

MODULE 21

WELDING

METALLURGY

OF

STEELS



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MODULE 21

**WELDING METALLURGY
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MODULE 21

WELDING METALLURGY OF STEELS

Introduction

The increasing sophistication of steel property requirements—especially fracture toughness—has resulted in a new generation of steels made by complex processing routes. These present important challenges to the welding engineer who must select welding consumables and develop procedures giving specific properties and which avoid cracking problems. This task is made all the more difficult by the growing sophistication of steel metallurgical design and demands a thorough knowledge of underlying metallurgical principles. This is the subject of this module.

Here you will learn about the structure of steels and how they obtain their strength and toughness. You will use various diagrams to determine the effects of heat treatment on the structure of steel. We will discuss what factors determine the properties of welds—both weld metal and the heat affected zone—and finally we show you how to determine a welding procedure that will avoid cracking problems. We do not discuss steel making or ways of classifying and specifying steel. These topics are covered in Module 8 and we recommend you study that Module in addition to this one.

Objectives

After successfully completing this module you will be able to:

- Describe the ways steels derive their strength and toughness
- Describe the effects of heat on steel properties
- Explain how the welding procedure affects the weld metal and heat affected zone properties
- Determine procedures necessary to avoid various cracking problems

STEEL METALLURGY

Steels form the largest group of commercially important alloys. There are several reasons for this:

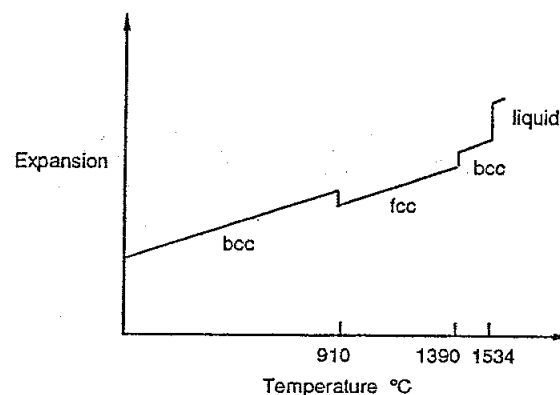
- The great abundance of iron in the earth's crust
- The relative ease of extraction and low cost
- The wide range of properties that can be achieved as a result of the solid state phase transformation
- Other unique properties such as magnetism

Steels are alloys of iron and less than 2% carbon plus a wide range of other elements. Some of these are added deliberately to impart special properties and others are impurities not completely removed during the steel making process. Elements may be present in solid solution or combined as intermetallic compounds with iron, carbon or other elements. Some elements, namely carbon, nitrogen, boron and hydrogen, form interstitial solutions with iron whereas others such as manganese and silicon form substitutional solutions. Beyond the limit of solubility these elements may also form intermetallic compounds with iron or other elements.

Crystal structure

Iron has the special property of existing in different crystallographic forms in the solid state. Below 910°C the structure is body-centred cubic (bcc) but between 910°C and 1390°C it assumes a face-centred cubic structure (fcc). Above 1390°C up to the melting point at 1534°C the structure reverts back to the body-centred cubic form. These are known as allotropic forms of iron. The face-centred cubic form is a close-packed structure being more dense than the body-centred cubic form. Consequently iron will actually *contract* as it is heated above 910°C when the structure transformation takes place. Fig. 1 illustrates this.

Figure 1. Heating a bar of iron causes it to expand but when transformation occurs from the bcc structure to the fcc structure at 910°C a small contraction takes place. The fcc crystal structure is close-packed and occupies less volume than the bcc structure.



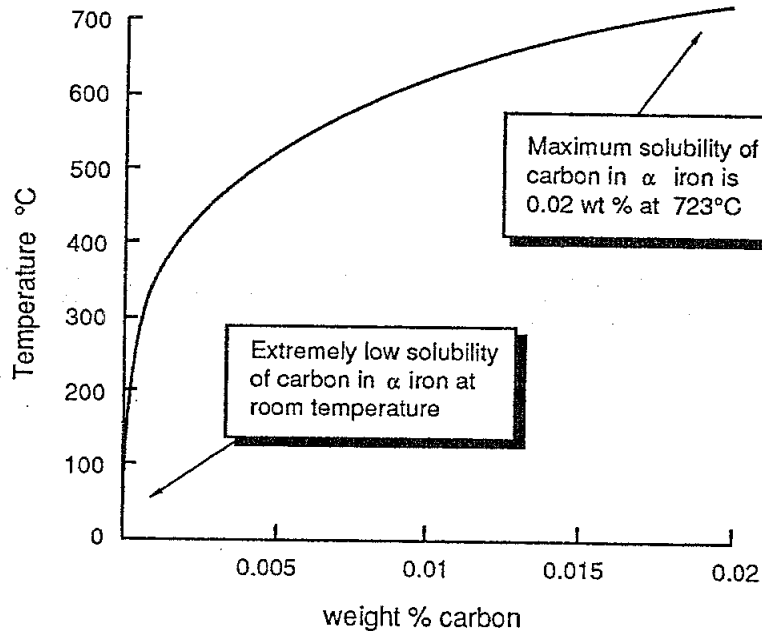


Figure 2. Solubility of carbon in α (bcc) iron as a function of temperature.

Solubility of carbon

The solubility of carbon in the bcc form of iron is very small, the maximum solubility being only about 0.02 weight per cent at 723°C. As Fig. 2 shows there is negligible solubility of carbon in iron at room temperature (less than 0.0001 wt %). Since steels nearly always have more carbon than this, the excess carbon is not in solution but rather present as the intermetallic compound iron-carbide Fe_3C known as cementite. *FORMS PERLITE*

In contrast the face-centred cubic form of iron dissolves up to 2.0 % carbon, well in excess of the usual carbon content of steels. A steel can therefore be heated to a temperature at which the structure changes to fcc and all carbon goes into solution. The way in which the carbon is obliged to redistribute itself upon cooling back below the transformation is the origin of the wide range of properties achievable in steels.

Iron-iron carbide phase diagram

Fundamental to a study of steels is an understanding of the iron-carbon phase diagram. The diagram we will discuss is actually the metastable iron-iron carbide system. The true stable form of carbon in iron is graphite, but except for cast irons this only occurs after prolonged heating. Since in steels carbon is normally present as iron carbide, it is this system that we consider. Fig. 3 shows the iron-iron carbide system up to 6 wt % carbon. We will now discuss several important features of this diagram.

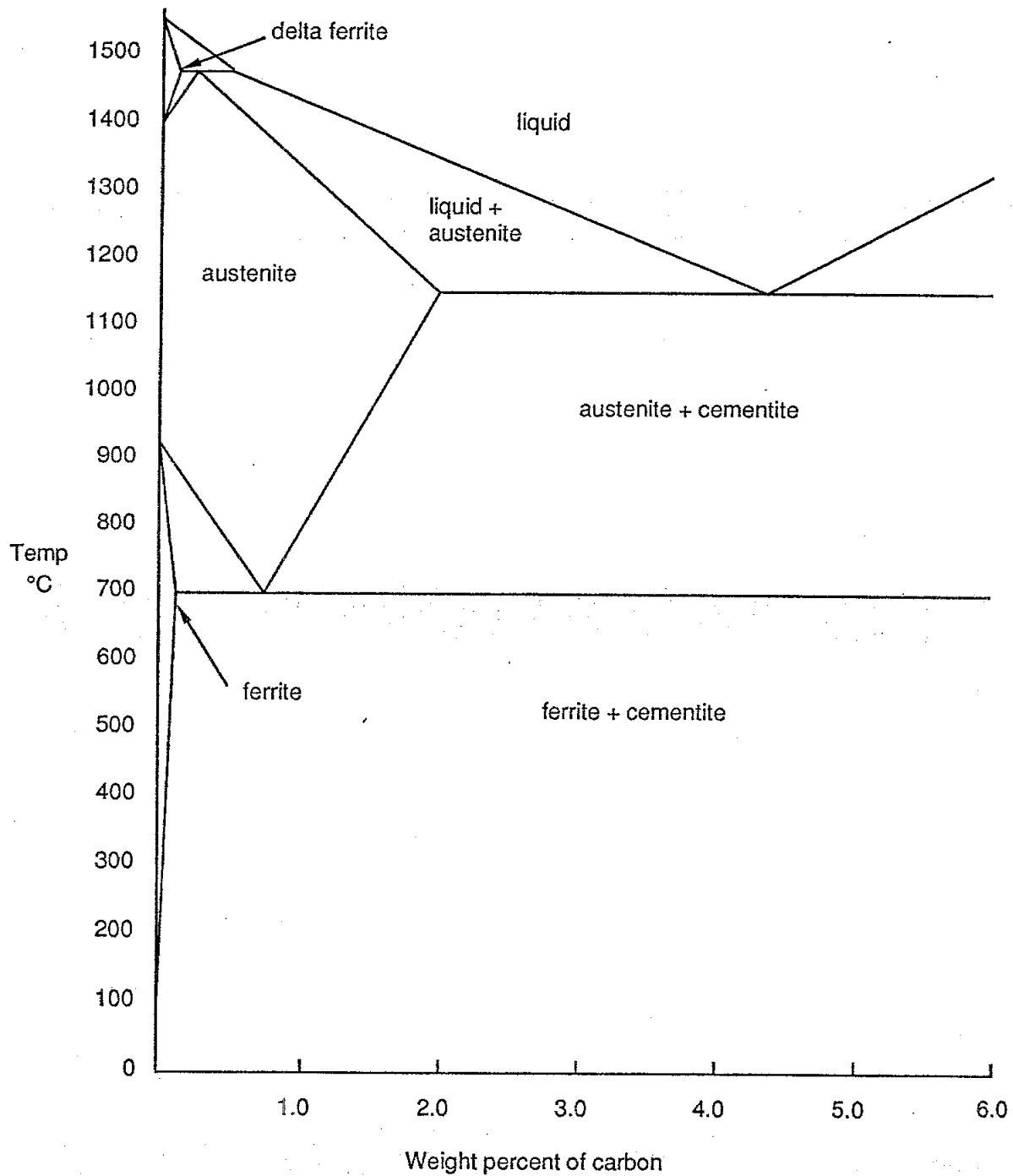
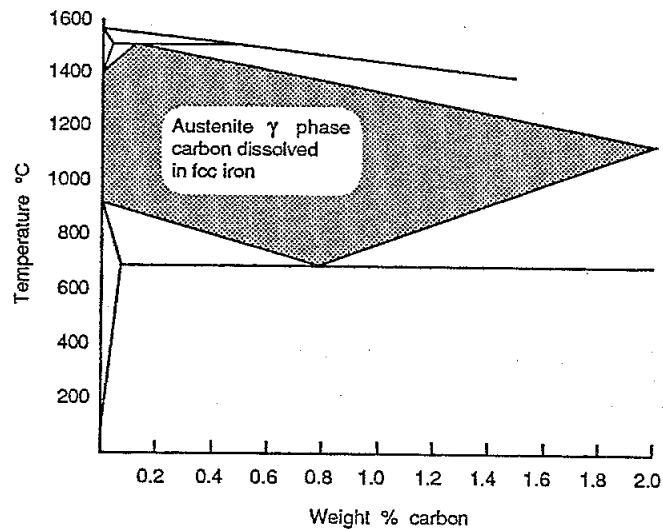


Figure 3. The iron-iron carbide phase diagram.

Figure 4. The austenite region of the iron-iron carbide diagram. Austenite dissolves up to about 2% carbon.



A eutectic is formed at 4.3% carbon. Liquid of this composition on solidification at 1147°C transforms to a mixture of two phases (austenite + cementite). This region is important when discussing cast irons but is not relevant to steels.

The region in which iron is face-centred cubic, identified in Fig. 4, dissolves up to 2% carbon. This phase is termed austenite or gamma phase. With no carbon present it begins at 910°C on heating but with 0.8% carbon it starts at 723°C. When a steel is heated into the austenite region all carbon and most other compounds dissolve to form a single phase.

The region shown in Fig. 5 where carbon is dissolved in bcc iron is very narrow, extending to only 0.02% carbon at 723°C. This phase is termed ferrite or alpha phase. Although the carbon content of ferrite is very low other elements may dissolve appreciably in it so ferrite cannot be considered as "pure iron".

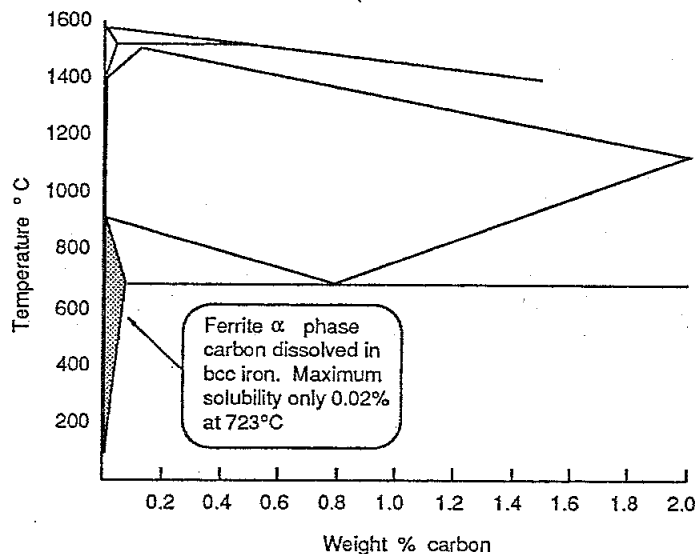
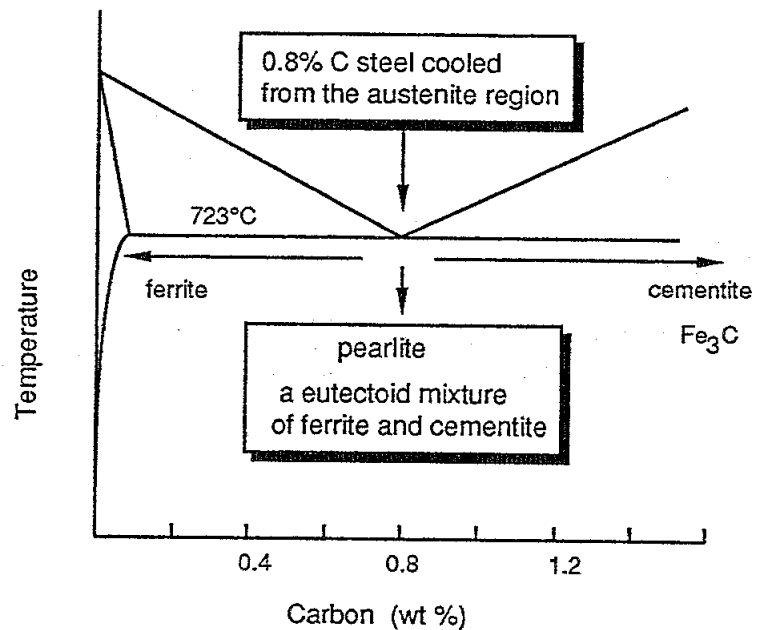


Figure 5. The ferrite region of the iron-iron carbide diagram.

Figure 6. The eutectoid part of the iron-iron carbide phase diagram.



Pearlite

At 0.8% carbon and 723°C a eutectoid is formed. This is similar to a eutectic but occurs completely in the solid. As Fig.6 illustrates a steel of eutectoid composition will, when cooled from the austenite region, transform at 723°C to a mixture of two phases, ferrite and cementite (Fe_3C). This eutectoid mixture is called pearlite. Fig. 7 shows an example. Note that pearlite is not a single phase itself but rather a constituent made of two phases. Pearlite is but one of many constituents given different names but made of the same two phases ferrite and iron carbide. Cementite itself is extremely hard—about 1150 Hv—but when mixed with the soft ferrite layers in pearlite the average hardness of pearlite is considerably less.

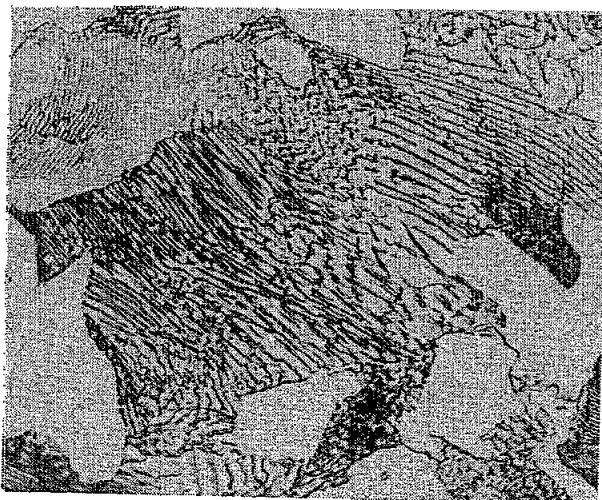


Figure 7. A typical example of pearlite showing the two phases ferrite and cementite (iron carbide) in layers. Light areas are ferrite and dark areas are cementite.

Peritectic

The region at the top left portion of the diagram enlarged in Fig. 8 is where iron reverts back to the bcc structure. Here again the solubility for carbon is low, only 0.1 wt % at 1493°C. This phase is known as delta ferrite. The part of the diagram at 0.16% carbon having the appearance of an inverted eutectoid is called a peritectic. At this point a two phase mixture of delta ferrite and liquid transforms on cooling to a single phase solid of austenite. This portion of the diagram will not be discussed in detail, but the general shape of the phase diagram in this region should be recognized since it has been invoked to explain various hot cracking phenomena in welding.

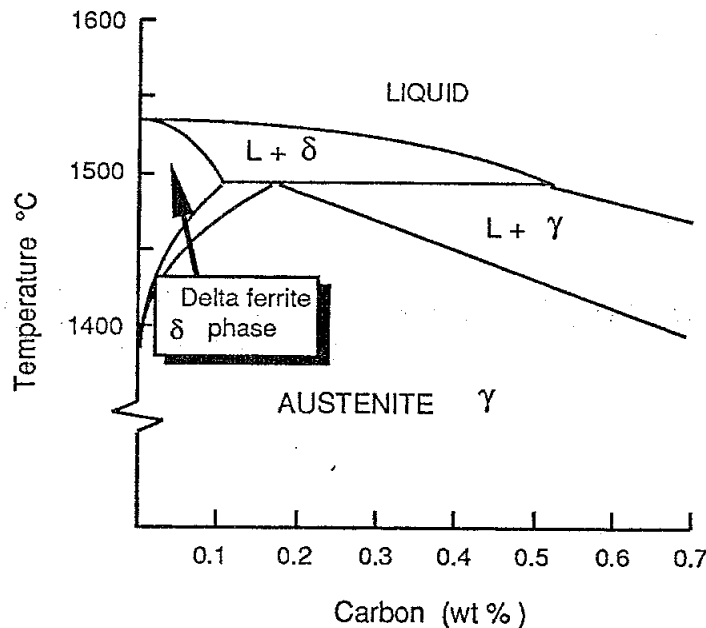


Figure 8. Peritectic region of the iron-iron carbide diagram.

Pearlite growth

The part of the diagram that is of most interest to a study of steels is the region containing the eutectoid in the bottom left corner of the iron-iron carbide diagram and we will discuss this in a little more detail.

A steel with 0.8 wt% carbon, it will be recalled, transforms on cooling through 723°C to the two phase eutectoid constituent pearlite. In pearlite the two phases ferrite and cementite are mixed closely together in fine layers. As the ferrite contains very little carbon while the cementite has 6.7%, carbon atoms must diffuse to the growing cementite plates as Fig. 9 shows. The distance they can diffuse—and hence the spacing of the plates—depends on how fast the pearlite is growing. A fast growth rate means less time for diffusion and a finer pearlite.

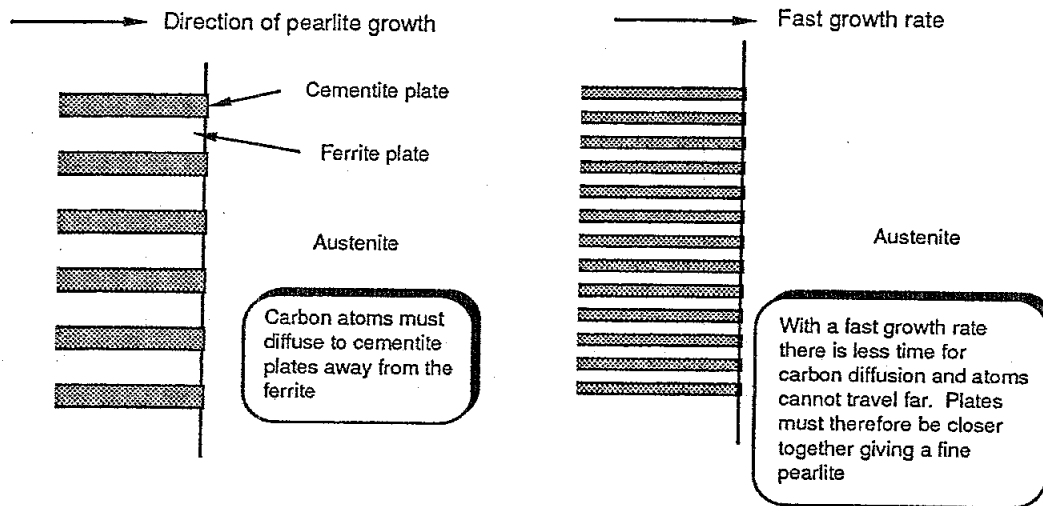


Figure 9. The growth of pearlite is diffusion controlled. The morphology therefore depends on the growth rate which in turn depends on the cooling rate of the steel.

Proeutectoid ferrite

If the steel has less than 0.8% carbon (termed hypo-eutectoid) ferrite will be formed first from the austenite. The example in Fig. 10 shows a steel of 0.4% carbon. This ferrite is called proeutectoid ferrite. As transformation continues and the temperature drops the remaining austenite becomes richer in carbon. At 723°C the steel comprises ferrite and remaining austenite with 0.8% carbon. The latter then transforms to pearlite so the final structure of the steel is a mixture of pro-eutectoid ferrite and pearlite. The amounts of pro-

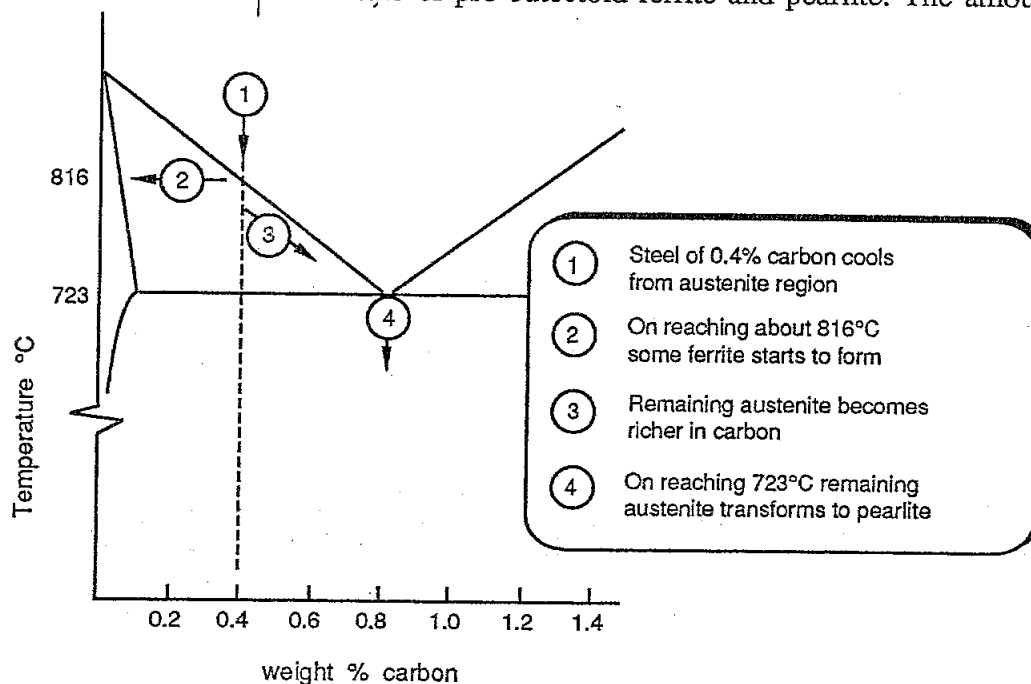


Figure 10. Diagram showing the sequence of transformation of a 0.4% carbon steel.

