CHAPTER 6

PROPERTIES OF ENGINEERING MATERIALS

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6.1 MATERIAL-SELECTION CRITERIA IN ENGINEERING DESIGN

The selection of materials for engineering components and devices depends upon knowledge of material properties and behavior in particular environmental states. Although a criterion for the choice of material in critically designed parts relates to the performance in a field test, it is usual in preliminary design to use appropriate data obtained from standardized tests. The following considerations are important in material selection:

1. Elastic properties: stiffness and rigidity
2. Plastic properties: yield conditions, stress-strain relations, and hysteresis
3. Time-dependent properties: elastic phenomenon (damping capacity), creep, relaxation, and strain-rate effect
4. Fracture phenomena: crack propagation, fatigue, and ductile-to-brittle transition
5. Thermal properties: thermal expansion, thermal conductivity, and specific heat
6. Chemical interactions with environment: oxidation, corrosion, and diffusion

It is good design practice to analyze the conditions under which test data were obtained and to use the data most pertinent to anticipated service conditions.

The challenge that an advancing technology imposes on the engineer, in specifying treatments to meet stringent material requirements, implies a need for a basic approach which relates properties to structure in metals. As a consequence of the mechanical, thermal, and metallurgical treatments of metals, it is advantageous to explore, for example, the nature of induced internal stresses as well as the processes of stress relief. Better material performance may ensue when particular treatments can be specified to alter the structure in metals so that the likelihood of premature failure in service is lessened. Some of the following concepts are both basic and important:

1. Lattice structure of metals: imperfections, anisotropy, and deformation mechanisms
2. Phase relations in alloys: equilibrium diagrams
3. Kinetic reactions in the solid state: heat treatment by nucleation and by diffusionless processes, precipitation hardening, diffusion, and oxidation
4. Surface treatments: chemical and structural changes in carburizing, nitriding, and localized heating
5. Metallurgical bonds: welded and brazed joints

6.2 STRENGTH PROPERTIES: TENSILE TEST AT ROOM TEMPERATURE

The yield strength determined by a specified offset, 0.2 percent strain, from a stress-strain diagram is an important and widely used property for the design of statically loaded members exhibiting elastic behavior. This property is derived from a test in which the following conditions are normally controlled: surface condition of standard specimen is specified; load is axial; the strain rate is low, i.e., about 10^{-3} in/(in·s); and grain size is known. Appropriate safety factors are applied to the yield strength to allow for uncertainties in the calculated stress and stress-concentration factors and for possible overloads in service. Since relatively small safety factors are used in critically stressed aircraft materials, a proof stress at 0.01 percent strain offset is used because this more nearly approaches the proportional limit for elastic behavior in the material. A typical stress-strain plot from a tensile test is shown in Fig. 6.1, indicating the elastic and plastic behaviors. In order to effect more meaningful comparisons in design strength properties among materials having different specific gravities, the strength property can be divided by the specific gravity, giving units of psi per pound per cubic inch.

The modulus of elasticity is a measure of the stiffness or rigidity in a material. Values of the modulus normally are not exactly determined quantities, and typical values are commonly reported for a given material. When a material is selected on the basis of a high modulus, the tendency toward whip and vibration in shaft or rod applications is reduced. These effects can lead to uneven wear. Furthermore the modulus assumes particular importance in the design of springs and diaphragms, which necessitate a definite degree of motion for a definite load. In this connection, selection of a high-modulus material can lead to a thinner cross section.

The ultimate tensile strength and the ductility, percent elongation in inches per inch or percent reduction in area at fracture are other properties frequently reported from tensile tests. These serve as qualitative measures reflecting the ability of a material in deforming plastically after being stressed beyond the elastic region. The strength properties and ductility of a material subjected to different treatments can vary widely. This is illustrated in Fig. 6.2. When the yield strength is raised by treatment to a high

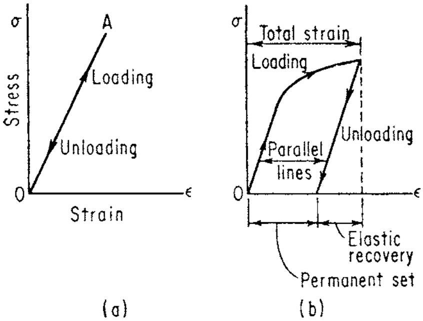


FIG. 6.1 Portions of tensile stress σ-strain ε curves in metals.[1] (*a*) Elastic behavior. (*b*) Elastic and plastic behaviors.

