

ASM HANDBOOK®

VOLUME 4 Heat Treating



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Foreword

Heat-treating technology has long been an area of deep interest and concern to ASM members. In fact, the origin of the Society can be traced back to 1913 when the Steel Treating Club was launched in Detroit. This group joined with the American Steel Treating Society to form the American Society for Steel Treating in 1920. It was the latter organization that issued the first bound Handbook in 1928, a volume that would serve as the prototype for future generations of the *ASM Handbook*.

During the ensuing six decades, many changes have taken place--both in terms of the positioning of the Society and the technology base it serves. In 1933 a name change to the American Society for Metals completed the transition from an organization concerned primarily with heat treating to one that served the interests of the entire metals industry. Finally in 1987, the technical scope of the Society was further broadened to include the processing, properties, and applications of all engineering/structural materials, and thus ASM International was born.

Despite these momentous changes, one fact has remained unchanged--ASM's recognition of heat treating as one of the foundations of the metals sciences and its unflagging commitment to this ever-changing technology. The publication of Volume 4 of the *ASM Handbook* is the most recent and significant example of the sustained leadership of the Society in addressing the needs of the heat treat community.

The present volume reflects the continuing research and effort that have led to a deeper understanding of the response of ferrous and nonferrous alloys to thermal treatments. For in the 10 years since publication of its 9th Edition predecessor, significant developments have taken place in quenching and hardenability studies, computer modelling of heat-treating operations, plasma-assisted case hardening methods, and improved quality control through advanced instrumentation and/or the application of statistical process control. These are but a few of the important topics that will undoubtedly contribute toward making the Heat Treating Handbook a timeless contribution to the literature.

Successful completion of such a formidable project, however, is dependent on the collective effort of a vast pool of knowledgeable and dedicated professionals. For their significant roles in this project, we are truly indebted to the ASM Heat Treating Technical Division and its subcommittees, to the Handbook Committee, to the hundreds of individual authors and reviewers, and the Handbook Editorial Staff. For their valuable contributions, we extend our thanks and gratitude.

- Stephen M. Copley
President
ASM International
- Edward L. Langer
Managing Director
ASM International

Preface

In compiling this new volume on heat treating, the challenge was to produce a book that contained subject matter strongly oriented toward industrial practice but that did not omit discussions of the underlying metallurgical fundamentals. With previously published ASM Handbooks devoted to heat treating, the omission of material on fundamentals was justified by either space limitations and/or the availability of other ASM books that described the physical metallurgy associated with thermal treatments. For example, when the 8th Edition was published in 1964, only 306 pages were related to heat treating (this Volume was divided between heat-treating technology and surface cleaning and finishing). As such, readers were referred to the classic book *Principles of Heat Treatment* by M.A. Grossmann and E.C. Bain, which was also published in 1964 by ASM. A similar situation arose in 1981 when the expanded 9th Edition Heat Treating Handbook was published. In the year prior to this publication, a completely revised version of the Grossmann/Bain book was prepared by G. Krauss and subsequently published by ASM.

The 1980s proved to be a dynamic period for heat-treating technology--a decade that witnessed the introduction of new alloys and processes as well as new "tools" for understanding the response of heat-treated materials. For example, new alloys under active development or brought to market during the 1980s that were not described in previous heat-treating Handbooks included duplex stainless steels, microalloyed (HSLA) steels, low-cobalt maraging steels, austempered ductile iron, directionally solidified and single-crystal superalloys, and aluminum-lithium alloys.

Changes in processing include improvements in continuous annealing, induction heating, and surface hardening operations using lasers or electron beams, the commercial viability of plasma-assisted case-hardening processes, and advances in thermomechanical processing.

But by far the most dramatic changes in heat-treat technology that have marked the past decade have been those involving newly developed tools for improving process characterization and process control. These include improved instrumentation for controlling furnace temperature, furnace atmosphere, and surface carbon content, the practical

application of statistical process control (SPC), and the use of computer modelling for both the prediction of hardness profiles after quenching and the quantitative modelling of properties after tempering or case hardening. It is this latter category of computer modelling that necessitates the inclusion of material on the basic principles or fundamentals of heat treating. For example, there are several articles in this Volume that deal with computer-assisted prediction of steel hardening and hardenability as a function of heat treatment parameters. In this regard, the primary measures of steel hardening are the end-quench hardenability curves (Jominy curves), isothermal transformation (IT) curves, and continuous cooling transformation (CCT) curves. In order to understand how computer programs can be used to calculate such diagrams, some brief background information is provided in several key articles to emphasize how these diagrams make possible the selection of steel and the design of proper heat treatments.

Principal Sections

Volume 4 has been organized into eight major sections:

- Heat Treating of Steel
- Surface Hardening of Steel
- Heat-Treating Equipment
- Process and Quality Control Considerations
- Heat Treating of Cast Irons
- Heat Treating of Tool Steels
- Heat Treating of Stainless Steels and Heat-Resistant Alloys
- Heat Treating of Nonferrous Alloys

A total of 71 articles are contained in these sections. Of these, 16 are new, 17 were completely rewritten, with the remaining articles revised and/or expanded. In addition, several important appendices supplement the Volume. These include a glossary of terms, a temper color chart for steels, and tabulated austenitizing temperatures for steels. A review of the content of the major sections is given below; highlighted are differences between the present volume and its 9th Edition predecessor. Table 1 summarizes the content of the principal sections.

Table 1 Summary of contents of Volume 4, Heat Treating, of the ASM Handbook

Section title	Number of articles	Pages	Figures ^(a)	Tables ^(b)	References
Heat Treating of Steel	16	253	355	123	430
Surface Hardening of Steel	18	203	305	69	324
Heat-Treating Equipment	6	62	83	17	43
Process and Quality Control Considerations	9	135	130	43	190
Heat Treating of Cast Irons	5	42	67	19	27
Heat Treating of Tool Steels	4	56	48	34	20
Heat Treating of Stainless Steels and Heat-Resistant Alloys	3	51	41	53	23
Heat Treating of Nonferrous Alloys	10	124	147	77	72

Total	71	926	1176	435	1129
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(a) Total number of figure captions; most figures include more than one illustration.

(b) Does not include in-text tables or tables that are part of figures

Heat Treating of Steel. This section begins with two entirely new articles that introduce the reader to the physical metallurgy of heat-treated steels and newly developed methodologies for quantitatively predicting transformation hardening in steels. These companion papers set the stage for a series of articles that describe specific types of heat treatments. Of particular note is the definitive treatise on "Quenching of Steel" by Bates, Totten, and Brennan. Featuring some 95 figures and 23 tables, this 55 page article has been substantially revised and expanded from previous Editions. Other highlights include new articles on continuous annealing, cryogenic treatment of steel, and thermomechanical processing of microalloyed steel. The section concludes with completely rewritten articles on heat-treat procedures for ultrahigh strength steels, maraging steels, and powder metallurgy ferrous alloys.

Surface Hardening of Steel. As explained in the introductory article to this section, emphasis has been placed on *thermally driven, diffusion processes* that induce solid-state transformation hardening. These processes include flame hardening, high-energy processes that utilize laser beams or electron beams, and conventional surface treatments such as carburizing, nitriding, and carbonitriding.

It is important to note the significant processing characteristics between the aforementioned processes and surface modification techniques 'such as ion implantation, PVD/CVD coatings, and surface melting/surface alloying processes that will be described in future volumes of this Handbook series. For example ion nitriding, which is described in this section, and nitrogen ion implantation are two distinctly different techniques for producing a case hardened surface layer. The implementation of each process, the characteristics of the case layers produced, the metallurgical strengthening mechanisms generated, and the economics and end use of each, are quite different.

Ion nitriding is a thermally driven, equilibrium, diffusion process that produces a relatively deep (100 to 400 μm), hardened, case layer. Nitrogen ion implantation is a non-thermal, non-equilibrium, physically driven, ballistic alloying process, which produces a relatively shallow (1 μm), extremely hard case layer. Ion nitriding is implemented at high temperatures in a glow discharge atmosphere, while nitrogen ion implantation is carried out at room temperature, at high vacuum, in a dedicated atomic particle accelerator. Case layer strengthening in ion nitrided surfaces is due primarily to formation of transition metal nitride precipitates, while strengthening in nitrogen ion implanted surfaces is due primarily to dislocation pinning. A summary of processing comparisons is given in Table 2.

Table 2 Process characteristics comparison

Process	Type	Process time, h	Process temperature		Process pressure, torr	Case depth ^(a) , μm	Hardness ^(a) , HRC
			°C	°F			
Ion nitriding	Thermal diffusion	10-30	500-1100	900-2000	0.2-5.0	400	62-67
Nitrogen ion	Physical	1-6	<150	<300	10 ⁻⁶	1	80-90

(a) Value for steel

Key additions to this section include articles that describe increasingly used processes such as plasma-assisted case hardening methods, boriding, and the Toyota diffusion process. Of critical importance to this section is the article "Microstructures and Properties of Carburized Steels" by G. Krauss which examines the correlation between processing, structure, and resulting fatigue, fracture, and wear properties of case-hardened steels.

Heat-Treating Equipment. Types of heat-treating furnaces, the materials used to construct furnaces, and the advantages and limitations associated with each are described next. More emphasis has been placed on furnace energy efficiency and proper design than in previous Editions.

Process and quality control considerations are more important than ever to heat treaters. Reliable sensors, computerized control equipment, and process control of heating and cooling and furnace atmospheres are described in detail in this section. Supplementing this material are new articles on the recognition and prevention of defects in heat-treated parts and the use of computer programs for designing heat-treat operations and predicting the properties of heat-treated steels.

Because the heat-treating industry is being challenged to introduce statistical concepts in order to minimize variability and ensure consistent quality of heat-treated parts, an important article on "Statistical Process Control of Heat-Treating Operations" is also included. Emphasis is on the practical application of SPC concepts in order to demonstrate to heat treaters how to identify critical process parameters that influence product quality and how to establish methods to monitor and evaluate such parameters.

Heat treating of cast irons is described in five articles. The "Introduction to Heat Treating of Cast Irons" was completely rewritten for this Volume. The remaining four articles contain new information on austempering of ductile iron and procedures for heat treating highly alloyed abrasion-resistant, corrosion-resistant, and heat-resistant cast irons.

Heat Treating of Tool Steels. Because tool steels must be processed to develop specific combinations of wear resistance, resistance to deformation or fracture under high loads, and resistance to softening under elevated temperatures, proper heat treating is critical. This section describes the procedures and equipment necessary to meet these criteria.

Heat Treating of Stainless Steels and Heat-Resistant Alloys. Procedures and process control for heat treating the principal types of stainless steels and superalloys are discussed in this section. The article on "Heat Treating of Superalloys" was completely rewritten for this Volume and includes information on both wrought and cast alloys, many of which are used in the aerospace industry. The article on refractory metals and alloys is completely new to the Handbook series.

Heat Treating of Nonferrous Alloys. The principles which govern heat treatment of nonferrous alloys are first described in this final section of the Handbook. Differences between ferrous and nonferrous processing are highlighted. Nine articles follow on heat treating of specific classes of nonferrous alloys.

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Principles of Heat Treating of Steels

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Introduction

A STEEL is usually defined as an alloy of iron and carbon with the carbon content between a few hundredths of a percent up to about 2 wt%. Other alloying elements can amount in total to about 5 wt% in low-alloy steels and higher in more highly alloyed steels such as tool steels and stainless steels. Steels can exhibit a wide variety of properties depending on composition as well as the phases and microconstituents present, which in turn depend on the heat treatment. In subsequent articles in this Section, various types of heat treatment are described in detail. In this article, an outline of the physical metallurgy associated with heat treating of steels will be given and some important "tools" will be introduced. Deliberately no microstructures are shown as a well-illustrated article exists in "Microstructures, Processing, and Properties of Steels," by G. Krauss in *Properties and Selection: Irons, Steels, and High-Performance Alloys*, Volume 1 of *ASM Handbook*. A companion article that emphasizes information systems for predicting microstructures and hardnesses of quenched steels follows (see the article "Quantitative Prediction of Transformation Hardening in Steels" in this Volume).

The Fe-C Phase Diagram

The basis for the understanding of the heat treatment of steels is the Fe-C phase diagram (Fig. 1). Because it is well explained in earlier volumes of *ASM Handbook*, formerly *Metals Handbook* (Ref 1, 2, 3), and in many elementary textbooks, it will be treated very briefly here. Figure 1 actually shows two diagrams; the stable iron-graphite diagram (dashed lines) and the metastable Fe-Fe₃C diagram. The stable condition usually takes a very long time to develop, especially in the low-temperature and low-carbon range, and therefore the metastable diagram is of more interest. The Fe-C diagram shows which phases are to be expected at equilibrium (or metastable equilibrium) for different combinations of carbon concentration and temperature. Table 1 provides a summary of important metallurgical phases and microconstituents. We distinguish at the low-carbon end ferrite (α -iron), which can at most dissolve 0.028 wt% C at 727 °C (1341 °F) and austenite (γ -iron), which can dissolve 2.11 wt% C at 1148 °C (2098 °F). At the carbon-rich side we find cementite (Fe₃C). Of less interest, except for highly alloyed steels, is the δ -ferrite existing at the highest temperatures. Between the single-phase fields are found regions with mixtures of two phases, such as ferrite + cementite, austenite + cementite, and ferrite + austenite. At the highest temperatures, the liquid phase field can be found and below this are the two phase fields liquid + austenite, liquid + cementite, and liquid + δ -ferrite. In heat treating of steels, the liquid phase is

always avoided. Some important boundaries at single-phase fields have been given special names that facilitate the discussion. These include:

- A_1 , the so-called eutectoid temperature, which is the minimum temperature for austenite
- A_3 , the lower-temperature boundary of the austenite region at low carbon contents, that is, the $\gamma/\gamma + \alpha$ boundary
- A_{cm} , the counterpart boundary for high carbon contents, that is, the $\gamma/\gamma + Fe_3C$ boundary

Sometimes the letters c, e, or r are included. Relevant definitions of terms associated with phase transformations of steels can be found in Table 2 as well as the Glossary of Terms in this Volume and Ref 3. The carbon content at which the minimum austenite temperature is attained is called the eutectoid carbon content (0.77 wt% C). The ferrite-cementite phase mixture of this composition formed during cooling has a characteristic appearance and is called pearlite and can be treated as a microstructural entity or microconstituent. It is an aggregate of alternating ferrite and cementite lamellae that degenerates ("spheroidizes" or "coarsens") into cementite particles dispersed with a ferrite matrix after extended holding close to A_1 .

Table 1 Important metallurgical phases and microconstituents

Phase (microconstituent)	Crystal structure of phases	Characteristics
Ferrite (α -iron)	bcc	Relatively soft low-temperature phase; stable equilibrium phase
δ -ferrite (δ -iron)	bcc	Isomorphous with α -iron; high-temperature phase; stable equilibrium phase
Austenite (γ -iron)	fcc	Relatively soft medium-temperature phase; stable equilibrium phase
Cementite (Fe_3C)	Complex orthorhombic	Hard metastable phase
Graphite	Hexagonal	Stable equilibrium phase
Pearlite		Metastable microconstituent; lamellar mixture of ferrite and cementite
Martensite	bct (supersaturated solution of carbon in ferrite)	Hard metastable phase; lath morphology when <0.6 wt% C; plate morphology when >1.0 wt% C and mixture of those in between
Bainite	...	Hard metastable microconstituent; nonlamellar mixture of ferrite and cementite on an extremely fine scale; upper bainite formed at higher temperatures has a feathery appearance; lower bainite formed at lower temperatures has an acicular appearance. The hardness of bainite increases with decreasing temperature of formation.

Table 2 Definitions of transformation temperatures in iron and steels

See the Glossary of Terms in this Volume for additional terminology.

Transformation temperature. The temperature at which a change in phase occurs. The term is sometimes used to denote the limiting temperature of a transformation range. The following symbols are used for iron and steels.

A_cm . In hypereutectoid steel, the temperature at which the solution of cementite in austenite is completed during heating.
A_c1 . The temperature at which austenite begins to form during heating, with the c being derived from the French <i>chauffant</i> .
A_c3 . The temperature at which transformation of ferrite to austenite is completed during heating.
A_em, A_e1, A_e3 . The temperatures of phase changes at equilibrium.
A_rm . In hypereutectoid steel, the temperature at which precipitation of cementite starts during cooling, with the r being derived from the French <i>refroidissant</i> .
A_r1 . The temperature at which transformation of austenite to ferrite or to ferrite plus cementite is completed during cooling.
A_r3 . The temperature at which austenite begins to transform to ferrite during cooling.
A_r4 . The temperature at which delta ferrite transforms to austenite during cooling.
M_s (or A_r'') . The temperature at which transformation of austenite to martensite starts during cooling.
M_f . The temperature at which martensite formation finishes during cooling.

Note: All of these changes, except the formation of martensite, occur at lower temperatures during cooling than during heating and depend on the rate of change of temperature.