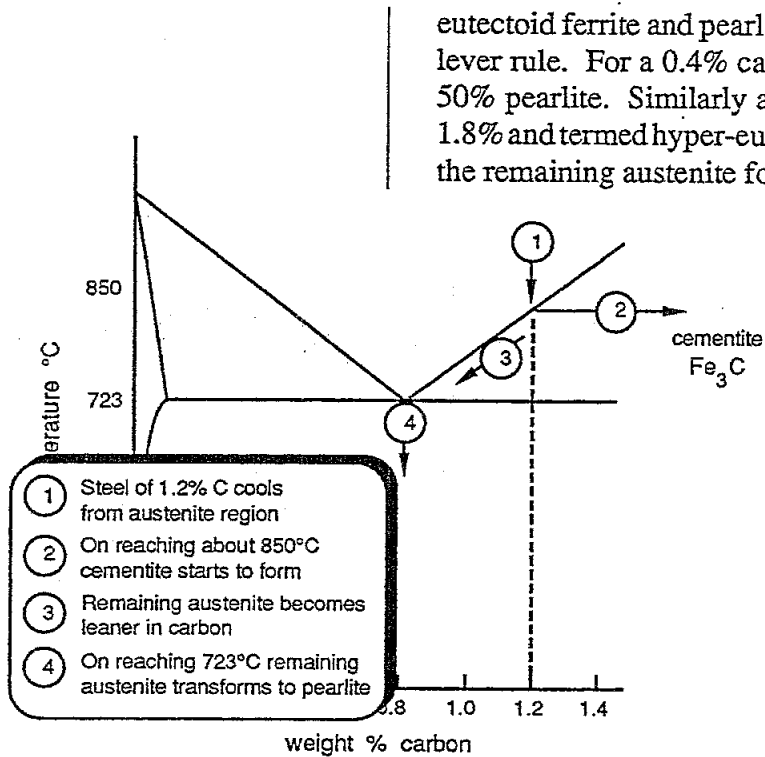
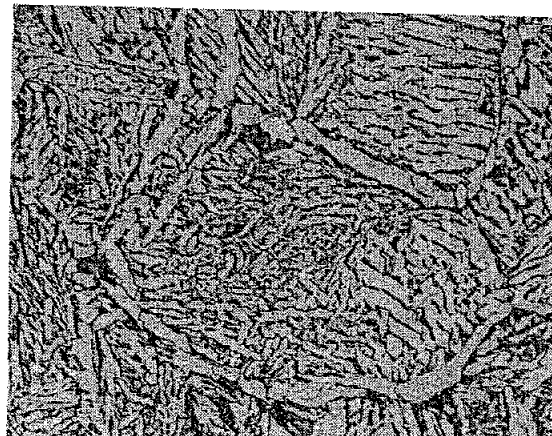


Figure 11. Sequence of phase transformations in a steel with greater than 0.8% carbon.

Morphology of proeutectoid ferrite

The appearance of polished sections viewed under the microscope is shown in Fig. 12. You can see that the proeutectoid ferrite (or cementite) forms initially in the grain boundaries of the austenite, nucleation taking place at several points around each austenite grain. Since each region of ferrite becomes an individual grain, its grain size will be very much smaller than that of the parent austenite. Ferrite continues to be formed and grow until the final transformation of remaining austenite to pearlite. The ferrite does not always appear as neat, equiaxed grains as shown in the figure but can occur as long spikes from the grain boundaries or even nucleate within the austenite grain especially in welds.

Figure 12. Photomicrograph of a steel with less than 0.8% carbon showing proeutectoid ferrite at the grain boundaries.



Grain growth

On reheating the steel the process reverses and the pearlite and ferrite grains transform back into single phase austenite, completely new grains being formed. The temperature required to get complete transformation depends on the carbon level as seen from the phase diagram and ranges from 910°C for zero carbon to 723°C for 0.8% carbon.

Fine grain steels

Heating to higher temperatures than those necessary to get complete transformation causes the austenite grains to grow. The final size of the austenite grains depends not only on the temperature reached but also on the type of steel. Some steels containing small precipitates, such as aluminum and vanadium nitride, retain small grain size up to high temperatures. These are termed fine grained steels. Steels may be deliberately made as coarse grain or fine grain. Fine grain is specified when toughness is a consideration but coarse grain may be desirable when high temperature creep resistance is important. For most structural applications fine grain steels are specified.

The effect of austenitizing temperature on grain size is shown in Fig. 13. It shows that although grain growth is restricted in a fine grain steel at a sufficiently high temperature the precipitates dissolve and

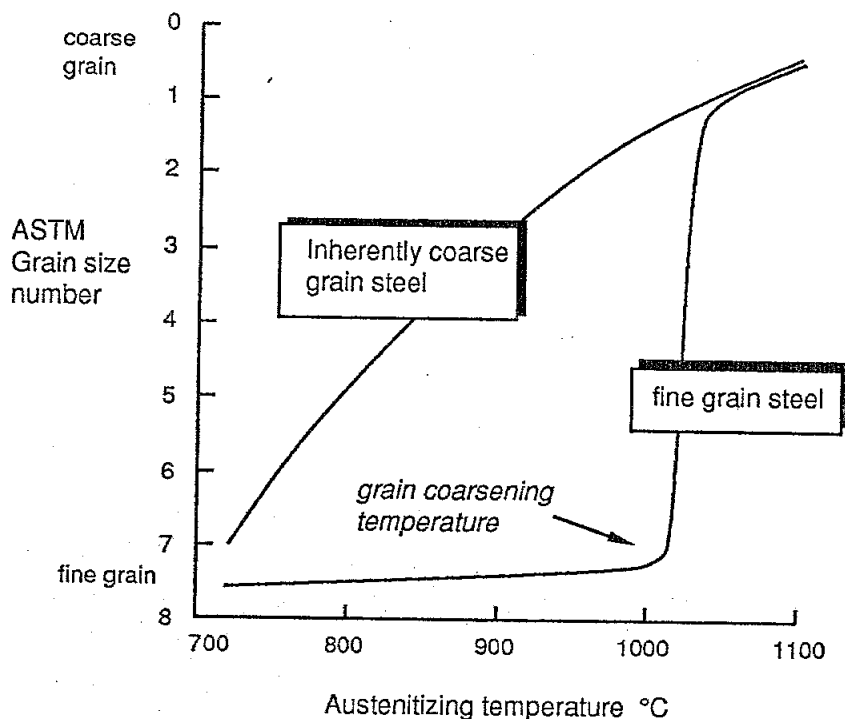
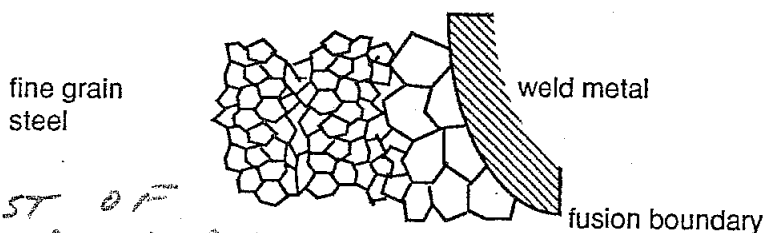
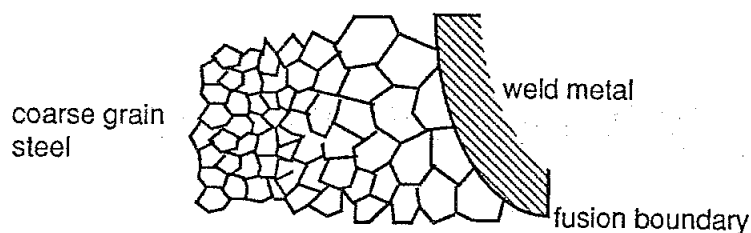


Figure 13. The effect of temperature on austenitic grain size in a coarse and a fine grain steel.

HAZ grain size

the steel behaves as a coarse grain steel. This effect is important in the heat affected zone (HAZ) of a weld. Fig. 14 illustrates the difference in HAZ grain size between a coarse grained and a fine grained steel. In the HAZ of the coarse grained steel there is a wide region where grain growth has occurred but the fine grained steel resists grain growth except in the narrow region immediately adjacent to the weld fusion boundary where temperatures are very high. Notice that in this narrow region the grain size is the same for the coarse and fine grained steels. Steels containing titanium are exceptions in that fine HAZ grain size may be achieved right up to the fusion boundary. Titanium nitride is very stable and may not completely dissolve in the HAZ even at the temperatures immediately adjacent to the fusion boundary. Modern steels, designed specifically to retain small grain size in the HAZ of high heat input welds, sometimes contain titanium for this purpose.

Figure 14. Austenitic grain size in the heat affected zone of a weld. Note that close to the fusion boundary the grain size is the same in the coarse and fine grain steels.



*MOST OF
OUR WORK.*

Non-equilibrium cooling

We only observe the phases and microstructures predicted by the iron-iron carbide diagram in steels cooled very slowly. Moreover the diagram assumes that carbon is the only alloying element present in the steel. If the concentration of other alloy elements is small the basic diagram can still be used except that it will be distorted and the lines may move to slightly different locations. For example the presence of alloy elements changes the eutectoid temperature as shown in

McQuaid-Ehn Test

- The actual grain size of a steel depends not only on its composition but on any prior heat treatment, such as normalizing, hot working, etc. A steel that is made to fine grain practice is inherently fine grain because of the presence of aluminum or other elements that form nitrides and resist grain growth when the steel is heated.
- To establish whether a steel is inherently fine or coarse grain the McQuaid-Ehn test is sometimes specified. This test reveals the austenite grain size after austenitizing the steel at $927 \pm 14^\circ\text{C}$ ($1700 \pm 25^\circ\text{F}$) for 8 hours.

Note that the test reveals the austenite grain size after a standard heat treatment and not the original grain size of the steel.

- Revealing the grain size

The austenite grain boundaries can be revealed by several methods. In one the steel is carburized by heating it at 927°C (1700°F) for 8 hours in a carburizing compound during which time carbon diffuses into the surface layers of the specimen raising the carbon content above 0.8%.

On cooling in a furnace proeutectoid cementite precipitates at the grain boundaries which shows up white after suitable etching.

- Reference to Fig. 13 shows a large difference in grain size between a coarse and a fine grain steel when heated at 927°C (1700°F). A grain size of ASTM 4 or larger (smaller number) is considered coarse grain and one of ASTM 5 or smaller (larger number) is fine grain although other values may be specified.
- Refer to ASTM E112-84 "Standard Method for Determining Average Grain Size" for details.

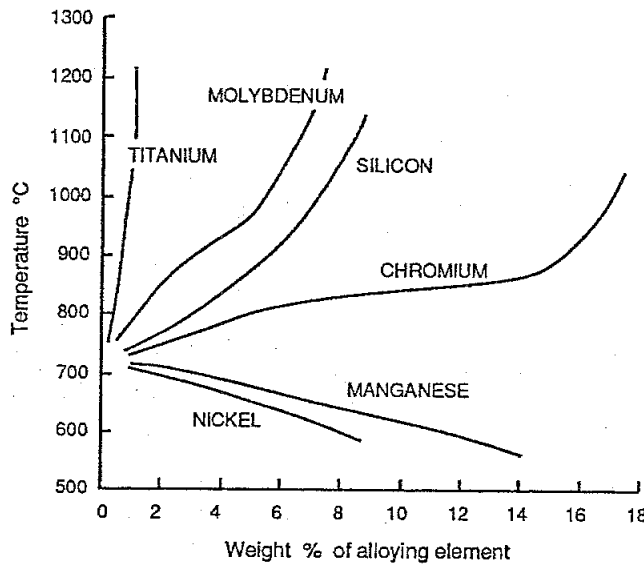


Figure 15. Diagram showing the effect of alloying elements on the eutectoid temperature in steel.

Fig. 15. In structural steels the concentration of alloys is quite small and the basic iron-iron carbide diagram is not distorted very much for equilibrium conditions.

The rate of cooling has a major effect on the types of microstructures formed and unless the steel cools slowly the iron-iron carbide diagram cannot be used. The reason is that the transformation of austenite to pearlite requires the diffusion of carbon to the sites of growing carbide, a process which takes time. We saw how a faster cooling rate produced finer pearlite. With even faster cooling rates less time is available for diffusion and pearlite cannot form. Alternate microstructures form with their exact morphology depending on just how quickly the steel cools. In a water quench, for example, the cooling rate is so rapid there is no time for any diffusion, and the carbon remains trapped in the same place as it was in the austenite. A rapid quench cannot suppress the crystal structure change from fcc to bcc but the presence of trapped carbon in the bcc phase distorts it to a tetragonal shape, shown in Fig. 16, rather than a true cubic structure.

Effect of rapid cooling

Quenching

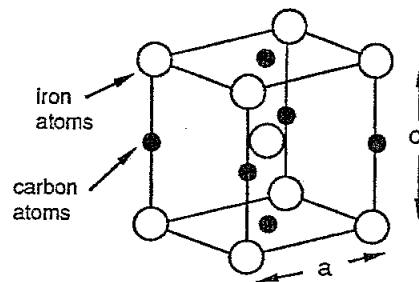


Figure 16. Carbon trapped in the martensite lattice causes a distortion to a tetragonal structure. The shape depends on the carbon level.

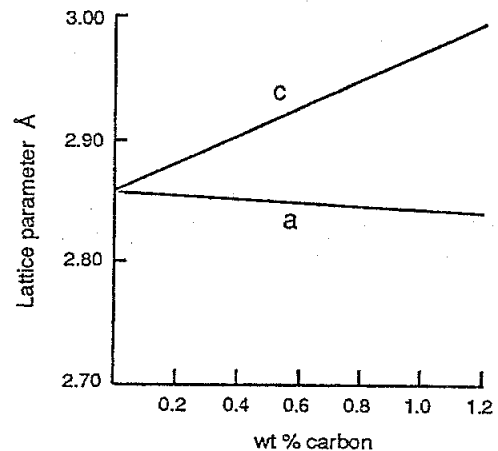
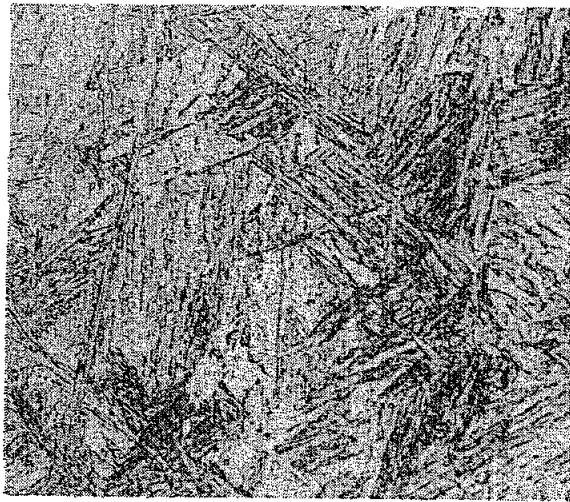


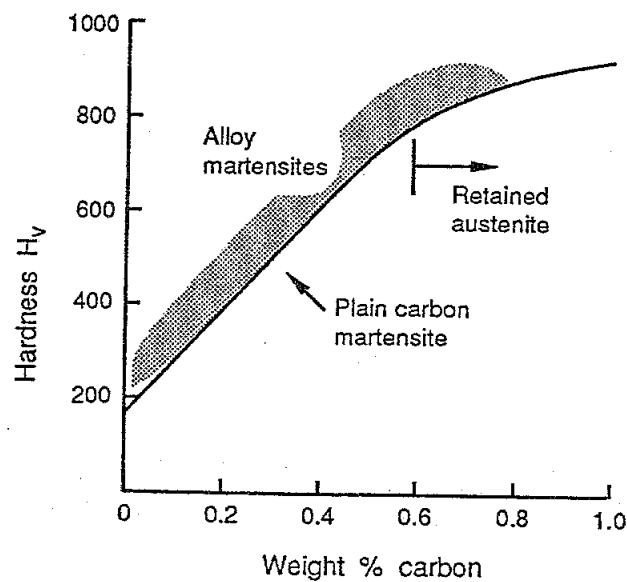
Figure 17. Photomicrograph of martensite.



Martensite

This constituent is known as martensite. Under the microscope martensite has the appearance (Fig. 17) of a mass of needles. Martensite can be a very hard and brittle constituent when it contains appreciable carbon. The hardness depends almost exclusively on the carbon content with other elements having little effect. Fig. 18 shows the relationship between martensite hardness and the carbon content.

Figure 18. Graph showing the effect of carbon content on the hardness of martensite. Other alloy elements have only a small effect on martensite hardness.



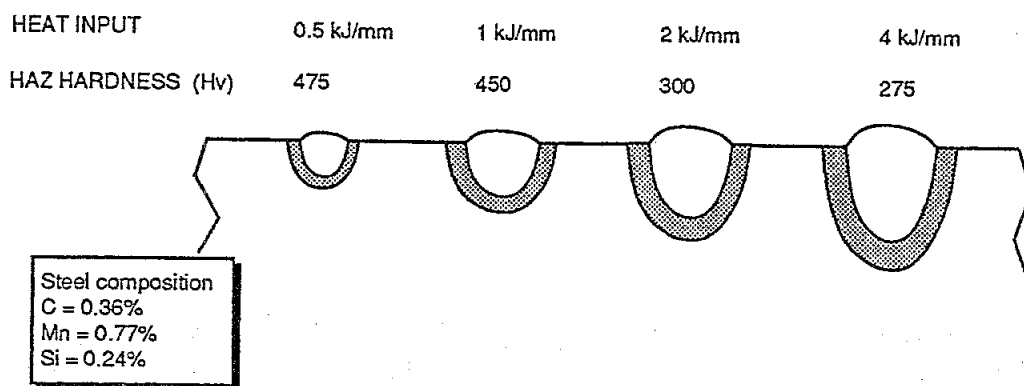
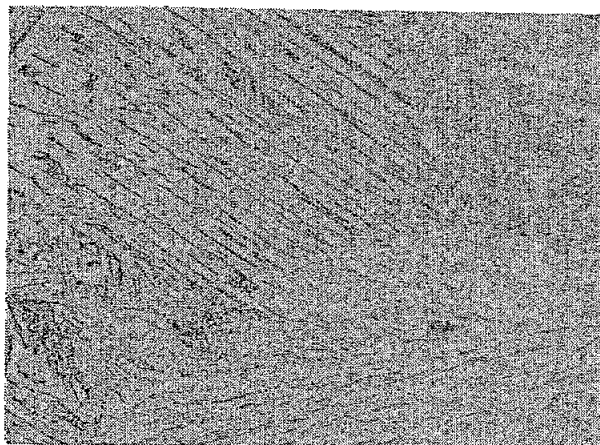


Figure 19. Examples of four welds of different heat input. The small welds cool rapidly enough to form martensite in the HAZ with a hardness exceeding 450Hv.

Illustrated in Fig. 19 are a number of welds in a carbon steel that have cooled at various rates. In each weld the heat affected zone transforms to a microstructure dependent on the cooling rate of that weld. For the small, rapidly cooled welds martensite is formed. For the large, slowly cooled welds the HAZ structure is pearlite. The hardness of the HAZ is much higher in those welds in which martensite is present.

Intermediate between a rapid quench that produces martensite and a slow cool producing pearlite other constituents may form, particularly in alloy steels. The most important of these is bainite. Bainite is still a two phase mixture of ferrite and iron carbide but unlike the cementite plates in pearlite the carbide in bainite is spherical (Fig. 20). Bainite formed above 300°C contains relatively coarse particles of the Fe_3C form of iron carbide (cementite) and is termed upper bainite. When formed below 300°C bainite has a much finer structure with the carbides tending to form striations across the ferrite laths. This is lower bainite. The carbides in lower bainite are $Fe_{24}C$ (ϵ carbide).

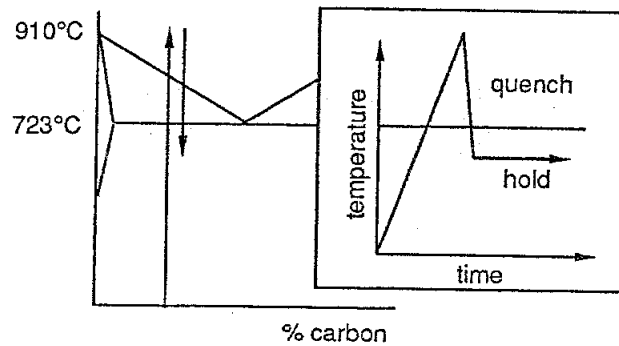
Figure 20. Photomicrograph of bainite. The growth rate is too rapid for carbides to form as plates of cementite as in pearlite, and instead the carbides form as particles.



Transformation diagrams

Since the iron-iron carbide diagram is only valid for very slow cooling rates, alternate diagrams for determining the constituents present in a more rapidly cooled steel have been developed. There are two types covering the case of continuous cooling and the case of isothermal transformation where the steel sample is held at a constant temperature until transformation is complete. We will consider the latter type first.

Figure 21. Procedure for determining time-temperature-transformation behaviour. Steel is heated to the austenitic region, quenched to a specific temperature then held at that temperature until transformation is complete.



Isothermal diagrams

Imagine a sample of steel heated until fully austenitic then quenched to some temperature below the equilibrium transformation temperature (Fig. 21). If we hold the steel at this temperature we find there is a delay before transformation begins and a further elapse of time while transformation takes place. The delay depends on the temperature at which the steel is held and we can plot this information on a diagram of temperature against time for a given steel composition.

An example of such a time-temperature-transformation (TTT) diagram for a carbon steel is shown in Fig. 22. Note that at high

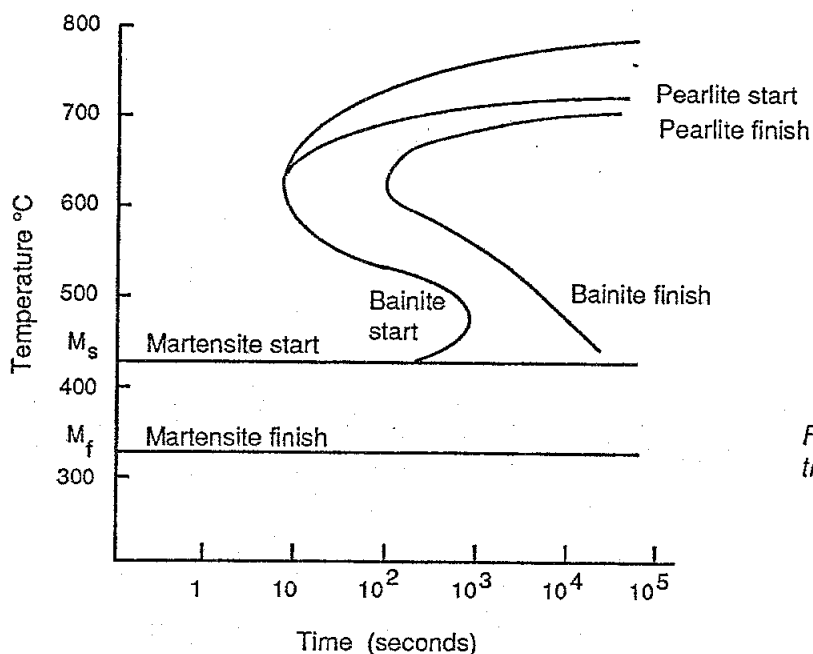


Figure 22. Typical TTT or isothermal transformation diagram.

