

SOME RECENT ADVANCES IN THE STUDY OF TRANSFORMATIONS IN ALLOY STEELS

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Abstract

Some effects of partial transformation of austenite at one subcritical temperature on the isothermal transformation characteristics of the residual austenite at a different temperature are described. Formulae are presented for calculating from the chemical composition of a steel the progress of martensite formation with decrease of temperature during rapid cooling. The general features of continuous-cooling transformation diagrams of alloy steels are discussed with reference to a selection of typical diagrams. The practical significance of diagrams of this type is indicated. The relative hardenabilities of a number of steels are compared using several available methods of assessment. The various methods are shown to give dissimilar results.

THE addition of alloying elements to steels has long been recognized as a means of securing increased strength thus permitting considerable saving in weight of components in many fields of steel application. This increase in strength arises from the influence of the alloying elements on the degree of hardening which can be achieved when an austenitized steel is cooled under a given set of conditions. The effect of alloying elements on the degree of hardening of a steel is the practical result, however, of the fundamental changes effected in the transformation characteristics of the steel. A knowledge of these changes has largely been revealed through isothermal transformation studies of alloy steels, and it is now well established that two critical cooling rates exist for many alloy steels. One of these must be exceeded if transformation to ferrite/pearlite structures is to be avoided during cooling, whilst if the other critical cooling rate is exceeded, transformation to bainite structures does not occur. It is

perhaps not widely appreciated that although pearlite is not often encountered in hardened alloy steels, the occurrence of fairly large proportions of bainite is not uncommon. Since bainite and martensite are the constituents mainly encountered in hardened alloy steels it is not surprising, therefore, that considerable attention has been given to the study of the mechanisms by which these products of austenite transformation are formed, whilst many investigations have been made of the various factors which may influence the progress of these transformations.

The present paper is devoted mainly to a description of the results of some investigations made in the laboratory of the Mond Nickel Co. Ltd. during the course of a general study of isothermal and athermal transformations in alloy steels.

The Bainite Reaction

Until recently a certain amount of doubt existed on the question of whether the formation of acicular bainite at relatively low temperatures was a nucleation and growth process, as the dependence of the progress of the reaction on time appeared to suggest, or whether plates of this constituent formed by an almost instantaneous shear type of reaction similar to that for martensite formation. The reason for this doubt could be ascribed to the great similarity of the acicular bainite structures with those of lightly tempered martensite. The work of Ko and Cottrell¹ has shown, however, that the process is one of nucleation and continuous growth. These authors used a hot-stage microscope for observing and photographically recording the changes in the structure of

steel during isothermal transformation. It was seen that the bainite needles or plates extended in length with time, but that lateral growth of the needles was comparatively slow. It was also shown that the bainite crystal lattice was coherent with that of the austenite from which it formed since the needles stood out in sharp relief from the plane surface of a polished sample in which partial transformation to bainite had been allowed to occur. Pro-pearlitic ferrite and pearlite, however, were found to be non-coherent with the lattice of the parent austenite.

The author recently made a detailed microscopic study of isothermal bainite formation in a 3 per cent nickel-chromium steel and the results were in accord with Ko's suggestion that the process was one of nucleation and growth at all temperatures at which the transformation occurred. It was clear, however, that whereas at low temperatures the growth of the bainite needles was mainly in the direction of their length, this characteristic was only evident in the initial stages of growth of a bainite crystal formed at relatively high temperatures in the bainite formation range. At these temperatures the growing crystals of bainite developed a needle-like shape initially, but when growth at the ends of the needles ceased on impingement with other crystals or at the austenite grain boundaries, growth continued at a significant rate in a direction normal to the major axes of the crystals. Thus as the transformation proceeded, the needle-like shape disappeared. It was further apparent that although the incubation period at low temperatures in the bainite region was greater than that required for the start of transformation to upper bainite, the rate of formation of new crystals proceeded more rapidly at the lower temperatures once a certain amount of bainite had formed. This was clearly due to the fact that each lower bainite needle rapidly nucleated the formation of other needles whereas at higher temperatures this tendency

was less marked. This process appears to be a major factor, therefore, in the mechanism of lower bainite formation. These differences between the mode of formation of upper and lower bainite are illustrated in Figs. 1-6.

In a series of experiments in which the 3 per cent nickel-chromium steel was austenitized, allowed to transform partly at one temperature, and then quenched into an isothermal bath at a second temperature, it was found that nucleation of lower bainite occurred preferentially at crystals of pro-pearlitic ferrite or upper bainite which had been formed in the first stage of the double isothermal treatment (Figs. 7 and 8). The provision of favourable sites for nucleation did not, however, significantly affect the incubation period of the lower bainite transformation or the subsequent rate of the reaction as indicated by the reaction curves shown in Fig. 9.

The prior formation of some needles of lower bainite in samples subsequently transformed at a temperature in the upper bainite region (475°C.), however, reduced the incubation period required for isothermal formation of upper bainite and also increased the reaction rate (Fig. 10). The photomicrographs shown in Figs. 11-14 reveal that growth of the lower bainite needles occurred at the upper bainitic temperature and the appearance of light-etching rims of upper bainite around carbon-rich cores may be noted in Fig. 14. In addition, new crystals of upper bainite were also formed at other sites and these were relatively carbide-free.

The presence of a small amount of pro-pearlitic ferrite in the austenite matrix also appeared to reduce the incubation period for isothermal formation of upper bainite at 475°C., but in this case the subsequent reaction rate was noticeably reduced (Fig. 15). This effect was rather surprising. It is possible, however, that the pro-pearlitic ferrite continued to grow as a non-coherent phase (with respect to the austenite lattice), but that its rate of growth was lower than



FIG. 1 — 180 SEC. AT 320°C. AND WATER-QUENCHED. $\times 750$



FIG. 2 — 240 SEC. AT 320°C. AND WATER-QUENCHED. $\times 750$



FIG. 3 — 360 SEC. AT 320°C. AND WATER-QUENCHED. $\times 750$

THREE STAGES IN THE FORMATION OF LOWER BAINITE IN A 3 PER CENT NICKEL-CHROMIUM STEEL ISOTHERMALLY TRANSFORMED AT 320°C.



FIG. 4 — 35 SEC. AT 475°C. AND WATER-QUENCHED. $\times 750$



FIG. 5 — 100 SEC. AT 475°C. AND WATER-QUENCHED. $\times 750$



FIG. 6 — 210 SEC. AT 475°C. AND WATER-QUENCHED. $\times 750$

THREE STAGES IN THE FORMATION OF UPPER BAINITE IN A 3 PER CENT NICKEL-CHROMIUM STEEL ISOTHERMALLY TRANSFORMED AT 475°C.

that of the coherent upper bainite, the nucleation of which might have been retarded by the presence of the pro-pearlitic ferrite at preferred sites for nucleation in the austenite lattice.