Source: Ref 2

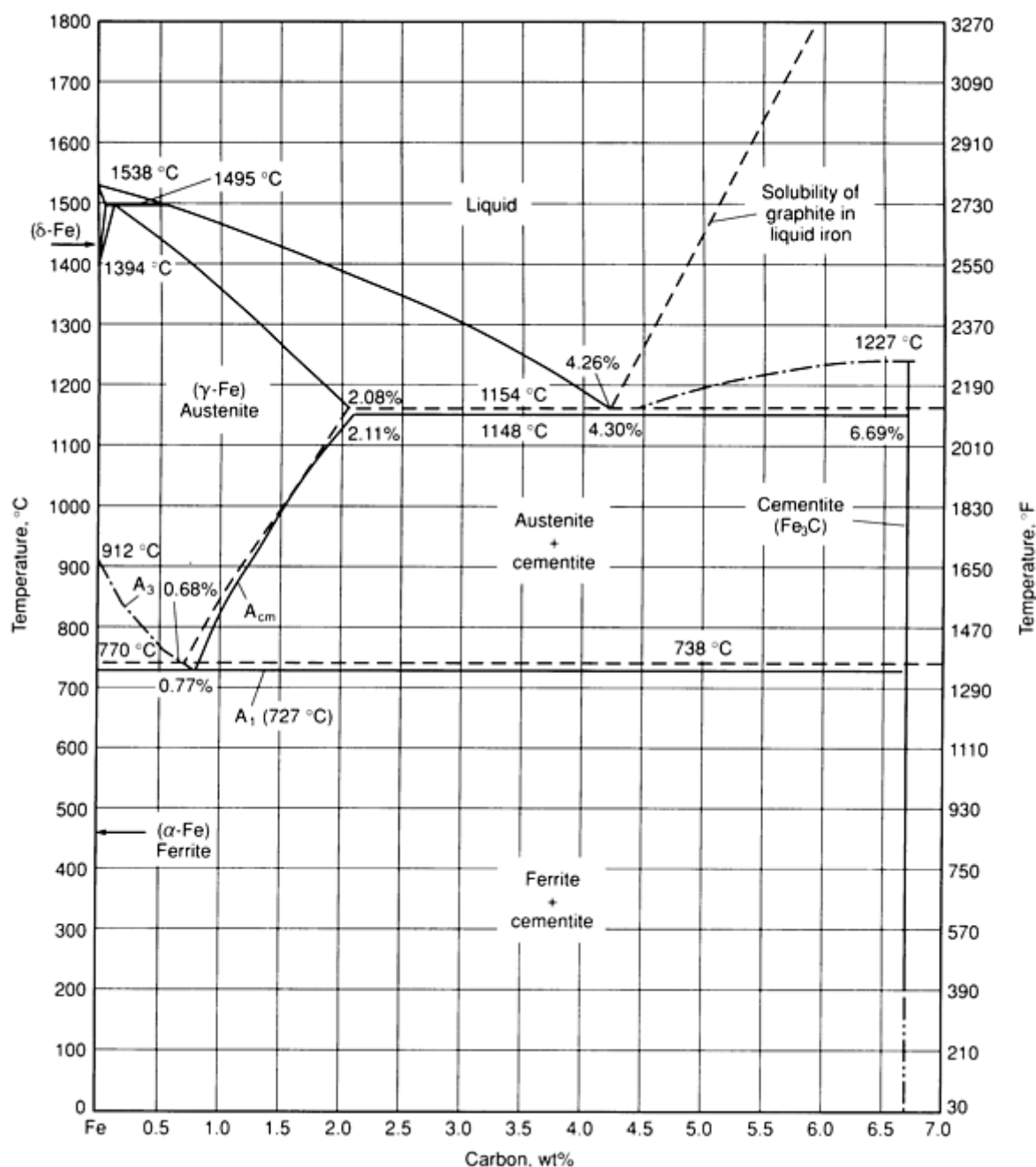


Fig. 1 The Fe-C equilibrium diagram up to 6.67 wt% C. Solid lines indicate Fe-Fe₃C diagram; dashed lines indicate iron-graphite diagram. Source: Ref 1

The Fe-C diagram in Fig. 1 is of experimental origin. The knowledge of the thermodynamic principles and modern thermodynamic data now permits very accurate calculations of this diagram (Ref 4). This is particularly useful when phase boundaries must be extrapolated and at low temperatures where the experimental equilibria are extremely slow to develop.

If alloying elements are added to the iron-carbon alloy (steel), the position of the A₁, A₃, and A_{cm} boundaries and the eutectoid composition are changed. Classical diagrams introduced by Bain (Ref 5) show the variation of A₁ and the eutectoid carbon content with increasing amount of a selected number of alloying elements (Fig. 2). It suffices here to mention that (1) all important alloying elements decrease the eutectoid carbon content, (2) the austenite-stabilizing elements manganese and nickel decrease A₁, and (3) the ferrite-stabilizing elements chromium, silicon, molybdenum, and tungsten increase A₁. These classifications relate directly to the synergisms in quench hardening as described in the articles "Quantitative Prediction of Transformation Hardening in Steels" and "Quenching of Steel".

Modern thermodynamic calculations allow accurate determinations of these shifts that affect the driving force for phase transformation (see below). These methods also permit calculation of complete ternary and higher-order phase diagrams including alloy carbides (Ref 6). Reference should be made to the Calphad computer system (Ref 7).

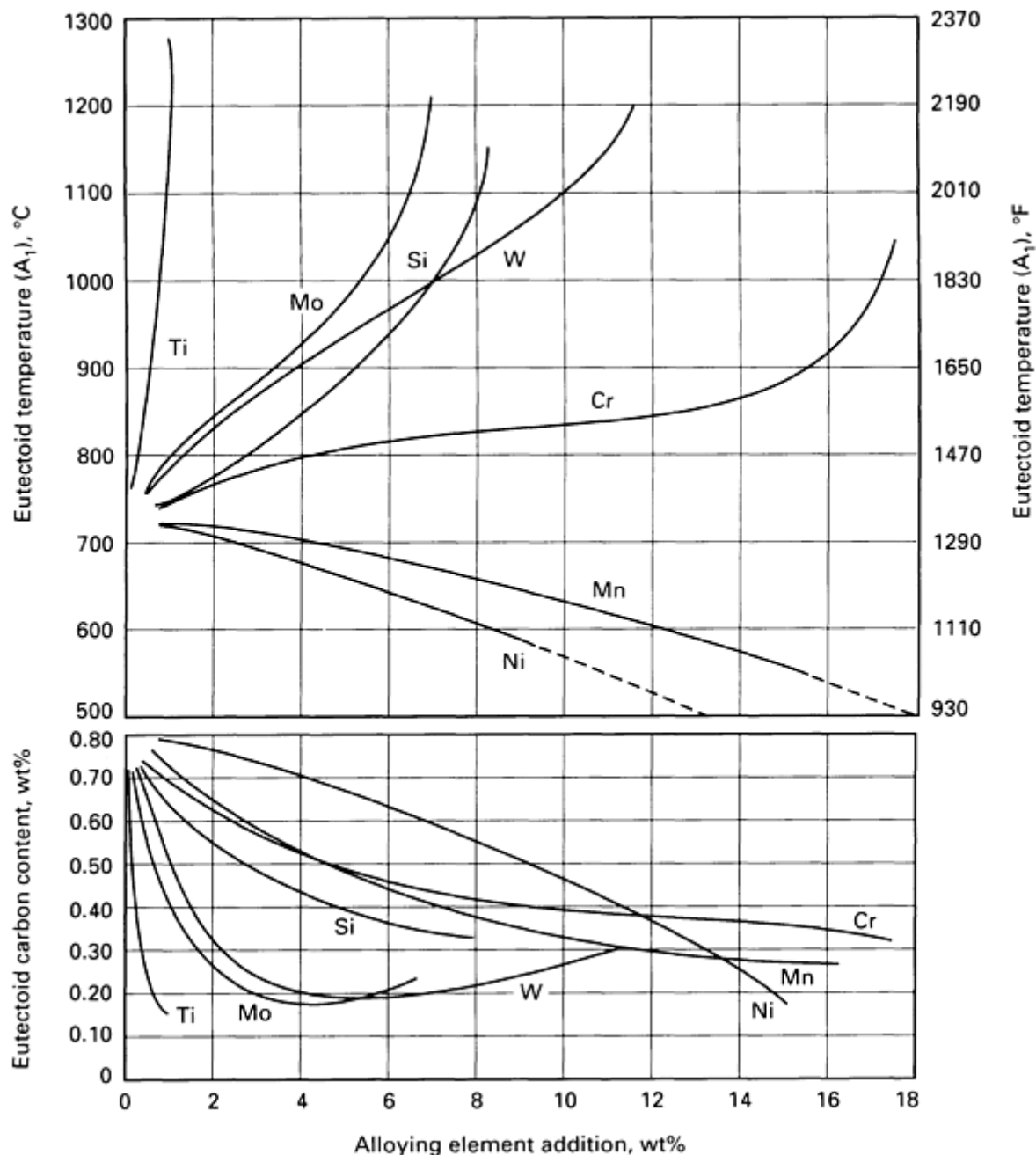


Fig. 2 Influence of alloying element additions on eutectoid temperature and eutectoid carbon content. Source: Ref 5

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Transformation Diagrams

The kinetic aspects of phase transformations are as important as the equilibrium diagrams for the heat treatment of steels. The metastable phase martensite and the morphologically metastable microconstituent bainite, which are of extreme importance to the properties of steels, can generally form with comparatively rapid cooling to ambient temperature, that is, when the diffusion of carbon and alloying elements is suppressed or limited to a very short range. Bainite is a eutectoid decomposition that is a mixture of ferrite and cementite. Martensite, the hardest constituent, forms during severe quenches from supersaturated austenite by a shear transformation. Its hardness increases monotonically with carbon content up to about 0.7 wt%. If these unstable metastable products are subsequently heated to a moderately elevated temperature, they decompose to more stable distributions of ferrite and carbide. The reheating process is sometimes known as tempering or annealing.

The transformation of an ambient temperature structure like ferrite-pearlite or tempered martensite to the elevated-temperature structure of austenite or austenite + carbide is also of importance in the heat treatment of steel.

One can conveniently describe what is happening during transformation with transformation diagrams. Four different types of such diagrams can be distinguished. These include:

- Isothermal transformation diagrams describing the *formation* of austenite, which will be referred to as ITh diagrams
- Isothermal transformation (IT) diagrams, also referred to as time-temperature-transformation (TTT) diagrams, describing the *decomposition* of austenite
- Continuous heating transformation (CHT) diagrams
- Continuous cooling transformation (CCT) diagrams

Isothermal Transformation Diagrams

This type of diagram shows what happens when a steel is held at a constant temperature for a prolonged period. The development of the microstructure with time can be followed by holding small specimens in a lead or salt bath and quenching them one at a time after increasing holding times and measuring the amount of phases formed in the microstructure with the aid of a microscope. An alternative method involves using a single specimen and a dilatometer which records the elongation of the specimen as a function of time. The basis for the dilatometer method is that the microconstituents undergo different volumetric changes (Table 3). A thorough description of the dilatometric method can be found in Ref 8.

Table 3 Volume changes due to different transformations

Transformation	Volume change, % ^(a)
Spheroidized pearlite-austenite	4.64-2.21 × (%C)
Austenite-martensite	4.64-0.53 × (%C)
Austenite-lower bainite	4.64-1.43 × (%C)
Austenite-upper bainite	4.64-2.21 × (%C)

(a) Linear changes are approximately one-third the volume changes.

ITh Diagrams (Formation of Austenite). During the formation of austenite from an original microstructure of ferrite and pearlite or tempered martensite, the volume (and hence the length) decreases with the formation of the dense austenite phase (see Fig. 3). From the elongation curves, the start and finish times for austenite formation, usually defined as 1% and 99% transformation, respectively, can be derived. These times are then conveniently plotted on a temperature-log time diagram (Fig. 4). Also plotted in this diagram are the Ac_1 and Ac_3 temperatures. Below Ac_1 no austenite can form, and between Ac_1 and Ac_3 the end product is a mixture of ferrite and austenite. Notice that a considerable overheating is required to complete the transformation in a short time. The original microstructure also plays a great role. A finely distributed structure like tempered martensite is more rapidly transformed to austenite than, for instance, a ferritic-pearlitic structure. This is particularly true for alloyed steels with carbide-forming alloying elements such as chromium and molybdenum. It is important that the heating rate to the hold temperature be very high if a true isothermal diagram is to be obtained.

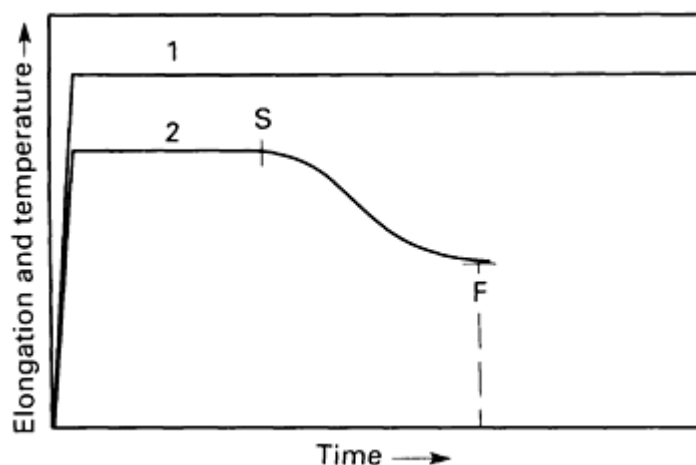


Fig. 3 The procedure for determining isothermal heating (ITh) diagrams. Line 1: Temperature versus time. Line 2: Elongation versus time. S represents the start and F the finish of the transformation of the original microstructure to austenite transformation, respectively.

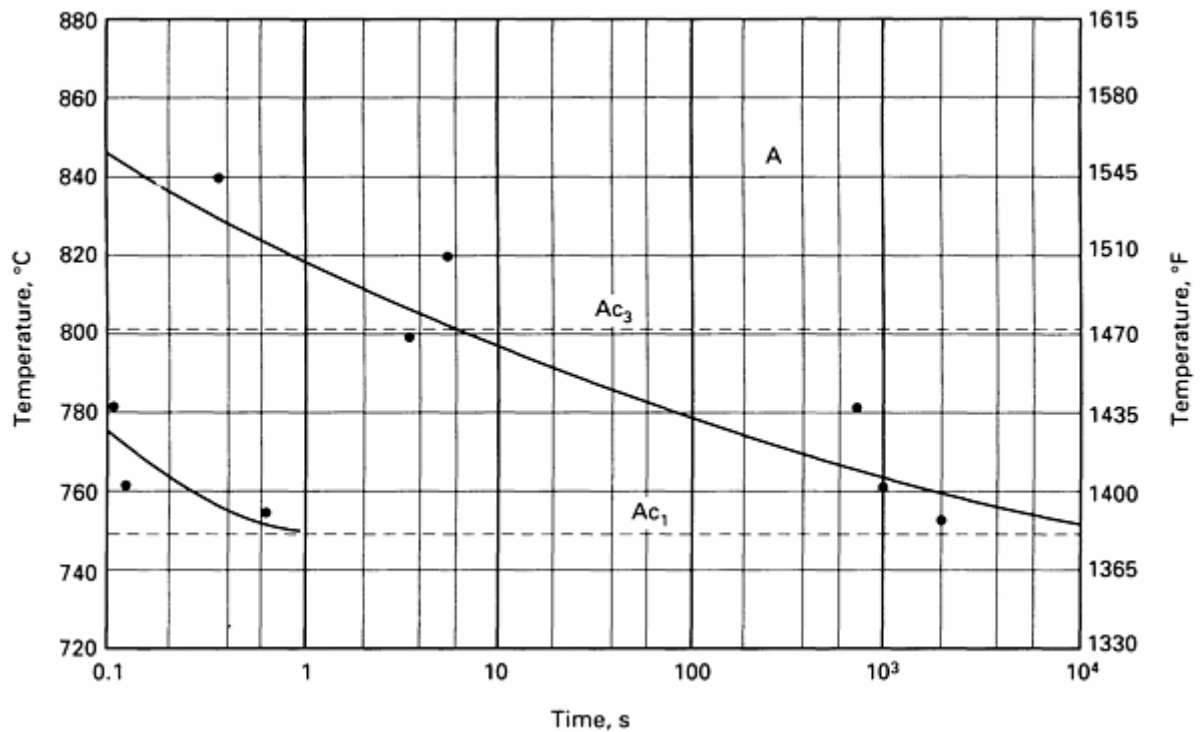


Fig. 4 Isothermal heating diagram for AISI 4140 steel. Heating rate to reach holding temperature is 1020 °C/s (1835 °F/s). Between Ac₃ and Ac₁, the final structure is a mixture of austenite and ferrite. A, austenite. Source: Ref 9

The ITh diagram is not as common as the isothermal cooling diagrams described next. Heating diagrams are useful in short time heat treatments like induction and laser hardening. An extensive collection of ITh diagrams can be found in Ref 10.

IT Diagrams (Decomposition of Austenite). The procedure starts at a high temperature, normally in the austenitic range after holding there long enough to obtain homogeneous austenite without undissolved carbides, followed by rapid cooling to the desired hold temperature (Fig. 5). An example of an IT diagram is given in Fig. 6. The cooling was started from 850 °C (1560 °F). The A₁ and A₃ temperatures are indicated as well as the hardness. Above A₃ no transformation can occur. Between A₁ and A₃ only ferrite can form from austenite. In Fig. 6, a series of isovolume fraction curves are shown; normally only the 1% and 99% curves are reproduced. Notice that the curves are C-shaped. This is typical for transformation curves. A higher-temperature set of C-shaped curves shows the transformation to pearlite and a lower-temperature set indicates the transformation to bainite. In between is found a so-called austenite bay, common for certain low-alloy steels containing appreciable amounts of carbide-forming alloying elements such as chromium or molybdenum.

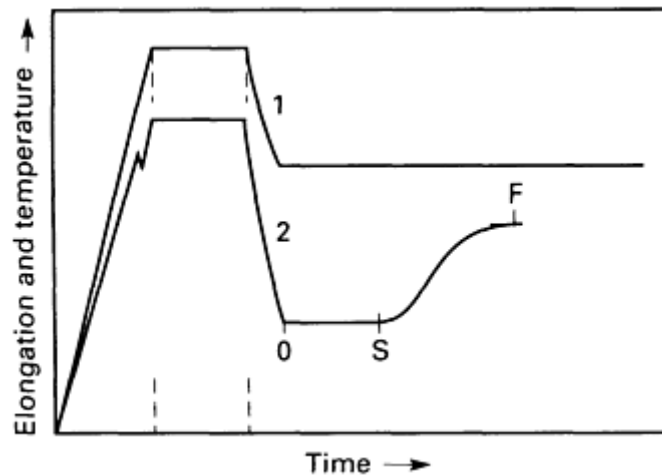


Fig. 5 The procedure for determining isothermal cooling (IT) diagrams. Line 1: Temperature versus time. Line 2: Elongation versus time. S represents the start and F the finish of austenite decomposition, respectively.

