(90–110 000)	621–758	181–188
Nodular iron, as-cast	—	—	293–302	230–250	(230–250 000)	1585–1724	84–92
Nodular iron, annealed	33	(510)	163–181	140–150	(140–150 000)	965–1034	144–168
Steel (screw stock)	18–22	(278–340)	80–100	205	205 000	1413	50
Steel (mild)	25–30	(386–463)	120–140	230	230 000	1586	100
Steel, hard	42–50	(649–772)	180–220	320	320 000	2206	51–54
Steel casting	25–30	(386–463)	135–160	252	252 000	1737	(90)
Brass	26	(402)	115	138	138 000	951	
Bronze	34	(525)	140	162	162 000	1117	
Silumin	16–20	(247–309)	70–90	122	122 000	841	
Copper, annealed[15]	14	(216)	45	171	171 000	1179	

To this must be added the horsepower to drive the machine, approximately one-quarter of the tool load, and the formula then becomes:

$$\frac{(A \times P)\ \text{lb} \times S\ \text{ft/min} \times 1{\cdot}25}{33\,000}$$

(5) A convenient term for defining the power required to remove metal at a given rate is unit horsepower, hp_o, horsepower per in^3/min (W/cm^3). The total horsepower required at the cutting tool is then the product of unit horsepower and rate of metal removal. Zlatin and Walton[9] give machining data based on this criterion. For several typical machining operations, the rate of metal removal is calculated as follows:

Turning = feed/rev × depth of cut × 12 × cutting speed in ft/min.
Milling = feed/tooth × no. of teeth × rpm × depth of cut × width of cut.
Drilling = $\pi D^2/4$ × feed/rev × rev/min.
Planing = feed/stroke × depth of cut × 12 × cutting speed in ft/min.

The horsepower of the motor is then obtained by dividing the cutter horsepower by the efficiency of the machine tool, which is usually about 80 per cent.

(6) Equivalent chip thickness: This concept used in Germany and developed in this country by Brewer and Rueda[15] is based on the idea that temperature at the tool tip in steady state conditions strikes a balance between heat generated and the heat removed. In machining, the heat generated (thermal equivalent of work done) is approximately proportional to the area of the cut, i.e. the product of the feed, f, and depth of cut, d. The heat removed, largely by conduction through the cutting tool in the region of the cut, is related to the total length of tool engagement, le, between the tool cutting edge and the work piece. Thus the general level of temperature is reflected by the ratio fd/le called the equivalent chip thickness. Brewer and Rueda showed that the cutting speed V60 for a 60 min tool life is clearly related to the equivalent chip thickness. The tangential cutting force, $Pt = Cfd$, where C is a constant for the material to be cut as described by Schlesinger and P is expressed in pounds per square inch of cut.

(7) Horsepower per inch of cutting edge: Another way of expressing the heat developed at the tool tip is in terms of horsepower absorbed at the tool tip (*see* paragraph 4) per unit length of cutting edge (*le*) for a given tool life. For carbide tipped tools based on 60 min life and 0·03 in (0·76 mm) tool wear for cast irons of different hardnesses, the following figures are an approximation to the maximum permissible horsepower per inch (W/mm) of cutting edge:

Hardness (HB)	*Horsepower/in of cutting edge*
120 (annealed)	15
190–200	7·5–10
Above 215	3–5

Table 146 POWER REQUIREMENTS FOR MACHINING CAST IRON WITH A SHARP CUTTER. Feed 0·010–0·012 in/rev (0·254–0·305 mm/rev) or per tooth etc.

Operation	Horsepower/in^3/min at cutter		
	Soft iron 160–193 BHN	*Medium iron* 193–220 BHN	*Hard iron* 220–240 BHN
Turning	0·6	0·7	0·75
Planing	0·45	0·5	0·55
Drilling	0·55	0·6	0·65
Milling	0·6	0·65	0·70

Note: in practice, the figures should be multiplied by 3 in order to allow for the dulling of the cutter.

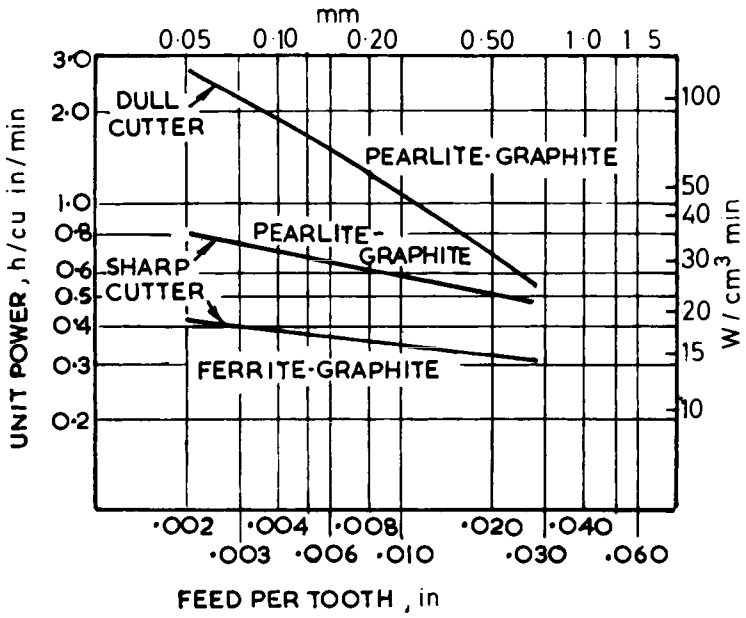


Figure 110. Unit power decreases as feed/tooth increases, as the metal is annealed and when the cutter is sharpened. Test made at 220 fpm cutting speed

Except in the fully annealed condition, all the grey cast irons require about the same power in machining. The differences are small when one realises that the power consumption may increase as much as 300 per cent as a result of the cutter becoming dull. General values for the power requirements of various machining operations, with sharp cutters, are listed in *Table 146*. Note that the horsepower (W) readings are given for feeds of about 0·010 in (0·254 mm). This is important, since the power used is a function of the feed. The horsepower/in^3/min (W/cm^3.min) required for cutting decreases as the feed is increased. This is demonstrated for milling in *Figure 110*. With a sharp

cutter, machining a pearlitic iron at a feed of 0·002 in/tooth (0·05 mm), the power requirement was 0·8 horsepower/in^3/min (36 W/cm^3.min), while at a feed of 0·020 in/tooth (0·51 mm) the power consumption was only 0·50 horsepower/in^3/min (23 W/cm^3.min).

The effect of sharpness of the cutter, the feed per tool and microstructure on power requirements[6, 15] is also illustrated in *Figure 110*. Note that the power requirements for milling a pearlitic iron increased markedly as the cutter dulled.

It is thus obvious that in determining the power requirement for a machine tool motor, allowance should be included for dulling of the cutter as well as for the efficiency of the transmission within the machine and the figures quoted in *Table 146* should therefore be multiplied by three to give an adequate margin of safety.

Machinability problems

Although the foregoing gives the effect of varying structures on machinability and the machinability of varying materials, complaints of poor machinability frequently arise which are not always easy to explain, particularly when hardness determinations appear normal.

POOR MACHINABILITY

When the complaint is a general one, that machinability seems poor but there is no complaint about hard spots, the following points should be examined:

(1) Excessive machining allowances: since it is customary to face mill castings from a point located on the casting itself rather than to operate with a fixed and determined cut depth, it is of the utmost importance that the machining allowance left on the casting by the foundry relative to the location point used by the machinist should be constantly checked. Many complaints of poor machinability originate from excessive machining allowances which have increased the depth of cut 50–100 per cent above the normal.

This difficulty can be quite marked where the same casting is supplied to different machinists who take their locating points on different portions of the casting.

Some complaints of poor machinability ascribed to pearlite rim in blackheart malleable castings have been shown to be due solely to excessive machining allowance.

(2) Whether castings are of higher strength and hardness than called for in service. If the designer calls for a higher grade of iron than is actually required in service, the machinability will be reduced and the risk of hard spots increased.

(3) Imperfectly ground tools: a dull tool may take three times the power

required by a sharp tool, and this variation may outweigh variations in intrinsic machinability between normal pearlitic irons.

(4) Insufficient rigidity or power in the machine.

(5) In malleable iron the machinability of ferritic (blackheart) structures is higher than that of pearlitic structures, and the presence of a deep pearlite rim around a blackheart casting may affect the milling operation where the cutting edges of the tools operate continuously within the pearlite rim. Nevertheless, for the effect to be really significant an appreciable thickness of pearlite is necessary and it must coincide with the depth of the cut. If the cut is below or above the rim the effect may be negligible and it rarely affects drilling operations.

The Brinell hardness is a fair guide to machinability in *structurally similar materials*. Lowering the carbon content, refining the pearlite, and the presence of dispersed carbide all increase the Brinell hardness and reduce machinability.

HARD SPOTS

The most common causes of 'hard spots' and 'hard edges' in grey iron are:

(1) Too low a carbon equivalent value for the section involved. This may be due to the designer calling for a higher grade of iron than the section warrants, but may also arise from variations in metal composition, such as silicon or carbon contents which are lower than normal. The avoidance of this has been dealt with later in the section on the application of specifications and data to design.

(2) Unbalanced sulphur/manganese ratio. The manganese content in grey iron should be $1{\cdot}7 \times$ sulphur content $+ 0{\cdot}3$ per cent if the defect known as inverse chill is to be avoided, particularly in light castings. This defect is characterised by a grey edge to the casting and a hard centre, and is most likely to occur with the first metal out of the cupola, which may be high in sulphur.

(3) In heavy castings, segregations of a phosphide/carbide complex may occur in the centre of the casting. This is particularly likely to occur if vanadium, molybdenum or chromium are present, or if the sulphur content is not balanced by manganese. The difficulty produced by these massive carbides is particularly pronounced with sawing and may be hardly noticeable in drilling.

HARD SKIN

Complaints of hard skin usually occur with fairly heavy grey iron castings or whiteheart malleable, and can rarely be substantiated by the Brinell test.

The complaint is usually associated with rapid tool wear rather than excessive power consumption.

Surface decarburisation in grey iron can produce a graphite-free pearlitic skin about 0·01 in (0·25 mm) thick with a hardness of 300–350 HV which may have a pronounced effect on tool wear.

An examination of several such complaints, however, has actually revealed an unusually soft skin as a result of decarburisation. In such cases the trouble may be caused by:

(1) Irregular depth of skin, so that the tool is occasionally cutting free ferrite with a hardness of 100–120 Brinell, and occasionally pearlite with a hardness of 180–220 Brinell. Such conditions produce non-uniform power requirements and cause tool vibration. This can occur with whiteheart malleable iron or with heavy castings which have been decarburised by slow cooling. A deeper cut will frequently overcome this difficulty[11, 12].

(2) Deep oxidation of the surface. This occasionally occurs in heavy castings that have cooled slowly in the mould. A long period at a temperature below the critical may produce a deep decarburised skin of low hardness, but in which the graphite flakes are surrounded by an oxide skin. There seems little doubt that the deep penetrating oxide can be the cause of the excessive tool wear. Since the main trouble is caused by loss of tool sharpness rather than tool face wear, a deep cut is most likely to overcome the trouble (see *Figures A* and *B*, p. 252).

(3) Burnt-on sand. This may range from imperfectly cleaned surfaces to actual metal penetration into the mould surface. The former is purely a cleaning problem and in heavy castings the practice of weathering serves to rust off the surface where shotblasting may be impracticable. Shotblasting, however, is the most effective method of cleaning and, from the point of view of tool life, is preferable to tumbling for fine-grained castings. High pressure shotblasting of coarse-grained castings can occasionally give trouble as a result of embedding of chilled shot in the surface of the casting.

Metal penetration into the mould or core face produces an intractable mass of sand and metal that can only be dealt with by chipping. It is particularly likely to happen at the base of deep castings where the pressure of liquid metal is high.

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SURFACE FINISH

Apart from purely decorative purposes, a surface is usually machined, scraped, ground, lapped or honed because at some stage it has to slide on or mate with an opposing surface. The character of the surface can also markedly affect the fatigue strength of a component. It has long been apparent that the mere verbal description of the means of surface development is not a satisfactory method of defining its texture, and various systems of numerical classification are available. These systems are based on either:

μin = one micro-inch = one millionth of an inch (British and American) = 0·0254 metre (microns) = 0·000 025 4 millimetres.

or

μ = one micron = 10^{-6} metres or 0·001 millimetres (ISO IR468).
1 micron = 39·4 micro-inches (approx. 40) = 0·000 394 inches

A machined surface (*A*) as in *Figure 111*, can be resolved into three main components[1]:

(1) Primary texture (*B*), dependent particularly on the character of the cutting tool, now called Roughness (BS 1134:1961).
(2) Secondary texture (*C*), dependent particularly upon the character of the machine tool itself, such as lead screw variations, now called waviness.
(3) Errors of form (*D*), dependent on machine tool and other errors.

The measurement of surface texture (primary) is defined by three main methods:

(a) Centre line average (CLA) BS 1134:1961 (and 1950) and ISO IR468.
(b) Root mean square (RMS) American. ASA B46.1 1955 is now based on CLA.
(c) Peak valley average (PVA), Swedish and German.

The centre line average value is defined as the 'average' height above an

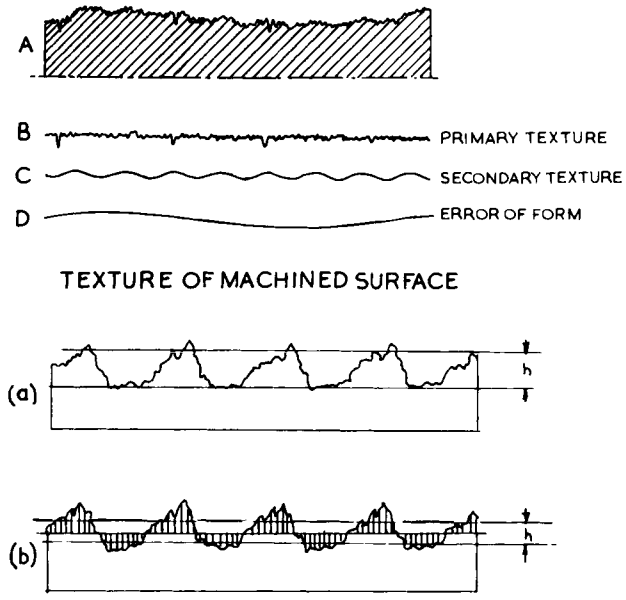


Figure 111. Methods of assessing (a) average peak-to-valley height (b) centre-line average values

established centre-line defined as a line parallel to the general direction of profile for which the areas embraced by the profile above and below the line are equal.

The root mean square system uses the same basis, but defines the variation from the centre-line position in terms of the root mean square of the peak valley heights. The peak valley average is taken as the distance between two lines, averaging the peaks and valleys.

Taking the CLA value as standard, for the same surface the RMS values would be 10–20 per cent greater and the PVA from 3 to 7 times greater (BS 1134:1961): 7 up to 12 μCLA and $\times 4$ for 300 μCLA^2 and above.

Table 147

	Standard CLA Grades	
	Micro-inches	μMetres or Microns
Polish	<1–2	0·025–0·20
Diamond grind, hone or lap	4–8	0·1–0·2
Cup grind	6–16	0·2–0·4
Hand scrape	16–32	0·4–0·8
Rough grind	20–32	0·5–0·8
Fine turning	32–63	0·8–1·6
Turning	63–125	1·6–3·2
Shaping	125–250	3·2–6·4
Rough turning	250 and upwards	6·4 and upwards

The PVA system has the advantage that it gives some indication of the dimensional change resulting from lapping off the bulk of the texture.

The CLA system has the advantage of ease of measurement and is the standard now employed in Great Britain, Canada and USA and the new ISO recommendations IR.468. For comparison of various standards of surface roughness see references 1 and 2.

The definition of surface texture is assuming increasing importance, particularly on sliding surfaces, and the preceding figures (*Table 147*) give an approximation to the values to be expected from various machining or finishing operations. Since it is difficult to measure surface texture accurately, the various grades are usually limited (*Table 147*).

Effect of graphite size on surface finish

There is a limit to the finish which can be obtained on grey iron surfaces, owing to the discontinuities provided by the graphite flakes. Specimens that have been prepared by careful polishing for microscopic examination have given the following figures[3]:

	CLA	
	μinches	(μm)
Ingot mould iron (hardness 119 Brinell), graphite size ASTM 1A	8–10	(0·2–0·25)
Light casting, approx. $\frac{1}{4}$ in (6 mm) thick as-cast (hardness 262 Brinell) graphite size ASTM 5–7B	2–3	(0·05–0·08)
Very fine undercooled graphite, ASTM 8D	$1\frac{1}{2}$–2	(0·04–0·05)

These figures are based on ground, polished and undistorted surfaces. Slightly finer finishes can be obtained by burnishing, in which metal is flowed over the graphite flakes.

Nodular, or SG, iron optically polished with diamond dust has given surface finish of 2 μinches (0·05 μm) CLA, which is of the same order as that found with fine graphite grey iron[4].

Although surface finish measurements, as given above, define certain aspects of primary texture, they do not completely define the surface, since profiles of the same height may have very different shapes. Note must be taken of secondary texture which is dependent upon the characteristics of the machine upon which the surface has been developed, e.g. lead screw variations, and so on.

Functional aspects of surface texture

It has been well established that perfect initial smoothness is not necessarily the best for sliding and bearing surfaces. Initially very smooth surfaces may

offer greater initial frictional resistance and a greater tendency to seize or scuff during running-in. For flat sliding surfaces, such as guides and slideways, scraped or cup ground surfaces are normally employed, partly on account of their oil-carrying capacity and their low initial starting load. Where cast iron is used as a shaft bearing, however, the surface finish of the shaft and the cast iron bearing itself are of great importance, particularly if a pearlitic cast iron is employed with hardness about 170 Brinell. The surface finish for both components should not be coarser than 32–63 CLA (0·8–1·6 μm) if scoring during running-in is to be avoided with heavy loads. Where one surface is of steel, the surface finish of the steel is more important than of the cast iron and should preferably not exceed 15 μinches (0·38 μm) CLA.

Normal unspecified machined finish should not be coarser than 125 CLA (3·2 μm). Diamond honing of BS 1452, Grade 10 iron, with 3·5 per cent carbon may give 2–8 μinches (0·05–0·20 μm) CLA. Scraped surfaces should give minimum bearing area of 20 per cent when tested by blueing and rubbing with a small surface plate—estimated by superimposing a 2×2 in (51×51 mm) glass plate sub-divided into $\frac{1}{10} \times \frac{1}{10}$ in ($2{\cdot}5 \times 2{\cdot}5$ mm) squares.

Cast iron cylinder bores range in surface finish from 20–60 μinches (0·51–1·5 μm) CLA and in automobiles usually from 20–35 μinches (0·51–0·89 μm) CLA when score resistance is at a maximum[5, 6] and finer or coarser finishes are said to score more readily. The upper limit is consistent with the view that the oil film is about 40–80 μinches (1·0–2·0 μm) in thickness. For steel, however, and possibly white iron, score resistance appears to improve the finer the finish. Heavy marine cast iron diesel liners (above 500 mm diameter) may deliberately operate with initial surface finishes of 120–200 CLA (3·0–5·1 μm) to facilitate running-in.

Crank pins are ground to 5–12 μinches (0·13–0·30 μm) CLA in automobiles and up to 32 μinches (0·81 μm) CLA for industrial use.

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5

Heat treatment of grey cast iron

HEAT TREATMENT

The heat treatment of cast iron falls into three main groups:

(1) *Stress relief:* Low temperature treatment for the relief of casting stresses.
(2) *Annealing:* To reduce hardness and to improve machinability.
(3) *Hardening:* By quenching to produce general or local increases in hardness.

Stress relief

It is practically impossible to produce a casting from the foundry completely free from internal stresses. In most cases these stresses are of little significance, but occasionally they can be of sufficient magnitude to cause fracture, and in other cases, to cause progressive distortion as machining operations continue. Internal casting stress is the result of one integral portion of the casting being too short or too long to fit into the space available for it so that it must be either stretched or compressed (*see* section on internal casting stress). It must therefore impose equal and opposite stresses on the adjacent parts. If the deformation is 'elastic', the casting stresses will be permanent. If plastic yielding occurs, however, progressive relief of stress will occur.

Plastic yielding can be made to occur if the temperature is raised high enough for creep to occur at the stress involved and in the time available (*see* section on heat resistance, creep). Some plastic yielding of grey iron occurs even at room temperature (*see* section on dimensional stability) and justifies the long established use of 'weathering' of critical castings before machining or at any intermediate stage of machining. Weathering, i.e. exposure to fluctuating atmosphere temperatures for long periods of time of up to months, however, has little effect on the major casting stresses and can only affect points of maximum stress. Long experience has shown that it does in fact improve dimensional stability. Heat treatment, however, is much more

certain and can be carried out in a few hours. Complete relief of stress may not necessarily be desirable since the material is then left in the condition that some plastic yielding will occur even at very low stresses and final dimensional stability under stress in grey iron will only occur after the casting has been stressed in service.

For complete dimensional stability of very critical castings they should be stressed up to service levels several times, or for a period of hours before final machining operations[1].

For normal castings, stress relief up to 80 per cent (*see Table 148*) should leave sufficient residual stress to ensure adequate stability for normal use. A residual level of stress of about $\frac{1}{2}$ to 1 ton/in^2 (8 to 15 N/mm^2) is reasonable but prestressing to service levels will ensure that casting accommodation has occurred in the direction required.

Table 148 COMPOSITION, %

	TC	*Si*	*Mn*	*S*	*P*	*Ni*	*Cr*	*Mo*
A	3·12–3·15	1·73–1·93	0·65–0·73	0·129–0·135	0·37–0·83	—	—	—
B	2·8	2·62	0·75	0·028	0·084	1·3	0·17	—
C	2·56	2·57	0·72	0·025	(0·1)	1·07	0·11	0·45

	Tensile range		*Percentage stress relief at*			
	ton/in^2	(N/mm^2)	*400°C*	*500°C*	*550°C*	*600°C*
A	18–22	(278–340)	−19–+40	28–32	56–70	78–88
B	22–25	(340–386)	0	39	51	70
C	27–32	(417–494)	24	52	67	71

The accepted practice of weathering castings by allowing them to stand in the open for several months before use serves the double purpose of removing the oxide and sand-contaminated surface by flaking corrosion, and of removing a very small percentage of the existing internal stress by subjecting the casting to fluctuating atmospheric conditions. This method, however, can only affect peak stresses since stress relief can occur only by plastic deformation and this is unlikely to be significant where residual casting stresses are less than one third of the maximum tensile strength[2, 3].

The figures in *Table 148* for percentage stress relief have been given for $\frac{1}{2}$ in to $\frac{3}{4}$ in (13 mm to 19 mm) sections for one hour at temperature, and furnace-cooled[4].

The time factor[5, 6] enters into this treatment, and *Table 149* gives figures for two grades of iron for varying temperatures and times.

It is important to bear in mind that prolonged exposure to temperatures above 620 to 650°C may produce slight structural changes that will result in some lowering of the tensile strength. Up to these temperatures, however, no appreciable change in tensile strength is likely to be experienced.

Nodular or SG iron castings may have residual stresses up to 13 to 19 ton/in^2 (201 to 293 N/mm^2) as cast, and after normalising and air cooling from 850°C, if cooling rates vary widely within the casting[5]. Precisely the same

mechanism of stress development as that given in the section on internal casting stress will occur after normalising heat treatment if cooling rates are uneven. Stress relief for 7 h at 550°C will reduce these stresses to 3 ton/in^2 (46 N/mm^2) or less.

It is of particular importance in stress-relieving to ensure that heating-up is sufficiently slow to avoid the imposition of further thermal stresses. A casting containing both thin and thick sections is particularly prone to cracking if the heating-up is rapid, as the thin sections may heat up more quickly than the thick ones. Rapid cooling may also impose further stresses, since the heavy section may cool more slowly than the light sections. Critical castings should be furnace-cooled down to at least 200°C before removing from the furnace. Heating-up speeds of large or critical castings should not exceed 10 to 15°C/h and cooling rates 20 to 40°C/h [8, 9, 10].

Table 149

Stress relief %	*Tensile strength*, 13 ton/in^2 *Anneal*, h (201 N/mm^2)				*Tensile strength*, 22 ton/in^2 *Anneal*, h (340 N/mm^2)			
	2	4	10	40	2	4	10	40
	°C	°C	°C	°C	°C	°C	°C	°C
50	482	460	454	438	—	538	482	399
80	582	566	538	499	593	566	540	540
90	595	590	566	540	621	593	566	550
100	620	610	600	593	649	621	593	593

It is occasionally suggested that treatment of castings at sub-zero temperatures may also provide some relief of stress. There is no evidence that this is the case with ordinary ferrous castings containing ferrite and pearlite, but rather that internal stresses may be increased by this treatment. There is some indication that in alloyed steels, where austenite is present, a sub-zero treatment followed by heat treatment at 200°C for 2 h can reduce internal stresses, and these circumstances may arise in certain alloyed cast irons. Sub-zero treatment, however, may cause austenite to transform to martensite or bainite (*see* section on austenitic cast iron).

Stress relief by vibration cannot be relied on for stress relief[11, 12] though it is possible it may aid dimensional stability on machining. Stress relief can only occur by replacement of elastic by plastic strain and it would appear that relief by vibration can only occur if further stresses are imposed at the points of previously stabilised maximum stress, permitting further plastic deformation to occur. If, however, the full capacity for plastic deformation has already been taken up, the imposition of further stress may cause fracture. Vibration can increase residual stress (ref. 11 discussion) but can reduce overall residual stress by 25% if the direction of vibration increases the original stress at the peak of the cycle[13].

Annealing

The process of annealing is normally carried out for the purpose of softening the casting by breaking down the pearlite. The critical temperature of cast iron (α–γ transformation) is usually between 750 and 850°C in unalloyed irons, and pearlite breakdown can take place below these temperatures; the rate of breakdown becomes appreciable at temperatures above 600°C. There is considerable evidence[14], however, to show that for unalloyed irons the most rapid and effective annealing occurs when the casting is maintained at or about 700 to 760°C, particularly when air cooling is to be employed.

The solubility of carbon in austenite rises rapidly with temperatures above the critical temperature, and with air cooling through the critical there is a greater tendency for pearlite to be retained[15]. It is advantageous, therefore, under these conditions, to keep the temperature of the casting within the range 700–760°C. Since in industrial practice the entire time cycle is of importance, however, it is customary to use furnace temperatures of 780–820°C for short-term annealing.

The presence of even small percentages of chromium (0·2 per cent to 0·3 per cent) will very markedly increase the stability of pearlite, and thus much longer annealing times will be required. The presence of nickel has little effect on the resistance of pearlite to breakdown, but will markedly increase the tendency to retain pearlite on moderately rapid cooling.

Where extreme softness is required, and the presence of free carbides is suspected, annealing at 900°C, followed by a furnace cool to 690°C holding for some hours at 690°C, followed by furnace cooling to room temperature, will serve both to break down resistant carbides and ensure freedom from retained pearlite.

There are two distinct sets of conditions in which annealing may be carried out:

(1) To break down free carbide or chill that may have formed owing to composition irregularities or the presence of rapidly cooled sections. Such chilled portions may have hardnesses ranging from 260–500 DPN, and it is important to realise that in such cases annealing must be carried out at temperatures of 850 to 900°C for periods ranging from 1 to 6 h, if complete breakdown of carbide is to be achieved[16]. This treatment should therefore be regarded as an emergency measure only, as this type of chill, wherever possible, should be avoided by using suitable compositions. The problem of designing castings to avoid the production of chilled edges is dealt with in the sections on the properties of commercial cast irons and the application of data to design.

(2) To provide castings requiring very high machining speeds. A fully pearlitic casting with a hardness of 200–240 Brinell can be annealed to give a mainly ferritic matrix with a hardness from 140 to 180 Brinell, depending on the phosphorus content. Exposure to temperatures of 780–820°C for $\frac{1}{2}$ to 2 h will usually suffice to break down the bulk of the pearlite to ferrite, provided the cooling rate is not too rapid. Air cooling of $\frac{1}{4}$ in (6 mm) sections is frequently employed. The use of

temperatures above 850°C, particularly in castings with section thicknesses of the order of $\frac{1}{4}$ in (6 mm) or less, may result in higher hardness, owing to the retention of slightly more pearlite during air cooling than would be found if temperatures of 780–820°C were employed[15].

Full annealing with slow cooling should be used where exceptionally high machining speeds are called for and a fully ferritic matrix is required. Pack annealing, using coke packing, is frequently employed for this purpose. With a 24-h cycle and a maximum temperature of 850°C, hardnesses of less than 140 Brinell can be regularly obtained with irons of up to 0·4 per cent phosphorus.

Growth of the order of $\frac{1}{64}$ in to $\frac{1}{16}$ in per foot (1/768 to 1/192) will occur during annealing on account of pearlitic breakdown.

EFFECT OF HEAT TREATMENT ON MECHANICAL PROPERTIES[17]

The hardness of annealed grey iron is dependent upon phosphorus content, graphite content and form, and silicon content.

The effect of phosphorus[17] is indicated in *Table 150.*

Table 150

TC %	*Si* %	*P* %	*Tensile strength* ton/in² (N/mm²)		*Brinell hardness*	
			As-cast	*Annealed*	*As-cast*	*Annealed*
3·3	1·8	0·08	17·4 (269)	9·9 (153)	235	100
3·3	2·0	1·2	15·3 (236)	12·8 (198)	247	147

Phosphorus has the most pronounced effect of all the elements normally present, by virtue of the high hardness of the phosphide eutectic. It also markedly retards the breakdown of pearlite and delays full annealing[15].

Ferry[18] states that the annealed hardness of irons of constant phosphorus content (0·3 per cent) is also dependent upon the total carbon content: the higher the carbon, the lower is the hardness. For irons of the same carbon content, the finer the graphite, the higher is the annealed hardness.

Silicon affects the fully annealed hardness by its effect on the ferrite; a silicoferrite with 0·82 per cent silicon has a hardness of 88, with 2·28 per cent silicon the hardness is 124, and with 3·4 per cent silicon the hardness is 150 (*see* section on hardness).

Ferry states that it is difficult to get hardnesses below 150 Brinell in cast bars of 0·4 in (10 mm) diameter after annealing, even with carbons of 3·6 per cent to 4·0 per cent, silicon 2·85 per cent and phosphorus 0·3 per cent. In this case, the effect of fine graphite, high silicon and 0·3 per cent phosphorus appears to offset the effect of very high carbon.

Annealing for 8 h at 900°C followed by slow cooling to 690°C, and holding

for 24 h, followed by furnace cooling, will give Brinell figures as low as 95 for irons with a phosphorus content of less than 0·1 per cent and silicon of about 1·6 per cent.

The effect of annealing upon other mechanical properties is also marked. The tensile and fatigue strengths may be reduced to 55 per cent to 65 per cent of the as-cast strengths. The impact strength may be slightly increased and Collaud states that the plastic and total energy at failure is always increased by annealing[19]. *Tables 151* and *152* illustrate the effect of annealing on mechanical properties of grey irons[20].

Table 151

Type of cast iron	*Composition* TC %	Si %	Mn %	S %	P %	*Tensile strength,* ton/in² (N/mm²)	*Elongation* %	BHN 3000/10	*Impact strength (BS 1349) unnotched bar* ft.lb (J)	*Impact strength* $\frac{5}{8}$ *in (16 mm) Charpy bar* ft.lb (J)
Pearlitic flake graphite cast iron	3·02	2·02	0·70	0·088	0·066	20·9 (323)	—	235	20–26·5 (27–36)	9·5 (13)
Annealed ferritic flake graphite cast iron	3·06	1·91	0·62	0·093	0·078	11·3 (175)	1	109	(20–35) (27–47)	10 (14)

Table 152

Type of cast iron	*Tensile strength* ton/in² (N/mm²)	*Unnotched bar* *Fatigue limit* ton/in² (N/mm²)	*Endurance ratio, Fatigue limit / Tensile strength*	*Notched bar* *Fatigue limit,* ton/in² (N/mm²)
Pearlitic flake graphite cast iron	20·9 (323)	10·0 (154)	0·48	7·0 (108)
Annealed ferritic flake graphite cast iron	11·3 (175)	5·5 (85)	0·49	4·5 (69)

The figures in *Tables 151* and *152* were taken from sand-cast 0·875 in (22 mm) diameter bars, 10 in (254 mm) long. The annealing cycle in both cases was 8 h at 900°C, followed by furnace cooling to 690°C, and holding at 690°C for 24 h, followed by furnace cooling to room temperature.

An investigation by Ferry[18] on unalloyed irons with phosphorus contents of about 0·3 per cent showed that annealing of pearlitic irons to the ferritic state (*Figure 112*):

(1) Caused the tensile strength to fall by 25 per cent to 40 per cent (average 27 per cent).

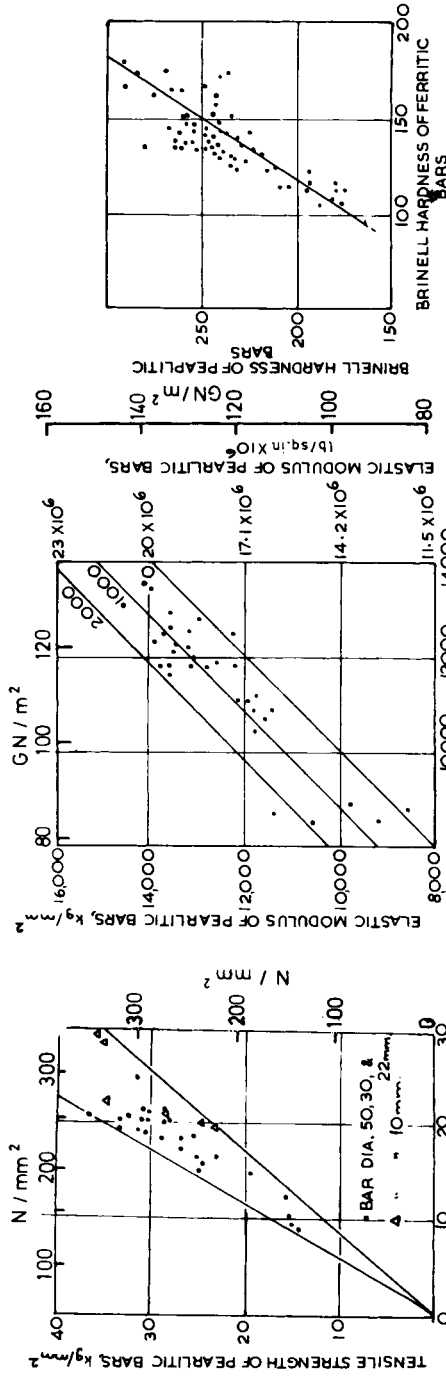


Figure 112. The effect of annealing on the mechanical properties of grey iron containing about 0·3% phosphorus

(2) Lowered the modulus of elasticity very slightly by increasing the amount of graphitic carbon. The reduction in value was practically constant over a wide range of elastic moduli and averaged approximately $1{\cdot}4 \times 10^6$ lb/in^2 (9·7 GN/m^2).
(3) Lowered the hardness from 30 to 50 per cent.
(4) Lowered the density by 0·1 to 0·2.

Irons containing chromium or molybdenum above about 0·3 per cent, however, do not readily become ferritic on annealing. Spheroidisation of the pearlite will occur and the tensile strength may fall only about 10 per cent, while the transverse rupture stress and deflection may increase (*see* section on impact strength, effect of annealing).

Annealing is also occasionally carried out on castings that are to be subjected in service to temperatures of 450–550°C. If pearlitic irons are employed within this range of temperature, slow breakdown may take place, resulting in permanent growth amounting to as much as $\frac{1}{16}$ in per ft. Preliminary annealing to remove pearlite can completely eliminate this growth within the temperature ranges quoted, so that the dimensional stability of parts operating at these temperatures can be assured[21]. The annealing temperature, however, should be sub-critical, i.e. 700–760°C.

This treatment must not be applied to castings operating in super-heated steam at temperatures above about 250–300°C[22] since it greatly increases the rate of corrosion attack.

RECLAMATION OF ANNEALED CASTINGS

Castings that have been annealed deliberately or accidentally and are too soft for their purpose, can be restored to the pearlitic state by reheating to 850–880°C and cooling in air, if light in section; by air blast if high in silicon or more than $\frac{1}{4}$ in (6 mm) in section; or by oil quenching and tempering at 650°C to the required hardness if full control of the final hardness is called for. The precise method of cooling will depend upon the section and composition.

Hardening by quenching and tempering

Heat treatment by quenching is frequently carried out where high surface hardness is needed on components requiring previous machining operations, and which therefore cannot be cast white. It is normally customary to oil-quench rather than to water-quench, as there is little advantage to be gained in hardness, and the risk of cracking appears to be substantially reduced. Untempered hardnesses of the order of 450–550 DPN are obtainable, but with a very substantial loss of tensile strength. It is, however, customary to temper the hardened castings, since although tempering up to 300°C may reduce the hardness from 10 to 50 points, it increases the tensile strength to the original unhardened figure or over. The tempered castings, therefore,

are much less susceptible to fracture. Hardening temperatures range from 820 to 880°C, but it is preferable to keep within the range of 850–880°C. The typical response of unalloyed, low-carbon, low-silicon grey iron castings to heat treatment is shown[23] in *Figure 113* and for a higher carbon alloyed iron[24] in *Figure 114*.

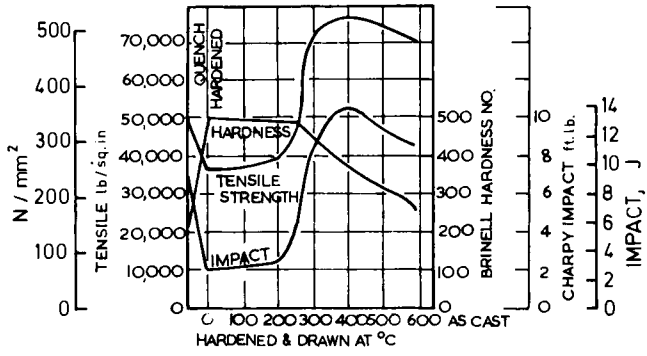


Figure 113. Effect of quench and draw on hardness, tensile and impact. GA Meehanite, approximate analysis: TC 2·75–3·0, Si 1·3–1·4, Mn 0·9–1·1, S 0·01–0·08, P 0·05–0·1%

Alloying is not strictly necessary for the production of satisfactorily hardened castings in sections of less than $\frac{3}{8}$ in (10 mm) to $\frac{1}{2}$ in (13 mm), but the presence of manganese, nickel or copper lowers the critical point

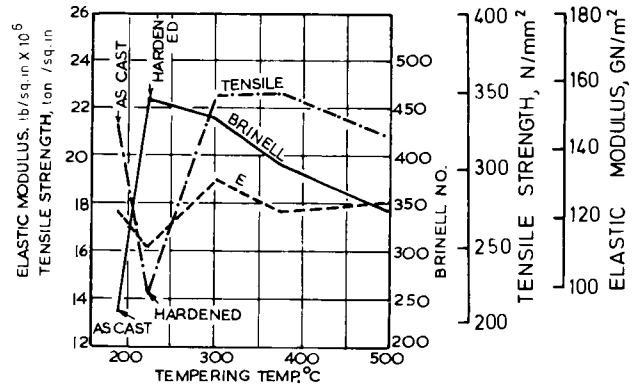


Figure 114. Brinell hardness, tensile strength and elastic modulus results of nickel–chromium cast iron (Ni 1·32%, Cr 0·44%), oil hardened from 850 to 875°C. TC 3·36%, CC 0·69%, Si 2·30%, Mn 0·67%, P 0·54%. Samples centrifugally cast in cylinders 9 in long (229 mm) × 4¼ in (108 mm) id × 10 mm radial thickness

and also decreases the critical quenching speed, as described in the section on hardenability and isothermal transformation. Additions of 1 per cent to 2 per cent nickel or copper are particularly useful in increasing the depth of hardness for a given section and quenching speed, and for this reason

the addition of nickel or copper is especially suitable for heavier sectioned castings. Additions of chromium and molybdenum are frequently made, but both these elements are very potent carbide formers, and the presence of any free carbide in the initial structure greatly increases the risk of cracking of difficult sections during quenching.

For complex castings, where molybdenum is included in the composition, the phosphorus content should be below 0·08 per cent in order to avoid the presence of a carbide/phosphide complex, which materially increases the risk of cracking in difficult sections. Simple sections, such as cylinder liners, piston rings, and so on, are not as a rule subject to cracking, and the presence of chromium and molybdenum carbides for wear-resisting purposes is quite common.

The object of heating above the critical point before quenching is to dissolve pearlite and form austenite, with as low a carbon gradient as possible. The pearlite in high silicon irons (2·5 per cent to 3 per cent silicon) is rapidly decomposed below the critical temperature, e.g. at temperatures of about 700°C, with the formation of graphite. When the temperature reaches the critical point, therefore, solution of carbon into austenite has to take place from the graphite centres instead of from the dispersed pearlite, and the carbon gradient in the austenite areas will be steep unless adequate soaking time is allowed.

Furthermore, the solubility of carbon in pure γ iron at the eutectoid point is lowered as the silicon content is increased (0·8–0·9% in pure γ iron and 0·55% to 0·7% in 2 per cent silicon iron) so that the quenched hardness of the martensite is reduced. Such irons are thus likely to show a poorer hardening response than irons low in silicon.

The effect of silicon and manganese upon the critical temperature is shown in the section on the constitution and structure of cast iron (*Figure 6*). Owing to the time involved in obtaining uniformity of temperature as well as full transformation and solution of carbon in austenite, it is essential to quench from temperatures 40–80°C above the nominal critical temperature; for iron within the normal range of 1·5 per cent to 2·2 per cent silicon and 0·5 per cent to 1·0 per cent manganese, a quenching temperature of 850–880°C is satisfactory. Temperatures much below 840°C may leave untransformed ferrite in the structure so that full hardening cannot be obtained. There is also increased risk of cracking caused by internal stresses set up by incomplete transformation.

The best response to hardening in unalloyed irons is obtained where the carbon and silicon contents are kept reasonably low; for example, total carbon 2·8 per cent to 3·2 per cent and silicon 1·3 per cent to 2·1 per cent. Phosphorus is preferably kept as low as possible to ensure freedom from brittleness, and the risk of cracking due to thermal shock. Manganese should be at least 0·3 per cent in excess of that required to balance the sulphur content ($1{\cdot}7 \times$ S per cent $+ 0{\cdot}3$).

The form of the graphite can have significant effects. In ferritic irons, fine graphite, particularly in the undercooled form, goes into solution much more readily than coarse flake graphite. Undercooled graphite structures, however, tend to have poor mechanical properties, and for all-round suit-

ability a medium-to-fine random flake graphite is most satisfactory. Castings that have been fully annealed, and are ferritic before heat treatment, tend to show a poorer response to heat treatment than a fully pearlitic casting for the reasons given previously. Where complex sections are being treated it is preferable to preheat the casting to 600–650°C, before transferring to the final furnace at the hardening temperature of 840–880°C. This ensures that the casting remains in the critical breakdown range (about 700°C) for as short a time as possible, and at the same time greatly reduces the risk of cracking due to thermal shock in heating up. The casting should be left in the furnace at final temperature for 1h, or until its colour blends with that of the furnace. Small castings, up to $\frac{1}{2}$ in (13 mm) in section and 1–2 lb (0·45–0·90 kg) in weight, may, on transfer from a furnace at 620°C to another at 840°C, merge into the furnace background too quickly to give adequate carbon diffusion, and should be held at temperature for not less than half an hour.

After soaking, the castings should be transferred rapidly to the quenching oil bath, and immersed with constant agitation if the oil itself is not being rapidly circulated. Castings prone to crack should be withdrawn from the oil bath at a temperature of approximately 150–200°C, and transferred immediately to the tempering furnace. Tempering from 1–2 h per in (25·4 mm) of casting thickness at 250°C will remove most of the quenching stresses without appreciably reducing the hardness. Tempering to 300–400°C restores fully the original tensile strength, and lowers the hardness by 10–50 points.

For heat treatment by quench and temper methods to be effective, the iron must have been heated above the α to γ transformation point for a sufficient length of time for the austenite to be saturated with carbon. The temperature at which the transformation commences is shown for irons of varying silicon content in *Figure 6*, and the critical temperature range is discussed in the section on constitution and structure. The temperatures given in *Figure 6*, however, indicate the commencement of the transformation and solution of carbon only, and it is important that the casting should be heated and soaked well above this temperature if heat treatment is to be successful. The temperature to which the casting is raised before quenching should be 60 to 100°C above the temperature given in the curve if the effective soaking time is to be kept within reasonable limits and complete α to γ transformation is to occur.

Pearlitic irons respond to heat treatment much better than annealed or ferritic irons, since the carbon is already dispersed in the matrix as pearlite, ready for solution when the α to γ transformation occurs. In a ferritic iron, however, carbon must diffuse from the graphite flakes into the matrix. Unless sufficient time is allowed for this to occur at a temperature 60–100°C above the critical, the carbon level in the austenite areas will be low and the iron will show a poor response to hardening.

The quenching process takes the casting through the α to γ change too rapidly for the carbon to be thrown out of solution as carbide to form pearlite, and a series of intermediate structures may be formed (martensite, bainite etc.) depending on the character of the quench. These intermediate structures are responsible for the changes of tensile and hardness which are

characteristic of the process. The speed of the quench is dependent upon the quenching medium, air, oil or water, and notably upon the section of the casting.

Quench cracking in steel is said to be avoided by the use of 0·05% polyvinyl alcohol. The alcohol should be dissolved in methylated spirits and then added to the water.

HARDENABILITY AND ISOTHERMAL QUENCHING

Hardenability is the term used to indicate the depth of hardening that will be found in a given section quenched in a given medium. This is of great importance since heavy sections cool much more slowly than light sections and require material of high hardenability if satisfactory response to quenching heat treatment is to be obtained. Davenport and Bain[25] studied the effect

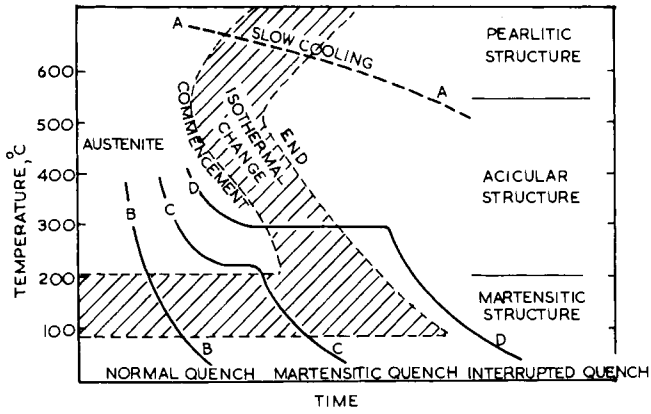
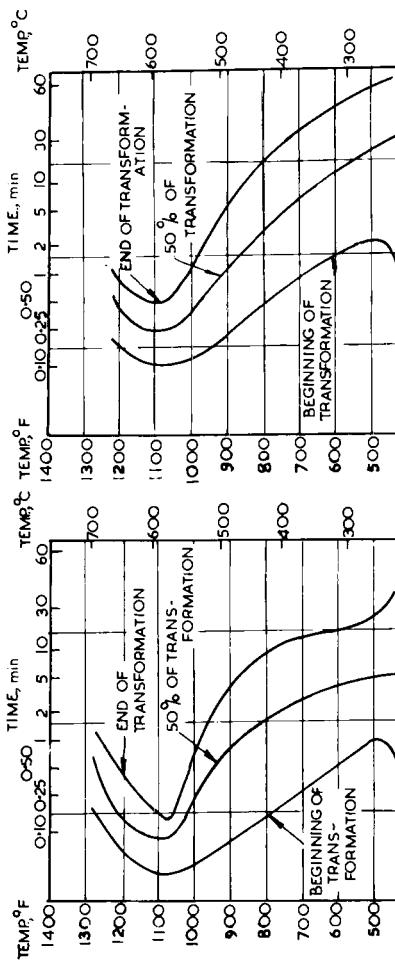


Figure 115

on steels of cooling below the critical temperature upon the formation of martensite and other intermediate structures. By examining the structures produced by holding quenched specimens at any desired temperature, they devised the S-curve to illustrate the effect of time at a given temperature upon structure. A great deal of work has been carried out since, and an idealised curve[26] is shown in *Figure 115*. Curves are shown in *Figure 116* for an unalloyed grey iron and one carrying 2 per cent of nickel[27].