The effect of partial transformation of austenite to one product accelerating the

initiation of transformation to another has also been observed when steel samples have been quenched to a temperature within the martensite formation range to form a proportion of martensite and then held for a period of time at this temperature. After a time isothermal transformation to bainite

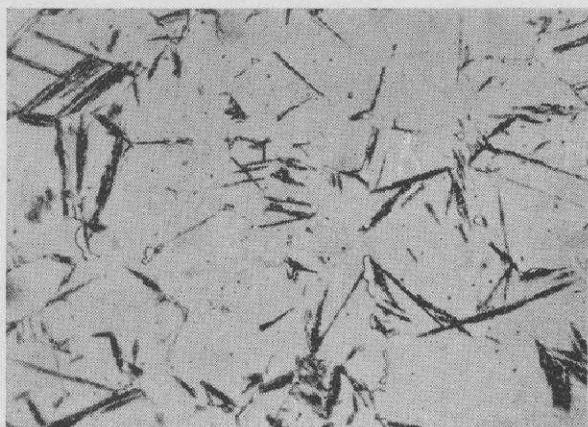


FIG. 7 — MICROSTRUCTURE OF A 3 PER CENT NICKEL-CHROMIUM STEEL ISOTHERMALLY TREATED AT 600°C. FOR 1 HR. FOLLOWED BY ISOTHERMAL TREATMENT AT 320°C. FOR 100 SEC. $\times 750$



FIG. 8 — MICROSTRUCTURE OF A 3 PER CENT NICKEL-CHROMIUM STEEL ISOTHERMALLY TREATED AT 475°C. FOR 30 SEC. FOLLOWED BY ISOTHERMAL TREATMENT AT 320°C. FOR 120 SEC. $\times 750$

occurs and the incubation period required for the start of this transformation has been noted to be less than would be expected from the extrapolation of the curves of the isothermal transformation diagram at temperatures above M_s . A similar effect of the presence of martensite in an austenite matrix has also been noted on the incubation periods required for isothermal transformation to bainite at temperatures above the M_s temperature of a steel. Data obtained by Steven and Haynes² illustrating these effects are presented in Fig. 16.

The practical significance of observations such as those which have been described is likely to be realized in certain special heat treatments for which increasing applications are being found. For example, it has been suggested by Elmendorf³ that the optimum combination of tensile strength and ductility may be obtained in steels of high carbon content if structures consisting of tempered martensite and bainite are obtained. Structures of this type can, of course, be secured most conveniently by a controlled quench to selected temperatures within the martensite formation range followed by isothermal transformation at any desired temperature in the bainite range. As indicated above the formation of martensite prior to the iso-

thermal transformation to bainite can often reduce the time necessary for completion of the transformation in comparison to the time which would be required if the two constituents were formed in the reverse order.

The Martensite Reaction

A knowledge of the temperatures at which martensite formation commences and of the progress of this reaction with decreasing temperature is frequently useful in the heat treatment of alloy steels. In addition, the

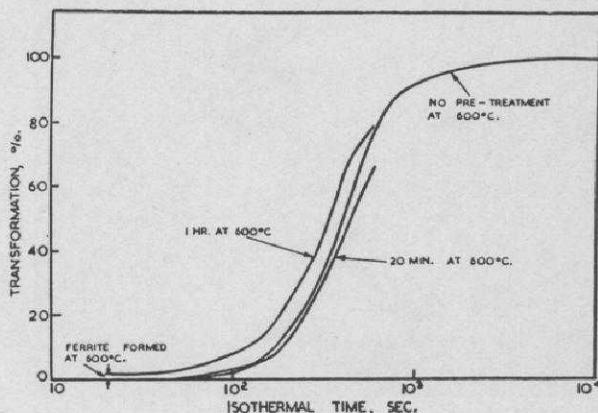


FIG. 9 — ISOTHERMAL REACTION CURVES FOR A 3 PER CENT NICKEL-CHROMIUM STEEL TRANSFORMED AT 320°C. WITH AND WITHOUT PRIOR TREATMENT AT 600°C.

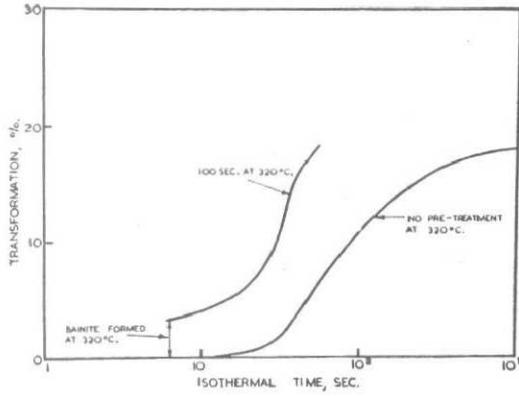


FIG. 10—ISOTHERMAL REACTION CURVES FOR A 3 PER CENT NICKEL-CHROMIUM STEEL TRANSFORMED AT 475°C. WITH AND WITHOUT PRIOR ISOTHERMAL TREATMENT AT 320°C.

recent work of Cottrell⁴ suggests that this would also provide a useful indication of the liability of a steel to crack formation after welding. Several formulae have been published⁵⁻⁹ for the calculation of M_s temperatures from the chemical composition of a steel. Steven and Haynes² compared M_s temperatures derived from these various formulae with those obtained experimentally on a wide selection of alloy steels and, in general, they found discrepancies between the two sets of values for steels with carbon contents below about 0.25 per cent. As a result of their investigations these authors suggested that the following formula gave a



FIG. 11—LOWER BAINITE CRYSTALS IN MARTENSITE MATRIX IN A 3 PER CENT NICKEL-CHROMIUM STEEL ISOTHERMALLY TRANSFORMED AT 320°C. FOR 100 SEC. AND WATER-QUENCHED. $\times 750$

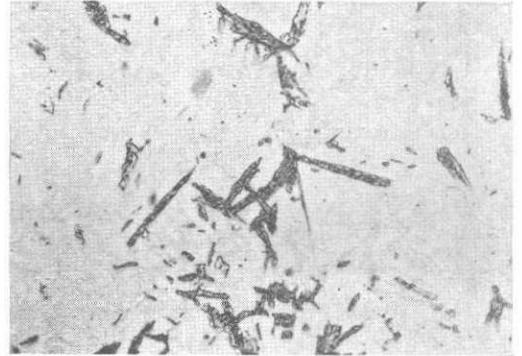


FIG. 12—3 PER CENT NICKEL-CHROMIUM STEEL ISOTHERMALLY TREATED AT 320°C. FOR 100 SEC. FOLLOWED BY ISOTHERMAL TREATMENT AT 475°C. FOR 11 SEC. $\times 750$