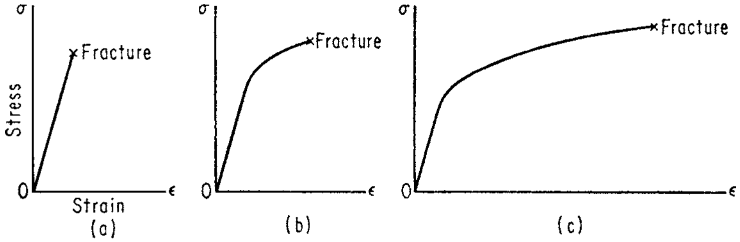


FIG. 6.2 The effects of treatments on tensile characteristics of a metal.[1] (*a*) Perfectly brittle (embrittled)—all elastic behavior. (*b*) Low ductility (hardened)—elastic plus plastic behaviors. (*c*) Ductile (softened)—elastic plus much plastic behaviors.

value, i.e., greater than two-thirds of the tensile strength, special concern should be given to the likelihood of tensile failures by small overloads in service. Members subjected solely to compressive stress may be made from high-yield-strength materials which result in weight reduction.

When failures are examined in statically loaded tensile specimens of circular section, they can exhibit a cup-and-cone fracture characteristic of a ductile material or on the other extreme a brittle fracture in which little or no necking down is apparent. Upon loading the specimen to the plastic region, axial, tangential, and radial stresses are induced. In a ductile material the initial crack forms in the center where the triaxial stresses become equally large, while at the surface the radial component is small and the deformation is principally by biaxial shear. On the other hand, an embrittled material exhibits no such tendency for shear and the fracture is normal to the loading axis. Some types of failures in round tensile specimens are shown in Fig. 6.3.

The properties of some wrought metals presented in Table 6.1 serve to show the significant differences relating to alloy content and treatment. Section 6.17 gives more information.

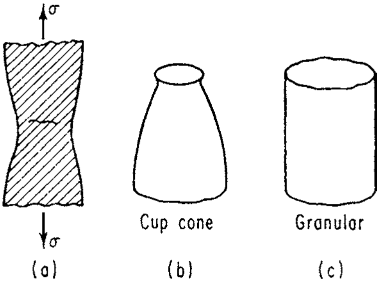


FIG. 6.3 Typical tensile-test fractures.[1] (*a*) Initial crack formation. (*b*) Ductile material. (*c*) Brittle material.

TABLE 6.1 Room-Temperature Tensile Properties for Some Wrought Metals

Metal	Condition	Ultimate tensile σ_{ult}, psi	Yield strength σ_y, psi	% elongation	Modulus of elasticity, psi	Density, lb/cu in.
Aluminum (pure)..	Annealed	13,000	5,000	35	10,000,000	0.098
7075T6...........	Heat-treated	76,000	67,000	11	10,400,000	0.101
Copper (pure).....	Annealed	32,000	10,000	45	17,000,000	0.321
Cu-Be(2%).......	Heat-treated	200,000	150,000	2	19,000,000	0.297
Magnesium	Annealed	33,000	18,000	17	6,500,000	0.064
Mg-Al(8.5)A780A.	Strain relieved	44,000	31,000	18	6,500,000	0.065
Nickel (pure).....	Annealed	65,000	20,000	40	30,000,000	0.321
K Monel.........	Heat-treated	190,000	140,000	5	....	0.306
Titanium (pure)...	Annealed	59,000	40,000	28	15,000,000	0.163
Ti-Al(6)-V(14)α-β.	Heat-treated	170,000	150,000	7	15,000,000	0.163

The tensile properties of metals are dependent upon the rate of straining, as shown for aluminum and copper in Fig. 6.4, and are significantly affected by the temperature, as shown in Fig. 6.5. For high-temperature applications it is important to base design on different criteria, notably the stress-rupture and creep characteristics in metals, both of which are also time-dependent phenomena. The use of metals at low temperatures requires a consideration of the possibility of brittleness, which can be measured in the impact test.

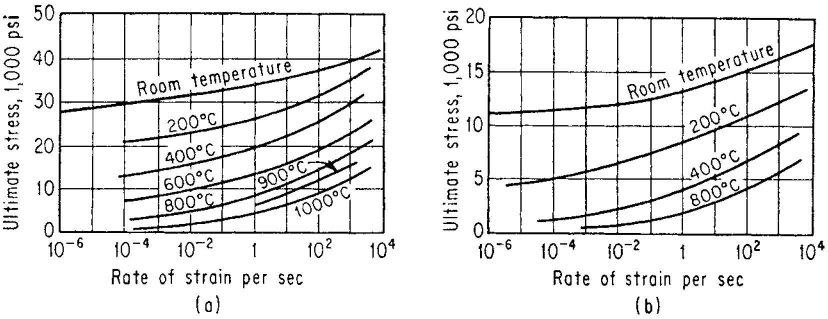


FIG. 6.4 Effects of strain rates and temperatures on tensile-strength properties of copper and aluminum.[1] (*a*) Copper. (*b*) Aluminum.