It will be seen that if transformation occurs above about 500–550°C, as in normal air cooling, a pearlitic structure will occur. If the specimen is held long enough at a temperature of, say, 300–450°C, an acicular or bainitic structure will be produced, and if cooling is sufficiently rapid, martensite will be formed. Acicular as-cast structures can be produced by additions of nickel or copper and molybdenum and are dealt with in the section on acicular cast iron.

The position of the nose of the curve at 500–600°C defines the hardenability of the iron–carbon alloy and is very markedly affected by various



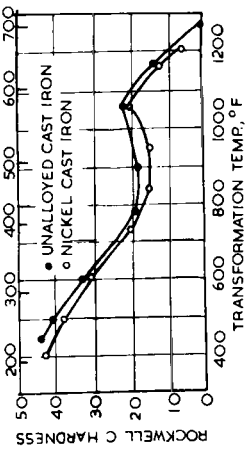


Figure 116. Isothermal transformation of two cast irons after austenitising at 845°C for 20 min

elements which may be present. Manganese, copper, nickel, chromium and molybdenum move the nose to the right and therefore increase hardenability. Carbon in the matrix up to the eutectoid content, and silicon and phosphorus, also increase hardenability, whereas cobalt moves the nose to the left and decreases hardenability.

These alloys also affect the temperature of the martensite change point (M_s) as indicated by the Nehrenburg formula:

$$M_s\ (°C) = 500 - 300 \text{ carbon} - 33 \text{ manganese} - 22 \text{ chromium} - 17 \text{ nickel} - 11 \text{ silicon} - 11 \text{ molybdenum}$$

In normal unalloyed cast irons the effects of silicon and manganese are of particular importance. An iron–carbon alloy containing 0·5 per cent carbon and very low manganese and silicon (less than 0·01 per cent) cannot be quenched successfully in any normal medium, since the nose of the S-curve is reached in less than half a second[23]. The effect of manganese is illustrated in three steels[28] in *Table 153.*

Table 153

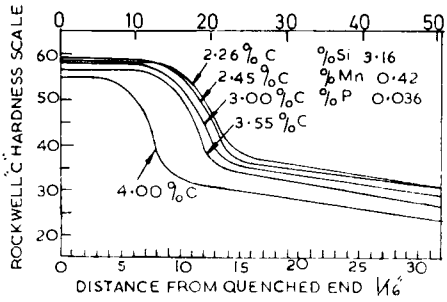
C %	*Si* %	*Mn* %	*Time for transformation* s	*Temperature* °C
0·88	0·41	0·41	9 (50%)	540–550
0·89	0·15	0·29	2 (50%)	560
0·95	0·19	0·45	9 (20%)	540

Silicon appears to have a similar, though less pronounced, effect. The results[28] for four grey irons and one malleable iron are shown in *Table 154.*

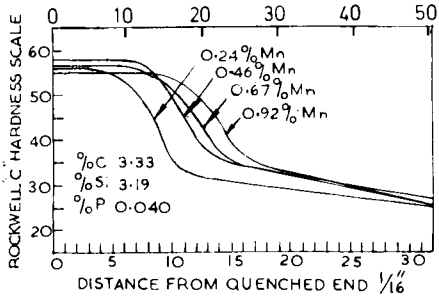
Table 154

Type of iron	*TC* %	*Si* %	*Mn* %	*S* %	*P* %	*Time for transformation*, s		*Temperature* °C
						Beginning	*End*	
Grey	2·5	2·5	0·85	0·1	0·1	9 (5%)	20	590–600
Grey	2·97	2·0	—	—	—	8	30–50	530–600
Grey	2·97	2·02	0·82	0·08	0·3	9	200	530–540
Grey	3·63	1·75	0·53	0·1	0·56	2–3	9	580–590
Malleable	2·55–2·65	1·3–1·45	0·42–0·47	0·11	0·055	3	28	570–590

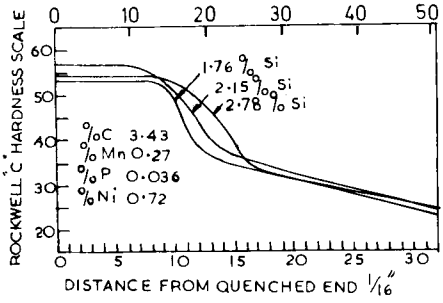
The effective manganese content in these irons is much lower than the figure given, since at least 1·7 times the sulphur content is locked up as manganese sulphide. The poor response to quenching of some pearlitic malleable irons is directly associated with the low silicon and manganese contents, and where hardening by oil quenching is required, the free manganese content after allowance has been made for sulphur should be not less than about 0·5 per cent for silicon of the order of 1 per cent. With a



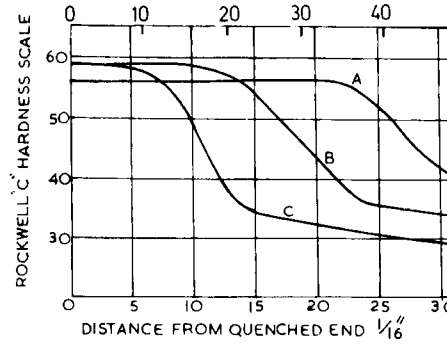
EFFECT OF CARBON ON NICKEL FREE IRON



EFFECT OF MANGANESE ON NICKEL FREE IRON



EFFECT OF SILICON ON 0·72% NICKEL IRON



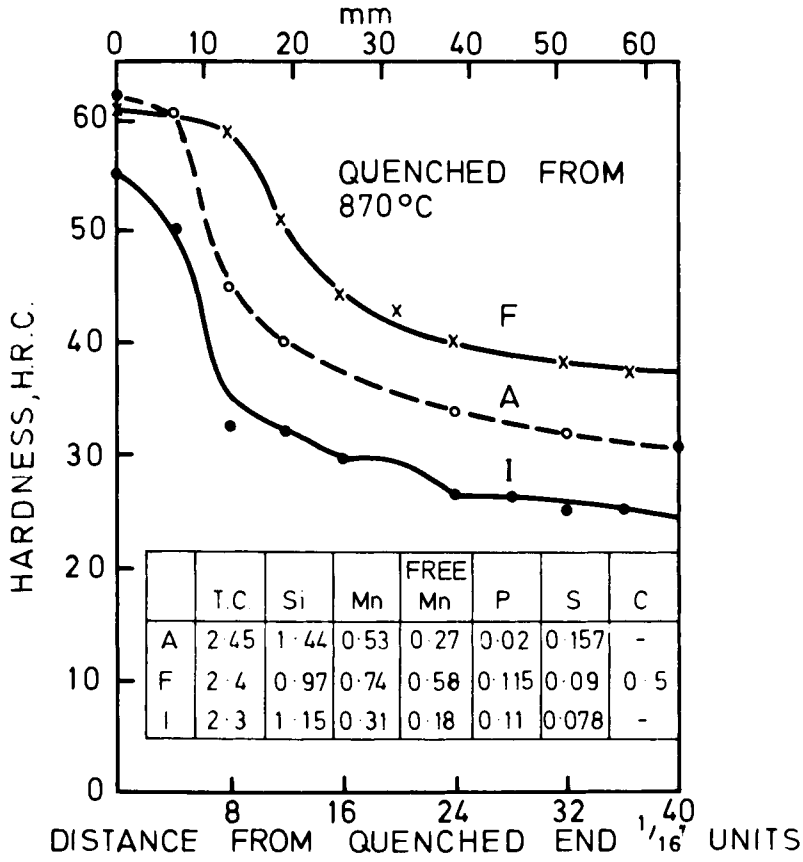
EFFECT OF MANGANESE AND NICKEL

SAMPLE	%C	%Si	%Mn	%Ni	%Mo
A	3·27	2·35	0·45	2·56	-
B	3·27	2·35	0·03	2·45	-
C	3·58	2·26	0·03	0·99	-

Figure 117. Jominy end-quench (water) test on 1 in (25 mm) diameter nodular iron bars

lower silicon content (0·5 to 0·8 per cent), manganese up to 1 per cent may be needed if complete hardening is required in sections of 1 in (25 mm) or more in thickness.

The effect of carbon, manganese, silicon and nickel on hardenability is shown in *Figure 117*, and the effect of manganese on the hardenability of three blackheart malleable irons is clearly shown in *Figure 118*.



	T.C	Si	Mn	FREE Mn	P	S	C
A	2·45	1·44	0·53	0·27	0·02	0·157	-
F	2·4	0·97	0·74	0·58	0·115	0·09	0·5
I	2·3	1·15	0·31	0·18	0·11	0·078	-

Figure 118. Effect of manganese on hardenability Jominy hardness tests on three blackheart malleable irons quenched from 872°C. Quenching from 817°C lowers the hardness below the surface by 3 to 7 points

Cooling rates, however, are also very dependent upon specimen size, and *Figures 119* and *120* show characteristic cooling rates in the centres of cylindrical steel bars of varying diameter, both with air and oil cooling[29]. If these curves are superimposed over the S-curves in *Figure 116*, it will be seen that only in very thin sections is the cooling rate by oil quenching fast enough to avoid intersecting the pearlite nose which represents the onset of transformation. It is not, however, strictly accurate to superimpose the cooling curves on the S-curves, since the latter indicate the period of

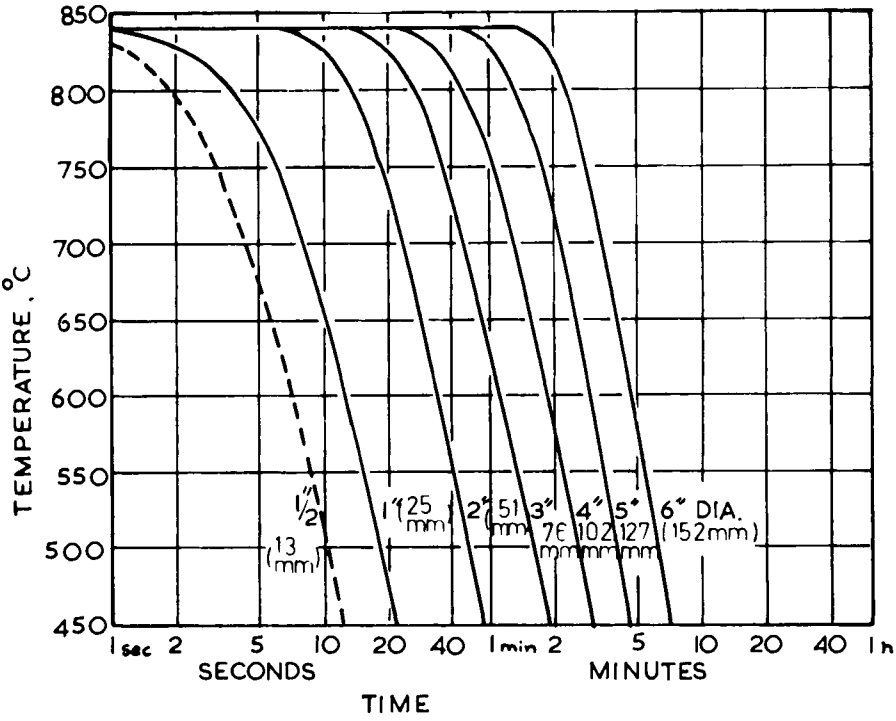


Figure 119. Air-cooling curves for the axial position of $\frac{1}{2}$ to 6 in (13–152 mm) diameter bars

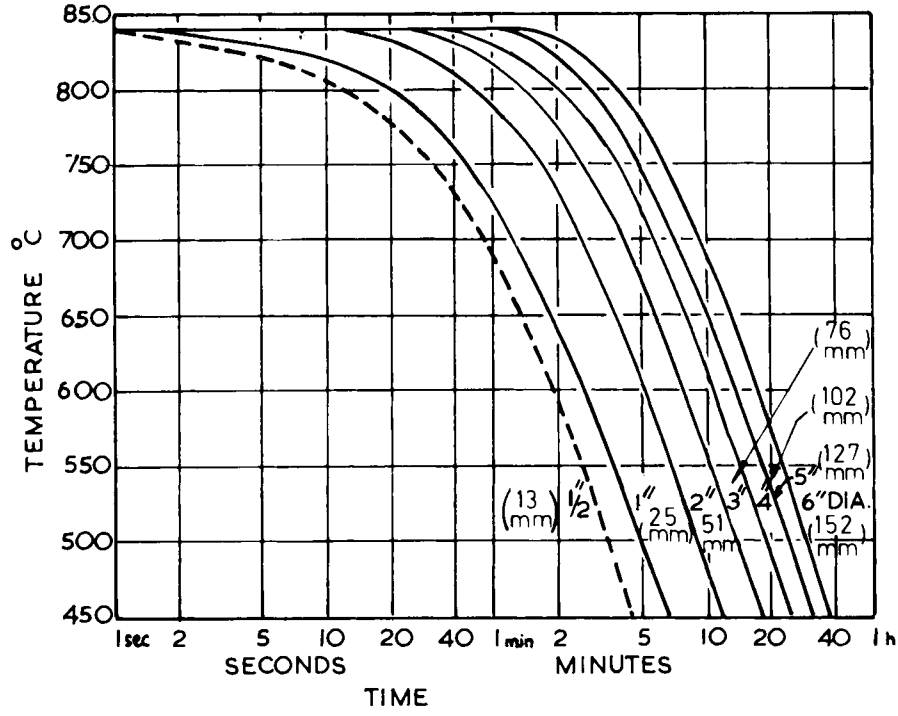


Figure 120. Oil-cooling curves for the axial position for cylinders of steel or varying diameter of $\frac{1}{2}$ to 6 in (13–152 mm) diameter bars

time *at* a given temperature, while the cooling curves illustrate only the temperature at a given moment of time, and the specimen may not have been within the pearlite transformation region long enough for transformation to begin. The cooling curves and the S-curves, however, illustrate the difficulty of deep hardening in unalloyed irons.

Hardenability can be determined by the Jominy[30] 'end-quench' method in which a cylindrical bar 1 in (25·4 mm) diameter and 3 to 4 in (76 to 102 mm) in length is heated to the required temperature, and after 30 min is quenched by exposing the flat bottom end of the bar to a standardised jet of water. By grinding flats along the bar it is possile to obtain a hardness gradient dependent upon the cooling speed of the bar and its hardenability. This method has been standardised in this country[31].

Hardenability curves for American pearlitic malleable irons have been published by Heine[32] (*Figure 118*) and for nodular iron[33, 34] with varying amounts of carbon, silicon, manganese and nickel (*Figure 117*).

INTERRUPTED QUENCH

The effect of interrupted quench is shown in *Figure 115*.

Curve A (slow cooling) passes through the entire pearlite change zone in the neighbourhood of 650°C, and the resulting structure will be pearlitic. Low transformation temperatures tend to produce finer pearlite than high temperatures.

Curve B illustrates normal quenching in which the speed of cooling has been sufficiently rapid to retain austenite down to temperatures of 200–250°C. The resulting structure will be martensitic.

Curve C illustrates a delayed martensite quench in which the austenitised specimen is quenched into a salt bath at 220–250°C before the commencement of transformation for a period of time sufficient to equalise the temperature in the various parts of the specimen. The transformation to martensite which then takes place on subsequent air cooling occurs more uniformly throughout the specimen, and avoids the differential volume changes which frequently cause cracking as a result of the transformation occurring at different parts of the casting.

According to Guillemeau[35] the period of delay in the salt bath should be approximately:

5 min for thicknesses below 20 mm.
5–10 min for thicknesses between 20–30 mm.
An extra 3 min for each increase in thickness of 10 mm.

Maintenance in a salt bath serves to equalise the temperature in the castings before any formation of martensite occurs, and the martensitic change takes place during the subsequent air cooling.

Curve D illustrates a quench designed to produce an acicular or bainitic structure and takes place in three stages:

(1) The specimen is quenched from the austenitising temperature in a salt bath at 300–450°C.

(2) The specimen is held in the bath at this temperature sufficiently long to permit the austenite to transform complete to the acicular or bainitic structure, which is softer and less brittle than the martensite.
(3) Air cooling.

If the time in the salt bath is too short, incomplete transformation of the austenite may occur, and on air cooling the retained austenite will transform to martensite, giving a mixed martensitic/bainitic structure which is both harder and more brittle than the fully bainitic structure.

The processes of interrupted quench require careful working out by preliminary tests for any given material, since the S-curves are approximate only and will vary in position with composition and the temperature of austenitisation. The most effective use of interrupted quench methods is attained with alloy irons containing, for example, nickel or molybdenum in which the transformation times have been extended, as shown in *Figure 116.*

SALT BATH HARDENING

Salt bath hardening of grey irons can be carried out by oil quenching from a cyanide bath. For example, finish-ground grey iron cam plates about $\frac{1}{2}$ in (13 mm) thick (TC 3·25, Si 2·24, Mn 0·62, S 0·14, P 0·14 per cent, unalloyed) heated for one hour at 850°C, oil quenched and tempered at 140–160°C for one hour gave surface hardness 400–410 HV and matrix (micro) hardness 690–760 HV (200 gram load). The bulk hardness below the surface makes the cam face more resistant to 'Brinelling' and rippling by the hard cam follower (sintered carbide) and an increase in life of at least 5 to 10 times has been observed.

The absence of pitting by scaling and distortion by heat treatment makes possible full machining and finish grinding before heat treatment.

THE EFFECT OF QUENCH AND TEMPER TREATMENT UPON THE MECHANICAL PROPERTIES OF CAST IRON

Direct quenching increases hardness, lowers tensile and modulus of elasticity and impact strength. Tempering will restore the tensile strength and may increase it markedly, while lowering the hardness 10 to 15 points. The impact strength and the modulus of elasticity are also restored.

The curves in *Figures 113* and *114* show typical response to heat treatment for two types of cast iron.

The two main properties that it is usually desired to augment are:

(1) Tensile strength.
(2) Hardness (usually to improve wear resistance).

Low total carbons and low silicon contents, with or without alloy additions, are required to obtain a substantial increase in tensile strength, but

there is some evidence to show that the maximum tensile strength obtainable from a given composition is dependent partly upon the graphite size, and provided the quenching speed exceeds a critical figure, a fairly fixed increment on the tensile strength may be expected owing to changes in the matrix.

In the examples given in *Table 155*, the same metal was poured into bars of differing section, all of which were oil-quenched in their original section and then tempered to approximately the same hardness. All bars were oil-quenched from 870°C.

Table 155

As-cast properties				*Heat-treated properties*				
Dia. as-cast, in (mm)	*Brinell hardness, as-cast*		*Tensile strength*, ton/in² (N/mm²)	*Tempering temperature* °C	*Tempering time* h	*Brinell hardness after tempering*		*Tensile strength*, ton/in² (N/mm²)
						Edge	*Centre*	
1·2 (30)	255		25·9 (400)	600	½	—	350	32·0 (494)
1·6 (41)	255		21·4 (331)	600	½	—	350	26·7 (412)
	Centre	*Edge*						23·2 (358)
3.0 (76)	229	241	17·7 (273)	650	1½	363	321	24·7 (381)

These figures show an increment of 5·3–6·3 ton/in² (82–97 N/mm²) in all cases on the basic as-cast tensile strength. 1·2 in (30 mm) and 1·6 in (41 mm) diameter bars were then cut from the centres of the 3 in (76 mm) diameter as-cast bars and treated similarly (*Table 156*).

Table 156

	Brinell hardness	*Tensile strength as-cast* ton/in² (N/mm²)	*Tensile strength after oil-quenching and tempering*, ton/in² (N/mm²)
1·2 in (30 mm) bar cut from centre of 3 in (76 mm) bar	229	17·7 (273)	23·4 (361)
1·6 in (41 mm) bar cut from centre of 3 in (76 mm) bar	229	17·7 (273)	23·2 (358)

It will be seen that the tensile specimens obtained from quenched and tempered 1·2 in (30 mm) and 1·6 in (41 mm) bars cut from the centre of the as-cast 3 in (76 mm) diameter bar gave tensiles similar to those obtained from the quenched 3 in (76 mm) bar, but markedly lower than those obtained from quenched and tempered as-cast 1·2 in (30 mm) and 1·6 in (41 mm) bars. Thus, increasing the quenching speed of material cast into a 3 in (76 mm) diameter bar has not increased the tensile strength to the figures given by similar quenching speeds on similar iron with a finer graphite size.

Table 157

Unalloyed iron composition %	*Treatment*	*Properties*	*2·1 in (53 mm) as-cast bar*				*4 in (102 mm) as-cast bar*					
			Edge		*Centre*		*Edge*		*Midway*		*Centre*	
TC, 3·06 Si, 1·36	As-cast	UTS, ton/in² (N/mm²) Brinell	19·1 210	(295)	14·6 200	(225)	18·6 195	(287)	15·5 200	(239)	14·4 203	(222)
Mn, 0·8 S, 0·063	Oil-quenched and tempered 550°C, 2–4 h	Tensile, ton/in² (N/mm²) Brinell	27·5 (+8·4) 286	(425) (+130)	21·2 272	(327)	27·5 (+8·9) 284	(425) (+138)	20·6 270	(318)	18·0 265	(278)
P, 0·053	1·2 in (30 mm) bars cut from centres of 2·1 in (53 mm) and 4 in (102 mm) bars and tempered 550°C, for 1 h	Tensile, ton/in² (N/mm²) Brinell	— —	—	22·4 285	(346)	— —	—	—	—	20·5 277	(317)

Table 158

Ni–Mo iron: composition %	*Treatment*	*Properties*	*2·1 in (53 mm) as-cast bar*				*4 in (102 mm) as-cast bar*					
			Edge		*Centre*		*Edge*		*Midway*		*Centre*	
TC, 2·90	As-cast	UTS, ton/in² (N/mm²)	23·8	(368)	19·3	(298)	21·2	(327)	17·95	(277)	17·3	(267)
Si, 1·26		Brinell	259		246		232		243		227	
Mn, 0·89	Oil-quenched and tempered	Tensile, ton/in² (N/mm²)	29·9 (+6·1)	(462) (+94)	25·0	(386)	28·2 (+7·0)	(436) (+109)	22·8	(352)	20·8	(321)
S, 0·065	550°C, 2–4 h	Brinell	331		317		325		307		300	
P, 0·056 Ni, 1·38 Mo, 0·41	1·2 in (30 mm) bars cut from centres of 2·1 in (53 mm) and 4 in (102 mm) bars oil-quenched and tempered 550°C, for 1 h	Tensile, ton/in² (N/mm²) Brinell	— —		24·6 328	(380)	— —		— —		22·6 337	(349)

Table 159

Cu–Mo iron: composition %	*Treatment*	*Properties*	*2·1 in (53 mm) as-cast bar*				*4 in (102 mm) as-cast bar*					
			Edge		*Centre*		*Edge*		*Midway*		*Centre*	
TC, 2·9	As-cast	UTS, ton/in^2 (N/mm^2)	24·9	(385)	19·1	(295)	21·6	(334)	19·1	(295)	18·3	(283)
Si, 1·31		Brinell	251		241		246		256		245	
Mn, 0·86	Oil-quenched and tempered 550°C, 2–4 h	Tensile, ton/in^2 (N/mm^2)	32·3 (+7·4)	(499) (+114)	24·0	(371)	30·8 (+9·2)	(476) (+142)	26·0	(402)	24·3	(375)
S, 0·07		Brinell	328		326		329		319		318	
P, 0·06	1·2 in (30 mm) bars cut from centres of 2·1 in (53 mm) and 4 in (102 mm) bars as cast, oil-quenched and tempered 550°C, for 1 h	Tensile, ton/in^2 (N/mm^2)	—		24·6	(380)	—		—		21·6	(334)
Cu, 1·27		Brinell	—		243		—		—		336	
Mo, 0·4												

A similar series of tests[36] was carried out with pearlitic unalloyed and alloyed irons (*Tables 157–159*). In all cases, castings were preheated to 650°C, taken to 870°C, soaked for 40 min, oil-quenched, and tempered 1 h per in (25·4 mm) section.

In each case it will be seen that the increment in strength from as-cast, to oil-quenched and tempered, is approximately the same at the edges both of the 2·1 in (53 mm) and 4·0 in (102 mm) bars. At the same time, with 1·2 in (30 mm) diameter bars cut from the centres of the 4·0 in (102 mm) bars and heat treated so that their quenching speed is the same as that of similar material cut from the centres of 2·1 in (53 mm) bars, the maximum tensile obtained is in all cases lower with the bars cut from the heavier section.

It is apparent therefore, that with heavy section castings the maximum tensile in the centre, due to heat treatment, is likely to be considerably less than that found in a thinner section of the same composition, and the difference in tensile from edge to centre of heat-treated heavy section castings is a function of the actual structure itself as well as the quenching speed. The resistance of the alloyed material to tempering is very marked. Castings should be removed from the quench bath at 150–200°C and transferred to a furnace at 160°C and held for 6 h or at 200°C for 2 h to provide stress relief after quenching.

FATIGUE STRENGTH

Work by Eagan[37] and Moore and Picco[38] has shown that heat treatment to obtain high tensile strength does not give a proportionate increase in endurance limit, and under some conditions the endurance limit may be lowered by such treatment. There is some indication that heavy sections are more likely to show no change, or an actual decrease, in fatigue strength. Before using heat treatment to increase the strength of a component subject to fatigue, the suitability of the material should be determined.

CORROSION OF HEAT-TREATED GREY IRON

According to Piwowarsky[39], an oil-hardened unalloyed grey iron showed a marked increase in corrosion rate when compared with the as-cast material. Tempering between 350 and 550°C increases the corrosion rate still further. Using *N*/5 HCl the following results were recorded:

(The iron was a 30 mm diameter sand-cast bar with approximately 3·01% total carbon and 1·71% silicon, unalloyed)

Treatment	*As-cast*	*Oil-quenched 850°C*	*Oil-quenched and tempered 350°C*	*Oil-quenched and tempered 700°C*
Weight loss in g/m²/day	100	300	680	110

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SURFACE HARDENING

There are many applications in engineering practice that call for a hard surface to be applied subsequent to the machining or grinding of cast iron components. Hardnesses of 470–550 DPN are obtainable with depths of

0·06–0·125 in (1·5–3·2 mm), although the martensitic layer may not exceed about 0·07 in (1·8 mm) on a flat surface. Surface hardening can be carried out using oxy-acetylene or oxy-coal gas flames, or by induction hardening followed by immediate water quenching. In either case, the maximum effective depth of hardening is about the same as that quoted above, except on cam noses, where owing to the configuration greater depths are possible without overheating the surface. Stress relief before treatment will reduce liability to cracking.

The greatest hardness appeared to be obtained on low phosphorus irons, with manganese contents of about 0·4 per cent to 0·5 per cent and silicon contents of 1·0 per cent to 1·6 per cent, when flame hardening will give figures of the order of 540–600 DPN. Hardnesses ranging from 490–596 have been obtained with phosphorus contents of 0·2 per cent to 0·8 per cent. An increase in manganese from 0·5 per cent to 1·0 per cent, or additions of nickel from 0·5 per cent to 2·4 per cent, lower the maximum hardness obtainable, which is of the order of 470 for normal silicon contents. Raising the silicon content above 2 per cent also tends to reduce the maximum hardness[1]. Additions of copper appear to act in the same way as nickel[2].

In general, the depth of hardness in these instances is of the order of 0·06–0·083 in (1·5–2·1 mm). With this depth of hardness no melting of phosphide eutectic is observed. Where a depth of hardness of 0·1 in (2·5 mm) or more is produced, particularly in the high phosphorus irons, there is a marked tendency for the phosphide eutectic to melt, since the surface temperatures may exceed 1000°C. Under these circumstances, the phosphide eutectic absorbs neighbouring graphite, and tends to segregate, giving rise ultimately to a pitted surface. This is particularly liable to occur where hardness depths of the order of 0·1–0·125 in (2·5–3·2 mm) are obtained[1].

Unalloyed irons respond well to surface hardening, provided the carbon equivalent value is below about 4·1 and the silicon is below 2·2 per cent. Parts designed for flame hardening should preferably have phosphorus contents below 0·4 per cent to reduce the risk of pitting or cracking, although hardening can be successfully carried out up to 0·9 per cent phosphorus provided the carbon equivalent value is kept below about 4·1[3]. It would be preferable to operate with phosphorus contents below 0·2 per cent, where maximum depth of hardness is called for. The addition of nickel[1] or copper[3] does not increase the depth of hardness, since in addition to speed of quench, the main factors governing hardness depth are the depth of skin of metal above the critical temperature and the time this layer remains at temperature before quenching.

Additions of chromium from 0·2 to 0·7 per cent, with a corresponding rise in the silicon content where necessary to keep the casting grey, have also been employed, in particular for camshafts, and nose hardnesses of over 600 DPN have been obtained $\frac{1}{32}$ in (0·8 mm) from the surface, falling away to approximately 465 at a distance of $\frac{7}{32}$ in (5·6 mm) from the nose. Chromium and molybdenum delay the solution of carbides in austenite and may slow up the treatment.

The tendency to produce pitting with flame hardening is not solely the result of the high phosphorus content, but is related also to the graphite size. Pitting is much more likely to occur with a coarse than with a fine graphite, and the closest control of the flame characteristics and the maximum surface temperature attained is essential if pitting is to be avoided with moderately coarse-grained graphite irons.

For effective hardening, the structure must be fully pearlitic, since if any substantial quantity of ferrite is present, either in the as-cast state or because of preliminary annealing, much higher surface temperatures and much longer periods of time are required to ensure satisfactory solution of carbon in the matrix from the graphite flakes. The response of ferritic irons will, therefore, be much less satisfactory than that of pearlitic irons. Free carbide should also be absent as it increases the risk of cracking.

Flame hardening of nodular cast iron can be carried out successfully provided the casting is pearlitic; hardnesses of up to 750 DPN are possible. With ferritic castings, imperfect response is obtained as martensite is formed only around the graphite nodules. Care must be observed with pearlitic irons to avoid cracking[1]. Water quenching will give highest hardnesses, but where sufficient mass is available and the castings are fully pearlitic, self-quenching is said[4] to give hardness figures up to 58 R_c.

Nitriding

Other methods of increasing the surface hardness of irons include nitriding[5]. This process can only be carried out on special cast irons containing both aluminium and chromium, which are normally supplied only for centrifugal castings. After machining and degreasing, the castings are heated for a prolonged period at 500°C in dry ammonia vapour, and surface hardnesses in excess of 900 DPN are obtainable.

Although this process can only be applied to alloyed irons of the type described and is not successfully applied to ordinary grey irons, austenitic cast iron is said to give hardnesses of about 500 Brinell when treated in a similar manner.

Salt bath nitriding[6]

Salt bath nitriding can be applied to grey, nodular or malleable cast irons by heat treatment for a half to two hours at 530 to 570°C in a bath of alkaline sodium cyanide followed by tempering at 300°C. Quenching from the original bath is not required. This provides a nitrided skin up to 0·0004 in (10 μm) thick with hardness up to 790 HV which improves greatly the lubricated sliding wear resistance and also increases fatigue resistance.

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FLAME HARDENING OF MALLEABLE IRONS

Pearlitic blackheart malleable iron can be flame hardened. The following are the points to be observed:

(1) Lamellar pearlite structures respond more readily than spheroidised structures.
(2) An increase in the manganese content increases the hardenability and also the depth of case in irons of similar microstructure provided the necessary depth of temperature penetration is achieved. Manganese should be 0·5 per cent or more (*see* section on heat treatment).
(3) The temperature is critical because a minimum temperature is required for austenite homogenisation—there is also a maximum permissible temperature, otherwise fusion or burning may take place.
(4) Increasing the heating time produces a more uniform and deeper case.
(5) Oil quenching is unsatisfactory, and water quenching is necessary to produce a hardness of about 482 BHN and a case depth of 0·060 in (1·5 mm).

Optimum results were obtained by quenching from 815°C into agitated water.

Ferritic blackheart malleable irons will only respond to this type of treatment if sufficient time is allowed for re-absorption of carbon into the matrix.

WELDING OF CAST IRON

Welding may be employed for the repair of broken or worn parts, for the reclamation of foundry or production defects, or for the actual addition of structural components.

Welds can be divided into the following categories:

(1) Non stress-bearing welds: cavity filling.
(2) Non stress-bearing welds: machinable.
(3) Stress-bearing welds.
(4) Build-up of wear-resisting areas.

In class (1) machinability may not be of importance, in class (2) the avoidance of hard spots is imperative, and in class (3) strength and reliability

are the most important requirements. Class (4) requires a harder grade of weld material than the others and may involve special requirements.

The commonly used methods of welding flake graphite cast irons can be broadly divided into: (a) fusion welding, in which the parts to be welded are melted; and (b) non-fusion welding, in which the parts to be welded are not melted, e.g. brazing and powder welding.

There is an extensive literature on the subject of cast iron welding, and reviews of the process have been published by Spraragen and Claussen[1] and by Pearce[2]. Two important papers dealing with the strength, hardness and brittleness of various forms of welding have been published by Dunn[3] and Morton[4]. Powder welding has been described by Westwood and Lamb[5,6] and Wirtz details methods and welding rod compositions for grey iron[7].

The choice of method for a particular application has been considered in an Institute of British Foundrymen report[8] published in 1950/51.

A detailed and comprehensive review of the welding, joining and cutting of grey iron, by C. O. Burgess, has been published by the Gray Iron Founders' Society[9].

The methods employed for fusion welding are:

(1) Oxy-acetylene gas welding using ferrous (cast iron) filler rods.
(2) Metal-arc welding using ferrous electrodes.
(3) Metal-arc welding using non-ferrous electrodes.

The method commonly employed for non-fusion welding is the so-called 'braze' welding process which has the advantage that it requires less pre-heat than fusion welding to produce dependable welds. It can also be used for joining white iron[9].

Oxy-acetylene gas welding using ferrous (cast iron) filler rods

This is a form of fusion welding in which the heat to affect the required fusion is supplied by means of an oxy-acetylene flame. It is usual to add the filler metal in the form of a filler rod to form the welded joint.

After suitable preparation and cleaning of the parts to be welded, it is customary to pre-heat the work, so that a uniform temperature not exceeding 600°C is attained. If this temperature is exceeded, there is a danger that structural changes will occur in the iron, with a consequent deterioration in the physical properties of the material. The primary object of pre-heating is to prevent too steep a temperature gradient occurring between the welded joint and the parent metal.

The method of making the weld has been described in detail in the literature[2,8,10,11]. After the weld has been completed it should be allowed to cool slowly in order to minimise internal stresses and the occurrence of hard spots in the weld zone.

With this method of welding, it is desirable to use a filler rod whose composition is such that the composition of the deposited metal approximates to the chemical composition of the part to be welded. In practice, it is customary to use a high silicon (3·5 per cent max. silicon) filler rod, to obtain a

machinable deposit, and to give good weldability. Filler rods with a lower silicon content are used where a harder deposit is required. Rods containing between 1·25 and 1·75 per cent nickel are also used where a high quality weld metal is required. Details of typical chemical compositions of these filler rods are given[12] in BS 1453:1957 (*Table 160*). It is necessary to employ a flux when using this method of welding.

Metal-arc welding using ferrous electrodes

In this method, the weld metal is deposited by means of an electric arc drawn between the parts to be welded and a metal electrode.

The parts are usually pre-heated and the sides of the joint are built up with the filler material before completing the joint. The parts are suitably prepared and pre-heated before carrying out the welding operation. Details regarding the welding current to be used and the technique to be employed are determined by the electrode used, and are usually based on details given by the electrode manufacturer.

The electrode normally used with this method is a steel electrode with a covering of flux which facilitates welding. The use of stainless steel electrodes has also been reported.

Metal-arc welding using non-ferrous electrodes

The principle of this method and the technique employed are essentially the same as those already described, where a ferrous electrode is used.

The part to be welded may be pre-heated, but with cast iron the more usual practice is to use the so-called 'cold-welding' process, where the sides of the joint are built up with the filler metal before completing the joint, and in this process no pre-heating is employed.

Flux-coated electrodes are used, the commonest electrode employed having a core wire of nickel–copper alloy, although electrodes based on nickel and of aluminium bronze are also used.

Non-fusion welding or braze welding*

This is particularly suitable where it is difficult to pre-heat the castings to the temperature required for fusion welding and free machinability is required. It is suitable for malleable iron. The cast iron is not melted, and the oxy-acetylene flame used supplies only the heat necessary to melt the filler rod, which adheres by 'wetting' the casting surface with as little surface alloying as possible.

The term 'bronze' welding is well-established, but is actually a misnomer,

* Brazing is distinguished from 'braze' welding in that in brazing the molten filler metal is drawn by capillary attraction into the space between closely adjacent surfaces and is not built up as a fillet (BS 499:1952)

since the filler metal consists of brass, composed basically of an alloy containing 60 per cent copper and 40 per cent zinc, which may incorporate quantities of silicon, manganese and/or nickel, and the term 'braze' welding describes the process more accurately. Typical chemical compositions of filler have been specified[13] but if strong welds free from interface brittleness are required, the silicon contents quoted in the standards for these rods are too high and should on no account exceed 0·1 per cent[3,4]. The interface brittleness is caused by diffusion of silicon from the brass filler metal into the iron giving a thin, brittle layer with 5–6·5 per cent silicon and with hardness 380–450 HV both with mild steel and cast iron. Nickel bronze with 9·7–15·6 per cent nickel and less than 0·17 per cent silicon does not show this effect and manganese bronze up to 0·25 per cent silicon shows less tendency.

After suitable preparation and cleaning, any graphite that has become exposed during the cleaning operation must be removed, in order to obtain a good bond between the weld deposit and the parent metal. Heating the surfaces to be welded to a temperature of about 400°C with the oxy-acetylene flame is sufficient for this purpose.

With this method, pre-heating is advisable, but is not so essential as in fusion welding, because of the lower temperatures employed. Complete pre-heating is usually necessary for complicated castings, and the pre-heating temperature should not exceed 400°C. Where total pre-heating is impracticable or small welds are required, local pre-heating with an oxy-acetylene flame should be used.

The method of making the weld has been described in detail in the literature[2,10,11] and a general description of the type of joint to be preferred has been specified[13].

As in the case of oxy-acetylene gas welding, it is necessary to employ a flux when using this method of welding.

Powder welding[5] is another form of non-fusion welding in which a finely divided metal powder is fed through a metering device into the oxygen stream of a modified oxy-acetylene torch. No flux is required. As the powder passes through the flame it is pre-heated and melts just as it reaches the surface of the base metal. Provided the temperature differential between the base metal and the heat source is of the right order, instantaneous wetting of the surface occurs when the stream of pre-heated powder makes contact with the base material which itself is not melted though some interface diffusion occurs. The nickel base powder welding alloys suitable for cast irons melt between 950 and 1000°C and the base cast iron is pre-heated up to 200–350°C and reaches 700–800°C during application of the powder filler but melting does not occur. A pre-sprayed coating over the area to be welded prevents subsequent oxidation while the weld metal is built up at the commencement of the junction itself. As the base metal is not melted, there is no carbide formation as in fusion welding and the rate of cooling can be controlled by the torch to prevent the formation of martensite. Such welds are therefore freely machinable.

Preparation of the surfaces is similar to that for fusion welding[9]. The process can be applied to grey and nodular irons, blackheart malleable and austenite nodular irons. In austenitic irons[6] the process appears to avoid

Table 160

Fusion Welding

Rod	*Composition*	*Welding*	*Suitability*	*Strength of welds expressed as percentage of original strength of casting*		*Max. Brinell*	
				Tensile strength	*Tensile strength*	*Parent metal*	*Filler metal*
Cast iron	BS 1453:1957 Grade B1. Total carbon, 3–3·6%; silicon, 3–3·5%; manganese, 0·5–1%; sulphur, <0·15%; phosphorus, <1·5%. The test figures quoted were taken with 0·6% phosphorus	Oxy-acetylene	Generally suitable for grey iron, but not for malleable	74–99%	66–98%	270–315	220–315
Nickel iron	53% nickel, 45% iron	Metal arc	Suitable for cast iron below 0·3% phosphorus	96–99% for 0·1% phosphorus iron	96–100% for $\frac{3}{8}$ in (9·5 mm) thick, 63–80% for 1 in (25·4 mm) thick	180–225	172–252

Nickel copper	Approx. 67% nickel; 30% copper; 1·25% iron; 1·25% manganese	Metal arc	Suitable for cast iron below 0·3% phosphorus	No test data		Usually machinable	
Pure nickel		Metal arc	Suitable up to 0·6% phosphorus, with silicon above 1·6%	No test data		Weld junctions may be above 300	
Dead mild steel		Metal arc	Suitable for cavity filling where machinability is not important in thin castings. Sometimes used for massive castings with adequate pre-heating and slow cooling. Not suitable for high phosphorus castings	No test data		Weld junctions may be above 300	
			Non-Fusion Welding				
'Braze'	Copper, 60–62%; zinc, 37–38%; tin up to 0·5%; silicon, <0·1%	Oxy-acetylene	Machinable and suitable for grey or malleable cast iron	60–80%	73–84%	244	90–103
Nickel-bronze	Copper, 43–48%; zinc, 39–40%; tin, 0–1%; nickel, 9–16%; silicon, preferably below 0·15%	Oxy-acetylene	Machinable and suitable for grey or malleable cast iron	55–100%	77–100%	277	97–214

parent metal cracking characteristic of fusion welding. The strength of the weld is comparable to that of the parent metal, residual stresses are much lower than for fusion welding and the cost of equipment much lower than for arc welding.

Properties of fusion welds

The main difficulties experienced in the welding of flake graphite cast irons are usually due to:

(1) Hard or unmachinable weld zones.
(2) Cracking, either in the parent metal or in the deposited weld metal.
(3) Lack of adherence between the deposit and the parent metal.
(4) Unsoundness at the weld.

If cast iron is rapidly cooled from temperatures near the melting point, as may happen in fusion welding, the carbon is retained in the combined form and the welded zone becomes hard and unmachinable. Where steel filler rods are employed, hardness in the weld zone may arise from interdiffusion of the parent grey cast iron and the steel filler material, with the formation of martensite.

It follows that if rapid cooling of the iron can be avoided during completion of the welding operation, the risk of obtaining a hard welded zone will be reduced. It must be remembered that when welding an isolated zone of a large casting, heat is conducted away from the welding zone at a fairly rapid rate, owing to the mass of cold metal surrounding the weld. Under these conditions, rapid cooling of the weld zone will occur, and this effect can be reduced by a pre-heating operation which prevents too steep a temperature gradient arising between the welded joint and the parent metal. Various methods of pre-heating are used, the oxy-acetylene blowpipe being frequently employed for small work or local pre-heating. For larger work, or where it is necessary to pre-heat a complete casting, temporary charcoal fire furnaces or gas muffles are used. With a charcoal fire, welding *in situ* is possible.

Pre-heating the parts to be welded also assists in the prevention of cracking by enabling the unequal expansion and contraction of the weld metal and parent metal to be equalised. The majority of iron castings are of closed or rigid construction, and any non-uniform expansion sets up considerable stresses in the colder portions of the casting. In the case of flake graphite cast iron, the lack of ductility of the material will accentuate the tendency to cracking if any non-uniform expansion or contraction takes place, and thus careful pre-heating is of particular importance if successful welded repairs are to be obtained.

A large proportion of the grey iron castings produced in Great Britain contain more than 0·6 per cent phosphorus. This element forms hard and brittle eutectic in cast iron, with a melting point some 200°C below that of the remainder of the iron.

When welding a high phosphorus cast iron, cracking in the weld area is likely to occur where conditions of thermal shock are severe, as in metal arc

welding without pre-heating. Contamination of the weld metal by phosphorus may also lead to brittle welds, particularly with nickel/copper non-ferrous electrodes. The use of arc welding methods is generally not recommended[8] for iron castings containing over 0·5 per cent phosphorus. Reference has been made[8] to a relationship between the phosphorus and silicon contents of grey iron and it is suggested that with a higher silicon content in the cast iron a higher phosphorus content can be tolerated, and good arc welds can be made when the silicon content exceeds 1·6 per cent, using electrodes made with pure nickel core wire.

Oxidation of the metal during welding, and the welding of improperly prepared surfaces, may give rise to lack of adherence between the parent cast iron and the deposited metal, and may also cause an unsound weld. For this reason, welding should be carefully carried out to ensure that oxidation products and any dross are removed by the flux and transferred into the slag which forms on the surface of the molten weld puddle.

The functions of the flux when welding cast iron are to keep the metal surface clean, protect the metal from oxidation during welding, and to dissolve oxides and other impurities. The flux may also be used to prevent loss from the molten metal by volatilisation, e.g. zinc in braze welding, and also to improve the fluidity of the weld metal.

During the welding operation, stresses will be set up in many castings owing to their closed or rigid construction (see p. 393). Although pre-heating will reduce these stresses, it is desirable that castings should be slowly cooled in order to reduce further internal stresses. In some cases it may be preferable to employ a stress-relieving heat treatment, depending on factors such as the extent of the welding operation, the complexity of the casting and the conditions of service required.

Table 160 summarises the main types of welding and filler rods and their uses, together with the likely strength and hardness of the resulting welds. Other grades of welding rod are available, both as British Standard and as proprietary materials.

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SOLDERING OF CAST IRON

Soldering of cast iron can be successfully carried out provided reasonable precautions are taken in the preparation of the surface.

Preparation of surface

The oxide skin must be removed by grinding, turning or shotblasting. Sandblasting is much less effective and contaminates the surface with minute particles of sand which cause the solder to 'roll off'. After shotblasting, the components need degreasing, either in trichloroethylene vapour or in boiling alkali cleanser.* Acid pickling is ineffective and leaves smears of graphite on the surface which resist tinning. Grinding, machining or filing also produce satisfactory surfaces.

After cleaning, the surfaces may be readily tinned using the normal types of soldering spirit [e.g. 6 oz commercial zinc chloride in 1 pint (299 g/litre) of water + 10 per cent ammonium chloride].

Resin fluxes are usually unsuitable and cored solders can only be used successfully on already tinned surfaces. Paste fluxes usually require to be rubbed vigorously on to the surface.

Where hot dip tinning is used, immersion in a bath of molten flux salts at 270°C is effective, provided that the articles are held in the bath until they have heated up sufficiently for chilled frozen flux to be remelted. From the flux bath the articles can be immersed in a solder pot passing through a layer of fused flux on its surface. The flux recommended is:

	%
Zinc chloride	75
Sodium chloride	15
Ammonium chloride	10

Tinning can be carried out with a heavy copper bit, using a large flat side rather than the point.

Strength of soldered joints

Some tests were carried out by Jolly at BCIRA, using a solder containing lead, 58·8 per cent; tin, 40·9 per cent; and iron, 0·16 per cent.

The material used was an ordinary low phosphorus grey iron and the experiments were made on the fractured ends of a 0·564 in (14·3 mm) diameter test bar so that the joint could be tested in tensile.