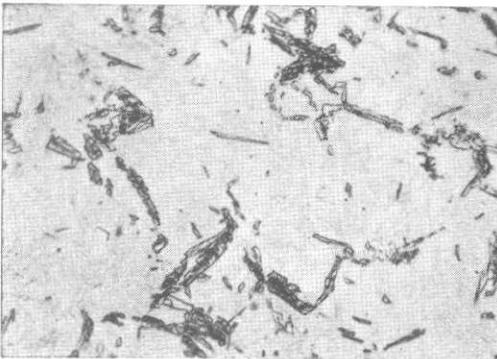


FIG. 13—3 PER CENT NICKEL-CHROMIUM STEEL ISOTHERMALLY TREATED AT 320°C. FOR 100 SEC. FOLLOWED BY ISOTHERMAL TREATMENT AT 475°C. FOR 30 SEC. $\times 750$

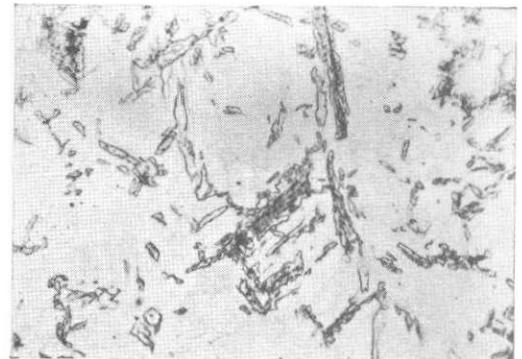


FIG. 14—3 PER CENT NICKEL-CHROMIUM STEEL ISOTHERMALLY TREATED AT 320°C. FOR 100 SEC. FOLLOWED BY ISOTHERMAL TREATMENT AT 475°C. FOR 60 SEC. $\times 750$

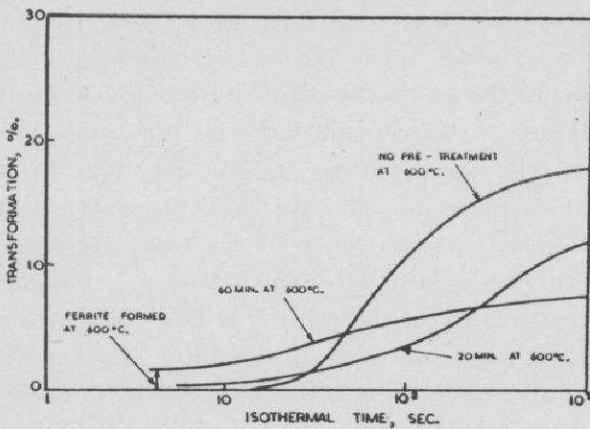


FIG. 15 — Isothermal reaction curves for a 3 per cent nickel-chromium steel transformed at 475°C. with and without prior isothermal treatment at 600°C.

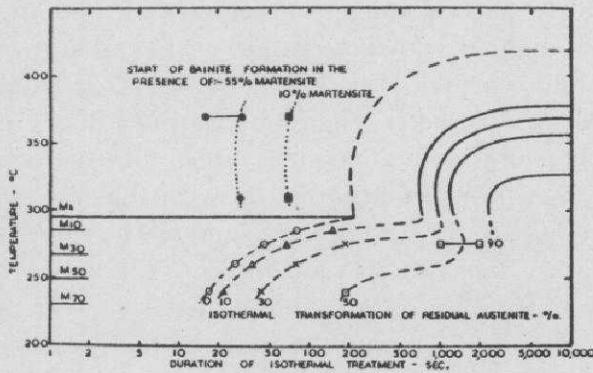


FIG. 16 — Effect of martensite in an austenite matrix on the isothermal transformation of the austenite to bainite at temperatures above and below M_s (after Steven and Haynes)

more reliable indication of M_s temperatures for steels in which complete carbide solution is effected during austenitizing:

$$M_s (\text{°C.}) = 561 - 474 (\%C) - 33 (\%Mn) - 17 (\%Ni) - 17 (\%Cr) - 21 (\%Mo)$$

The formula is applicable to steels having compositions within the following limits:

	Per cent
Carbon	0.1-0.55
Silicon	0.1-0.35
Manganese	0.2-1.70
Nickel	Trace to 5.0
Chromium	Trace to 3.5
Molybdenum	Trace to 1.0

The calculated M_s temperatures for a variety of alloy steels, using the above formula, are compared with the experimentally determined values in Fig. 17.

Steven and Haynes also studied the progress of martensite formation with decreasing temperature below M_s . For a selection of alloy steels containing 0.32-0.44 per cent carbon it was found that similar amounts of martensite were formed in all the steels for given decrements of temperature below the M_s value. Thus the temperatures at which various amounts of martensite were formed could be related to the M_s temperatures of the steels by the following simple relationships:

$$M_{10} (\text{°C.}) = M_s - 10$$

$$M_{50} (\text{°C.}) = M_s - 47$$

$$M_{90} (\text{°C.}) = M_s - 103$$

$$M_f (\text{°C.}) = M_s - 215$$

Continuous-cooling Transformation Diagrams

Although isothermal transformation studies provide useful information for the design

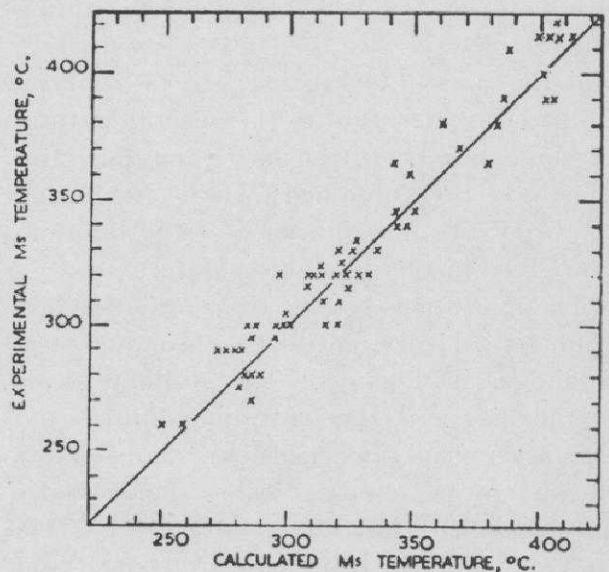


FIG. 17 — Comparison of experimentally determined M_s temperatures and values calculated from chemical composition (after Steven and Haynes)

and control of special heat treatments which can be used for steels, the application of isothermal transformation diagrams to normal heat treatments involving continuous cooling through the subcritical temperature range is of limited value. Various attempts have been made to derive information from isothermal diagrams on the transformations which are likely to occur in steels when austenitized and cooled at known rates. These attempts have not been entirely successful. Consequently in recent years much attention has been given to methods of determining the continuous-cooling transformation characteristics of steels and to the presentation of this information in the form of diagrams covering a range of cooling rates. If the cooling rates employed are related to those obtained in practice, then the diagrams obtained provide a direct indication of the transformation behaviour of the steels when cooled by normal heat treatment methods.