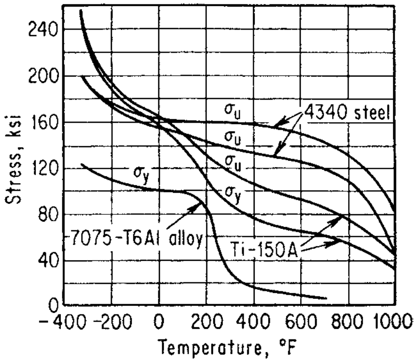


FIG. 6.5 Effects of temperatures on tensile properties. σ_u = ultimate tensile strength; σ_y = yield strength.

6.3 ATOMIC ARRANGEMENTS IN PURE METALS: CRYSTALLINITY

The basic structure of materials provides information upon which properties and behavior of metals may be generalized so that selection can be based on fundamental considerations. A regular and periodic array of atoms (in common metals whose atomic diameters are about one hundred-millionth of an inch) in space, in which a unit cell is the basic structure, is a fundamental characteristic of crystalline solids. Studies of these structures in metals lead to some important considerations of the behaviors in response to externally applied forces, temperature changes, as well as applied electrical and magnetic fields.

The body-centered cubic (bcc) cell shown in Fig. 6.6*a* is the atomic arrangement characteristic of αFe, W, Mo, Ta, ßTi, V, and Nb. It is among this class of metals that transitions from ductile to brittle behavior as a function of temperature are significant to investigate. This structure represents an atomic packing density where about 66 percent of the volume is populated by atoms while the remainder is free space. The elements Al, Cu, γFe, Ni, Pb, Ag, Au, and Pt have a closer packing of atoms in space constituting a face-centered cubic (fcc) cell shown in Fig. 6.6*b*. Characteristic of these are ductility properties which in many cases extend to very low temperatures. Another structure, common to Mg, Cd, Zn, αTi, and Be, is the hexagonal close-packed (hcp) cell in Fig. 6.6*c*. These metals are somewhat more difficult to deform plastically than the materials in the two other structures cited above.

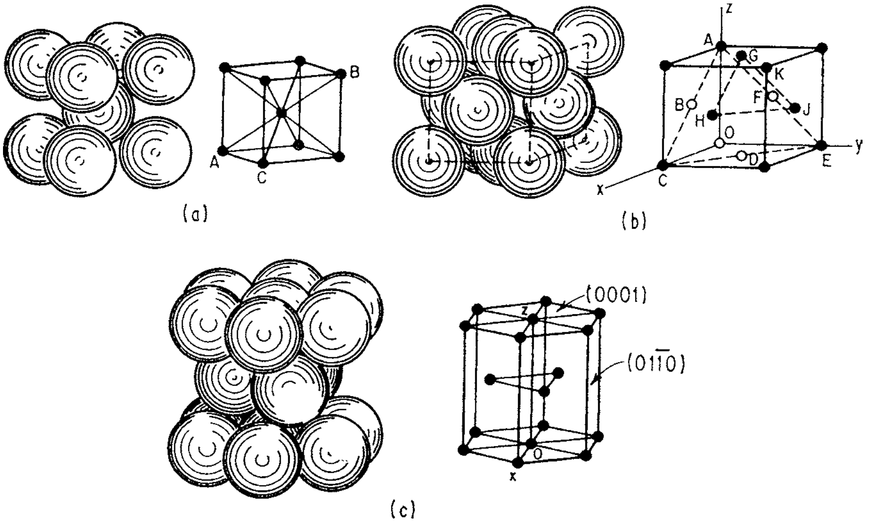


FIG. 6.6 Cell structure. (*a*) Body-centered cubic (bcc) unit cell structure. (*b*) Face-centered cubic (fcc) unit cell structure. (*c*) Hexagonal close-packed (hcp) unit cell structure.