* Shotblasting with grit not coarser than 70 grade, and pressures of 45–50 lb/in^2 (310–345 kPa) may give surfaces that will tin without subsequent cleaning

In the first test, the two halves were prepared by grinding flat, rubbing with emery and then degreasing. Soldering was then attempted using a resin flux. This failed as the surface would not 'tin'.

In the second test, the two surfaces were prepared as before, and soldering was attempted using a proprietary soldering fluid of the zinc ammonium chloride type. The surfaces tinned satisfactorily and were then placed against each other; the joint was heated, and when the solder had melted, the two halves were pressed together. The solder was then allowed to freeze.

Breaking load in tensile: 1·08 ton (10·8 kN); Max. stress: 4·3 ton/in^2 (66 N/mm^2).

Finally, another test piece was prepared by facing the two surfaces on a lathe and then soldering them with the use of the soldering fluid used in the second test.

Breaking load in tensile: 1·00 ton (10·0 kN); Max. stress: 4·0 ton/in^2 (62 N/mm^2).

These results show that very satisfactory strengths can be obtained from properly prepared soldered joints.

Plumbers solder (66 per cent lead: 33 per cent tin) appears to be less satisfactory, owing to the long pasty range, than a solder containing 40 per cent tin[2].

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6
Internal casting stresses

It is almost impossible, with conventional methods of founding, to produce castings in any metal that are completely free from cooling stresses. These stresses are usually too low to be of great significance, but they can reach levels comparable with those imposed in service, and in extreme instances may cause local fracture. There is a considerable but scattered literature upon the subject, but the development of present understanding of the subject is contained in the papers quoted at the end of this section[1–7].

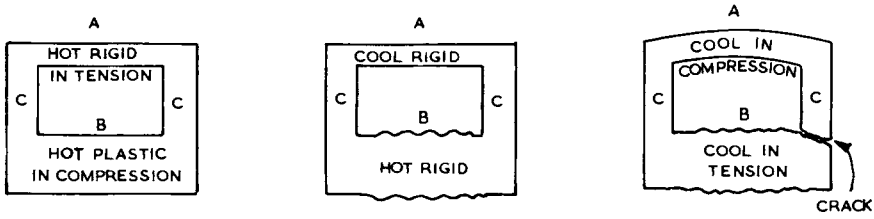
All metals contract on cooling; if a simple bar be allowed to cool and contract freely the amount of contraction for a given temperature drop is known quite accurately. In practice, however, even a simple bar does not cool uniformly; the ends cool more rapidly than the centre, and in a flat plate or disc the edges cool more rapidly than the centre. With differential cooling, therefore, differential contraction will take place, with the inevitable development of stress. This stress, however, should disappear again when cooling is complete and no temperature differential remains, unless some plastic deformation has taken place during the cooling period as a result of the differential contraction stress.

Deformation, however, can occur when the thinner or cooler portion of the castings has become rigid enough to support and impose stress while the thicker or hotter portion is still plastic. If at this stage the cooling and contraction rate of the thinner or cooler section remains more rapid than in the thicker or hotter portion, the hotter portion may yield plastically and relieve this initially imposed stress, and when in turn the hotter portion becomes cool enough to be rigid and support stress, since it is still hotter than the cooler portion, it will contract more in the final stages of cooling and impose compressive stresses on the cooler portion and tensile stresses on itself. These stresses will be permanent.

Thus casting stresses are mainly caused by temperature differences within a casting during cooling after solidification and in cast iron do not begin to develop permanently until the hottest member has cooled to a temperature of the order of 700–750°C at which temperature it is capable of sustaining appreciable stress. The initial cooling stresses that develop above this temperature range are opposite in direction to those finally left in the cold casting, as shown in *Figure 121* Stages 1–3.

Hot tearing usually occurs above 900°C and is distinguished by a heavily oxidised and frequently dendritic fracture. It may occur as a result of (a) these initial stresses, (b) mould restraint or (c) the collapse of the first formed skin of metal over a 'shrinkage' cavity.

Above about 700–750°C the capacity of grey iron to withstand appreciable stresses is so low that unless the cooling rate is rapid, deformation will prevent the development of appreciable stresses*. The final stresses that remain on



STAGE ① STAGE ② STAGE ③

① "A" HAS REACHED A LOW ENOUGH TEMPERATURE TO BE RIGID AND IS COOLING MORE RAPIDLY THAN "B" AND COMPRESSING IT PLASTICALLY

② BOTH "A" AND "B" ARE AT A TEMPERATURE LOW ENOUGH TO BE RIGID BUT "B" IS AT A HIGHER TEMPERATURE THAN "A". THE SYSTEM IS SUBSTANTIALLY STRESS FREE BECAUSE OF THE DEFORMATION THAT HAS OCCURRED IN "B"

③ "A" AND "B" ARE BOTH COOL BUT NOW "A" IS IN COMPRESSION AND "B" IN TENSION AS A RESULT OF THE CONTRACTION OF "B". CRACKING IS LIKELY TO OCCUR AT THE CORNERS OF THE CONNECTING MEMBERS "C"

Figure 121

cooling, however, are developed as a direct result of the inability of the hotter portions of the casting to withstand the crushing forces imposed on them by the contraction of the thinner or cooler sections while the casting is still hot.

The development of stress during cooling is summarised in *Figure 121*, which illustrates diagrammatically the development of stress in a simple grid with limbs of different thickness while cooling in the mould.

*Although stresses above 850°C are low because the material is weak, hot tearing may occur, particularly above about 920°C, when some liquid phase phosphide eutectic may be present. The stresses causing hot tearing, however, are usually opposite in sign to the residual stresses remaining when the casting is cold

Thus, the plastic deformation that occurs in the thicker or hotter member in the earlier stages of cooling, serves only to provide a system that is substantially stress-free at the time when the thicker or hotter member has become cool and rigid enough to support stress but is still hotter than the thinner member. The simple contraction of the thicker or hotter, but now rigid, member then imposes tensile stress upon itself and compressive stress on the thin or cooler section.

The dependence of residual casting stress upon section differences or temperature differences has been very clearly demonstrated[6], using the Ferranti strain grid shown in *Figure 122*. A series of these grids was made in which

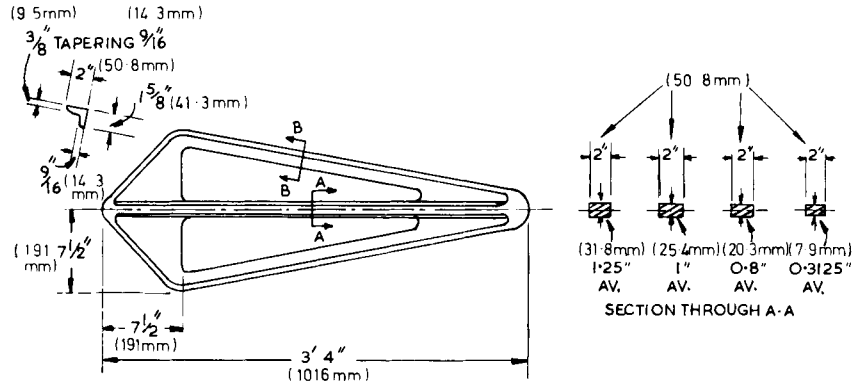


Figure 122. Ferranti strain grid used by IBF Committee TS 32

the centre member only was varied in thickness above and below that of the mean sections of the side members, which were about $\frac{1}{2}$ in (13 mm) thick. Tests with strain gauges showed that the centre member carried residual tensile stresses when thicker than the side members, and compressive stress when thinner than the side members (*see Table 161*).

Table 161

Average thickness of centre member of grid		*Average residual stress in centre member of grid*		
in	(mm)	ton/in²		(N/mm²)
1·25	(31·8)	3·0	tension	(46)
1·00	(25·4)	2·9	tension	(45)
0·80	(20·3)	1·4	tension	(22)
0·3125	(7·9)	−2·6	compression	(−40)

Differential cooling rates may arise in castings of uniform thickness. For example, the free edge of a flat plate cools much more rapidly than the centre, and it would be anticipated, therefore, that the centre of the plate would be in tension and the edge in compression. This is frequently the case, and will cause buckling of the plate or cracks in the centre not extending to the edge.

Stresses developed similarly, cause the well-known tendency of wheel spokes to crack in tension. Similar stresses can also be developed by welding (see p. 389).

METHODS OF RUNNING

In castings of uniform thickness, methods of running may cause temperature differentials to arise. If a grid of uniform thickness is gated through one member so that the whole of the metal flows through it, the heat transmitted to the sand by the continuous flow of metal may sufficiently reduce the cooling rate of this member to introduce substantial stresses on cooling. The runner bar itself may become the hot member and initiate stress.

Alternatively, in castings containing different section thicknesses, running through the thin member may effectively reduce the difference in cooling rate between it and the thicker members, and thereby reduce the final level of stress.

Time and temperature of knock-out from the mould may also be important. If a casting with thin and thick sections be knocked out of the mould when the thick section is about 800°C, then the rate of cooling of the thicker section will be accelerated much more than that of the thinner or cooler section, and hence will reduce the time available in which plastic deformation can occur. Under such conditions, hot knock-out will reduce the final level of stress, but will only be effective if carried out when the hotter members are sufficiently hot to deform plastically, i.e. above about 750°C, otherwise slow cooling in the mould may be preferable.

The manner in which these stresses develop and may cause cracks in actual castings, has been described[7].

The distortion of castings as a result of casting stresses is dealt with in the section on camber in casting.

Annealing of castings above 850°C does not necessarily produce a stress-free casting unless cooling is uniform and slow. Zingg[8] has shown that SG iron castings may carry substantial residual stresses after heat treatment at 850°C followed by air cooling. There is little doubt that the mechanism of stress development outlined in *Figure 121* is instrumental here also. He showed that for SG or nodular iron initial stresses of 13–19 ton/in^2 (201–293 N/mm^2) were reduced to below 3·8 ton/in^2 (59 N/mm^2) after 10 h at 550°C followed by slow cooling.

It is evident that if stress relief is required, any heat treatment above about 700–750°C should be followed by slow cooling. Stress relief by heat treatment is dealt with in the section on heat treatment.

Cracking by residual casting stress

The tendency of a casting with residual casting stresses to crack spontaneously increases as the grade or tensile strength of the iron increases, i.e. as the carbon equivalent value decreases.

With given casting strain, decreasing the carbon equivalent value increases

the residual stress, since the modulus of elasticity increases and the strain at failure (elastic and plastic) decreases, as shown in *Table 162*, and the risk of rupture increases.

Table 162

	Grade or tensile strength		*Total strain at failure, % (elastic + plastic) in tensile test*					
			Phosphorus <0·4%			*Phosphorus 0·4–1·0%*		
	ton/in²	(N/mm²)	*Elastic*	*Plastic*	*Total*	*Elastic*	*Plastic*	*Total*
	10	(154)	0·15	0·6	0·75	0·15	0·45	0·6
	12	(185)	0·17	0·53	0·70	0·17	0·33	0·5
Normal flake graphite	14	(216)	0·18	0·47	0·65	0·18	0·22	0·4
	17	(263)	0·20	0·38	0·58	—	—	—
	20	(309)	0·22	0·28	0·50	—	—	—
	23	(355)	0·25	0·25	0·50	—	—	—
Undercooled graphite	13·5	(208)	—	—	0·33	—	—	—
White iron	30–40	(463–618)	0·25	0·15–0·36	0·4–0·63	—	—	—

Thus a residual casting strain of 0·6 per cent would be expected to cause rupture in grades 17, 20 and 23 irons, but would not cause rupture in low phosphorus grades 10, 12 or 14 irons.

For this reason, thin plate castings are usually made in soft iron with a carbon equivalent value above 4·5, and the risk of cracking is reduced as the phosphorus content is decreased.

Relief of stress by static loading

Bühler and Schepp[9] measured the effect of static loading on residual casting stresses in grey and nodular iron bars with residual tensile stresses in the outerface and compressive stresses in the interior. Their results showed that stressing in tension and compression to 0·2 per cent proof stress reduced the level of residual stress. Similar static stresses applied to initially stress-free bars produced no residual stress.

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CAMBER IN CASTINGS

Owing to differences in cooling rate and mould resistance, it is extremely difficult to make castings completely free from internal strain, and as a result, most castings show some tendency to warp during cooling or after they have been removed from the mould. It is not easy to predict this tendency quantitatively.

The simplest form of distortion is caused as a result of differential cooling rates, as described in the section on casting stresses. If, for example, a casting is made with proportions similar to those shown in *Figure 121*, the thin

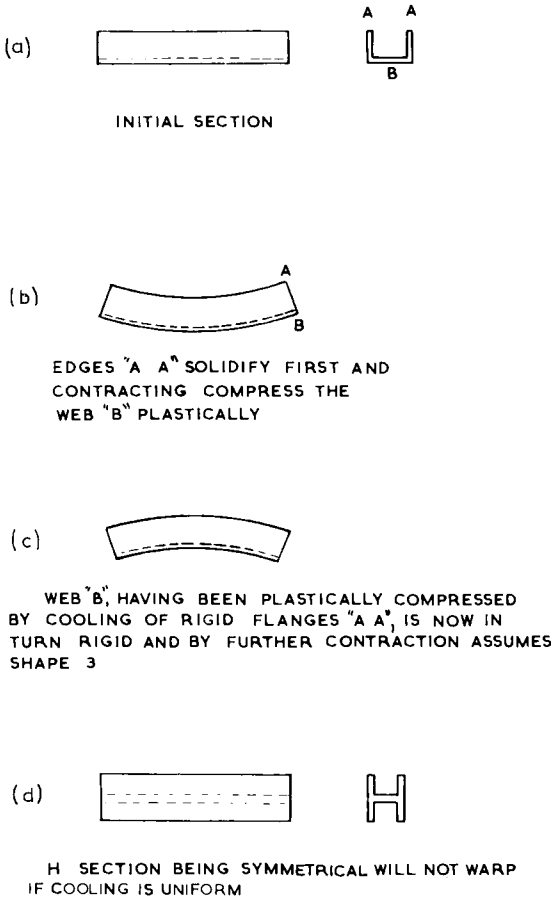
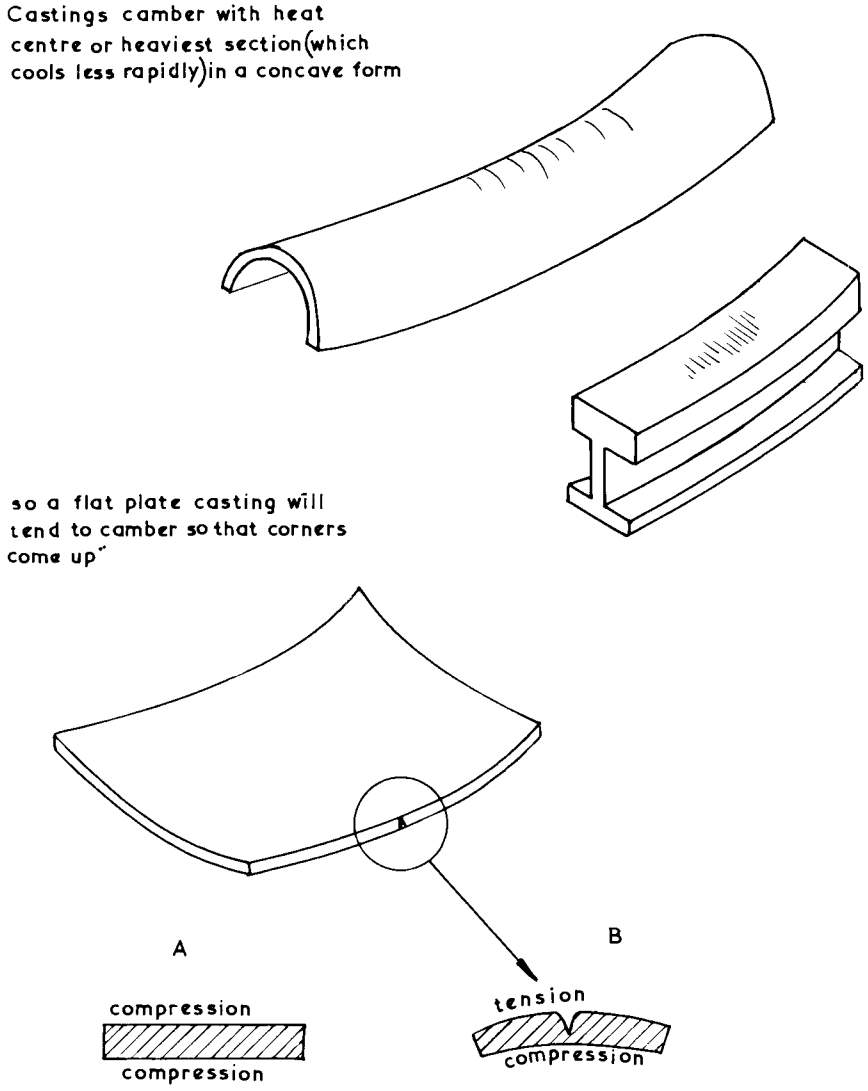


Figure 123

section will solidify first, and then contracting, will compress the still plastic heavier sections, causing them to yield. When these in turn become rigid, the thinner sections will be much cooler, and during solid contraction of the rigid but hotter thick members they will impose compressive stresses on the



Internal stresses may cause a flat plate casting to buckle and reverse the initial direction of stress at the free edge (A) to that shown in (B)

Figure 124

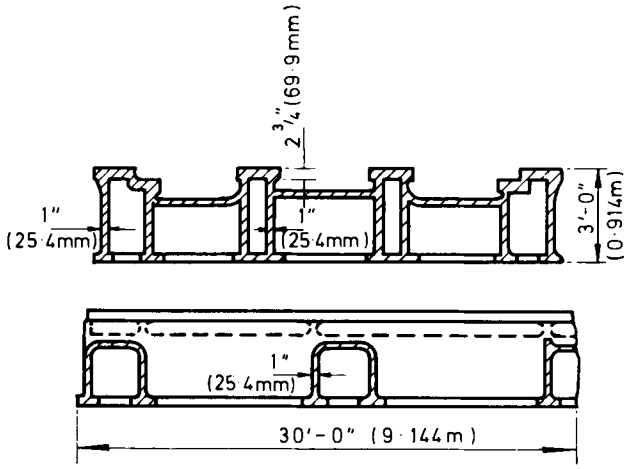


Figure 125

thin member and cause it to buckle in the manner indicated. This simple form of differential cooling stress can frequently be predicted from an examination of the drawing or pattern, and appropriate action can be taken before the casting is made[1]. The same effect is apparent in U-sections, where

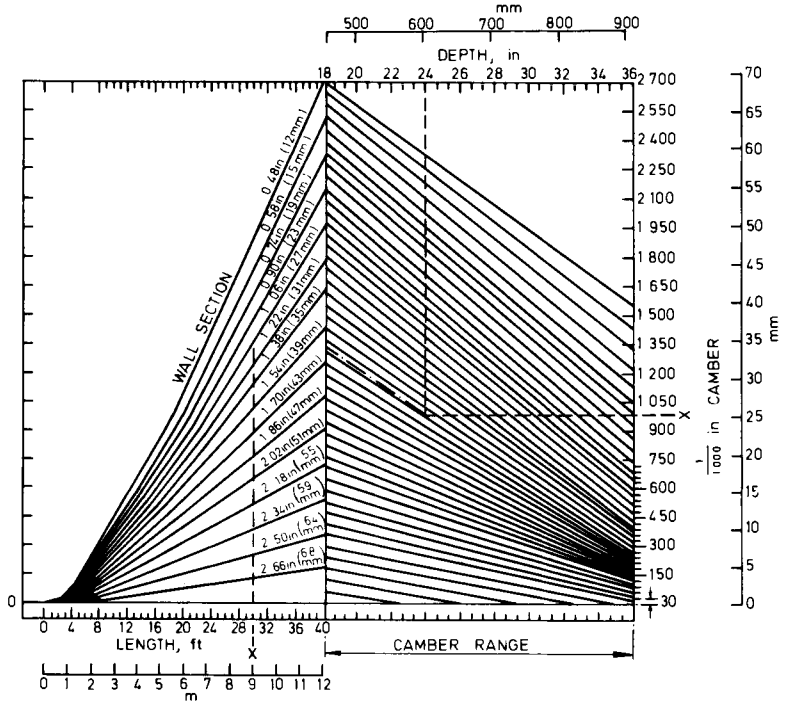


Figure 126

the edges cool more rapidly than the centre, as shown in *Figure 123*. In general, castings camber with the heat centre or heaviest section in a concave form (*Figure 124*).

This relatively simple distribution of stresses, however, does not always apply, particularly in large flat plates, where buckling may markedly redistribute the stresses originally present, and may cause actual reversal on one face of the casting, as shown in *Figure 124*. In very large castings, up to 30 ft (9·1 m) in length and 3 ft (0·91 m) in depth, a detailed study of camber correction has been made by Longden[2,3]. This is summarised in *Figures 125* and *126*. *Figure 125* shows the type of casting to which the diagram applies.

It will be noted that the nomograph in *Figure 126* is arranged in two divisions. The left-hand division covers the length and vertical sectional thicknesses of the casting and the right-hand division the depths and camber scale. The nomograph is based on a main vertical longitudinal wall section of 1·0 to 1·1 in (25 to 28 mm), the lower faces (as poured) carrying metal sections equivalent to between 250 and 300 per cent of the vertical sections and covering from 40 to 50 per cent of the area of the bottom face of the casting, as poured. Any reduction in the vertical wall section calls for an increase in camber allowances. A thickening of the wall sections requires a corresponding reduction in camber allowances.

How to use the nomograph

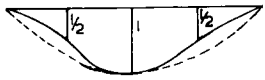
Example: Find the camber required for a bed 30 ft (9·1 m) in length, 24 in (610 mm) in depth, and with 1·06 in (27 mm) vertical wall sections.
From X, 30 ft (9·1 m), strike a vertical line until it joins the wall section of 1·06 in (27 mm). From this point draw a line at right angles until it meets, or is guided by, the first diagonal camber line and trace it down until it joins the vertical line descending from the 24 in (610 mm) depth point.
Answer: Camber needed 0·99 in (25 mm).

Generalisations on camber[2,3]

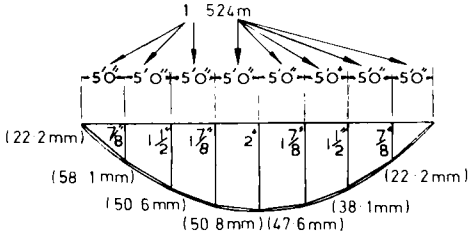
The following additional rules may be used in conjunction with the camber nomograph (*Figure 126*).

(1) Camber allowance increases with an increase in the length of a casting, other dimensions remaining constant.
(2) Camber allowance increases with a reduction in metal section of the vertical walls of the casting.
(3) Camber allowances are reduced with an increase in the depth of the casting, without any corresponding increase in length.
(4) Camber allowance is slightly decreased with a reduction in width.
(5) Camber allowances increase with an increase in the plated area on the lower face of the casting as poured.

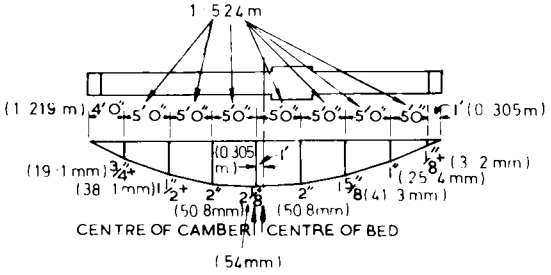
(6) Camber allowance increases with an increase in the section thickness of the metal on the bottom face of the casting.
(7) Camber allowance decreases with an increase in the plated area on the top side of the casting as poured.
(8) Camber allowances are modified slightly with pouring temperature, allowances increasing slightly with lower temperatures.



(a) MIS-SHAPEN CAMBER RESULTING FROM INCORRECT DISTRIBUTION OF CAMBER ALLOWANCES



(b) NEGLIGIBLE ERROR CAUSED BY CORRECT DISTRIBUTION OF CAMBER ALLOWANCES



(c) DISTRIBUTION OF CAMBER ALLOWANCES IN UNSYMMETRICAL CASTING

Figure 127

Nicholls and Kershaw[4] point out that while the camber correction given by the curves in *Figure 126* gives the maximum camber at one point, no indication is given as to the distribution of camber over the remainder of the casting, and if half of the full camber is used at points midway between the centre of the camber and the ends of the bed, the final casting may still be mis-shapen, as shown in *Figure 127(a)*. The distribution of the camber along

the length of the casting gives satisfactory results when it is distributed along an arc, as shown in *Figure 127(b)*. No difficulty occurs if this is made up of a series of straight strickles, which in a casting of 20–40 ft (6–12 m) long may be 5 ft (1·5 m) in length. Some variation in camber will naturally be required for castings which are unsymmetrical, e.g. *Figure 127(c)*. Camber corrections, however, are not always completely effective and may require to be supplemented by varying amounts of machining. The designer should therefore allow a reasonable clearance between unmachined surfaces and the machined faces to accommodate subsequently added moving parts located on the machined faces.

Elimination of warping by applying camber to the pattern does not necessarily reduce residual stress; it may increase it.

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THERMAL STRESS CRACKING

It is well known that surface cracking can occur in most metals if subjected to cyclic high temperatures on one face while the other face remains cool. In particular, this form of cracking occurs in diesel engine heads which are subjected to firing temperatures on one face while the other is water cooled, and in brakedrums where the rapid dissipation of energy on one face causes high local temperature while the remainder of the drum remains relatively cool.

The mechanism of failure is now fairly well understood and may be described as follows[1].

If one considers a flat square plate (*Figure 128(a)*) at a uniform initial temperature of zero, subjected to a linear thermal gradient such that one face now has a temperature t_r above zero, while the other face remains at zero temperature, and supposing the plate to be freely suspended, it will assume a spherical shape and will be free from stress (*Figure 128(b)*). The mid-plane of the plate will have expanded and the hot face will have expanded by twice as much, namely $L\alpha t_r$, where L = length of the side of the plate, α = coefficient of thermal expansion and t_r = temperature difference between the two faces.

Now thermal stresses arise through the suppression of expansion and the magnitude of the stress depends on the manner of suppression. Consider first the suppression of curvature only, as if by squeezing the plate between flat plattens (*Figure 128(c)*). The mid-plane will be a neutral axis and the faces will be subjected to stresses of equal magnitude but opposite in sense. The stretch on the cool face depends on the stress in the hot face and vice versa.

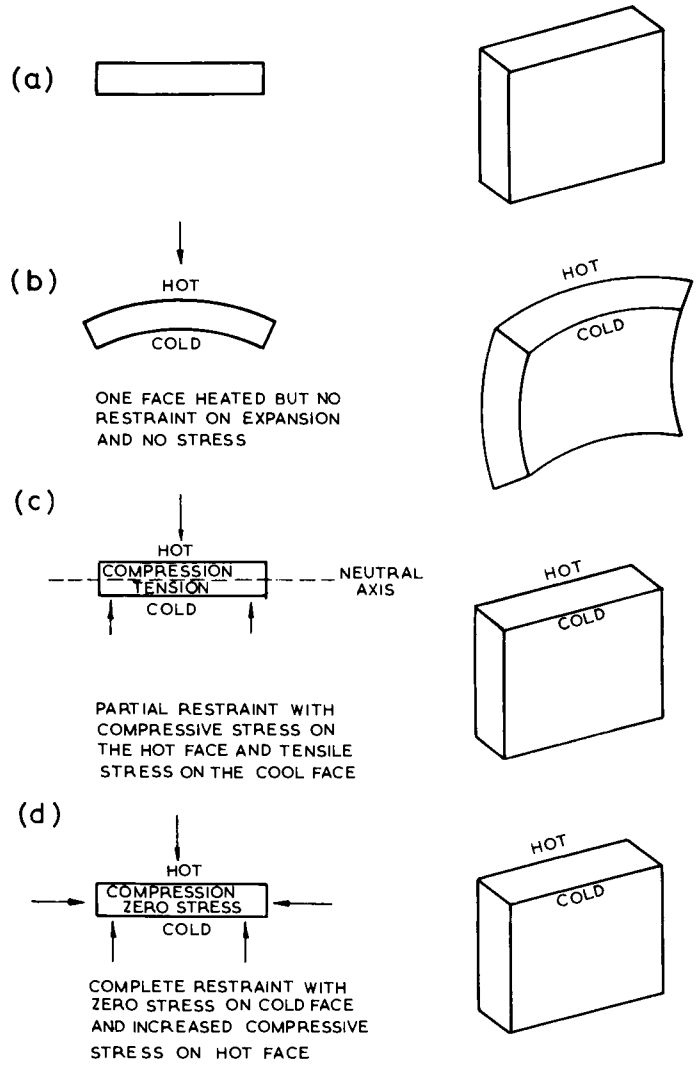


Figure 128. Development of thermal stress

The equation:

$$\text{Stress} = \frac{E\alpha t_r}{2(1-\mu)}$$

will give the stress values in either face, where E = elastic modulus and μ = Poisson's ratio. The value of stress given by this equation will be in tension on the cool face and in compression on the hot face.

This condition applies, for example, to a cylinder uniformly heated on one surface with uniform heat removal from the other surface or to a flat plate subjected to a similar temperature gradient in which curvature is restrained.

To obtain complete suppression of linear expansion as well as suppression of bending, one would now have to apply to the plate compressive edge forces of such a magnitude as to restore the plate to its original dimensions (*Figure 128(d)*). These forces would therefore have values which, if applied alone, would produce compressive stresses of $E\alpha t_r/2(1-\mu)$ throughout the plate. The resulting stress at the cool face is then zero and that at the hot face is twice the above, namely:

$$\text{Stress} = \frac{E\alpha t_r}{1-\mu}$$

This formula applies to a hot spot of great mechanical rigidity in which complete suppression of linear expansion is assumed.

These conditions can apply approximately in a diesel head on the flame face adjacent to the exhaust port or in a brakedrum where a hot spot may develop as distinct from uniform development of heat by friction over the whole of the swept surface. In the latter case the initial development of a hot spot for any reason, e.g. local sub-surface porosity, will be self intensifying since radial expansion at the hot spot will raise its level above that of the surrounding metal and this region will then be subjected to greater local pressure from the brake shoes, and the energy fed in at this point will be increased above that of the surrounding area.

If the compressive stresses so produced are within the elastic limit of the material they will be completely eliminated on cooling and no damage may occur. If, however, they are high enough *and of sufficient duration* to cause local creep of the hot surface, this surface on cooling will be permanently shortened and tensile stresses will develop as the hot face cools down with corresponding compressive stresses on the cool face. If the residual tensile stresses are high enough, surface cracking will occur on cooling down.

The maximum residual tensile stress on cooling must be at the original hot face but its value cannot exceed the numerical value of the initial compressive stress at the point because[1]

(1) All the stress (and therefore the elastic strain) cannot be relieved by creep within a finite running time.
(2) On cooling, all the plastic (creep) strain cannot be converted into elastic strain because of the elasticity of the remainder of the part.

The residual stress therefore cannot exceed numerically the initial compressive stress and fracture cannot take place unless the initial compressive stress is numerically greater than the fracture strength in tension.

For given conditions of heat flow or energy dissipation the tendency to develop residual stresses high enough to cause surface failure depends on:

(a) Thermal conductivity of the material which governs the maximum temperature attained at the hot face.
(b) Thermal expansion which governs the magnitude of the stress developed for a given temperature difference.
(c) Creep strength at the temperature developed, which governs the amount of plastic deformation occurring at the maximum temperature

and compressive stress developed at the hot surface *in the time available* and thereby the magnitude of the residual tensile stress.

(d) Modulus of elasticity and Poisson's ratio. The lower these are, other conditions being similar, the lower the stresses developed.

(e) Ductility or the ability to absorb small amounts of repeated plastic strain.

All of these factors, however, are interdependent; thus substituting a material of high thermal conductivity will lower the maximum temperature and other things being similar, the developed stresses will also be lower. If, however, the new material has a lower creep strength at the lower temperature than the original material had at the higher temperature, failure can still occur unless the ratio of creep strength to developed stress is increased. Similar consideration will apply to variations of the other properties, for example some non-ferrous alloys with high normal ductility may have their ductility at room temperature reduced to one-tenth to one-twentieth by prolonged exposure to high creep stresses.

The reduction of temperature by the substitution of non-ferrous alloys of high thermal conductivity may also be nullified by the increase in stress caused by their high thermal expansion.

The choice of materials for such circumstances must always therefore be a compromise and there will be optimum conditions of temperature and temperature difference for each material beyond which failure will be initiated.

For any given material any design modification that reduces the maximum temperature and the temperature difference between the two faces will increase its resistance to hot face cracking.

Cracking in diesel engine heads occurs as a result of hot face temperature rising above 350–400°C with the other face nominally water cooled and the mechanism of stress development just described is involved. The following figures give the stress to rupture of unalloyed grey iron [tensile 16·6 ton/in^2 (256 N/mm^2)] in 100 000 h at various temperatures (*see Figure 99*).

Temperature, °C	500	450	400	350	300
Stress to rupture,					
ton/in^2	2·2	4	6	10	13
(N/mm^2)	(34)	(62)	(93)	(154)	(201)

These figures indicate clearly the increased resistance to creep which occurs as the temperature is reduced. These figures are based on long-term tensile testing but the same pattern occurs in compression[4].

Heat transfer from the cooled face occurs by water flow which may be streamline or turbulent but in either case, adjacent to the surface the speed is greatly reduced and a very slow speed film of liquid is present which limits the heat transfer to the mass of the flowing liquid. The heat transfer coefficient varies as $V^{0·8}$ (V = velocity of water) for turbulent flow but falls to about $V^{0·3}$ for streamline flow[2]. The Nukijama diagram[3] (*Figure 129*) shows that as the metal surface temperature is increased above boiling point, heat transfer occurs by natural convection up to about 105–135°C aided by tiny steam bubbles (nucleate boiling). Above these temperatures transition boiling

occurs in which the bubbles increase markedly in size and the rate of heat transfer is reduced to one-fifth or one-tenth of the maximum until at surface temperatures of about 250°C vapour film boiling is established, the steam barrier may prevent contact between the wall and the water, and the metal wall temperature can rise catastrophically.

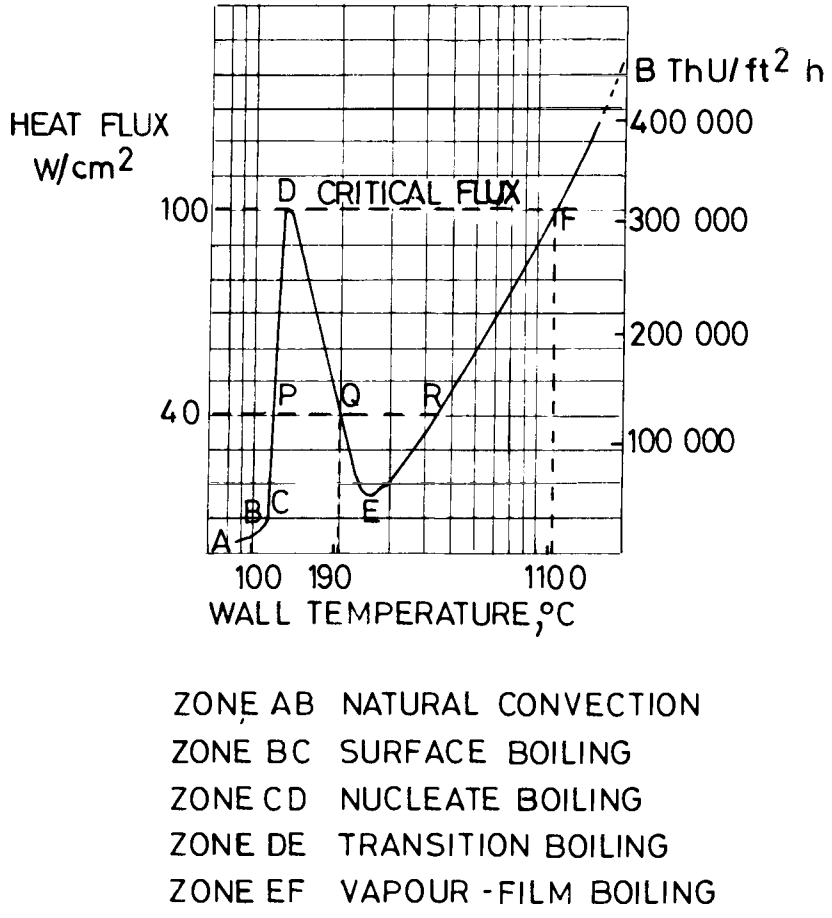


Figure 129. Nukijama curve (after Renè Morin, Euratom *Dec. 1966, V, No. 4)*

If the water flow is directed as a jet normal to, or at an angle to the metal wall instead of flowing parallel to it, the mass flow at a given point will be greatly increased, the transfer rate increased, the liability to film boiling reduced and the hot face temperature for a given water flow may be reduced by 15–30°C or more giving a marked increase in creep stress with reduced liability to surface cracking.

The significance of the special stress/strain response of cast iron to stress has been fully discussed by Dearden[4]. Northcott and Baron[5] also studied the effect in steel.

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CONTRACTION ALLOWANCES

Typical figures for grey iron range from $\frac{1}{8}$ to $\frac{1}{12}$ in/ft (1:96 to 1:144). For cast irons with strengths of 18 ton/in^2 (278 N/mm^2) or over, a contraction allowance of $\frac{5}{32}$ in to the foot (1:77) is often recommended. Figures given to the Institute of British Foundrymen[1] give 0·140 in/ft (1:86) for high carbon hematite cast iron and 0·158 in/ft (1:76) for high quality engineering iron.

It is well known that the contraction allowance that has to be applied in the manufacture of patterns varies appreciably from part to part of the casting, and indeed to some extent from foundry to foundry, for similar types of casting. In general, the longer the casting the smaller the contraction allowance.

As an indication of the way in which contraction can vary in a single casting, a cylinder block manufacturer recommends $\frac{1}{10}$ in/ft (1:120) for length, $\frac{1}{6}$ in/ft (1:72) for the width, and nothing at all for height. These figures will naturally depend upon whether the casting is made upon its side or vertically. American figures[2] for grey cast iron are given in *Table 163*.

Table 163

Pattern dimension		*Construction*	*Contraction per foot*	*Contraction ratio*
Up to 24 in	(610 mm)	Open construction	$\frac{1}{8}$ in (0·125 in)	1:96
25–48 in	(635–1219 mm)	Open construction	$\frac{1}{10}$ in (0·10 in)	1:120
Over 48 in	(1219 mm)	Open construction	$\frac{1}{12}$ in (0·084 in)	1:144
Up to 24 in	(610 mm)	Cored construction	$\frac{1}{8}$ in (0·125 in)	1:96
25–36 in	(635–914 mm)	Cored construction	$\frac{1}{10}$ in (0·10 in)	1:120
Over 36 in	(914 mm)	Cored construction	$\frac{1}{12}$ in (0·084 in)	1:144

Annealing and vitreous enamelling of grey iron

Grey cast iron is often annealed (*see* section on heat treatment) to improve machinability or is vitreous enamelled to improve corrosion resistance and appearance. The heat treatment involved in both these processes causes a permanent expansion because of the breakdown of pearlite. A construction therefore including enamelled (or annealed) and unannealed castings will call for different contraction allowances in the castings requiring heat treatment.

Expansion of a casting after annealing or vitreous enamelling may vary from $\frac{1}{64}$ in (1:768) to $\frac{1}{16}$ in/ft (1:192).

High silicon irons

For Silal containing up to 7 per cent of silicon, the contraction allowance is $\frac{1}{10}$ in/ft (1:120) for less than $\frac{1}{2}$ in (13 mm) thick, and $\frac{1}{8}$ in/ft (1:96) above $\frac{1}{2}$ in (13 mm) thick. For 14 per cent to 16 per cent silicon iron, $\frac{3}{16}$ to $\frac{19}{64}$ in/ft (1:64 to 1:40) contraction is normally allowed.

Austenitic cast irons

For austenitic irons, such as Ni-Resist and Nicrosilal, the allowance usually made is $\frac{5}{32}$ to $\frac{3}{16}$ in/ft (1:77 to 1:64).

White irons

The normal contraction allowance for white iron is $\frac{1}{4}$ in/ft (1:48)

Malleable irons

The process of heat treating white iron to produce a malleable cast iron will cause a change in the dimensions of the casting: such changes are varied and depend upon many factors. Carbide breakdown during annealing causes expansion to occur, owing to graphite deposition, and decarburisation causes contraction to take place. It is customary for the patternmaker to use a standard contraction of $\frac{1}{8}$ in (1:96) or $\frac{3}{16}$ in/ft (1:64), and any digression from this standard is adjusted by experience or by experiment.

Figures have been given[1] for the contraction of malleable cast iron:

Blackheart malleable cast iron	
Average contraction of unannealed iron	0·230 in/ft (1:52)
Total average contraction of annealed iron (unannealed contraction in mould plus annealed expansion owing to graphite deposition)	0·130 in/ft (1:92)
Whiteheart malleable cast iron	
Average contraction of unannealed iron	0·255 in/ft (1:47)
Total average contraction of annealed iron (unannealed contraction plus annealed contraction)	0·278 in/ft (1:43)

In the case of whiteheart malleable cast iron, the dimensions of the annealed casting depend upon the relative extents to which carbide breakdown and decarburisation occur during the annealing. Carbide breakdown causes an increase, and decarburisation causes a decrease, in the dimensions of the cast-

ing. The extent to which one or the other process controls the final size depends on a number of factors, but in general, castings of thick section (say, greater than $\frac{1}{4}$ in, i.e. 6 mm) will show an increase in dimensions after annealing, and thin section castings (say, less than $\frac{1}{4}$ in, i.e. 6 mm) will show a contraction after annealing. Thus, it is frequently difficult to predict the change in dimensions in whiteheart malleable iron castings following annealing, and the experience of the foundryman is the best guide. The figure given above for the average contraction of annealed whiteheart material will be applicable primarily to thin section castings.

Nodular or spheroidal graphite irons (BS 2789:1961)

These materials may be used in the (a) as-cast, (b) normalised, or (c) fully annealed state, and varying contraction allowances must be made to suit the thickness of casting and the type of heat treatment. The following figures may be employed as a guide:

Nodular (SG) iron		Contraction
Types SNG 24/17 and SNG 27/12 (annealed)	Castings over $1\frac{1}{4}$ in (32 mm) thick	$\frac{1}{10}$th/ft (1:120)
	Castings between $\frac{5}{8}$ in and $1\frac{1}{4}$ in (16–32 mm) thick	$\frac{1}{16}$th/ft (1:192)
	Castings less than $\frac{5}{8}$ in (16 mm) thick	Nil
Types SNG 32/7 and SNG 37/2	If this is normalised the above figures apply, but if the material is used as-cast the normal figure is $\frac{1}{10}$th/ft (1:120). This, however, may vary very considerably according to the composition of the iron used in relation to the section thickness of the casting. An adjustment after trial castings are made is usually necessary.	

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DIMENSIONAL TOLERANCES

While dimensional tolerances are closely linked with contraction allowances, the former are the prime concern of the purchaser of the casting, whereas contraction allowances are the concern of the supplier in his endeavour to meet the requirements of the purchaser.

It is not possible to give precise figures for the accuracy with which castings can be made without information on the type of patterns available, the number of castings required, and the methods of moulding to be used. It is axiomatic that castings cannot be as accurate as the patterns from which they have been made. A high degree of accuracy in the finished castings, therefore, demands the highest possible accuracy in the manufacture of the pattern, in the alignment of the machines and the maintenance of moulding equipment in general.

As mentioned in the section on contraction allowances, the amount to be allowed for contraction within the mould itself varies somewhat with the direction of measurement, with the weight of the casting and type of material employed. These variations are due partly to yielding of the mould cavity itself, partly to the plastic yielding of the casting under its own weight during cooling within the mould, and partly to the restriction by the mould material on the shrinkage of the casting. For these reasons, the amounts quoted for dimensional tolerances can be approximate only. Figures taken from various sources seem to indicate that the following allowances are reasonable:

Grey iron[1]

For miscellaneous castings, allow half of the maximum shrinkage allowance, i.e. $\pm\frac{1}{16}$ in (0·06 in)/ft (1:200). Where considerable experience in a particular product has been obtained, half this figure may occasionally be possible. Total contraction, and therefore dimensional tolerance, does not increase numerically with length, and although $\pm\frac{1}{16}$ in (1·6 mm) might be allowed on a casting 1 ft (305 mm) in length, on a casting 3 ft (914 mm) in length a tolerance of $\frac{3}{20}$ in (1:80) would probably be closer than $\pm\frac{3}{16}$ in (1:64).

It is important to remember that the figures quoted do not, as a rule, include joint lines. Dimensions over a joint line are likely to show up to double the normal dimensional tolerances. It is only with highly repetitive work that tolerances of the above order are likely to be improved upon in greensand moulding. A Committee of the Institute of British Foundrymen[2] found it impossible to give reliable figures for tolerances, and the foregoing figures must therefore be taken as approximations only.

Table 164 is quoted for American and Canadian practice[1, 3].

The degree of accuracy to which a casting can be made obviously affects the machining and grinding allowances. It is not really possible, therefore, to define machining allowances except with reference to a fixed datum point on the casting, and thus it is desirable to make the machining allowance at least twice the maximum dimensional tolerance based upon a particular datum point in the casting.

Moulding in cores increases dimensional control, and shell moulding can increase the accuracy most markedly. Claims are made for the shell moulding process that castings can be controlled within 0·005 in per in (1:200) excluding the joint line for which an additional allowance of 0·007 to 0·01 in (1:143 to 1:100) may be made. These close dimensional tolerances are also combined with a high surface finish. This high degree of reproducible accuracy is the

chief advantage claimed by the shell moulding process, and is partly inherent in the process itself and partly due to the high degree of accuracy with which the patterns used for this process are normally made, but the precise tolerances that can be attained depend very greatly upon the design of the casting. For malleable cast iron the tolerances in *Table 165* have been put forward[4].

Table 164 FINISH ALLOWANCE FOR GREY IRON CASTINGS

Approximate size of casting		*Approximate tolerances*[1] *to be expected for as-cast dimensions in sand castings*		*Tolerances*[3] *above machining allowance*	
		in	(mm)	in	(mm)
Up to 6–8 in	(152–203 mm)	$\frac{1}{16}$	(1·6)	$\pm\frac{1}{32}$	($\pm$0·8)
Up to 12–14 in	(305–356 mm)	$\frac{3}{32}$	(2·4)	$\pm\frac{1}{16}$	($\pm$1·6)
Up to 24 in	(610 mm)	$\frac{5}{32}$	(4·0)	$\pm\frac{3}{32}$	($\pm$2·4)
Up to 36 in	(914 mm)	$\frac{3}{16}$	(4·8)	$\pm\frac{5}{32}$	($\pm$4·0)
Above 36 in	(914 mm)	$\frac{1}{4}$	(6·4)	$\pm\frac{3}{16}$–$\frac{1}{4}$	($\pm$4·8–6·4)

Table 165

Dimensional range		*Repetition or mounted patterns*		*Non-repetition and loose patterns*	
Below 4 in	(102 mm)	$\pm\frac{1}{32}$ in on dimension	($\pm$0·8 mm)	$\pm\frac{3}{64}$ in on dimension	($\pm$1·2 mm)
4–8 in	(102 mm–203 mm)	$\pm\frac{5}{64}$ in on dimension	($\pm$1·2 mm)	$\pm\frac{1}{16}$ in on dimension	($\pm$1·6 mm)
Over 8 in	(203 mm)	$\pm\frac{1}{16}$ in per foot	($\pm$1:192)	$\pm\frac{3}{32}$ per foot	($\pm$1:128)

A schedule of dimensional tolerances and machining allowances has been put forward by the Supplement Journal d'Information, Techniques des Industries de la Fonderie, and a translation published in America[5]. The method of use is detailed and the values approximate to those previously given.