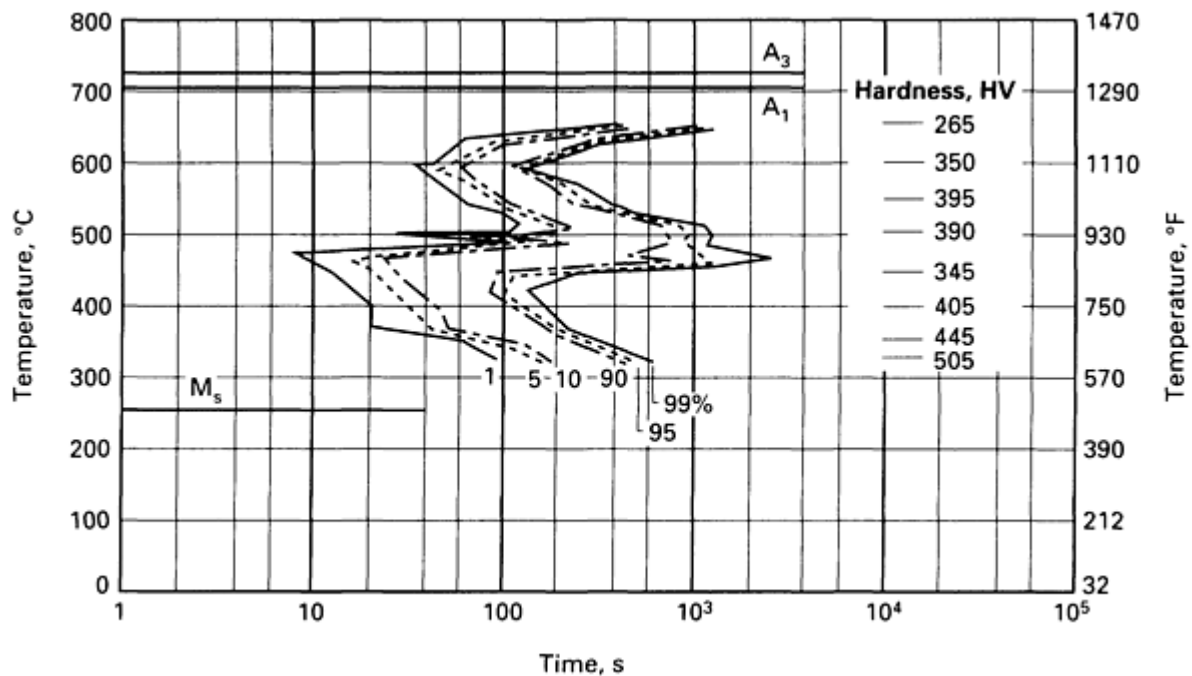


Fig. 6 Isothermal transformation diagram for a steel with 0.39% C, 0.86% Mn, 0.72% Cr, and 0.97% Ni. The upper C-shaped curves describe transformation to pearlite; the lower C-shaped curves to bainite. Ferrite is not visible. The column on the right side of the figure indicates the hardness after completed transformation measured at room temperature. Source: Ref 11

The transformation start curve shows as an upper-limiting estimate of the time (in seconds) for nucleation, τ . The isovolume fraction curves τ_x can semiempirically be described by a relation of the type (Ref 12):

$$t_x = \frac{\exp(Q/RT)}{2^{N/8} \cdot \Delta T^3} f \cdot I(X) \quad (\text{Eq 1})$$

where x is the volume fraction of the transformed phase, Q is an activation energy related to the boundary diffusion activation energies for the alloying elements, N is the ASTM grain size number for austenite, T is the temperature (in degrees Kelvin), ΔT is the undercooling ($A_3 - T$) for ferrite, ($A_1 - T$) for pearlite and an empirical value for bainite, f is a

linear function of the volume fractions of carbon and alloying elements and I is the volume fraction integral giving the dependence of the transformed phase on the volume fraction.

The combined effect of the $I/\Delta T^3$ factor, which increases with decreasing undercooling (that is, increasing temperature) and the $\exp(Q/RT)$ factor, which increases with decreasing temperature, results in long nucleation times, τ_x , for high and low temperatures and short nucleation times for intermediate temperatures. The C-shape can thus be obtained and understood. The factor $2^{N/8}$ is included to take into consideration the fact that the transformation rate is larger for smaller austenite grain sizes.

Figure 6 also indicates the martensite start temperature, M_s , which is time independent. A number of numerical expressions for the composition dependence of M_s have been suggested (Ref 13). One commonly used expression is (Ref 14):

$$M_s(^{\circ}\text{C}) = 512 - 453C - 16.9\text{Ni} + 15\text{Cr} - 9.5\text{Mo} + 217(C)^2 - 71.5(C)(\text{Mn}) - 67.6(C)(\text{Cr}) \quad (\text{Eq 2})$$

where the alloying elements are represented by their weight percents. If the hold time in the austenitic range is too short the C-shaped cooling curves may be shifted to shorter times. This is the case in Fig. 7 where an IT diagram obtained after 6 s at 950 °C (1740 °F) is compared with an IT diagram after 10 min at 860 °C (1580 °F). The shift may be due to incomplete carbide dissolution or a smaller grain size after short austenitizing time.

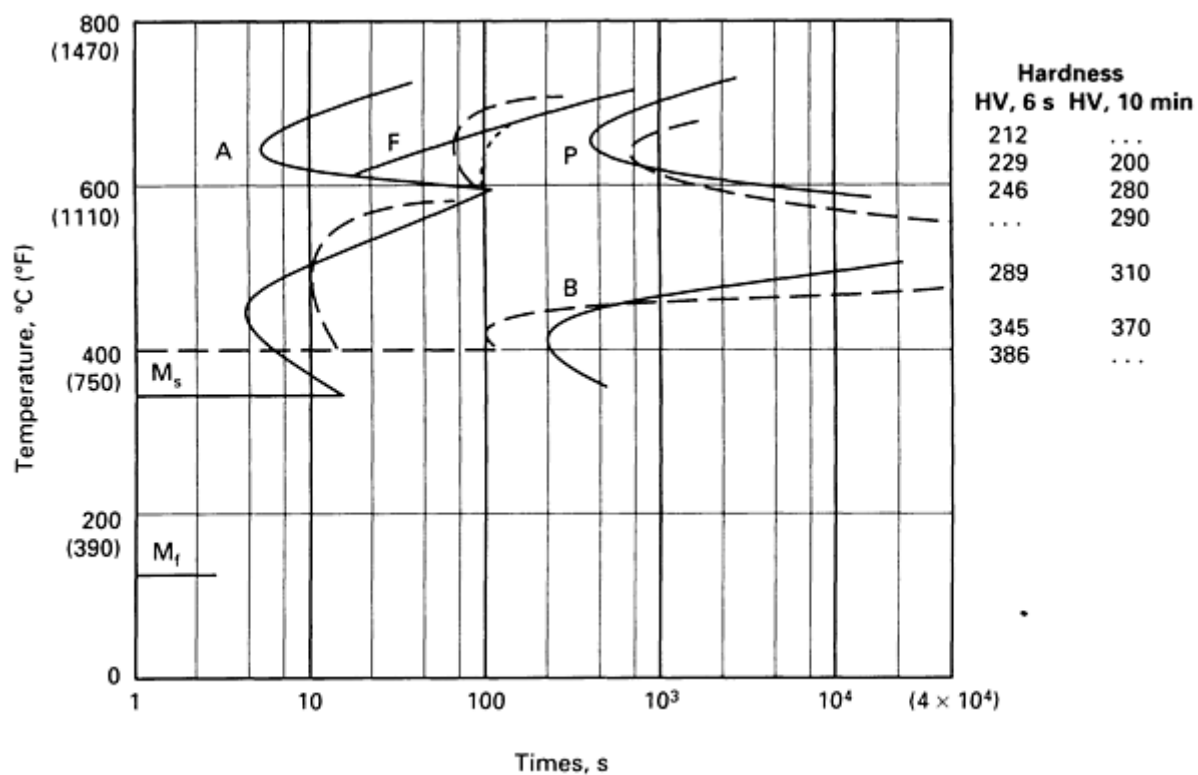


Fig. 7 Comparison of IT cooling diagrams for AISI 4140 steel after 6 s austenitizing time at 950 °C (1740 °F) (solid lines) and after 10 min austenitizing time at 860 °C (1580 °F) (dashed lines). Hardness values are also given for both heat-treated conditions. Source: Ref 9

Isothermal transformation curves can be found in standard graphs obtainable from the International Organization for Standardization (ISO), the Metallurgical Society of AIME, and ASM International (Ref 15). There is also a well-known German collection (Ref 10).

CHT Diagrams

In practical heat treatment situations, a constant temperature is not required, but rather a continuous changing temperature during either cooling or heating. Therefore, more directly applicable information is obtained if the diagram is constructed from dilatometric data using a continuously increasing or decreasing temperature. Figure 8 shows an example for continuous heating of the same steel as in Fig. 4. Indicated again are Ac_1 and Ac_3 and the same remarks as before are valid. The diagram was derived with constant heating rates and curves for 130 °C/s (265 °F/s), 10 °C/s (50 °F/s), and 0.2 °C/s (32 °F/s) are shown. Notice that the start and finish of the transformations are delayed relative to the isothermal diagram. This is generally true when one compares an isothermal and a continuous diagram, regardless of whether they are for heating or cooling.

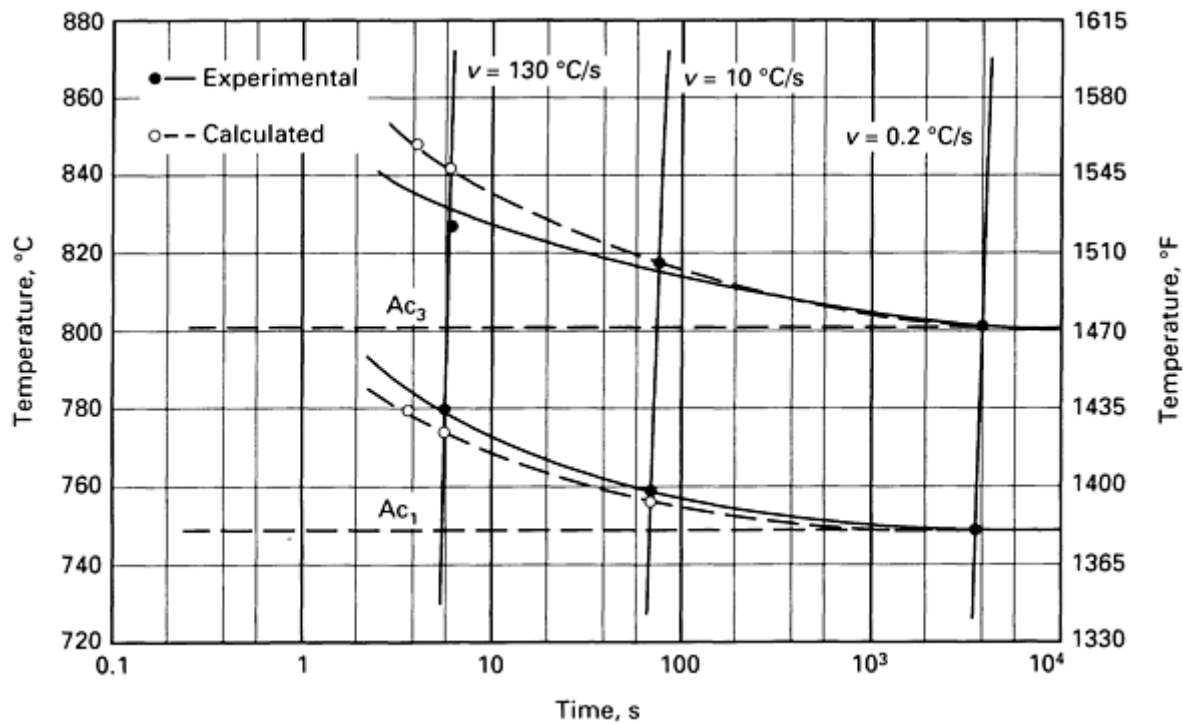


Fig. 8 Continuous heating transformation diagram for AISI 4140 steel. The phase being formed is austenite. Source: Ref 9

Like the ITh diagrams, the CHT diagrams are useful in predicting the effect of short-time austenitization that occurs in induction and laser hardening. One typical question is how high the maximum surface temperature should be in order to achieve complete austenitization for a given heating rate. Too high a temperature may cause unwanted austenite grain growth, which produces a more brittle martensitic microstructure. Reference 10 provides information on austenite grain growth as well as CHT diagrams.

CCT Diagrams

As for heating diagrams, it is important to clearly state what type of cooling curve the transformation diagram was derived from. Use of a constant cooling rate is very common in experimental practice. However, this regime rarely occurs in a practical situation. One can also find curves for so-called natural cooling rates according to Newton's law of cooling. These curves simulate the behavior in the interior of a large part such as the cooling rate of a Jominy bar at some distance from the quenched end. Close to the surface the characteristics of the cooling rate can be very complex as will be described below. In the lower part of Fig. 9 is shown a CCT diagram (fully drawn lines) for 4130 steel. Ferrite, pearlite, and bainite regions are indicated as well as the M_s temperature. Note that the M_s temperature is not constant when martensite formation is preceded by bainite formation, but typically decreases with longer times.

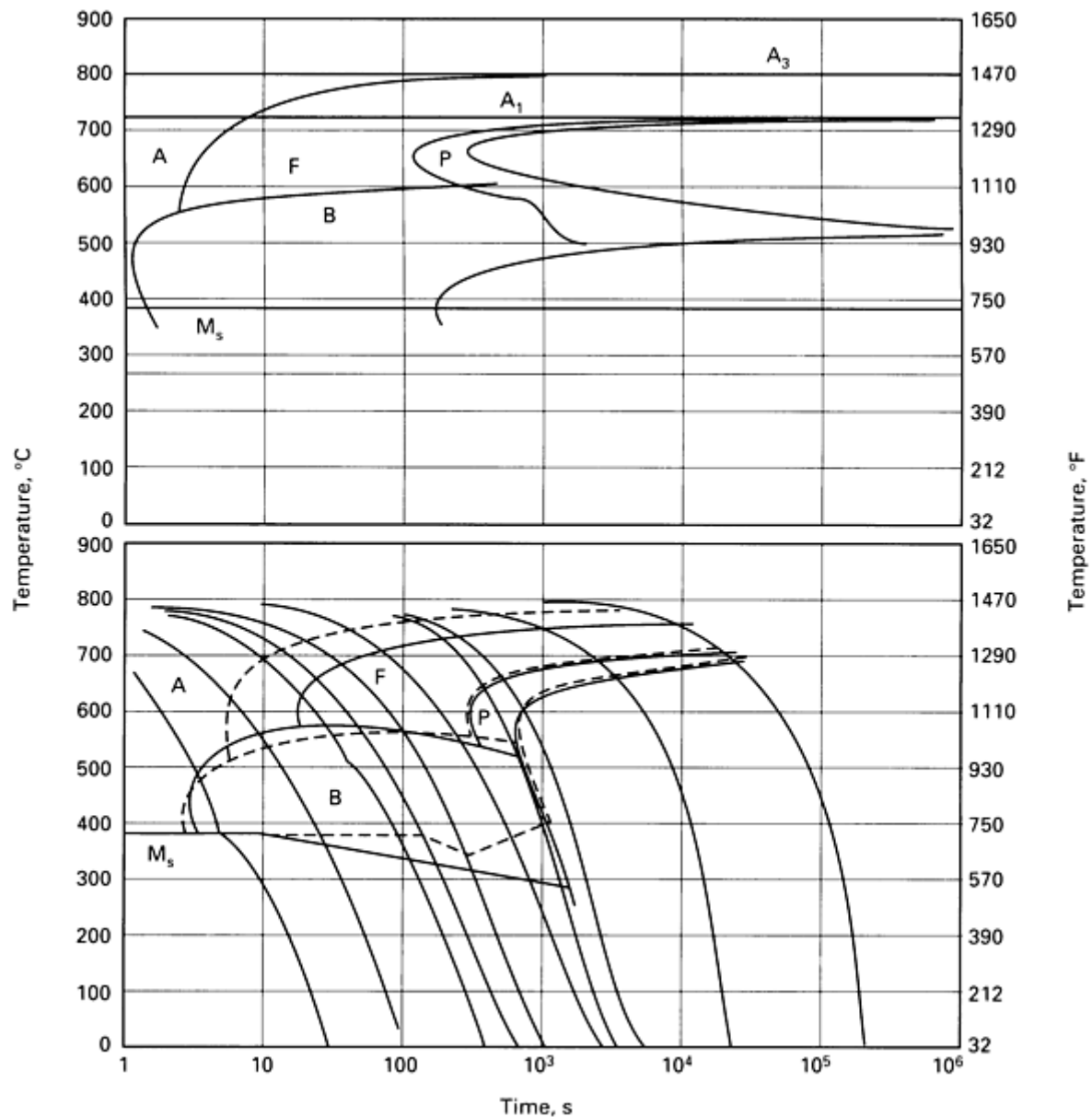


Fig. 9 Isothermal transformation (upper) and CCT (lower) diagrams for AISI 4130 steel containing 0.30% C, 0.64% Mn, 1.0% Cr, and 0.24% Mo. The IT diagram illustrates the input data representation for calculations described in the text. The CCT diagrams are computed (dashed lines) and experimentally determined (solid lines). Source: Ref 10, 11

The effect of different cooling curves is shown in Fig. 10. Each CCT diagram contains a family of curves representing the cooling rates at different depths of a cylinder with a 300 mm (12 in.) diameter. The slowest cooling rate represents the center of the cylinder. As shown in Fig. 10, the rate of cooling and the position of the CCT curves depend on the cooling medium (water produced the highest cooling rate followed by oil and air, respectively). The more severe the cooling medium, the longer the times to which the C-shaped curves are shifted. The M_s temperature is unaffected.