The method used for the determination of continuous-cooling transformation diagrams in the author's laboratory employs cooling schedules which simulate those of oil-quenched cylindrical bars varying in diameter between 1 and 6 in. A complete description of this method has been given elsewhere¹⁰, and the present discussion will be confined to directing attention to the general features of some types of transformation diagrams that have been obtained for alloy steels and to the practical significance of the information given in this type of diagram.

The continuous-cooling transformation diagram for a 1¼ per cent nickel-chromium steel is shown in Fig. 18. The ordinate scale of the diagram represents temperature and a range of oil-quenched bar diameters is shown on the abscissa scale. Three scales are included for different positions in the oil-quenched bars ($r/b = 0, 0.5, \text{ and } 0.8$, where $r = \text{distance from the axis of the bar, and } b = \text{radius of bar}$). Thus, for example, the cooling at the mid-radius

of a 2 in. diameter bar is represented by the 2 in. position along the $r/b = 0.5$ scale. The curves shown on the diagram indicate various stages of transformation and for each bar size the temperature can be obtained from these curves for the start and later stages of transformation at each of the three standard positions in the oil-quenched bar. The M_s temperature of the steel is indicated, and the hardness values shown along the lower edge of the diagram are the as-quenched values for the different bar sizes.

Fig. 18 indicates that in the 1¼ per cent nickel-chromium steel transformation of the austenite at the axes of oil-quenched bars greater than 4¼ in. diameter occurs over a fairly narrow range of temperature and the product of transformation consists of ferrite and pearlite. For bars between 2¼ and 4¼ in. diameter the temperature range of transformation is much greater. In this size range transformation starts with the separation of ferrite and pearlite, and as the temperature falls bainite also forms. The two reactions may occur simultaneously over a range of temperature, but as the temperature decreases the formation of pearlite is gradually replaced by the formation of bainite. At the upper end of this size range (i.e. 4¼ in. diameter) the transformation lines representing the later stages of transformation sweep sharply upwards. This indicates a critical diameter for the steel from a heat treatment viewpoint, and for bars of about this size slight variations in quenching rates, or in the transformation characteristics of the steel from bar to bar or from cast to cast, would lead to wide variations in the microstructure of the hardened steel and consequently variations may be obtained in the mechanical properties after tempering. The effects would, of course, be minimized the higher the tempering temperature employed. With decreasing bar diameter below 4¼ in. the amount of ferrite and pearlite formed decreases until at 2¼ in. diameter little or no pearlite is formed. This variation in the

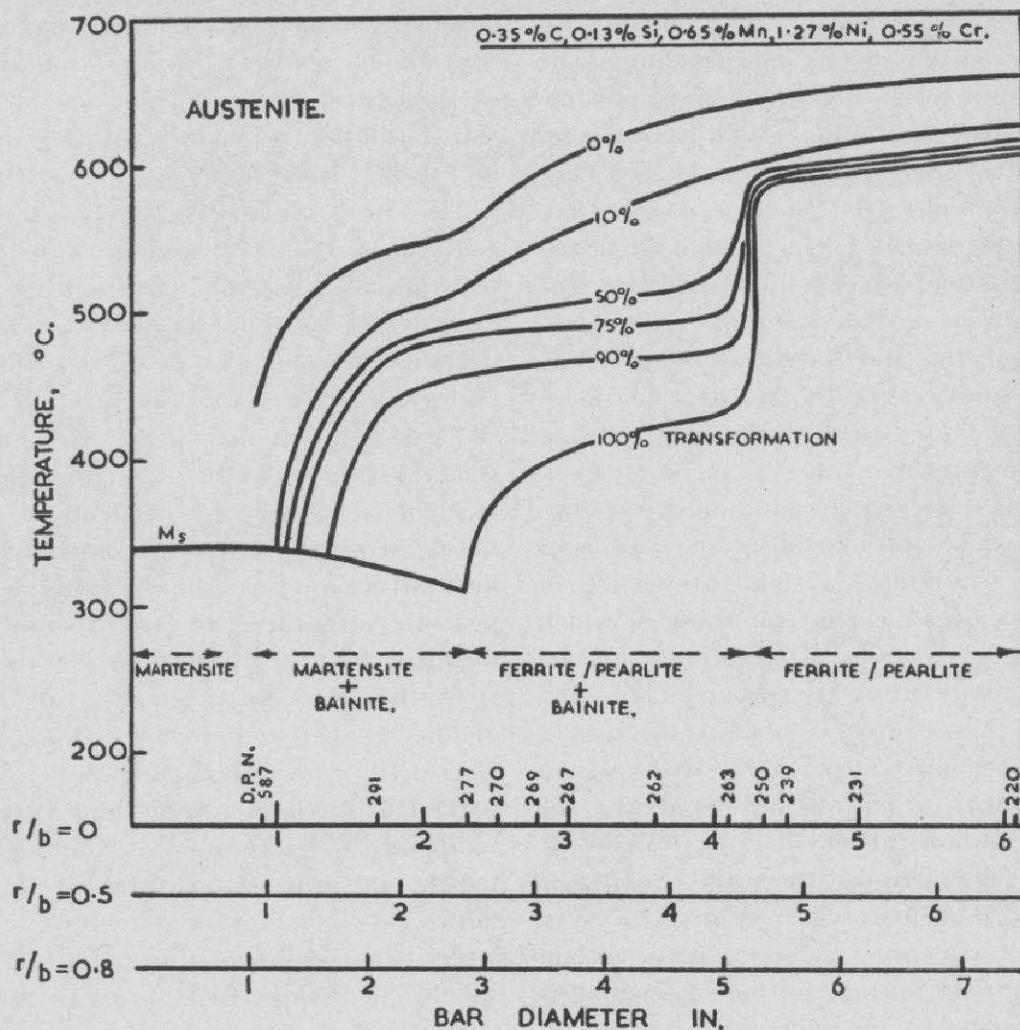


FIG. 18 — CONTINUOUS-COOLING TRANSFORMATION DIAGRAM FOR OIL-QUENCHED BARS OF A 1½ PER CENT NICKEL-CHROMIUM STEEL AUSTENITIZED AT 845°C.

proportions of the different constituents formed at the different bar sizes is not indicated precisely on continuous-cooling transformation diagrams of the type shown in Fig. 18. From the temperature ranges shown for the various stages of transformation, and a knowledge from isothermal studies of the nature of the constituents formed at different temperatures, it is possible, however, to obtain an indication of the relative amounts of these constituents in the as-quenched microstructure.