Extracts for dimensional tolerances are given in *Table 166*.

Table 166

Dimension		*$\pm$Tolerance L, General*		*$\pm$Tolerance A, wooden patterns on plates*		*$\pm$Tolerance B, metal patterns*	
in	(mm)	in	(mm)	in	(mm)	in	(mm)
10 × 10	(254 × 254)	0·12	(3·0)	0·08	(2·0)	0·06	(1·5)
10 × 25	(254 × 635)	0·18	(4·6)	0·12	(3·0)	0·10	(2·5)
25 × 10	(635 × 254)	0·16	(4·1)	0·1	(2·5)	0·06	(1·5)
25 × 25	(635 × 635)	0·18	(4·6)	0·14	(3·6)	0·10	(2·5)
25 × 40	(635 × 1016)	0·26	(6·6)	0·18	(4·6)	0·14	(3·6)
25 × 64	(635 × 1626)	0·34	(8·6)	0·24	(6·1)	—	—
64 × 10	(1626 × 254)	0·22	(5·6)	0·16	(4·1)	—	—
64 × 25	(1626 × 635)	0·30	(7·6)	0·20	(5·1)	—	—
64 × 64	(1626 × 1626)	0·42	(10·7)	0·28	(7·1)	—	—
64 × 160	(1626 × 4064)	0·64	(16·3)	0·42	(10·7)	—	—

No clear standards of dimensional tolerances, however, have been laid down for any type of casting and the figures quoted must not be taken as standards. Agreement on casting tolerances must be reached between the user and the producer.

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7

Components with high local loadings

When two curved surfaces slide or roll over each other (*Figure 130*) theoretically contact is made along a line or at a single point, and assuming no deformation to take place the ultimate load should be infinitely high. In practice, of course, local yielding ensures that contact does take place over a finite area and the effective pressures (mean and maximum) can be determined theoretically by means of the methods devised by Hertz. It is not easy to determine how far these pressures are actually achieved in practice but the figures do provide a basis whereby two different sets of conditions can be compared.

The figures are here expressed in lb/in^2 and Imperial units as these are the units in which the original work was reported.

GEARS

For a cylinder on a flat plate or a cylinder on cylinder with parallel axes the Hertz formula[1]:

$$S_{max} = 0{\cdot}418 \times \sqrt{\left(\frac{F_1 E_e}{R_r}\right)}$$

gives the theoretical maximum stress, where:

F_1 = the load per inch of line contact normal to surfaces.

$E_e = \dfrac{2E_1 \times E_2}{E_1 + E_2}$ = harmonic mean of respective moduli of elasticity ($E_e = E$ if materials are the same on both faces).

$R_r = \dfrac{R_1 \times R_2}{R_1 + R_2}$ = relative radius of curvature.

Where one surface is concave the value for R is negative.

This figure S_{max} assumes isotropic materials and static loading, and it is

of value in providing a figure for loading which may be compared with a maximum determined by practical experience.

The use of 'pounds per inch of line contact' for comparison purposes can be misleading, since the intensity of load increases rapidly as the radius of

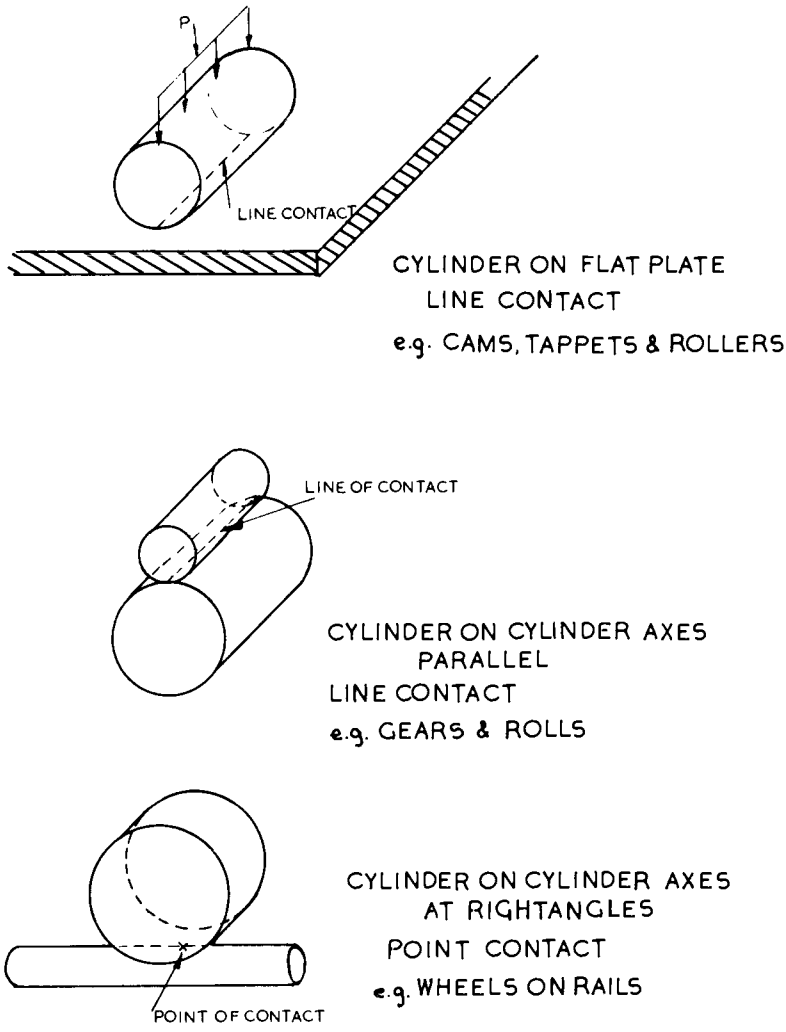


Figure 130

curvature of one or both surfaces is reduced, but using the principle of similarity, the static stress conditions are the same if the applied load per inch of line contact is proportional to the relative radius of curvature. The following criterion has been developed on this basis:

$$S_c = \text{surface stress value (compressive)} = \frac{F_1}{R_r}$$

$$= \frac{\text{load per inch of line contact (lb)}}{\text{relative radius of curvature (in)}}$$

Although the S_c value is occasionally expressed as pounds per square inch, it must be remembered that it is actually a criterion expressed in pound and inch units and not a stress in pounds per square inch[1], since it is the equivalent load per inch of line contact for a 1 in radius (2 in diameter) cylinder on a flat plate and must not be expressed in pounds per square inch.

The S_c value is occasionally expressed as a stress in pounds per square inch on the assumption that R_r can be interpreted as a dimension. R_r, however, should be regarded as a factor and not as a dimension and it is then a reasonable interpretation to express it as a load per unit length of contact, e.g. pounds per inch of line contact for a standard cylinder on a flat plate.

Merritt[1] expresses this clearly in the second edition (1946) of his book on 'Gears' (p. 262) where he states '... the S_c value is a criterion in pound and inch units and not a stress in pounds per square inch ... it is in fact the equivalent load for a 1 in radius (2 in diameter) cylinder on a flat plate and must not be expressed as a stress.'

Thus, if the surface stress value in a given case be, say, 2000 lb and inch units the stress conditions can be likened to those occurring in a cylinder 1 in radius and 1 in long, and a flat plate of corresponding materials, pressed together under a load of 2000 lb.

Thus, if the maximum safe surface stress value (S_c) is 2000 lb/in of line contact for a curved surface 1 in radius (2 in diameter) and 1 in long operating against a flat plate for a given pair of materials, a 10 in radius wheel and a flat rail of these materials can be loaded up to 20 000 lb/in of line contact before this limit is reached.

Thus, the loading limit per inch of line contact for a given S_c value for a given pair of materials equals $S_c \times$ radius $\times$ length of line contact, or

$$S_c \times \frac{\text{diameter}}{2} \times \text{length of line contact}$$

The approximate maximum surface pressure, ignoring oil film, can be obtained from:

$$S_{max} = 0{\cdot}418 \times \sqrt{(S_c E_e)} \quad \text{lb/in}^2$$

The maximum shear stress equals approximately $0{\cdot}3S_{max}$ and occurs at a depth of $0{\cdot}39b$ below the surface, where b is breadth of contact zone[2].
For two cylindrical surfaces:

$$b = 3{\cdot}04 \times \sqrt{\left(\frac{F_1 R_1 R_2}{E_e(R_1 + R_2)}\right)}$$

For a cylinder on a plane surface:

$$b = 3{\cdot}04 \times \sqrt{\left(\frac{F_1 \times R}{E_e}\right)}$$

For bearings where a shaft revolves or oscillates in a bearing of slightly larger diameter a modified Hertz equation gives the width (b) of contact:

$$b = 2{\cdot}15\sqrt{\left(\frac{WD^2}{LEC}\right)}$$

where W = load, L = bearing length, D = nominal bearing diameter, C = diametrical clearance, and E = modulus of elasticity. Thus 'b' varies inversely as the square root of C and becomes large for small values of C. Large clearance, therefore, reduces contact area and increases contact loading excessively. Reducing clearance too much, however, reduces oil flow. The modified Hertz equation is valid for new surfaces and the projected area becomes more valuable after running-in[3].

The following values for S_{max} are given for five combinations of materials with differing modulus of elasticity:

(1) For steel versus steel, $E_1 = E_2 = 30 \times 10^6$: $S_{max} = 2290\sqrt{S_c}$
(2) For chilled iron versus chilled iron, $E_1 = E_2 = 30 \times 10^6$: $S_{max} = 2290\sqrt{S_c}$
(3) For grey iron (16–20 ton/in^2, $E_1 = 18 \times 10^6$) versus hardened steel ($E_2 = 30 \times 10^6$), $E_e = 22{\cdot}5 \times 10^6$: $S_{max} = 1980\sqrt{S_c}$
(4) For grey iron versus grey iron (16–20 ton/in^2 tensile) where E_1 and $E_2 = 18 \times 10^6$: $S_{max} = 1810\sqrt{S_c}$
(5) For grey iron versus grey iron (10–12 ton/in^2 tensile) where E_1 and $E_2 = 15 \times 10^6$: $S_{max} = 1620\sqrt{S_c}$

These figures differ slightly from those given by Merritt for cast iron as they are based on more recently determined E values.

The S_c factor for an iron such as (3) is given[4] as 1450 for gears where conditions of sliding and rolling occur, and therefore $S_{max} = 76\,000$ lb/in^2 for spur and helical gears.

Surface stress characteristics for varying degrees of rolling and sliding have been examined by Talbourdet[5] using lubricated rollers with initial surface finish of 90–120 RMS (approximately 80–100 CLA), giving sliding to rolling percentages of from 75 per cent sliding to pure rolling under varying loads by estimating the number of cycles to failure. The results follow the characteristic S/N curve of fatigue testing and the S_c values given in *Table 167* are taken from this work and represent the limiting values for 100 million cycles. The surface fatigue of these materials is influenced by the number of cycles of stress, but not, apparently, by the rapidity of the cycle. The figures are given in *Table 167* and the values of mechanical properties shown in round brackets are interpolated when full information is lacking in the paper. The figures for chilled iron are based on the limits for cam and tappet practice given by Turkish[6].

The working figures given in *Table 168* are based on BS 545:1949 and BS 436, Amendment 1960, which also give a correction chart for speed to keep the working figures within the limits of surface fatigue failure. The chilled iron figures are based on British automobile cam and tappet practice and are included for comparison, although the suitability of the material for gears is governed by its shock resistance.

Table 167 SURFACE STRESS FACTORS FOR VARIOUS COMBINATIONS OF CAST IRON AND STEEL UNDER VARYING CONDITIONS OF ROLLING AND SLIDING[5]

	Mechanical properties			*Rolling + % sliding, S_c values*					
	Tensile strength ton/in^2	*E* lb/in^2 × 10^6	*Brinell hardness no.*	*Rolling*	*8–9%*	*30%*	*50%*	*75%*	*100% sliding*
Cast iron *v.* Cast iron	8·9	(10–14)	130–160	1300	1100	900	540	480	—
Cast iron *v.* Cast iron	(24·5) (24·5)	(20–23) (20–23)	190–240 190–240	1950	1500	1300	950	600	—
Cast iron (interrupted quench) *v.* Cast iron (interrupted quench)	(28–32) (28–32)	(18–20) (18–20)	270–290 270–290	4400	3400	2800	2500	2000	—
Unalloyed grey iron *v.* Hardened steel	11·1–13·4 —	(14–16) 30	140–190 58–62R_c	1000–1300	900	—	—	—	—
Alloy cast iron *v.* Hardened steel	13·4–15·6 —	(15–17) 30	225–255 58–62R_c	2300	2100	—	—	—	—
Cast iron (interrupted quench) *v.* Hardened steel	(28–32) —	(22–28) 30	270–290 60–62R_c	3100	2800	—	—	—	—
Low carbon steel *v.* Hardened steel	37·5 —	30 30	130–150 60–62R_c	1700	1350	—	—	—	—
Low carbon steel carburised *v.* Hardened steel	— —	30 30	52–58R_c 60–62R_c	13 000	8500	—	—	—	—
Chilled iron *v.* Hardened steel	— —	26 30	350–400 60–62R_c	—	—	—	—	—	8000
Chilled iron *v.* Chilled iron	— —	26 26	350–400 350–400	—	—	—	—	—	8000

Table 168 CAST IRON GEARS[4, 5]

	Tensile strength ton/in^2	*E* lb/in^2 × 10^6	*Brinell hardness no.*	*Surface stress factor,* S_c	S_{max}	S_{max} *v. hardened steel* lb/in^2	*Bending stress,* S_b lb/in^2
Cast iron (as-cast)	12	12–16	165	1025	$1700\sqrt{S_c}$	54 500	6000
Cast iron (as-cast)	16	15–18	210	1350	$1700\sqrt{S_c}$	63 000	7600
Cast iron (as-cast)	22	18–22	220	1450	$2000\sqrt{S_c}$	76 000	10 500
Alloyed iron	20	18–22	240–280	1750	$2000\sqrt{S_c}$	84 000	10 000
Cast iron, high strength	22	18–22	300	2100	$2000\sqrt{S_c}$	92 000	10 000
Malleable iron	20	24–27	140	850	$2150\sqrt{S_c}$	63 000	12 000
Nodular iron, pearlitic/ferritic[7] BS 2789 SNG 42/2	42	23–26	230	1900	$2100\sqrt{S_c}$	91 000	20 000
Nodular iron, hardened and tempered BS 2789 SNG 47/2	47	23–26	250	2100	$2100\sqrt{S_c}$	96 000	22 000
Chilled iron	28–35	30	350–400	2400*	$2290\sqrt{S_c}$	116 000	—
Chilled iron *v.* chilled iron	28–35	30	350–400	3180*	$2290\sqrt{S_c}$	129 000	—
NiCrMo case-hardened steel†	85	30	710 HV (case)	13 000	$2290\sqrt{S_c}$	261 000	50 000

* These are normal figures, but in exceptional cases S_c values of 7000 are employed in cam and tappet operation requiring S_{max} values of 179 000. With 30 in or 100 in spherical faced chilled iron tappets on hardened steel or chilled iron cams[6], S_{max} should not exceed 190 000 lb/in^2. Tentative figures for 30 per cent Cr, 1 per cent C cast iron (tensile 25–30 ton/in^2; HB: 280–300; E: 30 × 10^6 lb/in^2) are $S_c = 2100$, $S_{max} = 105\,000$, $S_b = 12\,000$ tested with water lubrication.
† En 39 A and B (BS436 amendment No. 51960).

According to Merritt[1], cast iron gears should have hardnesses of 170–220 for free machinability, and higher in special cases. The total carbon content should be not more than 3·4 per cent, and the phosphorus content up to 0·8 per cent, as this improved wear-resistance.

It has been stated that according to Hertz's theoretical work the permissible load varies as the square of the hardness. Practical experience, however, indicates that the safe loading varies directly as the hardness (*see Figure 131*).

Figure 131 summarises the data given for bevel gears (BS 545:1949) and Talbourdet's results. It will be seen that the figures from BS 545 are roughly comparable with Talbourdet's figures for rolling with 30 per cent sliding. Talbourdet's figures indicate that in pure rolling higher S_c values may be tolerated with cast iron versus cast iron, than with cast iron versus hardened steel, possibly owing to the lower modulus of elasticity.

For comparison purposes, the values for steel and bronze are given in *Tables 169* and *170*.

For different mating surfaces Merritt[1] gives,

For steel and steel:	$S_{max} = 2250\sqrt{S_c}$
For steel and cast iron:	$S_{max} = 1695\sqrt{S_c}$
For steel and bronze:	$S_{max} = 1870\sqrt{S_c}$
For cast iron and cast iron:	$S_{max} = 1425\sqrt{S_c}$

These figures differ slightly from those given earlier, as lower values for E for cast iron were used by Merritt.

The figure for cast iron is consistent with $E - 11{\cdot}6 \times 10^6$ lb/in^2 which would apply to an iron lower than BS 1452, Grade 10 or 12, whereas Grades 17–23 are normally employed for gears and E values of $18–21 \times 10^6$ lb/in^2 would be expected.

Worm gears

The above figures do not apply to worm and worm wheels which are used to connect skew shafts, usually but not necessarily at right angles and where the ratio of slide to travel is high and cast iron, grey or nodular, shows up less favourably. The figures given in *Table 167* for 75 per cent sliding should be regarded as the absolute limit and used with a suitable safety factor. Like spur gears, they give line contact between the teeth but the sliding velocity is high and special attention must be given to choice of material and lubricants.

For grey irons[1] the tooth rubbing speed for worm gears should not exceed 8–10 ft/s when operated against phosphor bronze or cast iron and the material has been regarded as suitable for hand-operated motions only against steel[4].

However, a recent paper by Niemann and Michaels[8] records tests of various materials against hardened steel and showed that a pearlitic grey iron worm gave excellent life comparable with or better than phosphor bronze

Table 169 STEEL, HELICAL, STRAIGHT SPUR AND BEVEL GEARS

	Tensile strength ton/in^2	$E \times 10^6$ lb/in^2	*Brinell hardness no.*	*Surface stress* S_c *factor*	*Bending stress* S_b lb/in^2
0·35%–0·45% carbon steel normalised (BS 436:1960)	35	30	152	1400	19 000
0·50%–0·55% carbon steel, surface-hardened (BS 436:1960)	35	30	152–500	2800	17 000
	45	30	201–520	4000	21 500
			DPN		
0·12%–0·22% carbon steel, surface-hardened (BS 545:1949)	32	30	140–750	9200	40 000
5% nickel steel, case-hardened (BS 545:1949)	65	30	293–710	12 000	50 000

Table 170 BRONZE BEVEL GEARS

	Tensile strength ton/in^2	$E \times 10^6$ lb/in^2	*Brinell hardness no.*	*Surface stress* S_c *factor*	*Bending stress* S_b lb/in^2
Phosphor bronze*[4] sand cast (BS 421)	12	16	69	710	7200
Phosphor bronze[4] chill cast (BS 421)	15	16	82	880	9100
Phosphor bronze[4] centrifugally cast	17	16	90	1000	10 000

* Tin, 11–13%; phosphorus, not less than 0·15%; lead not more than 0·5%; nickel not more than 0·5%; zinc not more than 0·3%; all other elements not more than 0·15%; remainder copper. Test samples cut from casting portion representing the teeth of the finished wheel. Not annealed or treated otherwise than the casting.

when the worm, but not the steel gear, was cadmium plated. The conditions of test were as follows:

Worm: pearlitic grey iron (GG 26) roughly BS 1452 Grade 14–17, HB 180–210.
Cadmium plated 0·0002–0·0007 in thick.
Surface finish 10 μinches CLA.

Gear: hardened steel HV 760.
Surface finish 10 μinches CLA.
Surface speed 3 m/s (9·4 ft/s) 100 per cent sliding.
Surface loading 1340–1820 (in and lb units). $S_{max} = 62\,000$–$72\,500$ lb/in^2.
Lubricant, hypoid running-in oil for 3 h followed by hypoid oil.

Cadmium plating of one surface gave better results than plating with copper or tin or phosphating one or both surfaces.

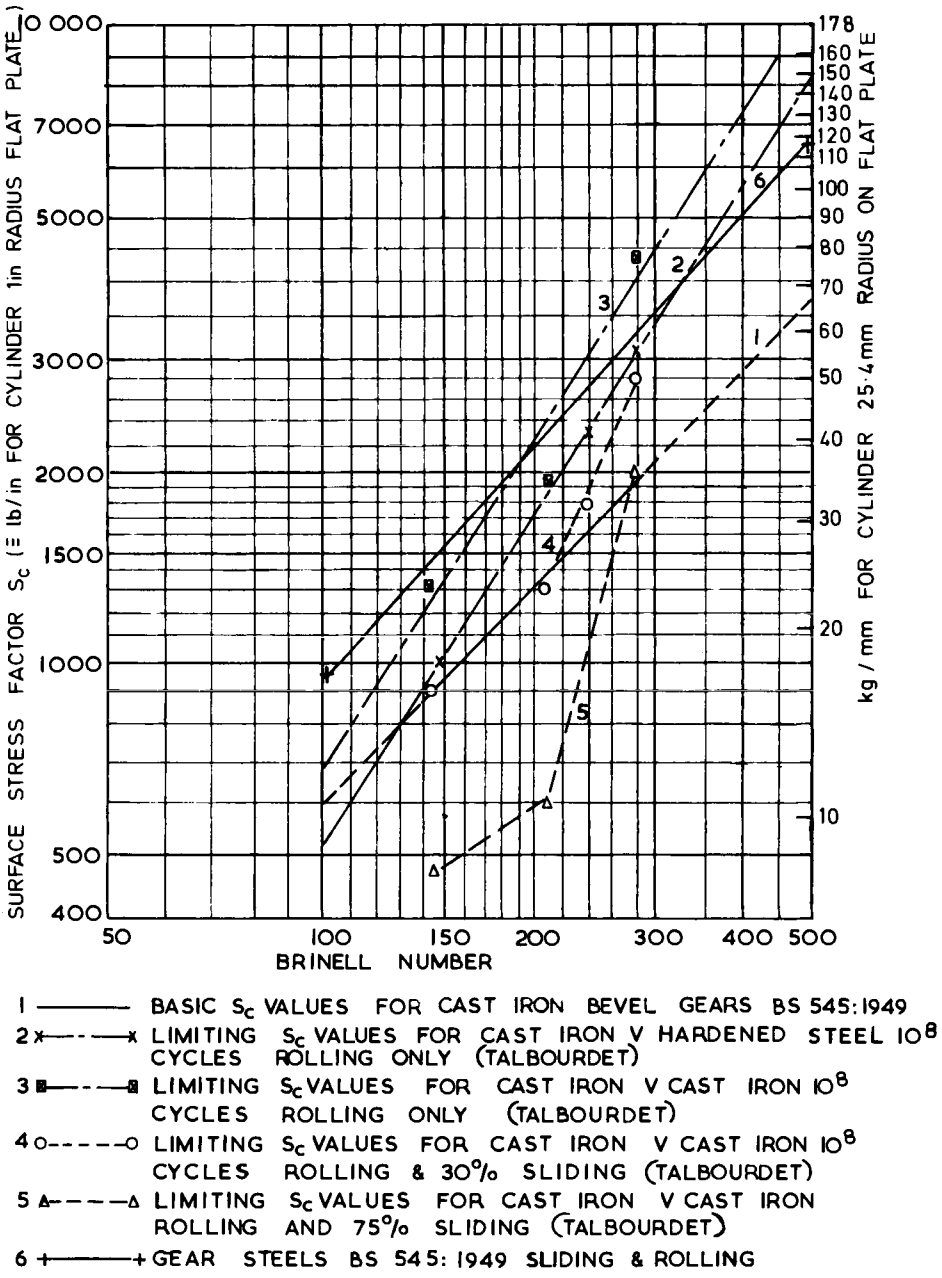


Figure 131. Surface stress factors for malleable and cast iron of varying hardness under varying degrees of rolling and sliding

These figures are comparable with those quoted in *Table 168* for gears and should be subject to service test before general use.

In chilled irons[6] the tooth rubbing speed of gears should not exceed 40 ft/s.

Spiral gears are used to connect skew shafts and the teeth make point contact instead of line contact and there is longitudinal sliding so they are used only for comparatively light loads (e.g. timing gear) and grey iron with hardnesses of 240–280 is in general use, as for camshafts.

Corrections to safe surface loadings given above may be applied[1, 3]. The use of loadings of nodular or SG iron for gears has been discussed by Rettig[9].

Surface finish is of great importance and should not exceed 10–20 μinches CLA and finish machining or grinding should eliminate the surface layers distorted by previous heavier machining cuts.

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Cams and tappets

Cams for use in internal combustion engines may be of hardened steel, flame-hardened cast iron, or chilled cast iron.

The contact against the tappet face is almost 100 per cent sliding.

Burwell and Strang[1] studied sliding wear using a lubricated pin-disc technique with a disc of hardened steel tempered to 550 HB and steel pins of, respectively, 223 and 430 HB and sliding speeds of 20 cm/s. They showed that a change from constant mild wear rate to very rapid wear occurred when the pressure of contact exceeded $H/3$ kg/mm^2, where H is the hardness of the pin or the softer of the two components. Using a pin of 430 HB, the transition occurred at 145 kg/mm^2.

The hardness of chilled cams and tappets in normal use is between 430 and 550 HB and $H/3 = 143$–183 kg/mm^2. The normal commercial operating loads give S_{max} values of the order of 55–80 kg/mm^2 (80–120 000 lb/in^2).

Normal automobile practice, therefore, gives a factor of safety of about $2\frac{1}{2}$ when compared with Burwell and Strang's maximum value which indicates that their contact stress value of $H/3$ kg/mm^2 is a reasonably reliable index of the maximum stress for the softer of two mating surfaces in lubricated sliding contact.

Table 171 TYPICAL CAMSHAFT COMPOSITIONS AND PROPERTIES

Composition								*Mechanical properties*			
TC %	*Si* %	*Mn* %	*S* %	*P* %	*Ni* %	*Cr* %	*Mo* %	*Tensile strength (0·875 in bar)* ton/in^2	*Transverse rupture stress (0·875 in bar)* ton/in^2	*Shaft*	*Nose*
3·1–3·4	2·0–2·4	0·5–0·8	0·1 max	0·2 max	0·15–0·25	0·8–1·0	0·15–0·25	19	32	240–280 Brinell	45–50 R_c
3·3–3·5	2·3–2·5	0·6–0·8	0·1 max	0·2 max	0·3–0·5	0·3–0·5	—	—	—	—	450–550 HV

In this country, the cast iron camshaft with a chilled nose to the cam is very widely employed. A typical camshaft composition and properties are given in *Table 171*.

A chill is incorporated in the mould at each cam nose so that at this point the metal solidifies white, although the flanks remain grey (*Figure 132*).

There is no sharply defined hard case on a relatively soft matrix as with case-hardened or flame-hardened steel. The amount of free carbide, and therefore the hardness, declines progressively from the nose of the cam until it merges into the light mottle in the body of the shaft.

The hardness of the nose is approximately 40–45 Rockwell C* and the flanks 240–320 Brinell.

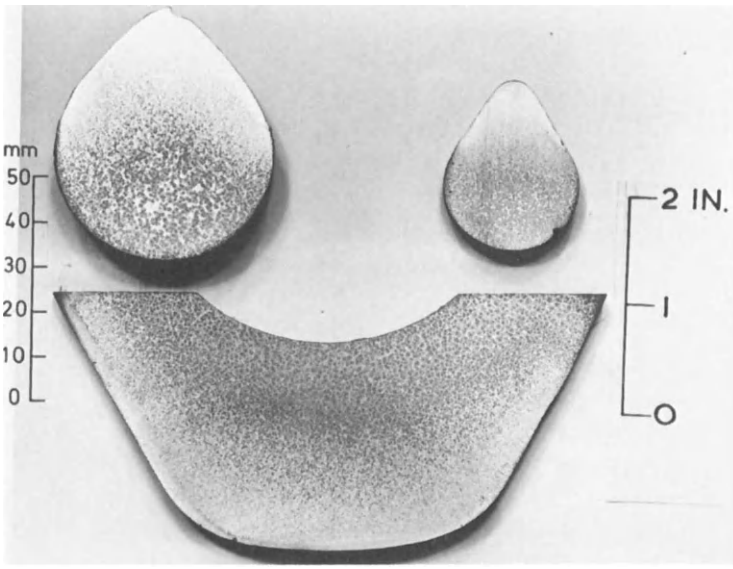


Figure 132. Three sizes of chilled nose cast iron cams (courtesy of Midland Motor Cylinder Co, Ltd)

The chilled nose structure consists basically of carbide and pearlite. It is possible to increase the hardness by increasing the amount of carbide or by alloying to produce martensite in place of pearlite, but there is evidence to show that increasing the hardness above about R_c 47 may increase the liability to surface crumbling.

Chilled tappets used in conjunction with the camshafts normally have a hardness from R_c 45 to R_c 50. The chilled structure is basically similar to that of the cam nose. Hardnesses lower than the range quoted may result in wear, and hardnesses above R_c 52–55 may result in chipping or spalling

* The Rockwell scale of hardness is widely employed in cam and tappet practice and conversion to the Diamond Pyramid scale cannot be regarded as absolute. The conversion figures quoted in the section on hardness (BS 860:1939) are a reasonable approximation

of the surface. It is interesting to observe that the hardness of the tappet is usually appreciably higher than that of the cam nose.

For special purposes tappet faces of tungsten carbide may be brazed on to cast iron tappets.

Satisfactory life will be likely if stress loadings at the cam nose do not exceed about 120 000 lb/in^2 (Hertz[2]).

$$S_m = 0{\cdot}591 \times \sqrt{\left\{ WR\left(\frac{1}{E_2}+\frac{1}{E_1}\right)\right\}^{P}} \text{ lb/in}^2$$

Where S_m = contact stress in lb/in^2 max
P = load, lb
W = cam width, in
R = nose radius, in or R_t where the follower is curved
E_1 and E_2 = moduli of contacting materials[2].

The S_c and S_{max} values quoted previously (*Table 168*) appear to satisfy the conditions for cams and tappets as a first approximation.

Surface speeds[2] over the cam may range in petrol engines from 8 to 40 ft/s (480 to 2400 ft/min).

Spring loads may range from 100 to 250 lb, nose radii from 0·1 to 0·225 in, nose widths from 0·4–0·5 in, and maximum stress (Hertz) from 80 000 to 128 000 lb/in^2. Turkish states that S_{max} should not exceed 120 000 lb/in^2 (S_c 3450) for chilled iron tappets and hardened steel cams, flat faced.

These figures[1] are based on flat-face chilled tappets and chilled nose camshafts. Increasing the cam width in order to spread its load is only effective if accurate alignment can be maintained. For high speed engines, cam widths of $\frac{1}{2}$ in are common, but for slow speed diesels, cam widths of $1\frac{1}{2}$ in are not uncommon.

Tappets with faces ground to 30 in or 100 in spherical contour are commonly employed, although this would appear to increase the stress concentration compared with flat-face tappets.

Turkish* states that for 30 in or 100 in spherical chilled cast iron tappets on steel or cast iron cams, the calculated stress should not exceed S_c 8000 or 190 000 lb/in^2. He gives a method for the calculation of these stresses on spherical surfaces. British practice confirms that under these conditions, loads of 160 000–180 000 lb/in^2 may be used and these values are comparable with those used in American railroad practice for chilled iron wheels.

The increase in surface stress with spherical ground tappet faces may be more apparent than real, since with flat-face tappets, uniform contact pressures will only be obtained with extreme accuracy of alignment.

Inertia[2] and spring load on flanks of cams at 400 rpm is several times greater than the spring load on the nose of the cam at zero speed, but the stress on the nose is greater because of its small radius.

Since chilled iron has a high fatigue stress in compression but a lower

* Turkish uses (a) for steel: $E = 30 \times 10^6$ lb/in^2 and Poisson's ratio of 0·25; (b) for chilled cast iron: $E = 20 \times 10^6$ lb/in^2 and Poisson's ratio of 0·25. The E value for white iron, however, appears to be nearer to 30×10^6 than 20×10^6

fatigue stress in shear, failure under surface stresses commences below and parallel to the surface and spalling may occur, forming a pit. It appears to be general experience that cam and tappet failure by pitting is more likely to arise at low than at high speeds. It is possible that full hydrodynamic lubrication occurs more readily at high surface speeds and so prevents metallic contact between the mating surfaces.

Failure in white iron and steel commences as microscopic surface cracks parallel to the surface and about 0·04–0·06 in below it. This is the region of maximum Hertzian stress. These cracks break out to the surface and form pits. Craze cracking normal to the surface can also occur following the dendrite pattern.

For hardened steel roller followers with chilled iron cams in marine diesels running at 200–400 rpm, a 1000 lb load per in of line contact is about the safe maximum, and loads of 800 to 1000 lb per in are normal. Loads of 1300 lb are occasionally employed successfully. A combination of hardened steel cam versus grey iron surface has been found to operate successfully (provided that the Hertzian load did not exceed 50 000–60 000 lb/in^2) under conditions where case-hardened surfaces operating against each other wore out rapidly under any load. In the case of an hydraulically operated motor, a heat-treated cast iron ring operated successfully (at about 20 rpm) against a heat-treated grey cast iron pusher at a Hertzian load of 80 000 lb/in^2 when hardened steel components had failed by pitting. The cast iron components had an approximate composition of: total carbon, 2·75 per cent; silicon, 0·83 per cent; manganese, 1·1 per cent; sulphur, 0·01 per cent; phosphorus, 0·37 per cent; nickel, 0·07 per cent; chromium, 0·06 per cent; and molybdenum, 0·06 per cent; and were oil quenched and tempered to give a hardness of 520–580 DPN.

The finish of cams and tappets is of vital importance, particularly where high loadings are employed (120 000 lb/in^2 or more). Under such loads, the tappets should be finished to 4–6 μinches CLA, or hand-lapped if of hardened steel. The cams are not so critical as the tappets and a finish of 25 μinches CLA may be acceptable.

Although for valve operation where the cam is spring and inertia loaded the flank need not exceed 240–280 Brinell, for diesel pumps which are cam operated the flanks need to be as hard as the nose, since the load is a fluid flow resistance load.

Flame-hardened cams are widely used in America, partly for production reasons, and it is not uncommon to flame-harden the entire cam surface. This appears to offer advantages chiefly when flank wear is high, for example, in fuel pumps. Many people in this country are satisfied that flame-hardened martensitic cast iron cams with a hardness of 50 R_c or above are not so satisfactory as chilled iron cams with hardness from 40 R_c and which are less susceptible to variations in lubricant than hardened cast iron or steel cams.

The oil-retaining properties of a flame-hardened grey iron cam would appear to be better than those of a chilled iron cam, but high stress concentrations on the edges of exposed graphite flakes may possibly initiate crumbling. There is also the possibility of developing high oil pressures in front of the

sliding component at the root of exposed graphite flakes, which might under some conditions contribute to failure.

Chilled white iron carries two constituents of widely differing hardness (cementite and pearlite) and which wear at different rates so that some degree of oil retention is still possible without the presence of weakening discontinuities.

Nevertheless, flame or oil hardened cast iron cams and tappets with nominal loadings as given above are widely used in the USA with success* and the choice between chilled or independently hardened iron appears to rest on production requirements and local conditions.

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WHEELS AND ROLLERS

Wheel tread loadings can be most readily described in terms of xD lbs or tons per inch tread width, where x = number of pounds or tons and D is the diameter of the wheel or roller in inches or other consistent units. This value is one half of the S_c value given in the section on gears which is based on radius. The values in terms of diameter and related to hardness are given in *Table 172* at the end of the section. Some values for temperatures of 800–850°C are given in the section on furnace wheels and rollers.

The safe vertical load for a flanged wheel is largely based upon the section of the flange that will safely guide the load, rather than on the ability of the body of the wheel to sustain the load, although this must also be considered[1] In the following notes, axle loads refer to the load on one wheel only.

With slow speeds, higher axle loads can be employed because of less severe impact on the flange. In the case of grey iron wheels, wear of the tread and flange must also be considered.

The following figures[1] have been employed in British practice for crane wheels, expressed in lb per linear inch contact (d = wheel diameter in inches):

Grey iron	$200d$
Chilled iron	$672d$
Cast steel	$900d$
Rolled steel tyres (0·6%C)	$1350d$

For live steel rollers in swing bridges etc., the working load at rest is not

* Typical analyses of flame or induction hardened cams or tappets:
TC, 3·1–3·25%; Si, 2·1–2·3%; Mn, 0·7–0·8%; S<0·12%; P<0·12%; Mo, 0·5–0·65%; Cr, 1·0–1·1%; Ni or Cu, up to 0·5%. Tappets are fully hardened by oil quenching from 900°C and tempered at 200°C. American chilled cast cams are frequently made from an iron such as: TC, 3·6%; Si, 2·1%; Mn, 0·85%; Cr, 1·1%; Mo, 0·6%

to exceed $530d$ and in motion $375d$. The figure for grey iron, however, takes no account of hardness and is in fact only suitable for irons above 200 Brinell.

The resistance to wear and distortion under dry rolling and abrasive conditions increases with increasing hardness, and the presence of free carbide in the microstructure either by alloy additions or by surface chilling increases the surface resistance of the wheel up to the point where the surface is entirely white. Strength and machining requirements usually make it desirable to have grey webs and centres.

With chilled iron wheel surfaces, the limit to the loading appears to be the ability of the rail, rather than the wheel surface, to carry the load.

CHILLED RAIL WHEELS

The wheels are cast in sand moulds with the tread against a continuous metal ring chill.

The wheels are removed from the mould at about 800–850°C, and transferred to a soaking pit for approximately 1–3 days and cooled to 620–680°C (below the critical) at a rate not exceeding 20–22°C/h. They are then transferred to a secondary pit where cooling is more rapid. The depth of chill from the tread is approximately $\frac{1}{2}$ in.

According to the Association of American Railroads, Specification M 403–48, the hardness of a chilled iron wheel should be a minimum of 352 Brinell at a depth of $\frac{3}{8}$ in below the tread surface*, and should not exceed 215 Brinell in the centre of the rim and web junction. The basic chemical composition is given as:

	%
Total carbon	3·0 min.
Combined carbon	0·90 max.
Silicon (not specified)	0·60–0·75
Manganese	0·5 min.
Sulphur	0·14 max.
Phosphorus	0·35 max.

Railway wheels of chilled iron were standard on American railways for many years and have been used for high speed freight service carrying loads up to 26 250 lb per wheel (33 in diameter). It is stated that chilled iron wheels show no appreciable flattening by any load below 250 000 lb, although such loads are obviously not permissible in service.

Some early experimental work was carried out by Lyndon and Vial[2], using flat-topped and crowned rails with chilled wheels of varying diameter.

Figure 133 indicates the type of deformation that occurs in the rail. Typical total depressions obtained on an 8 in top radius rail are given in *Table 172*.

As distinct from the total depression, the permanent indentation in a rail varies as the applied load, and also varies inversely as the square root of the diameter of the wheel. The areas of the surfaces in contact between rail and

* No figures are given for the tread face but the hardness is likely to be 400–500 Brinell (420–550 DPN)

Table 172 TOTAL DEPRESSION ON 8 in TOP RADIUS RAIL

Load lb	*15 in wheel* in	*33 in wheel* in	*33 in on 14 in radius rail* in
20 000	0·004	0·008	0·0023
60 000	0·014	0·021	0·0023
100 000	0·023	0·032	0·0087
140 000	0·035	0·043	0·0087

wheel for varying loads and diameters are nearly proportional to the applied wheel load, and are not materially affected by the wheel diameter. The first permanent set in a rail occurs with a depression of approximately 0·007 in.

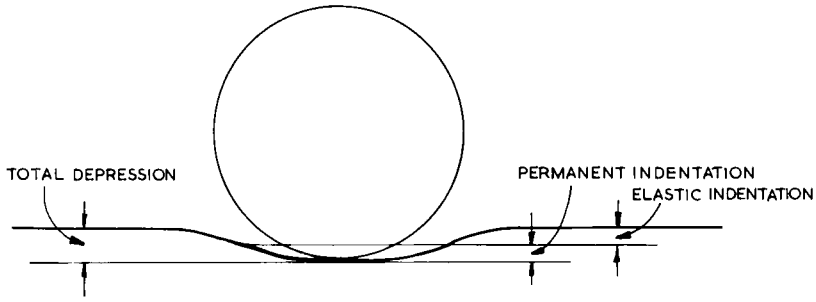


Figure 133. Diagrammatic illustration of rail deformation under a loaded wheel

Table 173 indicates areas of contact by direct measurement, and average pressure per in^2 on a flat-topped rail 2 in wide, under a direct wheel load of 100 000 lb (44·6 ton) using chilled iron wheels.

Table 173 FLAT-TOPPED 2 in WIDE RAILS: 100 000 lb LOAD

Diameter of wheel in	*Area of contact (100 000 lb load)* in^2	*Depth of penetration* in	*Permanent set in rail*	*Average pressure on area of contact* lb/in^2
12	0·950	0·0031	none	105 000
16	0·960	0·0027	none	104 000
20	0·970	0·0024	none	103 000
24	0·980	0·0022	none	102 000
30	0·990	0·0020	none	101 000
33	1·000	0·0019	none	100 000
36	1·010	0·0018	none	99 000

The average pressure (as distinct from maximum pressure) in lb/in^2 between a 33 in chilled iron wheel[3, 4, 5] and a 0·6 per cent carbon rail steel under varying loads, is shown in *Figure 134*.

In practice, the 33 in diameter chilled cast iron railway wheel of varying weights with a nominal tread width of $3\frac{3}{4}$ in carried an axle load not exceeding 26 250 lb per wheel on standard steel rails with 12 in radius crown[3]. The

average contact pressure with rail crown radii of 8 in and 14 in under this load is (by interpolation of direct area measurements[2, 4]) approximately 105 000–115 000 lb/in^2. The calculated maximum Hertzian stress between a chilled wheel 33 in in diameter and a 12 in radius crown rail under a load of 26 250 lb/in^2, is 180 000 lb/in^2 assuming a value of 30×10^6 for modulus of elasticity of white iron, and a value of 0·25 for Poisson's ratio for both steel and white iron[6] (*see also* pp. 425 and 436).

The average service in miles of a 33 in wheel so loaded has been given as 29 200 miles[2], equivalent to 15×10^6 applications of load.

The calculated safe maximum surface compressive stresses for chilled iron railway wheels, therefore, appear to be of the same order as those employed as a maximum in cam and tappet practice. In cam and tappet practice, however, both surfaces have hardnesses in excess of 400 and so no plastic yielding

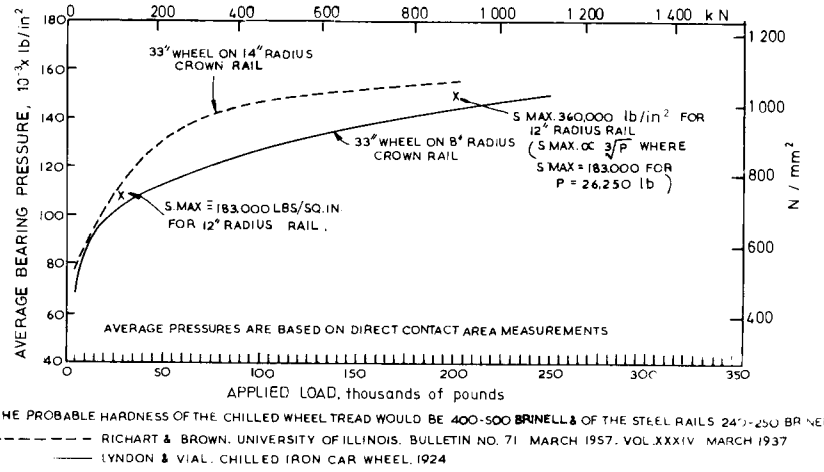


Figure 134. Average pressures between 33 in chilled iron wheels (Association of American Railroads Standard) and 0·6–0·7%C rail steel under varying applied loads

can occur to spread the load. In the case of chilled wheels on steel rails, there is no doubt that some plastic deformation of the rail occurs, so that the true maximum Hertzian stresses will be lower than those calculated (e.g. 183 000 lb/in^2). Nevertheless, these repeated stresses appear high in relation to the maximum compressive strength (approximately 224 000 lb/in^2) of white iron, but although there are no data on the actual maximum stresses as compared with the observed average stresses, the calculated figures give a basis for comparison.

The limiting factors governing these recommended wheel loadings were the stresses in the flange and throat of the wheel as a result of impact, pressure caused by rail curvature etc., and not the surface contact stresses on the wheel tread, which apparently have a large factor of safety even under railway speeds of 40–60 mph.

Crane wheels The figures in *Table 174* have been given in America for chilled iron crane wheels with flat-face treads operating on steel rails with 12 in

Table 174 DIAMETER AND WEIGHTS OF DOUBLE FLANGE CHILLED IRON CRANE WHEELS FOR VARIOUS LOADS AND FLANGE PRESSURES

Diameter of wheel in	*Weight* lb	*Weight of rail used* lb	*Tread of wheel* in	*Height of flange* in	*Thickness of flange* in	*Thickness of throat* in	*Maximum safe flange pressure* lb	*Maximum safe vertical load* lb	*672d* (British practice) lb per in tread width*
12	95	40	2¼	¾	1	1¼	3 000	4 000	8 064
12	130	50	2½	¾	1¼	1½	5 000	7 000	—
12	165	60	2¾	¾	1½	2	8 000	11 000	—
16	145	40	2¼	¾	1	1¼	5 000	7 000	10 752
16	200	50	2½	¾	1¼	1½	8 000	11 000	—
16	250	60	2¾	¾	1½	2	11 000	15 000	—
20	250	50	2¼	¾	1	1½	8 000	11 000	13 440
20	325	60	3	1	1¼	1⅞	11 000	15 000	—
20	400	80	3¼	1	1½	2¼	15 000	20 000	—
24	415	60	3	1	1¼	1¾	12 000	16 000	16 128
24	530	80	3¼	1	1½	2⅛	18 000	24 000	—
24	640	100	3⅜	1	1¾	2½	25 000	34 000	—
30	650	60	3	1	1¼	1¾	16 000	22 000	20 160
30	760	80	3¼	1	1½	2	22 000	30 000	—
30	875	100	3⅜	1	1¾	2½	32 000	43 000	—
30	1000	100	3⅜	1	2	3	47 000	63 000	—
30	1275	150	4⅛	1	2¼	3½	69 000	92 000	—
33	740	60	3	1	1¼	1¾	18 000	24 000	22 176
33	850	80	3¼	1	1½	2	24 000	32 000	—
33	1040	100	3⅜	1	1¾	2½	35 000	46 000	—
33	1200	100	3⅜	1	2	3	53 000	71 000	—
33	1525	150	4⅛	1	2¼	3½	78 000	104 000	—
36	850	60	3	1	1¼	1¾	20 000	27 000	24 192
36	985	80	3¼	1	1½	2	26 000	35 000	—
36	1200	100	3⅜	1	1¾	2½	38 000	51 000	—
36	1425	100	3⅜	1	2	3	56 000	75 000	—
36	1800	150	4⅛	1	3½	3½	86 000	115 000	—

* d = wheel diameter (in).

radius[5]. *Figures 135* and *136*, and the final column in *Table 174*, show for comparison the British figures based on 672 lb per linear in per in wheel diameter[2].