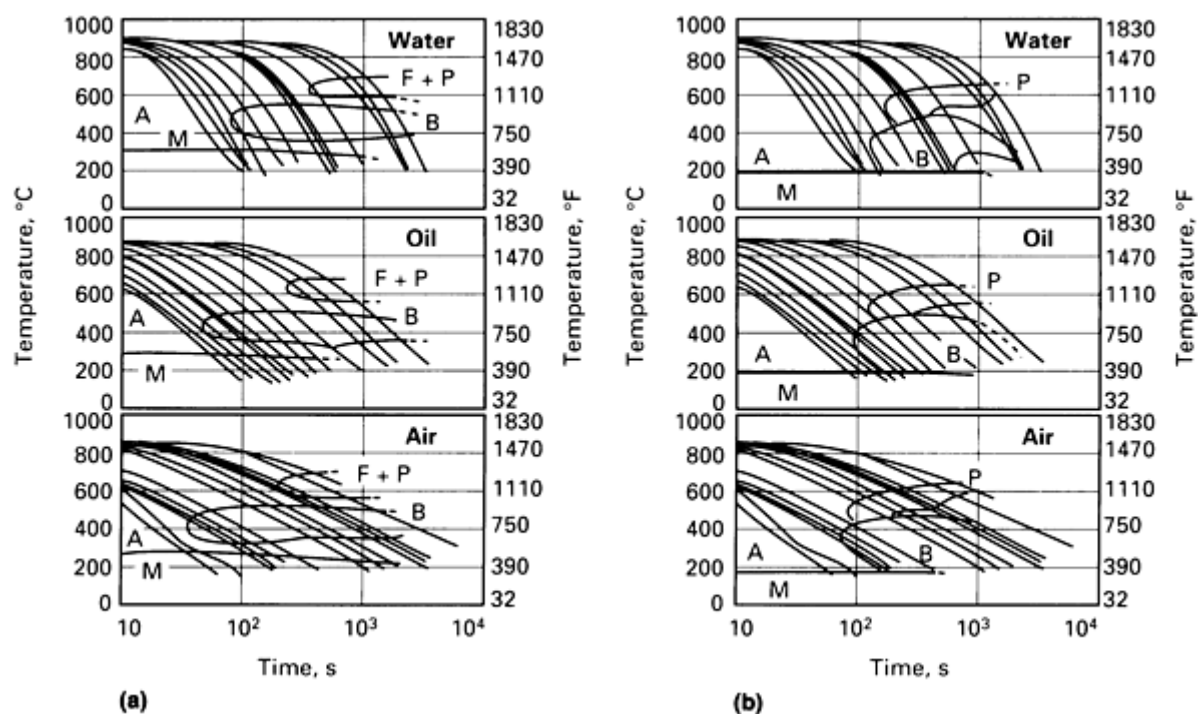


Fig. 10 Examples of CCT diagrams for low-alloy steels. (a) CCT diagrams of a chromium-molybdenum steel using simulated cooling curves for water, oil, and air. Source: Ref 16. (b) Computer-calculated CCT diagrams of a nickel-chromium steel containing 0.77 wt% C based on the cooling curves for water, oil, and air given in (a). Source: Ref 17

It should be noted, however, that transformation diagrams cannot be used to predict the response to thermal histories that are very much different from the ones used to construct the diagrams. For instance, first cooling rapidly to slightly above M_s and then reheating to a higher temperature will give more rapid transformation than shown in the IT diagram because nucleation is greatly accelerated during the introductory quench. It should also be remembered that the transformation diagrams are sensitive to the exact alloying content within the allowable composition range.

Various attempts have been made to facilitate the use of CCT diagrams by including the effect of the dimension of the workpiece and cooling medium on the microstructure and hardness obtained. Two such approaches have been described by Atkins (Ref 18) and Thelning (Ref 19). Experimental cooling curves were derived by Atkins for round bars of various diameters at fractional radii (R) 0.5 and 0.8 from the center during cooling in water, oil, and air. The cooling curves were used in dilatometer measurements to get the transformation temperatures and the resulting microstructures and hardnesses were also noted. A diagram of this type is shown in Fig. 11. From the diagram, one can correlate that a specimen, 2 mm (0.08 in.) in diameter being cooled in air, corresponds to a bar somewhat less than 40 mm (1.6 in.) in diameter cooled in oil, or to a bar 50 mm (2 in.) in diameter quenched in water. Shown in the upper part of the figure are the amount and type of microstructure; the lower part gives hardnesses after hardening and tempering at various temperatures.

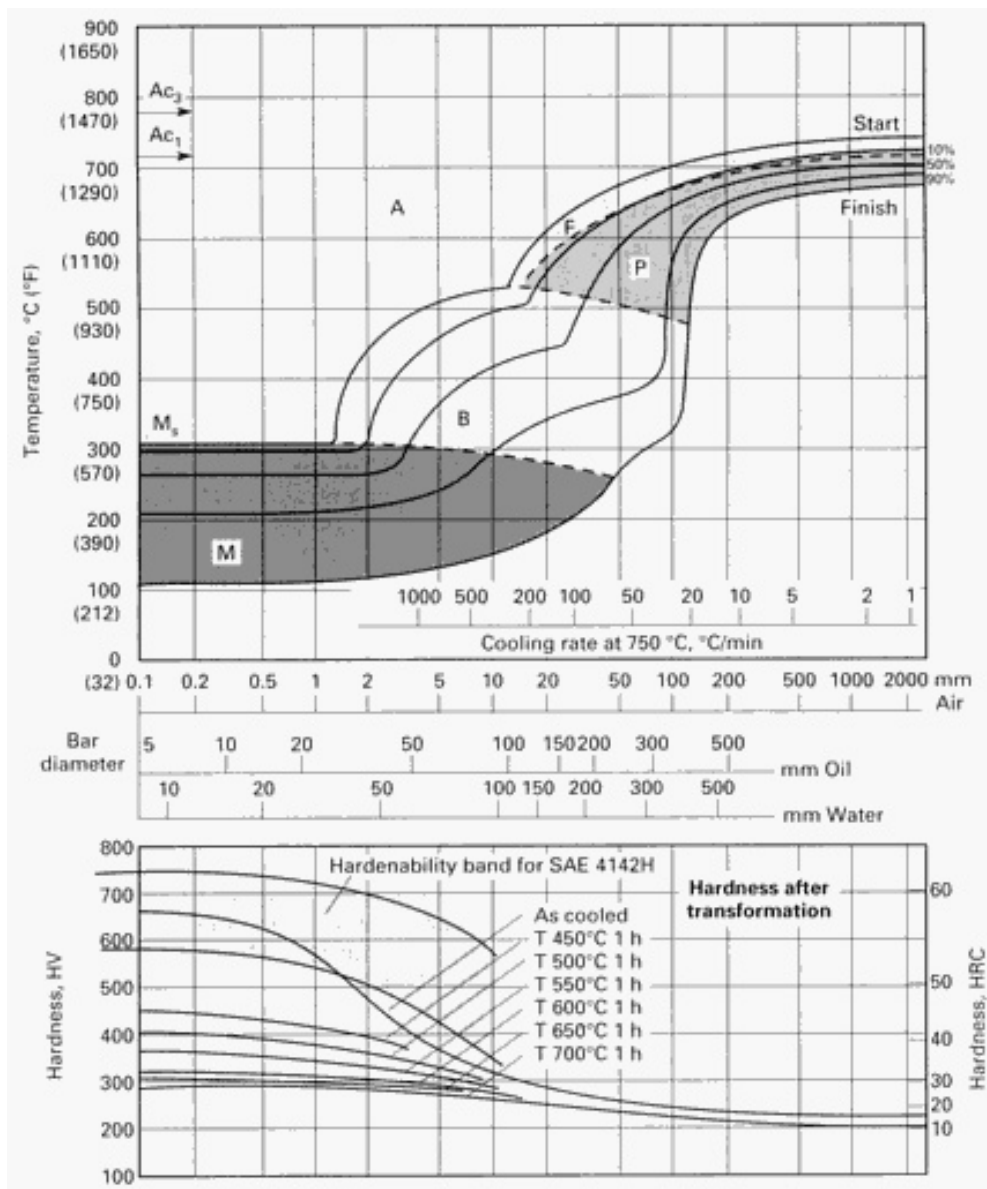


Fig. 11 A CCT diagram for 1.25Cr-0.20 Mo steel (SAE/AISI 4140-4142) that was austenitized at 860 °C (1580 °F). The vertical lines in the upper diagram give the cooling rate for the center of bars with different diameters when quenching in different media. The lower part of the figure shows the hardness after hardening and tempering (T) at various temperatures. Source: Ref 18

Thelning used the cooling curves just mentioned to derive dilatometer curves and presented the results in diagrams of the type shown in Fig. 12. From this diagram the structural constituents and hardness values that exist at 0.8 R and at the center of bars of various dimensions after cooling in water or oil can be determined.

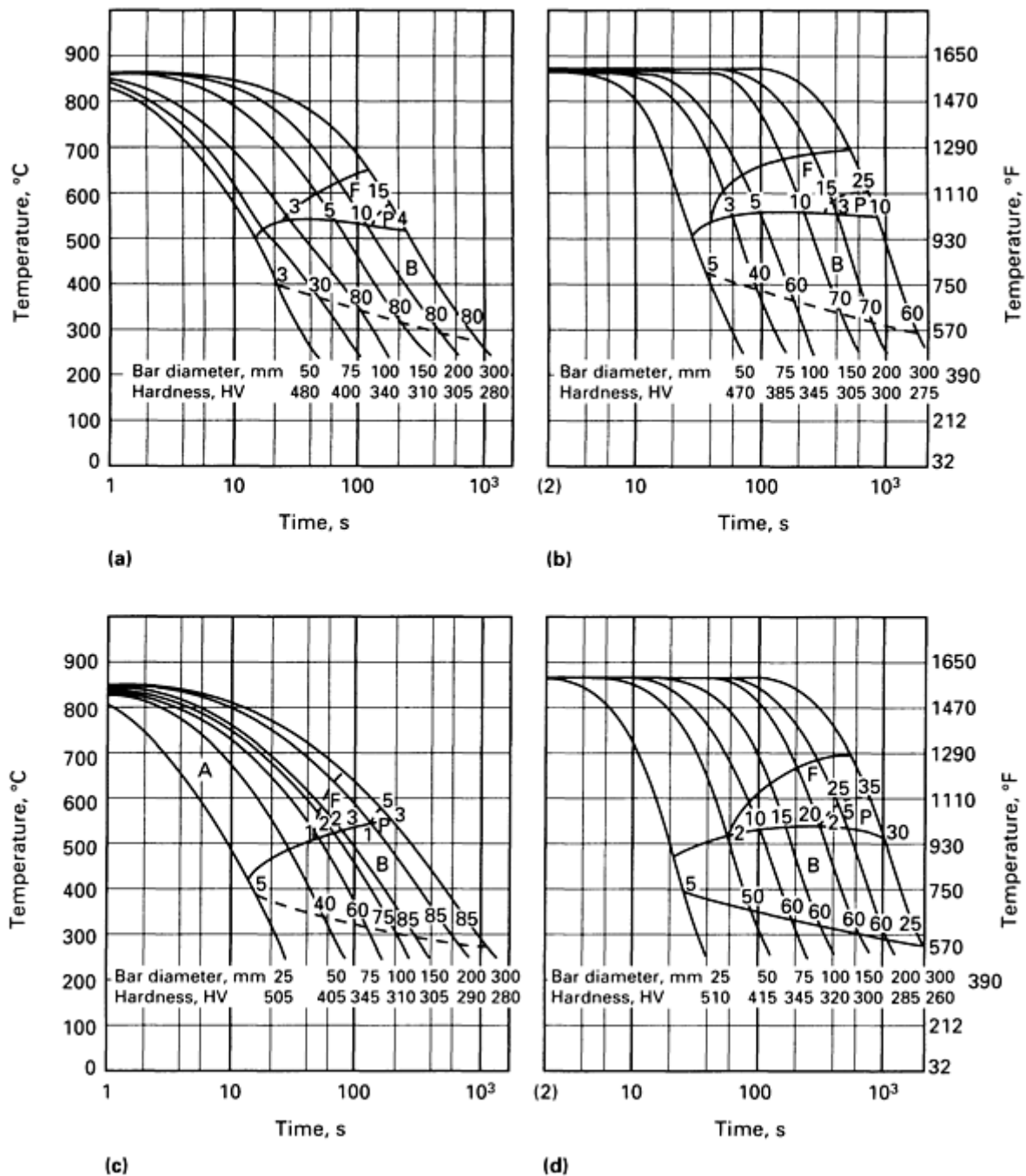


Fig. 12 CCT diagrams for AISI 4130 steel. (a) Water quench, 0.8 *R* specimen. (b) Water quench, center of specimen. (c) Oil quench, 0.8 *R* specimen. (d) Oil quench, center of specimen. Source: Ref 19

Computer Simulation of Transformation Diagrams

Because the experimental determination of transformation diagrams is costly and tedious, there is great interest in calculating these diagrams (Ref 20, 21). An effective computer model would allow the exact shape of the cooling curves discussed above to be determined.

The problems associated with developing such programs are twofold and involve the calculation of the TTT diagram and the calculation of the CCT diagram from the isothermal one. Because solving the first problem from first principles for a given steel composition is a formidable task, empirical correlations or semiempirical procedures are in current use (Ref 12). The second problem is much easier to handle, and several successful approaches have been tried. One of these will be presented below while alternative approaches are described in the following article in this Section (see "Quantitative Prediction of Transformation Hardening in Steels").

As a basis for the calculation, the Johnson-Mehl-Avrami expression that describes the diffusional transformation of austenite to either ferrite, pearlite, or bainite can be used:

$$v = 1 - \exp(-b \cdot t^n) \quad (\text{Eq 3})$$

where v is the transformed volume fraction, t is the time, and b and n are temperature-dependent constants; b and n are evaluated from the given TTT diagram except for ferrite for which it is assumed that $n = 1$. More generally, it can be stated that:

$$n = \frac{\log[\ln(1-v_s)/\ln(1-v_f)]}{\log[t_s(T)/t_f(T)]} \quad (\text{Eq 4})$$

$$b = \frac{-\ln(1-v_s)}{t_s(T)} \quad (\text{Eq 5})$$

where the subscripts s and f indicate start and finish, respectively. In a TTT diagram, v_s is usually chosen to be 1% and v_f 99%, but other percentages may also be chosen. The experimental C-shaped curves in the TTT diagrams are approximated by spline functions when stored in the computer. The cooling curve is approximated by a staircase, the step length being equal to the time step. The phase transformation is then calculated isothermally during each time step. This is equivalent to using the Scheil-Avrami additivity rule, which is described in the article "Quenching of Steel" in this Volume (see the discussion on quench factor analysis).

Note that in the general case several transformation products coexist and compete. For each phase being formed, a separate set of equations is written. For ferrite, the maximum volume fraction possible above A_1 is less than unity and is given by application of the lever rule to the multi-component phase diagram.

The martensitic transformation relation is based on the following expression (Ref 22):

$$v_M = \frac{(1 - v_F - v_P - v_B - v_C)}{(1 - \exp(-0.011 \cdot (M_s - T)))} \quad (\text{Eq 6})$$

where F , P , B , and C represent ferrite, pearlite, bainite, and cementite, respectively.

A comparison between measured and calculated CCT diagrams derived by linear cooling is shown in Fig. 9. The agreement is quite satisfactory. Figure 10 shows that the effect of different cooling rates on the CCT diagram is well reproduced by the calculations.

The model described above has been refined by incorporating a nucleation stage before the diffusional-growth stage (Ref 21). Scheil's principle of additivity of incubation fractions has been used, that is, for each time step, denoted as Δt_i , the fraction $\Delta t_i/\tau_{iIT}$ is calculated where τ_{iIT} is the time to the start of the transformation of the temperature prevailing during the time step. The fractions are then summed:

$$\sum \frac{\Delta t_i}{\tau_{iIT}}$$

When this sum attains unity, the incubation period is over. The nonadditivity of pearlitic and bainitic nucleation is accounted for by a correction factor of the Scheil sum at the transition from pearlite to bainite.

The calculation of the CHT diagram from the ITh diagram on heating is quite analogous. Good agreement between calculated and measured CHT curves where linear heating was applied is shown in Fig. 8.

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Hardenability Concepts

The goal of heat treatment of steel is very often to attain a satisfactory hardness. The important microstructural phase is then normally martensite, which is the hardest constituent in low-alloy steels. The hardness of martensite is primarily dependent on its carbon content as is shown in Fig. 13. If the microstructure is not fully martensitic, its hardness is lower. In practical heat treatment, it is important to achieve full hardness to a certain minimum depth after cooling, that is, to obtain a fully martensitic microstructure to a certain minimum depth, which also represents a critical cooling rate. If a given steel does not permit a martensitic structure to be formed to this depth, one has to choose another steel with a higher hardenability (the possibility of increasing the cooling rate at the minimum depth will be discussed later). There are various ways to characterize the hardenability of a steel. Certain aspects of this will be discussed in the following article in the Section and has also been described in detail in previous *ASM Handbooks*, formerly *Metals Handbooks* (Ref 23). The CCT diagram can serve this purpose if one knows the cooling rate at the minimum depth. The CCT diagrams constructed according to Atkins or Thelning presented above are particularly suitable.

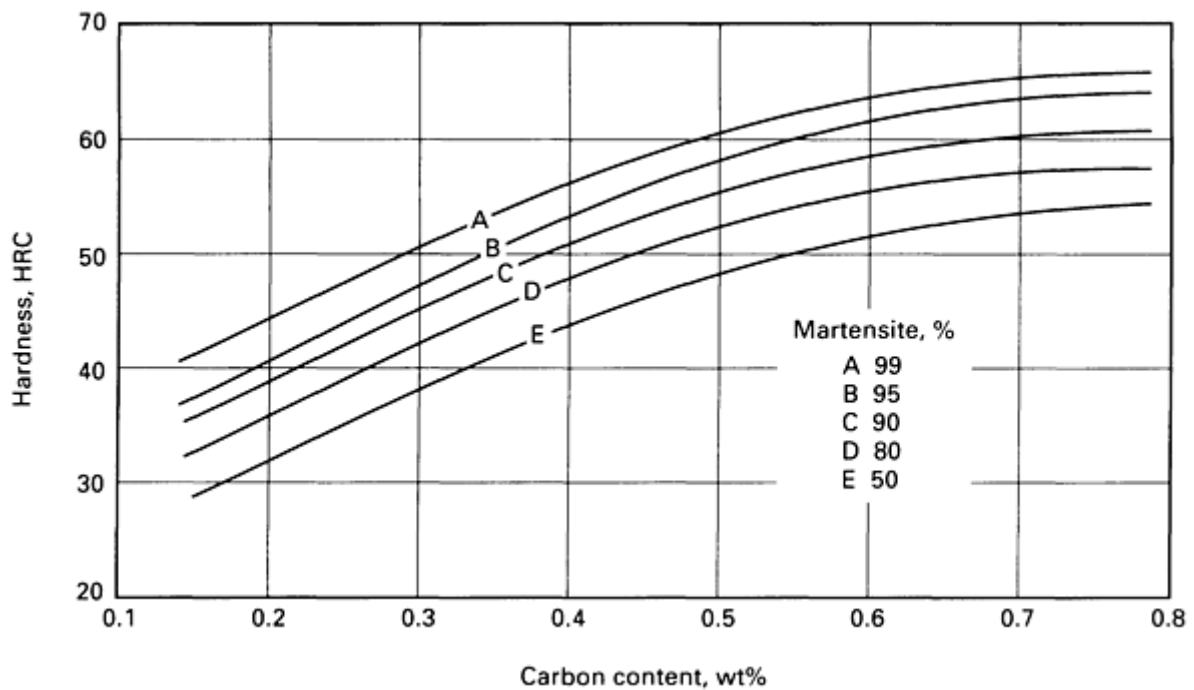


Fig. 13 Relationship between hardness, carbon content, and amount of martensite

Grossmann Hardenability Test. Another method using cylindrical test bars is the classical one of Grossmann (Ref 24). A number of cylindrical steel bars of different diameters are hardened in a given cooling medium. By means of metallographic examination, the bar that has 50% martensite at its center is singled out and the diameter of this bar is designated as the critical diameter (D_0). This D_0 value is valid for the particular cooling medium used as well as its cooling intensity. An ideal cooling situation is when the surface of the test bar is immediately cooled to ambient temperature, that is, an infinite cooling rate at the surface. Although such cooling cannot be carried out in practice, one can mathematically extrapolate this situation and derive the ideal Grossmann diameter, D_1 , defined as the bar diameter that, when the surface is cooled at an infinite rate, will yield a structure containing 50% martensite at the center of the specimen. The value of D_1 obtained is a measure of the hardenability of the steel and is independent of the cooling medium. Using this method, D_1 values have been determined for low- and medium-carbon steels with different carbon contents and grain sizes. The additional effect of alloying elements has been determined for a number of alloying elements and is expressed as hardenability factors, f , which are multiplying factors corresponding to the percentage of alloying element present. The ideal diameter, D_1 , for a certain steel can then be expressed as:

$$D_1 = D_0 \cdot f_1 \cdot f_2 \cdot f_3 \dots$$

where f_1, f_2, f_3 are the factors for alloying elements 1, 2, 3, ..., respectively, and D_0 applies to Fe-C alloys.