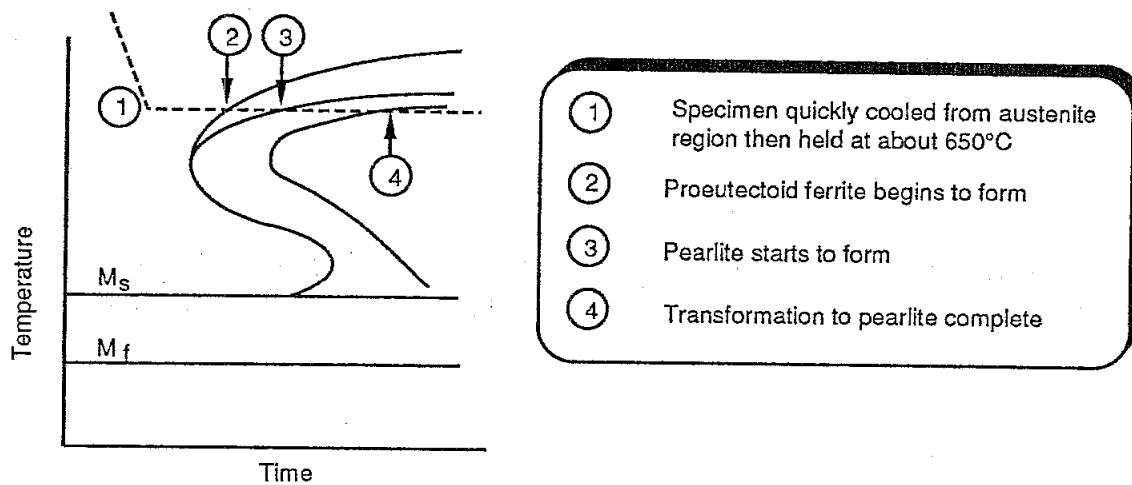


Figure 23. Steel undergoing isothermal transformation at about 650°C.

temperatures (Fig. 23) the steel transforms to proeutectoid ferrite followed by pearlite. At lower temperatures less proeutectoid ferrite is formed and the pearlite is finer. At about 550°C the pearlite forms in the shortest time and there is no pro-eutectoid ferrite (Fig. 24). Below this range (Fig. 25) transformation to bainite occurs, taking a longer time for lower temperatures. At a sufficiently low temperature

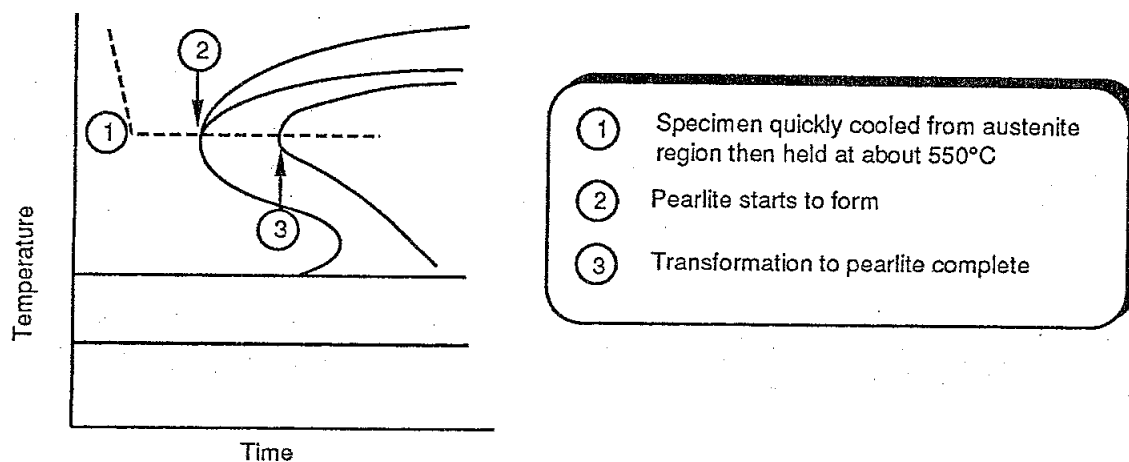


Figure 24. Steel transformed at about 550°C. No proeutectoid ferrite is produced, only pearlite.

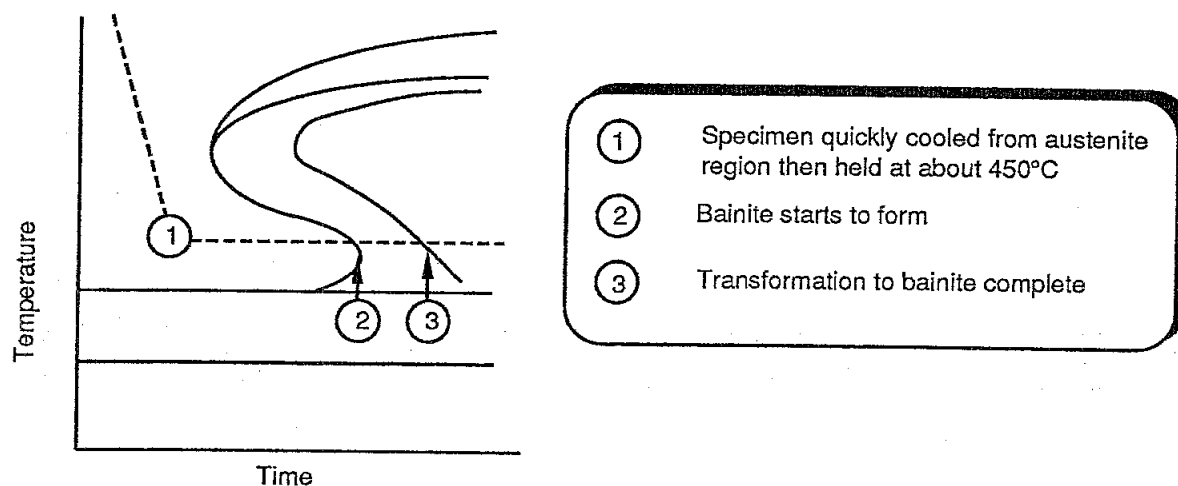


Figure 25. Transformation at about 450°C to bainite.

Martensite formation

(Fig. 26) martensite begins to form. Note that it forms almost instantaneously and does not grow as a function of time. For each steel there is a fixed temperature M_s at which martensite starts to form and a fixed temperature M_f at which transformation is complete. The percentage of martensite formed therefore depends only on the temperature to which the steel is cooled and not on how long it is held there.

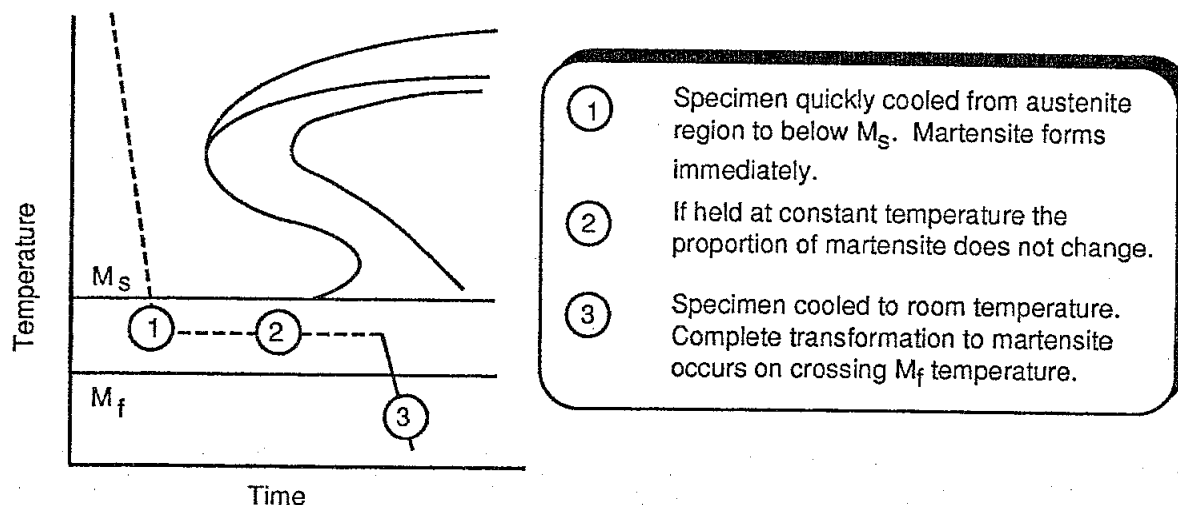


Figure 26. Specimen quenched to temperature below the martensite start. Martensite forms essentially instantaneously but transformation is not complete until the specimen is cooled below the martensite finish temperature.

CCT diagrams

Now consider the case of continuous cooling. We may superimpose a cooling curve on the TTT diagram in order to get an idea of what microstructures form, but it is more accurate to use a diagram established under continuous cooling conditions. The continuous cooling transformation (CCT) diagram is slightly different from the TTT diagram. The example for the steel in Fig. 27 reveals that there is a critical cooling rate at which the nose of the curve is missed and martensite forms. This is an important concept since the cooling rate at which martensite can form strongly influences the risk of cracking during welding and gives an indication of the 'weldability'.

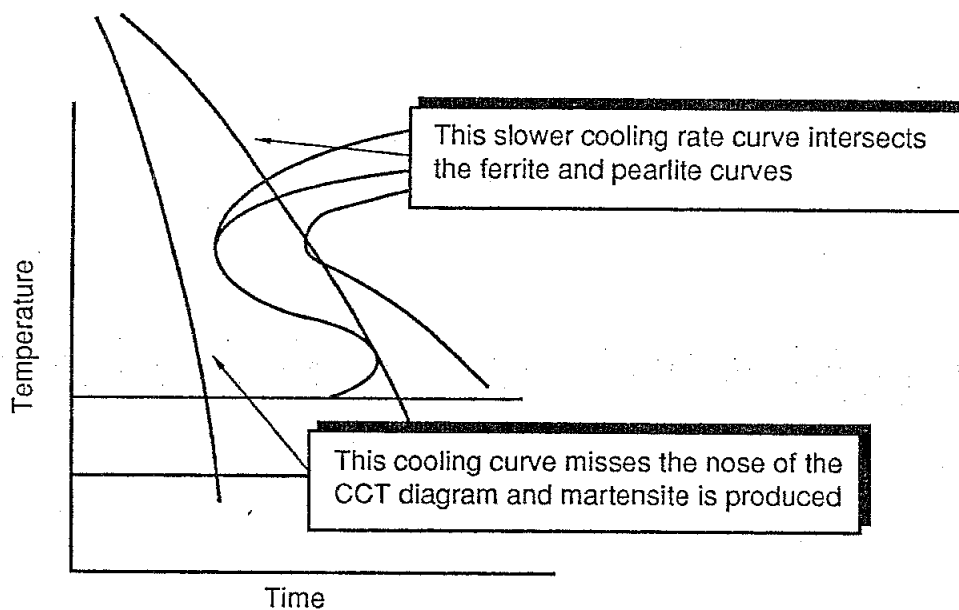


Figure 27. Transformation diagram with cooling curves superimposed. Continuous cooling transformation diagrams can be used with cooling curves representing weld cooling rates to determine microstructures produced in welds and heat affected zones.

Critical cooling rates

You should note that in plain carbon steels bainite cannot form during continuous cooling because of the shape of the transformation diagram. The bainite region is tucked under the pearlite area so a cooling curve either hits the pearlite curve or misses it completely. At cooling rates fast enough to miss the nose of the curve martensite is formed and at slower rates the cooling curve intercepts the ferrite start and a ferrite/pearlite structure results. This is also apparent in the example shown in Fig. 19 where the HAZ is martensitic for small welds and pearlitic for large, slowly cooled welds.

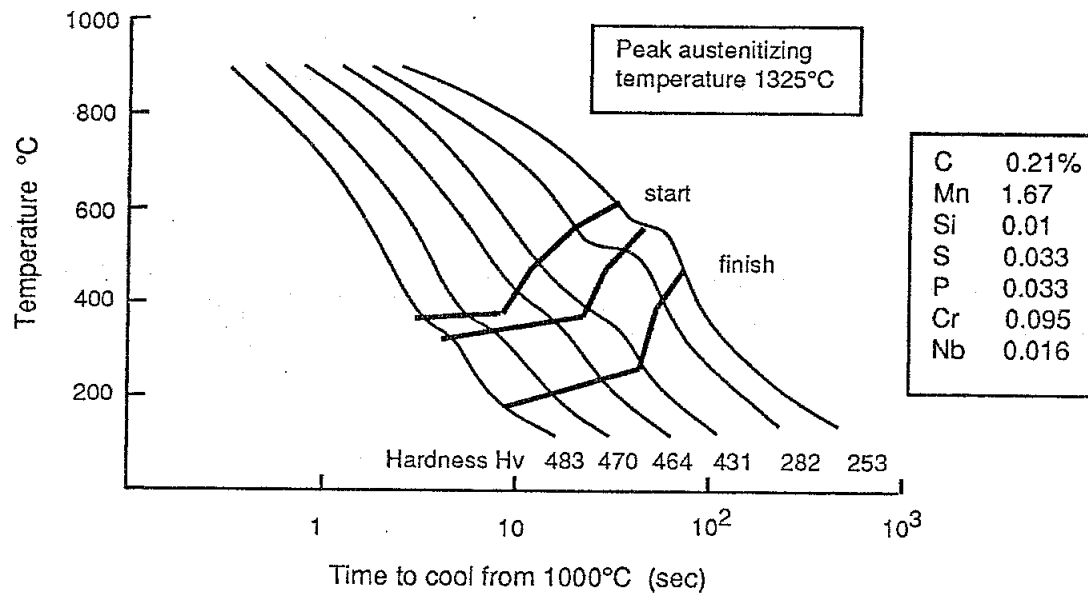


Figure 28. Typical CCT diagram for a C-Mn-Nb steel showing six cooling rates and the corresponding hardnesses. Welds cooling at these rates would be expected to show these hardness levels in the HAZ.

Determining CCT diagrams

The exact shape of a CCT curve depends on the details of the heating and cooling cycles. CCT diagrams can be established specifically for the conditions pertaining to welds but are different from the conventional curves in that they must be determined with high austenitizing temperatures to simulate the conditions in the HAZ. This can be done using a dilatometer in which the start of transformation is detected by sensing the changes in length of a small sample when it cools. An alternate method for establishing CCT diagrams is by thermal analysis on actual welds. A typical CCT diagram for the HAZ of a low alloy steel is shown in Fig. 28 for six standard cooling rates.

Effect of alloy elements

Alloy elements have significant effects on the shape of the CCT and TTT diagrams which allow different microstructures to be produced in alloy steels. Molybdenum and boron, for example, shift the top (pearlite) part of the curve to the right, to longer times, thus exposing the bainite region. Steels containing these elements may produce bainite on continuous cooling. The entire curve may also move to greater times allowing martensite to form at much slower cooling rates. This increases the hardenability of the steel. The other notable effect of alloy elements is to change the martensite start and finish temperatures. Increasing carbon, for example, depresses the

M_s and M_f temperatures

martensite start to lower temperatures as we show in Fig. 29. Other elements affect martensite formation and the combined effect can be approximated by the following equation:

$$M_s(^{\circ}\text{C}) = 550 - 350 \times \%C - 40 \times \%Mn - 35 \times \%V - 20 \times \%Cr - 17 \times \%Ni \\ - 10 \times \%Cu - 10 \times \%Mo - 5 \times \%W + 15 \times \%Co + 30 \times \%Al$$

If M_f is below room temperature martensite transformation is not completed and the steel contains retained austenite.

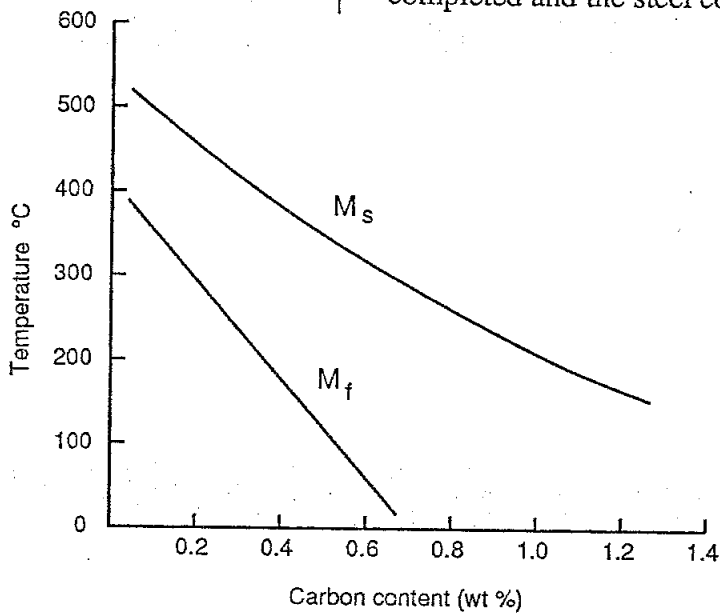
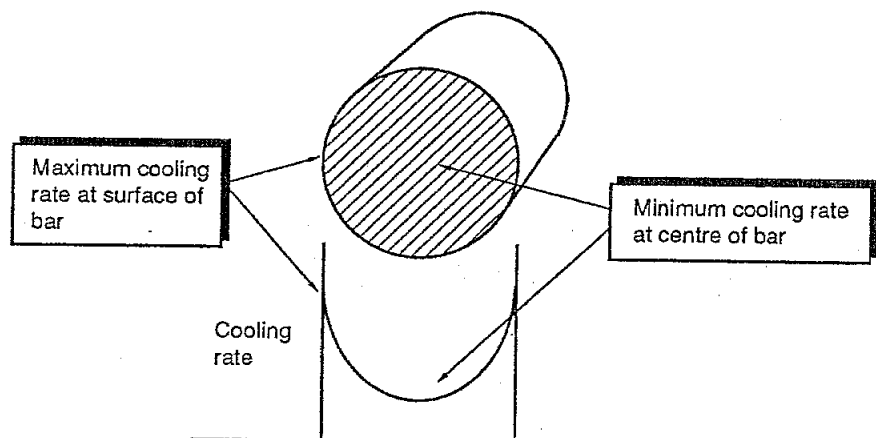


Figure 29. The effect of carbon content on the martensite start and finish temperatures.

Heat treatment

If we heat a bar of steel in the austenitic region (above 910°C) then quench it in water or brine (salt water) the cooling rate at the surface of the bar will be much higher than at the centre of the bar (Fig. 30). Thus the microstructure formed in the surface region could be quite different from that formed in the interior. In a large bar the centre

Figure 30. When a hot bar is quenched the cooling rate is a maximum at the surface but is slower in the centre of the bar.



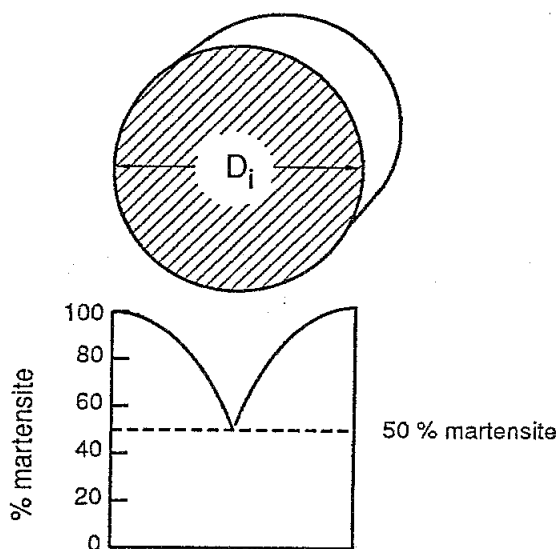


Figure 31. The ideal critical diameter is that which will just produce 50% martensite at the centre of the bar under an ideal quench.

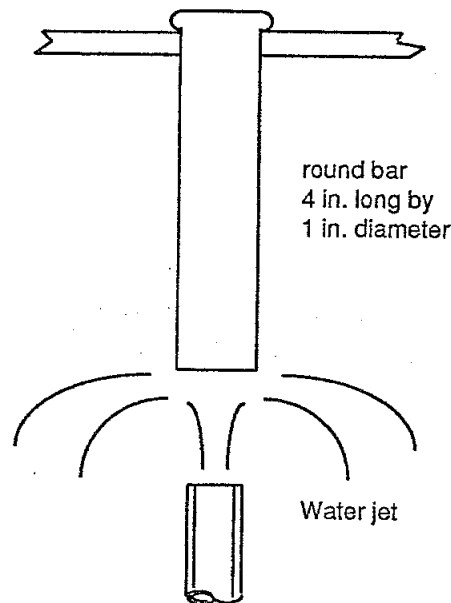


Figure 32. Arrangement for carrying out the Jominy end quench test for measuring hardenability.

Hardenability

could cool so slowly that no martensite is formed and the steel would not be hardened at all. The distance below the surface that is hardened is of great practical importance since it determines the diameter of bar that can be hardened right through. The diameter of bar that will just produce 50% martensite in the centre under ideal, perfect quenching conditions is known as the ideal critical diameter D_i as we show in Fig. 31. It is a measure of the steel's hardenability. If we add alloy elements that allow a larger diameter bar to produce 50% martensite at the centre then the hardenability has been increased.

Measuring hardenability

We can determine the hardenability by performing a Jominy end quench test illustrated in Fig. 32. The test specimen is a cylinder one inch diameter and four inches long which is heated to the hardening temperature then placed in a fixture where it is quenched by water impinging on one end. The fastest quench is on the end of the specimen with progressively slower cooler rates being experienced away from the end. The specimen is examined by grinding two flats on opposite sides and making hardness measurements at intervals along the length from the quenched end. The distance from the end at which 50% martensite is produced can be correlated directly with the hardenability D_i .

Calculating hardenability

Many researchers have attempted to establish quantitative relationships between the composition of the steel and the hardenability. The classic work of Grossman led to the concept of multiplying factors for each element. For a given carbon level and austenite grain size a base diameter $D_{I \text{ carbon}}$ is determined and these are listed in Table 1. We then multiply this by a factor for each of the alloying elements depending on their concentration. Factors are shown graphically in Fig. 33. The example illustrates how this works. This method has been superseded by more sophisticated computer based models but is still useful for approximations. Notice that since the hardenability depends on the *product* of various factors we can obtain high hardenability by using small quantities of several alloy elements rather than a large quantity of just one element.

Table 1. Base diameters (in inches) for carbon steels of various austenitic grain sizes.

Carbon %	Base diameter $D_{I \text{ carbon}}$		
	ASTM Grain Size		
	#6	#7	#8
0.05	0.0814	0.0750	0.0697
0.10	0.1153	0.1065	0.0995
0.15	0.1413	0.1315	0.1212
0.20	0.1623	0.1509	0.1400
0.25	0.1820	0.1678	0.1560
0.30	0.1991	0.1849	0.1700
0.35	0.2154	0.2000	0.1842
0.40	0.2300	0.2130	0.1976
0.45	0.2440	0.2259	0.2090
0.50	0.2580	0.2380	0.2200
0.55	0.273	0.251	0.231
0.60	0.284	0.262	0.241
0.65	0.295	0.273	0.251
0.70	0.306	0.283	0.260
0.75	0.316	0.293	0.270
0.80	0.326	0.303	0.278
0.85	0.336	0.312	0.287
0.90	0.346	0.321	0.296

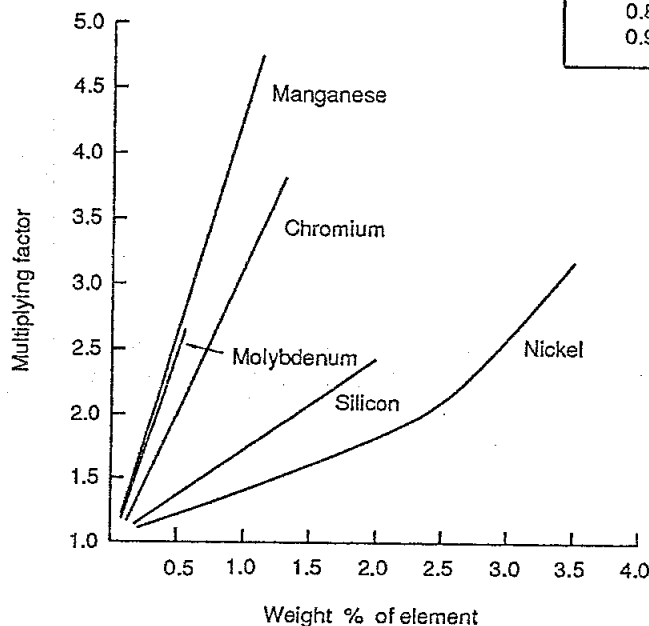


Figure 33. Values of multiplying factors for various alloy elements as a function of the alloy content. For greater accuracy refer to the values published in tabular form.

EXAMPLE OF HARDENABILITY CALCULATION

In this example we calculate the hardenability of AISI 4340 steel. Hardenability is measured by the Ideal Critical Diameter. This is the diameter of a round bar that will produce 50% martensite at the centre when given an ideal quench.

- Step 1.** Determine the base diameter from the carbon content using Table 1. We will assume the grain size of the austenite before quenching is ASTM 8.

Base diameter = 0.1976 in.