These figures are based on experimentally determined maximum safe horizontal flange pressures which are given with a factor of safety of 5. The maxi-

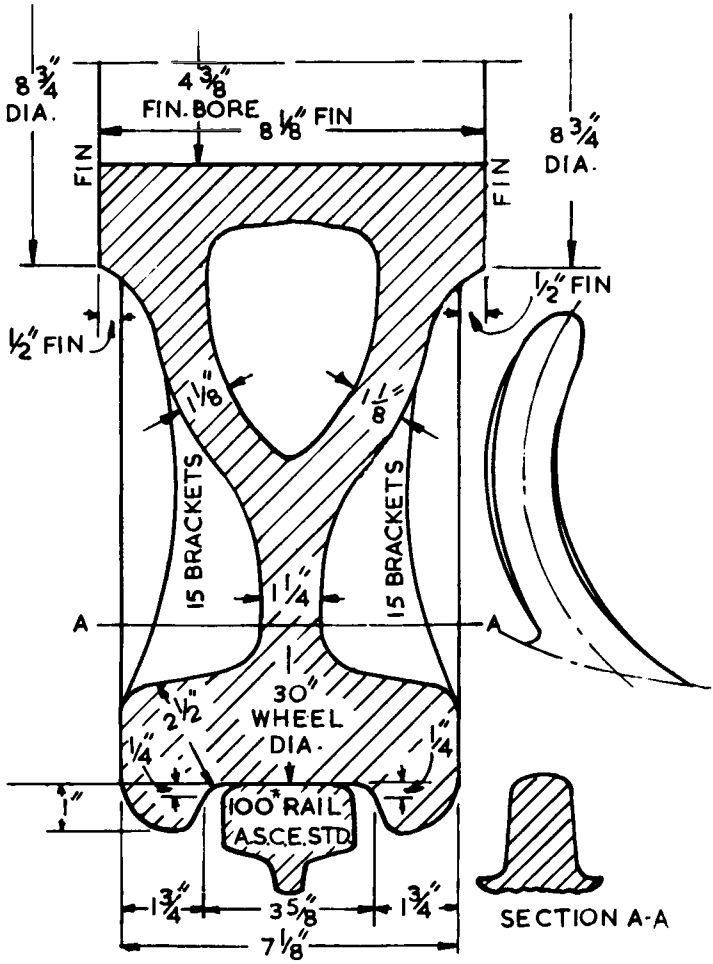


Figure 135. 30 in, 875 lb double flange chilled iron crane wheel, double plate type, for maximum vertical load 43 000 lb, maximum safe flange pressure 32 000 lb

mum safe vertical load is taken as approximately 1·25 times the safe flange pressure and bears no direct relation to the ability of the chilled tread to carry the load, which is many times greater than the figures shown, and is in fact, greater than the surface load carrying ability of a normal rail steel (0·06–0·07 per cent carbon, Brinell 240–250).

The successful production of chilled iron wheels to operate under these

arduous conditions is entirely dependent upon foundry technique and the data given have been based upon wheels made by specialised foundries.

Grey cast iron wheels These are quite widely employed, particularly in relatively low speed uses, and have very substantial load-carrying capacity. They require a less specialised production technique than chilled iron wheels.

The load-carrying capacity of the tread of a given wheel appears to vary

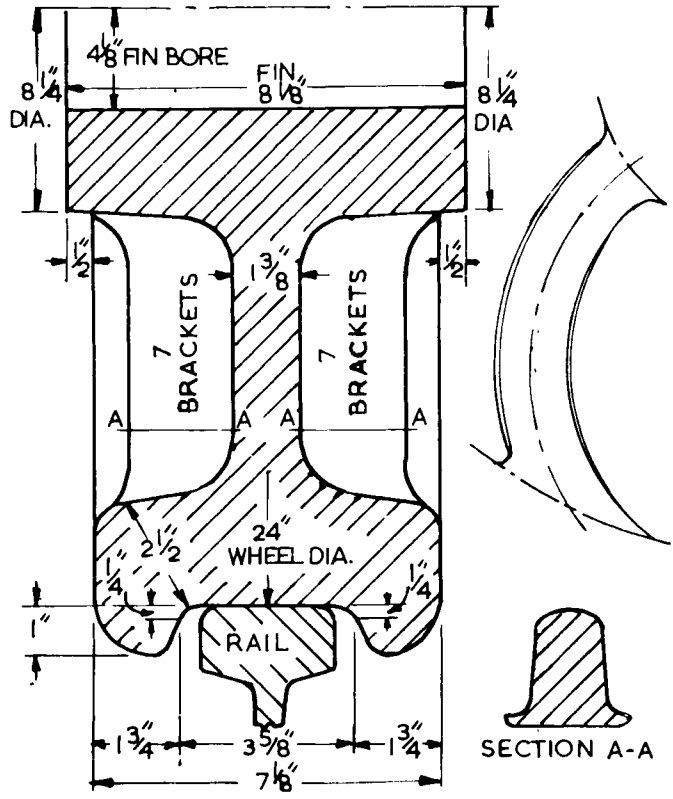
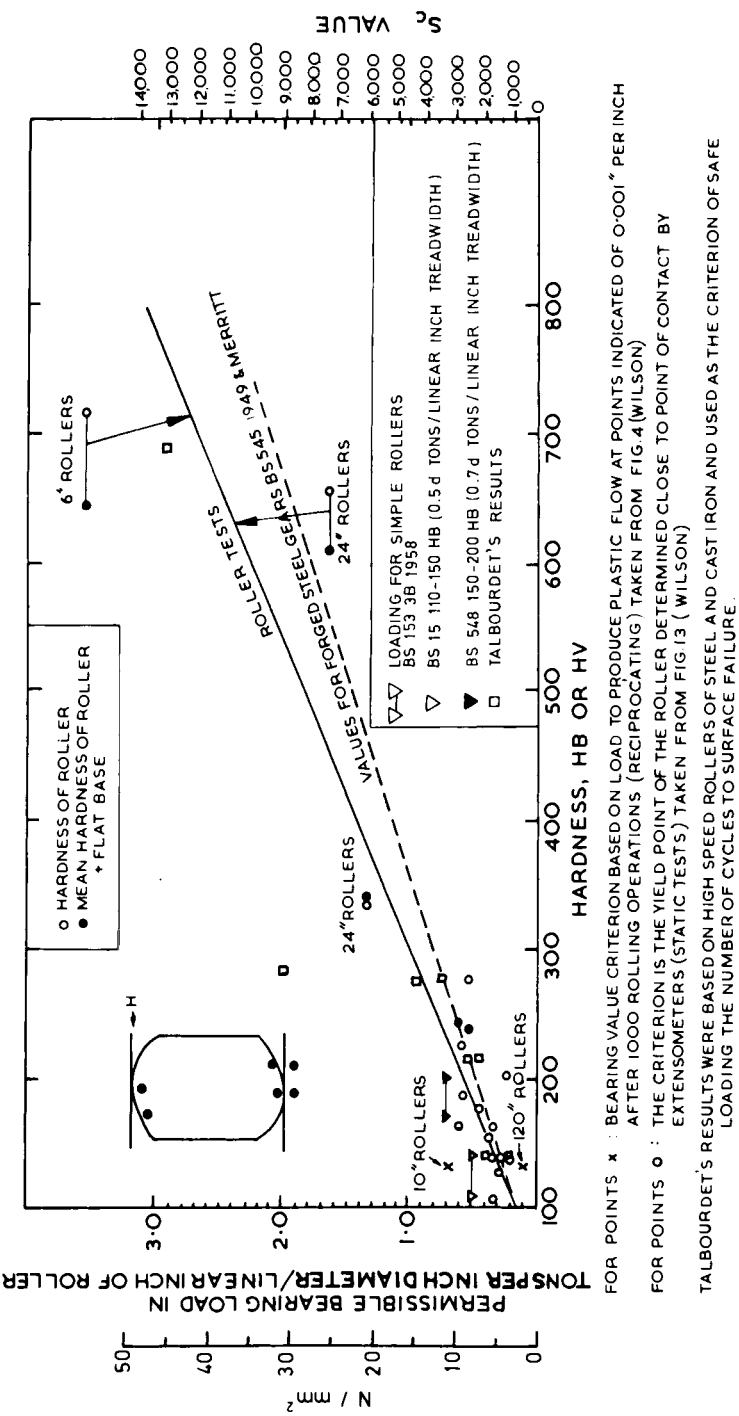


Figure 136. 24 in, 640 lb, double flange chilled iron crane wheel, single plate type, for maximum vertical load 34 000 lb, maximum safe flange pressure 25 000 lb

directly as the hardness, as shown in *Figures 131* and *137*, rather than as the square of the hardness. Lines 2 and 3 (*Figure 131*) are taken from Talbourdet's results in pure rolling of cast iron *v.* cast iron, and cast iron *v.* steel. Results by Wilson[7] on steel rollers (120 HB to 700 HV) for bridge construction straddle these curves (*Figure 137*). Line 6 (*Figure 131*) which indicates the relationship between S_c values and hardness for gears (sliding and rolling) from BS 545:1949, falls within the same region.

The values for cast iron wheels of varying hardness on flat steel rails given in *Table 175* are based on Talbourdet's results (*Figure 131*, line 2).



FOR POINTS x : BEARING VALUE CRITERION BASED ON LOAD TO PRODUCE PLASTIC FLOW AT POINTS INDICATED OF 0·001" PER INCH AFTER 1000 ROLLING OPERATIONS (RECIPROCATING) TAKEN FROM FIG.4 (WILSON)

FOR POINTS o : THE CRITERION IS THE YIELD POINT OF THE ROLLER DETERMINED CLOSE TO POINT OF CONTACT BY EXTENSOMETERS (STATIC TESTS) TAKEN FROM FIG 13 (WILSON)

TALBOURDET'S RESULTS WERE BASED ON HIGH SPEED ROLLERS OF STEEL AND CAST IRON AND USED AS THE CRITERION OF SAFE LOADING THE NUMBER OF CYCLES TO SURFACE FAILURE.

N.B THESE FIGURES INDICATE THAT PERMISSIBLE BEARING LOAD VARIES DIRECTLY WITH BRINELL HARDNESS

Figure 137. Safe bearing loads of steel and cast iron rollers

Table 175 MAXIMUM AND SAFE LOADINGS FOR WHEELS AND ROLLERS OF DIFFERENT HARDNESSES

Hardness: kg/mm²		*100–140*	*140–180*	*180–220*	*220–260*	*260–300*	*450–550*
(1) S_{max} (lb/in²) for point contact on crowned rails based on chilled wheel data and extrapolated for hardness	lb/in²	84 500	113 000	141 000	169 000	197 000	352 000*
	kg/mm²	59·4	79·5	99·0	118·8	138·5	247·5
(2) Probable safe value of calculated S_{max} on crowned steel rails in lb/in²	lb/in²	43 000	58 000	72 000	86 000	100 000	180 000†
	kg/mm²	30·2	40·8	50·6	60·4	70·3	126·4
(3) Line contact on flat steel rails: limiting S_c values in lb/in of line contact, based on Talbourdet (*Figure 131*, Curve 2) 10^8 wheel revolutions; *see* p. 415	lb/in	700	1200	1700	2300	3100	(8500)
	kg/mm	12·5	21·4	30·3	41·0	55·2	152·0
(4) Equivalent S_{max} values	lb/in²	48 800	59 880	67 170	97 000	112 800	211 000‡
	kg/mm²	34·2	42·0	47·1	68·2	78·4	148·0

* Based on Lyndon and Vial's value of 200 000 lb load on 33 in chilled iron wheels on crowned rails. At this load no permanent set of the chilled tread occurs.
† Based on 26 250 lb normal load on chilled iron rail wheels 33 in diameter on 8 in and 12 in crowned rails[3] where $S_{max} = 180\,000$ lb/in² and $S_c = 6200$.
‡ $S_{max} = 0{\cdot}418 \times \sqrt{(S_c \times E_e)}$; E for HB $500 = 30 \times 10^6$ and $E_e = 30 \times 10^6$; E for HB $240 = 18 \times 10^6$ and $E_e = 23{\cdot}5 \times 10^6$; E for HB $120 = 12 \times 10^6$ and $E_e = 17{\cdot}1 \times 10^6$; E for steel $= 30 \times 10^6$ lb/in².
Conversion: 1 lb/in² = 0·000 703 kg/mm² : 1 kg/mm² = 1422·3 lb/in² (S_{max} values)
1 lb/in line contact = 0·0178 kg/mm line contact; 1 kg/mm line contact = 56 lb/in (S_c values)

Note: The limiting S_{max} values for crowned rails are approximately HB/2 kg/mm² and flat-topped rails are approximately HB/3·5 kg/mm². These values are of the same order as HB/3 kg/mm² quoted by Burwell and Strang[9] as the surface stress above which surface breakdown accelerates rapidly. *See* section on cams and tappets, p. 422.

The extrapolation to 500 HB (chilled iron wheels) gives a value of $S_c = 8500$ (4250 D per inch tread width) which is somewhat higher than that used with chilled iron railway wheels ($S_c = 6200$). This value applies to pure rolling and takes no account of the risk of spalling of the chilled tread if severe braking causes dry sliding to occur.

For nominal point contact on crowned rails (radius 8–12 in) it is therefore reasonable to assume that the Brinell hardness is a good index of the ability of a surface to carry a load, and the following information is available on three ranges of hardness to which maximum Hertzian loadings have been related:

(1) Vial[5] has shown that a 33 in diameter chilled wheel of about 500 Brinell has limiting load capacity of about 200 000 lb, without permanent deformation, on a 12 in radius crowned steel rail. This is equivalent to a maximum Hertzian stress of approximately 352 000 lb/in^2. The normal maximum load on a 33 in chilled railway wheel[3] on a 12 in radius crown rail is 26 250 lb ($S_{max} = 180\,000$ lb/in^2).

(2) A grey iron wheel with a Brinell of 250 may therefore be tentatively assumed to be at its limit with a maximum Hertzian load of 176 000 lb/in^2. Grey iron wheels of 14 in diameter with a surface hardness of 220–260 Brinell are known to give indefinite life at low speeds on 8 in radius crowned steel rails with loads of $4\frac{1}{2}$ ton per wheel, equivalent to calculated maximum Hertzian loads of 144 000 lb/in^2.

(3) A grey iron wheel with a Brinell of 110 might be expected to be at its limit with a maximum Hertzian stress of 43 000 lb/in^2. Soft grey iron wheels of 14 in diameter with a surface hardness of 105–125 Brinell have been known to fail by tread disintegration on 8 in radius crowned steel rails with loads of $2\frac{1}{2}$ ton per wheel, equivalent to a calculated maximum Hertzian stress of 97 000 lb/in^2.

The limiting and safe Hertzian stresses versus crowned and flat steel rails for cast iron wheels of varying hardness appear to be of the order shown in *Table 175*.

The figures given for probable safe values of maximum Hertzian stress with point loading (crowned rails) are very close to the maximum compressive strengths of these irons. Any plastic deformation of rail or wheel, however, would very materially reduce the calculated and, of course, the true stresses.

The S_{max} value of 180 000 lb/in^2 used in American practice for chilled iron railway wheels is much the same as the maximum used in chilled iron cam and tappet practice where the contact is in pure lubricated sliding, *see* p. 425.

The figures given for safe values of S_c (lb/in of line contact) when multiplied by the wheel radius (*not* the diameter) may therefore be regarded as giving the safe loading for grey iron wheels on crowned rails, and give an ample margin of safety against tread disintegration for flat-topped rails with line contact greater than 1 in.

The correspondence between Section 2 (*Table 175*) based on chilled iron railway wheel practice (USA) and Section 4, based on Talbourdet's experimental results for grey iron is sufficiently close to give reasonable confidence in the results.

The line contact loadings on flat steel rails or tracks given above may be more conveniently expressed as load (tons or pounds) per inch of tread width per inch of wheel diameter and the following figures may be more readily used. The values are numerically one-half of those given for the S_c value in Section 3 (*Table 175*) as they are based on diameter and not radius.

The figures in *Table 176* are approximate and include S_c values given for gears by BS 545:1949. The values are based on flat-topped rails and the higher values apply in pure rolling, and the lower values with up to 30 per cent sliding motion. Pure rolling on crowned rails (radius 8–12 in) probably falls in the upper region of the values quoted.

For bridge rollers the maximum load is 0·45 d to 0·7 d tons per inch of line contact for cast iron to BS 1452, Grade 23 (HB > 220). For chilled iron crane wheels the maximum loading has been given as 672d pounds per inch tread width[1] which gives a large factor of safety as far as wheel tread is concerned.

While the tensile and transverse bending strengths of white and grey irons are comparable, the bend deflection and impact strengths of most grey irons of 160 Brinell or more are greater than for white iron. The flange dimensions given in *Table 174*, therefore, would be quite adequate for grey iron as far as strength and impact resistance are concerned. The resistance of grey iron flanges to abrasion, however, is markedly less than white iron, and grey iron wheels should, therefore, be used where speeds and side loadings are reasonable.

Table 176

Hardness (HB)		100–140	140–180	180–220	220–260	260–300	450–550
S_c value		600–700	1000–1400	1300–1800	1450–2400	1700–3100	3000–8500
Load per inch width per inch wheel diameter d	lb:	300–350d	500–700d	650–900d	725–1200d	850–1550d	1500–4250d
	ton:	0·13–0·16d	0·2–0·3d	0·3–0·4d	0·3–0·5d	0·4–0·7d	0·7–1·9d

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FURNACE WHEELS AND ROLLERS AT HIGH TEMPERATURES

Wheels and rollers are employed in continuous heat treatment furnaces at temperatures up to 850°C and present problems both on the wheel tread and the bearings which must operate without lubrication. The stresses on the wheel tread which can be used in service are most readily based on room temperature properties as described previously.

Some service experience has been gained with integrally cast rollers and spindles in 30–33 per cent chromium iron in which the roller diameter was 6 in, the tread width 1·75 in and the two bearing spindles were 1·0 in diameter × 1·25 in long. The roller tread operated in pure rolling and the spindles in rotational sliding. The rollers were cast to a composition of TC 2·8 per cent, Si 3·33 per cent, Mn 0·48 per cent, Cr 32·6 to 33·8 per cent and had an as-cast hardness of 373 HB. Hot hardness values (HV 20) of a similar iron were 600°C = 79 HV; 800°C = 34 HV; 900°C = 25 HV. The hot hardness was therefore low.

The tread loadings were 50–100d pounds per inch tread width (d = diameter in in), i.e. S_c 100–200, and S_{max} 23 000 to 32 000 lb/in^2.

The high chromium iron spindles operated in high phosphorus austenitic cast iron bearings without lubrication. Bearing analysis TC 2·93 per cent, Si 2·28 per cent, Mn 0·42 per cent, S 0·024 per cent, P 1·01 per cent, Ni 14·9 per cent, Cr 1·95 per cent, Cu 5·8 per cent with as-cast hardness of 148 HB with hot hardness (HV 20), 600°C = 90 HV; 800°C = 35 HV; 900°C = 30 HV. The bearing i/d was 1·125 in and contact length 1·25 in × 2 and the S_c value 45–55.

The S_c and S_{max} values are based on room temperature properties.

These rollers and bearings were still operating successfully after 6 months at a temperature of 800–830°C on rails of heat resisting steel.

It may be significant that austenitic steel roller spindles and bearings tended to seize although their hot hardness was higher than that of the 30 per cent chromium iron spindles (ferritic) or the high phosphorus austenitic cast iron bearings which showed no sign of seizure or wear after 4 to 5 times the life of austenitic steel spindles and bearings at the same loading. It may therefore be desirable under these conditions of unlubricated sliding wear to use ferritic material for one component and austenitic for the other to avoid seizure. The ability of these materials to withstand pure sliding at 800–830°C in air is remarkable and appears to be due to the formation of a highly polished skin rather than to the maintenance of high hot hardness.

FLYWHEELS

There seems to be little experimental work published on actual bursting stresses in iron flywheels since that carried out by Benjamin[1] in 1899. The validity of this work has never been questioned and the conclusions drawn

are still valid for spoked, built-up flywheels, in which it is regarded as undesirable to exceed rim speeds of 60 to 100 ft/s (rim stress, 0·43 ton/in^2).

In automobile and diesel engine practice, however, the great improvement in casting technique, and the increased strength of the irons available makes very much higher stressing possible. The stresses built up in a flywheel rim during rotation depend almost entirely upon the rim velocity and not upon the radius of the wheel.

The nominal hoop stress in the rim of the flywheel rises as the square of the rim velocity, and is given by the fomula[2]:

$$f = \frac{12WV^2}{g} \text{ lb/in}^2 = 0{\cdot}000\,043\,V^2 \text{ ton/in}^2 \text{ for cast iron}$$

where W = density of metal, lb/in^3 (for cast iron = 0·26 lb/in^3)
V = velocity of rim, ft/s
g = 32·2 ft/s^2

Petrol engine manufacturers using unspoked cast iron flywheels normally regard a rim velocity of 230–250 ft/s as the maximum, and this is equivalent to a hoop stress of 2·3–2·7 ton/in^2.

Diesel engine manufacturers using cast iron flywheels normally operate at 150–180 ft/s, which is equivalent to a nominal hoop stress of 0·97–1·4 ton/in^2. In an actual flywheel test, failure occurred at a rim speed in excess of 430 ft/s (calculated rim stress of more than 8 ton/in^2). The tensile strength of the specimen cut from the rim was 9·2 ton/in^2, equivalent to a rim speed of 463 ft/s which gives good agreement.

The margin of safety provided by these limiting speeds is between 8:1 and 3:1 in the case of a Grade 12 iron in a 4 in section where the minimum actual tensile strength is about 8 ton/in^2. Most flywheels for automobiles and diesel engines are made in Grade 17 iron which, in a similar section, would give approximately 12 ton/in^2, so that the factor of safety at 250 ft/s rim speed would be over four to one.

The presence of casting stresses may modify these factors to some extent. There is a marked tendency, particularly in spoked wheels for the spokes to retain substantial tensile cooling stresses and for the rim to be in compression. Although, therefore, the compressive stress in the rim increases its own factor of safety, as the wheel speed increases and the initial compressive stresses in the rim are counterbalanced by the rotational hoop stresses, the release of compressive strain in the rim will increase the tensile stress in the spokes. This uneven distribution of stress will cause irregularity in the rim as the speed increases. High speed flywheels should therefore be made with continuous webs rather than spokes, and should preferably be stress-relieved.

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DIE BLOCKS FOR COLD FORMING AND HOT FORMING

COLD FORMING DIES

Die blocks for deep drawing and pressing require sufficient strength to resist the stresses imposed by deformation and flow of the sheet metal over the surface and sufficient hardness and resistance to seizure to avoid surface deformation and pressure welding where metal is forced to flow over the surface.

A wide variety of materials can be employed, depending upon the severity of the draw, the metal to be drawn and the number of components required from a given die. Die materials range from wood and plastics, cast iron, bronze, hardened steel to carbide faced tools.

Cast iron dies are particularly useful for cold forming[1] where the drag of the sheet is not excessive over small radius corners since the graphite cavities carry lubricant and reduce the risk of seizure and tearing as the sheet flows over the surface. The surface of the die should be fully pearlitic with a hardness not less than 190–200 HB. With heavy castings the as-cast surface skin may be partially decarburised and appreciably softer than the main body. A machining allowance of $\frac{5}{32}$ in (3·97 mm) should remove this surface layer. Since high graphite content lowers hardness even with a fully pearlitic structure, the total carbon value should be kept below 3·3 per cent to 3·4 per cent in unalloyed irons.

The radius of a cast iron die corner over which sheet is drawn is usually 6 to 8 times the thickness of the sheet being drawn. Cast iron dies usually fail either by attrition, or if the iron is soft and the pressures high, by plastic deformation. The harder the iron the greater is its resistance to both forms of failure and for severe duty the irons may be oil hardened or inserts of hardened tool steel may be used. Most shells can be drawn in one operation to a depth of 4 to 6 times the corner radius, but increased clearance between punch and die is required in the corner regions owing to thickening of the blank.

Cast iron dies can be used for steel sheet at least up to 0·14 in (3·56 mm) thick. In many cases it is customary to make tools for sheet metal forming, extruding, pressing, forging and so on, from tool steels. These materials, which range from the normal eutectoid carbon, non-shrink steels, to the more expensive, highly alloyed steels are basically expensive, and the cost of machining from solid blanks is often quite considerable. The use of these special steels, in many instances, is not necessary, and cast iron can be employed, with the associated advantages of: (1) cheapness; (2) ease of machinability; (3) ability to be cast roughly to shape.

The following grades of cast iron cover many of the requirements of the press shop.

HIGH PHOSPHORIC IRON

Because of the presence of the dispersed phosphide eutectic with hardness of about 420–740 HV and its resistance to lubricated sliding wear, this material is very suitable for dies where foundry techniques can produce sound castings. It is not suitable for flame hardening.

This material is available from some foundries, either in stock bar form or cast closely to shape to reduce machining. The material is comparatively low in strength [9–12 ton/in^2 (139–185 N/mm^2)], but has a hardness HB ranging from 180–240 in the as-cast state, depending on section. A composition of an actual die block [overall diameter 13 in (330 mm) internal diameter 6 in (152 mm), wall thickness $3\frac{1}{2}$ in (89 mm)], which gave very satisfactory results in service is:

	%
Total carbon	3·2
Silicon	2·1
Manganese	0·4
Sulphur	0·12
Phosphorus	1·5
Nickel	0·03
Chromium	0·06

This phosphorus content is exceptionally high and would not be normally regarded as acceptable for foundry reasons. The example indicates, however, that where high phosphorus irons are readily available they can give results comparable or better than alloyed irons in the as-cast condition provided the foundry technique can produce sound castings. Suitable ranges of composition for various sections in high phosphoric irons are given below:

	Section		
	1 in	*2* in	*3* in
Total carbon, %	3·0–3·3	3·0–3·3	3·0–3·3
Silicon, %	1·8–2·5	1·5–2·1	1·0–1·6
Manganese, %	0·5–0·9	0·5–0·9	0·5–0·9
Sulphur, %	<0·12	<0·12	<0·12
Phosphorus, %	0·7–0·9	0·7–1·0	0·7–1·0

In most cases the precise chemical composition is of much less importance than the soundness and close grain of the machined surface, and these figures are to be regarded only as a guide.

This is the cheapest grade of iron that can be employed, and is frequently used for very large die blocks employed in the pressing of automobile bodies. Its wear-resisting properties are excellent because of the large amount of phosphide eutectic present. Tools from this material may require external reinforcement if bursting loads are great.

LOW PHOSPHORIC IRONS

These irons should preferably contain silicon ranging from 1·0 to 1·5 per cent for sections above $1\frac{1}{2}$ in and 1·5 to 2·0 per cent for sections less than $1\frac{1}{2}$ in. They are also particularly suitable for dies in which very high surface pressures are not developed, or for which subsequent flame hardening or full heat treatment is required. Such dies are likely to be softer (170–200 HB as-cast) than the high phosphorus dies, but will show a greater toughness and bursting strength.

Additions of 0·1 per cent tin or 0·07 per cent antimony will produce fully pearlitic structures in unalloyed irons with a corresponding rise in hardness from 10 to 20 points. Such irons will respond to flame hardening as well as irons containing nickel, chromium or molybdenum, but depth of hardening with full heat treatment and quench will be lower. If more tin or antimony is used than is required for complete ferrite suppression some loss of tensile or impact properties may occur.

ALLOY IRONS

The following figures have been quoted as particularly suitable for dies and die blocks:

	A1	*B1*	*C*
Total carbon, %	3·0–3·3	2·9–3·2	2·6–3·5
Silicon, %	0·8–1·6	1·0–1·5	1·2–2·2
Manganese, %	0·5–1·0	0·7–1·0	0·6–0·9
Sulphur, %	0·12 max.	0·12 max.	0·12 max.
Phosphorus, %	0·3	0·3	0·25 max.
Nickel, %	1·2–1·75	2·5–3·5	—
Chromium, %	0·4–0·8	0·6–1·0	0·35–0·50
Molybdenum, %	—	—	0·35–0·50
Hardness, HB, as-cast	200/220	250/280	200/250
Tensile strength as-cast ton/in^2	16–19	18–20	16–19
(N/mm^2)	(247–293)	(278–309)	(247–293)

These irons are mainly used in the as-cast condition but may be locally hardened by induction or flame hardening techniques if so desired.

The object of the alloy additions is to provide a fully pearlitic matrix, therefore minimising scoring and 'pick up' in service. A phosphorus content greater than 0·15 per cent may lead to the formation of free carbides which will improve the wear resistance of the blocks at the expense of machinability, particularly at the finishing stages.

NICKEL-MOLYBDENUM CAST IRONS

These irons are particularly suitable for heavier duty work and have given excellent results for shell nosing dies for hot forging. There are two distinct ranges of molybdenum-bearing irons that may be used.

(a) Pearlitic—The pearlitic material gives hardnesses ranging from 220 to 240 HB, and tensile strengths from 17 to 22 ton/in^2 in sections of up to 2 in to 3 in. A typical composition is:

	%
Total carbon	2·9–3·1
Silicon	1·8–2·0
Manganese	0·7–0·9
Sulphur	Below 0·12
Phosphorus	Below 0·1
Nickel	1·0–1·5
Molybdenum	0·5–0·8

(b) Acicular molybdenum-bearing iron—This material has strengths ranging from 20 to 30 ton/in^2, depending on section, and hardness ranging from 250 to 320 HB. For sections between $1\frac{1}{4}$ and $2\frac{1}{2}$ in a suitable composition is:

	%
Total carbon	2·9–3·1
Silicon	1·4–1·6
Manganese	0·8–1·0
Sulphur	Below 0·12
Phosphorus	0·05
Nickel	1·4–1·6
Molybdenum	1·5–1·7

For sections of about 6 in (152 mm) the nickel should be raised to 3 per cent.

GREY MACHINABLE MARTENSITIC IRON

This iron has a hardness as-cast of 350–420 HB, but as it contains no structural constituent of very high hardness (e.g. free carbide) it can be machined with suitable tools and cutting speeds. A suitable range of composition is:

	%
Total carbon	3·0–3·4
Silicon	1·0–1·5
Manganese	0·7–1·0
Sulphur	0·12
Phosphorus	Below 0·2
Nickel	5·0–6·0
Chromium	0·5–1·5

Hardness, HB, as-cast	380–420
Tensile strength, as-cast, ton/in^2	19–26
(N/mm^2)	(293–402)

IRONS SUITABLE FOR HEAT TREATMENT

Small simple dies in any of the low phosphorus unalloyed irons or alloyed irons containing nickel and chromium, can be heat treated by oil quenching from a temperature of 850 to 870°C, to give hardnesses ranging from 350 to 450 HB. Hardening treatment may be followed by tempering between 400 and 550°C, depending upon the hardness required.

The quenched unalloyed irons will give hardnesses from 350 to 450 HB, depending on section, but alloyed irons should give hardnesses ranging from 420 to 500 HB when quenched.

The following die composition is quoted[1] as particularly suitable for heat treatment:

	%
Total carbon	2·9–3·2
Silicon	1·2–1·6
Manganese	0·7–1·0
Sulphur	0·12
Phosphorus	Below 0·15
Nickel	2·5–4·0
Chromium	0·5–1·0

Such an iron will give a hardness on heat treatment of the order of 300–350 HB, and a maximum tensile strength ranging from 24 to 26 ton/in^2 (371 to 402 N/mm^2).

Stress relief For complex dies, stress relief heat treatment may be desirable before machining.

Flame hardening This can be successfully carried out on all the compositions quoted for pearlitic cast irons, but is not recommended for the high phosphorus-cast irons (more than 0·4 per cent phosphorus) owing to the likelihood of surface pitting.

NODULAR CAST IRON

Where heavy stresses must be withstood, pearlitic nodular cast irons have proved extremely satisfactory, both for small dies, where accuracy of form and long life are required, and for heavy bolsters, in which the stresses involved are too great for normal cast iron. This material is probably best supplied in the as-cast or stress-relieved state, giving hardnesses from 200 to 260 HB, and a tensile strength from 40 to 55 ton/in^2 (618 to 849 N/mm^2). In this form, different combinations of strength, toughness and wear resistance can be obtained by suitable heat treatment.

Glass bottle moulds These moulds or dies are subject to exceptionally severe conditions[2] and the material must possess certain combinations of properties to be successful:

(1) Machinability.
(2) Good surface finish.
(3) Resistance to surface heat checking and major cracking (thermal shock).
(4) Resistance to growth and oxidation.
(5) Resistance to wear.
(6) High thermal conductivity.

High-carbon cast iron has a high thermal conductivity and a low modulus of elasticity giving marked resistance to thermal shock, but may have a coarse, open-grained surface with a poor initial finish and a tendency to deep oxidation, together with poor wear resistance. Surface densening, however, can overcome these disadvantages.

The addition of alloys such as nickel, chromium and molybdenum, may increase the strength, hardness and growth resistance, but will reduce thermal conductivity, and appears to increase the risk of cracking.

Nodular cast irons have high thermal shock resistance and good oxidation resistance, but have a slightly lower thermal conductivity than grey irons, and if the nodules of graphite are large, the surface finish will be affected.

Grey cast iron A high carbon, unalloyed grey cast iron with a densened surface giving undercooled graphite. A typical composition is:

	%
Total carbon	3·5–3·7
Silicon	1·6–2·3
Manganese	0·45–0·65
Sulphur	<0·12
Phosphorus	Up to 0·35

The surface should be densened to give a fine surface finish. Full annealing, to give a ferritic structure, will provide the best resistance to cracking but annealing to give a spheroidised pearlite appears to give the best combination of wear resistance and resistance to cracking. An addition of 0·15–0·25 per cent titanium, particularly if CO_2 or any other inert gas is bubbled through the melt, will ensure the complete production of undercooled graphite, which will markedly improve the surface finish but may slightly increase the risk of major cracking.

Nodular cast iron Moulds of this material, made by any process, have shown a promising increase in life due to their high thermal shock resistance, particularly in the ferritic state, and good oxidation resistance. They tend to run slightly hotter on account of their lower thermal conductivity, and in heavy sections coarse graphite nodules may give difficulty with surface finish, but where these disadvantages can be tolerated or overcome, nodular cast iron moulds have shown a significant improvement in life.

Hot forming Dies for hot forming such as shell nosing have been successfully made in cast iron where the speed of operation ensures that the die does not exceed 400–460°C. Acicular iron, containing TC about 3·0 per cent, Si 2·0 per cent, Mn 1·0 per cent, P 0·08 per cent, Ni 2·2 per cent and Mo 0·95 per cent heat treated at 350°C for 7 h, furnace cooled, tensile 21–26 ton/in^2 (324–402 N/mm^2) and HB 290–310, has given satisfactory life in forging steel at 1050–1100°C. Die temperatures did not exceed 420°C. Pearlitic nodular iron, normalised for 2 h at 900°C, cooled to 870°C and air cooled giving tensile 55–60 ton/in^2 (849–927 N/mm^2), elongation 3% and HB 300–320 also gave satisfactory life with maximum die temperature of 460°C.

Martensitic white iron of the Ni-Hard type has also been employed for hot drawing dies for steel tube.

Dies for drop stamping For drop stamping particularly on non-ferrous sheet, grey cast iron dies operating with a 'force' or impact die of lead or white metal give good service if the impression is not too intricate and the surface hardness is above 200 HB. Pearlitic nodular iron dies with hardness up to 300 HB are very suitable but if intricate patterns are involved, a higher hardness may be required.

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INTERFERENCE AND SHRINKAGE FITS

There are three typical methods of making fits:

(1) To bore the hole slightly smaller than the shaft which is to be fitted, and press the shaft through hydraulically, both components being cold. This method is frequently used when fitting steel shafts to wheel hubs.
(2) To bore the hole in the component smaller than the size of the shaft, and to heat the component so that it will slide over the shaft and contract on to it when cooling. This method is often used where the length of shaft in contact with the component is great.
(3) To freeze the centre component in liquid air or solid CO_2 so that it contracts to a diameter smaller than that of the external component and then to assemble the two so that when the temperature is restored to normal the central component expands into position. This method is particularly common for cylinder liners.

In all the above cases, it is important that the interference fits do not impose bursting stresses upon the outer shell greater than it can withstand. The interference is usually expressed as a proportion of the diameter of the shaft, e.g. 0·001 in per in diameter of shaft, and the greater the interference, the greater the hoop stress which is applied to the outside shell. The hoop stresses in-

volved are also partly dependent on the thickness of the surrounding shell, and *Figure 138* shows a convenient method of calculating the stresses involved, based on the assumption that the elastic modulus for steel is 30×10^6 lb/in.^{2} and for cast iron is 15×10^6 lb/in.^{2} This figure would be obtained with an iron giving approximately 12–15 ton/in^2 tensile strength.

In the diagram, a = ratio of wall thickness of ring or hub to diameter of shaft. Thus, if the wall thickness of a hub of cast iron is 0·4 times the bore,

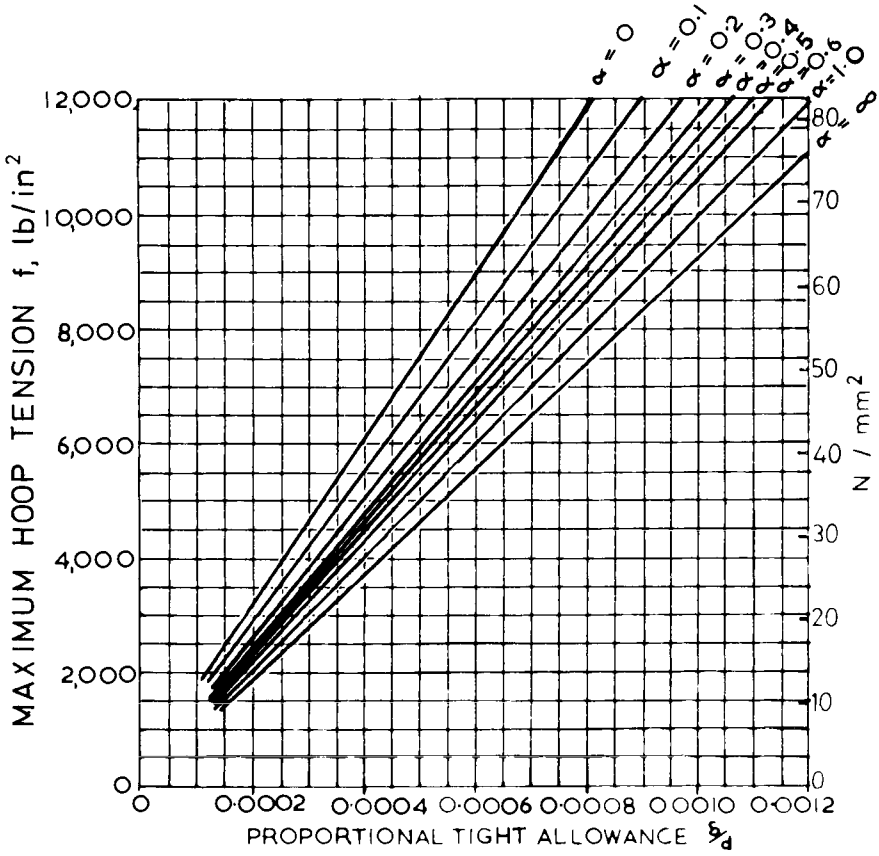


Figure 138. Stresses produced by force fits. Cast iron hub on steel shaft

the interference allowance on the diameter of shaft which is required if the bursting stress is not to exceed 3000 lb/in^2 can be seen from the curve to be 0·000 27 in per in diameter of shaft[1]. It is desirable to allow a factor of safety of at least 4:1 between the tensile strength of the casting and the calculated stress.

Heating the outer shell to 350–440°C is sufficient to expand the outer member by approximately 0·004–0·005 in per in of diameter, which is usually sufficient to allow adequate clearance beyond the interference allowance.

These calculations assume that close contact is maintained between the

two surfaces and make no allowance for the collapse of the tips of machined ridges* so that in heavy engineering practice the stresses calculated may be higher than those that actually occur in practice. For example, in the assembly of chilled iron railway wheels on to steel axles, which is done cold, the wheels are bored 0·001 to 0·0015 in per in bore diameter less than the axle diameter, and an hydraulic force of approximately 1 ton per in diameter per in length is applied axially to force the axle into the wheel[2]. With a ratio of hub thickness to shaft diameter of 0·5, the figure of 0·0015 in per in interference quoted gives a nominal hoop stress of 7·6 ton/in^2 according to the diagram, on the assumption that the machining is sufficiently fine to ensure complete contact over the whole surface.

Tests have been carried out on chilled iron railway wheels (grey centre, tensile 11·2–14·3 ton/in^2, Brinell approximately 240) which were pressed on to steel shafts, the mounting surfaces being turned, with an interference of about 0·0014 in per in diameter—an actual interference of 0·01 in for 7 in shaft diameter. On first dismantling, the measured interference was reduced by 0·002 to 0·005 in, partly by smoothing the surface and partly by permanent set in the bore. Of this, approximately 0·001 in was stated to be due to smoothing of surface roughness. There was very little subsequent loss of interference by repeated mounting and dismounting, but in spite of the reduced interference, the actual mounting pressures were from 28 to 41 per cent greater than for the initial mounting.

When similar tests were carried out with ground instead of turned surfaces, the mounting pressures were approximately doubled, but the measured hub strains only increased by 4 to 20 per cent and were not proportional to the mounting pressures.

It was estimated that the tensile stresses produced by interference fits of this order with thick hubs were less than 7·8 ton/in^2, so that the agreement with the figures derived from *Figure 138* is quite good.

The figure of 0·0015 in can therefore be regarded as the maximum safe interference fit which can be employed where the ratio of hub thickness to shaft diameter exceeds about 0·3 to 0·5 and the actual strength of metal in the hub is not less than about 12 ton/in^2. For thinner hub thicknesses it is probably wise to limit interference to a maximum of 0·001 in per in of shaft diameter, with nominal stresses less than 8000 lb/in^2 or one quarter of the tensile strength.

Dismantling pressures are from 1·6 to 2·0 times the assembly pressures[2, 3]. With honed surfaces dismounting force may be more than doubled[3].

Interference fits quoted in engineering practice range from 0·0005 in to 0·0024 in per in diameter for steel/cast iron, and from 0·0005 in to 0·0034 in per in diameter for steel/steel[4]. Some plastic deformation occurs with steel/steel interferences above about 0·001 in, and Lloyd's Register of Shipping rules give limits of 0·001 to 0·0018 in per in for marine cranks, which permits a small amount of plastic yield when both components are of steel[3]. There seems little doubt that with commercial rough machining, the maximum safe interference is 0·0015 in per in with very thick cast iron hubs, and the thinner the hub and the finer the surface finish, the smaller is the safe interference.

*In some cases for shrink fits surface finishes of 16 CLA are employed

Insufficient interference, however, may involve slipping between the two surfaces, and if cyclic variations occur, as in shaft and flywheel, fretting corrosion may take place between the surfaces. Fretting between components assembled by interference increases the liability to fatigue failure, which may commence from the fretted area[5, 6].

Where one component of a shrunk assembly has a markedly different modulus of elasticity from the other, the application of heavy rolling local stress may cause movement between the two faces. For example, a steel ring shrunk on to a cast iron centre, as with a steel tyre or gear ring, may show axial or circumferential movement between the two components when the assembly has been in service. Conventional British marine engine practice, however, has for many years successfully used shrink fits of 0·0005–0·000 75 in per in for forged steel gear rings shrunk on to grey iron centres and dowelled axially at the junction of the ring and the centre. Interference fits of 0·0003 in per in with radial dowelling may allow slipping to occur. Surface finish of the mating surfaces is about 70–100 μinches CLA. One advantage of the grey iron centre appears to lie in its high damping capacity[7].

Similarly, if a thick cylindrical cast iron roller is shrunk on to a steel shaft, with an interference of 0·0005 in per in of shaft diameter, and submitted to prolonged and heavy rolling pressure, it may move axially along the shaft[8], unless provision is made for machining a raised ring on the shaft to mate with a corresponding groove in the roller. The depth of the ring must obviously be less than the increase in radius when the roller is heated.

Where one component has a markedly different coefficient of thermal expansion from the other, changes of temperature will markedly affect tightness of the fit, and under stress may permit movement to occur.

Temperature differences between two components of similar expansion characteristics can also loosen interference fits or cause seizure. Thus, seizure between piston and cylinder (which may occur if the piston temperature rises above the cylinder temperature with insufficient initial clearance) is a form of interference which is normally allowed for in cast iron pistons and liners with a clearance of 0·001 to 0·0015 in per in diameter in the centre of the piston wall, and up to 0·01 in per in diameter at the piston crown where the temperature rise is greatest.

Dry liners in internal combustion engine practice use interference fits from 0·0020 in per in diameters for cylinders 1·125–1·5 in diameter to 0·000 65 in per in for cylinder $10\frac{1}{2}$–$12\frac{1}{2}$ in diameter (BS 3525: 1962).

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8

Cast iron beams and columns

BEAMS

A great deal of work has been done on this subject. In particular, the varying forms of section have been studied and the anomalies in their response to stress have been discussed. This work has indicated the difficulty of estimating the breaking stress of a given section when making use of conventional formulae, owing to the non-elastic response to stress, the variation between tensile and compressive properties and the shift of the neutral axis towards the compression side of a stressed iron beam.

Work up to 1948 has been summarised by Grant[1]. Since that date further work has been carried out to elucidate and, if possible, to simplify our understanding of the problem.

The following figures (*Table 177*) have been given[2,3] for the shift of the neutral axis to the compression side in rectangular beams, expressed as a percentage of a beam depth under given percentages of the ultimate load (four-point loading).

Table 177

Approx. tensile strength		*Beam size and span*		*Percentage of ultimate load*					
ton/in²	(N/mm²)	in	(mm)	*20%*	*40%*	*53%*	*75%*	*80%*	*100%*
10	(154)	2·2 × 1·4 × 24	(56 × 36 × 610)	—	—	3·4	4·6	—	6·5
24·9	(385)	4 × 2 × 48	(102 × 51 × 1219)	0	0·4–1·2	—	—	3·2–4·0	—
29·5	(456)	2½ × 1 × 33	(64 × 25 × 838)	0	1·0–1·2	—	—	2·7–2·8	—

The transverse rupture stress (maximum fibre stress) for round test bars calculated from the conventional formula $f = M/Z$, where f = maximum stress in tension and compression, M = bending moment applied to the beam and Z = modulus of section, gives figures from 1·6 to 2·0 times the actual tensile strength. In 1951 Gilbert[2] obtained results for an iron giving a tensile of approximately 10 ton/in² (154 N/mm²) in a 1·2 in (30 mm) bar. Using a beam

of rectangular section and electrical resistance strain gauges, he found that over a stress range 7–30 per cent of the ultimate breaking load, the calculated stresses were about 20–25 per cent in excess of the true stresses.

Smalley[3] and Kuivinen have indicated that with irons giving 24–29·5 ton/in^2 (371–456 N/mm^2) tensile in a 1·2 in (30 mm) bar, at stress levels of up to 30 per cent of the ultimate tensile, the standard bending formulae will give stress levels within 10 per cent of the true levels.

Gilbert[4] later carried out extensive tests on beams of cast iron with a true breaking stress of 21 ton/in^2 (324 N/mm^2). *Figure 139* summarises his results, both for calculated and true fibre stress, and shift of neutral axis. These curves confirm that for fibre stresses up to 30 per cent of the maximum tensile strength the error in the stresses calculated by the conventional formulae is less than 10 per cent, and this error is not likely to be greatly exceeded for irons down to 9–10 ton/in^2 (139–154 N/mm^2) actual tensile strength, utilising the standard beam formulae and the tensile modulus. In any case, the calculated stress levels will tend to be higher than the true level of stress.

A satisfactory margin of safety can therefore be attained by keeping the calculated stresses with conventional beam formulae less than 25 per cent of the actual tensile strength[4].