Diagrams have been developed that convert D_1 to D_0 for the cooling medium of interest. The ISO has also developed diagrams to convert from rectangular to round sections.

The Grossmann method is frequently used but gives a very approximate measure of hardenability. However, it can be reliably employed to compare steels of different compositions. More detailed information on the Grossmann method and the calculation of D_1 values from the chemical composition can be found in Ref 19 and in the article "Quenching of Steel" in this Volume.

Jominy End-Quench Test. The most commonly used experimental method for hardenability is the well-known Jominy test (Ref 23). For this test a round bar specimen that is 100 mm (4 in.) in length and 25 mm (1 in.) in diameter is used. The specimen is heated to the austenitizing temperature of the steel with a holding time of 20 min. One endface of the specimen is quenched by spraying it with a jet of water. This causes the rate of cooling to decrease progressively from the quenched end along the length of the bar. When it is cool, two diametrically opposite flats that are 0.4 mm (0.015 in.)

deep and parallel to the axis of the bar, are ground and the hardness is measured along the flats. The hardness values are plotted on a diagram at specified intervals from the quenched end. To get good reproducibility, the time and temperature of austenitizing, the grinding of the flats to avoid burning, and the placement of the specimen in the hardness tester should be carefully controlled. It is also important to protect the specimen against decarburization. Less critical are the water temperature, the diameter of the spraying nozzle, the height of the water jet, and the time to move the specimen from the furnace to the fixture. For steels with very high or very low hardenability, neither the Grossmann nor the Jominy methods are well suited and other methods are used. These methods are well covered in Volume 1 of *ASM Handbook*, formerly 10th Edition of *Metals Handbook* (Ref 23) and in Ref 24.

The rate of cooling at different distances from the quenched end is approximately independent of the steel used because the thermal conductivity and heat capacity of hardenable steels do not vary very much and the heat transfer at the cooled end is steel independent. Therefore, the Jominy bar presents a range of cooling curves that can be used to estimate a CCT diagram. The cooling rates are not linear but rather of the natural type according to Newton's law of cooling. The volume fractions of the various constituents of the microstructure are evaluated quantitatively at different Jominy distances as well as hardness measurements. Conversely, the Jominy curve can be calculated from the TTT diagram using the computer methods described above and expressions that give the hardness for the different phases (Ref 18). The total hardness is then the weighted average. Figure 14 shows the agreement between calculated and experimental Jominy curves that can be achieved.

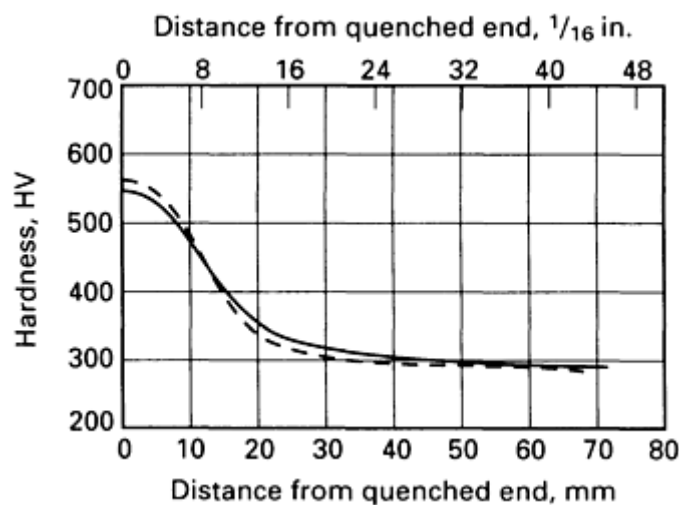


Fig. 14 Calculated hardness (dashed line) and reported hardness (solid line) from a Jominy test of AISI 4130 steel. Source: Ref 10, 11

There is an economic trend to reduce the amount of experimental Jominy testing and replace it by the calculation of Jominy curves from the chemical composition and the grain size. Several regression formulas have been developed for different grades of steel which are quite accurate for a limited composition range. A more general approach will be described in the article "Quantitative Prediction of Transformation Hardening in Steels" that immediately follows.

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Principles of Tempering of Steels

As pointed out earlier in this article, martensite is a very hard phase in steel. It owes its high hardness to a strong supersaturation of carbon in the iron lattice and to a high density of crystal defects, especially dislocations, and high- and low-angle boundaries. However, except at low carbon contents, martensitic steels have insufficient toughness for many applications. Tempering of martensitic steels, by heating for a certain time at temperatures below the A_1 , is therefore introduced to exchange some of the strength for greater ductility through reduction of the carbon supersaturation initially present and replacing it with more stable structures. Additionally, the retained austenite associated with martensite in steels containing more than about 0.7 wt% C can be decomposed during the tempering process. In carbon steels containing small percentages of the common alloying elements, one distinguishes the following stages during tempering (Ref 3):

- **Stage 1** (20 to 100 °C, or 70 to 212 °F). Short-range diffusion of carbon atoms to dislocations and martensite plate boundaries; formation of carbon clusters
- **Stage 2** (100 to 200 °C, or 212 to 390 °F). Precipitation of transition carbides, especially ϵ -carbide (Fe_2C)
- **Stage 3** (200 to 350 °C, or 390 to 660 °F). Transformation of retained austenite to ferrite and cementite
- **Stage 4** (250 to 700 °C, or 480 to 1290 °F). Formation of ferrite and cementite. Spheroidized carbides in a matrix of equiaxed ferrite grains comprises well-tempered steels, that is, tempered for long times at temperatures approaching 700 °C (1290 °F)

In steels alloyed with chromium, molybdenum, vanadium, or tungsten, formation of alloy carbides occurs in the temperature range 500 to 700 °C (930 to 1290 °F).

During stage 1, the hardness increases slightly while during stage 2, 3, and 4 the hardness decreases. Figure 15 shows the hardness of some steels after tempering for 2 h at different temperatures.

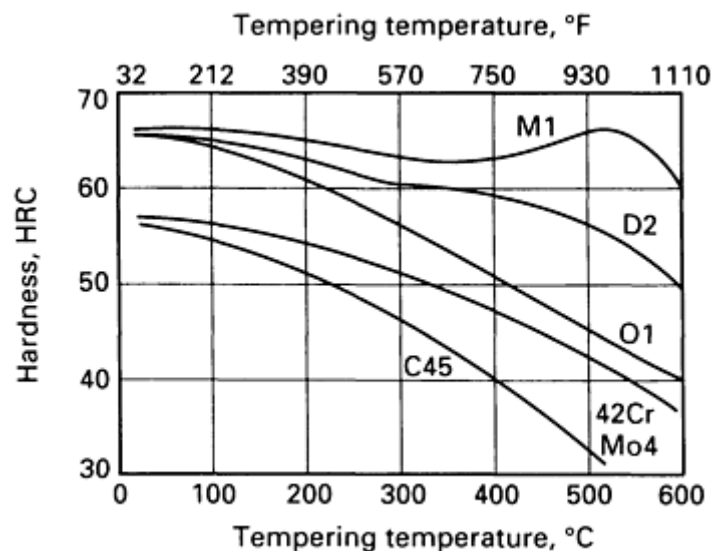


Fig. 15 Tempering curves for some current steels. The steel 42CrMo4 is equivalent to AISI 4142 and C45 to AISI 1045. Source: Ref 19

Hollomon and Jaffe (Ref 25) showed that the relation between hardness and time and temperature of tempering can be expressed graphically by plotting hardness as a function of a tempering parameter P , where:

$$P = T(k + \log t)$$

In this expression, T is the temperature (in degrees Kelvin), k is a constant with a value of about 20, and t is the time in hours. The physical background to this empirical expression is the fact that the various tempering reactions are thermally activated diffusional processes.

The toughness of a steel increases with decreasing hardness. However, when certain impurities such as arsenic, phosphorus, antimony, and tin are present, a toughness minimum termed "temper embrittlement" may occur in the temperature range 350 to 600 °C (660 to 1100 °F) due to segregation of impurities to grain boundaries. Temper embrittlement is a problem when parts are exposed to temperatures in the critical range for rather long times and is a concern for parts exposed to these temperatures while in service or when heat treating very massive parts which require long times to heat and cool. It is not a concern, even for susceptible alloys, if small parts are exposed to these temperatures for an hour or so during heat treatment, then used at ambient temperatures. Nuts and bolts, for example, made of various types of steels, are tempered in this temperature range with no problems as long as they are used at lower temperatures. The time for embrittlement to occur at differing tempering temperatures shows a C-shaped curve behavior in a temperature-log time diagram similar to what is found in transformation diagrams (Fig. 16). The long times in the upper temperature range are due to low thermodynamic driving force and in the lower temperature range to slow diffusion. Temper embrittlement can be removed by reheating the steel above 600 °C (1110 °F) followed by rapid cooling, for example, water quenching.

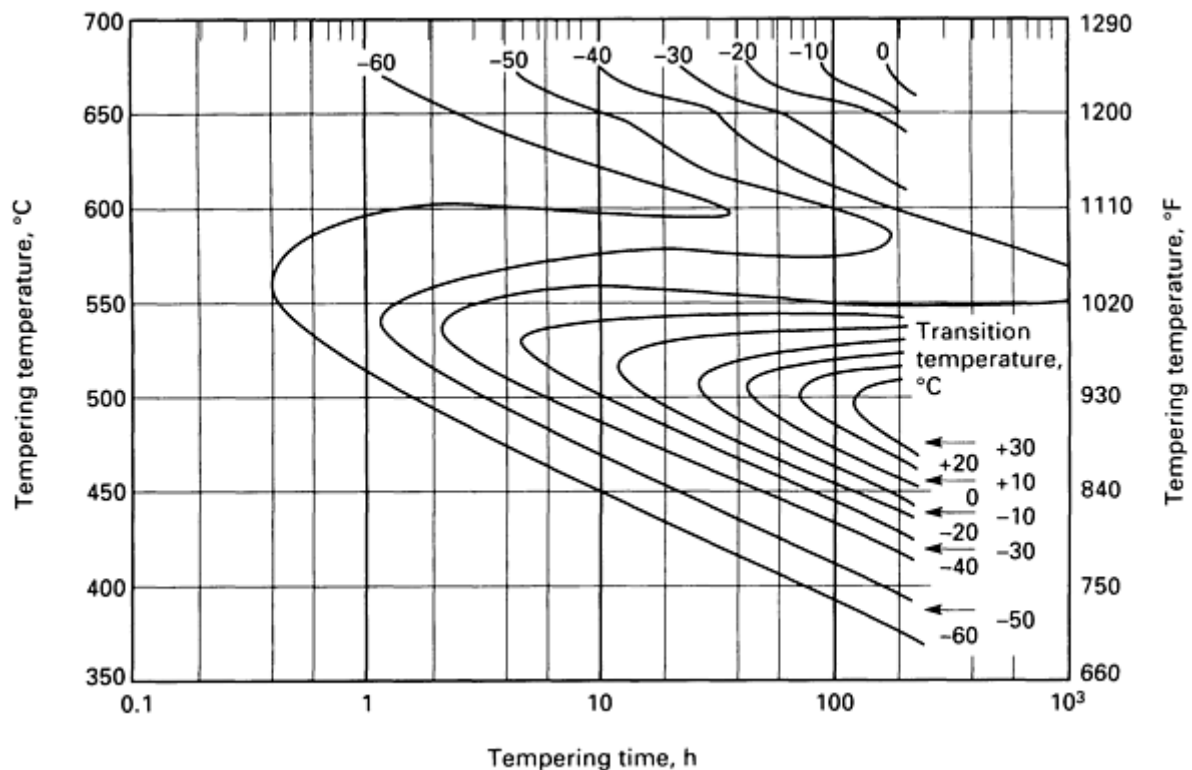


Fig. 16 Dependence of transition temperature on tempering temperature and time for SAE 3140 steel containing 0.40% C, 0.80% Mn, 0.60% Cr, and 1.25% Ni. Water quenched from 900 °C (1650 °F). Source: Ref 26

Another type of embrittlement that affects high-strength alloy steels is tempered martensite embrittlement (also known as 350 °C, or 500 °F, embrittlement), which occurs upon tempering in the range of 205 to 370 °C (400 to 700 °F). It differs from temper embrittlement in the strength of the material and the temperature exposure range. In temper embrittlement, the steel is usually tempered at a relatively high temperature, producing lower strength and hardness, and embrittlement occurs upon slow cooling after tempering and during service at temperatures within the embrittlement range. In tempered martensite embrittlement, the steel is tempered within the embrittlement range, and service exposure is usually at room temperature. Therefore, temper embrittlement is often called two-step temper embrittlement, while tempered martensite embrittlement is often called one-step temper embrittlement. Detailed information on both temper embrittlement and

tempered martensite embrittlement, including time-temperature diagrams for embrittled steels, can be found in Volume 1 of *ASM Handbook*, formerly 10th Edition *Metals Handbook* (see Ref 27).

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Cooling Media and Quench Intensity

The depth of hardness at a given work-piece dimension is determined by the chemical composition of the steel, the austenite grain size as established during the austenitizing treatment, and the cooling rate. The steel is normally chosen on the basis of hardenability. The choice of cooling medium, on the other hand, is less exact and crude rules are normally applied (unalloyed steel is quenched in water, alloy steels in oil, and high-alloy steels in air). Molten salt is often used for bainitic hardening of medium-carbon steels and martempering of carburized parts (salt temperature above M_s for the case). Judicious selection of cooling medium is critical for obtaining optimum mechanical properties, avoiding quench cracks, minimizing distortion (to be discussed below), and improving reproducibility in hardening. Of most interest are the liquid quenching media, and they also show the most complicated cooling process. More detailed information on cooling (quench) media can be found in the article "Quenching of Steel" in this Volume.

Three Stages of Quenching. The most commonly used liquid quenching media are water and its solutions (brine and caustic solutions), oils, and polymer solutions. During the quenching of steel in liquid media, the process may be split up into the following three stages:

- The vapor blanket stage
- The boiling stage
- The convection stage

Figure 17 shows a typical cooling curve for an oil and at the same time shows what happens at the surface of a steel that is being quenched.

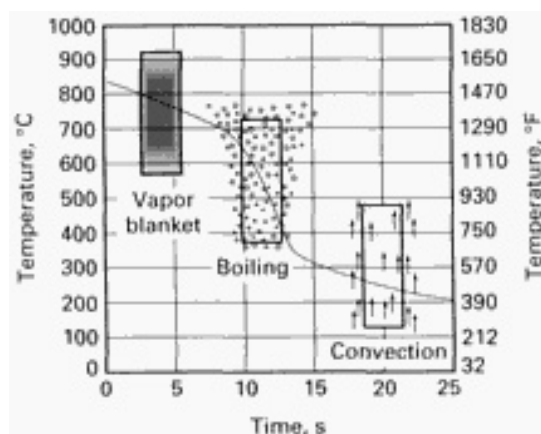


Fig. 17 The three stages of quenching. Source: Ref 19

In the vapor blanket stage the surface temperature is so high that the cooling medium is vaporized and a thin vapor is formed around the part. Heat transfer occurs by radiation through the vapor blanket, which thus acts as an insulating layer. The cooling rate is therefore low. The time duration of the vapor blanket stage can vary considerably among different cooling liquids and is also affected by the surface condition (for example, the presence of oxide scale) and how densely the workpieces are loaded on the furnace tray.

The boiling stage starts when the surface temperature has decreased to the point where the heat of radiation does not sustain a stable vapor blanket. The liquid that is brought into contact with the hot surface boils immediately and vapor bubbles leave the metal surface and provide efficient heat transfer. This gives a high rate of heat extraction.

The convection stage starts when the surface temperature has decreased to the boiling point of the cooling medium. The heat transfer occurs by direct contact between surface and liquid. The cooling rate is low and is affected primarily by quench viscosity and convection rate.

Evaluating Quenching Media. There are a great number of different methods for evaluating quenching media. Frequently, a 12.5 mm diameter by 60 mm long (0.5 × 2.5 in.) cylinder (probe) made of a nickel-base alloy or stainless steel is used. A thermocouple is located at the center of the testing body and is connected to a recording instrument. Examples of such cooling curves are shown in Fig. 18. Of more interest in many situations is the cooling rate, also shown in Fig. 18, obtained by differentiation of the cooling curves. These types of curves are useful for comparison between quenching media and judging the condition of a certain medium. Earlier a silver ball, 20 mm (0.08 in.) in diameter, was employed as the probe. The cooling curves from such measurements are even more difficult to apply to hardness predictions because silver has very different thermal properties and the physical properties at the surface are not representative of steel.