In the size range in which ferrite/pearlite structures are the first to separate during cooling, the temperature for the start-of-transformation usually decreases gradually

as the bar size decreases, but in the size range in which bainite is the first constituent to form there is often a range of bar sizes in which the temperature for the start-of-transformation remains constant, or falls less rapidly. Thus the minimum diameter for partial pearlite formation is often indicated by an inflection in the curve representing the start-of-transformation. In the diagram shown in Fig. 18 an inflection of this type occurs at the 2¼ in. diameter position. For bars smaller than 2¼ in. diameter the product of transformation is, therefore, substantially bainitic, or martensitic, or consists of various proportions of each of these constituents. This diameter represents, therefore, the

maximum diameter for the avoidance of pearlite formation during oil-quenching. The critical diameter below which it is possible to obtain full hardening is not shown precisely in the diagram for the $1\frac{1}{4}$ per cent nickel-chromium steel, since it is less than the smallest diameter which could be tested by the particular technique used for the determination of the diagram. An extrapolation of the start-of-transformation line suggests, however, that it is about 0.8 in.

It will be noted that the line representing the M_s temperature of the $1\frac{1}{4}$ per cent nickel-chromium steel is horizontal over the size range in which full martensitic structures or structures containing a small proportion of bainite are obtained, but for those in which increasing proportions of bainite are formed the M_s temperature is progressively depressed. This can be ascribed to the enrichment of the austenite matrix with carbon during bainite formation, and the greater the amount of low-carbon bainite formed at relatively high temperatures, the greater is the effect. This feature is not shown by every steel, however, since the amount of low-carbon bainite formed during continuous cooling is often insufficient to increase significantly the carbon content of the austenite matrix.

A similar type of diagram to that of the $1\frac{1}{4}$ per cent nickel-chromium steel, but with the ferrite/pearlite transformation extending over a greater range of bar size, is shown in Fig. 19 for a 1 per cent nickel steel.

The continuous-cooling diagram for a 1 per cent chromium steel (FIG. 20) shows zones of ferrite/pearlite and bainite transformation which are sharply separated. For bar diameters greater than about 6 in. the former type of structure is formed throughout the section of each bar, whilst bars between approximately 3 and 5 in. diameter would have completely bainitic structures.

Many of the deeper hardening steels do not transform to ferrite/pearlite structures when

oil-quenched as 1-6 in. diameter bars and the continuous-cooling transformation diagrams of these steels show only a zone of bainite transformation. Typical examples are shown in Figs. 21 and 22.

The transformation data given by the continuous-cooling diagrams for quenched bars provide a useful indication of the suitability of a steel for applications in which particular mechanical properties after tempering are required. In general, fully hardened (100 per cent martensite) and tempered structures provide the best combination of mechanical properties, particularly at high tensile strengths. Bainitic or partially bainitic structures are slightly inferior, whilst pearlitic structures are only to be recommended for the lower ranges of tensile strength. Bars up to about $\frac{3}{4}$ in. diameter of the $1\frac{1}{4}$ per cent nickel-chromium steel (FIG. 18), for example, would be fully hardened by oil-quenching, and after lightly tempering a satisfactory combination of tensile and impact properties should be obtained at high tensile strengths. Slightly larger bars containing a proportion of bainite should be tempered to somewhat lower strengths to obtain comparable impact resistance. The proportion of bainite in the oil-quenched bars of this steel increases rapidly, however, with increase in bar diameter above the critical size for full hardening and it is probable that sufficient control of heat treatment could not be effected to secure consistent properties at relatively high tensile strengths. Tempering to intermediate strength levels is, therefore, advisable for all bar sizes up to about 2 in. diameter. Bars between 2 and 4 in. diameter should be tempered to still lower tensile strengths to secure satisfactory impact properties, and those greater than about 4 in. diameter would only be suitable for relatively low strength applications.

The diagram for the 3 per cent nickel-chromium-molybdenum steel shows that it could be satisfactorily oil-hardened in bar