Assumed average composition of AISI 4340 (wt%)

C	Si	Mn	Cr	Mo	Ni
0.40	0.25	0.70	0.8	0.25	1.8

- Step 2** Determine multiplying factors for each of the alloying elements. Use Figure 33 or published tables for greater accuracy.

Alloy element	wt %	Multiplying factor
Silicon	0.25	1.2
Manganese	0.70	3.3
Chromium	0.80	2.7
Molybdenum	0.25	1.75
Nickel	1.8	1.7

- Step 3** Calculate ideal critical diameter

$$D_i = D_{i \text{ carbon}} \times \text{multiplying factors}$$

$$D_i = 0.1976 \times 1.2 \times 3.3 \times 2.7 \times 1.75 \times 1.7 = 6.29 \text{ in.}$$

Note: A quench in practice is never ideal. Real quenching media always cool the specimen slower than the ideal. The actual critical diameter is therefore less than the ideal. In the present example a water quench without any agitation will reduce the ideal critical diameter to a value of about 5.5 in.

Hardenability in the HAZ

Carbon equivalent

Similar attempts have been made to develop models for hardenability in welding. Depth of hardening or critical diameters are not relevant concepts in a welding situation, but we are interested in the hardness produced at a given cooling rate or the critical cooling rate to produce a given hardness in the heat affected zone. The simplest model is one in which the effects of individual alloying elements are added together (a linear model) in a carbon equivalent which in turn relates to the critical cooling rate to produce a given hardness. Fig. 34 shows the carbon equivalent plotted against critical cooling rate at 540°C to give a hardness of 350 Hv, and a reasonable correlation exists. As might be expected from a consideration of the multiplying effect apparent in the Grossman approach, the linear carbon equivalent method tends to over estimate hardenability for plain carbon steels and under estimate it for multialloyed steels. It is useful, nevertheless, in estimating welding conditions for the normal composition ranges of common weldable structural and pressure vessel steels. Carbon equivalents are used widely in industry as measures of weldability and it is important to understand their origin. Several different formulae have been proposed and some are incorporated into national standards and codes.

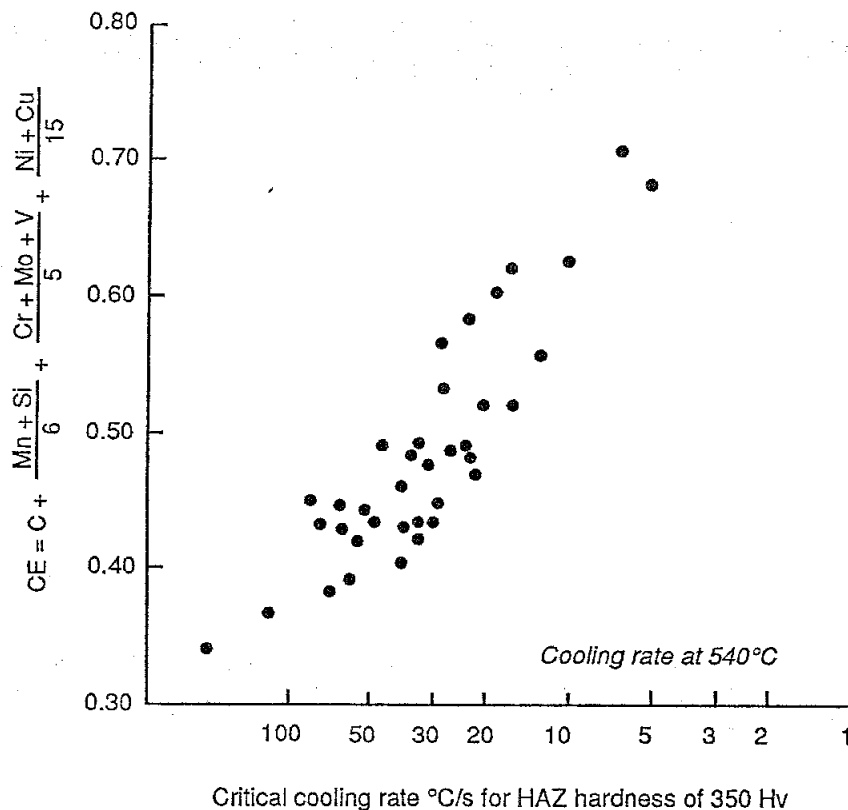


Figure 34. Relation between the carbon equivalent and the critical cooling rate in a weld to produce a heat affected zone hardness of 350Hv.

HAZ hardness

Several models have been proposed for predicting the hardness of the heat affected zone and we give one example here for structural steels:

$$H_v = 2019[(1 - 0.5 \log_{t_{8/5}})C + 0.3(CE_B - C)] + 66(1 - 0.8 \log_{t_{8/5}})$$

H_v is the Vickers (diamond pyramid) hardness of the HAZ. The term CE_B represents the steel composition as a linear carbon equivalent:

$$CE_B(\%) = C + \frac{Si}{11} + \frac{Mn}{8} + \frac{Cr}{5} + \frac{Mo}{6} + \frac{V}{3} + \frac{Ni}{17} + \frac{Cu}{9}$$

and $t_{8/5}$ is the time, in seconds, for the HAZ to cool from 800°C to 500°C.

Tempering

Martensite produced in a quenched steel is hard and brittle and in most cases the steel is unusable in that form. The toughness may be improved by a process of tempering. This involves reheating the steel to a temperature below that for transformation, holding for a period of time, then cooling to room temperature. During tempering the carbon trapped as an interstitial in the martensite tetragonal structure is released. Carbon atoms diffuse and precipitate as small carbides. With enough time and at sufficiently high temperatures cementite forms, not as plates as in pearlite, but as round particles. Improvement in toughness is accompanied by a loss of hardness which is a function of both temperature and time. Over limited ranges these two factors may be combined into a single parameter as proposed by Holloman and Jaffe:

$$H = T (20 + \log_{10} t)$$

where T is the temperature in degrees kelvin ($^{\circ}K = ^{\circ}C + 273$) and t is the holding time in hours. Fig. 35 shows the hardness of a carbon steel

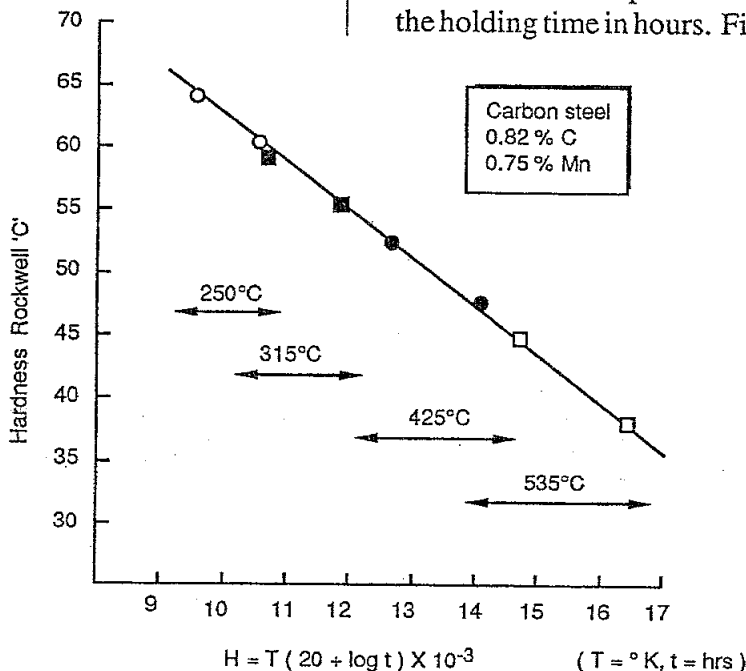
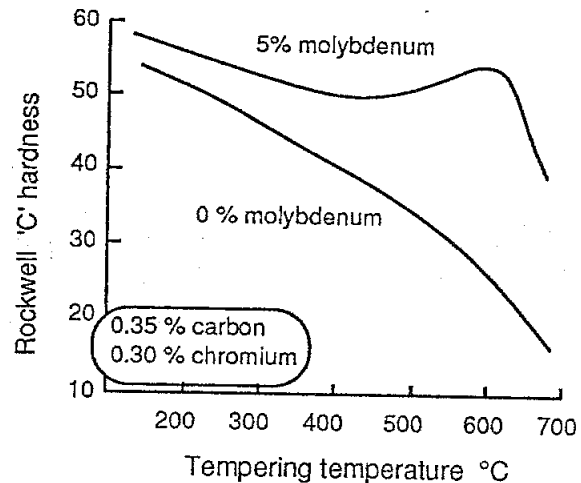


Figure 35. The effect of tempering time and temperature on hardness of a quenched steel. Temperature and time are combined in the Holloman-Jaffe parameter, H .

Figure 36. Hardness as a function of tempering temperature for a plain carbon and a molybdenum steel. The molybdenum steel shows secondary hardening.



Secondary hardening

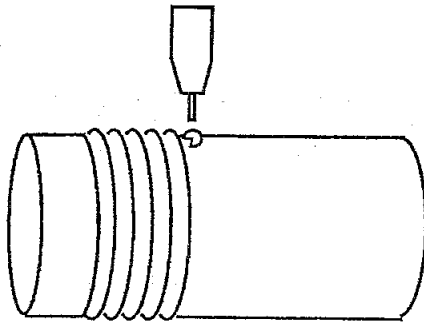
quenched and tempered at various temperatures and times. Notice that the hardness loss is far more sensitive to temperature than time. An increase of 50°C (at 600°C) is equivalent to increasing the time *ten* times. The temperatures typically selected for thermally stress relieving welded steel structures are high enough to cause rapid tempering. Many of the benefits derived from a thermal stress relief are actually due to tempering.

In some steels containing specific alloy elements tempering may actually cause an increase in hardness as the tempering temperature is raised as Fig. 36 shows. Strong carbide forming elements such as molybdenum (Mo), chromium (Cr), and tungsten (W), cause alloy carbides to be precipitated in certain temperature ranges. This behaviour of secondary hardening is put to good use in the design of tool steels (high speed steels) that need to retain their hardness at the high temperatures experienced in machining.

Welding Q&T steels

Welding of quenched and tempered steels may pose several problems and a careful study of the steel and intended application is necessary before specifying the welding procedure. If welding is carried out after the heat treatment it may be difficult to achieve uniform properties across the joint. In some cases, therefore, it is preferable to perform the quenching and tempering after completing the weld. This assumes that the component is small enough to allow this. The weld metal in this case must be selected to have a matching composition so that uniform properties are achieved after quenching and tempering.

Preheat for Roll Build-up



A roll is to be built up by welding. The roll is turned and the weld metal deposited by SAW in a continuous spiral bead. A uniform hardness is required in the weld metal and HAZ to avoid uneven wear in service.



	C	Mn	Cr	Si	Ni	Mo
Roll	0.5	1.0	1.5	0.5	1.0	0.5
Deposit	0.15	1.9	12.0	1.0	-	-

To avoid each bead tempering the one before the weld metal and HAZ must be held above the M_s so that transformation is delayed long enough for several beads to be deposited.

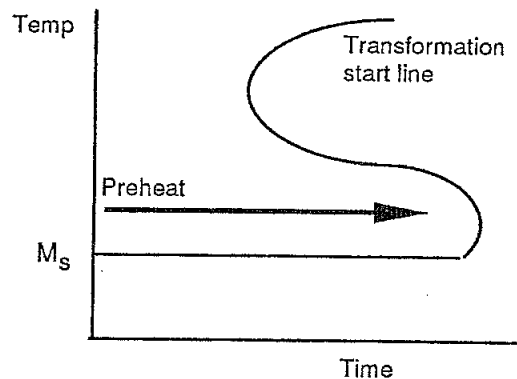
Deposit must be held at a temperature in the "bay" of the TTT diagram. This gives the longest delay before transformation. Preheat is above the M_s .

From the formula on page 21 estimate the M_s for the roll and deposit metal.

M_s (roll) $\approx 283^\circ\text{C}$

M_s (deposit) $\approx 183^\circ\text{C}$

Preheat specified 300°C



Effects of preheating

A second method of achieving uniform properties without heat treating after welding is to apply sufficient preheat to the component to hold it above the martensite start temperature until welding is complete. All deposited metal and the HAZ remain austenitic during welding and transform together on cooling producing a uniform structure. This avoids any tempering of adjacent weld beads as each bead is deposited. In applying this approach the TTT diagram for the steel can be studied to determine the preheat temperature and the maximum allowable time for completion of welding. The method is applied for build up and surfacing of alloy steel shafts but is not used on structural steels. The example shown illustrates how the TTT diagram can be used to develop a procedure for this type of application.

Stress relief of Q&T steels

If the weld is not to be heat treated or the isothermal method cannot be used then the problem will be welding the steel without cracking and obtaining adequate weld properties. The high hardenability of a Q and T steel means the HAZ probably transforms to martensite on cooling and this must be prevented from cracking by applying adequate preheat. A high preheat may cause autotempering as the martensite forms but stays hot. The weld metal must be selected to provide adequate strength without the benefit of a subsequent quench and temper.

It is common practice to apply a postweld temper (stress relief) heat treatment which softens the HAZ, removes hydrogen and lowers residual stresses. The temperature for this treatment must not be higher than the original tempering temperature of the steel otherwise the steel strength could drop below specification. To reduce the risk of cracking the post weld heat treatment may be carried out immediately after welding without letting the weld cool down.

PROPERTIES OF STEELS

Effect of carbon

Plain carbon steels represent the simplest family of steels in which various strength levels are achieved through the appropriate carbon content. In a slowly cooled steel the structure is pearlitic and the strength is proportional to the amount of pearlite formed. The general relation between carbon and strength is shown in Fig. 37. Increasing the percentage of pearlite has a stronger effect on the ultimate tensile strength than on the yield strength. Carbon steels, therefore, tend to have a lower yield to ultimate ratio than low alloy steels. This is one reason carbon steels are popular for pressure vessels where the UTS usually determines design requirements. In structural applications (e.g., bridges) where the yield stress controls design low alloy steels with higher yield to ultimate ratios are favoured.

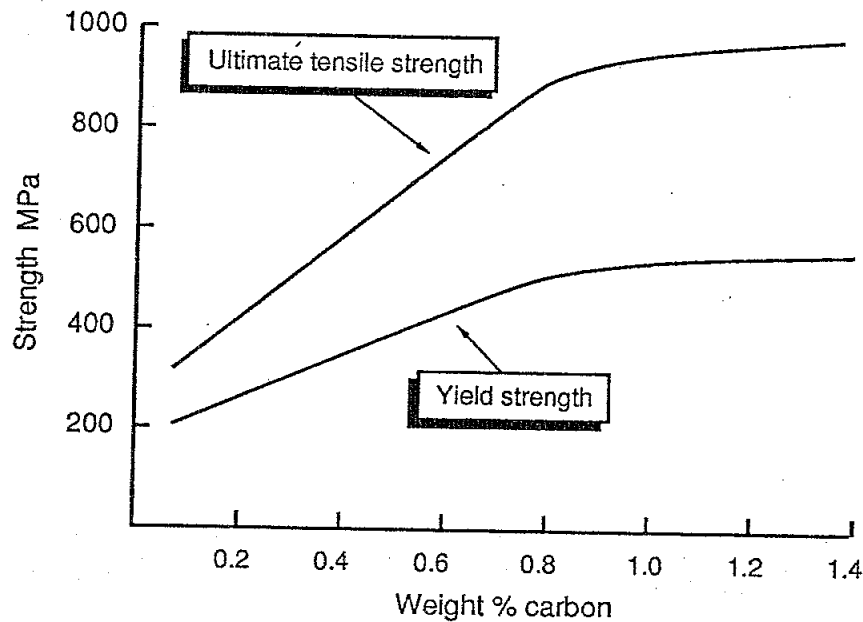


Figure 37. The effect of carbon content on yield and ultimate strength in a plain carbon steel.

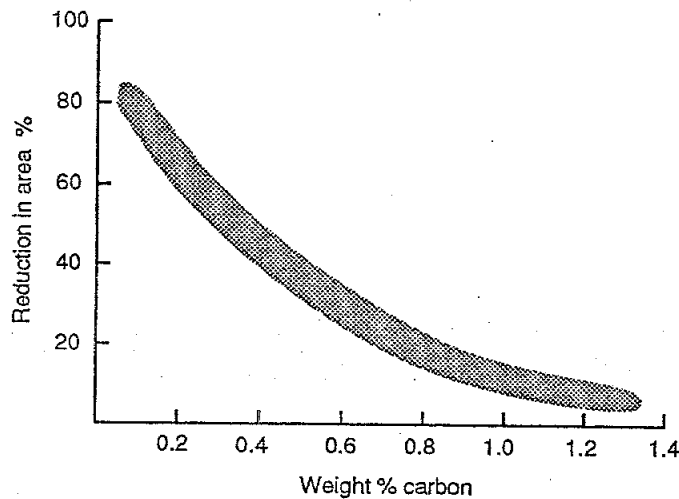


Figure 38. The effect of carbon content on ductility in plain carbon steels.

Effect on ductility

At 0.8% the steel is 100% pearlite and the maximum strength is obtained. Any increases of carbon above 0.8% serve only to produce pro-eutectoid cementite with no further increase in strength. Plain carbon steels of about 0.8% carbon are often specified where maximum strength is desired, for example rails and piano wire.

* Increasing carbon content, however, reduces the ductility (as shown in Fig. 38) and this often limits the applications for plain carbon steels. The impact toughness is also lowered and the effect of carbon content on the charpy transition curve is illustrated in Fig. 39.

The strength of a plain carbon steel depends to a certain extent on the coarseness of the pearlite, a very coarse pearlite giving a strength about 20 ksi (137 MPa) less than a fine pearlite. Very fine pearlite can be achieved in an isothermal transformation by selecting a temperature level with the nose of the TTT diagram where transformation occurs in the shortest time. In the wire industry this heat treatment is known as patenting and is used to impart very high strength and good ductility on the wire enabling it to undergo the drawing operation.

Figure 39. Effect of carbon content on the charpy V-notch impact energy transition curves of plain carbon steels.

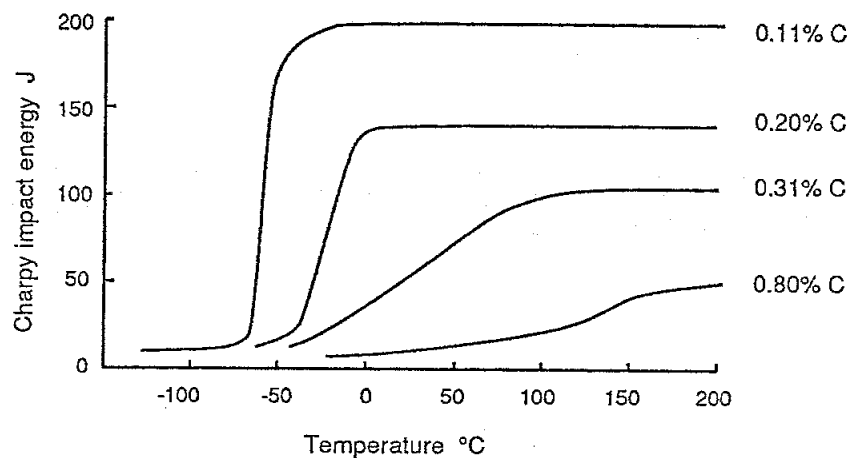


Table 2. Maximum specified carbon levels for various thicknesses of ASTM A516.

Thickness range	Carbon content (max) for Grade 70
1 in. (25 mm) and under	0.31
Over 1 to 2 in. (25 to 50 mm), incl.	0.33
Over 2 to 4 in. (50 to 100 mm), incl.	0.35
Over 4 to 8 in. (100 to 200 mm), incl.	0.35
Over 8 in. (200 mm)	0.35

Effect of thickness

Under continuous cooling conditions the coarseness of pearlite depends on cooling rate. A thick plate cooling slower produces a coarser structure with lower strength. This is reflected in the specifications for carbon steels which may have lower specified strength requirements for thick sections or may allow higher carbon levels to compensate. An example for A516 is shown in Table 2. This has important implications for welding since not only does a weld on a thick plate cool faster than one on a thin plate, but the thick plate is likely to have a higher carbon content. These two factors greatly increase the risk of cracking in thick plates.

The coarsest pearlite and ferrite structure is achieved in a carbon steel after a full anneal, i.e. heating to fully austenitize at (900°-950°C) followed by a very slow cool in a furnace. A full anneal is used when maximum softening is required but the coarse microstructure has very low toughness. Although the ductility in a tensile test is improved, the impact toughness is poor and the fully annealed condition is rarely desirable in a steel for structural use. A full anneal might be specified prior to machining, particularly when this is followed by a further heat treatment.

Normalizing

The toughness of carbon structural and pressure vessel steels can be improved by normalizing to reduce the grain size (Fig. 40). This treatment involves austenitizing at about 900°C followed by an air

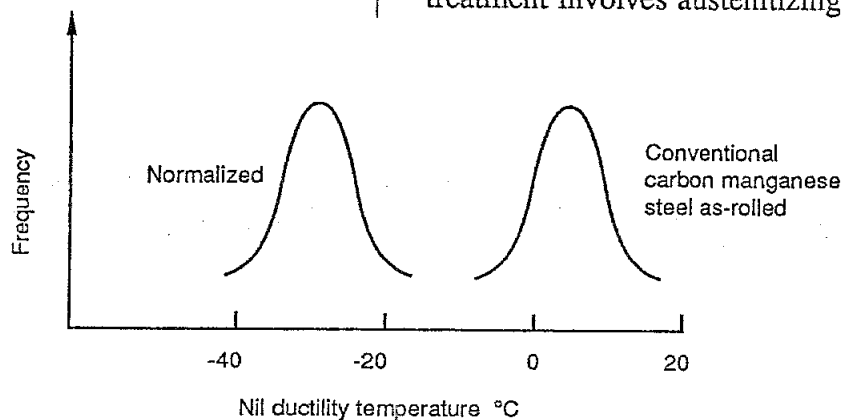


Figure 40. The toughness of carbon steels is improved by normalizing. The diagram shows a typical distribution of NDT temperatures for as-rolled and normalized steels.

T.G. CTOD - what is it & what does it show

CRACK TEST OPENING DISPLACEMENT TEST.

MOD 12 PPS1

Welding Metallurgy of Steels

Normalizing temperature

cool. The temperature is the lowest practical temperature at which the steel becomes fully austenitized, and the holding time need only be sufficient to ensure complete austenitization has occurred. Prolonged heating in fact may cause grain growth which is undesirable. A holding time of 1/2 hour at temperature for carbon steels up to 75 mm is typical although various codes and standards may specify differently.

Some of the older steel specifications allow a hot forming operation carried out by the fabricator to count as the normalize. If the fabricator chooses this option there are two important things to note: First, the steel may be supplied in the as-rolled (un-normalized) state, but the mill test certificates will report tests on coupons normalized in the laboratory. Second, the temperature for hot forming must not significantly exceed the normalizing temperature otherwise grain growth and toughness deterioration could occur. In this regard fine grain steels are more tolerant to higher temperatures than those with coarse grain.

Cooling after normalizing

To prevent coarse pearlite forming, normalizing is followed by an air cool rather than a furnace cool. An inadequate cooling rate could result in toughness deterioration. If several components are normalized together they must be properly stacked to allow good air circulation between them. With thick sections a forced air cool using fans may be specified, but care must be taken to avoid uneven cooling that results in distortion.

Alternate methods of strengthening

The negative effects of carbon on toughness and weldability have led steel makers to employ other strengthening mechanisms. The most recent and important for structural steels are precipitation hardening and grain refinement.

The first generation of high-strength low-alloy structural steels used carbo-nitride forming elements vanadium and niobium (columbium) as precipitation hardening agents. These elements had the advantage they could be added to semi-killed steels because of their limited tendency to form oxides. The increased strength due to precipitation hardening allows a reduction in carbon content so improving weldability and toughness. Typical of these steels are ASTM A441 and A572.