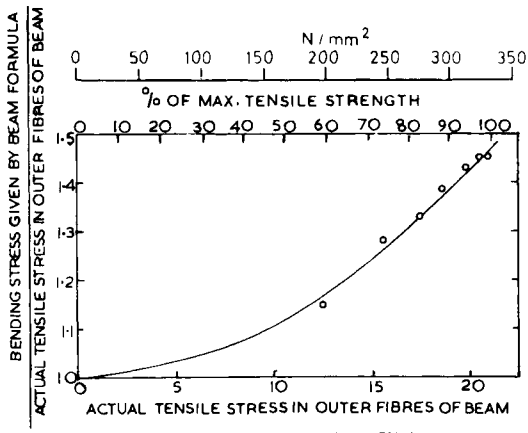
Gilbert[5] has put forward a more detailed method whereby the strength of a cast iron beam in any section shape can be determined, by consideration of the conditions for equilibrium in a beam under load and utilising the basic tensile and compression test data.

Where it is desired to carry out comparison trials with a standard section, Gilbert has indicated that the ratios of strength of a range of sections with a standard section are the same for irons with tensile strengths varying from 10 to 24 ton/in^2 (154 to 371 N/mm^2).

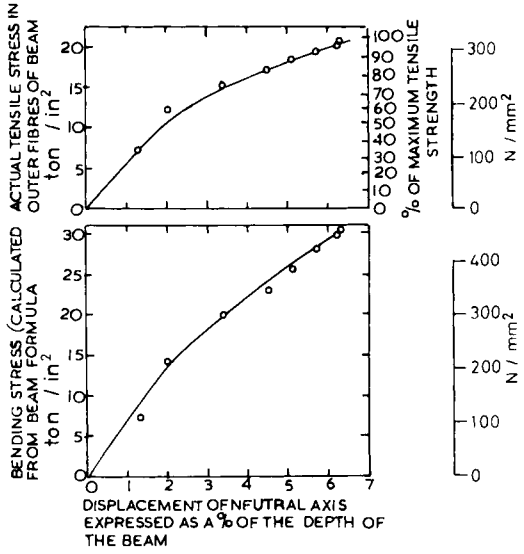
An extremely interesting and detailed examination of old cast iron girder bridges was carried out by Chettoe, Davey and Mitchell[6]. The basic tensile strength of the cast iron appeared to range between 8 and 12 ton/in^2 (124 and 185 N/mm^2), and bending tests of the actual beams and of laboratory specimens cut from them were taken. Full-scale tests on beams up to 30 ft long with beam depths $\frac{1}{12}$th to $\frac{1}{20}$th of the span gave modulus of rupture figures of 7·5 to 8·7 (average 7·9) ton/in^2 [116 to 134 (av. 122) N/mm^2], and Young's modulus figures from 10·0 to 20×10^6 (average $14{\cdot}1 \times 10^6$) lb/in^2 [69 to 138 (av. 97) GN/m^2].

Laboratory tests on samples cut from the same beams gave modulus of rupture figures of 16·3 to 25·2 (average 21·5) ton/in^2 [252 to 389 (av. 332) N/mm^2], and Young's modulus figures of 12·0 to $18{\cdot}2 \times 10^6$ (average $13{\cdot}7 \times 10^6$) lb/in^2 [83 to 125 (av. 94) GN/m^2].

As a result of destruction tests carried out on the bridges themselves, it was concluded that tensile stresses up to 2·5 ton/in^2 (39 N/mm^2) were permissible in the cast iron beams. Since this stress must be divided between dead load and live load, the Building Research Station suggested the following formula[7]: $5{\cdot}5f_l + 2{\cdot}2f_d = 8$ ($5{\cdot}5f_l + 2{\cdot}2f_d = 123{\cdot}6$) for the permissible live load, where f_l and f_d are the respective live and dead load stresses, in ton/in^2 (N/mm^2), calculated on the assumption that the effective section modulus of the com-



VARIATION OF THE BENDING STRESS GIVEN BY THE BEAM FORMULA WITH THE TRUE MAXIMUM TENSILE STRESS



DISPLACEMENT OF THE NEUTRAL AXIS WITH CHANGE OF STRESS.

Figure 139. Tests carried out on beams of cast iron with maximum tensile strength of 21 ton/in²

posite beam equals that of the bare girder multiplied by D/d, where D is the overall depth of the bridge and d the depth of the girder at midspan.

BS 153 Part B Girder Bridges (1937) quotes maximum permissible stress for cast iron in compression of 10 ton/in² (154 N/mm²), and for bridges other than railway, 2½ ton/in² (39 N/mm²) in tensile.

The London Building Act 1930, for purposes other than pillars, quotes the following permissible stresses for cast iron:

Tension:	1·5 ton/in^2 (23 N/mm^2)
Compression:	8·0 ton/in^2 (124 N/mm^2)
Shearing:	1·5 ton/in^2 (23 N/mm^2)
Bearing:	10·0 ton/in^2 (154 N/mm^2)

These figures can be assumed to apply to cast iron components whose actual tensile strength does not exceed 8–10 ton/in^2 (124–154 N/mm^2). The tensile figure is markedly lower than that given for girder bridges and can be assumed to have a considerable factor of safety. These figures can safely be increased proportionally for irons of higher strength.

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COLUMNS

Under its maximum compression load a short column may fail by direct compression, but a long column may fail by buckling. The safe maximum stress will depend, therefore, upon the 'slenderness ratio'. This may be expressed as l/R, where l = length and R = least radius of gyration, or as l/d, where d = least dimension of the section. When comparing data from different sources it is important to determine which units are employed.

There appears to be no recent work on this subject for cast iron, and although various formulae are available for the calculation of permissible loads for cast iron columns, the data given by Goodman[1] appear to be soundly based. He includes two grades of cast iron:

(1) A soft cast iron with a maximum compression stress of 36 ton/in^2 (556 N/mm^2) [equivalent to Grade 10 iron in BS 1452, 10 ton/in^2 (154 N/mm^2) tensile].
(2) A hard, close-grained cast iron with a maximum compressive stress of 58 ton/in^2 (896 N/mm^2) [roughly equivalent to BS 1452, Grade 14–17, or 14–17 ton/in^2 (216–263 N/mm^2) tensile].

The slenderness ratio is expressed as l/d, where l = length of column and d = least dimension of section, and l/d is expressed in consistent units, inches or feet etc.

Where the Euler formula is used l/d must exceed 12.

The recommended factor of safety for the buckling loads in *Table 178*

Table 178 THE BUCKLING LOAD OF CAST IRON COLUMNS. BUCKLING LOAD OF STRUTS OR COLUMNS IN lb/in² (N/mm²) OF SECTION

$\frac{l}{d}$	*Cast iron (soft)*								*Cast iron (close grained)*							
																
	lb/in²	(N/mm²)	lb/in²	(N/mm²)	lb/in²	(N/mm²)	lb/in²	(N/mm²)	lb/in²	(N/mm²)	lb/in²	(N/mm²)	lb/in²	(N/mm²)	lb/in²	(N/mm²)
5	67 100	(463)	64 000	(441)	69 200	(477)	58 900	(406)	109 000	(752)	104 000	(717)	112 000	(772)	95 700	(660)
10	45 200	(312)	40 000	(276)	49 200	(339)	39 900	(275)	73 500	(507)	65 000	(448)	80 000	(552)	53 500	(369)
20	19 600	(135)	16 000	(110)	22 900	(158)	11 500	(79)	31 800	(219)	26 000	(179)	37 200	(256)	19 300	(133)
30	10 100	(70)	8 000	(55)	12 100	(83)	5 800	(40)	16 400	(113)	13 000	(90)	19 700	(136)	9 400	(65)
40	6 000	(41)	4 700	(32)	7 300	(50)	3 400	(23)	9 700	(67)	7 600	(52)	11 900	(82)	5 500	(38)
50	3 900	(27)	3 100	(21)	4 800	(33)	2 200	(15)	6 300	(43)	5 000	(34)	7 800	(54)	3 600	(25)
60	2 800	(19)	2 200	(15)	3 400	(23)	1 500	(10)	4 600	(32)	3 600	(25)	5 500	(38)	2 400	(17)
70	2 100	(14)	1 600	(11)	2 500	(17)	1 100	(8)	3 400	(23)	2 600	(18)	4 100	(28)	1 800	(12)
80	1 600	(11)	1 200	(8)	2 000	(14)	870	(6)	2 600	(18)	1 900	(13)	3 300	(23)	1 400	(10)
90	1 300	(9)	980	(7)	1 600	(11)	690	(5)	2 100	(14)	1 600	(11)	2 600	(18)	1 100	(8)
100	1 000	(7)	790	(5)	1 300	(9)	560	(4)	1 600	(11)	1 300	(9)	2 100	(14)	900	(6)

(*Figure 140*) is 6 for dead loads and 12 for live loads. This, in the case of short columns, is equivalent to a maximum safe compressive stress of:

5·1 ton/in^2 (79 N/mm^2) for soft irons under dead loads.
2·5 ton/in^2 (39 N/mm^2) for soft irons under live loads.
8·3 ton/in^2 (128 N/mm^2) for close-grained irons under dead loads.
4·15 ton/in^2 (64 N/mm^2) for close-grained irons under live loads.

Goodman states that slenderness ratio, l/d, should not exceed 15.

Ros[2] gives a curve of permissible stresses based on slenderness ratios, l/R, where R = least radius of gyration. Curves are given for two irons (a) 13·9 ton/in^2 (215 N/mm^2) tensile and 57·2 ton/in^2 (883 N/mm^2) compression, and (b) 8·9 ton/in^2 (137 N/mm^2) tensile and 44·5 ton/in^2 (687 N/mm^2) compression. The maximum permitted compression stresses ($l/R = 0$) quoted

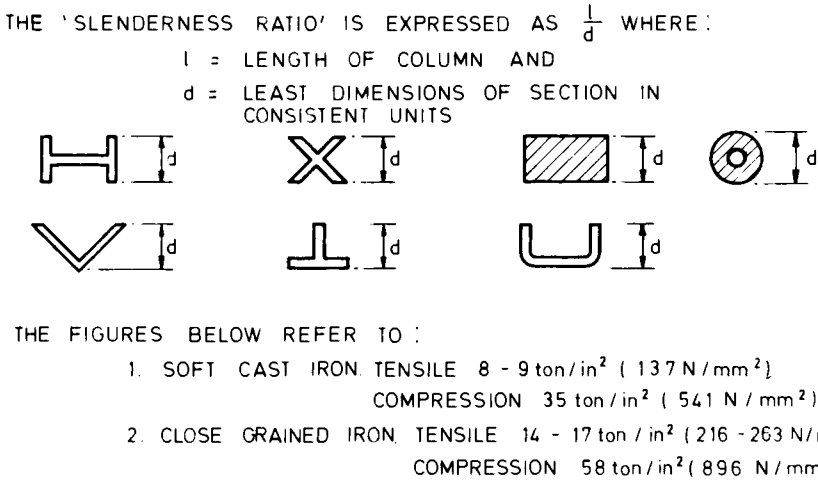


Figure 140. The buckling load of cast iron columns

are respectively 10·2 ton/in^2 (158 N/mm^2) and 7·6 ton/in^2 (117 N/mm^2), which are higher than those given by Goodman.

The London Building Act, 1930, issues limits of working stress for cast iron columns[3], based on l/R slenderness ratio, which show a maximum working stress of 5·5 ton/in^2 (85 N/mm^2) for a column of zero length. For purposes other than columns, the same Act quotes a maximum compressive stress of 8·0 ton/in^2 (124 N/mm^2), a maximum bearing stress of 10·0 ton/in^2 (154 N/mm^2), a maximum shear stress of 1·5 ton/in^2 (23 N/mm^2) and a maximum tensile stress of 1·5 ton/in^2 (23 N/mm^2).

Figure 141 shows the compression stress for the three sets of data, based on columns 12 in (305 mm) outside diameter × 1 in (25·4 mm) wall thickness.

For slenderness ratios (l/R) above about 50, the agreement between the various estimates is quite good, but for shorter columns the LCC figures are conservative (*Table 179*).

Side loading by means of brackets should be avoided wherever possible, as

Table 179 SAFE CONCENTRIC DEAD LOAD ON CAST IRON COLUMNS WITH 'FIXED ENDS', IN ACCORDANCE WITH LCC FORMULA

Outside dia.	*Thickness of metal*		*Safe load in tons (kN) for lengths in feet (m)*															
in (mm)	in	(mm)	6	(1·83)	8	(2·44)	10	(3·05)	12	(3·66)	14	(4·27)	16	(4·88)	18	(5·49)	20	(6·10)
4 (102)	$\frac{1}{2}$	(12·7)	11·3	113	8·3	83	6·05	60	4·56	45	3·77	38	—	—	—	—	—	—
	$\frac{5}{8}$	(15·9)	13·6	136	10·0	100	7·31	73	5·5	55	4·55	45	—	—	—	—	—	—
	$\frac{3}{4}$	(19·1)	15·8	157	11·6	116	8·45	84	6·36	63	5·25	52	—	—	—	—	—	—
5 (127)	$\frac{1}{2}$	(12·7)	17·8	177	14·5	144	10·7	107	8·62	86	6·44	64	5·35	53	—	—	—	—
	$\frac{5}{8}$	(15·9)	21·6	215	17·7	176	13·0	130	10·5	105	7·82	78	6·5	65	—	—	—	—
	$\frac{3}{4}$	(19·1)	25·2	251	20·6	205	15·1	150	12·2	122	9·1	91	7·55	75	—	—	—	—
6 (152)	$\frac{1}{2}$	(12·7)	23·8	237	19·7	196	17·8	177	13·0	130	10·5	105	8·64	86	7·16	71	5·93	59
	$\frac{5}{8}$	(15·9)	29·2	291	24·2	241	21·8	217	16·0	159	12·9	129	10·6	106	8·8	88	7·26	72
	$\frac{3}{4}$	(19·1)	34·2	341	28·2	281	25·6	255	18·7	186	15·1	150	12·4	124	10·3	103	8·5	85
	1	(25·4)	40·5	404	33·5	334	30·3	302	21·7	216	17·9	178	14·7	146	12·2	122	10·1	101
7 (178)	$\frac{1}{2}$	(12·7)	30·6	305	25·4	253	21·0	209	18·9	188	15·4	153	12·5	125	10·8	108	9·3	93
	$\frac{3}{4}$	(19·1)	47·1	469	39·6	395	32·4	323	29·0	289	23·7	236	19·2	191	16·7	166	14·3	142
	1	(25·4)	56·4	562	47·3	471	38·8	387	34·8	347	28·4	283	23·0	229	19·9	198	17·1	170
8 (203)	$\frac{1}{2}$	(12·7)	35·4	353	32·6	325	28·3	282	24·3	242	19·4	193	17·8	177	15·2	151	13·0	130
	$\frac{3}{4}$	(19·1)	51·2	510	47·1	469	41·0	409	35·2	351	28·0	279	25·8	257	22·0	219	18·8	187
	1	(25·4)	66·0	658	60·7	605	52·8	526	45·3	451	36·1	360	32·2	321	28·4	283	24·2	241
9 (229)	$\frac{1}{2}$	(12·7)	—	—	40·2	401	34·8	347	30·6	305	26·8	267	22·8	227	20·2	201	17·4	173
	$\frac{3}{4}$	(19·1)	—	—	58·2	580	50·5	503	44·3	441	38·8	387	33·0	329	29·3	292	25·2	251
	1	(25·4)	—	—	75·3	750	65·2	650	57·2	570	50·2	500	42·6	424	37·9	378	32·7	326
	$1\frac{1}{4}$	(31·8)	—	—	91·2	909	76·0	757	69·4	692	60·8	606	51·6	514	45·9	457	39·5	394
10 (254)	$\frac{3}{4}$	(19·1)	—	—	65·4	652	60·1	599	55·0	548	48·0	478	43·6	434	35·8	357	32·9	328
	1	(25·4)	—	—	84·9	846	78·0	777	71·3	710	62·2	620	56·6	564	46·4	462	42·7	425
	$1\frac{1}{4}$	(31·8)	—	—	110	1096	94·9	946	86·6	863	75·6	753	68·8	686	56·5	563	52·0	518
11 (279)	$\frac{3}{4}$	(19·1)	—	—	—	—	69·0	688	63·0	628	58·0	578	51·0	508	44·8	446	40·0	399
	1	(25·4)	—	—	—	—	89·5	892	81·6	813	75·4	751	66·0	658	58·1	579	51·8	516
	$1\frac{1}{4}$	(31·8)	—	—	—	—	109	1086	99·5	991	92·0	917	80·5	802	70·8	705	63·1	629
12 (305)	$\frac{3}{4}$	(19·1)	—	—	—	—	79·8	795	73·5	732	67·0	668	60·5	603	55·0	548	49·2	490
	1	(25·4)	—	—	—	—	104	1036	95·5	952	87·1	868	78·9	786	71·3	710	64·0	638
	$1\frac{1}{4}$	(31·8)	—	—	—	—	127	1265	116	1156	106	1056	96·1	958	87·0	867	78·0	777

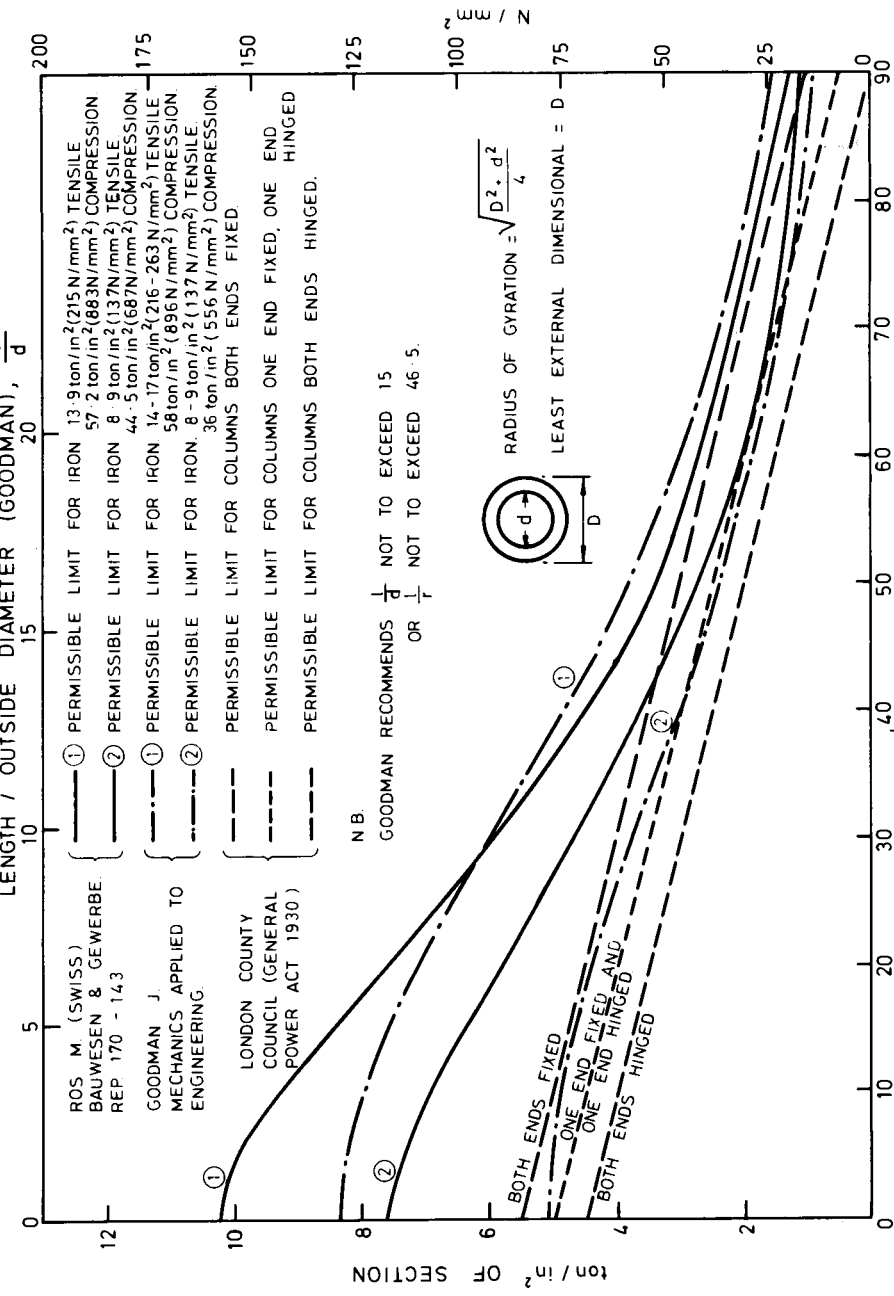


Figure 141. Permissible specific loading for hollow circular cast iron columns concentrically loaded

it greatly increases the buckling tendency. With side brackets and where loads are eccentric, more detailed treatment will be necessary, and reference should be made to standard engineering works of reference.

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PIPES, CYLINDERS AND PRESSURE VESSELS

These components can be divided into three main groups:

(1) Pipes and fittings for the transmission of liquids or gases.
(2) Cylinders for internal combustion engines, pumps etc.
(3) Vessels or containers for liquids or gases.

These can be sub-divided into:

(a) Pipes or vessels operating at pressures up to (i) 200 lb/in^2 (1379 kPa), (ii) 900 lb/in^2 (6205 kPa), (iii) above 1200 lb/in^2 (8274 kPa).
(b) Pipes or vessels operating at temperatures (i) room to 100°C, (ii) above 100°C.

Cast iron is widely employed in all groups for the conveyance of gas, water and oil, for internal combustion engine cylinders, for chemical vessels, hydraulic rams, jacks and fittings etc.

The largest tonnage is undoubtedly supplied in the form of pipes and fittings for the conveyance of gas, water and sewage, and these rarely operate at pressures above 250 lb/in^2 (1724 kPa) or temperatures above 100°C.

In British practice these are covered by BS 78 *Cast Iron Spigot and Socket Pipes* (*Vertically Cast*) and *Spigot and Socket Fittings* Part 1:1961, *Pipes* and Part 2:1965 *Fittings*, and BS 2035:1966 (*Cast Iron Flanged Pipes and Fittings*). These specifications cover four grades (A, B, C and D), and BS 2035:1953 quoted working pressures of 43, 87, 130 and 173 lb/in^2 (296, 600, 896 and 1193 kPa) respectively with installed test pressures of double the working pressure, and works test pressures of $2\frac{1}{4}$–3 times the working pressure. The recommended working pressures are not quoted in the latter specification but the test pressures are the same.

There is, however, an appreciable tonnage employed in hydraulic pressure fittings, valves etc., in which pressures may rise to 1200 lb/in^2 (8274 kPa). These were mainly covered by British Standard Specification 44:1948, now withdrawn [Class A up to 900 lb/in^2 (621 kPa), Class B 900–1200 lb/in^2

(8274 kPa)], with test pressures of 2500 and 3300 lb/in^2 (17·2 and 22·8 MPa) respectively.

Pressures up to 5000–6000 lb/in^2 (34·5–41·4 MPa) are, however, occasionally employed in hydraulic rams and jacks, for which cast iron is also employed both in the cylinders and rams.

Conditions of high temperature, together with high pressure, may be found in chemical reaction vessels, superheated steam valves and fittings, or diesel engine liners and heads.

The British Standards lay down recommended procedures for metal strength, test pressures and wall thickness, and for components operating under the prescribed conditions these standards are entirely adequate (BS 806: 1967). Where standards do not exist, however, some guidance about section thickness, estimated stress and factors of safety may be of some value.

In thick-walled cylinders, the maximum stress occurs on the inner wall, and Lamé's formula for thick cylinders is widely used in the following forms:

$$S_H = \frac{P(R^2 + r^2)}{R^2 - r^2} \tag{1}$$

where S_H = hoop stress in lb/in^2 (N/mm^2)
P = internal pressure in lb/in^2 (MPa)
R = outside radius in in (mm)
r = inside radius in in (mm)

Where the thickness is required for a given pressure and a maximum hoop stress 'S', the formula becomes:

$$t = r\left[\sqrt{\left(\frac{S+P}{S-P}\right)} - 1\right]$$

where t = thickness in in (mm).

For thin cylinders, e.g. in which the wall thickness is less than $\frac{1}{20}$th the diameter, the stress over the whole thickness may be assumed to be uniform and the following expressions may be used:

$$S_H = \frac{pd}{2t} \tag{2}$$

where d = internal diameter. Formula (1) gives higher results than Formula (2).

$$S_L = \frac{pd}{4t} = \tfrac{1}{2} S_H$$

where S_H = hoop stress in lb/in^2 (N/mm^2)
S_L = longitudinal stress in lb/in^2 (N/mm^2)
p = pressure in lb/in^2 (MPa)
t = thickness in in (mm)
d = diameter in in (mm)

For a sphere, using the same notation, the stress S in the shell is given by:

$$S = \frac{pd}{4t}$$

While the calculated stresses given by these formulae are fairly widely used, they apply only to simple shapes, and the stresses likely to be present in more complex fittings, such as branch pipes, may be much higher.

Strain gauge tests on 20 in (508 mm) internal diameter 45° cast iron branch pipes with 1·2 in (30 mm) wall thickness, have shown maximum external skin tensile stresses of 3·0 to 3·5 times the values calculated and found in the cylindrical positions of the pipe. The regions of maximum stress occur on the sides of the junction where the branch runs into the main pipe.

For welded steel branch pipes and tee junctions, stress intensification factors of 3·9 for axial loading and 4·9 for internal pressure loading have been determined so that the bursting strength of such fittings if unreinforced can be assumed to be $\frac{1}{4}$–$\frac{1}{5}$ of that for a simple pipe or cylinder.

Ordinary pipes and fittings are usually made in iron, Grade 10 or 12 (BS 1452), and in sections of $\frac{3}{8}$ in (10 mm) or more the actual tensile strength is likely to be of the order of 9 to 16 ton/in^2 (139 to 247 N/mm^2) depending upon section. Sand cast pipes will give the lower range of tensiles and centrifugally-cast pipes the upper range.

In assessing the stresses which can safely be applied, the test pressures can be taken as one criterion. Thus, if one assumes for cast iron that one quarter the tensile strength is the maximum beyond which marked deviation from elastic behaviour occurs, it is obviously desirable that the test pressure should not produce stresses significantly above this, and the stresses produced at working pressures will automatically follow once a ratio between working and test pressures is established.

In normal water and gas transmission practice for working pressures up to 173 lb/in^2 (1193 kPa), a works test of $2\frac{1}{4}$ to 3 times and an assembled test pressure of twice the working pressure was adopted in the BS 2035: 1953.

For hydraulic power transmission up to 1200 lb/in^2 (8273 kPa), a works test of 2·75 times the rated maximum pressure is used in BS 44:1948. Tests on fully machined cylinders have shown that the tangential stress at fracture caused by internal pressure equals the tensile strength[1,2].

For various reasons the bursting stresses calculated from unmachined cylinders tested to destruction may be 15–25 per cent below the true tensile strength. Thus, normal cast iron pipes to Grade 10 or 12 (BS 1452) will have an actual tensile strength (specimens cut from wall of pipe) of about 9–16 ton/in^2 (139–247 N/mm^2) for sand castings, and 14–19 ton/in^2 (216–293 N/mm^2) for annealed centrifugally-cast pipes. For the purposes of calculation, however, the American standards for similar grades of iron work to a minimum bursting stress (tensile) of 11 000 lb/in^2 (76 N/mm^2) (4·9 ton/in^2) for sand cast pipes and 18 000 lb/in^2 (124 N/mm^2) (8·04 ton/in^2) for centrifugally-spun pipes, using the formula $S = pd/2t$. These values have been obtained by calculation from actual bursting tests and are approximately half of the tensile strength determined directly. In the larger pipes (8 in [203 mm] internal diameter and above) the wall sections in the American standards are calculated on the assumption that the bursting stress will be at least $2\frac{1}{2}$ times the working stress calculated from the normal working pressure, plus an allowance for water hammer, as given below. Higher figures are possible with

sudden valve closure but the figures appear to take into account the time required for valve closure.

Pipe diameter: in	4–10	12–14	16–18	20	24	30	36	42–60
mm	102–254	305–356	406–457	508	610	762	914	1067–1524
Increase in pressure for water hammer:								
lb/in^2	120	110	100	90	85	80	75	70
kPa	827	758	689	621	586	552	517	483

A more detailed treatment of pressure rise caused by water hammer is given in the handbook of the Staveley Iron and Chemical Co. Ltd. (Section H). For instantaneous valve closure the maximum rise in pressure, $P_m = 63{\cdot}5\,Ve$ where P_m is in lb/in^2, V = water velocity in ft/s and e is a constant varying from 0·64 to 0·88 depending on diameter of pipe and class of pipe ($P_m = 1436\,Ve$, where P_m is in kPa and V in m/s).

For pipes smaller than about 8 in (203 mm), British and American standard thicknesses are appreciably greater than calculations required, to allow adequate strength for handling, so that factors of safety against bursting are substantially higher than for sizes above 8 in (203 mm) diameter.

Allowance, however, may have to be made for externally applied loads, for which reference must be made to the appropriate standards and authorities.

The calculated stresses caused by internal pressure at the working loads of normal British Standard Pipes, Class D [maximum working pressure 173 lb/in^2 (1193 kPa) quoted only in BS 2035: 1953] range from about 0·42 ton/in^2 (6·5 N/mm^2) for 4 in (102 mm) pipes to 1·21 ton/in^2 (18·7 N/mm^2) for 48 in (1219 mm) pipes, and at the works test pressure of 390 lb/in^2 (2689 kPa) BS 2035:1966, the calculated stress for 4 in (102 mm) pipes is 0·95 ton/in^2 (14·7 N/mm^2) and for 48 in (1219 mm) pipes is 2·7 ton/in^2 (41·7 N/mm^2).

The nearest corresponding American standards for 200 lb/in^2 (1379 kPa) internal working pressure [field condition C and depth of cover 2½ ft (762 mm)] give stresses, due to internal pressure only, of 0·5 ton/in^2 (7·7 N/mm^2) for 4 in (102 mm) pipes and 1·28 ton/in^2 (19·8 N/mm^2) for 48 in (1219 mm) pipes [or, with additional allowance for water hammer, 1·73 ton/in^2 (26·7 N/mm^2)].

Thus the maximum calculated working stresses caused by internal pressure accepted by both groups of standards is approximately 1·2 to 1·3 ton/in^2 (18·5 to 20·0 N/mm^2) for castings with an actual tensile strength of 9·16 ton/in^2 (141 N/mm^2).

If a calculated bursting stress of 11 000 lb/in^2 (76 N/mm^2), as in the American standards (4·9 ton/in^2), is adopted for iron pipes with a tensile strength of 9 ton/in^2 (139 N/mm^2), these figures are equivalent to a factor of safety of about 2·8 if water hammer is included, and 4 if it is excluded. The American standards work to a minimum overall factor of safety of 2½ where all sources of load are taken into consideration.

Since bursting stresses vary directly with pressure, higher grades of iron will permit higher bursting stresses, roughly proportional to the increase in actual tensile strength.

The factors of safety actually employed in a given circumstance may well vary. Thus, in transmission work with liquids such as water or oil, fracture may lead to very substantial damage by flooding, and with steam, fracture at very high pressures may be explosive. Under such conditions, large factors of safety are desirable.

In certain hydraulic systems, however, where the reserve volume is small—as in hydraulic jacks—the risks of failure are confined to settlement of the load and lower factors of safety may occasionally be regarded as acceptable.

In general, the maximum test pressure should not give calculated stresses of more than one quarter the actual tensile strength, and the working stresses should be correspondingly lower.

As a general guide, the following maximum fibre stresses give a satisfactory factor of safety:

BS 1452 Grade	10	12	14	17	20	23	26
Max. working fibre stress:							
ton/in^2	0·5–1·3	0·6–1·4	0·7–1·8	0·9–2·2	1·0–2·6	1·15–3·0	1·3–3·4
lb/in^2	1100–2900	1350–3100	1570–4000	2000–4900	2240–5800	2500–6700	2900–7600
N/mm^2	7·7–20·0	9·3–21·6	10·8–27·8	13·9–34·0	15·4–40·2	17·8–46·3	20·0–52·5
Australian Std, CBI	2000	2450	2900	3350	9000	4700	5350

The Australian standard suggests working stress for grey iron $\geqslant \frac{1}{7}$ tensile; for nodular $\geqslant \frac{1}{5}$ tensile.

Since the maximum stresses at branches and flanges are practically incalculable but are certainly appreciably higher than those calculated for cylinders, additional webs and sections should be employed at these points and hydraulic test pressures of $1\frac{1}{2}$ to $2\frac{1}{2}$ times the working pressure should be allowed for.

Where complex designs are employed, not only will the actual stresses vary from the calculated, but casting stresses may also be present, and under such circumstances stress relief heat treatment at 500–550° C should be carried out (*see* section on heat treatment).

Vessels operating at high temperatures

The importance of stress relief heat treatment applies in particular to vessels or pipes operating up to 400°C, and the effect of temperature on the creep strength of the material to be used must be taken into account. This is dealt with in the section on heat resistance.

A special case exists with cast iron used for superheated steam at temperatures above about 230°C. Work carried out by Burgess in America indicated the importance of using only cast irons of high tensile if weakening were not to occur owing to accelerated growth. Annealed or ferrite-containing grey

irons may be particularly prone to such deterioration in superheated steam, and it is accordingly recommended that for all purposes above 450°F (230°C) cast iron should have a carbon equivalent value of not more than 3·8 and should have a tensile strength at least equivalent to Grades 17 or 20 (BS 1452). Cast irons which conform to these recommendations do not suffer this form of deterioration, show no measurable growth in air after 11 000 h at 750°F (400°C), and less than 0·1 per cent growth after 8750 h in superheated steam at 750°F (400°C). They are, therefore, entirely satisfactory for use in superheated steam up to 650°F (343°C). Full annealing of such irons might make them susceptible to deterioration in superheated steam. Recent work by Gilbert has confirmed that grey irons above 15·5 ton/in^2 (239 N/mm^2) tensile show 0·01 per cent growth or less after exposure to superheated steam at (800°F) 427°C and that blackheart malleable and nodular cast irons are also unaffected.

The ASME Boiler and Pressure Vessel Code 1968 SA 47 requires, for malleable fittings, the following properties: tensile 53 000 lb/in^2 (365 N/mm^2), yield 35 000 lb/in^2 (241 N/mm^2), elongation on 2 in (51 mm) 18 per cent (p. 8) and for cupola malleable (p. 70): tensile 40 000 lb/in^2 (276 N/mm^2), yield 30 000 lb/in^2 (207 N/mm^2), elongation on 2 in (51 mm) 5 per cent.

For grey iron above 450°F (232°C) only classes 40, 50 and 60 are to be used, stress relieved for 2–12 h at 566–649°C.

Heads and cylinders in internal combustion engines—petrol or diesel—are subject not only to the bending and hoop stresses imposed by the firing pressure, but also to steep temperature gradients which can impose severe surface stresses. Under such conditions the design of the component must allow for both contingencies. Where cracking occurs as a result of both forms of stress, the use of material with a low modulus of elasticity may well be preferable to a stronger material with a high elastic modulus.

Stress due to thermal gradients

The stresses imposed by temperature differences between two faces of a restrained flat plate (no bending) or a cylinder on which one face is heated and the other cooled are as follows:

$$\text{Stress} = \pm \frac{E \times \alpha \times (t_1 - t_2)}{2(1-\mu)}$$

where E = modulus of elasticity
α = coefficient of thermal expansion
$t_1 - t_2$ = temperature difference between the two faces
μ = Poisson's ratio (0·25–0·3 for cast iron)

Thus for a pearlitic cast iron where the temperature difference is 62·3°C:

$$E = 17 \times 10^6 \text{ lb/in}^2 (117 \text{ GN/m}^2)$$
$$\alpha = 12 \times 10^{-6} \text{ per °C}$$
$$\mu = 0{\cdot}25$$

the tangential stress,

$$S_t = \frac{62{\cdot}3 \times 12 \times 10^{-6} \times 17 \times 10^6}{2(1-0{\cdot}25) \times 2240} = +3{\cdot}78 \text{ ton/in}^2$$

$$\left(S_t = \frac{62{\cdot}3 \times 12 \times 10^{-6} \times 117 \times 10^3}{2(1-0{\cdot}25)} = 58{\cdot}3 \text{ N/mm}^2 \right)$$

in tension on the cooled face and compression in the hot face.

REFERENCES

1. CLOUGH, W. R. and SHANK, M. E., *Trans. Am. Soc. for Metals*, **49**, 241–262 (1957)
2. COFFIN, L. F. Jun., *J. appl. Mechanics*, **17**, No. 3, 233–248 (Sept. 1950)

SELECTED REFERENCES ON PIPES, CYLINDERS AND PRESSURE VESSELS:

British Standard Specification 78: *see* p. 460
British Standard Specification 44:1948, Cast iron pipes for hydraulic power, 30 pp. (now withdrawn)
British Standard Specification 1211:1958, Centrifugally cast (spun) iron pressure pipes for water, gas and sewage, 24 pp.
British Standard Specification 2035:1966, Cast iron flanged pipes and flanged fittings, 48 pp.
British Standard Specification 806:1967, Ferrous pipes and piping installations for and in connection with land boilers, 36 pp.
American Standards Association Specification A21.1–1957 and 1967 (American Water Works Association C.101–57) Manual for the computation of strength and thickness of cast iron pipe, 159 pp.
Australian Standard CBI Parts I–IV

9

Application of specifications and data to design

Although specifications usually cover more than one property, one of the principal purposes of a specification is to establish a standard by which a given material may be judged. The basic property chosen for the standard is usually one capable of fairly simple and accurate determination, but which may nevertheless, have little direct bearing upon the purpose for which the material is required. It is important, therefore, when formulating specifications, to use as standards only those properties actually required or known to be closely related to the requirements.

Before specifying the use and grade of cast iron for any component, the following questions should be answered:

(1) What stress must it withstand, static or fatigue?
(2) What impact stress must it withstand?
(3) Which parts require machining?
(4) Is it subject to sliding or abrasive wear?
(5) At what temperature will it operate?
(6) How much distortion is permissible under load?
(7) Is it subject to thermal shock?
(8) Is corrosion likely to occur?
(9) Must the casting be hydraulically sound?

The first and second questions are fundamental, since if the castings fracture they are useless for their purpose.

TENSILE STRENGTH

This is often the least important property of a casting, but in conjunction with the phosphorus content does provide the most reliable index to all the other normally required mechanical properties. It is possible, in a given section, to predict with some certainty from the tensile test:

(1) Fatigue strength.
(2) Impact resistance (at a given phosphorus level).
(3) Transverse strength and deflection (at a given phosphorus level).
(4) Compression strength.
(5) Shear strength.
(6) Torsional strength.
(7) Hardness (at a given phosphorus level).

The grade number assigned to a particular iron in almost all national specifications is the minimum tensile strength required when the metal is cast into a 1·2 in (or 30 mm) diameter sand cast bar. Thus, in BS 1452, Grade 14 requires a tensile strength of 14 ton/in^2 (216 N/mm^2), in a 1·2 in (30 mm) bar; the American ASTM Specification A48–64 Grade 30 requires a tensile of 30 000 lb/in^2 (207 N/mm^2) in a 1·2 in (30 mm) bar; and the Swedish, German and Swiss Grade 20 requires 20 kg/mm^2 (196 N/mm^2) in a 30 mm bar. These strengths, however, will be found only in those sections of the casting that cool at the same rate as the centres of the 1·2 in (or 30 mm) test bar. Reference to this is made in Chapter 2. It will be seen from *Figure 39*, p. 41, that a Grade 14 iron, giving a minimum tensile of 14 ton/in^2 (216 N/mm^2) in a 1·2 in (30 mm) bar would be expected to give 16 ton/in^2 (247 N/mm^2) in a wall section of 0·35 in (9 mm) and 10 ton/in^2 (154 N/mm^2) in a 3–4 in (76–102 mm) section; all the related properties, fatigue, transverse, hardness, and so on, would vary correspondingly with the actual tensile strength roughly in accordance with the data given in the appropriate sections.

The tensile strength is merely an index of strength and not of quality. The best quality of iron is that which best fulfils its purpose: if tensile and fatigue strengths are of overriding importance, then a high tensile iron must be specified, having due regard to the effect on hardness and machinability. If free machinability is the criterion, a high tensile iron is not the best, but an annealed low-phosphorus iron with a tensile of about 10 ton/in^2 (154 N/mm^2) is more suitable. If fluidity and cheapness are called for, then a high phosphorus Grade 10 or Grade 12 iron may be most suitable. Grey irons can be provided (usually with alloys) up to about 28 ton/in^2 (432 N/mm^2), but consideration must also be given to the malleable irons which will give tensiles of 22–38 ton/in^2 (340–587 N/mm^2) with considerable shock resistance and ductility in sections up to 1–1$\frac{1}{2}$ in.

For tensiles over 28–32 ton/in^2 (432–494 N/mm^2), the nodular or spheroidal graphite irons are suitable, giving up to 60 ton/in^2 (927 N/mm^2) in the pearlitic grades, with comparatively low ductility and shock resistance, and from 24 to 35 ton/in^2 (371 to 541 N/mm^2) in the ferritic grades with very substantial impact resistance and ductility in almost any section thickness (*see* section on nodular cast iron, p. 201 et seq.).

MODULUS OF ELASTICITY

It is important not to underestimate the significance of the modulus of elasticity. There are many circumstances in which a low modulus is much more important than a high tensile strength.

The following conditions may be better served by irons with a low modulus of elasticity:

(1) Conditions where vibration may cause unwanted noise or develop resonant stresses, and high damping capacity is desirable.
(2) Conditions of thermal shock, as in ingot moulds.
(3) Structures usually subject to differential thermal stresses where one component must yield, e.g. tube plates in chemical vessels.
(4) Components operating with substantial thermal gradients without thermal shock, as in diesel pistons and heads, and subject to relaxation stresses.
(5) Brake surfaces operating with substantial thermal gradients and considerable thermal shock.

Under conditions of thermal shock as in ingot moulds the use of irons with a high nitrogen content, such as may occur when cupola melting high steel charges, may by the production of compacted graphite yield metal with higher tensile strength and hardness and raise the modulus of elasticity from $11–13 \times 10^6$ lb/in^2 (76–90 GN/m^2) to $17–18 \times 10^6$ lb/in^2 (117–124 GN/m^2).

To use high-strength, high modulus irons in such conditions may frequently lead to disappointing results. The difficult combination of high tensile and low modulus is particularly desirable under the conditions under (4).

On the other hand, when using materials with different moduli of elasticity, such as cast iron and steel, the effect of the differing deflection under identical stresses must be borne in mind.

MACHINABILITY*

The most important property of many engineering castings is machinability, and machinability tends to fall markedly as the tensile strength increases. To specify an iron with a tensile strength higher than is required by the stress to which the component is subjected is to lower the permissible cutting speeds and thus penalise the output of the machine shop. This is true even when the section of the iron yields a machinable and fully pearlitic structure (up to 250 Brinell), in every portion. Higher strength irons, however, are more prone to chill and produce hard spots (255–500 Brinell), particularly at free edges and corners, so that the maximum strength that should be specified

Table 180 MINIMUM CHILL-FREE EDGE OR EQUIVALENT DIAMETER

BS 1452:1961 Grade				
10	*12*	*14*	*17*	*20*
0·1–0·2 in	0·2–0·3 in	0·3–0·4 in	0·4–0·6 in	0·6–1·0 in
2·5–5·1 mm	5·1–7·6 mm	7·6–10 mm	10–15 mm	15–25 mm

* See also p. 332.

is one that will still produce a soft casting even at the point of maximum cooling rate. For the British Standard grades, the minimum safe equivalent diameters (or free edge thicknesses) are given in *Table 180.*

Where thinner sections than these are required, it will be necessary to inoculate the cast iron before pouring if chill or hard spots are to be avoided. The section on strength, composition and structure (machinability) in Chapter 3 deals more fully with the effect of variations in casting thickness.

MACHINABILITY AND ANNEALING

This is dealt with more fully in the sections on heat treatment, and machinability. Where an iron meets a given strength specification but hard spots are present, these can be removed by annealing. The iron will, however, no longer meet the original strength specification as the tensile will fall by about one-third and the fatigue strength will fall to about one-half of that before annealing. The impact strength, however, may be slightly raised and the total energy to rupture is said to be increased appreciably by annealing (*see* sections on impact strength and heat treatment).

HARDNESS

Although some Continental standards specify maximum hardness, together with strength, these conditions can be safely met only under clearly defined conditions of tensile strength, section thickness, and phosphorus and alloy content. The method of surface preparation for the hardness test is also of importance, since certain methods of casting may lead to the formation of a decarburised skin, particularly with heavy castings. Levels of acceptable hardness and methods of test must therefore be agreed upon between the customer and the foundryman.

Hardness affects machinability and wear resistance, and may also be an index of strength, but there is no clearly defined relationship between the three properties that is applicable under all circumstances. The most that can be said is that in general:

(1) The higher the hardness, the poorer the machinability.
(2) The higher the hardness, the better the wear resistance (this statement is subject to many qualifications and is by no means always true except under fully abrasive conditions).
(3) For pearlitic and ferritic irons of similar graphite form and phosphorus and alloy content, the higher the hardness the higher the tensile strength.

IMPACT STRENGTH

Where castings are subjected to sudden shock loads, the lower the phosphorus content the greater the shock resistance, and although grey cast iron is essentially a brittle material, the difference in single blow shock resistance over a range of 0·2 to 1·0 per cent phosphorus is very marked and significant. It is not easy, however, to interpret the conventional Izod or Charpy impact figures in terms of engineering usage. For example, shoemakers' lasts, which are subject to repeated hammering, were normally made in the more brittle high-phosphorus irons, and all that can be said is that if made in a low-phosphorus iron (0·3 per cent phosphorus or less) they would have a much greater safety margin against fracture by single blow, but not necessarily against repeated blows of low intensity.

In the section on impact strength it has been shown that the impact fatigue strength of irons with phosphorus contents at least up to 0·81 per cent is not significantly different from that of low-phosphorus irons for repeated blows of low intensity. There are differing degrees of brittleness; for example, railway lines of 0·6 per cent carbon steel which are subject to the battering of main line traffic are also brittle by conventional standards, with notched Izod impact figures of 2–4 ft.lb (3–5 J), and although grey cast iron is obviously unsuitable for this class of work there are innumerable instances of engineering castings, such as press and forge tool frames, gears and camshafts, made in grey iron—much of it high-phosphorus—which withstand repeated impact successfully during the whole life of the component. There are some circumstances in which ductile failure may prove more of an embarrassment than brittle failure. Thus, cast iron gears will frequently operate for long periods of time with one tooth missing, when a bent mild steel tooth would stop the mechanism. Cast iron propellers also are widely used in coastal waters, not merely because of their low cost, but also because of the ability of the blades to break when hitting an obstruction, thus reducing the liability of damage to the driving gear.

There are, of course, many circumstances in which brittle failure under single shock loading cannot be tolerated, and where experience has shown that the low-phosphorus irons are still too brittle it will be necessary to use either whiteheart or blackheart malleable iron. Typical cases are car door hinges, cycle and motorcycle brackets, conveyor chains, motor axle boxes, and so on. These irons are much softer than the normal as-cast grey irons, are very freely machinable, and have very substantial ductility and shock resistance (*see* section on malleable cast iron). They are cast in section down to 0·1 in (2·5 mm), but are not normally cast in sections thicker than 1–1½ in (25–38 mm).

Brackets of well decarburised whiteheart malleable iron can be successfully welded to mild steel structures, and provided suitable material is specified, and the welding technique and the design are satisfactory, fully shock resistant structures will be obtained.