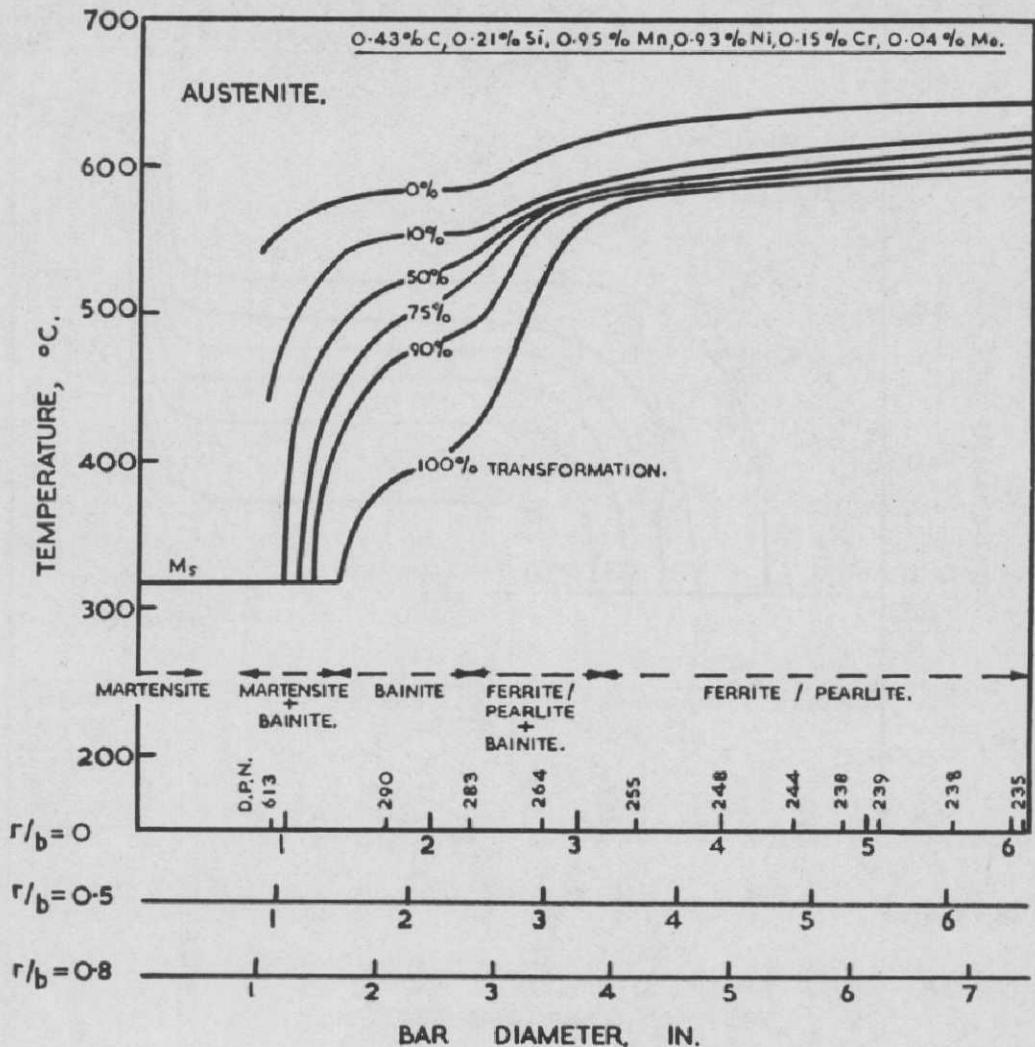


FIG. 19 — CONTINUOUS-COOLING TRANSFORMATION DIAGRAM FOR OIL-QUENCHED BARS OF A 1 PER CENT NICKEL STEEL AUSTENITIZED AT 845°C.

diameters up to 4 in. At 5 in. diameter the proportion of bainite formed is still not excessive and it is evident, therefore, that a steel of this type is suitable for many relatively high strength applications.

Comparison of the Hardenabilities of Steels by Several Methods

As already indicated, continuous-cooling transformation diagrams provide a useful means of assessing the hardenabilities of the steels. Isothermal transformation diagrams have been used to assess relative hardenabilities and several other methods have also

been described. These include the end-quench hardenability test¹¹ and values calculated from chemical composition¹²⁻¹⁴. The end-quench test is not suitable, however, for the deeper hardening alloy steels.

It has been suggested by Scheil¹⁵ that the temperature at which transformation begins during continuous cooling can be derived from the isothermal transformation diagram of a steel if the cooling schedule is known. The method which he proposed involves integrating the times spent in cooling through small intervals of temperature expressed as fractions of the isothermal incubation periods for these temperature intervals. The

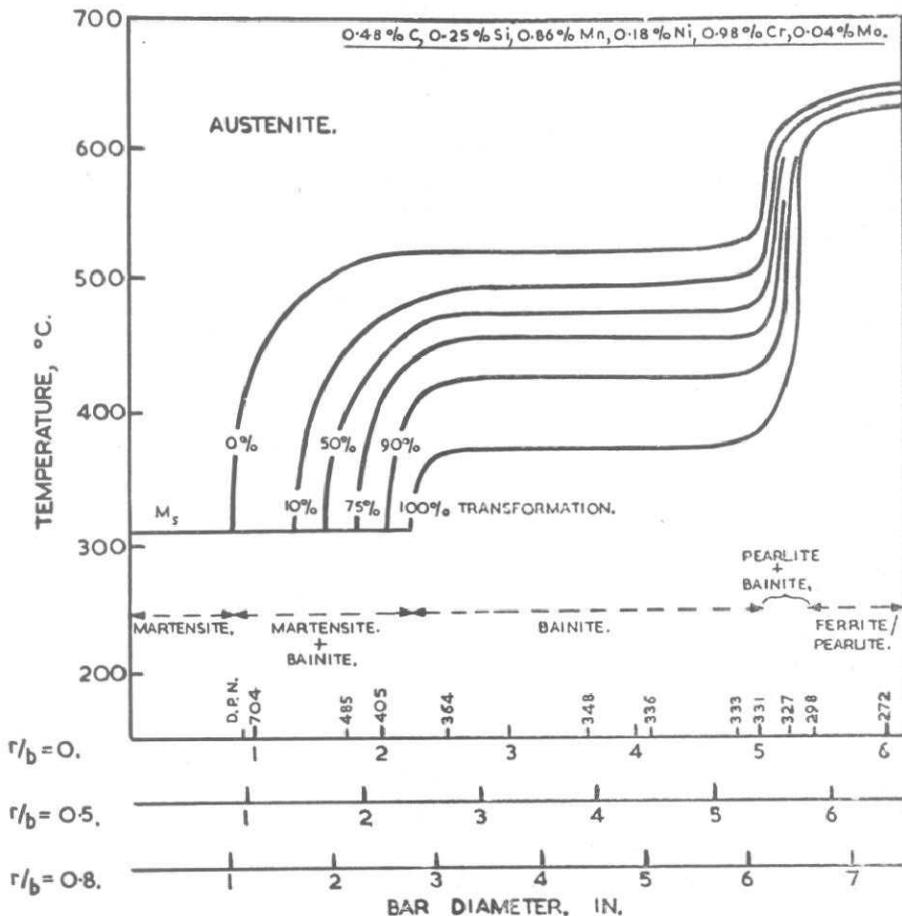


FIG. 20—CONTINUOUS-COOLING TRANSFORMATION DIAGRAM FOR OIL-QUENCHED BARS OF A 1 PER CENT CHROMIUM STEEL AUSTENITIZED AT 860°C.

temperature at which the integration of these fractions equals unity is then assumed to be that at which transformation begins during cooling. By applying this method to known cooling schedules for oil-quenched bars of various sizes it is possible to determine the critical bar diameter below which a steel can be fully hardened. Manning and Lorig¹⁶ later suggested that the time spent in cooling through the pearlite formation temperature range does not contribute to the time required for transformation in the bainite range.

During the course of a transformation study using a variety of alloy steels their continuous-cooling and isothermal diagrams

were determined and the Scheil method of calculating critical diameters for complete hardening on oil-quenching was applied to the isothermal diagrams. The values obtained were significantly different from those derived from the continuous-cooling transformation diagrams. It was apparent, therefore, that this method of assessing hardenability was not reliable. Nevertheless, although no satisfactory method was available for deriving absolute hardenability data from isothermal transformation diagrams, it was expected that these diagrams should, at least, provide an indication of the relative hardenabilities of different steels. Analysis of the available data, however, did not