More recent steels have employed grain refinement in addition to precipitation hardening to produce high strength coupled with excellent toughness. Fine grain is produced by thermo-mechanical processing in which the reduction of plate thickness by rolling is integrated

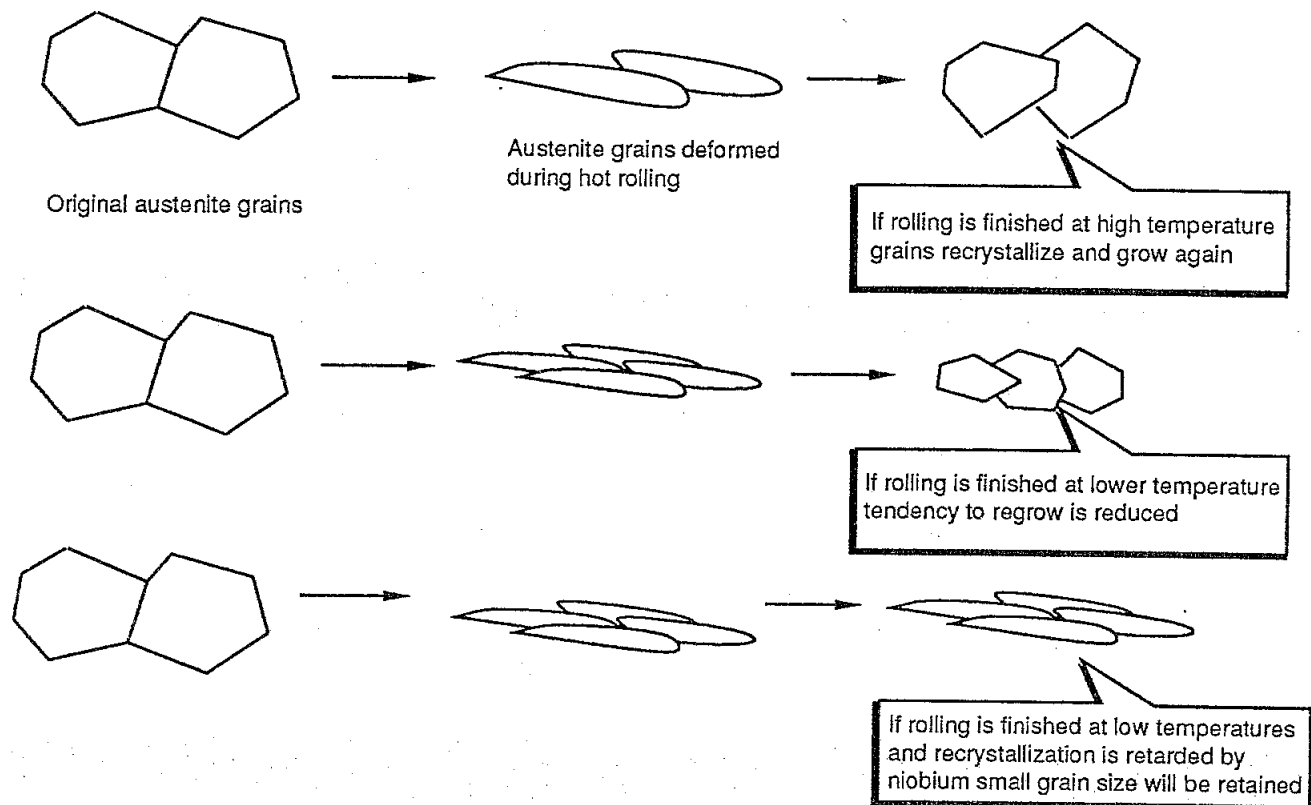


Figure 41. Using lower finishing temperatures during hot rolling and micro-alloy elements that retard recrystallization can result in steels of very fine grain size.

Controlled rolling

with temperature. During hot rolling the austenite grains breakup, but at high temperatures they recrystallize and grow again (Fig. 41). If rolling is performed at lower temperatures (controlled rolling) the tendency to regrow austenite grains is less. The recrystallization of the austenite grains is further retarded by the presence of microalloy elements—particularly niobium—and the austenite range is extended to lower temperatures by alloys such as manganese.

A steel maker may use controlled rolling for the lighter sections of plate and normalizing for thick sections when excellent toughness is required. Several other devices are being developed to further improve properties among them being on-line accelerated cooling and direct quenching.

Effect of work hardening

Since heating a steel above 400°C causes recovery and recrystallization to begin, any process involving temperatures above this removes the effects of work hardening. Thus castings, hot forged, hot rolled, or any heat treated product will not gain any strength from work hardening. But any cold finished product or one subject to subsequent cold working such as bending or forming will have a strength component that results from work hardening. In Fig. 42 we show the increase in strength due to cold deformation while Fig. 43 reveals the accompanying reduction in ductility. Work hardening as a strengthening method is particularly valuable in the production of wires. The wire drawing process itself provides a greater degree of reduction than is achieved with any other method. High strength achieved in low carbon steel wires is particularly valuable in making gas metal arc welding electrodes. GMAW wires are usually pushed by a wire feeder along a conduit to the welding gun, and a high column strength is desirable to prevent kinking. Previous editions of AWS A5.18 "Specification for Carbon Steel Filler Metals for Gas Shielded

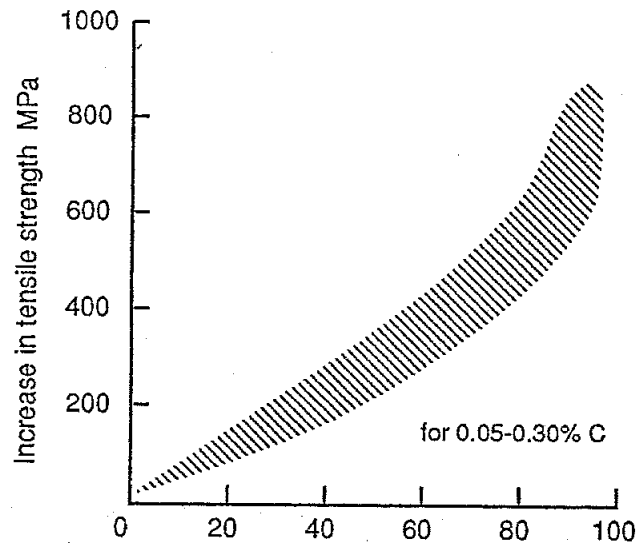


Figure 42. The effect of cold rolling on tensile strength of steel.

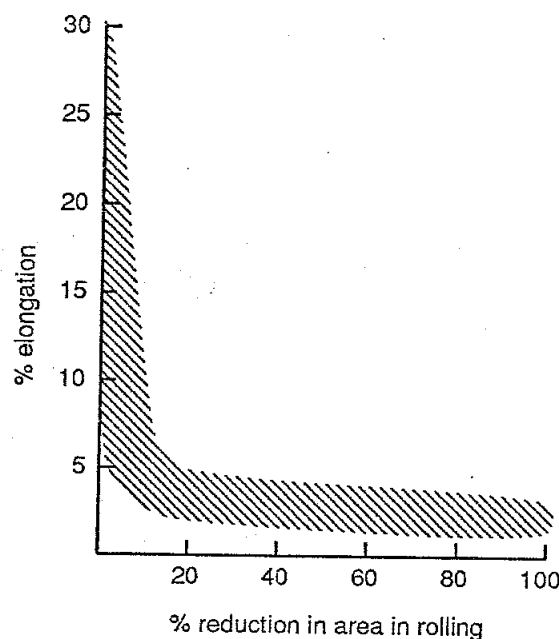


Figure 43. Effect of cold rolling on ductility as measured by elongation in a tensile test.

Restoring ductility

"Arc Welding" have included a requirement for the minimum tensile strength of the wire in addition to the weld deposit. For wires up to 1.2 mm (0.045 in.) the minimum strength is 930 MPa (135 ksi).

Notice in Fig. 43 that the loss of ductility is substantial for the first ten percent of deformation. In bending steel plates for structural applications it may be necessary to give the plate a thermal stress relief to restore ductility when deformation levels are substantial. ASME Section VIII Division 1 clause UCS-79 requires a stress relief heat treatment for most vessel shell sections, heads and other pressure boundary parts of carbon and low alloy steels if the extreme fibre elongation exceeds 5%.

When a cold worked material is welded the heat from welding may be sufficient to cause a marked loss of strength in the HAZ where recrystallization takes place. This may be important, for instance, when resistance welding sheet that has a hard temper. The loss of strength or width of softened zone can be reduced by modifying the welding procedure to reduce the time.

THE HEAT AFFECTED ZONE

Structure

Adjacent to the weld the metal undergoes various changes according to the peak temperature and cooling rate experienced at each location. Close to the fusion boundary the peak temperature will be high enough to cause complete transformation to austenite and some grain growth. At some distance from the fusion boundary the peak temperature is insufficient to cause any microstructural changes although other effects such as strain aging may occur. In between a range of mixed structures may be observed. Fig. 44 illustrates the various zones adjacent to the weld metal.

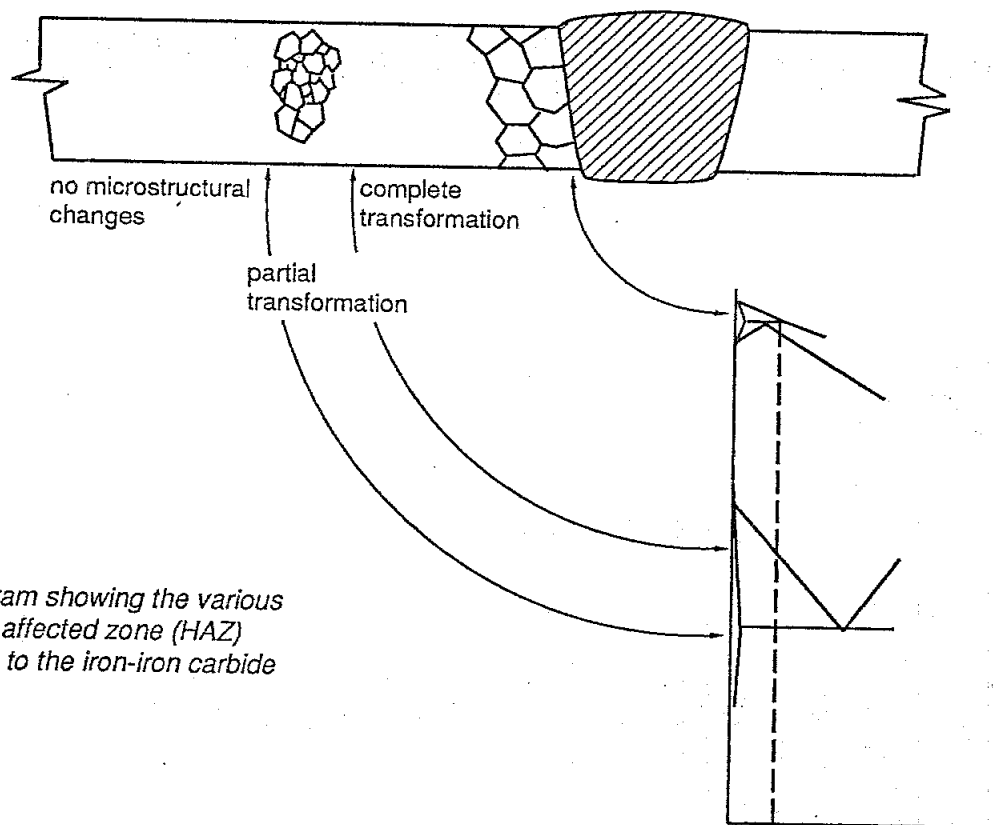
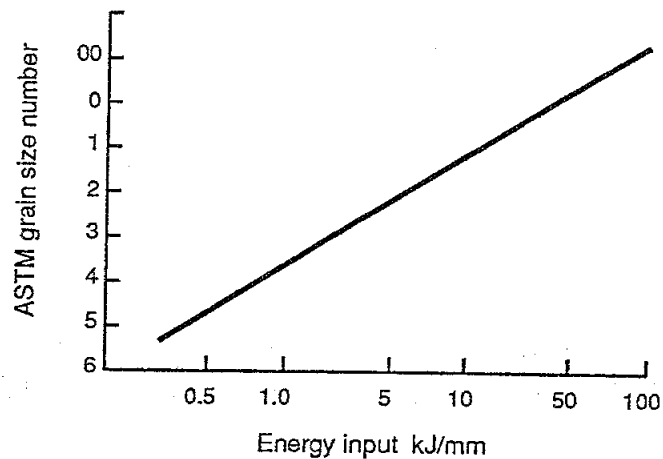


Figure 44. Diagram showing the various parts of the heat affected zone (HAZ) and their relation to the iron-iron carbide phase diagram.

Figure 45. Austenitic grain size close to the fusion boundary as a function of weld heat input.



HAZ grain size

The austenite grain size in the HAZ is controlled mainly by the weld heat input but it is also influenced by the shape of the fusion boundary. The effect of heat input on austenite grain size in the HAZ is shown in Fig. 45. In very high heat input processes, such as electroslag, the HAZ is very wide and the grains can grow to their equivalent isothermal size, i.e. the same size grains as in a specimen uniformly heated in a furnace at the same temperature. The heat input affects the time at peak temperature and also the temperature gradient in the HAZ. A steep gradient from a low heat input weld restricts the size to which grains can grow. This effect is seen in welds that have non-circular fusion boundary shapes. The concave region of the submerged arc weld shown in Fig. 46 has a large austenite grain size compared with the convex region despite the same energy input. The effect can influence metallurgical behaviour. For example the hardest region in the HAZ is usually in the concave region because the larger grain size increases hardenability.

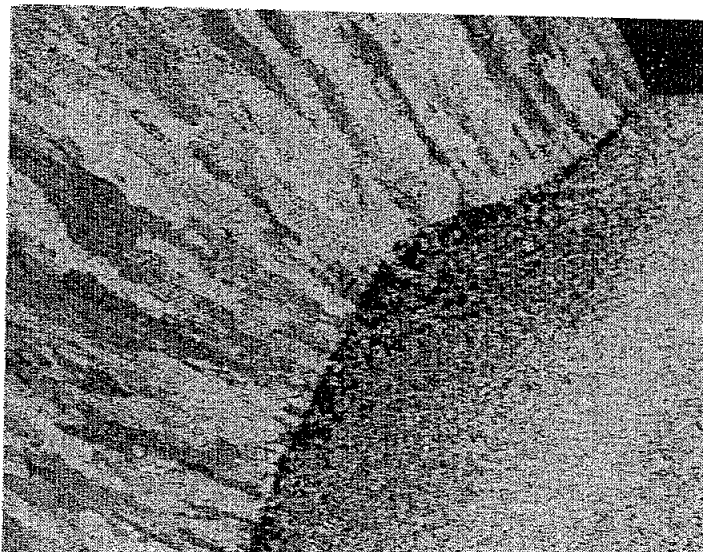


Figure 46. Section of submerged arc weld showing how the shape of the fusion boundary affects the grain size in the HAZ.

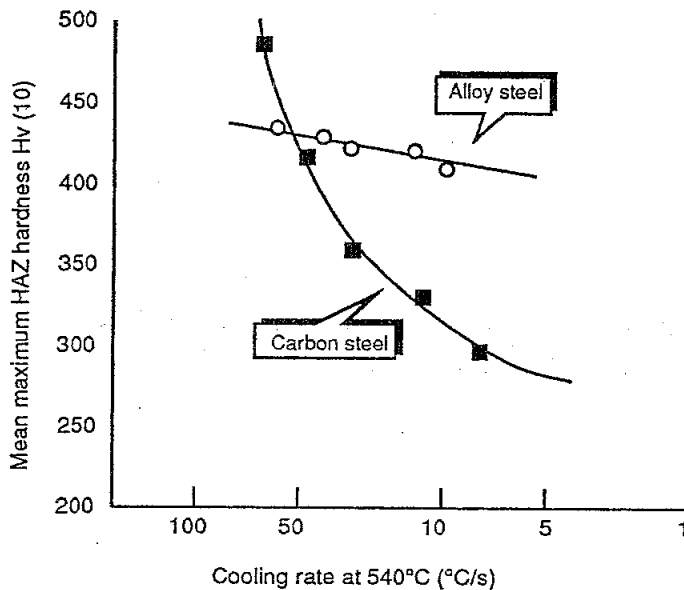


Figure 47. Plot of HAZ hardness versus cooling rate (hardening curves) for two steels. Note the HAZ hardness of the alloy steel does not change much with cooling rate. For the carbon steel there is a large change of hardness with cooling rate.

Width of coarse grain HAZ

Partial transformation

Hardening curves

The austenite grain size is affected by the composition of the steel, a coarse grain steel having a wide grain coarsened austenite region in the HAZ and a fine grain steel a narrow one. Except for the recent titanium steels the grain size immediately adjacent to the fusion boundary will be the same for all types of steel. In the titanium steels the presence of very stable titanium nitrides prevents grain coarsening, fine grains being retained right up to the fusion boundary.

Adjacent to the fully austenitic zone is a region where partial transformation to austenite can occur. Sometimes, particularly in small welds in higher carbon steels that cool rapidly, small patches of high carbon (white etching) martensite can be seen in this zone. These often have hardnesses higher than would be expected from the average composition of the steel. They occur in the region that just exceeds 723°C where partial transformation to austenite higher in carbon occurs and rapid cooling results in the formation of martensite. The martensite regions are localized and it is not certain whether the microstructure as a whole becomes any more susceptible to cracking.

On cooling, the austenite region of the HAZ transforms to a structure dependent on the cooling rate and the hardenability of the steel. Continuous cooling transformation diagrams are helpful in determining the structures and hardness that occur. In cases where only the hardness is of interest the standard hardening curve determined directly from weld samples of various heat inputs can be used. The hardening curve is helpful in establishing welding conditions that prevent HAZs forming with hardnesses higher than a given value. Two examples of hardening curves are shown in Fig. 47. One, for a carbon steel has a steep transition from high hardnesses associated

HAZ toughness

with martensite to a low hardness where the microstructure is pearlitic. The other example is for a quenched and tempered alloy steel. The hardenability is high enough for a hard HAZ to be formed at any cooling rate experienced in normal welding operations.

Studies of brittle fractures in welded structures show the site of initiation of the fracture is often the HAZ. The HAZ is usually at a point of stress concentration, is subjected to high residual stresses, and may contain defects. If, in addition, the HAZ toughness is low it becomes a prime candidate for starting a brittle fracture. HAZ toughness is quite difficult to measure because the HAZ is so small and the notches of fracture and impact specimens are hard to locate precisely. Nevertheless, testing for toughness in the HAZ is increasingly being specified and it is important to understand the factors which control it.

Martensite

Examining the hardening curve of a typical steel indicates that at fast cooling rates the HAZ transforms to a hard martensitic structure, the hardness depending on the carbon content. This microstructure, unless tempered by a stress relieving heat treatment, is very brittle—particularly when the carbon is high. For steels with greater than about 0.2% carbon, hardnesses exceeding 450 Hv are obtained in the HAZ of rapidly cooled welds, and the toughness of the HAZ will be unacceptably low for many applications.

Arc strikes

Very fast cooling rates are encountered in arc strikes, tack welds, and small 'cosmetic' weld beads and can induce brittle HAZs in steels with sufficient carbon. Many of the older structural steels had carbon levels over 0.2%, and brittle fractures occurring in these steels could often be traced to these small welds. As a result, welding standards such as CSA W59 and AWS D1.1 state that arc strikes are to be avoided, and if they occur on dynamically loaded structures they must be ground. Pipeline standards such as API 1104 require arc strikes to be repaired or removed.

To avoid the presence of hard, brittle martensitic structures in the HAZ, welding energy input and preheat are selected to ensure an adequately slow cooling rate. The hardening curves provide guidance on how slowly the weld must cool, and the critical cooling rate can be equated to minimum heat input or preheat using published cooling rate diagrams. In most of the newer structural steels (e.g., those used for pipelines and offshore oil and gas structures) the carbon content is sufficiently low that martensite formed in the HAZ at very fast cooling rates is not excessively hard and is, in fact, quite tough. In fact, with most modern HSLA steels it is the microstructures formed at *slow*

cooling rates that are sometimes of low toughness. The general trend, therefore, is for HAZ toughness to decrease with increasing heat input as we illustrate in Fig. 48.

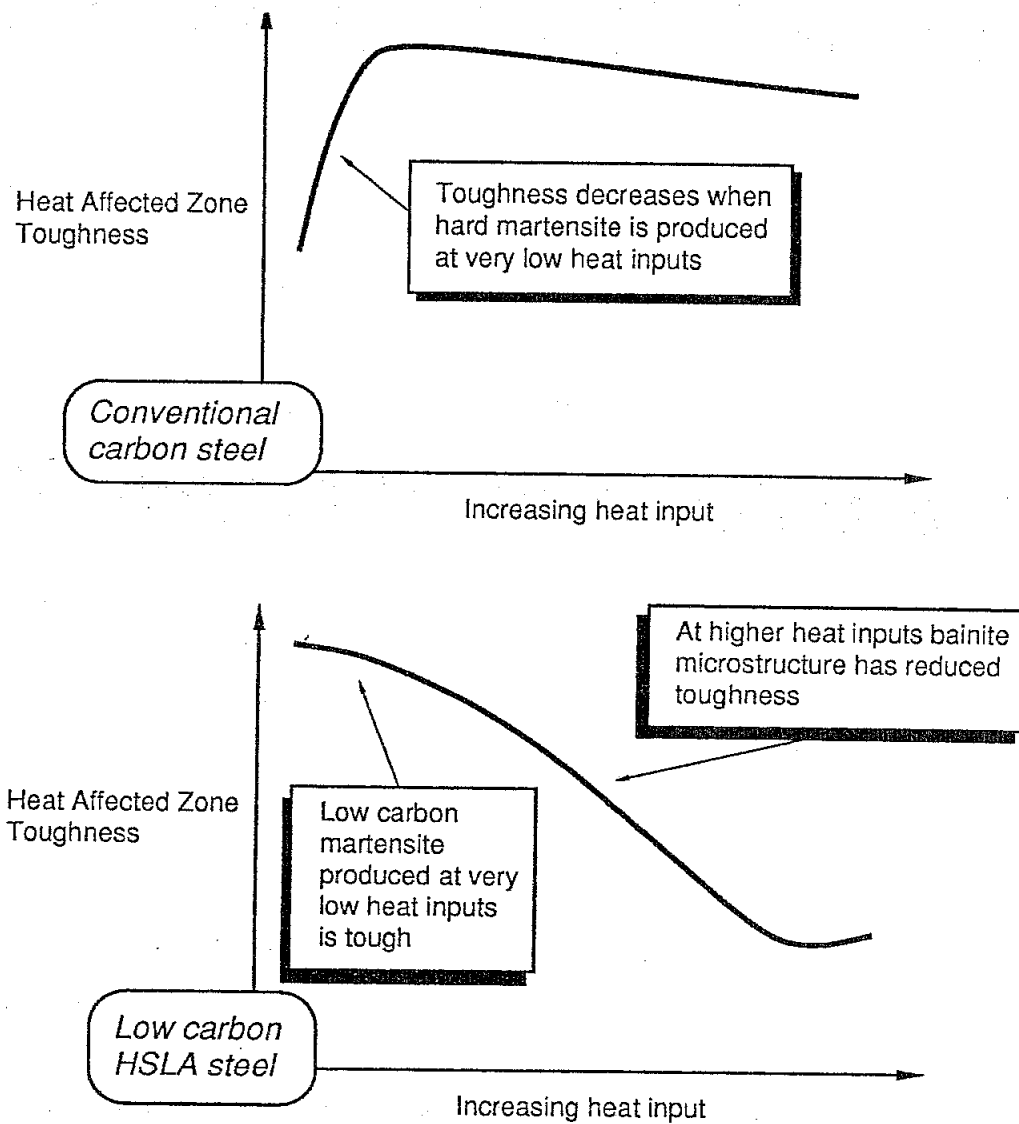
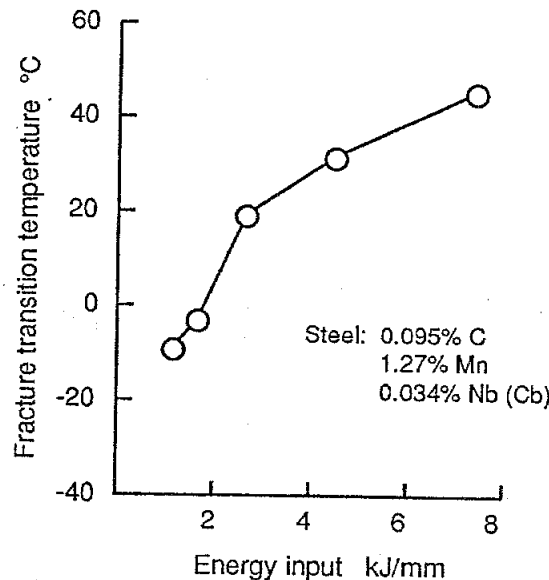


Figure 48. Diagram illustrating the general effects of heat input on HAZ toughness.