Since some of the resistance to fracture by bending of a whiteheart iron is given by the deep decarburised skin, deep machining or threading may

reach the pearlitic centre and reduce quite appreciably the shock resistance or resistance to fracture by bending (*see* section on malleable cast iron). This effect is less marked in the case of blackheart iron where the body of the iron is ferritic.

Where shock resistance is required in sections greater than 1–1½ in (25–38 mm), the use of the ferritic nodular or spheroidal graphite irons may be necessary. Such irons can combine high tensile strength [24–32 ton/in^2 (371–494 N/mm^2)] with shock resistances comparable with some cast steels [BS notched Izod, 6–14 ft.lb (8–19 J)]. For temperatures below -10°C, where many normally ductile materials become brittle, suitable ferritic nodular irons and malleable irons can offer shock resistance comparable with that of many steels (*see* section on impact strength in Chapter 2).

FATIGUE RESISTANCE

Where components are to be subject to alternating stresses, the possibility of fatigue failure must be considered. Even the comparatively slow rate of alternation of 1 per second will give over 31 000 000 reversals with continuous running for 12 months, and with cast iron it is wise to assume that failure can occur up to 20 000 000 alternations if the stresses are high enough. Failure is unlikely to occur at a given level of stress after 20 000 000 reversals.

The section on fatigue strength in Chapter 2 gives data on the fatigue strength of various cast irons based mainly on the rotating beam Wöhler type test. This test assumes a zero mean stress with alternating stresses equal and opposite, and the ratio of fatigue strength to tensile strength may range from 0·35 to 0·45. This type of stressing is characteristic of the rotating shaft or crankshaft. Thus, a component of cast iron whose actual strength is 20 ton/in^2 (309 N/mm^2), would be expected to fail up to 20 000 000 reversals of stress of 7 to 9 ton/in^2 (108 to 139 N/mm^2). These figures, however, are based on carefully machined specimens and accurately aligned testing machines. Very little rotating machinery can provide such accuracy and alignment, nor in calculating the stresses can it be assumed that the calculations have taken fully into account every factor, e.g. inertia, load cycle variation and so on.

Although grey iron is not regarded as notch-sensitive, notches which increase the effective stress above the safe limit will cause failure and the subsequent provision of suitable fillets at such points, while desirable, is less effective than with steel unless the section has been markedly increased.

The general experience of the BCIRA, based on the examination of crankshaft failures, indicates that calculated bending stresses should not exceed ± 3 to 3½ ton/in^2 (46–54 N/mm^2) for shafts in which the actual tensile strength is not less than 20 ton/in^2 (309 N/mm^2). If the section thickness exceeds 1 to 1½ in (25 to 38 mm), or if there is any element of doubt about the accuracy of the stress calculations or the precise strength of the iron, a figure of 2½ to 3 ton/in^2 (39 to 46 N/mm^2) should not be exceeded.

Gassner and Schütz[1] state that fatigue stresses in crankshafts should be based on maximum firing pressure. Diesel engines have peak pressures up to 1400 lb/in^2 (9653 kPa) and, with supercharging, up to 2000 lb/in^2

(13 790 kPa). Petrol engines have normal peak pressures between 575 and 850 lb/in^2 (3964 and 5861 kPa). 'Pinking' may increase the pressure up to 30 per cent.

Fillet radii of crankpins should be $\frac{1}{10}$ crankpin diameter.

Pearlitic nodular irons may have limiting fatigue stresses of ± 12 ton/in^2 (185 N/mm^2) or over, but in the absence of critical operational experience these also should operate with a substantial safety margin, and calculated stresses should not exceed ± 6 ton/in^2 (93 N/mm^2). Steel shafts are rarely stressed above ± 9 ton/in^2 (139 N/mm^2), in spite of limiting fatigue figures of ± 18 ton/in^2 (278 N/mm^2).

If for any reason either the grey or nodular irons have been annealed, both the tensile and the limiting safe fatigue stress will be approximately halved, although the shock resistance may have been increased.

There are many cases, however, when the alternating stresses are not bending, but direct, as in connecting rods, and notably when they occur as hoop stresses in wet liners of internal combustion engines. In the latter case the stress alternates from positive to zero with virtually no compression component. The permissible range of stress under these conditions of direct loading appears to be lower (fatigue/tensile ratio of 0·25) than for rotating beam fatigue where the stresses alternate from tension to compression (fatigue/tensile ratio of 0·35); where the calculated loads are likely to be subject to mechanical inaccuracy, a factor of safety of 3 should be employed. For gas loadings, as in internal combustion engine liners, hoop stresses in excess of $+3\frac{1}{2}$ ton/in^2 (54 N/mm^2) to zero should be avoided.

The limiting fatigue stress increases as the compression component increases (*see* section on fatigue strength).

Fatigue limits are never specified for cast iron since the figures can be deduced with sufficient accuracy from the tensile strength, and the data given previously in this handbook provide reasonable safety margins.

Increasing the tensile strength by quenching and tempering, however, cannot be relied on to increase the limiting fatigue stress.

The fatigue strength appears to be increased at temperatures up to 400°C, but no data are available for sub-zero temperatures.

REFERENCE

1. GASSNER, E. and SCHÜTZ, M., *Engineers Digest*, **22**, No. 9, 79–81 (Sept. 1961)

WEAR AND HARDNESS*

High overall component hardness appears to be desirable only when abrasive conditions are present or high local pressures exist, as in cams, rollers or gears. Where poor or marginal lubrication is present, the relatively soft, high-carbon, high-phosphorus irons frequently give better performance than the tougher, high-tensile, low-carbon irons. The graphite acts as an anti-welding solid lubricant in the early stages of running in; the graphite cavities act as

* See also pp. 286–307.

oil reservoirs and as points of termination for local damage, while the phosphide eutectic, which corrodes less than the matrix in contact with the lubricant, tends to stand proud of the main surface and carries the opposing surface. The surrounding area, at a slightly lower level, ensures an adequate spread of oil around all the bearing points.

Nevertheless, the aim of the designer should always be to provide the most adequate lubrication, since the effect of metal structure is insignificant where full hydrodynamic lubrication is present.

Metal structure is obviously of no significance where coatings of chromium are applied. In such cases, the only requirements of the base iron are strength to carry the load and soundness to carry the coating.

CORROSION AND ABRASION

These conditions usually require a combination of hardness and corrosion resistance, and the hardened grey irons may prove disappointing, as their corrosion resistance may be lower than in the unhardened state[1].

The 30 per cent chromium irons and 14–18 per cent silicon irons combine high corrosion resistance with high hardness (320–420 DPN) and frequently offer suitable solutions to otherwise intractable problems, e.g. lightly loaded bearing rollers operating under water, and guides and rests operating in abrasive slurries. The 30 per cent chromium iron can be machined with suitable tools and speeds, but 14–18 per cent silicon iron is usually ground and is very brittle.

The white alloyed or unalloyed irons are suitable where the main difficulty is abrasion, although they are also appreciably more corrosion resistant than the grey irons.

Where corrosion is the main problem, the austenitic irons of the Ni-Resist type may be suitable, as their corrosion resistance is high and their resistance to abrasion is higher than would be expected from their low hardness (170 Brinell).

For general service, where conditions are not severe and large tonnages are called for and cost is a major consideration, the ordinary high-phosphoric irons can give excellent service and are widely employed.

Tables 181 and *182* summarise the behaviour of various materials under different conditions of wear (see pp. 310, 324, 328, 329).

REFERENCE

1. PIWOWARSKY, E., *Giesserei*, **24**, 97–106 (Feb. 1937)

Table 181 LUBRICATED SLIDING SURFACES: LOW SURFACE LOADING [LESS THAN 1500 lb/in^2 (10 N/mm^2)]

Continuous operation (complete lubrication)	*Intermittent operation (complete lubrication while running)*	*Sliding with marginal lubrication*	*Combined corrosion and abrasion*
Continuously operated shaft bearings	*Shaft bearings and slides*	*Internal combustion engine cylinders, piston rings, marginal lubrication*	*Pumps, pipes, valves, clay working dies etc.*
Any cast iron is suitable	Any cast iron is suitable under favourable conditions. Greatest safety margin with iron of total carbon exceeding 3·3%, with random graphite and phosphorus over 0·6%. With high phosphorus the matrix may be either pearlitic to give a hard bearing (180–220 Brinell) or ferritic to give a soft bearing (140–180 Brinell). With low phosphorus (less than 0·4%) the matrix must be pearlitic (min. 170 Brinell)	(a) Alloyed, pearlitic or sorbitic matrix with dispersed phosphide/carbide. Random graphite. Heat-treated martensitic or sorbitic. Nitrided. Chrome plated. (b) Unalloyed 0·6–1% phosphorus in pearlite. (c) Unalloyed low phosphorus with random graphite, pearlitic matrix with less than 5% ferrite. Group (a) For severe duty, the type used depending upon conditions. (b) Normal automobile liners and rings. (c) Automobile cylinder blocks. For piston rings a total carbon of 3·4% or more is preferable	Approximate order of corrosion and abrasion resistance: (a) 30% chromium 14–16% silicon. (b) Austenitic iron, e.g. Ni-Resist (primarily corrosion resistant). (c) White iron, alloyed or unalloyed (primarily abrasion-resistant)

Table 182 UNLUBRICATED AND LUBRICATED SLIDING OR ROLLING SURFACES: HIGH SURFACE LOADING [1500–180 000 lb/in^2 (10–1241 N/mm^2)]

Dry abrasion	*Unlubricated rolling*	*Lubricated rolling and/or sliding*		
Rolls, crushers, chutes etc.	*Rolls, wheels, rollers*	*Rollers (100% rolling)*	*Gears (varying amounts of sliding and rolling)*	*Cams and tappets (100% sliding)*
Order of merit against abrasion: (1) White alloyed iron. Hardness 600–900 DPN. (2) White unalloyed iron. Hardness 400–600 DPN. (3) Hardened and tempered alloy iron. Hardness 300–530 DPN. (4) Mottled or dispersed carbide iron. Hardness 280–350 DPN. (5) High-phosphorus irons, 0·7–1·4% phosphorus. Hardness 220–260 DPN.	White chilled iron for high surface finish and wear resistance (450–650 DPN). Mottled or dispersed carbide for toughness and wear resistance (240–280 DPN). Hardened and tempered with or without dispersed carbide (300–550 DPN). Grey for toughness and shock resistance (220–260 Brinell).	White or chilled iron for loading (350–650 DPN). S_{max} 80–180 000. Mottled or dispersed carbide iron (240–280 DPN). S_{max} 70–90 000. Flame or induction hardened surfaces. S_{max} 70–100 000. Oil quenched and tempered. S_{max} 70–100 000. Hard grey iron (220–260 Brinell). S_{max} 50–80 000.	Hard grey (220–260 Brinell) with free phosphide or carbide for wear resistance and machinability. Without free carbide or phosphide for toughness. S_{max} 50–85 000. Flame or induction hardened or oil quenched. S_{max} 92 000. White or chilled surfaces might be suitable where profile can be ground. Teeth would be rather more brittle than hard grey iron.	White or chilled surfaces (350–650 DPN). S_{max} 80–160 000. Flame or induction hardened grey (350–500 DPN). S_{max} 80–120 000. Oil quenched and tempered grey. S_{max} 70–100 000.

DESIGN STRESSES AND FACTORS OF SAFETY

The stresses that can safely be applied to a cast iron component have never been concisely summarised, but *Table 183c* gives tentative estimates of the maximum stress which can be applied without shock loading. *Tables 183a* and *b* summarise the basic data for the given grades of iron upon which the maximum design stresses have been based (*see* footnotes).

These stresses are presented only as a guide, and the experience of the designer must decide what, if any, additional factors of safety require to be employed with them.

In direct static tensile it may be assumed that the fatigue strength is a reasonable criterion, since if the iron will withstand 20×10^6 reversals of a given stress, it is not unreasonable to suppose that it may be stressed statically up to this load without risk of failure. Alternatively 0·75 of the limit of proportionality may be employed or $\frac{1}{4}$ of the tensile strength of grey iron. British girder bridge practice permits the use of a BS 1452 Grade 10 iron in tensile up to $2\frac{1}{2}$ ton/in^2 (39 N/mm^2), but not for railway bridges where it can only be used in compression. It will be seen that these figures are below the apparent limit of proportionality. If any shock loading is possible, a factor of safety will have to be employed which can only be determined by experience, but an additional factor of safety of 2 is usually regarded as adequate. For components operating under steam or gas pressure where an explosion risk exists, a further factor of safety is required.

The London Building Act (1930) quotes the working stresses presented in *Table 184*.

The figures for cast iron can be presumed to apply to an iron not exceeding BS 1452 Grade 10, with an actual tensile strength of 8–10 ton/in^2 (124–154 N/mm^2).

The American Society of Mechanical Engineers' criteria for design stresses for unfired pressure vessels in forged materials are based upon the following:

(1) One-fourth of the minimum tensile strength.
(2) Two-thirds of the minimum specified yield.
(3) The stress producing a secondary creep of 0·1 per cent in 10 000 h.
(4) Four-fifths of the stress producing rupture in 100 000 h.

In Germany the general basis for design stress is yield stress divided by 1·8, and the accepted test pressure for steel vessels is 1·3 × maximum attainable pressure.

In rotary bending fatigue, direct experience with grey iron crankshafts at the BCIRA indicates that failure usually occurs in bending rather than torsion, and places the maximum calculated bending stress at ± 3 to $3\frac{1}{2}$ ton/in^2 (46 to 54 N/mm^2) for motor cycle shafts in which the actual tensile strength is not less than 20 ton/in^2 (309 N/mm^2). A factor of 0·60 × minimum fatigue value may also be used.

For stationary engines these stresses could probably be increased by 15 to 25 per cent, but where size is of extreme significance, it is probably better

Table 183a BASIC TENSILE DATA AT NORMAL TEMPERATURES[1]

Grade	*Tensile strength in stressed portion** ton/in²	(N/mm²)	*Proof stress, 0·1%* ton/in²	(N/mm²)	*Proportional limit, 0·01% deviation from E_0* ton/in²	(N/mm²)	*Total plastic strain at failure*	*Young's modulus* $10^6 \times$ lb/in²	(GN/m²)
10	10	(154)	6·5	(100)	2·5	(39)	0·45–0·60	11–15	(76–103)
12	12	(185)	7·8	(120)	3·3	(51)	0·33–0·53	12–16	(83–110)
14	14	(216)	9·1	(141)	3·9	(60)	0·22–0·47	14–17·4	(97–120)
17	17	(263)	11·1–12·5	(171–193)	4·7	(73)	0·38	16–19	(110–131)
20	20	(309)	13·0	(201)	5·5	(85)	0·28	17–21	(117–145)
23	23	(355)	15·0–17·5	(232–270)	6·3	(97)	0·25	18–22	(124–152)
26	26	(402)	16·9	(261)	7·2	(111)	0·28	18–22	(124–152)
Whiteheart malleable	20–24	(309–371)	11·6–12·7	(179–196)	6·3–7·0	(97–198)	Elongn. % 4–8	25·5 (21–26)	(176) (145–179)
Blackheart malleable	18–22	(278–340)	11–14	(170–216)	8·4–10·4	(130–161)	6–14 min.	24·5 (21–27)	(169) (145–186)
Pearlitic malleable	28–33	(432–510)	17·1–20·6	(264–318)	10·3–12·4	(159–192)	6–4 min.	25 (21–27)	(172) (145–186)
Nodular pearlitic SNG	32–47	(494–726)	20–25	(309–386)	12–15	(185–232)	7–2	24·5–25·5	(169–176)
Nodular ferritic SNG 27/12	27	(417)	16·9	(261)	11·8	(182)	12	24·5	(169)
Nodular ferritic SNG 24/17	24	(371)	14·5	(224)	10·9	(168)	17	24·5	(169)

* The tensile strength quoted in column 2 is that obtained in a standard 1·2 in (30 mm) as-cast test bar of the grade quoted for grey iron. If the stressed portion is of much heavier section, the tensile strength may be much lower (*see Figure 39*) and the maximum design stress will be correspondingly lower in accordance with the criteria given.

Reference

1. Engineering Data on Cast Iron, BCIRA, 1969, Alvechurch, 72 pp.

Table 183b BASIC TEST DATA AT NORMAL TEMPERATURES: COMPRESSION, HARDNESS, IMPACT, FATIGUE†

	Compression										
			Proof stress					*Impact strength*		*Fatigue strength*	
	Maximum		*0·1%*		*0·2%*		*Brinell*	(*BS 1349*) (*unnotched*)		(*0·25 to 0·45 × tensile*)	
Grade	ton/in²	(N/mm²)	ton/in²	(N/mm²)	ton/in²	(N/mm²)	*hardness*	ft.lb	(J)	ton/in²	(N/mm²)
10	35–40	(541–618)	13·0	(201)	—	—	140–210	6–12	(8–16)	±2·5–4·5	(±39–69)
12	36–48	(556–741)	15·6	(241)	—	—	160–220	6–12	(8–16)	±3·0–5·4	(±46–83)
14	40–56	(618–865)	18·2	(281)	—	—	165–225	6–12	(8–16)	±3·5–6·3	(±54–97)
17	50–65	(772–1004)	22·2	(343)	—	—	170–290	10–23	(14–31)	±4·3–7·7	(±66–119)
20	55–65	(849–1004)	26	(402)	—	—	200–290	10–23	(14–31)	±5–9	(±77–139)
23	60–90	(927–1390)	30	(463)	—	—	205–320	18–35	(24–47)	±5·8–9·8	(±90–151)
26	70–90	(1081–1390)	33·8	(522)	—	—	220–350	18–35	(24–47)	±6·5–9·9	(±100–153)
								BS 131			
Whiteheart malleable	—	—	11·5–13·7	(178–212)	11·7–13·9	(181–215)	110–248	2–3 notched	(3–4 notched)	±9–11	(±139–170)
Blackheart malleable	no definite failure		12·1–14·7	(187–227)	12·2–14·8	(188–229)	109–145	up to 16	(up to 22)	±8·1–9·9	(±125–153)
Pearlitic malleable	—	—	18·4–21·7	(284–335)	18·6–22·1	(287–341)	168–240	—	—	±12·6–14·4	(±195–222)
								BS 2789 & BS 131			
Nodular pearlitic SNG	115–144	(1776–2224)	21·3–25·8	(329–398)	21·8–27·6	(337–426)	260–320	—	—	±14·4–18·8	(±222–290)
Nodular ferritic SNG 27/12	110–120	(1699–1853)	18·4	(284)	18·6	(287)	130–187	2–10	(3–14)	±13·2	(±204)
Nodular ferritic SNG 24/17	—	—	16·0	(247)	16·2	(250)	130–159	10	(14)	±12	(±185)

† *Important*. Factors of safety to be used with these stresses must be related to conditions of service. The maximum direct or fibre static stress may be based on 0·6 times the minimum fatigue strength, 0·75 of the limit of proportionality, or for grey iron 0·25 of the tensile strength which is about 0·01% proof stress value.

Table 183c

Maximum design stresses†

Grade	*Direct or fibre static tensile (based on min. fatigue)* ton/in²	(N/mm²)	*Alternating direct tensile/compression stress (0·5 min. fatigue)* ton/in²	(N/mm²)	*Rotary bending fatigue* ton/in²	(N/mm²)	*Compression stress (approx. 0·3 maximum compression), l/d not over 2* ton/in²	(N/mm²)
10	1·5–2·5	(23–39)	1·25	(19)	±1·5	(±23)	10	(154)
12	3	(46)	1·5	(23)	±1·8	(±28)	12·5	(193)
14	3·5	(54)	1·75	(27)	±2·1	(±32)	14·6	(225)
17	4·2	(65)	2	(31)	±2·9	(±45)	17·8	(275)
20	5	(77)	2·5	(39)	±3	(±46)	16–20·8	(247–321)
23	5·7	(88)	3	(46)	±3·3	(±51)	18–24	(278–371)
26	6·5	(100)	3·2	(49)	±3·3–4	(±51–62)	20–27	(309–417)
Whiteheart malleable	4·7–5·7	(73–88)	—	—	±3·0–3·6	(±46–56)	6·9–8·2	(107–127)
Blackheart malleable	6·3–7·7	(97–119)	4	(62)	±2·7–3·3	(±42–51)	7·3–8·8	(113–136)
Pearlitic malleable	7·7–9·3	(119–144)	6·7	(103)	±4·2–4·8	(±65–74)	11–13	(170–201)
Nodular pearlitic SNG	9–11·3	(139–175)	7–9	(108–139)	±4·8–6·3	(±74–97)	12·8–15·5	(198–239)
Nodular ferritic SNG 27/12	9·5	(147)	6·5	(100)	±4·4	(±68)	11	(170)
Nodular ferritic SNG 24/17	9·5	(147)	6	(93)	±4·4	(±68)	11	(170)

† *Important.* Factors of safety to be used with these stresses must be related to conditions of service. The maximum direct or fibre static stress may be based on 0·6 times the minimum fatigue strength, 0·75 of the limit of proportionality, or for grey iron 0·25 of the tensile strength which is about 0·01% proof stress value.

Table 183d

Basic test data at elevated temperatures: stress

Grade		*350°C*		*400°C*		*500°C*		*600°C*	
		ton/in²	(N/mm²)	ton/in²	(N/mm²)	ton/in²	(N/mm²)	ton/in²	(N/mm²)
10–12	Short term tensile	[10–12]	[154–185]	[10–12] 11	[(154–185) 170]	7·4	(114)	3·3	(51)
	Stress to rupture 100 000 h	[6–7]	[93–108]	[4]	(62)	—	—	—	—
14–17	Short term tensile	[14–16·8]	[216–259]	[14–16·8]	(216–259)	10–13·8	(154–213)	4–8·3	(62–128)
	Stress to rupture 100 000 h	9–10	(139–154)	6	(93)	1·5	(23)	—	—
	Total creep for 100 000 h 0·1%	5	(77)	1·8	(28)	—	—	—	—
20–26	Short term tensile	20–25	(309–386)	19–23	(293–355)	12–15	(185–232)	5–9	(77–139)
	Stress to rupture 100 000 h	17	(263)	8–11	(124–170)	2–3	(31–46)	—	—
Whiteheart malleable	Stress to rupture in 100 000 h	9–14	(139–216)	5–10	(77–154)	—	—	—	—
	Total creep 0·2% in 100 000 h	2–3	(31–46)	1–1·2	(15–19)	—	—	—	—
Blackheart malleable	Stress to rupture in 100 000 h	9	(139)	5–6	(77–93)	2·5	(39)	—	—
	Creep rate for 1000 h 0·1%	—	—	5·4	(83)	1·0	(15)	—	—
Pearlitic malleable	Stress to rupture in 100 000 h	16	(247)	10	(154)	2–3	(31–46)	—	—
	Stress to rupture in 100 000 h	20–25	(309–386)	13–15	(201–232)	—	—	—	—
Nodular pearlitic	Short term tensile	(Room 59·4) 44·5	([917] 687)	36·0	(556)	27·4	(423)	16·2	(250)
	Short term 0·1% proof	(Room 34·8) 30·7	([537] 474)	28·2	(436)	19·2	(297)	7–11	(108–170)
	Total creep for 0·1% 100 000 h	7–9	(108–139)	2–3	(31–46)	—	—	—	—
Nodular ferritic	Stress to rupture in 100 000 h	12–14	(185–216)	7–8	(108–124)	2–3	(31–46)	—	—
	Tensile short term	—	—	3·8–6·5 (427°C)	(59–100)	1·79	(28)	0·6	9
	Total creep for 100 000 h 0·1%	5–6	(77–93)	1–1·5	(15–23)	—	—	—	—

Table 183d (*continued*)

Design stresses‡ at elevated temperatures

Grade		*350°C* ton/in²	(N/mm²)	*400°C* ton/in²	(N/mm²)	*500°C* ton/in²	(N/mm²)	*600°C* ton/in²	(N/mm²)
10–12	Tensile	2·5	(39)	0·75–1	(12–15)	0·75	(12)	<0·5	(<8)
	Compression	7–10	(108–154)	<7	(<108)	<7	(<108)	<4·5	(<69)
14–17	Tensile	3	(46)	1–2	(15–31)	1	(15)	<0·4	(<6)
	Compression	8–12	(124–185)	<7	(<108)	<7	(<108)	<4·5	(<69)
20–26	Tensile	5–6	(77–93)	1–2	(15–31)	1–1·5	(15–23)	<1	(<15)
	Compression	[12–18]	[185–278]	<7	(<108)	<7	(<108)	<4·5	(<69)
Whiteheart malleable	Tensile	3–4	(46–62)	2·5	(39)	0·25	(4)	—	—
Blackheart malleable	Tensile	3	(46)	2·5	(39)	(0·25)	(4)	—	—
Pearlitic malleable	Tensile	5	(77)	2·5	(39)	0·25	(4)	—	—
Nodular pearlitic	Tensile	6–8	(93–124)	<6	(<93)	<0·75	(<12)	<0·5	(<8)
	Compression	[12–18]	[185–278]	—	—	—	—	—	—
Nodular ferritic	Tensile	4	(62)	2–3	(31–46)	<1·8	(<28)	<0·6	(<9)
	Compression	8–10	(124–154)	—	—	—	—	(<4·5)	(<69)

Values in square brackets are interpolated. Most of the test data figures are taken from the section on heat resistance (pp. 274–276). The design stress values at 350°C are $\frac{1}{3}$ of the stress to rupture values.

‡*Important*. Factors of safety to be used with these stresses must be related to conditions of service.

to use a nodular or spheroidal graphite cast iron with a much higher fatigue strength.

It is not infrequently possible to stress a component well above the limiting fatigue stress if for some other reason the life of the component is short. Thus, for example, in rolling mill practice, the machined surface of the roll may be worn out in less than 5 million stressed revolutions, and the bearing necks may therefore be stressed above the limiting stress, depending upon the iron and the likely number of stress reversals. Surface rolling of fillets may increase both the limiting stress and the maximum stress to fracture for a definite number of reversals (*see* section on fatigue).

For alternating direct tensile/compression stress, the safe range of stress appears to be lower than in reversed bending, and an arbitrary value of 0·5 × minimum fatigue limit has been applied to all grades.

In compression, an arbitrary value of one-third of the maximum compressive stress has been employed. This figure is well inside the 0·1 per cent proof

Table 184

	Working stresses ton/in² (N/mm²)			
	Tension	*Compression*	*Shearing*	*Bearing*
Cast iron	1·5 (23)	8·0 (124)	1·5 (23)	10·0 (154)
Wrought iron	5·0 (77)	5·0 (77)	4·0 (62)	7·0 (108)
Mild steel	7·5 (116)	7·5 (116)	5·5 (85)	11·0 (170)

stress and is based upon British girder bridge practice which permits compressive loadings up to 10 ton/in² (154 N/mm²) for a BS 1452 Grade 10 iron, which will have a compression strength of 30 to 40 ton/in² (463 to 618 N/mm²).

Compressive stresses, however, cannot be dissociated from the 'slenderness' of the component, since with long columns deflection may occur which may put one side of the component in tension. The figures quoted for compression stress, therefore, apply only to short components in which the length does not much exceed the minimum lateral dimensions by about 2 to 1. This problem is dealt with in greater detail in the section on columns.

Compression loading applied repeatedly without undue impact, as in dies used in automatic presses, can cause surface deformation or spreading to occur, particularly in the case of soft irons, and for this purpose a strong shock-resistant iron with a low hardness, such as a low-phosphorus grey iron or a malleable iron, may give poorer results than weaker and possibly more brittle iron with higher hardness. Higher hardness and, therefore, resistance to compression distortion, may be obtained by local densening, by higher phosphorus and lower carbon contents, by alloying with chromium or molybdenum, or by surface hardening.

Where such conditions of repeatedly applied stress arise, the maximum compression loading should not exceed one half of the figures given in the final column of *Table 183c*.

High temperatures

At elevated temperatures the stress to rupture in 100 000 h at 350°C has been taken as the criterion and one third of this value has been taken as the maximum design stress. Nevertheless, in spite of the low values, the grey irons have appreciable load-carrying capacity up to about 750°C, probably of the order of 200 to 300 lb/in^2 (1·4 to 2·1 N/mm^2) in compression, and 30 per cent chromium oxidation resistant iron is quite capable of supporting its own weight at temperatures of 1000 to 1020°C in structures or plates in which the base compression load is 20 lb/in^2 (0·14 N/mm^2). Molybdenum acts as a pearlite stabiliser and ferritic strengthener and may be used on grey, malleable or nodular iron to increase the creep strength in stress-carrying castings operating at or above 350°C.

Low temperatures

No attempt has been made to summarise safe loadings at temperatures below normal. In general, the grey irons suffer only a gradual reduction in impact strength and a slight increase in tensile strength as the temperature falls.

The more ductile materials, such as steel, malleable iron, nodular iron or spheroidal graphite iron, however, can show marked reduction in ductility and shock resistance at low temperatures. Provided it is known, however, that they are to be subjected to low temperatures, all these materials can be provided in qualities which have very low ductile/brittle transition temperatures.

General considerations

The foregoing gives some idea of the functional properties that may be called for. These, however, cannot be dissociated from the economic requirements, and the cheapest material may well prove the most expensive in service in some circumstances, while in other cases a short service life may be justified by low cost and ease of replacement.

Appendix

Standard specifications for grey, nodular and malleable cast iron

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* in	mm	*Diameter of as-cast test bar* in	mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Suggested Brinell hardness*
Great Britain BS1452: 1961	10	$<\frac{3}{8}$*	<9·5	0·6*	15·2	170	17·3	11·0*	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	162	16·5	10·5	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	154	15·7	10·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	147	15·0	9·5	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	139	14·2	9·0	—
	12	$<\frac{3}{8}$*	<9·5	0·6*	15·2	201	20·5	13·0	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	193	19·7	12·5	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	185	18·9	12·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	178	18·1	11·5	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	170	17·3	11·0	—
	14	$<\frac{3}{8}$*	<9·5	0·6*	15·2	247	25·2	16·0	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	232	23·6	15·0	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	216	22·0	14·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	209	21·3	13·5	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	201	20·5	13·0	—
	17	$<\frac{3}{8}$*	<9·5	0·6*	15·2	293	30·0	19·0	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	278	28·3	18·0	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	263	26·8	17·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	247	25·2	16·0	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	232	23·6	15·0	—
	20	$<\frac{3}{8}$*	<9·5	0·6*	15·2	340	34·6	22·0	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	324	33·1	21·0	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	309	31·5	20·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	293	30·0	19·0	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	278	28·3	18·0	—

*Specifications in columns throughout Appendix marked with asterisk may have changed to SI units from metric or imperial ton and inch units since these figures were printed. Where tonf/in² units are quoted the imperial ton of 2240 lb is used.

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* in	mm	*Diameter of as-cast test bar* in	mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Suggested Brinell hardness*
	23	$<\frac{3}{8}$*	<9·5	0·6*	15·2	386	39·4	25·0*	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	371	37·8	24·0	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	355	36·2	23·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	340	34·6	22·0	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	324	33·1	21·0	—
	26	$<\frac{3}{8}$*	<9·5	0·6*	15·2	432	44·1	28·0	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	417	42·5	27·0	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	402	41·0	26·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	386	39·4	25·0	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	371	37·8	24·0	—
Argentina IRAM 629	Fg14	—	—	—	—	137	14*	8·9	187 max.
	Fg17	—	—	—	—	167	17	10·8	165–200
	Fg20	—	—	—	—	196	20	12·7	170–223
	Fg24	—	—	—	—	235	24	15·2	187–241
	Fg28	—	—	—	—	275	28	17·8	202–255
	Fg35	—	—	—	—	343	35	22·2	233–290
	Fg42	—	—	—	—	412	42	26·7	250–310
Australia G8–1969	10	$<\frac{3}{8}$*	<9·5	0·6*	15·2	170	17·3	11·0*	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	162	16·5	10·5	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	154	15·7	10·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	147	15·0	9·5	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	139	14·2	9·0	—
	12	$<\frac{3}{8}$*	<9·5	0·6*	15·2	201	20·5	13·0	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	193	19·7	12·5	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	185	18·9	12·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	178	18·1	11·5	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	170	17·3	11·0	—
	14	$<\frac{3}{8}$*	<9·5	0·6*	15·2	247	25·2	16·0	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	232	23·6	15·0	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	216	22·0	14·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	209	21·3	13·5	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	201	20·5	13·0	—
	15.5	$<\frac{3}{8}$*	<9·5	0·6*	15·2	270	27·6	17·5	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	255	26·0	16·5	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	239	24·4	15·5	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	224	22·8	14·5	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	216	22·0	14·0	—

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* in	mm	*Diameter of as-cast test bar* in	mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Suggested Brinell hardness*
Australia	17	$<\frac{3}{8}$*	<9·5	0·6*	15·2	293	30·0	19·0	—
G8–1969		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	278	28·3	18·0	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	263	26·8	17·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	247	25·2	16·0	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	232	23·6	15·0	—
	20	$<\frac{3}{8}$*	<9·5	0·6*	15·2	340	34·6	22·0	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	324	33·1	21·0	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	309	31·5	20·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	293	30·0	19·0	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	278	28·3	18·0	—
	23	$<\frac{3}{8}$*	<9·5	0·6*	15·2	386	39·4	25·0	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	371	37·8	24·0	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	355	36·2	23·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	340	34·6	22·0	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	324	33·1	21·0	—
	26	$<\frac{3}{8}$*	<9·5	0·6*	15·2	432	44·1	28·0	—
		$\frac{3}{8}$–$\frac{3}{4}$	9·5–19·0	0·875	22·2	417	42·5	27·0	—
		$\frac{3}{4}$–$1\frac{1}{8}$	19·0–29·6	1·2	30·5	402	41·0	26·0	—
		$1\frac{1}{8}$–$1\frac{5}{8}$	29·6–41·3	1·6	40·6	386	39·4	25·0	—
		$>1\frac{5}{8}$	>41·3	2·1	53·3	371	37·8	24·0	—
Austria	GG10		—	30		98	10*	6·3	—
Ö NORM	GG15		—	30		147	15	9·5	—
M 3191	GG20		—	30		196	20	12·7	—
1971	GG25		—	30		245	25	15·9	—
	GG30		—	30		294	30	19·0	—
	GG35		—	30		343	35	22·2	
Belgium	FGG00		—	Mechanical properties not specified					
NBN	FGG10		—	30		98	10*	6·3	—
830–01	FGG15		—	30		147	15	9·5	—
1970	FGG20		—	30		196	20	12·7	—
	FGG25		—	30		245	25	15·9	—
	FGG30		—	30		294	30	19·0	—
	FGG35		—	30		343	35	22·2	—
	FGG40		—	30		392	40	25·4	—
Czechoslovakia	42 24 10		4–30	30		98	10*	6·3	180
CSN 42	42 24 15		4–30	30		147	15	9·5	200
1241	42 24 20		8–30	30		196	20	12·7	220
1964	42 24 25		15–45	30		245	25	15·9	240
	42 24 30		15–80	30		294	30	19·0	260
	42 24 35		15–80	30		343	35	22·2	270

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* mm	*Diameter of as-cast test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Suggested Brinell hardness*
East	GGL00	—	Mechanical properties not specified				
Germany	GGL15	—	30	147	15*	9·5	200
(German	GGL20	—	30	196	20	12·7	225
Democratic	GGL25	—	30	245	25	15·9	245
Republic)	GGL30	—	30	294	30	19·0	260
TGL	GGL35	—	30	343	35	22·2	275
14 400	GGL40	—	30	392	40	25·4	285
1962							
Finland	GRS00	—	Mechanical properties not specified				
H.II.51	GRS10	—	30	98	10*	6·3	—
1965	GRS15	—	30	147	15	9·5	180 max.
	GRS20	—	30	196	20	12·7	—
	GRS25	—	30	245	25	15·9	—
	GRS30	—	30	294	30	19·0	—
	GRS35	—	30	343	35	22·2	—
	GRS40	—	30	392	40	25·4	—
France	Ft 10 D	13–22	30	100*	10·2	6·5	—
NF A32–101	E	22–40	50				
1965	S	>40	(1)				
	Ft 15 B	6–9	13	150	15·3	9·7	—
	C	9–13	20				
	D	13–22	30				
	E	22–40	50				
	S	>40	(1)				
	Ft 20 A	<6	9	200	20·4	12·9	—
	B	6–9	13				
	C	9–13	20				
	D	13–22	30				
	E	22–40	50				
	S	>40	(1)				
	Ft 25 A	<6	9	250	25·2	16·2	—
	B	6–9	13				
	C	9–13	20				
	D	13–22	30				
	E	22–40	50				
	S	>40	(1)				
	Ft 30 A	<6	9	300	30·6	19·4	—
	B	6–9	13				
	C	9–13	20				
	D	13–22	30				
	E	22–40	50				
	S	>40	(1)				

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* mm	*Diameter of as-cast test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Suggested Brinell hardness*
France	Ft 35 A	<6	9	350*	35·7	22·7	—
NF A32–101	B	6–9	13				
1965	C	9–13	20				
	D	13–22	30				
	E	22–40	50				
	S	>40	(1)				
	Ft 40 B	6–9	13	400	40·8	25·8	—
	C	9–13	20				
	D	13–22	30				
	S	>40	(1)				
	(The dimensions of bar S are to be agreed between the parties)						
Hungary	Öv10	—	—	98	10*	6·3	—
MSZ 8280–66	Öv15	10–20	30	147	15	9·5	212 max.
C81	Öv20	10–20	30	196	20	12·7	225 max.
1968	Öv25	10–20	30	245	25	15·9	245 max.
	Öv30	10–20	30	294	30	19·0	260 max.
	Öv35	10–20	30	343	35	22·2	277 max.
	Öv40	10–20	30	392	40	25·4	292 max.
India	15	4–8*	13	186	19*	12·1	130–180
IS 210–		8–15	20	167	17	10·8	
1970		15–30	30	147	15	9·5	
		30–50	45	127	13	8·3	
	20	4–8	13	235	24	15·2	160–220
		8–15	20	216	22	14·0	
		15–30	30	196	20	12·7	
		30–50	45	167	17	10·8	
	25	4–8	13	275	28	17·8	180–230
		8–15	20	255	26	16·5	
		15–30	30	245	25	15·9	
		30–50	45	216	22	14·0	
	30	8–15	20	304	31	19·7	180–230
		15–30	30	294	30	19·0	
		30–50	45	265	27	17·1	
	35	15–30	30	343	35	22·2	207–241
		30–50	45	314	32	20·3	
	40	15–30	30	392	40	25·4	207–270
		30–50	45	363	37	23·5	

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* mm	*Diameter of as-cast test bar* mm	*Minimum tensile strength* N/mm^2	kgf/mm^2	$tonf/in^2$	*Suggested Brinell hardness*
Italy	G00		Mechanical properties not specified				
UNI 5007	G10	3·5–7·5	30	157	16*	10·2	215
1969		7·5–15	30	98	10	6·3	175
	G15	3·5–7·5	30	206	21	13·3	225
		7·5–15	30	147	15	9·5	185
		15–30	30	127	13	8·3	165
	G20	3·5–7·5	30	255	26	16·5	235
		7·5–15	30	196	20	12·7	205
		15–30	30	147	15	9·5	175
		30–50	30	118	12	7·6	155
	G25	3·5–7·5	30	304	31	19·7	250
		7·5–15	30	245	25	15·9	220
		15–30	30	206	21	13·3	195
		30–50	30	157	16	10·2	170
	G30	7·5–15	30	294	30	19·0	235
		15–30	30	265	27	17·1	215
		30–50	30	206	21	13·3	190
	G35	7·5–15	30	343	35	22·2	255
		15–30	30	314	32	20·3	235
		30–50	30	265	27	17·1	210
Japan	FC10	4–50	30	98	10*	6·3	201 max.
JIS	(Grade 1)						
G5501	FC15	4–8	13	186	19	12·1	241 max.
1972	(Grade 2)	8–15	20	167	17	10·8	223 max.
		15–30	30	147	15	9·5	212 max.
		30–50	45	127	13	8·3	201 max.
	FC20	4–8	13	235	24	15·2	255 max.
	(Grade 3)	8–15	20	216	22	14·0	235 max.
		15–30	30	196	20	12·7	223 max.
		30–50	45	167	17	10·8	217 max.
	FC25	4–8	13	275	28	17·8	269 max.
	(Grade 4)	8–15	20	255	26	16·5	248 max.
		15–30	30	245	25	15·9	241 max.
		30–50	45	216	22	14·0	229 max.
	FC30	8–15	20	304	31	19·7	269 max.
	(Grade 5)	15–30	30	294	30	19·0	262 max.
		30–50	45	265	27	17·1	248 max.
	FC35	15–30	30	343	35	22·2	277
	(Grade 6)	30–50	45	314	32	20·3	269

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* mm	*Diameter of as-cast test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Suggested Brinell hardness*
Netherlands	GG15	—	30	147	15*	9·5	180
NEN	GG20	—	30	196	20	12·7	200
6002 A	GG25	—	30	245	25	15·9	215
1966	GG30	—	30	294	30	19·0	230
	GG35	—	30	343	35	22·2	240
	GG40	—	30	392	40	25·4	—
Norway	SjG00	—	Mechanical properties not specified				—
NS 722	SjG10	—	Tensile strength not specified				180 max.
1963	SjG15	—	30	147	15*	9·5	—
	SjG20	—	30	196	20	12·7	—
	SjG25	—	30	245	25	15·9	—
	SjG30	—	30	294	30	19·0	—
	SjG35	—	30	343	35	22·2	—
	SjG40	—	30	392	40	25·4	—
Poland	Z1X	—	Mechanical properties not specified				—
PN 63	Z110	—	30	98	10*	6·3	—
H–83101	Z115	—	30	147	15	9·5	159–212
1963	Z120	—	30	196	20	12·7	170–241
	Z125	—	30	245	25	15·9	183–262
	Z130	—	30	294	30	19·0	192–269
	Z135	—	30	343	35	22·2	201–277
	Z140	—	30	392	40	25·4	> 223
Romania	Fc10	—	30	98	10*	6·3	120–187
STAS	Fc15	—	30	147	15	9·5	143–229
568–67	Fc20	—	30	196	20	12·7	156–285
1968	Fc25	—	30	245	25	15·9	170–302
	Fc30	—	30	294	30	19·0	187–321
	Fc35	—	30	343	35	22·2	207–321
	Fc40	—	30	392	40	25·4	207–363
Spain	F1.0.10	—	30	98	10*	6·3	—
UNE	F1.0.15	—	30	147	15	9·5	—
36 111	F1.0.20	—	30	196	20	12·7	—
	F1.0.25	—	30	245	25	15·9	—
	F1.0.30	—	30	294	30	19·0	—
	F1.0.35	—	30	343	35	22·2	—
	F1.0.40	—	30	392	40	25·4	—

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* mm	*Diameter of as-cast test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Suggested Brinell hardness*
Sweden	00–00	—	Mechanical properties not specified				—
SIS	10–00	—	30	98	10*	6·3	180 max.
14 01	15–00	—	30	147	15	9·5	190
1971	20–00	—	30	196	20	12·7	210
	25–00	—	30	245	25	15·9	230
	30–00	—	30	294	30	19·0	250
	35–00	—	30	343	35	22·2	260
	40–00	—	30	392	40	25·4	—
Switzerland	Ft/GG10	3–8	30	98	10*	6·3	—
VSM	Ft/GG15	5	30	147	15	9·5	—
10 691	Ft/GG20	>10	30	196	20	12·7	—
1959	Ft/GG25	>20	30	245	25	15·9	—
	Ft/GG30	>30	30	294	30	19·0	—
Turkey	DDL–10	—	30	98	10*	6·3	—
TS 552	DDL–15	—	30	147	15	9·5	—
1967	DDL–20	—	30	196	20	12·7	—
	DDL–25	—	30	245	25	15·9	—
	DDL–30	—	30	294	30	19·0	—
	DDL–35	—	30	343	35	22·2	—
	DDL–40	—	30	392	40	25·5	—

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* in	mm	*Diameter of as-cast test bar* in	mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	ksi†	*Suggested Brinell hardness*
United	20A	½*	6–12	0·875*	22·4	138	14·0	8·9	20*	—
States	B	½–1	13–25	1·2	30·5					
ASTM	C	1–2	26–50	2·0	50·8					
A48–64 1971	S		<6 or >50		Bar S[a]					
	25A	½*	6–12	0·875*	22·4	172	17·5	11·1	25	—
	B	½–1	13–25	1·2	30·5					
	C	1–2	26–50	2·0	50·8					
	S		<6 or >50		Bar S[a]					

†ksi = kilopounds per square inch.
[a] All dimensions of test bar S shall be agreed upon between the manufacturer and the purchaser.
[b] These irons are specified on hardness and microstructure. The mechanical properties are material descriptions only and not part of the standard.

Grey cast iron										
Country and specification	*Grade*	*Nominal section thickness* in	mm	*Diameter of as-cast test bar* in	mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	ksi†	*Suggested Brinell hardness*
United States ASTM A48–64 1971	30A	½*	6–12	0·875*	22·4	207	21·1	13·4	30*	—
	B	½–1	13–25	1·2	30·5					
	C	1–2	26–50	2·0	50·8					
	S		<6 or >50		Bar S[a]					
	35A	½*	6–12	0·875*	22·4	241	24·6	15·6	35	—
	B	½–1	13–25	1·2	30·5					
	C	1–2	26–50	2·0	50·8					
	S		<6 or >50		Bar S[a]					
	40A	½*	6–12	0·875*	22·4	276	28·2	17·9	40	—
	B	½–1	13–25	1·2	30·5					
	C	1–2	26–50	2·0	50·8					
	S		<6 or >50		Bar S[a]					
	45A	½*	6–12	0·875*	22·4	310	33·1	20·1	45	—
	B	½–1	13–25	1·2	30·5					
	C	1–2	26–50	2·0	50·8					
	S		<6 or >50		Bar S[a]					
	50A	½*	6–12	0·875*	22·4	345	35·1	22·3	50	—
	B	½–1	13–25	1·2	30·5					
	C	1–2	26–50	2·0	50·8					
	S		<6 or >50		Bar S[a]					
	55A	½*	6–12	0·875*	22·4	379	38·6	24·5	55	—
	B	½–1	13–25	1·2	30·5					
	C	1–2	26–50	2·0	50·8					
	S		<6 or >50		Bar S[a]					
	60A	½*	6–12	0·875*	22·4	414	42·2	26·8	60	—
	B	½–1	13–25	1·2	30·5					
	C	1–2	26–50	2·0	50·8					
	S		<6 or >50		Bar S[a]					
ASTM A159–72[b] (automotive grey iron castings) SAE J431b	G1800	—	—	—	—	(137)	(14)	(8·9)	(18)*	187 max.
	G2500	—	—	—	—	(172)	(17·5)	(11·1)	(25)	170–229
	G3000	—	—	—	—	(206)	(21)	(13·3)	(30)	187–241
	G3500	—	—	—	—	(240)	(24·5)	(15·5)	(35)	207–255
	G4000	—	—	—	—	(275)	(28)	(17·8)	(40)	217–269
	G2500a	—	—	—	—	(172)	(17·5)	(11·1)	(25)	170–229
	G3500b	—	—	—	—	(240)	(24·5)	(15·5)	(35)	207–255
	G3500c	—	—	—	—	(240)	(24·5)	(15·5)	(35)	207–255

† ksi = kilopounds per square inch.
[a] All dimensions of test bar S shall be agreed upon between the manufacturer and the purchaser.
[b] These irons are specified on hardness nad microstructure. The mechanical properties are material descriptions only and not part of the standard.