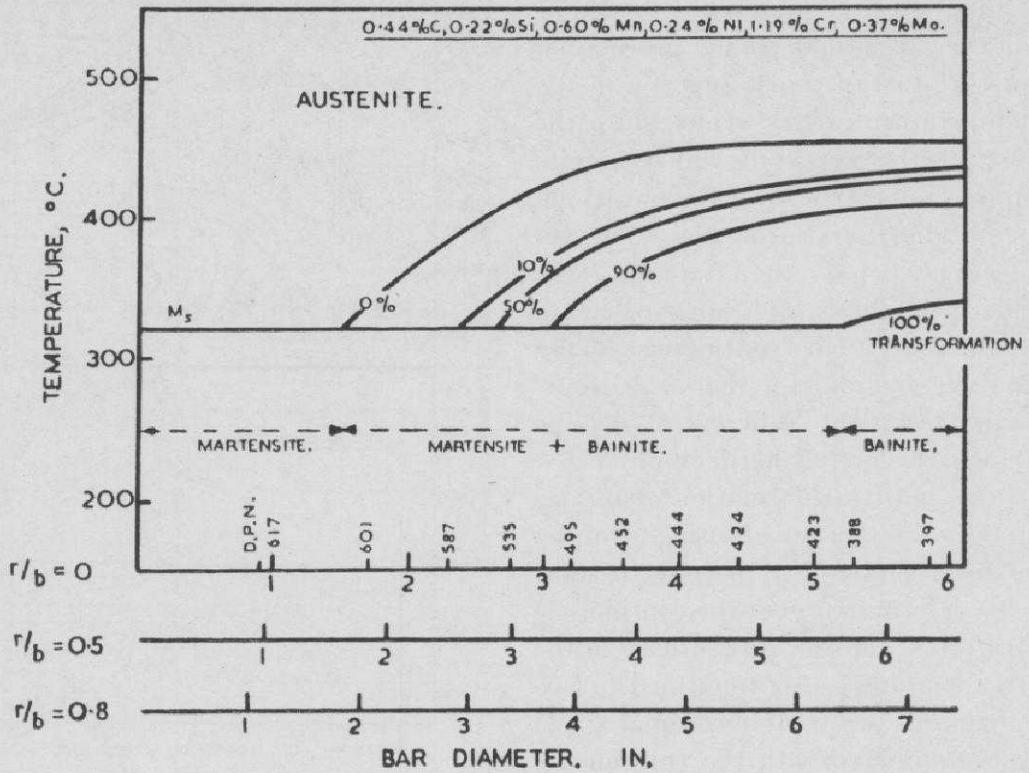


FIG. 21 — CONTINUOUS-COOLING TRANSFORMATION DIAGRAM FOR OIL-QUENCHED BARS OF A 1 PER CENT CHROMIUM-MOLYBDENUM STEEL AUSTENITIZED AT 850°C.

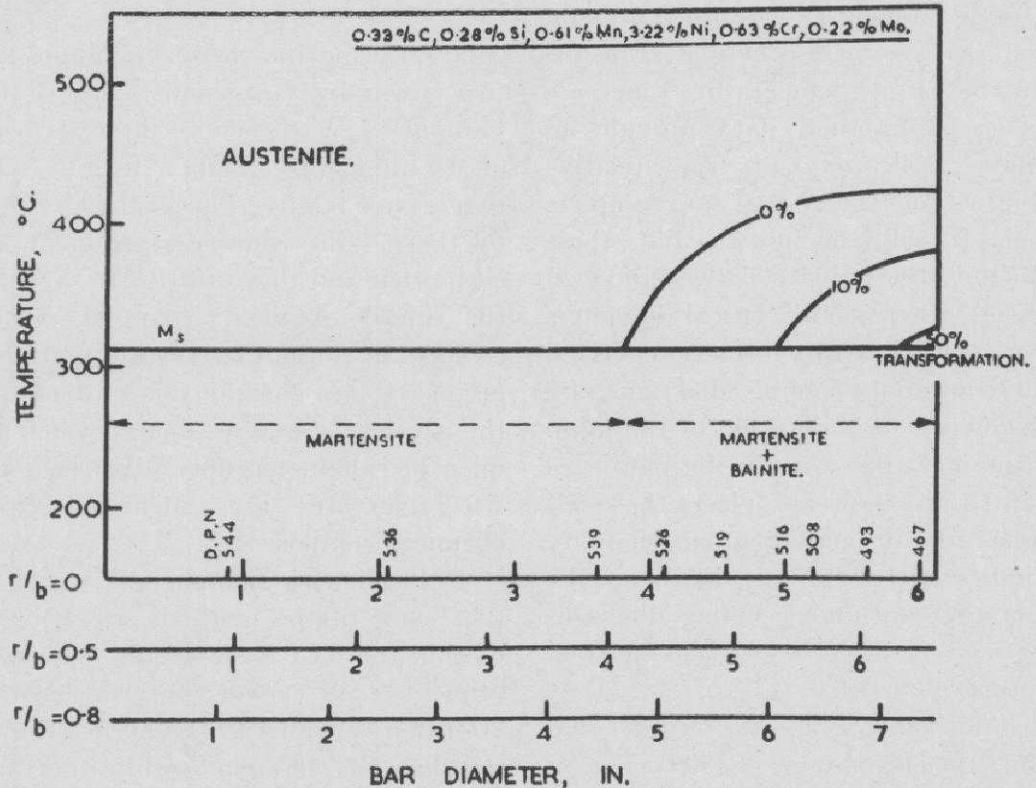


FIG. 22 — CONTINUOUS-COOLING TRANSFORMATION DIAGRAM FOR OIL-QUENCHED BARS OF A 3 PER CENT NICKEL-CHROMIUM-MOLYBDENUM STEEL AUSTENITIZED AT 835°C.

support this view. It was noted that several of the steels had similar isothermal curves for the start-of-transformation in the bainite temperature range. Thus, since the steels were of the types in which bainite formation had to be avoided if full hardening was to be obtained, and using the ability to harden completely as the criterion, the relative hardenabilities did not appear to be very different. The continuous-cooling transformation diagrams of the steels, however, showed marked differences in the critical diameters for full hardening.

The isothermal transformation diagrams were carefully redetermined using samples machined from similar bar material to that used for the determination of the continuous-cooling diagrams, and the authenticity of the isothermal diagrams was confirmed. Two of these diagrams are reproduced in Fig. 23 and may be compared with the continuous-cooling transformation diagrams for the same steels shown in Figs. 21 and 22.

The compositions of the several steels are given in Table 1, and in Table 2 the minimum incubation periods for isothermal transformation in the bainite temperature range are given. These isothermal data provide an approximate indication of the relative hardenabilities of the steels for complete hardening. It will be noted that these values do not discriminate between five of the six steels, whereas the critical diameters for 100 per cent martensite formation derived from the continuous-cooling diagrams are quite dissimilar. Consideration of the minimum isothermal time for the formation of 50 per cent bainite, however, places the steels in the same order of increasing hardenability as that indicated by the bar diameters derived from the continuous-cooling diagrams for 50 per cent bainite - 50 per cent martensite structures.