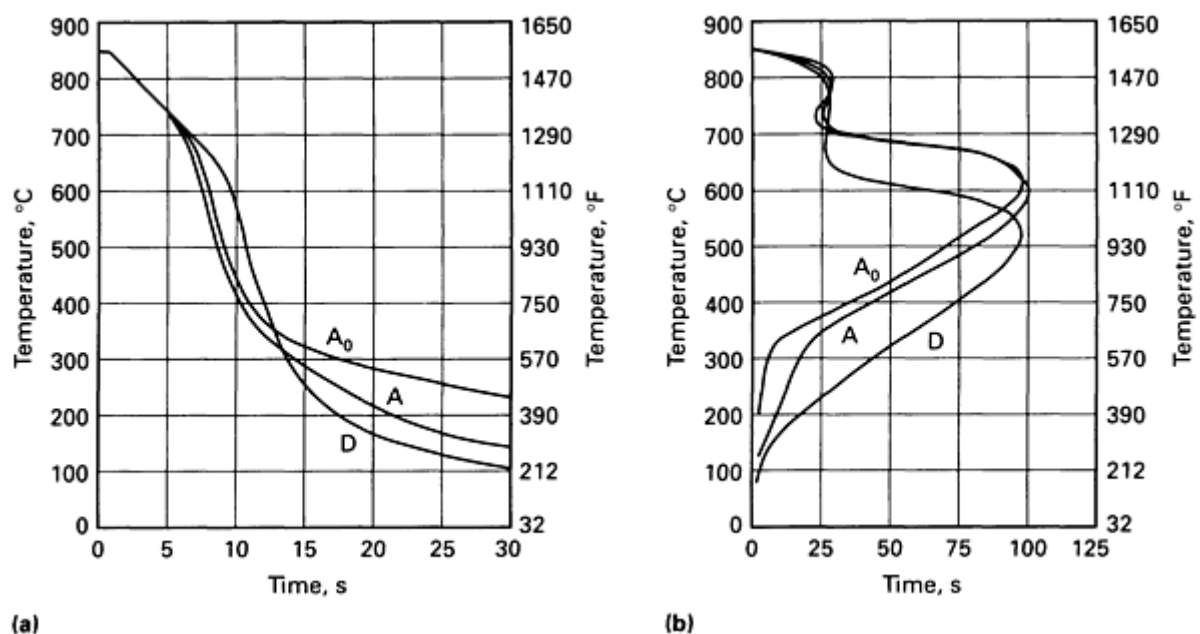


Fig. 18 Temperature-time curves (a) and temperature-cooling rate curves (b) obtained using an Alloy 600 probe. A and D are two oils with slight agitation, and A₀ is the oil A with no agitation. Source: Ref 28

Visual observations have shown that the breakdown of the vapor blanket starts at edges where the radius of curvature is small. The length of the test specimen can therefore affect the duration of the vapor blanket stage.

Figure 19 shows the temperature at different depths below the surface of an Inconel probe quenched in oil. This figure illustrates that the different quench stages are more pronounced closer to the specimen surface and that the resulting

temperature curves differ more from either linear or natural cooling curves. The conventional CCT curves must then be used with caution for hardening response predictions.

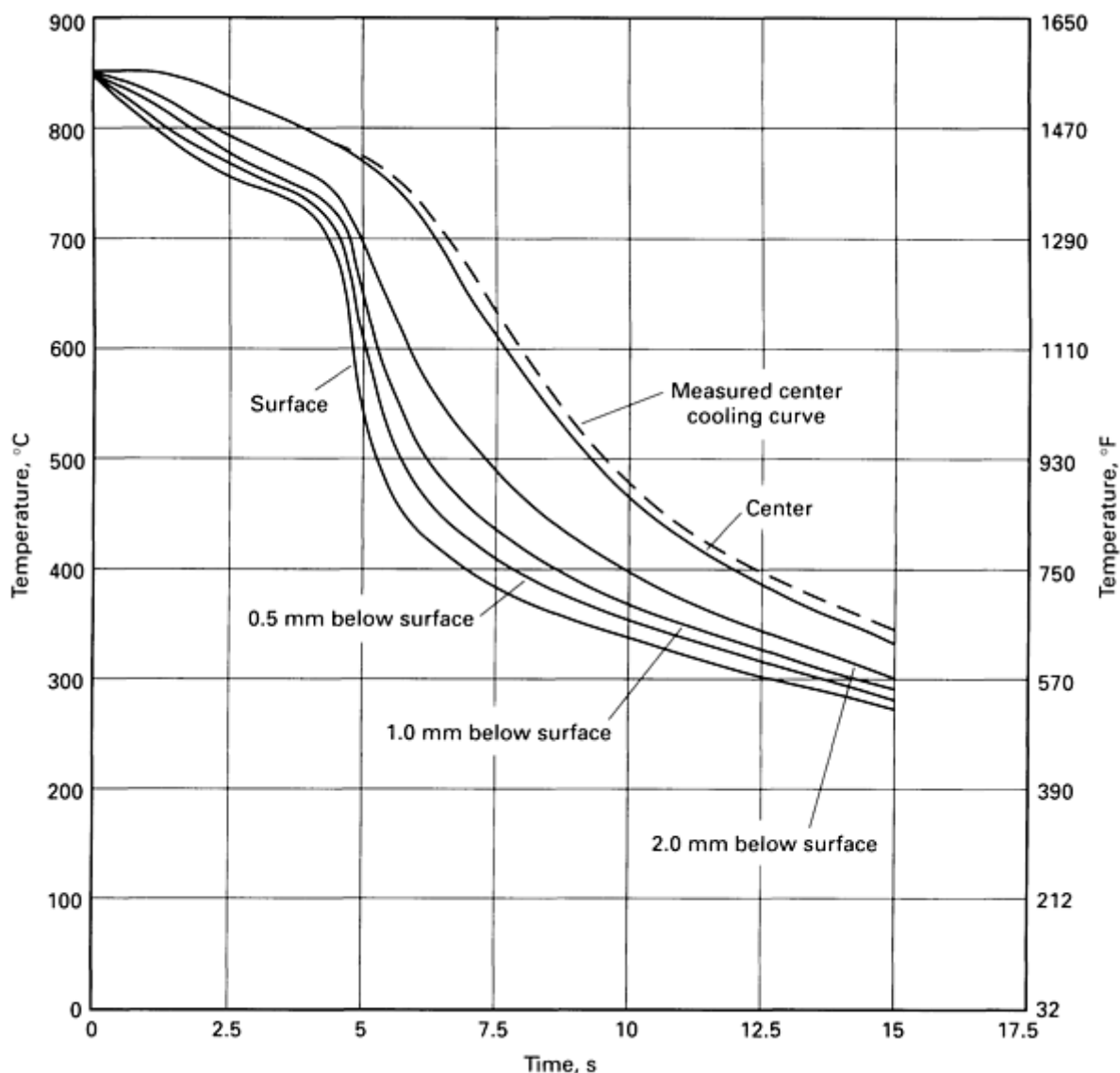


Fig. 19 Calculated cooling curves obtained using an Alloy 600 probe based on the heat flow curve in Fig. 20. Source: Ref 29

The curves in Fig. 19 are calculated using the temperature measured 2 mm (0.08 in.) below the surface. The calculation is based on the solution of the heat-flow equation. The experimental curve is assumed to be the boundary condition when determining the temperature-time history of the probe. The calculated temperature gradient through the surface is then employed to solve the heat transfer through the outer surface, which can then be expressed as a function of the surface temperature (Fig. 20). The heat-transfer function thus derived can be used to solve the heat-flow equation for the part of interest and to derive the temperature distribution in the part as a function of time, provided that the same state of quenching medium agitation prevails. A more accurate heat-transfer function is obtained if the experimental curve is measured closer to the surface. However, this is experimentally more difficult.

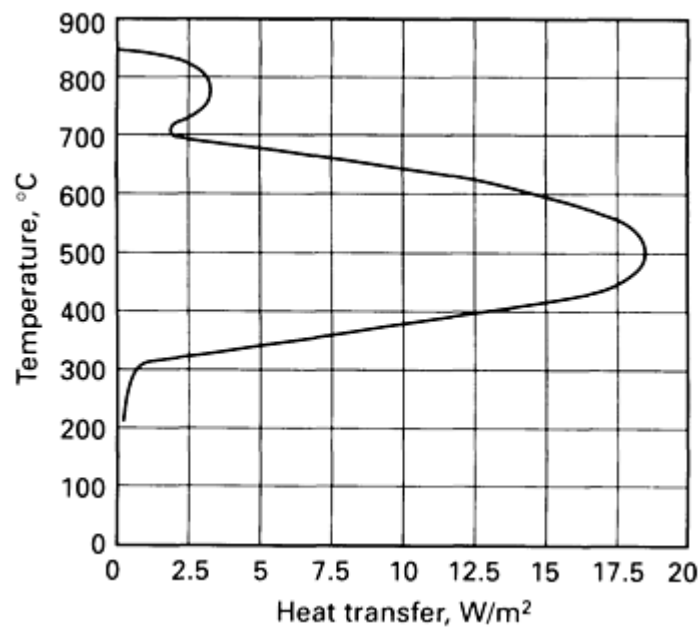


Fig. 20 Heat transfer as a function of surface temperature of the test probe used in Fig. 19. The curve is calculated from temperature measurements 2.0 mm (0.08 in.) below the probe surface. Source: Ref 29

In the classical literature, the quench severity concept, or the Grossmann number, is used. This is a single temperature independent value for each state of agitation of oil, water, and brines, respectively. It is a much cruder measure of the quench intensity than the heat-transfer function. The Grossmann number (H -value) is defined by $H = h/2k$, where h is the coefficient of heat transfer at the metal-quenchant interface and k is the thermal conductivity. See the article "Quenching of Steel" in this Volume for a more detailed discussion of the Grossmann number.

Very often it is stated that a short-lived vapor blanket stage will enhance the hardening result. This is not always correct as has been pointed out by Thelning (Ref 19). The most important consideration is the rate of cooling through the temperature ranges in which the diffusion-dependent transformations take place. Therefore, the characteristics of the steel must be considered at the same time as the quenching medium. It should also be noted that by adding certain organic substances to an oil or water quenchant, the cooling characteristics can be modified by extending the vapor blanket stage or delaying the transition from the boiling stage to the convection stage.

A more recently developed method is computer-controlled spray cooling (Ref 30). Using this technique, it is possible to program fast cooling past the pearlite nose on the CCT diagram, have a very slow cooling just above M_s , followed by a moderate cooling during the martensite transformation.

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Thermal Stresses during and Residual Stresses after Heat Treatment

Heat treatment of steel, especially martensitic hardening, is usually accompanied by the evolution of large residual stresses, that is, stresses that exist without any external load on the part considered. Causes for such stresses include:

- Thermal expansion or contraction of a homogeneous material in a temperature gradient field
- Different thermal expansion coefficients of the various phases in a multiphase material
- Density changes due to phase transformations in the metal
- Growth stresses of reaction products formed on the surface or as precipitates, for example, external and internal oxidation

Residual stresses can be divided into three categories. A macroresidual stress is the average of the residual stress in many adjacent grains of the material. If a work-piece is cut or material is removed, the presence of macroresidual stress will cause a distortion. The introduction of macroresidual stresses into a workpiece by heat treatment or plastic deformation may also cause a distortion of the part. The pseudo-macroresidual stress is the average of the residual stress in many grains of one phase in a multiphase material minus the macroresidual stress. The microresidual stress in a part is the total residual stress minus the macroresidual and the pseudo-macroresidual stress. The residual stresses considered in this section are of the macrotype. Stresses that exist during the entire heat-treatment process will also be discussed in the following paragraphs.

The subject of residual stresses after heat treatment of steel has been studied extensively in the recent literature (Ref 31, 32, 33) and is also discussed in the article "Defects and Distortion in Heat-Treated Parts" in this Volume. The principle for the creation of thermal stresses on cooling is shown in Fig. 21 for a 100 mm (4 in.) diameter bar that was water quenched from the austenitizing temperature of 850 °C (1560 °F). The surface temperature (S) decreases more rapidly than the core temperature (C), and at time w , the temperature difference between the surface and core is at a maximum of about 550 °C (1020 °F). This means that the specific volume is greater in the core than in the surface. The volume contraction in the surface is prevented by the higher specific volume in the core. The thermal stress is approximately proportional to the temperature difference and is tensile in the surface and compressive in the core. Large thermal stresses are favored by low thermal conductivity, high heat capacity, and high thermal expansion coefficient. Other factors increasing the temperature difference and thermal stresses are large thickness dimensions and high-cooling intensity of the cooling medium. A large yield stress at elevated temperatures will decrease the degree of plastic flow and thus the residual stress, while the yield stress at the ambient temperature puts an upper limit on the residual stress.

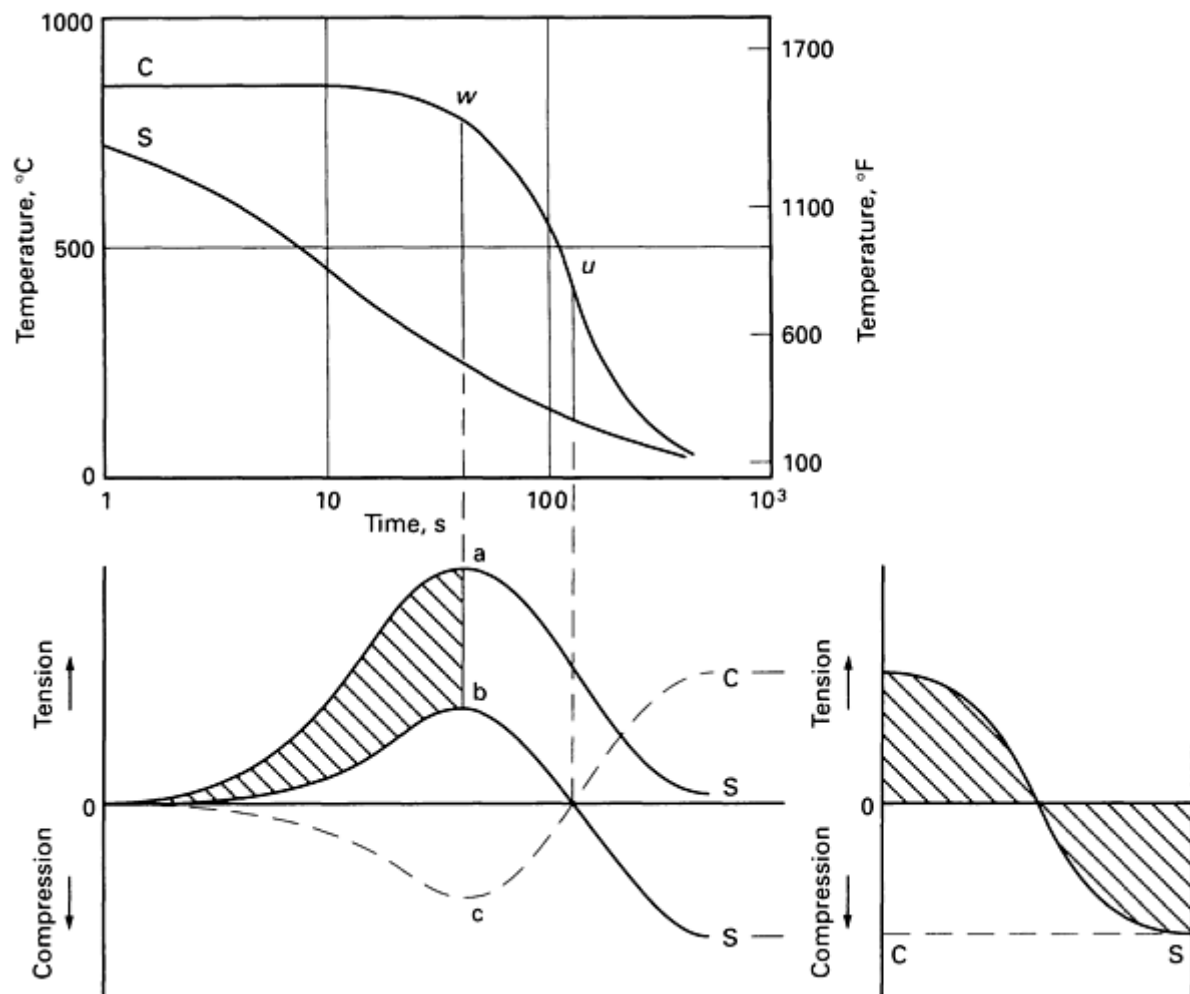


Fig. 21 Formation of thermal stresses on cooling in a 100 mm (4 in.) steel specimen. C designates the core, S the surface, u the stress reversal time instant, and w the time instant of maximum temperature difference. The top graph shows the temperature variation with time at the surface and in the core; the graph below shows the hypothetical thermal stress, a , which is proportional to the temperature difference between the surface and the core, the actual stress at the surface, b , which can never exceed the yield stress, and the actual stress in the core, c . To the right is shown the residual stress distribution after completed cooling as a function of the specimen radius. Source. Ref 33

The added effect of transformation of austenite to martensite in steel is demonstrated in Fig. 22. At time t_1 , the surface temperature falls below the M_s temperature and the surface starts to transform. The surface expands and the thermal tensile stresses are counteracted. The stress reversal takes place earlier than when transformation stresses are not taken into consideration. At time t_2 , the core transforms, causing another stress reversal. After cooling, transformation-induced tensile stresses at the surface dominate over the thermally induced compressive stresses.