Figure 49. Effect of heat input on HAZ toughness of a niobium steel. Toughness measured by the instrumented charpy test.



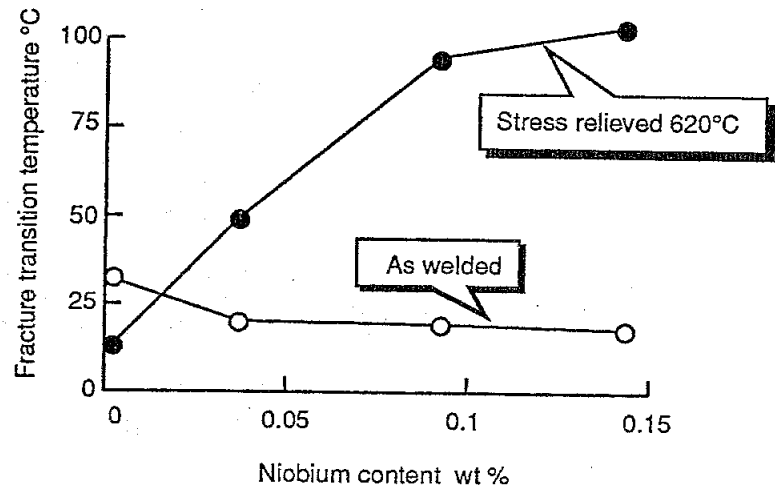
Effect of grain size

Research is not sufficiently advanced to provide rules for predicting the toughness of the HAZ, but certain factors are known to influence it. Grain size is one of the most important, and an increasing austenitic grain size in the HAZ is likely to result in deteriorating toughness. The grain size is determined largely by the heat input and it may be necessary for this to be restricted in order to achieve adequate toughness. In practical terms, restrictions on heat input may mean some welding processes, such as electroslag or tandem submerged arc, cannot be used; there may be a maximum interpass temperature; and the width of weave of a vertical SMAW weld may be limited. Some results of fracture toughness tests on an experimental steel are shown in Fig. 49 and illustrate dramatically the effect of heat input on HAZ toughness.

Precipitation hardening

Precipitation hardening from the presence of micro-alloy elements can also lower the HAZ toughness. Again, precipitation is encouraged by high heat input because of the longer times at high temperature and the slower cooling rates. A postweld heat treatment (stress relief) can cause a considerable amount of precipitation with a substantial decrease in toughness in the HAZ as Fig. 50 shows. Contrary to the plain carbon and carbon manganese steels a postweld heat treatment of some micro-alloyed steels can cause an increase in the HAZ hardness due to precipitation of fine carbides. Fig. 51 shows the effect of postweld heat treatment temperature on the HAZ hardness of a niobium-vanadium steel.

Figure 50. Effect of niobium content on HAZ toughness in the as-welded and stress relieved condition. Precipitation during stress relief results in deterioration of toughness at high niobium levels.



Other causes of HAZ embrittlement

The microstructure itself may have an effect on the toughness in the HAZ. Some researchers have observed low toughness associated with certain bainitic structures, but there is insufficient data to predict precisely when such structures might occur. A further source of embrittlement in the HAZ is strain aging. The contraction of a cooling weld may cause substantial plastic deformation, particularly around defects residing in the HAZ. In certain parts of the HAZ this may result in dynamic strain aging with a consequent loss of toughness. Because the phenomenon is associated with the presence of free nitrogen it is most likely in steels that are not killed, aluminum treated grades.

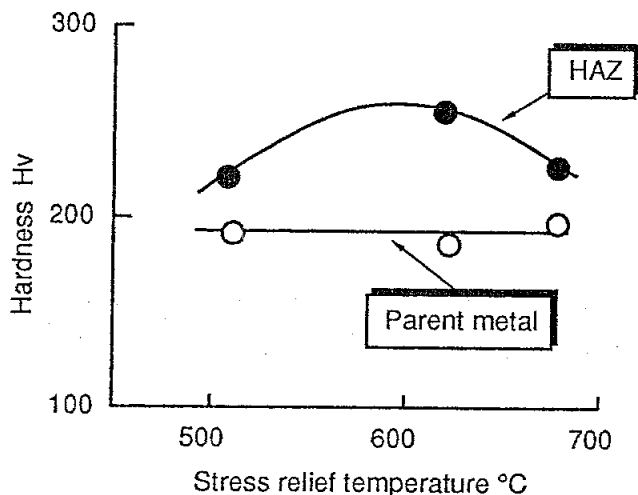
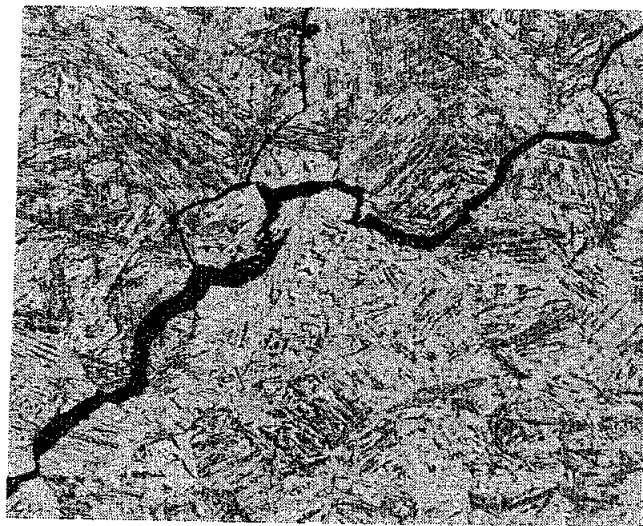


Figure 51. Effect of stress relief temperature on hardness in a micro-alloyed steel.

Figure 52. Example of a hydrogen crack in the HAZ of a weld.



Hydrogen cracking in the HAZ

Martensite formed in the HAZ is highly susceptible to hydrogen cracking like that shown in Fig. 52. In carbon or carbon manganese steels (i.e., those with steep hardening curves) welding conditions can be selected to avoid the cooling rates at which martensite is produced. But in low alloy steels or those where a hard HAZ cannot be avoided, other steps must be taken to prevent cracks. These often involve applying preheat to accelerate the diffusion of hydrogen out of the weld. Fig. 53 shows that quite moderate temperatures are highly effective in removing hydrogen.

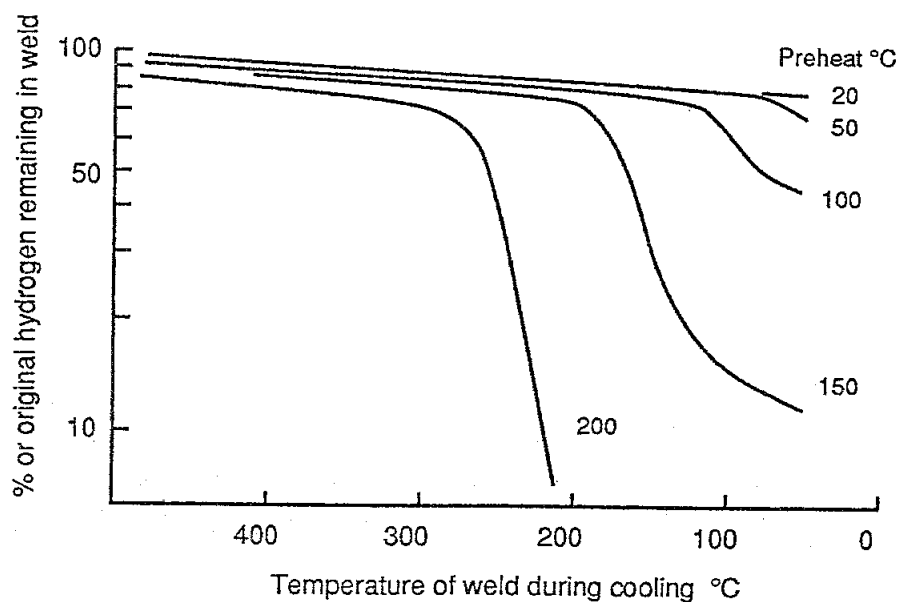


Figure 53. Diagram showing the fraction of hydrogen remaining in an example weld as it cools. When preheat is used very little hydrogen remains by the time the weld cools to room temperature.

Pcm formula

The preheat temperature required depends on the susceptibility of the HAZ to hydrogen cracking, and much work has been done to find compositional formulae to indicate this. One widely used formula due to Ito and Bessyo is

$$P_{cm} = C + \frac{Si}{30} + \frac{Mn}{20} + \frac{Cu}{20} + \frac{Ni}{60} + \frac{Cr}{20} + \frac{Mo}{15} + \frac{V}{10} + 5B$$

You should note that this relates directly to cracking susceptibility in steels of less than about 0.2% carbon and should be distinguished from the hardenability carbon equivalent given earlier.

The level of preheat also depends on the original hydrogen level in the weld metal. These two factors—the plate composition and the hydrogen—can be combined in an index of susceptibility equal to

$$12P_{cm} + \log H$$

where H is the hydrogen determined by the standard IIW method. This index is used in CSA W59-Appendix P and AWS D1.1-Appendix XI as part of an alternate guide for estimating preheat levels.

Stress relief cracking

During thermal stress relief the residual elastic strains are converted to plastic strains by a drop in the material yield point and creep. If these strains are not spread uniformly through the metal but are concentrated in local regions, they may be sufficient to cause cracking. This phenomenon, known as stress relief or re-heat cracking, has been observed in the heat affected zones of welds in particular steels. Cracking is more likely in highly restrained, very heavy sections and where there is a stress concentration such as an unfused root or the toe of a large fillet weld.

Precipitation of carbides of secondary hardening elements during stress relief is the primary cause of cracking. Precipitation occurs preferentially within the grain causing the grain interior to strengthen relative to the grain boundary. This shifts the strain onto the grain boundary where cracks may form.

Studies of stress relief cracking reveal that composition has a dominant effect, and empirical relations have been established to predict whether a steel is sensitive to this type of defect. One formula suggested in the research literature is:

$$P_{sr} = Cr + Cu + 2Mo + 10V + 7Nb + Ti - 2$$

When $P_{sr} \geq 0$ stress relief cracks may occur. This formula does not work for low carbon (<0.1%) or high chromium (>1.5%), these steels being resistant to stress relief cracking.

Structure

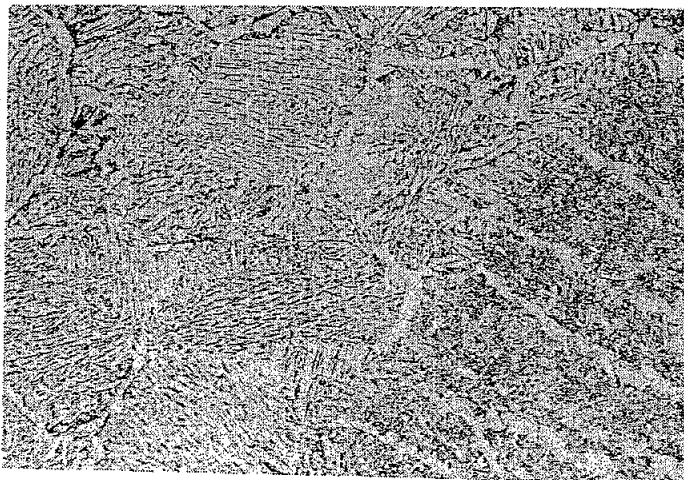
WELD METAL

Since weld metal starts off as a small casting, cools rapidly and does not undergo heavy deformation it is not surprising that the properties of weld metal can be quite different from those of the parent plate and heat affected zone. Consequently the optimum compositions of weld metals are rarely exactly the same as the steels they are joining. For example, a thick plate of pressure vessel steel of 0.25% carbon may be joined with a weld metal containing only 0.1% carbon. On the other hand, when alloy steels are selected for specific properties such as corrosion resistance or high temperature strength, the weld metal will likely have a similar composition to the base metal.

Columnar grains

The microstructure of weld metal is influenced first by the solidification structure that exists after the metal freezes. Usually long austenite grains extend from the fusion boundary towards the centre of the weld (Fig. 54). They are epitaxial with the grains in the HAZ meaning they grow directly from them. This is an important observation since the grain size of the weld metal can be influenced by the grain size of the HAZ. In most steels the HAZ grain size at the fusion boundary depends only on the heat input and not on the steel composition so the weld metal grain size is also independent of the steel composition. As we saw earlier, however, some modern titanium steels have very small HAZ grain size, and this is reflected in a smaller weld metal grain size.

Figure 54. Weld metal structure showing austenite grains extending from the fusion boundary where they are epitaxial with the HAZ grains.



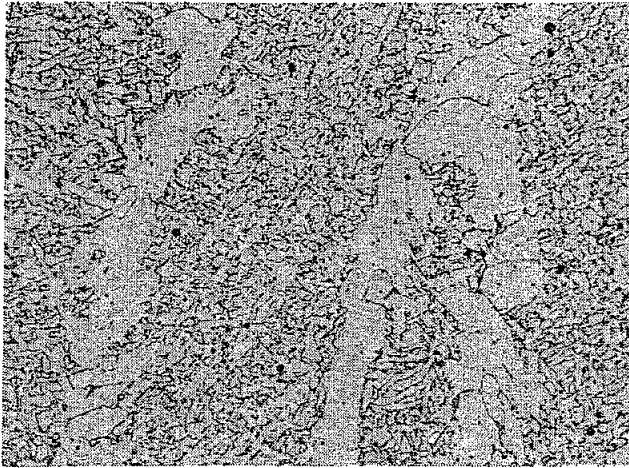
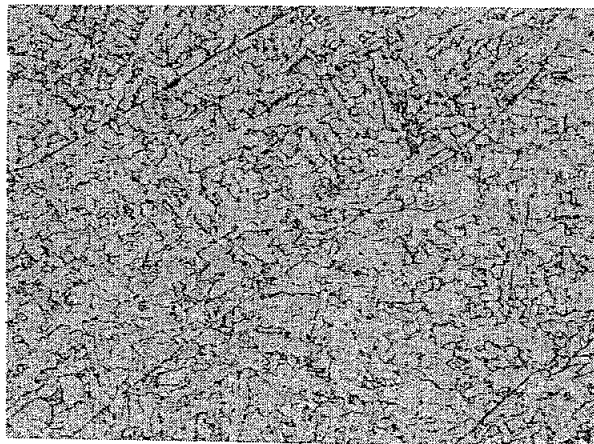


Figure 55. Proeutectoid ferrite 'veins' formed at the austenite grain boundaries in a C-Mn-Si weld metal.

Ferrite morphology

In a typical carbon manganese weld metal, austenite grains transform to a mixture of ferrite and carbide. Ferrite may form first at the grain boundaries of the austenite and give rise to the familiar 'veins' commonly observed in weld metals and visible in Fig. 55. Ferrite may also form within the grain if nucleation sites are available. Small oxide particles provide excellent sites for nucleation and a weld metal having more than about 250 ppm of oxygen present has enough oxide particles to nucleate many tiny grains of ferrite within the austenite grain. This intragranular ferrite, shown in Fig. 56, is termed *acicular* ferrite and is associated with weld metals having good toughness. Ferrite may also be present in the form of spikes or laths growing into the grain from the edge. This form, sometimes termed side plate ferrite, has been associated with poor toughness. After ferrite is formed the remaining austenite becomes richer in carbon and can transform to a number of different structures. The carbon may therefore be present as a carbide, in small islands of martensite or in retained austenite.

Figure 56. Acicular ferrite formed within the austenite grain of weld metal. The very small grain size of this intragranular ferrite contributes to good toughness.



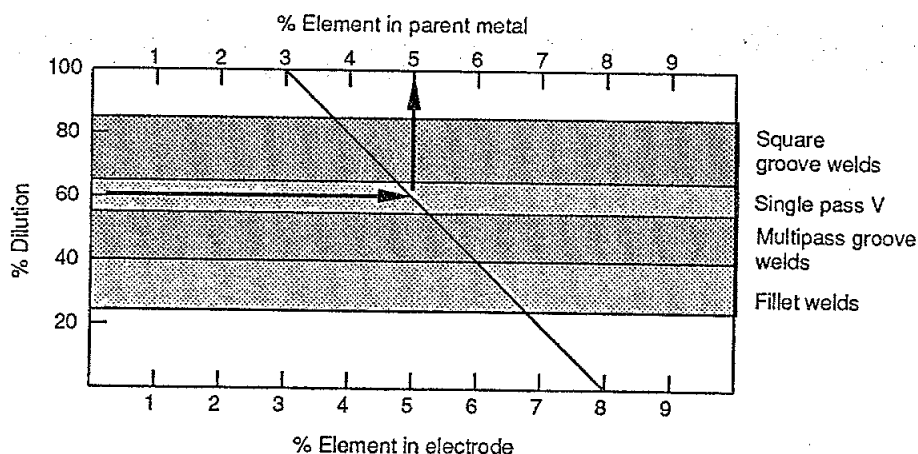
Weld metal composition

The final composition of the weld metal is determined by three factors:

- The filler metal composition
- Dilution from the parent metal
- Chemical reactions

Notice that the filler metal composition includes elements coming from the flux and other ingredients present on the electrode coating or in a cored wire. It will not be easy to determine the composition if fluxes are present, and for this reason standards may specify composition in terms of a deposit chemistry. This is not necessarily the same as the electrode, but this approach is useful for specification purposes if the method of making the deposit is standardized. For example, CSA W48.5-M specifies flux cored electrodes in terms of the deposit chemistry. In the case of solid wires for gas shielded or submerged arc welding the composition can be specified in terms of the actual wire chemistry. In this case it is important to recognize that the weld deposit may not be the same.

Figure 57. Diagram to estimate weld metal composition from dilution. The example shows that an electrode of (say) 8% Ni on a plate of 3% Ni with a 60% dilution weld gives a weld metal of 5% Ni.



Dilution diagram

The effects of dilution can be predicted quite accurately for some elements by taking the proportion of the element contributed by the parent metal and adding it to that contributed by the electrode and assuming complete mixing. This only works for those elements that do not enter into chemical reactions, such as nickel and molybdenum, but does not work for active elements like carbon or manganese. The simple diagram shown in Fig. 57 assists in determining the final weld composition. The major difficulty in using this method is that the percentage dilution is not known with any degree of certainty, but it is a useful method for setting ranges.

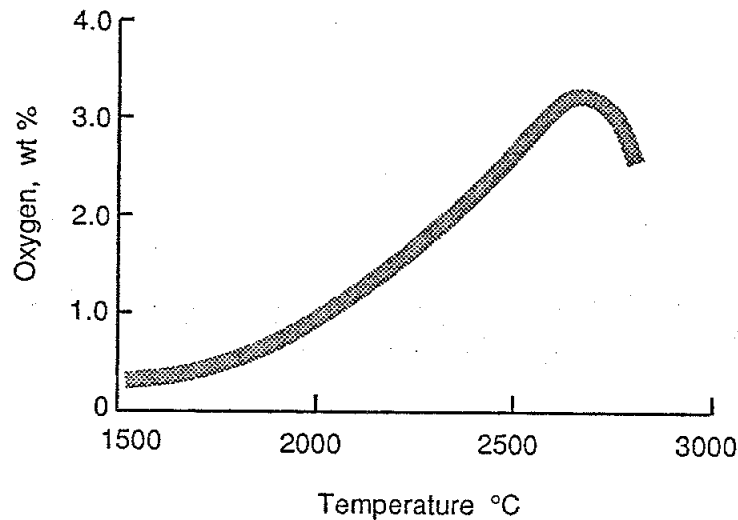


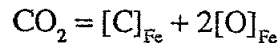
Figure 58. Solubility of oxygen in liquid steel in equilibrium with pure iron oxide.

Chemical reactions

Chemical reactions that take place between the weld metal and the slag or the surrounding gas can result in significant changes to the weld metal composition. We consider first the reactions between shielding gases and the weld metal in gas metal arc welding.

The main purpose of the shielding gas is to exclude oxygen and nitrogen from the weld pool but this is not always achieved. Some welding grade gases may deliberately contain oxygen or active gases and these will react with the weld metal. If iron is in contact with pure oxygen gas the oxygen continues to dissolve up to the solubility limit then form FeO as a slag on the surface of the liquid iron. At 1550°C the solubility is about 0.2% [O] but at higher temperatures it increases to more than 2% (Fig. 58).

If we surround a pool of liquid iron by a carbon dioxide CO₂ atmosphere, the CO₂ reacts to produce carbon and oxygen which both dissolve in the liquid iron:

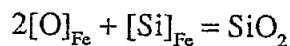


Since there is a steady supply of CO₂ gas the reaction above tends to result in equilibrium concentrations of carbon and oxygen in the iron regardless of the original levels of carbon and oxygen present. In the case of a weld in which the weld pool is moving, fresh metal is being melted ahead of the arc while solidification takes place behind and this influences the weld metal composition. But the constant supply of CO₂ shielding gas to the weld pool results in a similar reaction in which the oxygen and carbon content of the pool tend to an equilibrium value.

At the back of the pool where solidification takes place the falling temperature causes oxygen and carbon to react together producing carbon monoxide:



The carbon monoxide tries to escape but, getting trapped, forms porosity. Porosity can be prevented by lowering the oxygen content in the weld pool before solidification takes place. This is achieved by adding deoxidants, such as silicon, to the electrode wire. Silicon reacts with oxygen to form a slag of silica which floats out on the surface of the weld metal:



The effects of adding silicon to the pool are:

- It lowers the oxygen content
- It raises the carbon level at which carbon monoxide could form.
- It raises the carbon level that is in equilibrium with CO_2 gas.

A typical CO_2 weld where silicon is added through the wire results in a weld metal carbon content of about 0.1% C regardless of the original wire or plate carbon level. In practice wires can be purchased having various levels of deoxidants to accommodate the range of oxygen that may be met. Welding rimming steel or rusty plate requires a higher level of deoxidation than welding a clean, fully killed steel. The effectiveness of silicon as a deoxidant is increased by the addition of manganese. Further deoxidation is obtained by adding titanium, zirconium or aluminum (triple deoxidized wires). The deoxidizing power of a solid wire electrode depends on the combination of silicon, manganese and other deoxidants. The list in Table 3 shows a number of commercially available wires in ascending order of deoxidizing power and a guide to their selection.

Fluxes are used in several welding processes such as submerged arc welding and flux cored arc welding. The prime function of the flux in submerged arc welding is to form a slag that protects the molten weld metal from the air as Fig. 59 shows, but there are secondary functions such as arc stabilization and weld pool support. Naturally the liquid slag formed should be immiscible with the molten steel so that two layers form. The melting point of the slag and its density should be lower than those of the steel so that the slag floats out. Since

Deoxidation

GMAW electrodes

Fluxes

Table 3. GMAW electrodes and their relative deoxidizing power.