Included in Table 2 are the critical diameters for the formation of 50 per cent martensite structures at the axes of oil-quenched bars which were derived by cal-

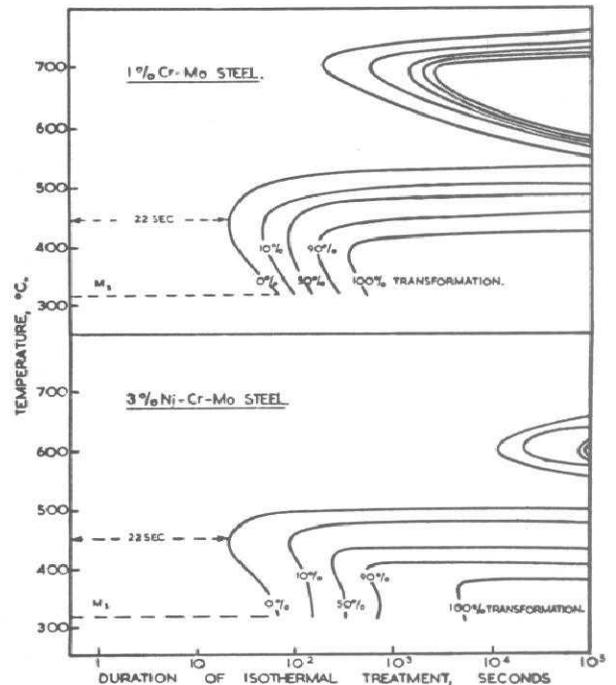


FIG. 23 — ISOTHERMAL TRANSFORMATION DIAGRAMS FOR 1 PER CENT CHROMIUM-MOLYBDENUM AND 3 PER CENT NICKEL-CHROMIUM-MOLYBDENUM STEELS

culating using the chemical composition factors given by Grossman¹² and Crafts and Lamont¹³. A quenching intensity value (h) of 0.8 in.⁻¹ was assumed for the oil. The two sets of relative hardenabilities indicated by these values show disagreement between each other and they are also at variance with the results obtained from the continuous-cooling or isothermal transformation measurements. In general, the hardenabilities of the steels are over-estimated when the oil-quenched diameters for 50 per cent martensite structures are calculated from the chemical compositions.

From this comparison it is evident that some of the methods which have been used in the past for assessing the hardenabilities of steels can be quite misleading, and attention is drawn to the need for an understanding of the relationships between isothermal and continuous-cooling transformation characteristics. Until these have been

TABLE 1 — ANALYSES OF STEELS USED FOR THE ASSESSMENT OF HARDENABILITIES BY SEVERAL METHODS

STEEL No.	COMPOSITION WT. %						AUSTENITIZING TEMPERATURE, °C.	AS-QUENCHED GRAIN SIZE (A.S.T.M.)
	C	Si	Mn	Ni	Cr	Mo		
1	0.44	0.22	0.60	0.24	1.19	0.37	860	6
1	0.38	0.20	0.69	1.58	0.95	0.26	855	7 and 8
3	0.40	0.21	1.34	1.03	0.53	0.22	845	6
4	0.38	0.15	0.56	2.42	0.74	0.46	835	8
5	0.32	0.28	0.61	3.22	0.63	0.22	835	7
6	0.31	0.20	0.62	2.63	0.64	0.58	835	6/7

TABLE 2 — COMPARISON OF RELATIVE HARDENABILITIES AS ASSESSED BY SEVERAL METHODS

STEEL No.	CONTINUOUS-COOLING DIAGRAM. OIL-QUENCHED DIAMETER FOR		ISOTHERMAL DATA		OIL-QUENCHED DIA. FOR 50% MARTENSITE (IN.) CALCULATED FROM COMPOSITION USING FACTORS OF	
	100% martensite at bar axis, in.	50% bainite at bar axis, in.	Minimum bainite incubation period, sec.	Minimum time for formation of 50% bainite, sec.	Grossman	Crafts & Lamont
	1	1.5	2.6	22	90	4.40
2	2.3	3.6	20	105	4.20	4.5
3	3.0	4.4	22	120	6.00	5.0
4	3.6	4.6	22	135	4.50	5.4
5	3.8	5.8	22	230	3.40	5.0
6	4.4	5.0	40	210	7.25	7.5

established, it is suggested that satisfactory determinations of hardenability can only be made by measurements on continuously cooled samples using practical cooling schedules.

Summary

Recent work has provided evidence that the transformation of austenite to bainite occurs by a process of nucleation and continuous growth at all temperatures in the bainite formation range and that the acicular type of bainite formed at relatively low temperatures does not form by a very rapid shear mechanism of the type associated with martensite formation.

The presence of crystals of pro-pearlitic ferrite or upper bainite in an austenite matrix of a 3 per cent nickel-chromium steel pro-

vided favourable sites for the nucleation of the acicular type of lower bainite, but this did not significantly affect the isothermal incubation period required for lower bainite formation or the rate at which this reaction proceeded. Partial transformation to lower bainite followed by isothermal transformation at a temperature in the upper bainite range was found to reduce the incubation period at the latter temperature and to increase the rate of upper bainite formation. A similar effect on the isothermal transformation of austenite to bainite at temperatures above and below the martensite range has been observed when a proportion of martensite is formed in a steel prior to the isothermal treatment.

Prior formation of a small amount of pro-pearlitic ferrite reduced the incubation period

required for upper bainite formation in the 3 per cent nickel-chromium steel investigated, but the subsequent rate at which the upper bainite reaction proceeded was reduced.

The temperature at which martensite formation commences (M_s) in a rapidly cooled steel and the progress of this reaction with decreasing temperature can be calculated from the chemical composition of the steel. Several formulae which enable these calculations to be made have been presented.

Continuous-cooling transformation diagrams are more directly applicable to normal heat treatment practice than isothermal transformation diagrams. The general features of the continuous-cooling transformation diagrams of alloy steels have been discussed with reference to the particular diagrams obtained for several types of steel. Attention is drawn to the practical significance of the diagrams.

A comparison of the hardenabilities of a selection of steels obtained by several methods indicated that isothermal transformation diagrams were not sufficiently discriminating for assessing relative hardenabilities with respect to full hardening (100 per cent martensite). For hardening to 50 per cent martensite structures the relative hardenabilities obtained from isothermal diagrams were in broad agreement with continuous-cooling transformation data. The results obtained by calculating from chemical compositions bar diameters for the formation of 50 per cent martensite structures at the axes of oil-quenched bars were also at variance with the values derived from continuous-cooling transformation diagrams. In view of these discrepancies it is suggested that only the results of measurements made on continuously cooled samples using practical cooling schedules should be accepted as indicating the hardenabilities of steels. It is likely, however, that with continued

attention to both continuous-cooling and isothermal transformations a greater understanding of the relationships between the two processes will eventually be obtained, thus permitting broader application of the wealth of isothermal data which are now available.

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