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* mm	*Diameter of as-cast test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Suggested Brinell hardness*
USSR	SCh00		Mechanical properties not specified				—
GOST	SCh12–28	—	30	118	12*	7·6	143–229
1412–54		> 50	50				
1967	SCh15–32	—	30	147	15	9·5	163–229
		> 50	50				
	SCh18–36	—	30	177	18	11·4	170–229
		> 50	50				
	SCh21–40	—	30	206	21	13·3	170–241
		> 50	50				
	SCh24–44	—	30	235	24	15·2	170–241
		> 50	50				
	SCh28–48	—	30	275	28	17·8	170–241
		> 50	50				
	SCh32–52		30	314	32	20·3	187–255
			50				
	SCh35–56		30	343	35	22·2	197–269
			50				
	SCh38–60		30	373	38	24·1	207–269
			50				
West	GG10		30	98	10*	6·3	—
Germany	GG15		13	226	23	14·6	—
DIN			20	177	18	11·4	—
1691			30	147	15	9·5	—
1964			45	108	11	7·0	—
	GG20		13	275	28	17·8	—
			20	226	23	14·6	—
			30	196	20	12·7	—
			45	157	16	10·2	—
	GG25		13	324	33	21·0	—
			20	275	28	17·8	—
			30	245	25	15·9	—
			45	206	21	13·3	—
	GG30		20	324	33	21·0	—
			30	294	30	19·0	—
			45	255	26	16·5	—
	GG35		20	373	38	24·1	—
			30	343	35	22·2	—
			45	304	31	19·7	—
	GG40		30	392	40	25·4	
			45	353	36	22·9	

Grey cast iron

Country and specification	*Grade*	*Nominal section thickness* mm	*Diameter of as-cast test bar* mm	*Minimum tensile strength*			*Suggested Brinell hardness*
				N/mm^2	kgf/mm^2	$tonf/in^2$	
Yugoslavia	SL 14	15–30	30	137	14*	8·9	160–230
JUS C J	18	15–30	30	177	18	11·4	170–230
2020	22	15–30	30	216	22	14·0	170–240
1958	26	15–30	30	255	26	16·5	170–240
	30	15–30	30	294	30	19·0	180–260
	(Type)						
International	10		30	98	10*	6·3	—
Organisation	15		30	147	15	9·5	—
for	20		30	196	20	12·7	—
Standard-	25		30	245	25	15·9	—
isation	30		30	294	30	19·0	—
ISO R185	35		30	343	35	22·2	—
1961	40		30	392	40	25·5	—

Nodular cast iron (see also Table 183b on p. 478)

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
Great Britain BS 2789 1973	370/17†	370*	37·7	24·0	230	23·4	14·9	17	< 179	F
	420/12	420	42·8	27·2	250	25·5	16·2	12	< 201	F
	500/7	500	51·0	32·4	310	31·6	20·1	7	170–241	P–F
	600/3	600	61·2	38·8	350	35·7	22·7	3	192–269	P–F
	700/2	700	71·4	45·3	400	40·8	25·9	2	229–302	P
	800/2	800	81·6	51·8	460	46·9	29·8	2	248–352	P

† Grade 370/17 has the following impact value at ambient temperatures BS 131 Parts 2 and 3:

	Mean value (3 tests) J	kgf.m/cm²	ft.lbf	*Individual value* J	kgf.m/cm²	ft.lbf
Charpy V-notch	13·0	1·3	9·4	12·0	1·2	8·7
Charpy U-notch (2 mm notch depth)	15·0	1·5	10·8	13·0	1·3	10·8

[a] *Note:* Throughout specifications for nodular cast iron the following abbreviations are used: F = ferritic, P = pearlitic, F–P = ferritic pearlitic, F^{+}_{+} = mainly ferritic, P^{+}_{+} = mainly pearlitic, S = sorbite.

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
Austria	SG42	412	42	26·7	275	28	17·8	12		F‡
Ö NORM	SG50	490	50	31·7	343	35	22·2	7		F P
M3193	SG60	588	60	38·1	412	42	26·7	2		P‡
1964	SG38 (special)	373	38	24·1	245	25	15·9	17		F‡
	SG70 (special)	686	70	44·4	490	50	31·7	2		P‡
Belgium	FNG	490	50	31·7				2		P–F
NBN	FNG 38–17†	373	38	24·1	235	24	15·2	17	140–180	F
830–02	FNG42–12	412	42	26·7	275	28	17·8	12	150–200	F
1970	FNG50–7	490	50	31·7	343	35	22·2	7	170–240	P–F
	FNG60–2	588	60	38·1	392	40	25·4	2	210–280	P
	FNG70–2	686	70	44·4	441	45	28·6	2	230–300	P‡
	FNG80–2	785	80	50·8	490	50	31·7	2	240–320	P–S

† Grade FNG 38–17 has the following impact values:

	Mean value (3 tests) J	kgf.m/cm²	ft.lbf	*Individual value* J	kgf.m/cm²	ft.lbf
Charpy V-notch	16·6	1·7	12·3	14·7	1·5	10·9
Charpy U-notch	18·6	1·9	13·7	15·7	1·6	11·6

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
Czechoslo-	42 2438	373	38*	24·1	245	25	15·9	17	140–200	—
vakia	42 2442	412	42	26·7	275	28	17·8	12	160–220	—
ČSN 42	42 2450	490	50	31·7	343	35	22·2	7	200–260	—
1242	42 2460	588	60	38·1	412	42	26·7	2	220–280	—
	42 2470	686	70	44·4	490	50	31·7	2	240–300	—
East	GGG 40	392	40*	25·4	275	28	17·8	12	150–190	F
Germany	GGG 45	—	45	—	343	35	22·2	—	160–240	F–P
TGL	GGG 50	490	50	31·7	343	35	22·2	5	180–240	F–P
8189	GGG 60	588	60	38·1	412	42	26·7	2	220–300	P
1962	GGG 70	686	70	44·4	490	50	31·7	2	260–320	P
Finland	GRP 38†	373	38*	24·1	245	25	15·9	17	—	F
SFS	GRP 40	392	40	25·4	275	28	17·8	12	—	F
2113	GRP 50	490	50	31·7	343	35	22·2	7	—	F–P
1966	GRP 60	588	60	38·1	392	40	25·4	2	—	P
	GRP 70	686	70	44·4	461	47	29·8	2	—	P

†Grade GRP 38 has the following impact values:

Mean value (3 tests) J	kgf.m/cm²	ft.lbf	*Individual value* J	kgf.m/cm²	ft.lbf
18·6	1·9	13·7	15·7	1·6	11·6

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
France	FGS38–15	372	38*	24·1	235	24	15·2	15	<200	F
NF A32–	FGS42–12	412	42	26·7	275	28	17·8	12	<220	F
201	FGS50–7	490	50	31·7	343	35	22·2	7	210–260	F–P
1966	FGS60–2	588	60	38·1	392	40	25·4	2	230–280	P
	FGS70–2	686	70	44·4	461	47	29·8	2	240–300	P
India	SG38/17†	373	38*	24·1	235	24	15·2	17	171 max.	—
IS 1865	SG42/12†	412	42	26·7	275	28	17·8	12	187 max.	—
1968	SG50/7	490	50	31·7	343	35	22·2	7	170–240	—
	SG60/2	588	60	38·1	392	40	25·4	2	210–280	—
	SG70/2	686	70	44·4	441	45	28·6	2	230–300	—
	SG80/2	784	80	50·8	471	48	30·5	2	260–330	—

† Grades SG38/17 and SG42/12 have the following impact values:

	Mean value (3 tests) J	kgf.m/cm²	ft.lbf	*Individual value* J	kgf.m/cm²	ft.lbf
SG38/17	16·6	1·7	12·3	14·7	1·5	10·9
SG42/12	8·8	0·9 (optional)	6·5	5·9	0·6 (optional)	4·3

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
Italy	GS42/10	412	42*	26·7	294	30	19·0	10	140–190	F^{+}_{+}
UNI	GSQ42/15†	412	42	26·7	275	28	17·8	15	130–170	F
4544	GS50/5	490	50	31·7	373	38	24·1	5	160–220	P–F
1960	GS55/2	539	55	34·9	392	40	25·4	2	190–250	P^{+}_{+}

† Grade GSQ42/15 has the following impact values:

Mean value (3 tests) J	kgf.m/cm²	ft.lbf	*Individual value* J	kgf.m/cm²	ft.lbf
19·6	2	14·4	14·7	1·5	10·8

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
Japan JIS G5502 1971	FCD40 (Class) 1)	392	40*	25·4	255	26	16·5	15	121–197	—
	FCD45 (Class 2)	441	45	28·6	294	30	19·0	10	143–217	—
	FCD50 (Class 3)	490	50	31·7	343	35	22·2	7	170–241	—
	FCD60 (Class 4)	588	60	38·1	392	40	25·4	2	207–285	—
	FCD70 (Class 5)	686	70	44·4	441	45	28·6	2	229–321	—
Netherlands NEN 6002–D 1966	GN	490–686	50–70*	31·7–44·4	343	35	22·2	2	170–320	P–F
	GN38†	373	38	24·1	245	25	15·9	17	<180	F
	GN42	412	42	26·7	275	28	17·8	12	<200	F
	GN50	490	50	31·7	343	35	22·2	7	170–240	F–P
	GN60	588	60	38·1	412	42	26·7	2	200–275	P
	GN70	686	70	44·4	490	50	31·7	2	220–320	P

†Grade GN38 has the following impact values at 20°C (mean of 3 tests):

J	kgf.m/cm²	ft.lbf
16.6	1·7	12·3

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
Norway	NS11 388†	373	38*	24·1	235	24	15·2	20	<150	—
NS 11	NS11 342	412	42	26·7	275	28	17·8	15	140–180	—
301	NS11 350	490	50	31·7	343	35	22·2	7	170–230	—
1971	NS11 360	588	60	38·1	392	40	25·4	5	200–260	—
	NS11 370	686	70	44·4	441	45	28·6	3	230–300	—

† Grade NS11 338 has the following impact values:

Mean value (3 tests) J	kgf.m/cm²	ft.lbf	*Individual value* J	kgf.m/cm²	ft.lbf
18·6	1·9	13·7	15·7	1·6	11·6

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
Poland	ZsF05	392–490	40–50*	26–32	—	—	—	5	155–210	F
PN–62	ZsF10	392–490	40–50	26–32	—	—	—	10	155–210	F
H–83123	ZsP45f†	441	45	28·6	—	—	—	—	260–300	P
1962	ZsP50f†	490	50	31·7	—	—	—	—	260–300	P
	ZsP45	441	45	28·6	—	—	—	2	200–280	P
	ZsP50	490	50	31·7	—	—	—	2	200–280	P
	ZsP55	539	55	34·9	—	—	—	2	200–280	P
	ZsP60	588	60	38·1	—	—	—	2	200–300	P
Romania	Fgn38–11	373	38*	24·1	235	24	15·2	11	140–180	F
STAS	Fgn42–12	412	42	26·7	274	28	17·8	12	150–200	F
6071–	Fgn45–5	441	45	28·6	313	32	20·3	5	160–240	F
70	Fgn50–7	490	50	31·7	392	40	25·4	7	170–240	F + P
1970	Fgn60–22	588	60	38·1	392	40	25·4	2	210–280	P
	Fgn70–2	686	70	44·4	460	47	29·8	2	230–300	P

† P content 0·2–0·3 per cent; other grades 0·15 per cent max.

Impact values

	daJ/cm²	kgf.m/cm²	ft.lbf
Fgn38–11	1·6	1·6	11·6
Fgn42–12	2·0	2·0	14·4
Fgn45–5	1·5	1·5	10·8
Fgn50–7	—	—	—
Fgn60–22	1·5	1·5	10·8
Fgn70–2	2·0	2·0	14·4

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
South	SG38	373	38*	23·8	245	25	15·5	17	180	F
Africa	SG42	412	42	26·4	275	28	17·5	12	200	F
SABS 936	SG50	490	50	31·4	345	35	21·9	7	170–240	F–P
1969	SG60	588	60	37·8	390	40	25·0	4	210–280	P
(amended	SG70	686	70	44·0	440	45	28·3	3	230–300	P
1970)	SG80	785	80	50·5	490	50	31·4	2	265–330	tempered martensite or bainite
Sweden	17–02	392	40	25·4	245	25	15·9	18	130–170	—
SIS	17–12	373	38	24·1	245	25	15·9	20	160 max.	—
1407	17–15	343	35	22·2	216	22	14·0	22	150 max.	—
1971	27–02	490	50	31·7	343	35	22·2	7	170–230	—
	32–03	588	60	38·1	392	40	25·4	5	200–260	—
	37–01	686	70	74·4	441	45	28·6	3	230–300	—

Impact values

	J	kgf.m/cm²	ft.lbf	
17–12	19·6	2	14·5	(−20°C)
17–15	19·6	2	14·5	(−40°C)

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
Turkey	DDK–38	373	38*	23·8	245	25	15·9	17	—	F
TS 526	DDK–42	412	42	26·4	275	28	17·5	12	—	F
1967	DDK–45	441	45	28·6	343	35	22·2	5	—	F
	DDK–50	490	50	31·7	343	35	22·2	7	—	F–P
	DDK–60	588	60	38·1	412	42	26·7	2	—	F–P
	DDK–70	686	70	44·4	490	50	31·7	2	—	P

Country and Specification	*Grade*	*Minimum Tensile Strength* N/mm²	kgf/mm²	tonf/in²	psi	*Proof Stress (min).* N/mm²	kgf/mm²	tonf/in²	psi	*Elongation (min.)* %	*Hardness HB*	*Structure (a)*
USA	60–40–18	414	42·2	26·8	60 000	276	28·1	17·9	40 000*	18	—	—
A536–	65–45–12	448	45·7	29·0	65 000	310	31·6	20·1	45 000	12	—	—
72	80–55–06	552	56·3	35·7	80 000	379	38·6	24·5	55 000	6	—	—
1972	100–70–03	689	70·2	44·6	100 000	483	49·2	31·2	70 000	3	—	—
	120–90–02	827	84·3	53·6	120 000	621	63·3	40·2	90 000	2	—	—
	†80–60–03	552	56·3	35·7	80 000	414	42·2	26·8	60 000	3	—	—
USA	D4018	(414)	(42·0)	(26·8)	(60 000)	(276)	(28·0)	(17·9)	(40 000)*	18	170 max.	F
AE434b†	D4512	(448)	(45·6)	(29·0)	(65 000)	(310)	(31·5)	(20·1)	(45 000)	12	156–217	F–P
(automotive	D5506	(552)	(56·0)	(35·7)	(80 000)	(379)	(39·0)	(24·5)	(55 000)	6	187–255	F–P
ductile iron	D7003	(689)	(70·0)	(44·6)	(100 000)	(483)	(49·0)	(31·2)	(70 000)	3	241–302	P
castings)	DQ&T‡	—	—	—	—	—	—	—	—	—		Marten-sitic

† Non-standard grade which is widely used.

† These irons are primarily specified on hardness and structure. The mechanical properties are material descriptions.
‡ Quenched and temperated grade. Hardness to be agreed between supplier and purchaser.

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
USSR	VCh45–0	441	45*	28·6	353	36	22·9	—	187–255	—
GOST	VCh50–1·5	490	50	31·7	372	38	24·1	1·5	187–255	—
7293–	VCh60–2	588	60	38·1	412	42	26·7	2·0	197–269	—
54	VCh45–5	441	45	28·6	324	33	21	5·0	170–207	—
1954	VCh40–10	392	40	25·4	294	30	19	10·0	156–197	—

Impact values

	J	kgf.m/cm²	ft.lbf
VCh50–1·5	14·7	1·5	10·9
VCh60–2	14·7	1·5	10·9
VCh45–5	19·6	2·0	14·5
VCh40–10	29·4	3·0	21·8

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
West	GGG–40	400*	40·8	25·9	250	25·5	16·2	15	120–180	F
Germany	GGG–50	500	51·0	32·4	320	32·6	20·7	7	150–240	F–P
DIN	GGG–60	600	61·2	38·8	380	38·7	24·6	3	175–290	F–P
1693	GGG–70	700	71·4	45·3	440	44·9	28·5	2	210–320	P
1973	GGG–80	800	81·6	51·8	500	51·0	32·4	2	230–360	P
	GGG–35·3†	350	35·7	22·7	220	22·4	14·2	22	—	—
	GGG–40·3†	400	40·8	25·9	250	25·5	16·2	18	—	—

† Grades with guaranteed impact properties.

Impact values

	Mean value (3 tests) J	kgf.m/cm²	ft.lbf	*Individual values* J	kgf.m/cm²	ft.lbf
At room temperature:						
GGG–35·3	19	1·9	14·0	17	1·7	12·5
GGG–40·3	16	1·6	11·8	14	1·4	10·3
At lower temperatures:						
GGG–35·3 (at −40°C)	14	1·4	10·3	11	1·1	8·1
GGG–40·3 (at −20°C)	14	1·4	10·3	11	1·1	8·1

Nodular cast iron

Country and specification	*Grade*	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*Proof stress (min.)* N/mm²	kgf/mm²	tonf/in²	*Elongation (min.)* %	*Hardness HB*	*Structure*[a]
International	38–17†	373	38	24·1	235	24	15·2	17	140–180	F
Organisation	42–12	412	42	26·7	275	28	17·8	12	150–200	F
for	50–7	490	50	31·7	343	35	22·2	7	170–240	F–P
Standardisa-	60–2	588	60	38·1	392	40	25·4	2	210–280	P
tion (ISO)	70–2	686	70	44·4	441	45	29·6	2	230–300	P
R1083										
1969										

† Type 38–17 has the following impact values:

	Mean value (3 tests) J	kgf.m/cm²	ft.lbf	*Individual value* J	kgf.m/cm²	ft.lbf
V—notch (Charpy)	16·6	1·7	12·3	14·7	1·5	10·9
U—notch (Mesnager)	18·6	1·9	13·7	15·7	1·6	11·6

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
Great	W340/3	9	270*	27·5	17·5	—	—	—	7	—
Britain		12	310	31·6	20·1	—	—	—	4	—
BS 309:1972		15	340	34·7	22·0	—	—	—	3	—
(Whiteheart)	W410/4	9	350	35·7	22·7	190	19·4	12·3	10	—
		12	390	39·8	25·2	230	23·5	14·9	6	—
		15	410	41·8	26·5	250	25·5	16·2	4	—
BS 310:1972	B290/6	15	290	29·6	18·8	170	17·3	11·0	6	—
(Blackheart)	B310/10	15	310	31·6	20·1	190	19·4	12·3	10	—
	B340/12	15	340	34·7	22·0	200	20·4	12·9	12	—
BS3333:1972	P440/7	15	440	44·9	28·5	270	27·5	17·5	7	(149–197)
(Pearlitic)	P510/4	15	510	52·0	33·0	310	31·6	20·1	4	(170–229)
	P540/5	15	540	55·1	35·0	340	34·7	22·0	5	(179–229)
	P570/3	15	570	58·1	36·9	420	42·8	27·2	3	(197–241)
	P690/2	15	690	70·4	44·7	540	55·1	35·0	2	(241–285)

Values are also specified for 0·2% proof stress as follows:

		Dia.	N/mm²	kgf/mm²	tonf/in²
BS 309:1972	W410/4	9	170*	17·3	11·0
		12	210	21·4	13·6
		15	220	22·4	14·2
BS 310:1972	B310/10	—	180	18·4	11·7
	B340/12	—	190	19·4	12·3
BS3333:1972	P440/7	—	250	25·5	16·2
	P510/4	—	290	29·6	18·8
	P540/5	—	320	32·6	20·7
	P570/3	—	400	40·8	25·9
	P690/2	—	520	53·0	33·7

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
Argentina IRAM 526 1948 (Whiteheart)	Fm32a	—	314	32*	20·3	177	18	11·4	2	(125–190)
	Fm38	—	373	38	24·1	206	21	13·3	4	(125–190)
(Blackheart)	Fm32b	—	314	32	20·3	177	18	11·4	2	(110–140)
	Fm35	—	343	35	22·2	186	19	12·1	9	(110–140)
	Fm37	—	363	37	23·5	245	25	15·9	†	(110–140)

† To be agreed between supplier and purchaser.

Country and Specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	lbf/in²	*0·5 Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	lbf/in²	*Elongation,* % (min.)	*Hardness*
Australia AS G14 1969 (Pearlitic)	PM63–40–09	14·3	434	44·3	28·1	63 000	276	28·1	17·9	40 000*	9	152–187
	PM68–45–07	14·3	469	47·8	30·4	68 000	310	31·6	20·1	45 000	7	163–217
	PM70–48–05	14·3	483	49·2	31·2	70 000	331	33·8	21·4	48 000	5	170–229
	PM75–50–07	14·3	517	52·7	33·5	75 000	345	35·2	22·3	50 000	7	179–229
	PM80–60–03	14·3	552	56·2	35·7	80 000	414	42·2	26·8	60 000	3	197–255
	PM90–70–02	14·3	621	63·3	40·2	90 000	483	49·2	31·2	70 000	2	229–269
	PM100–80–02	14·3	689	70·3	44·6	100 000	552	56·2	35·7	80 000	2	241–269
	PM105–85–02	14·3	724	73·8	46·9	105 000	586	59·8	37·9	85 000	2	269–302

Whiteheart and blackheart as BS309 and BS310 respectively.

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm² (0·2%)	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
Austria	GTW–35	9	333	34*	21·6	—	—	—	6	220
Ö NORM		12	343	35	22·2	—	—	—	4	—
M3192		15	353	36	22·9	—	—	—	3	—
1970	GTW–40	9	353	36	22·9	196	20	12·7	10	220
(Whiteheart)		12	392	40	25·4	216	22	14·0	5	—
		15	412	42	26·7	226	23	14·6	3	—
	GTW–45	9	392	40	25·4	226	23	14·6	12	200
		12	441	45	28·6	255	26	16·5	7	—
		15	471	48	30·5	275	28	17·8	5	—
	GTW–55	9	510	52	33·0	333	34	21·6	7	240
		12	539	55	34·9	353	36	22·9	5	—
		15	559	57	36·2	363	37	23·5	4	—
	GTW–65	9	608	62	39·4	402	41	26·0	4	270
		12	637	65	41·3	422	43	27·3	3	—
		15	657	67	42·5	431	44	27·9	2	—
	GTW–S38	9	314	32	20·3	167	17	10·8	15	200
		12	373	38	24·1	196	20	12·7	12	—
		15	392	40	25·4	206	21	13·3	8	—
(Blackheart)	GTS–35	12 15	343	35*	22·2	196	20	12·7	12	150 max.
	GTS–45	12 15	441	45	28·6	294	30	19·0	7	160–200
	GTS–55	12 15	539	55	34·9	353	36	22·9	5	180–220
	GTS–65	12 15	637	65	41·3	422	43	27·3	3	210–250
	GTS–70	12 15	686	70	44·4	539	55	34·9	2	240–270

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
							(0·2%)			
Belgium	MES35–10	9								
NBN		12	350*	35·7	22·7	190	19·4	12·3	10	(150)
A22–204		15								
1972	MES45–7	9								
(Whiteheart)		12	450	45·9	29·1	260	26·5	16·8	7	(150–200)
		15								
	MES55–4	9								
		12	550	56·1	35·6	330	33·7	21·4	4	(190–240)
		15								
	MES65–3	9								
		12	650	66·3	42·1	390	39·8	25·3	3	(210–250)
		15								
	MES70–2	9								
		12	700	71·4	45·3	500	51·0	32·4	2	(240–285)
		15								
NBN	MAS35–12	15	350	35·7	22·7	190	19·4	12·3	12	(150)
A22–203	MAS45–7	15	450	45·9	29·1	260	26·5	16·8	7	(150–200)
1972	MAS55–4	15	550	56·1	35·6	330	33·7	21·4	4	(190–240)
Blackheart)	MAS65–3	15	650	66·3	42·1	390	39·8	25·3	3	(210–250)
	MAS70–2	15	700	71·4	45·3	550	51·0	32·4	4	(240–285)
							(0·5%)			
Czecho-	TLČ 42	12	294	30*	19·0	—	—	—	6	180 max.
slovakia	2506	15								
ČSN	TLČ 42	12	314	32	20·3	167	17	10·8	8	180 max.
42–1243	2508	15								
1964	TLČ 42	12	324	33	21·0	177	18	11·4	10	180 max.
(Whiteheart)	2510	15								
	TLČ 42	12	343	35	22·2	186	19	12·1	12	160 max.
	2512	15								

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
Czecho-slovakia (Blackheart)	TLB 42 2535	9	294	30*	19·0	—	—	—	7	—
		12	343	35	22·2	—	—	—	4	220 max.
		15	353	36	22·9	—	—	—	3	—
	TLB 42 2540	9	353	36	22·9	177	18	11·4	10	—
		12	392	40	25·4	216	22	14·0	6	220 max.
		15	412	42	26·7	226	23	14·6	4	—
(Pearlitic)	TLP 42 2545	12 15	441	45	28·6	255	26	16·5	7	200 max.
	TLP 42 2550	12 15	490	50	31·7	294	30	19·0	5	230 max.
	TLP 42 2555	12 15	539	55	34·9	324	33	21·0	4	240 max.
France NF A32–701 1967 (Whiteheart)	MB 35–7	9	343	35*	22·2	216	(0·2%) 22	14·0	7	160–220
	MB 40–10	9	392	40	25·4	245	25	15·9	10	160–220
NF A32–702 1968 (Blackheart)	MN 32–8	15	314	32*	20·3	206	21	13·3	8	156 max.
	MN 35–10	15	343	35	22·2	225	23	14·6	10	150 max.
	MN 38–18	15	372	38	24·1	245	25	15·9	18	156 max.
NF A32–703 1968 (Pearlitic)	MP 50–5	15	490	50*	31·8	323	33	21·0	5	170–230
	MP 60–3	15	588	60	38·1	392	40	25·4	3	200–245
	MP 70–2	15	686	70	44·4	490	50	31·8	2	240–285

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
Hungary	Tö 35	9	333	34*	21·6	—	—	—	6	220 max.
MSZ		12	343	35	22·2	—	—	—	4	220 max.
8282–66		15	353	36	22·9	—	—	—	3	220
1968	Tö 40	9	353	36	22·9	196	20	12·7	10	220
(Whiteheart)		12	392	40	25·4	235	24	15·2	5	220
		15	412	42	26·7	255	26	16·5	3	220
(Blackheart)	Tö fk 30–06	15	294	30	19·0	—	—	—	6	190 max.
	Tö fk 32–08	15	314	32	20·3	186	19	12·1	8	190 max.
	Tö fk 32–10	15	314	32	20·3	186	19	12·1	10	190 max.
	Tö fk 35–10	15	343	35	22·2	206	21	13·3	10	180 max.
	Tö fk 35–12	15	343	35	22·2	206	21	13·3	12	170 max.
(Pearlitic)	Tö p 45	15	441	45	28·6	275	28	17·8	7	150–220
	Tö p 50	15	490	50	31·7	314	32	20·3	5	170–240
	Tö p 55	15	539	55	34·9	353	36	22·9	4	190–260
	Tö p 65	15	637	65	41·3	422	43	27·3	3	210–270
	Tö p 70	15	686	70	44·4	539	55	34·9	2	240–280
India	A	9	353	36	22·9	196	20	12·7	10	217 max.
IS 2107–		12	392	40	25·4	235	24	15·2	6	217 max.
1962		15	412	42	26·7	255	26	16·5	4	217 max.
(Whiteheart)	B	9	275	28	17·8	—	—	—	7	217 max.
		12	314	32	20·3	—	—	—	4	217 max.
		15	343	35	22·2	—	—	—	3	217 max.
IS 2108–	A	15	343	35	22·2	206	21	13·3	14	149 max.
1962	B	15	314	32	20·3	186	19	21·1	10	149 max.
(Blackheart)	C	15	294	30	19·0	—	—	—	6	163 max.

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
India	A	15	686	70	44·4	539	55	34·9	2	241–285
IS 2640–	B	15	637	65	41·3	422	43	27·3	3	212–248
1964	C	15	539	55	34·9	353	36	22·9	4	192–241
(Pearlitic)	D	15	490	50	31·7	314	32	20·3	5	170–229
	E	15	441	45	28·6	275	28	17·8	7	149–201
							(0·2%)			
Italy	GMB35	9	333	34*	21·6	—	—	—	7	125–200
UNI		12	343	35	22·2	186	19	12·1	6	
3779–69	GMB40	9	373	38	24·1	—	—	—	12	125–200
1969		12	392	40	25·4	216	22	14·0	10	
(Whiteheart)	GMB45	12	441	45	28·6	255	26	16·5	7	150–210
	GMB50	12	490	50	31·7	294	30	19·0	5	170–230
	GMB55	12	539	55	34·9	324	33	21·0	4	190–240
	GMB65	12	637	65	41·3	382	39	24·8	3	210–250
	GMB70	12	686	70	44·4	490	50	31·7	2	240–285
(Blackheart)	GMN35	12	343	35	22·2	186	19	12·1	12	110–150
	GMN37	12	363	37	23·5	216	22	14·0	14	120–160
	GMN45	12	441	45	28·6	255	26	16·5	7	150–210
	GMN50	12	490	50	31·7	294	30	19·0	5	170–230
	GMN55	12	539	55	34·9	324	33	21·0	4	190–240
	GMN65	12	637	65	41·3	382	39	24·8	3	210–250
	GMN70	12	686	70	44·4	490	50	31·7	2	240–285

Malleable cast iron										
Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength*			*0·5% Proof stress, min.*			*Elongation* % (min.)	*Hardness*
			N/mm²	kgf/mm²	tonf/in²	N/mm²	kgf/mm² (0·5%)*	tonf/in²		
Japan	FCMW34	6	314	32*	20·3	—	—	—	8	—
JIS G	(Class 1)	10	333	34	21·6	167	17	10·8	5	—
5703		14	353	36	22·9	196	20	12·7	3	—
1969	FCMW38	6	353	36	22·9	—	—	—	14	—
(Whiteheart)	(Class 2)	10	373	38	24·1	186	19	12·1	8	—
		14	392	40	25·4	216	22	14·0	6	—
Japan	FCMB28	12 14	275	28	17·8	167	17	10·8	5	—
JIS G	(Class 1)									
5702	FCMB32	12 14	314	32	20·3	186	19	12·1	8	—
1969	(Class 2)									
(Blackheart)	FCMB35	12 14	343	35	22·2	196	20	12·7	10	—
	(Class 3)									
	FCMB37	12 14	363	37	23·5	206	21	13·3	14	—
	(Class 4)									
JIS G	FCMP45	14	441	45	28·6	265	27	17·1	6	(149–207)
5704	(Class 1)									
1969	FCMP50	14	490	50	31·7	304	31	19·7	4	(167–229)
(Pearlitic)	(Class 2)									
	FCMP55	14	539	55	34·9	343	35	22·2	3	(183–241)
	(Class 3)									
	FCMP60	14	588	60	38·1	392	40	25·4	3	(204–269)
	(Class 4)									
	FCMP70	14	686	70	44·4	510	52	33·0	2	(229–285)
	(Class 5)									

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
Netherlands		9	353	36*	22·9	196	20	12·7	10	220 max.
NEN	GSmF	12	392	40	25·4	235	24	15·2	6	220 max.
6002–C		15	412	42	26·7	255	28	16·5	4	220 max.
1966 (Whiteheart)										
(Blackheart)	GSmT32	15	314	32	20·3	186	19	12·1	10	160 max.
	GSmT35	15	343	35	22·2	206	21	13·3	12	150 max.
(Pearlitic)	GSmP45	15	441	45	28·6	275	28	17·8	7	150–200
	GSmP50	15	490	50	31·7	314	32	20·3	5	170–230
	GSmP55	15	539	55	34·9	353	36	22·9	4	180–240
	GSmP65	15	637	65	41·3	422	42	27·3	3	210–250
	GSmP70	15	686	70	44·4	539	55	34·9	2	240–285

Values are also specified for 0·2% proof stress as follows:

	N/mm²	kgf/mm²	tonf/in²
GSmT32	167	17	10·8
GSmT35	186	19	12·1
GSmP45	255	26	16·5
GSmP50	294	30	19·0
GSmP55	324	33	21·0
GSmP65	382	39	24·8
GSmP70	490	50	31·7

Malleable cast iron										
Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength*			*0·5% Proof stress, min.*			*Elongation* % (min.)	*Hardness*
			N/mm²	kgf/mm²	tonf/in²	N/mm²	kgf/mm²	tonf/in²		
Norway							(0·2%)			
NS 11501	NS11533	—	324	33*	21·0	186	19	12·1	14	150 max.
1971	NS11550	—	490	50	31·7	294	30	19·0	5	160–210
(Blackheart)	NS11560	—	588	60	38·1	353	36	22·9	4	190–250
Poland							(0·5%)			
PN–68	ZcB3504	12	343	35*	22·2	—	—	—	4	220 max.
H–83221	ZcB4005	12	392	40	25·4	236	24	15·2	5	220 max.
(Whiteheart)	ZcB4505	12	441	45	28·6	255	26	16·5	5	220 max.
(Blackheart)	ZcC3006	15	294	30	19·0	—	—	—	6	190 max.
	ZcC3208	15	314	32	20·3	186	19	12·1	8	190 max.
	ZcC3510	15	343	35	22·2	206	21	13·3	10	180 max.
	ZcC3512	15	343	35	22·2	206	21	13·3	12	170 max.
(Pearlitic)	ZcP4507	15	441	45	28·6	275	28	17·8	7	220 max.
	ZcP5005	15	490	50	31·7	314	32	20·3	5	240 max.
	ZcP5504	15	539	55	34·9	353	36	22·9	4	260 max.
	ZcP6503	15	637	65	41·3	422	43	27·3	3	270 max.
	ZcP7002	15	686	70	44·4	539	55	34·9	2	280 max.
Romania		9	274	28*	17·8	—	—	—	7	200 max.
STAS	Fma35.03	12	314	32	20·3	—	—	—	4	220 max.
569–70		15	343	35	22·2	—	—	—	3	240 max.
1071		9	355	36	22·9	216	22	14·0	10	200 max.
(Whiteheart)	Fma42.04	12	392	40	25·4	235	24	15·2	6	220 max.
		15	412	42	26·7	255	26	16·5	4	240 max.

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
Romania										
(Blackheart)	Fmn30.06	15	294	30*	19·0	—	—	—	6	160 max.
	Fmn32.08	15	314	32	20·3	—	—	—	8	160 max.
	Fmn32.10	15	314	32	20·3	186	19	12·1	10	160 max.
	Fmn35.10	15	343	35	22·2	186	19	12·1	10	150 max.
	Fmn35.12	15	343	35	22·2	206	21	13·3	12	150 max.
	Fmn37.12	15	363	37	23·5	206	21	13·3	12	150 max.
	Fmn37.14	15	363	37	23·5	206	21	13·3	14	150 max.
(Pearlitic)	Fmp45.07	15	441	45	28·6	275	28	17·8	7	220 max.
	Fmp50.05	15	490	50	31·7	314	32	20·3	5	240 max.
	Fmp55.04	15	539	55	34·9	353	36	22·9	4	260 max.
	Fmp60.03	15	588	60	38·1	382	39	24·8	3	270 max.
	Fmp65.03	15	637	65	41·3	412	43	27·3	3	270 max.
	Fmp70.02	15	686	70	44·4	539	55	34·9	2	280 max.

Values are also specified for 0·2% proof stress, as follows:

Grade	*Diameter of test bar* mm	N/mm²	kgf/mm²	tonf/in²
Fma42.04	9	196	20	12·7
	12	216	22	14·0
	15	235	24	15·2
Fmn32.10	15	168	17	10·8
Fmn35.10	15	168	17	10·8
Fmn35.12	15	186	19	12·1
Fmn37.12	15	186	19	12·1
Fmn37.14	15	186	19	12·1
Fmp45.07	15	255	26	16·5
Fmp50.05	15	294	30	19·0
Fmp55.04	15	323	33	21·0
Fmp60.03	15	353	36	22·9
Fmp65.03	15	382	39	24·8
Fmp70.02	15	490	50	31·7

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
Spain	A	9	353	36	22·9	196	20	12·7	10	
UNE		12	392	40	25·4	235	24	15·2	6	220 max.
36 113		15	412	42	26·7	255	26	16·5	4	
(1963)	B	9	274	28	17·8	—	—	—	7	
(Whiteheart)		12	314	32	20·3	—	—	—	4	220 max.
		15	343	35	22·2	—	—	—	3	
UNE	A	15	343	35	22·2	206	21	13·3	12	150 max.
36 114	B	15	314	32	20·3	186	19	12·1	10	160 max.
(1963)	C	15	294	30	19·0	—	—	—	6	160 max.
(Blackheart)										
(Spain)	A	15	686	70	44·4	539	55	34·9	2	240–285
UNE	B	15	637	65	41·3	422	43	27·3	3	210–250
36 116	C	15	539	55	34·9	353	36	22·9	4	190–240
(Pearlitic)	D	15	490	50	31·7	314	32	20·3	5	170–230
	E	15	441	45	28·6	274	28	17·8	7	150–200

Values are also specified for 0·2% proof stress, as follows:

Grade	*Diameter of test bar* mm	N/mm²	kgf/mm²	tonf/in²
UNE				
36 113 A	9	177	18	11·4
	12	216	22	14·0
	15	225	23	14·6
UNE				
36 114 A	15	186	19	12·1
B	15	167	17	10·8
UNE				
36 116 A	15	490	50	31·7
B	15	382	39	24·8
C	15	323	33	21·0
D	15	294	30	19·0
E	15	255	26	16·5

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm² (0·2%)	tonf/in²	*Elongation* % (min.)	*Hardness*
Sweden SIS	08 10–02	—	324	33	21·0	186	19	12·1	14	150 max.
1408 1971	08 54–03	—	490	50	31·7	294	30	19·0	5	160–210
(Blackheart)	08 56–03	—	588	60	38·1	353	36	22·9	4	190–250
Switzerland	GTW–40	9	353	36*	22·9	196	20	12·7	10	
VSM		12	392	40	25·4	216	22	14·0	5	220 max.
10692		15	412	42	26·7	226	23	14·6	3	
1971	GTW–45	9	392	40	25·4	226	23	14·6	12	
(Whiteheart)		12	441	45	28·6	255	26	16·5	7	200 max.
		15	471	48	30·5	275	28	17·8	5	
	GTW–S38	9	314	32	20·3	167	17	10·8	15	
		12	373	38	24·1	196	20	12·7	12	200 max.
		15	392	40	25·4	206	21	13·3	8	
(Blackheart)	GTS–35	12 15	343	35	22·2	196	20	12·7	12	150 max.
	GTS–45	12 15	441	45	28·6	294	30	19·0	7	160–200
	GTS–55	12 15	539	55	34·9	353	36	22·9	5	180–220
	GTS–65	12 15	637	65	41·3	422	43	27·3	3	210–250
	GTS–70	12 15	686	70	44·4	539	55	34·9	2	240–270

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm² (0·2%)	tonf/in²	*Elongation* % (min.)	*Hardness*
Turkey	TDB–35	9	333	34*	21·6	—	—	—	6	
TS 519		12	343	35	22·2	—	—	—	4	220 max.
1967		15	353	36	22·9	—	—	—	3	
(Whiteheart)	TDB–40	9	353	36	22·9	196	20	12·7	10	
		12	392	40	25·4	216	22	14·0	5	220 max.
		15	412	42	26·7	226	23	14·6	3	
	TDB–45	9	392	40	25·4	226	23	14·6	12	
		12	441	45	28·6	255	26	16·5	7	200 max.
		15	471	48	30·5	275	28	17·8	5	
	TDB–55	9	510	52	33·0	333	34	21·6	7	
		12	539	55	34·9	353	36	22·9	5	240 max.
		15	559	57	36·2	363	37	23·5	4	
	TDB–65	9	608	62	39·4	402	41	26·0	4	
		12	637	65	41·3	422	43	27·3	3	270 max.
		15	657	67	42·5	431	44	27·9	2	
	TDB–K38	9	314	32	20·3	167	17	10·8	15	
		12	373	38	24·1	196	20	12·7	12	200 max.
		15	392	40	25·4	206	21	13·3	8	
(Blackheart)	TDS–35	12 15	343	35	22·2	196	20	12·7	12	150 max.
	TDS–45	12 15	441	45	28·6	294	30	19·0	7	100–200
	TDS–65	12 15	539	55	34·9	353	36	22·9	5	180–220
	TDS–70	12 15	686	70	44·4	539	55	34·9	2	240–270

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	ksi	*Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	ksi	*Elongation, % (min.)*	*Hardness*
								(0·5%)				
USA ASTM A47–68 Ferritic	32510	15·9	345	35·2	22·3	50	221	22·5	14·3	32*	10	
	35018	15·9	365	37·3	23·7	53	241	24·6	15·6	35	18	
ASTM A.197–47 (Reapproved 1971) Cupola malleable iron		15·9	276	28·1	17·9	40	207	21·1	13·4	30	5	
								(0·2%)				
ASTM A.220–71 (Pearlitic)	40010	15·9	414	42·2	26·8	60	276	28·1	17·9	40*	10	(149–197)
	45008	15·9	448	45·7	29·0	65	310	31·6	20·1	45	8	(156–197)
	45006	15·9	448	45·7	29·0	65	310	31·6	20·1	45	6	(156–207)
	50005	15·9	483	49·2	31·2	70	345	35·2	22·3	50	5	(179–229)
	60004	15·9	552	56·2	35·7	80	414	42·2	26·8	60	4	(197–241)
	70003	15·9	586	59·8	37·9	85	483	49·2	31·2	70	3	(217–269)
	80002	15·9	655	66·8	42·4	95	552	56·2	35·7	80	2	(241–285)
	90001	15·9	724	73·8	46·9	105	621	63·3	40·2	90	1	(269–321)
								(0·2%)				
ASTM A.602–70 SAE J 158 Automotive malleable iron castings†	M3210	(	345	35·2	22·3	50	221	22·5	14·3	32	10)	156 max.
	M4504	(	448	45·7	29·0	65	310	31·6	20·1	45	4)	163–217
	M5003	(	517	52·7	33·5	75	345	35·2	22·3	50	3)	187–241
	M5503	(	517	52·7	33·5	75	380	38·7	24·6	55	3)	187–241
	M7002	(	621	63·3	40·2	90	483	49·2	31·2	70	2)	229–269
	M8501	(	724	73·8	46·9	105	586	59·8	37·9	85	1)	269–302

†Specified by hardness and heat treatment. Mechanical properties are for material description only and not part of the standard.

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
USSR GOST 1215–59 (reissued 1964)	KCh30–6	8								
		12	294	30*	19·0	—	—	—	6	163 max.
		16								
	KCh33–8	8								
		12	324	33	21·0	—	—	—	8	163 max.
		16								
	KCh35–10	8								
		12	343	35	22·2	—	—	—	10	163 max.
		16								
	KCh37–12	8								
		12	363	37	23·5	—	—	—	12	163 max.
		16								
	KCh45–6	8								
		12	441	45	28·6	—	—	—	6	241 max.
		16								
	KCh50–4	8								
		12	490	50	31·7	—	—	—	4	241 max.
		16								
	KCh56–4	8								
		12	549	56	35·6	—	—	—	4	269 max.
		16								
	KCh60–3	8								
		12	588	60	38·1	—	—	—	3	269 max.
		16								
	KCh63–2	8								
		12	618	63	40·0	—	—	—	2	269 max.
		16								

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength*			*0·5% Proof stress, min.*			*Elongation* % (min.)	*Hardness*
			N/mm²	kgf/mm²	tonf/in²	N/mm²	kgf/mm² (0·2%)	tonf/in²		
West Germany (Federal German Republic) DIN 1692 1963 (Whiteheart)	GTW–35	9	333	34*	21·6	—	—	—	6	
		12	343	35	22·2	—	—	—	4	220 max.
		15	353	36	22·9	—	—	—	3	
	GTW–40	9	353	36	22·9	196	20	12·7	10	
		12	392	40	25·4	216	22	14·0	5	220 max.
		15	412	42	26·7	226	23	14·6	3	
	GTW–45	9	392	40	25·4	226	23	14·6	12	
		12	441	45	28·6	255	26	16·5	7	200 max.
		15	471	48	30·5	275	28	17·8	5	
	GTW–55	9	510	52	33·0	333	34	21·6	7	
		12	539	55	34·9	353	36	22·9	5	240 max.
		15	559	57	36·2	363	37	23·5	4	
	GTW–65	9	608	62	39·4	402	41	26·0	4	
		12	637	65	41·3	422	43	27·3	3	270 max.
		15	657	67	42·5	431	44	27·9	2	
	GTW–S38	9	314	32	20·3	167	17	10·8	15	
		12	373	38	24·1	196	20	12·7	12	200 max.
		15	392	40	25·4	206	21	13·3	8	
(Blackheart)	GTS–35	12 15	343	35	22·2	196	20	12·7	12	150 max.
	GTS–45	12 15	441	45	28·6	294	30	19·0	7	160–200
	GTS–55	12 15	539	55	34·9	353	36	22·9	5	180–220
	GTS–65	12 15	537	65	41·3	422	43	27·3	3	210–250
	GTS–70	12 15	686	70	44·4	539	55	34·9	2	240–270

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength*			*0·5% Proof stress, min.*			*Elongation* % (min.)	*Hardness*
			N/mm²	kgf/mm²	tonf/in²	N/mm²	kgf/mm²	tonf/in²		
Yugoslavia JUS C.J2.021 (Whiteheart)	BTeL35	9	333	34*	21·6	—	—	—	5	200 max.
		12							3	
		15							3	
		18							2	
	BTeL40	9	373	38	24·1	—	—	—	10	200 max.
		12							5	
		15							3	
		18							3	
	BTeL00	—	—	—	—	—	—	—	—	250 max.
(Blackheart)	CTeL35	9	333	33	21·0	—	—	—	8	150 max.
		12								
		15								
		18								
	CTeL38	9	363	37	23·5	—	—	—	12	140 max.
		12								
		15								
		18								
(Pearlitic)	CTeL45	9	392	40	25·4	—	—	—	5	250
		12								
		15								
		18								

Malleable cast iron

Country and specification	*Grade*	*Diameter of test bar* mm	*Minimum tensile strength* N/mm²	kgf/mm²	tonf/in²	*0·5% Proof stress, min.* N/mm²	kgf/mm²	tonf/in²	*Elongation* % (min.)	*Hardness*
International Organisation for Standardisation (ISO) ISO/R 942 1969 (Whiteheart)	A	9	353	36*	22·9	196	20	12·7	10	
		12	392	40	25·4	235	24	15·2	6	(220 max.)
		15	412	42	26·7	255	26	16·5	4	
	B	9	275	28	17·8	—	—	—	7	
		12	314	32	20·3	—	—	—	4	(220 max.)
		15	343	35	22·2	—	—	—	3	
(ISO) ISO/R 943 1969 (Blackheart)	A	15	343	35	22·2	206	21	13·3	12	(150 max.)
	B	15	314	32	20·3	186	19	12·1	10	(160 max.)
	C	15	294	30	19·0	—	—	—	6	(160 max.)
ISO/R 944 1969	A	15	686	70	44·4	539	55	34·9	2	(240–285)
	B	15	637	65	41·3	422	43	27·3	3	(210–250)
	C	15	539	55	34·9	353	36	22·9	4	(190–240)
	D	15	490	50	31·7	314	32	20·3	5	(170–230)
	E	15	441	45	28·6	275	28	17·8	7	(150–200)

Values are also specified for 0·2% proof stress, as follows:

Grade	*Diameter of test bar* mm	N/mm²	kgf/mm²	tonf/in²
942A	9	177	18	11·4
	12	216	22	14·0
	15	226	23	14·6
943A	15	186	19	12·1
B	15	167	17	10·8
944A	15	490	50	31·7
B	15	382	39	24·8
C	15	324	33	21·0
D	15	294	30	19·0
E	15	255	26	16·5

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