Classification		Typical electrode analysis				Increasing deoxidizing power
CSA W48.4-M	AWS A5.18-79	C	Mn	Si	Other	
ER480S-3	ER70S-3	0.10	1.15	0.60		
ER480S-4	ER70S-4	0.10	1.25	0.75		
ER480S-6	ER70S-6	0.10	1.60	0.95		
Low alloy high toughness (A5.28-79 ER80S-D2)		0.11	1.90	0.60	Mo 0.5 Ni 0.15 Cr 0.08 Ti&Zr 0.017	
Low alloy high strength (A5.28-79 ER80S-D2)		0.10	1.85	0.65	Mo 0.50	
ER480S-2		0.05	1.15	0.50	Al 0.05 Ti 0.07 Zr 0.10	

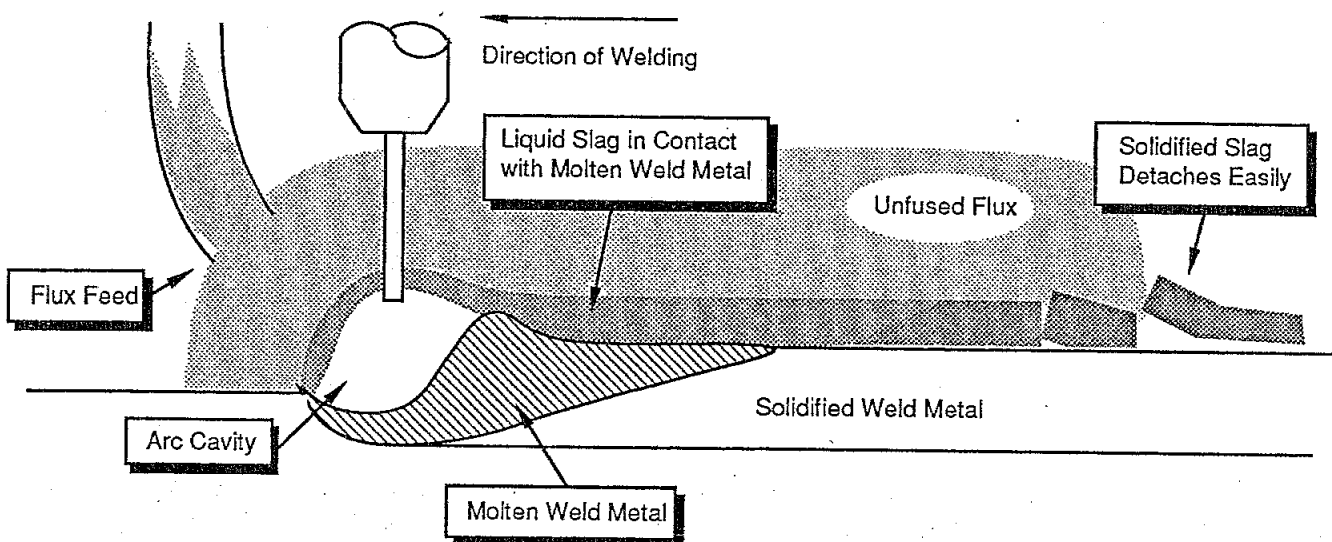
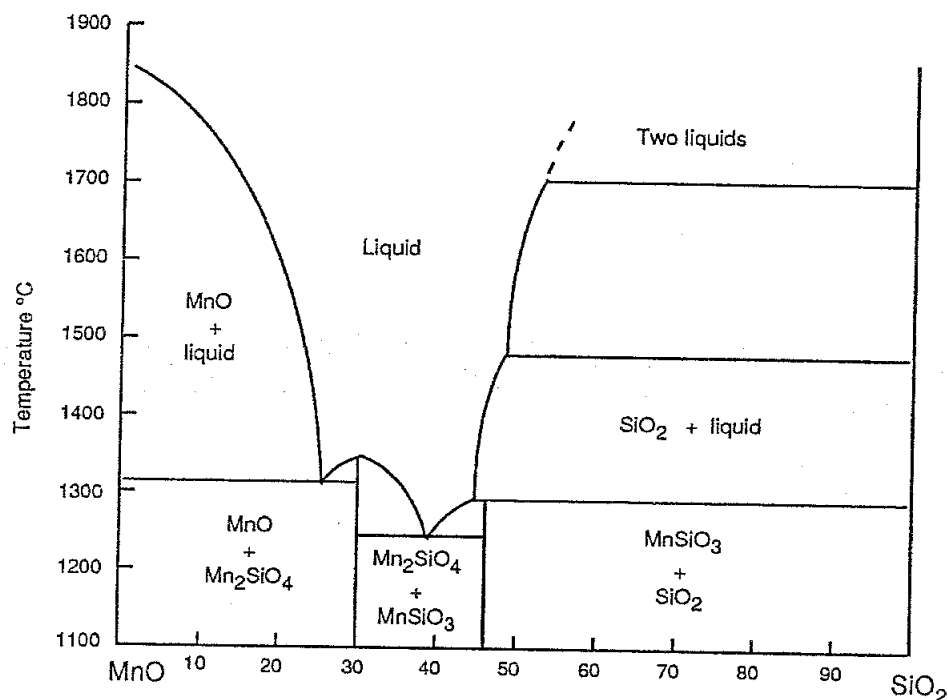


Figure 59. The main features of the submerged arc process. Weld metal composition is affected markedly by chemical reactions between the weld pool and the molten slag.

Figure 60. MnO-SiO_2 phase diagram showing how certain compositions may produce a flux with a melting temperature suitable for welding steel.



Submerged arc fluxes

some mechanical support to the weld metal is also desirable, the melting temperature of the slag should not be too far below that of the steel. A solidification range, as opposed to a sharp freezing point, is preferable since a suitable viscosity will assist in giving a good bead shape. In addition, the thermal contraction of the slag relative to the steel must allow the slag to detach easily.

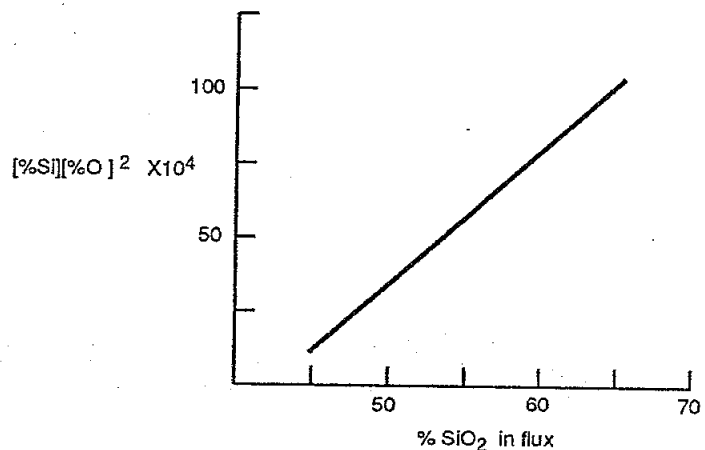
Fluxes of metal oxides melting in the range 1200°C - 1400°C are satisfactory and are chemically quite stable. There are, however, some important chemical reactions that can occur between the slag and the weld metal. Because most single metal oxides melt well above the melting temperature of steel, two or more oxides are used in combination, often as naturally occurring minerals. Fig. 60 shows the phase diagram for the MnO-SiO_2 system which has been used as the basis of many commercial fluxes, and it is clear that appropriate melting temperatures can be found by carefully selecting the composition.

Slag-metal reactions

Although silica (SiO_2) has many desirable properties for a welding flux it does become chemically active at high temperatures. Consider for a moment a pool of liquid iron on which floats a slag of pure silica. Oxygen and silicon (Si) in the metal will be in equilibrium with the silica (SiO_2) such that the concentrations satisfy the equilibrium equation:

$$K \rightleftharpoons \frac{[\text{SiO}_2]}{[\text{Si}][\text{O}]^2}$$

Figure 61. Experimental data showing the relation between the flux silica content and the oxygen and silicon levels in the weld metal.



Oxygen and silicon in weld metal

where K is the equilibrium constant. $[]$ represents the activity of the component. For pure silica the activity is one so the product $[Si][O]^2$ is a constant at any given temperature. If the silicon content of the iron is raised by adding silicon, the oxygen level must go down to restore equilibrium. If no silicon or oxygen were initially present in the iron, the equilibrium levels would be the stoichiometric values where one atom of silicon from the silica enters the liquid metal for every two oxygen atoms.

If instead of being pure the silica is mixed with other components, the activity of the silica is reduced, depending on the concentration. In the case of submerged arc welding, equilibrium conditions cannot be assumed, but it has been shown experimentally (Fig. 61) that the above equation is approximately true in the sense that the term $[Si][O]^2$ is proportional to the SiO_2 content of the flux. The data in Fig. 61 may be redrawn in terms of weld metal oxygen and silicon levels for various proportions of silica in the flux. Fig. 62 shows this.

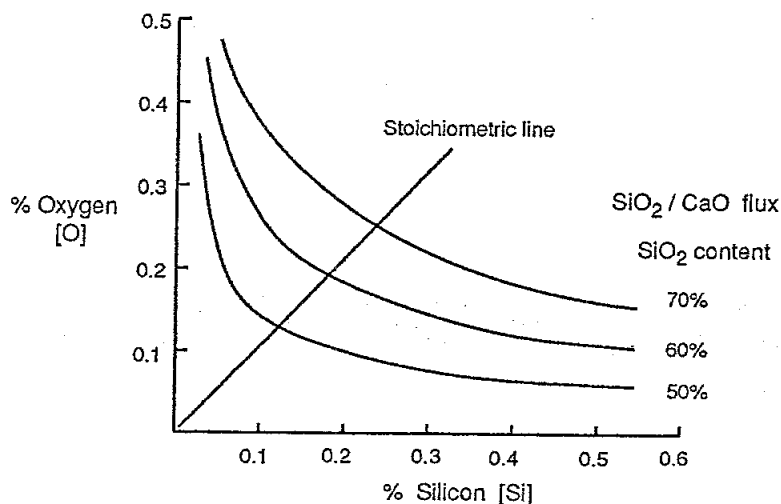


Figure 62. Weld metal oxygen and silicon levels for various silica contents in the flux. Drawn for the data shown in Figure 61.

The implications of this are important. If the flux contains a high percentage of silica then the value of $[\text{Si}][\text{O}]^2$ is high. In effect, both silicon and oxygen are transferred to the weld metal from the flux. A lower oxygen could be achieved by adding more silicon $[\text{Si}]$ from the wire giving a very high silicon level but keeping the value of $[\text{Si}][\text{O}]^2$ the same, but to achieve both lower silicon and lower oxygen the silica content of the flux must be reduced. Silicon in small quantities may have an apparent beneficial effect on weld metal toughness because it lowers the oxygen content. But as you can see from Fig. 62 high levels of silicon have a diminishing effect on oxygen, and high levels of silicon increase the yield strength and lower the toughness.

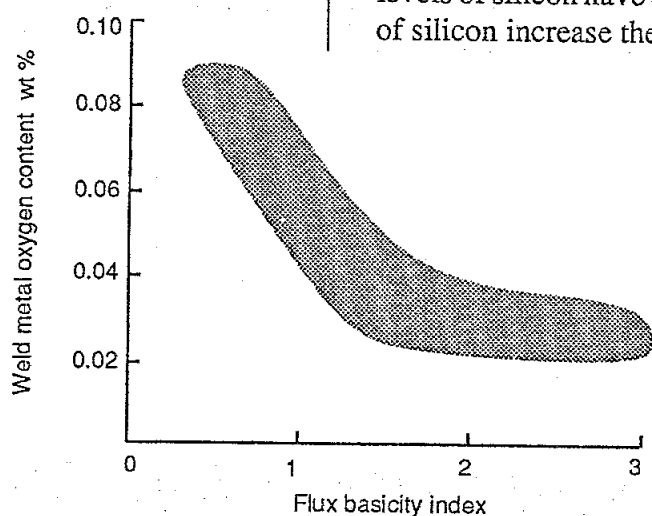


Figure 63. Effect of flux basicity on the oxygen content of weld metal.

Oxide Inclusions

Oxygen does not remain in solution when the metal solidifies but forms small, round inclusions. They have a marked effect on the energy to cause ductile fracture in the steel, a measure of which is provided by the upper shelf of the charpy transition curve. Thus the oxygen content of the weld metal directly influences the charpy impact properties. This becomes an important factor in selecting a flux when impact properties are a consideration. High silica fluxes have many desirable properties, such as high current capacity and excellent bead shape control, but their use should be restricted to those applications where high toughness is not a prime consideration. Some commercial high silica fluxes are only recommended for single pass welds.

Flux basicity

The concept of oxygen transfer from fluxes can be extended to other flux components and has given rise to the basicity index as a measure of potential oxygen transfer. Although not originally based on oxygen reactions, experiments have shown that the basicity index relates quite well to the weld metal oxygen as Fig. 63 demonstrates. The basicity index is given by:

$$\text{BI} = \frac{\text{CaO} + \text{MgO} + \text{Na}_2\text{O} + \text{K}_2\text{O} + \text{Li}_2\text{O} + \text{CaF}_2 + 1/2(\text{MnO} + \text{FeO})}{\text{SiO}_2 + 1/2(\text{Al}_2\text{O}_3 + \text{TiO}_2 + \text{ZrO}_2)}$$

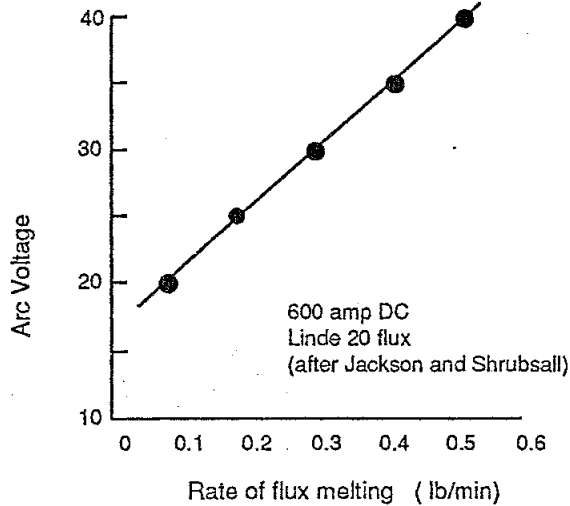


Figure 64. The amount of flux melted is directly proportional to the arc voltage.

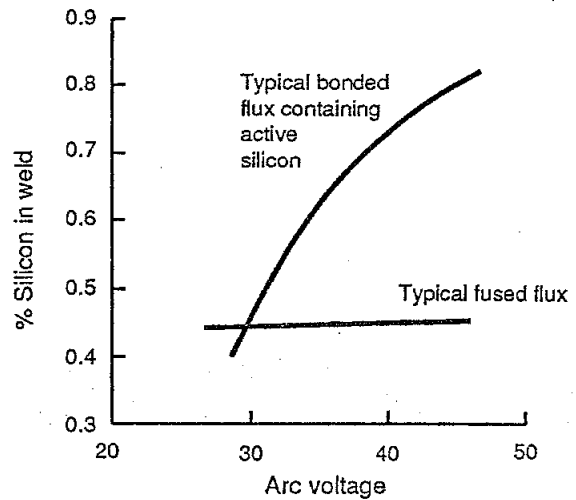


Figure 65. The amount of silicon transferred from active fluxes depends on the arc voltage.

Active fluxes

Commercial basic fluxes with high basicity are often specified when excellent toughness at low temperatures is required.

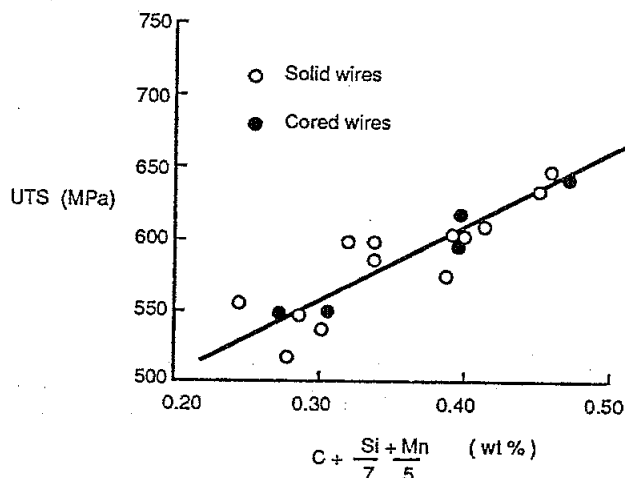
What we have discussed up to now has been concerned with chemical reactions with the oxide components of the flux. Not all the flux components, however, are oxides. In some fluxes made by the bonding or agglomeration process 'active' deoxidants may be present. Silicon in the form of ferro-silicon may be added to improve the deoxidizing capacity of the flux allowing it to handle rusty plate. The amount of silicon transferred to the weld pool this way depends on the quantity of flux melted which in turn depends on the arc length and voltage. A high voltage causes a long arc within the flux and results in more flux being melted. Fig. 64 shows typical results illustrating the effect of voltage on flux consumption. Since voltage has little effect on the electrode melting rate, an increase in voltage increases the concentration of any active element transferred from the flux.

Effect of voltage

T.R. FOR LEVEL III

Fig. 65 shows the weld metal silicon contents for two types of flux as a function of voltage. Voltage sensitivity is important when making multipass welds since it is quite common practice to increase the voltage on the final layers to spread the bead and get a good weld profile. With voltage sensitive fluxes this could result in a silicon (or manganese or other elements) build-up leading to undesirable properties. Other alloy elements such as chromium are present in some alloy fluxes which are used for welding weathering steels, stainless steels and other alloy steels. If fluxes of this type are employed, very careful attention to voltage is required to ensure proper control of weld metal composition.

Figure 66. Relation between the tensile strength and the weld metal composition for CO₂ welds.



Properties of welds

In most cases the strength of weld metal is determined by the composition and whether the weld is subject to postweld heat treatment. Welding heat input may also have an effect since this controls the cooling rate of the weld. Although accurate prediction of weld metal strength from composition is not possible, some general trends have been established for carbon-manganese-silicon weld metals. In most cases the ultimate tensile strength (UTS) is of greatest importance since weld metal yield strength is usually substantially higher than that of the plate when the ultimate strength matches.

Tensile strength

Fig. 66 shows a relation between composition and the UTS of GMA welds made under CO₂ shielding. Wires of different deoxidizing capability produce a range of manganese and silicon levels in the weld metal. High levels of these elements lead to weld metal of high strength and hardness which may have unacceptably low ductility. Fig. 67 shows a similar relation for submerged arc welds, and the wide

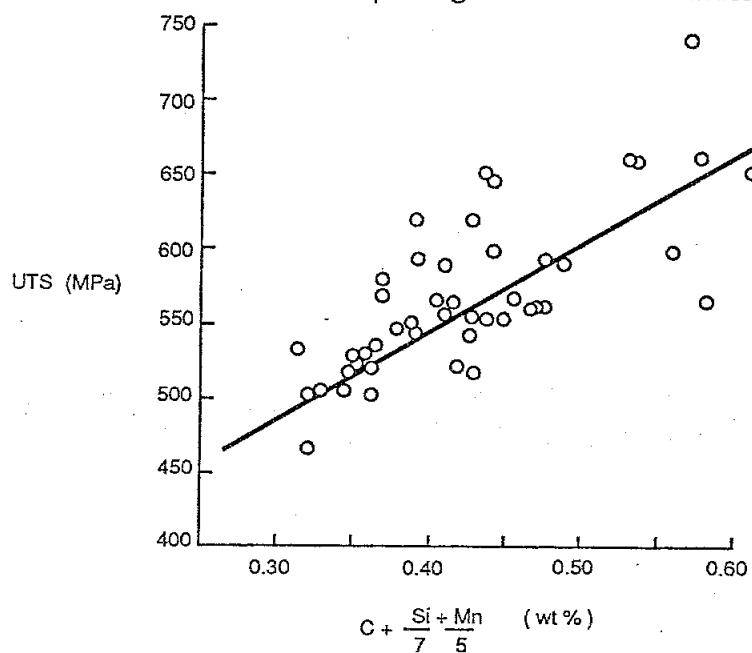


Figure 67. Relation between the tensile strength and the weld metal composition for submerged arc welds.

Toughness

Ductile fracture

Effect of microstructure

scatter of results indicates that several factors other than composition affect the strength. Toughness cannot be simply related to composition and a considerable amount of research has been done to determine the factors which control toughness. An understanding of these is helpful when faced with the problem of selecting welding consumables that must meet certain toughness requirements.

Ductile fracture in steel results from the growth and subsequent joining of microvoids which form around inclusions within the metal. We have seen how the presence of inclusions from deoxidation influence ductile fracture as represented by the charpy energy. Reduction in the number of inclusions (particularly large inclusions) by lowering the weld metal oxygen can lead to substantial improvements in toughness. Low oxygen is achieved through the use of basic fluxes or processes with low oxygen potential such as inert gas shielded welding. In the case of flux cored arc welding the effect of the flux can override that of the gas so that using an inert gas alone is not sufficient to ensure high toughness. The progress made in developing flux cored wires with very high toughness was recognized in the CSA standard W48.5-M1982 (August 1984 revision) which provided for classification of wires with impacts tested down to -40°C .

Microstructure has a pronounced effect on toughness although it is not always possible to predict the microstructure from the composition. CCT or structure diagrams for weld metal are useful in helping determine the microstructure to be expected. Two examples of weld metal structure diagrams are shown in Fig. 68 and Fig. 69. They show the proportion of each microstructural constituent expected at various

Figure 68. Weld metal structure diagram showing the percentage of each constituent present. This weld metal exhibited a low percentage of acicular or intragranular ferrite.

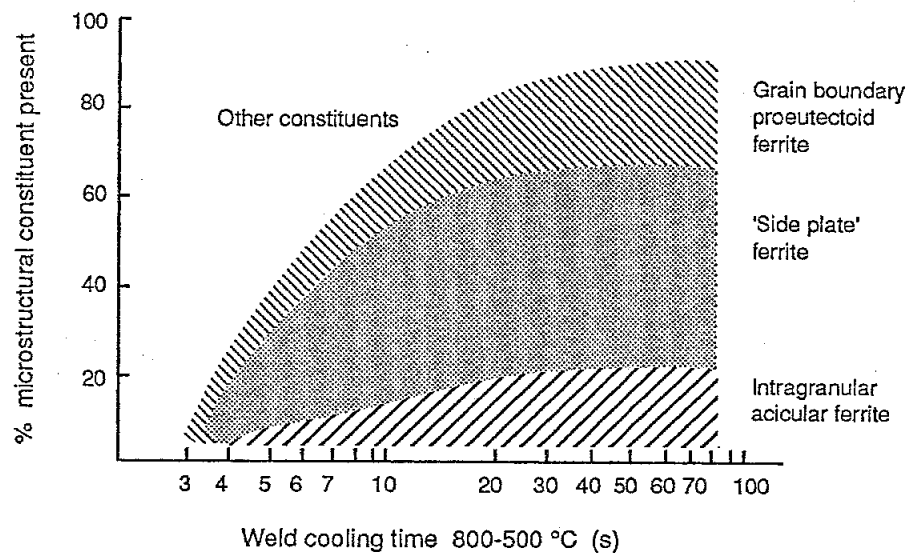
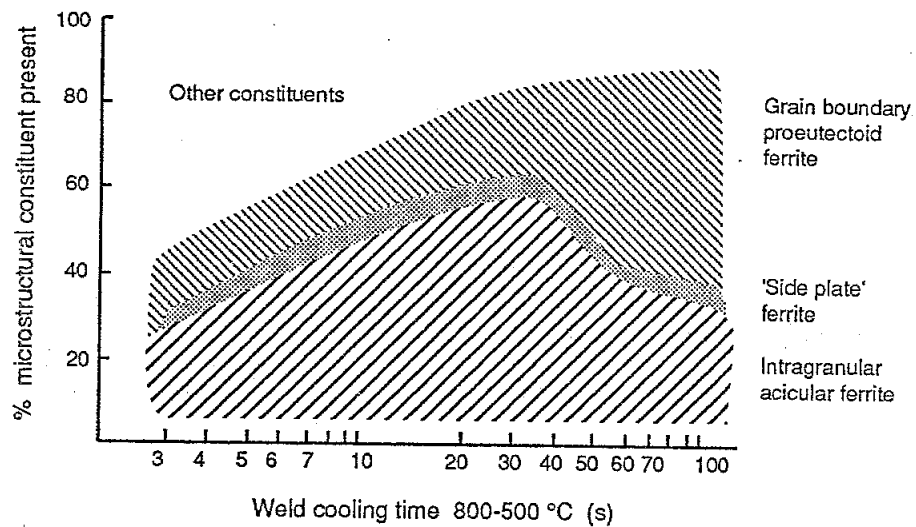


Figure 69. Weld metal structure diagram. This weld metal had a medium oxygen level and showed a high proportion of intragranular acicular ferrite.



cooling rates. Research has shown that structures containing a large percentage of intragranular, acicular ferrite are associated with good toughness.

As with the heat affected zone the grain size appears to play a prominent role in determining weld metal toughness, and low heat input and fast cooling rates generally give better toughness. In multipass welds the heat from subsequent passes refines the grain structure and may improve the properties. A welding procedure giving a high proportion of refined structure can result in better toughness even though the heat input per pass may be higher. Lower impact values are usually recorded in the root region of groove welds in thick sections, the factors believed to contribute to the loss of toughness being dynamic strain aging, high dilution and rapid cooling rates. Many specifications call for impact specimens to be tested from this region (Fig. 70).