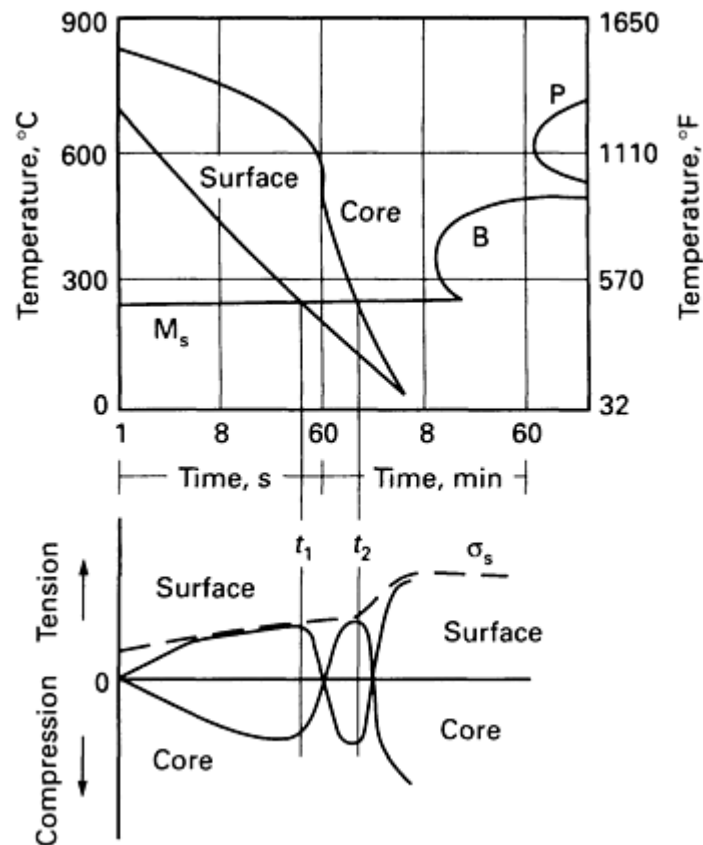


Fig. 22 Formation of residual stress on cooling considering thermal expansion and the austenite to martensite transformation. The dashed line is the yield stress, σ_s , at the surface. See text for details. Source: Ref 34

In order to predict hardening stresses quantitatively, it is necessary to consider interactions among various factors. As shown in Fig. 23, these include: (1) phase transformations, (2) latent heat, (3) thermal stress, (4) transformation stress and plasticity, (5) heat generation due to deformation, and (6) mechanically induced transformation. The most important of these are interactions 1, 3, and 5. Interaction 6, however, is also a very important factor. When discussing mechanically induced transformation, at least three different effects should be mentioned (Ref 36). The first is that the M_s temperature is decreased by hydrostatic pressure and raised by tensile stress (see Fig. 24, which shows an increase of $\sim 15^\circ\text{C}$ ($\sim 27^\circ\text{F}$) for a high-carbon steel). The second effect is the transformation plasticity, which is a permanent strain that occurs during an ongoing phase transformation under applied stress lower than the yield stress. It is displayed in Fig. 24 as an increase in the elongation from about 1% under an applied stress of 18 MPa (2.6 ksi) to 3% under 285 MPa (41 ksi) applied stress. The third effect is the incubation time of the nonmartensitic transformations, which is prolonged by hydrostatic pressure (Ref 37) and shortened by tensile stress (Ref 38). This is particularly important for large dimensions. It has also been shown that a separate steel sample that is inserted into a cylinder of the same hardenable steel (Carney-type test) has a higher hardness value than the material in the same position in a homogeneous steel cylinder. The inserted steel specimen has the same temperature history, but is not exposed to hardening stresses.

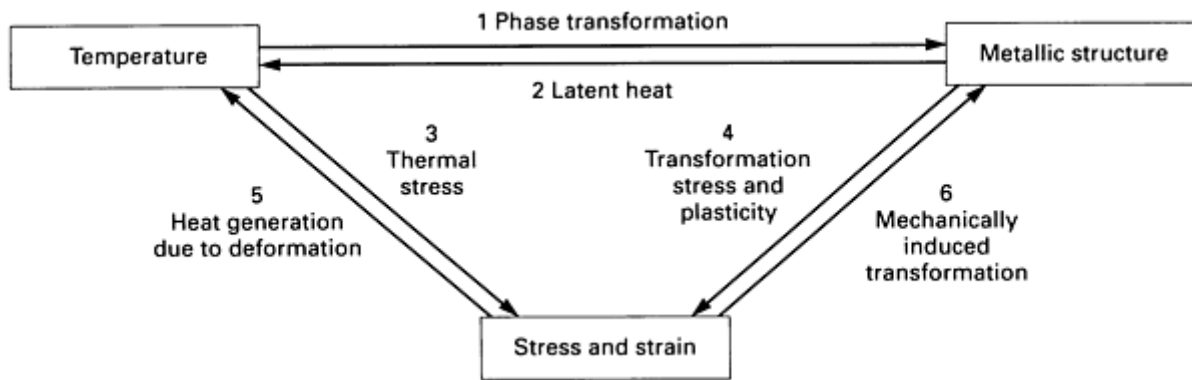


Fig. 23 The interactions between various factors of importance for residual stress generation. Source: Ref 35

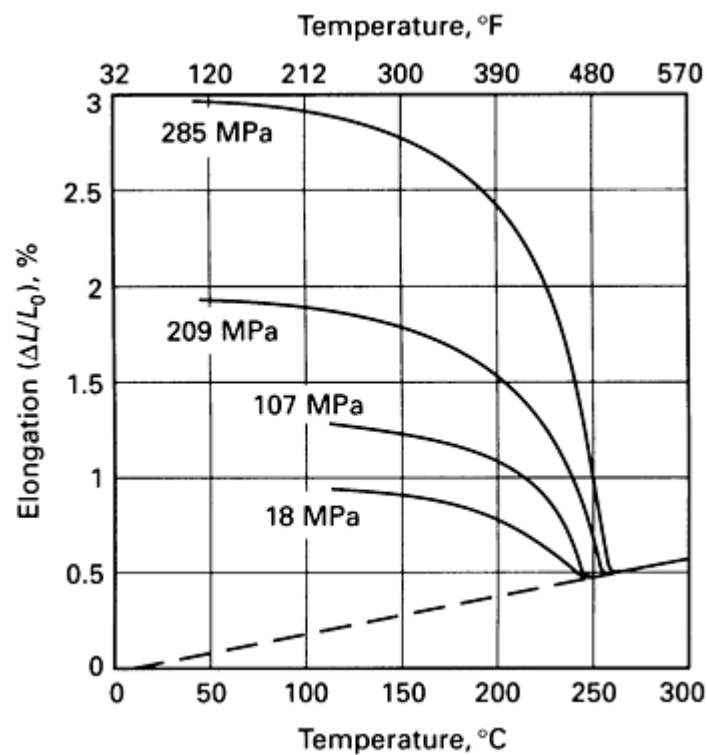


Fig. 24 Dilatometer curves for a steel with 0.6 wt% C for different applied tensile stresses. Source: Ref 36

Through Hardening. A series of computer calculations for a 50 mm (2 in.) diameter 0.6 wt% low-alloy steel cylinder illustrating the stress variation with time and the effect of the cooling medium (that is, the heat-transfer function) are shown in Fig. 25. In Fig. 25, which shows quenching in 20 °C (70 °F) oil, the transformation takes place well after the stress reversal in both surface and core. Some slight plastic deformation occurs in the surface in connection with the surface transformation, but otherwise the behavior is elastic. The effect of the transformation on the residual stress distribution is very slight. Quenching in 20 °C (70 °F) water is shown in Fig. 26. The transformation of the surface occurs before the stress reversal and transformation of the core occurs well after the stress reversal. The surface transformation helps to push the surface stress rapidly into compression, and compression yielding occurs. Before the core transformation, the core deforms plastically. With the onset of the core transformation, the tensile stress in the core is decreased, and the resulting deformation is elastic.

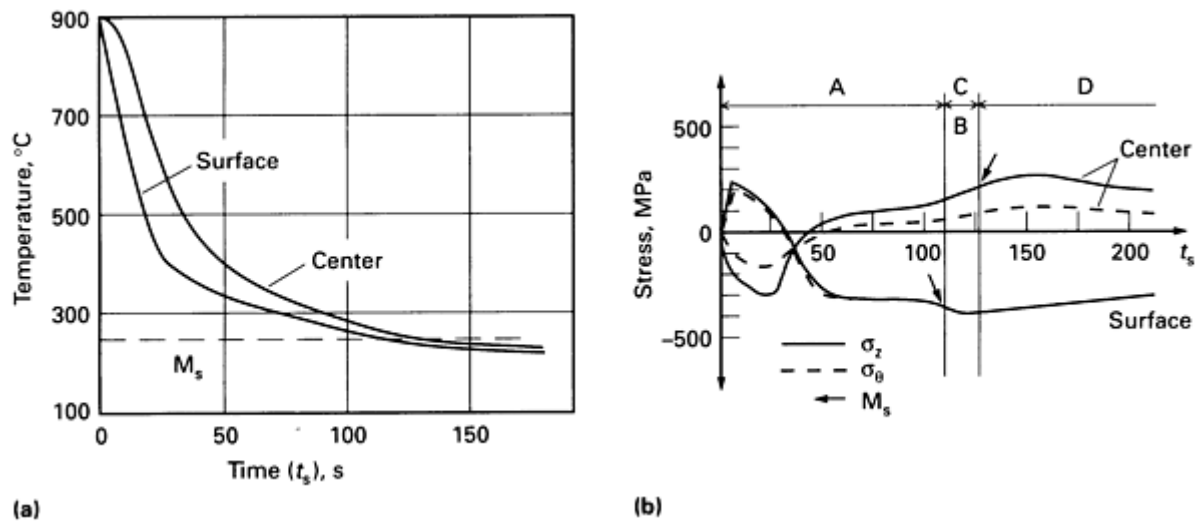


Fig. 25 Temperatures (a) and calculated stresses (b) in a long 0.6 wt% C low-alloy steel cylinder with 50 mm (2 in.) diameter quenched in 20 °C (70 °F) oil. In region A, the temperature is above M_s ; in region D, it is below M_s . In region C, B, the temperature is above M_s at the center, but below M_s at the surface. σ_z , axial stress; σ_θ , tangential stress. Source: Ref 39

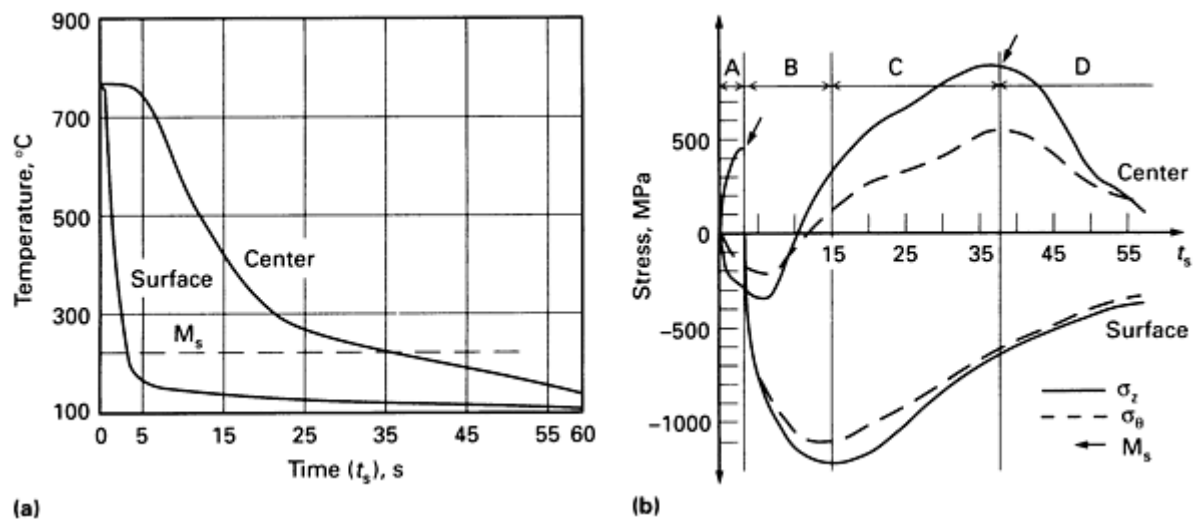


Fig. 26 Same type of diagram for the same steel and cylinder as in Fig. 25, but the specimen is quenched in 20 °C (70 °F) water. The final microstructure is completely martensitic. Source: Ref 39

In these calculations, the transformation plasticity was not considered. Figure 27 shows that it is important to include this effect in order to obtain the best agreement with measured residual stress values.

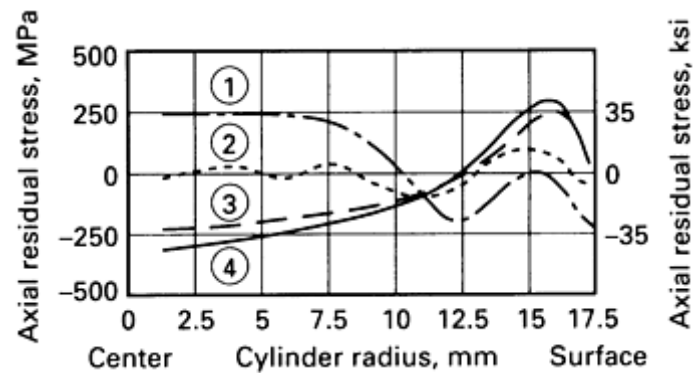


Fig. 27 Calculated residual axial stress profiles after martensitic quenching of a 0.6 wt% C low-alloy steel in 20 °C (70 °F) water. Curve 1: the effect of internal stresses on the kinetics of martensitic transformation. Curve 2: calculated without stress/transformation interactions. Curve 3: calculated with transformation plasticity alone. Curve 4: calculated with both stress and transformation effects considered. Source: Ref 36

Another series of calculations has been made to study the effect of cylinder diameter and cooling medium on residual stresses in 1045 steel. In the case of oil, ferrite-pearlite mixtures are formed; during water quenching, martensite and bainite are also formed. Figure 28 shows the stress development during water quenching. The 10 mm (0.4 in.) diameter cylinder starts to transform to martensite at the surface and the transformation front moves gradually inward resulting in a typical tensile stress at the surface. The large-diameter cylinders first transform to ferrite-pearlite at intermediate radii and then to martensite at the surface. This causes two stress minima seen in the dashed curves in Fig. 28(a), (b), and (c). The final residual stress is compressive at the surface and tensile in the core. Good agreement was obtained with x-ray stress measurements in several cases, as well as with older mechanically measured data. The dependence on specimen diameter and quenching medium is summarized in Fig. 29. The difference between oil and water quenching decreases with increasing diameter.

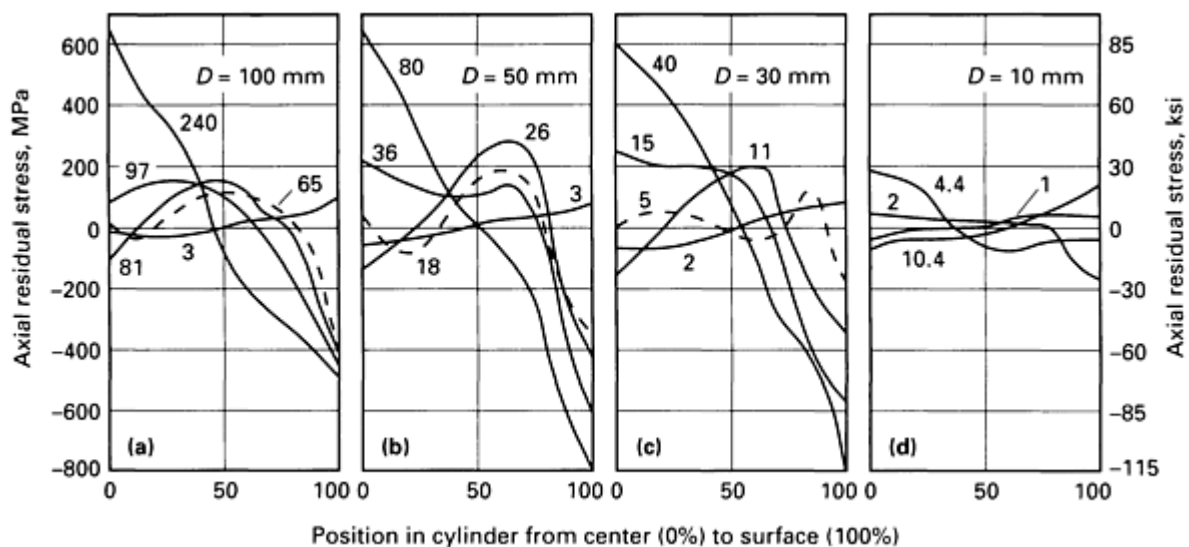


Fig. 28 Axial residual stress distributions for various AISI 1045 steel cylinder diameters, D , at selected times (in seconds) after the start of quenching from 850 °C (1560 °F) in 20 °C (70 °F) water. The final microstructure of the 10 mm (0.4 in.) diameter cylinder is completely martensitic, while the others have a ferritic-pearlitic core. Source: Ref 40

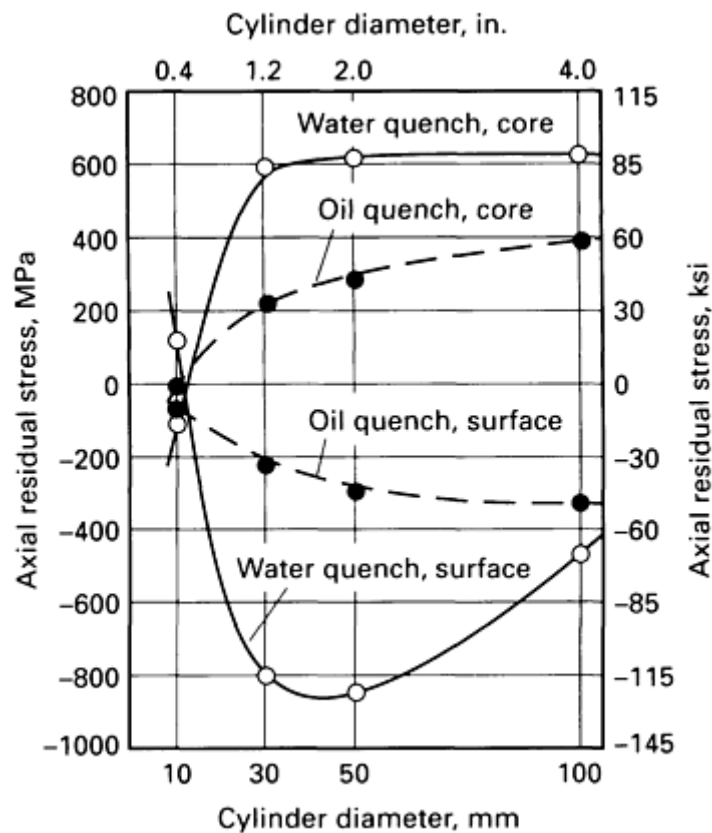


Fig. 29 The dependence of axial residual stresses on cylinder diameter. Same steel as in Fig. 28. The core is martensitic for 10 mm (0.4 in.) diameter, but ferritic-pearlitic for larger diameters. Source: Ref 40

Case hardening (that is, carburizing and quenching) is a rather complicated process. During the carburizing stage at about 900 °C (1650 °F), carbon diffuses into the steel from the surrounding environment, most often as a gas mixture. The carburizing takes several hours, and usually the temperature is lowered 50 to 100 °C (90 to 180 °F) for about one hour before quenching to reduce the risk of distortion. It is reasonable to assume that the material is stress-free when the quenching takes place.