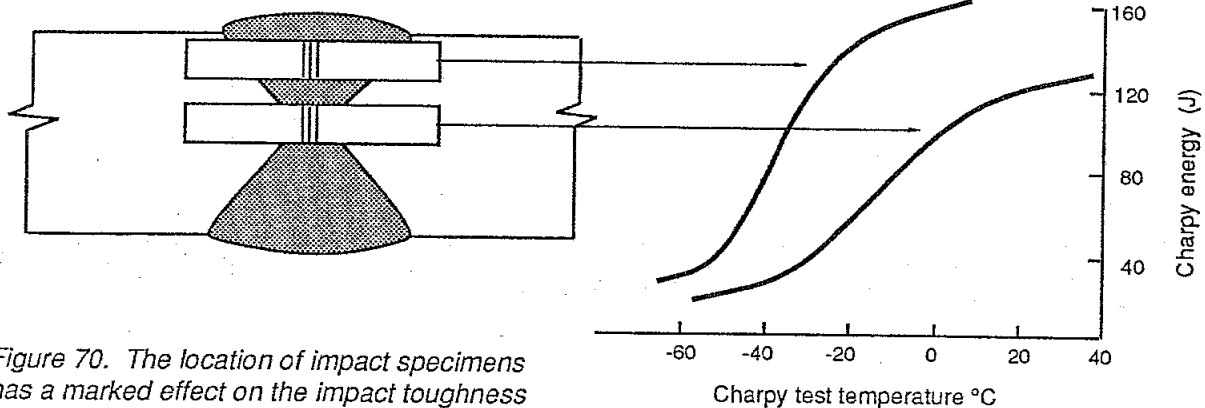


Figure 70. The location of impact specimens has a marked effect on the impact toughness measured. The root region usually shows the lowest.

Effect of postweld heat treatment

Both strength and toughness can be affected by postweld heat treatment. There is usually a reduction in strength but toughness may either increase or decrease. The extent of strength loss depends on both the temperature and the time of postweld heat treatment and it is possible to combine these two factors into the single Holloman-Jaffe parameter introduced earlier:

$$H = T(20 + \log t)$$

This applies to a steady temperature. If the temperature is changing, as for instance during the heating and cooling stage of a stress relief treatment, then a correction factor can be added to the time. The correction is

$$t_e = T / 2.3k[20 - \log k]$$

where k is the heating or cooling rate in $^{\circ}\text{K}$ per hour.

In Fig. 71 the UTS of typical carbon-manganese and molybdenum weld metals is plotted against the parameter H and over the practical range of interest shows a linear relation. Knowing the initial strength of a weld this graph could be used, for example, to determine the maximum stress relief time at a given temperature before the weld metal strength drops below a specified value.

Figure 71. Drop in the tensile strength of weld metal resulting from stress relief. The drop depends on both time and temperature and the weld metal composition.

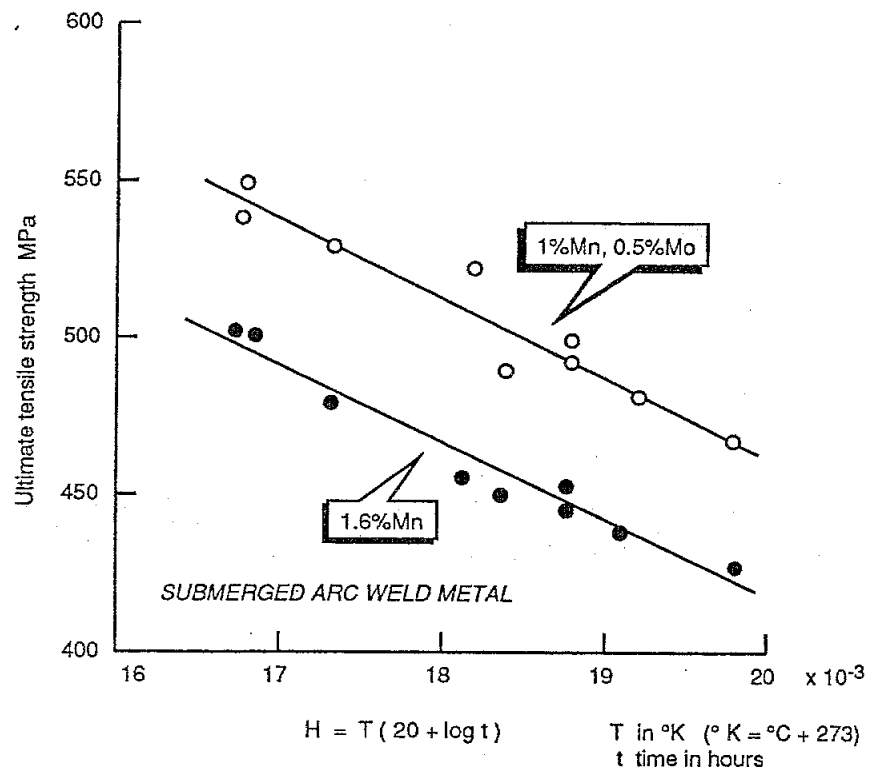
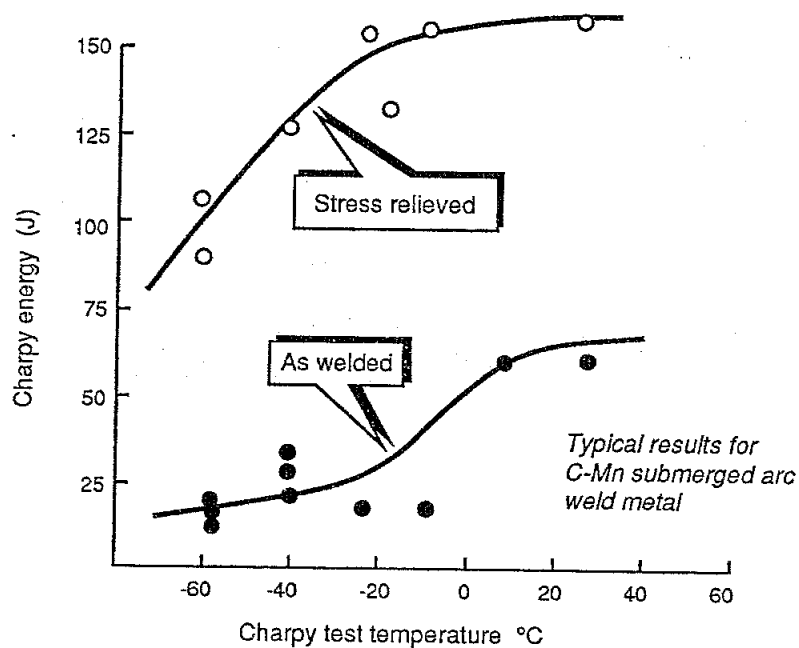


Figure 72. A stress relief (PWHT) generally improves the toughness of C-Mn weld metal as shown in these typical results.



Effect on toughness

Unfortunately it is not possible to predict the change in toughness of a weld metal subject to a PWHT stress relief. In many cases the toughness can improve by a substantial amount as shown in the example in Fig. 72. Conversely, deterioration of toughness could occur, for example, due to precipitation hardening. As with the heat affected zone the presence of certain micro-alloy elements in sufficient quantities, such as niobium or vanadium, causes precipitation during stress relief resulting in increased hardness and a loss of toughness. Fig. 73 illustrates this effect. CSA W59-89, for example, places a limit of 0.05% on weld metal vanadium content when welds are to be stress relieved.

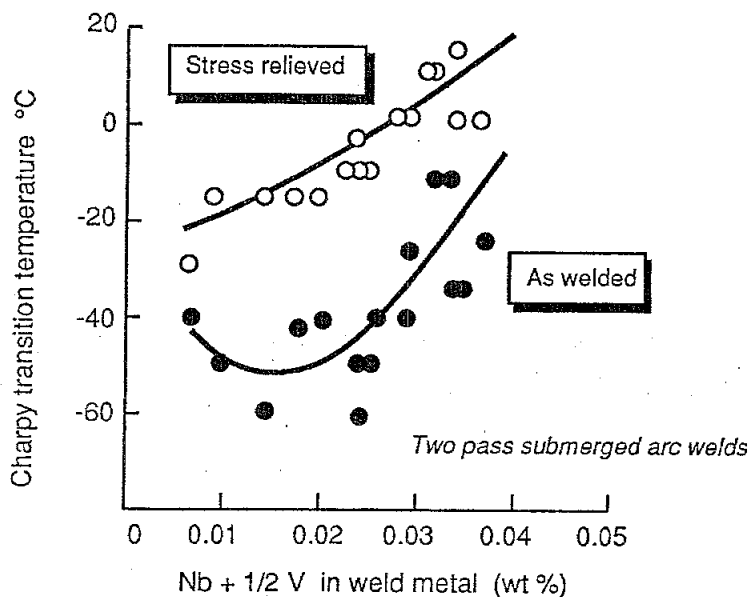


Figure 73. Stress relief of weld metal containing significant levels of vanadium or niobium (columbium) usually causes a deterioration of toughness as shown here by the increase in toughness transition temperature.

Weld metal hydrogen cracking

Hydrogen cracking can occur in either the heat affected zone or the weld metal and be either transverse or longitudinal to the weld axis. The level of preheat or other precautions necessary to avoid cracking will depend on which region, weld metal or HAZ, is the more sensitive. In the older carbon-manganese medium strength structural steels the HAZ was usually the more critical region and weld metal rarely caused a problem. Modern steels with low carbon are now very resistant to hydrogen cracking and so the weld metal may be the controlling factor.

Hydrogen cracking in weld metal depends on the same fundamental factors as in the HAZ i.e., hydrogen content, microstructure and stress. The controlling variables in practice are usually: strength level, hydrogen content, restraint and stress concentrations. In single pass welds and the root run of multipass welds the root gap provides a stress concentration to the transverse stress and this leads to longitudinal cracks in the weld metal.

The higher dilution of the root run often results in a harder weld bead more likely to crack. This is the predominant form of cracking in such applications as pipelines. In cases where transverse notches exist, such as an unspliced backing bar, the longitudinal stress becomes the controlling factor and transverse cracks may occur. In heavy multiple-pass welds cracking will generally be in the transverse direction either normal to the surface or at an angle of about 45 degrees. The latter type has been called 'chevron cracking'. The solution to these cracking problems rests with proper control of hydrogen through handling of consumables, drying of fluxes and electrodes, and adequate preheat.

Summary

Steels may employ a wide variety of methods to achieve specific combinations of desirable properties. Modern structural steels with their demands for strength, toughness, and good welding behaviour have evolved to depend less on carbon content as a strengthening agent and more on fine grain size and precipitation hardening. This has meant a shift in the class of problems faced by the welding engineer who may now have to determine a welding procedure and select consumables to meet stringent property requirements as well as avoiding cracking and other defects during welding.

In this Module we have described how steel may be heat treated to alter its properties and how such changes may take place in the heat affected zone of a weld. We have discussed the effects of welding on the properties of the heat affected zone and those aspects of the welding procedure that must be addressed to achieve adequate properties. The effects of various fabrication heat treatments, such as postweld stress relief, have been discussed together with the effects on properties.

Several major types of cracking problems have been covered, and we have presented guides to the procedural controls necessary to avoid them. Having successfully completed this Module you will be able to identify how the welding procedure variables affect the properties of welds in steel and estimate the procedures necessary to avoid metallurgical problems.

Additional Resources

Below is a brief list of additional sources of information that you may find valuable in your study of the Welding Metallurgy of Steels.

1. A comprehensive work on all aspects of steel making but not covering welding or fabrication is:
 - *The Making, Shaping, and Treating of Steel.*
United States Steel, Pittsburgh, Pa., U.S.A.
2. The basic principles of heat treatment of steel are covered in:
 - *Physical Metallurgy Principles* by Reed-Hill, R. E.
Published by D. Van Nostrand Company, Inc.
3. Recommended sources for information on welding of steel are:
 - *Welding Handbook*, Seventh Edition, Vol. 4
American Welding Society, 550 N.W. LeJeune Road,
Miami, FL 33126, U.S.A.
 - *Metals Handbook*, Tenth Edition, Vol 6
American Society for Metals, Metals Park, Ohio 44073,
U.S.A.
4. If you are interested in specific welding problems such as hydrogen cracking, solidification cracking, and so forth, you should consult the research literature published in major journals such as the AWS Welding Journal.

GUIDES AND EXERCISES

MODULE 21

WELDING METALLURGY OF STEELS

WELDING METALLURGY OF STEELS

MODULE 21

Guides & Exercises

To obtain maximum benefit from this module we suggest that you follow this guide and complete the exercises as indicated. It is important that you work through the text methodically, studying each section thoroughly before moving on. The exercises are designed to give you an indication of whether you have learned the material and can move on or whether you need to go back and study the section again.

Do the exercises honestly. They will not help you unless you take them seriously. If you get a question wrong go back through the text until you understand where you have gone wrong and know the correct answer.

The length of time required to complete the module will vary from student to student. Find your own pace. Do not rush. Remember you are trying to learn something, not win a race.

Some people like to underline sections when they read a text. We suggest that you use caution if you do this. What you think is important first time you read it may be different after reading it three times. We suggest you read a section three times thoroughly before highlighting anything.

The last exercise is designed to give you an indication of whether you are ready to take the closed-book exams. The exercise questions are of a similar standard to the official exams. Do not take the exam until you feel you are ready, and you may wish to study several Modules before taking the exams on each. The details of the examination procedures are on a separate sheet.

If you have any difficulties with this Module do not hesitate to ask for help. You may find you learn more by attending seminars, and you should also contact your local WIC Chapter to see if they can be of assistance. Remember, if you fail the examination you can always try again later.

MODULE 21

Guide 1

Carefully read pages 2 to 11 and answer the following questions:

1. The maximum solubility of carbon in bcc iron is—
 - a) 2%
 - b) 0.02%
 - c) no solubility
2. True or false?

"When a steel is heated into the austenite region all carbon and most other compounds dissolve to form a single phase"
3. Is pearlite a single phase?
4. What is the resulting structure of a 1.2% carbon steel slowly cooled from 900°C to room temperature?
5. Most fine grained steels retain their fine grain—
 - a) up to the melting temperature
 - b) up to the grain coarsening temperature but above that behave as coarse grained steels
 - c) only up to the transformation temperature

Check your answers for accuracy. If any of your answers are wrong re-study the guide subject matter in the text until you understand it.

MODULE 21

Guide 2

Carefully read pages 11 to 24 and answer the following questions:

1. Why is martensite produced rather than pearlite when a steel is rapidly quenched from 900°C?
2. The hardness of martensite depends mainly on—
 - a) the alloy content
 - b) the rate of cooling
 - c) the carbon content
3. Can bainite normally be produced in continuous cooling of a plain carbon steel?
4. If a steel is isothermally transformed at about 550°C in the shortest possible time how much proeutectoid ferrite is formed?
5. Increasing the carbon content—
 - a) lowers the martensite start temperature M_s
 - b) raises the M_s
 - c) has no effect on the M_s
6. Calculate the ideal critical diameter (hardenability) for a steel with 0.2%C, 0.25%Si, 1.0%Mn, 0.5%Cr, 0.5%Mo, 1.0%Ni. (Assume an initial grain size of ASTM 8)

Check your answers for accuracy. If any of your answers are wrong re-study the guide subject matter in the text until you understand it.

MODULE 21

Guide 3

Carefully read pages 25 to 29 and answer the following questions:

1. Using Fig. 34 estimate the fastest rate a weld could cool without the HAZ becoming harder than 350Hv for a steel of 0.2%C, 0.18%Si, 1.2%Mn, 0.35%Cr.
2. A tempering treatment specifies 5 hours at 500°C. What temperature would be required to achieve the same tempering effect in 1 hour?
3. Secondary hardening during tempering results from—
 - a) retained austenite
 - b) a slow heating rate
 - c) presence of strong carbide formers
4. If a thermal stress relief is applied to a welded quenched and tempered steel the temperature should not exceed—
 - a) the transformation temperature
 - b) the original tempering temperature
 - c) the grain coarsening temperature

Check your answers for accuracy. If any of your answers are wrong re-study the guide subject matter in the text until you understand it.

MODULE 21

Guide 4

Carefully read pages 30 to 36 and answer the following questions:

1. The effect of increasing carbon content of a steel on the charpy transition curve is—
 - a) to lower the transition temperature
 - b) to raise the upper shelf
 - c) to raise the transition temperature and lower the upper shelf
2. What simple heat treatment is widely used to improve the toughness of structural steels?
3. What is the main purpose of using a lower finishing temperature during controlled rolling?
 - a) to limit regrowth of austenite grains
 - b) to allow time for precipitates to dissolve
 - c) to reduce the force necessary for rolling
4. If plates are deformed more than 5% during fabrication what simple heat treatment is often specified to restore ductility?
 - a) full furnace anneal at 900°C
 - b) normalize
 - c) stress relief, typically at 625°C

Check your answers for accuracy. If any of your answers are wrong re-study the guide subject matter in the text until you understand it.

MODULE 21

Guide 5

Carefully read pages 37 to 45 and answer the following questions:

1. The austenitic grain size close to the fusion boundary is determined mainly by—
 - a) the composition of the steel
 - b) the cooling rate
 - c) the weld heat input
2. What is one of the main reasons for limiting the maximum heat input when welding HSLA steels?
 - a) to keep the HAZ hardness below 350Hv
 - b) to preserve the HAZ toughness
 - c) to promote a bainite structure
3. True or false?

"A postweld heat treatment (stress relief) always improves the fracture toughness of the HAZ"
4. Name two ways in which preheating may reduce the risk of hydrogen cracking in structural steels.
5. For a carbon increase of 0.025% in a HSLA steel estimate, from the susceptibility index given on page 45, what reduction in hydrogen content is needed to maintain the same susceptibility to hydrogen cracking.
6. Would you expect a 1%Cr, 1%Mo, 0.35%V steel to be sensitive to stress relief cracking?

Check your answers for accuracy. If any of your answers are wrong re-study the guide subject matter in the text until you understand it.

MODULE 21

Guide 6

Carefully read pages 46 to 55 and answer the following questions:

1. Acicular ferrite in weld metal is—
 - a) proeutectoid ferrite formed at the grain boundaries
 - b) a type of martensite
 - c) ferrite nucleated within the austenite grains
2. Wires for CO₂ welding usually contain silicon. Why?
 - a) to increase the strength of the weld metal
 - b) to allow easy feeding of the wire
 - c) to deoxidize the weld pool
3. Which wire has the greater deoxidizing power: ER480S-2 or ER480S-3?
4. If you wished to obtain a deposit with very low silicon using SAW which combination would you use?
 - a) a low silicon wire with an acidic (non-basic) flux
 - b) a high silicon wire with a basic flux
 - c) a low silicon wire with a basic flux
5. When might it be appropriate to specify basic fluxes for SAW?
6. Which welding parameter has the largest effect on the transfer of elements from an active flux?

Check your answers for accuracy. If any of your answers are wrong re-study the guide subject matter in the text until you understand it.

MODULE 21

Guide 7

Carefully read pages 56 to 62 and answer the following questions:

1. Will the use of argon shielding with flux cored arc welding always ensure a low oxygen deposit? Explain.
2. For what region of a weld are the lowest charpy impact values usually recorded?
 - a) close to the fusion boundary
 - b) root region
 - c) surface
3. A weld metal had a strength of 500 MPa after a postweld stress relief of 2 hours. If the time were extended to 4 hours is it likely that the weld metal would still exceed a specified minimum strength of 480 MPa?
4. Does a postweld stress relief always improve the toughness of weld metal?

Check your answers for accuracy. If any of your answers are wrong re-study the guide subject matter in the text until you understand it.

MODULE 21

ANSWERS

Guide 1

1. b)
2. True
3. No
4. Pearlite surrounded by proeutectoid cementite
5. b)

Guide 2

1. There is insufficient time for diffusion of carbon to form pearlite
2. c)
3. No
4. none
5. a)
6. 5.0

Guide 3

1. 18°C/s at 540°C
2. 527°C
3. c)
4. b)

Guide 4

1. c)
2. Normalizing
3. a)
4. c)

Guide 5

1. c)
2. b)
3. False
4. Retarding the cooling rate to give a softer structure and facilitating the diffusion of hydrogen out of the weld
5. Reduce the hydrogen by 50%
6. Yes

Answers continued

Guide 6

1. c)
2. c)
3. ER480S-2
4. c)
5. When improved weld metal toughness is needed
6. Voltage

Guide 7

1. No. The flux will also affect the oxygen content
2. b)
3. Yes
4. No

MODULE 21

TEST

This test is designed to determine whether you are ready to attempt the formal examination.

Complete the ANSWER SHEET and compare the results with the TEST KEY. If you score less than 70% we suggest you re-study the material.

1. What is the maximum solubility of carbon in bcc iron?
 - (a) 0.8%
 - (b) none
 - (c) 2%
 - (d) 0.02%
 - (e) 6.67%
2. What is the maximum solubility of carbon in fcc iron?
 - (a) 0.8%
 - (b) none
 - (c) 2%
 - (d) 0.02%
 - (e) 6.67%
3. How can the toughness of martensite be improved?
 - (a) cooling more quickly
 - (b) tempering
 - (c) normalizing
 - (d) increasing carbon content
 - (e) cold working
4. If excellent HAZ toughness were required which of the following options would you choose in developing a welding procedure?
 - (a) use the least number of passes
 - (b) do all welding in the vertical up position
 - (c) specify a high preheat and interpass temperature
 - (d) use low heat input
 - (e) specify only SAW
5. How does normalizing improve the toughness of carbon steels?
 - (a) by causing precipitation hardening
 - (b) by refining the austenite grains and promoting fine pearlite
 - (c) by tempering martensite
 - (d) by lowering the oxygen content
 - (e) by removing stresses

MODULE 21

TEST continued

6. Which element has the greatest effect on susceptibility to hydrogen cracking?
 - (a) nickel
 - (b) manganese
 - (c) vanadium
 - (d) carbon
 - (e) copper
7. If a wire with 0.10%C, 0.05%Si, 0.5%Mn were used for CO₂ welding what problem might you expect?
 - (a) excessive hydrogen cracking
 - (b) greater degree of undercut
 - (c) porosity
 - (d) poor HAZ toughness
 - (e) hard weld metal
8. A minimum energy input may be specified for welding a carbon steel when no preheat is used—
 - (a) to prevent precipitation hardening
 - (b) to reduce the HAZ grain size
 - (c) to remove moisture
 - (d) to keep below the critical cooling rate and prevent a hard HAZ forming
 - (e) to reduce stresses
9. Under what conditions is the HAZ likely to be harder after a stress relief?
 - (a) never
 - (b) when the weld was made under high restraint
 - (c) when very long stress relief times are used
 - (d) when the steel contains carbide formers such as vanadium
 - (e) when the carbon content exceeds 0.2%
10. Which weld is likely to have the best charpy properties?
 - (a) a multipass weld with a high percentage of refined structure
 - (b) a high heat input weld
 - (c) weld metal deposited with low basicity fluxes
 - (d) weld metal with a minimum of acicular ferrite
 - (e) single pass welds

**Welding Institute of Canada
Answer Sheet
Module 21**

Complete the "Answer Sheet" and compare your answers with those in the "Test Key." If you score less than 70% we suggest you re-study the material.

Please circle only ONE letter corresponding to the answer you think is most correct.

QUESTION	ANSWER				
1	a	b	c	d	e
2	a	b	c	d	e
3	a	b	c	d	e
4	a	b	c	d	e
5	a	b	c	d	e
6	a	b	c	d	e
7	a	b	c	d	e
8	a	b	c	d	e
9	a	b	c	d	e
10	a	b	c	d	e

The answer sheet below is provided in the event you wish to re-test yourself.

QUESTION	ANSWER				
1	a	b	c	d	e
2	a	b	c	d	e
3	a	b	c	d	e
4	a	b	c	d	e
5	a	b	c	d	e
6	a	b	c	d	e
7	a	b	c	d	e
8	a	b	c	d	e
9	a	b	c	d	e
10	a	b	c	d	e

**Welding Institute of Canada
Test Key
Module 21**

Compare your answer sheet to this key

QUESTION	ANSWER				
1	a	b	c	(d)	e
2	a	b	(c)	d	e
3	a	(b)	c	d	e
4	a	b	c	(d)	e
5	a	(b)	c	d	e
6	a	b	c	(d)	e
7	a	b	(c)	d	e
8	a	b	c	(d)	e
9	a	b	c	(d)	e
10	(a)	b	c	d	e