Figure 30 shows the development of the residual stresses during the case hardening of a 17 mm (0.70 in.) diameter cylinder with a carburizing depth of 1 mm (0.04 in.). The onset of martensite formation can clearly be seen below the surface where, due to the combination of temperature and carbon content, the M_s is first attained. The transformation to martensite then moves toward the surface and the transformation to bainite moves toward the core. The transformation of austenite at the surface occurs very late and some retained austenite remains. The residual stress is first tensile at the surface due to thermal effects. The stress is shifted in the compressive direction where the martensite has formed, and simultaneously the tensile stresses are increased near the surface in the untransformed austenite as a compensating effect. Under these conditions, the hard and brittle martensite is never exposed to tensile stresses.

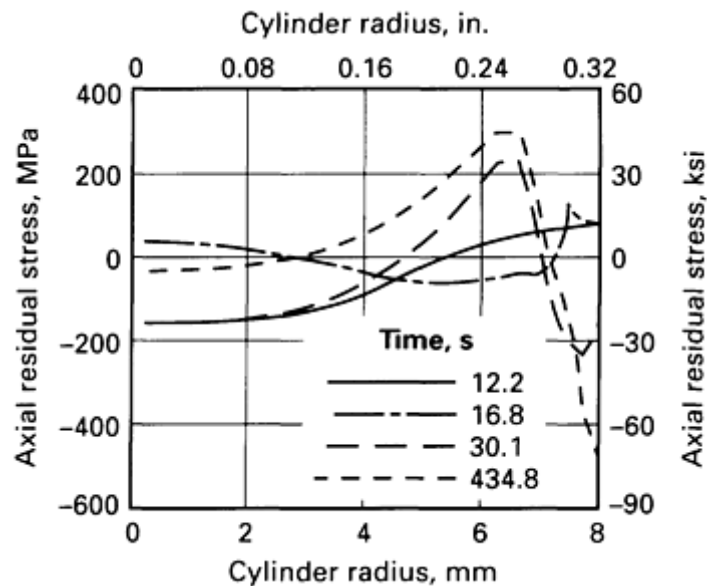


Fig. 30 Calculated axial residual stress for various times following the start of quenching after carburizing a 17 mm (0.70 in.) diameter cylinder made of AISI 3138 steel. Source: Ref 41

The final residual stress distribution has its minimum slightly below the surface. This is due to the presence of retained austenite near the surface. The position of the residual stress minimum occurs approximately at 50 to 60% of the total case depth corresponding to a carbon concentration at a position where the case carbon content is 0.5 wt% C. Below the hardened case, the residual stress may show a tensile peak.

This type of computer prediction can give good agreement with experimentally measured residual stress distributions (Fig. 31). The observed tensile stress at the surface is due to the presence of a soft pearlitic layer caused by the slight internal oxidation of the alloying elements chromium and molybdenum, which has reduced the hardenability of the surface layer. This reduction in hardness can be included in the calculations by adding a transformation diagram relevant to this surface layer (Ref 42).

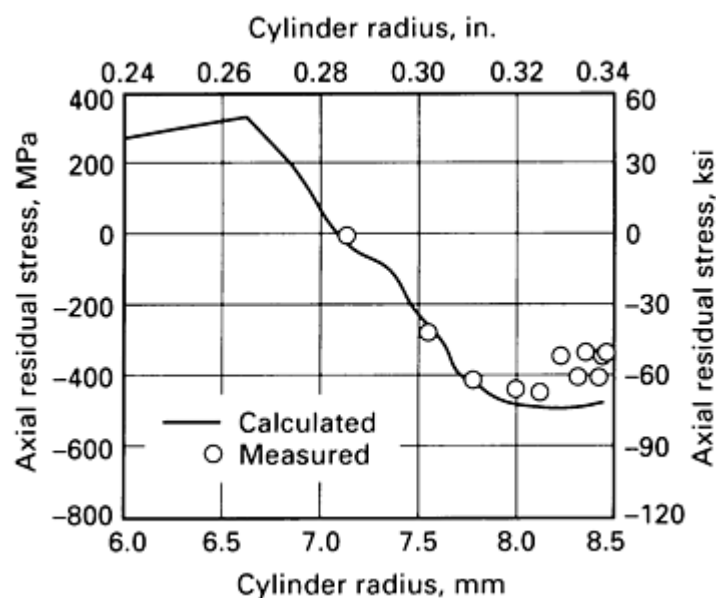


Fig. 31 Calculated and measured axial residual stress for a 17 mm (0.70 in.) diameter cylinder made of AISI 3138 steel. Source: Ref 41

With access to a good computer model, the effect of a number of variables, such as increased base carbon content, different cooling medium, and higher or lower base hardenability, can be studied. Increased base carbon content does not affect the magnitude of compressive stress significantly but the tensile peak below the case disappears. A more intense cooling medium like water instead of oil changes the compressive residual stress value to a higher value of 600 MPa (87 ksi) compared to 500 MPa (73 ksi), but does not increase the depth of the compressed layer.

Induction hardening is more complicated than carburizing. Both the heating stage as well as the cooling stage have been calculated (Ref 43). Static, or single-shot induction hardening, can be distinguished from progressive induction hardening. Using the single-shot method, a multiturn coil totally surrounds the area to be hardened; using the progressive technique, the workpiece and a single-turn coil are moved relative to each other. The single-shot method is more readily calculated and will be described here.

To calculate the heating during induction hardening, the magnetic field equation must be solved and the magnetic permeability, μ is determined. A function fitted to experimental curves is:

$$m = \frac{a}{b + H} + c \quad (\text{Eq 7})$$

where a , b , and c are constants and H is the magnetic field strength. The electrical resistivity must also be determined as a function of temperature and phases present.

Quenching by a water spray requires experimental data for the corresponding heat-transfer function. For the description of the transformation to austenite from the starting structure, an isothermal diagram for heating (ITh diagram) is needed. To describe the transformation behavior during the subsequent cooling, an IT diagram representative of a short austenitizing time is necessary. The austenitizing time during induction hardening is normally less than one minute. This means that carbide dissolution has little time to take place, and therefore the type and distribution of carbides in the starting structure are important. It is well known that structures with fine carbides harden to greater depths than structures with coarse carbides.

Calculated values for single-shot induction hardening of a 4140 steel specimen are shown in Fig. 32. Experimentally measured residual stresses for the same steel are compared with the calculated values in Fig. 33. In Fig. 32(a), the evolution of the temperature with time at different depths is shown. It is clearly seen that the heating rate decreases at the Curie temperature and that the transformation to austenite absorbs heat (constant temperature at the surface between 10 and 20 s). A brief air cooling step is followed by water spraying. The final martensite distribution is shown in Fig. 32(b). The evolution of the stress versus depth at different times and stress versus time for different depths is shown in Fig. 32(c) and (d), respectively. At the outset, the stress becomes compressive in the heated zone, but is shifted in the tensile direction when the austenite formation sets in. Only the axial stress component is shown in Fig. 32 and 33.

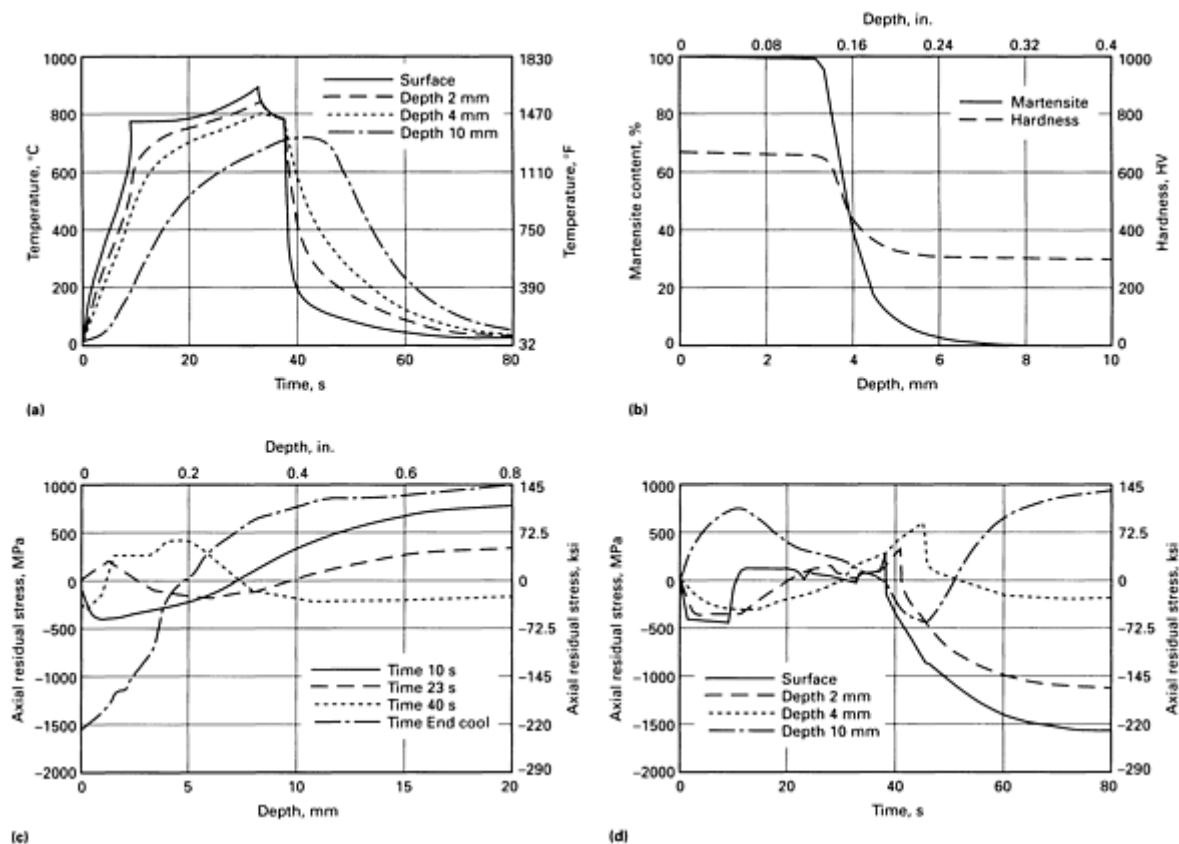


Fig. 32 Calculated values for single-shot induction hardening of a 40 mm (1.6 in.) diameter cylinder made of AISI 4140 steel. Induction frequency: 300 kHz. (a) Temperature versus time for different depths. (b) The final martensite and hardness distributions. (c) Axial residual stress versus depth at different times. (d) Axial residual stress versus time for different depths. Source: Ref 43

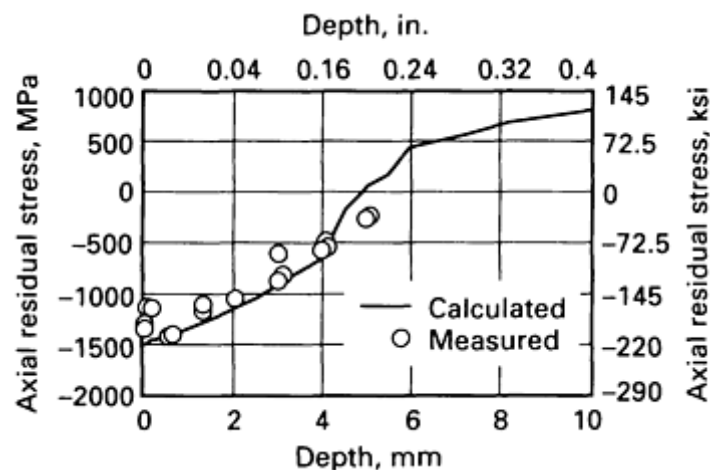


Fig. 33 Calculated and experimental axial residual stress after induction hardening versus depth for a 40 mm (1.6 in.) diameter steel specimen made of AISI 4140 steel. Source: Ref 43

The computer predictions are reasonably good as can be seen in Fig. 33. The experimental data show a smaller compressive stress at the surface than is predicted. The effect of variables such as the quenching medium and the ITh diagram has also been taken into consideration. Figure 34(a) shows how the hardening depth is increased when the transformation to austenite at elevated temperature occurs more easily. The different transformation rates can be due to different steel compositions or to different prior microstructures of the same steel. The cooling is rapid enough to transform all austenite that has formed to martensite and the residual stress distribution is only affected by the hardening

in Fig. 34(b). If the water spray is replaced by oil quenching, the calculation shows that the magnitude of the compressive stress peak is reduced while the hardening depth is the same because the depth of austenitization is the same.

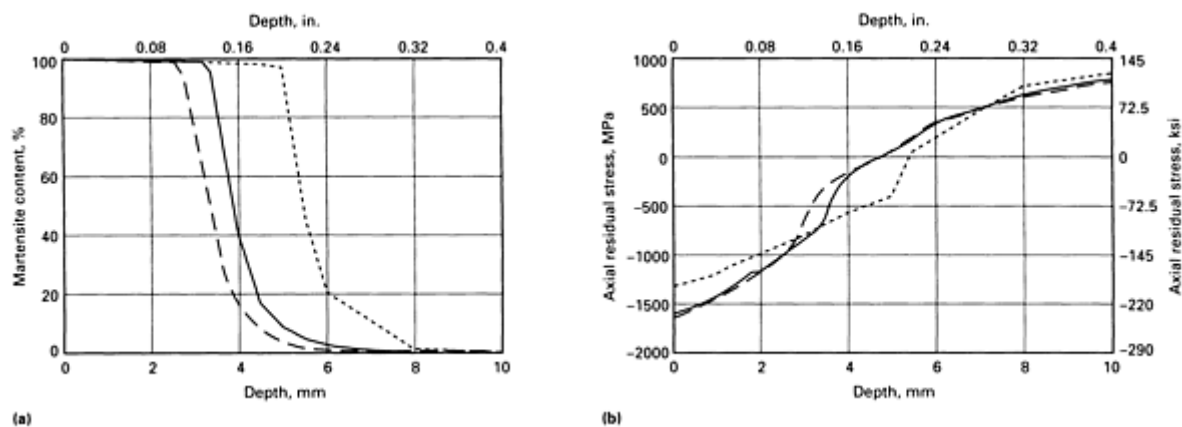


Fig. 34 Calculated values of martensite distribution and residual stress after induction hardening of a cylinder with a 40 mm (1.6 in.) diameter. The solid curve is calculated using the ITh diagram in Fig. 4, and the two others are hypothetical displacements of this diagram to shorter and longer times, respectively. This can illustrate the effect of different prior microstructures or different steel compositions. (a) Final martensite distribution. (b) Axial residual stress profiles for the three cases in (a). Source: Ref 43

Cracking and Distortion due to Hardening. There is a risk for cracking of a workpiece if large tensile stresses, transient or residual, are combined with the presence of a brittle microstructure (particularly martensite). Thermal stresses during cooling generally increase with the size of a workpiece. For phase transformation-induced stresses, geometric dimension, hardenability of the steel, and quench intensity interact in a complicated manner as has been described in earlier paragraphs. However, as a general rule it holds that the use of a more efficient cooling medium, for example, water as compared to oil, will lead to larger stresses as demonstrated in Fig. 25, 26, and 29. The presence of geometric stress raisers increases the risk of cracking.

Figure 22 indicates that tensile stresses are present at the surface when the surface transformation to martensite is complete and the core transformation is in progress. As such, there is the risk of surface cracking. However, it is shown in Fig. 25 and 26 that through hardening does not necessarily lead to tensile stresses at the surface. In case hardening, surface cracking is not to be expected (see Fig. 30, 31, and 33).

Large tensile stresses in the core at lower temperatures may lead to center cracks even if the microstructure is not martensitic. Figure 29 shows that such a situation exists for larger diameter cylinders with a martensitic surface and a ferritic-pearlitic core. Case hardening may also lead to core cracking.

Hardening is usually accompanied by distortion of a workpiece. The degree of distortion depends on the magnitude of the residual stresses. Hardening procedures that minimize transient and residual stresses are beneficial as well as the use of fixtures (press hardening). Distortion can also occur during tempering or annealing due to release of residual stresses or phase transformations during tempering as described previously in this article.

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Introduction

HARDENABILITY STEELS, or H-steels, offer a wide range of mechanical properties that depend on the development of tempered martensite after quenching and tempering. Typical room-temperature properties of quenched and tempered steels can vary as follows:

- Hardness values of 130 to 700 HV (30 kgf load)
- Tensile strengths of 400 to 2000 MPa (58 to 290 ksi)
- Yield strengths of 300 to 1800 MPa (43 to 261 ksi)
- Elongation of 8 to 28% in 50 mm (2 in.)

This broad range of properties depends on the maximum (as-quenched) hardness and the degree of softening (tempering) after quenching.

The maximum (100% martensite) hardness of heat-treated steel depends primarily on the carbon content (Table 1), until carbon levels reach about 0.50 wt%. Above 0.50 wt%, carbon has little effect on hardness but does improve hardenability. Alloying elements have little effect on the maximum hardness that can be developed in steel, but they profoundly affect the depth to which this maximum hardness can be developed in a part of specific size and shape. Thus, for a specific application, one of the first decisions to be made is what carbon level is required to obtain the desired hardness. The next step is to determine what alloy content will give the proper hardening response in the section size involved. This is not to imply that tempered martensitic steels are alike in every respect, regardless of composition, because the alloy content is responsible for differences in the preservation of strength at elevated temperatures; in abrasion resistance; in resistance to corrosion; and even, to a certain extent, in toughness. However, the similarities are sufficiently marked to permit reasonably accurate predictions of mechanical properties from hardness rather than from composition, thereby justifying the emphasis on hardenability as the most important function of the alloying elements. The basic effects of alloying elements on hardening are described in the article "Principles of Heat Treating of Steels" in this Volume.

Table 1 Steel hardness for various carbon contents and percentages of martensite for some low-alloy steels

Carbon, %	Rockwell C hardness (HRC) with martensite contents of:
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