It is apparent, from the atomic arrays represented in these structures, that the closest approach of atoms can vary markedly in different crystallographic directions. Properties in materials are anisotropic when they show significant variations in different directions. Such tendencies are dependent on the particular structure and can be especially pronounced in single crystals (one orientation of the lattices in space). Some examples of these are given in Table 6.2. When materials are processed so that

TABLE 6.2 Examples of Anisotropic Properties in Single Crystals

Property	Material and structure	Properties relation
Elastic module E in tension	αFe (bcc)	$E_{\{AB\}} \sim 2.2E_{\{AC\}}$
Elastic module G in shear	Ag (fcc)	$G_{\{OC\}} \sim 2.3G_{\{OK\}}$
Magnetization	αFe (bcc)	Ease of magnetization
Thermal expansion coefficient—α	Zn (hcp)	$\alpha_{\{OZ\}} \sim 4\alpha_{\{OA\}}$

their final grain size is large (each grain represents one orientation of the lattices) or that the grains are preferentially oriented, as in extrusions, drawn wire, rolled sheet, sometimes in forgings and castings, special evaluation of anisotropy should be made. In the event that directional properties influence design considerations, particular attention must be given to metallurgical treatments which may control the degree of anisotropy. The magnetic anisotropy in a single crystal of iron is shown in Fig. 6.7.

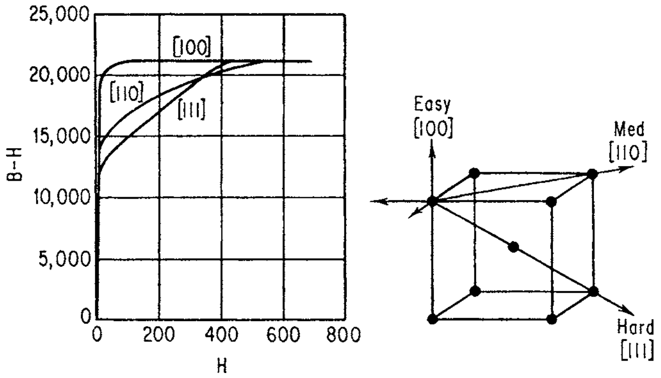


FIG. 6.7 Magnetic anisotropy in a single crystal of iron[2]: $I = (B - H)/4\pi$, where I = intensity of magnetization; B = magnetic induction, gauss; H = field strength, oersteds.

6.4 PLASTIC DEFORMATION OF METALS

When metals are externally loaded past the elastic limit, so that permanent changes in shape occur, it is important to consider the induced internal stresses, property changes, and the mechanisms of plastic deformation. These are matters of practical consideration in the following: materials that are to be strengthened by cold work, machining of cold-worked metals, flow of metals in deep-drawing and impact extrusion operations, forgings where the grain flow patterns may affect the internal soundness, localized surface deformation to enhance fatigue properties, and cold working of some magnetic materials. Experimental studies provide the key by which important phenomena are revealed as a result of the plastic-deformation process. These studies indicate some treatments that may be employed to minimize unfavorable internal-stress distributions and undesirable grain-orientation distributions.

Plastic deformation in metals occurs by a glide or slip process along densely packed planes fixed by the particular lattice structure in a metal. Therefore, an applied

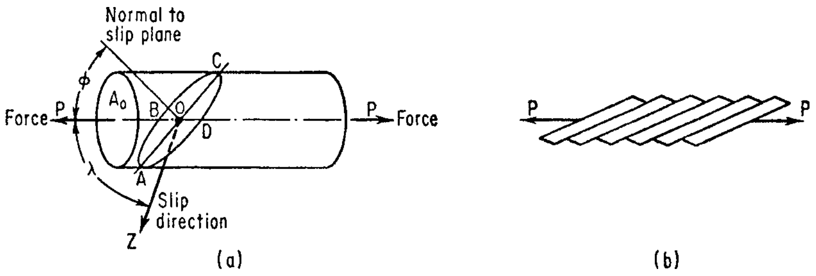


FIG. 6.8 Slip deformation in single crystals. (*a*) Resolved shear stress = $P/A_0 \cos \phi \cos \lambda$. *ABCD* is plane of slip. *OZ* is slip direction. (*b*) Sketch of single crystal after yielding.

load is resolved as a shear stress, on those particular glide elements (planes and directions) requiring the least amount of deformation work on the system. An example of this deformation process is shown in Fig. 6.8. Face-centered cubic (fcc) structured metals, such as Cu, Al, and Ni, are more ductile than the hexagonal structured metals, such as Mg, Cd, and Zn, at room temperature because in the fcc structure there are four times as many possible slip systems as in a hexagonal structure. Slip is initiated at much lower stresses in metals than theoretical calculations based on a perfect array of atoms would indicate. In real crystals there are inherent structural imperfections termed dislocations (atomic misfits) as shown in Fig. 6.9, which account for the observed yielding phenomenon in metals. In addition, dislocations are made mobile by mechanical and thermal excitations and they can interact to result in strain hardening of metals by cold work. Strength properties can be increased while the ductility is

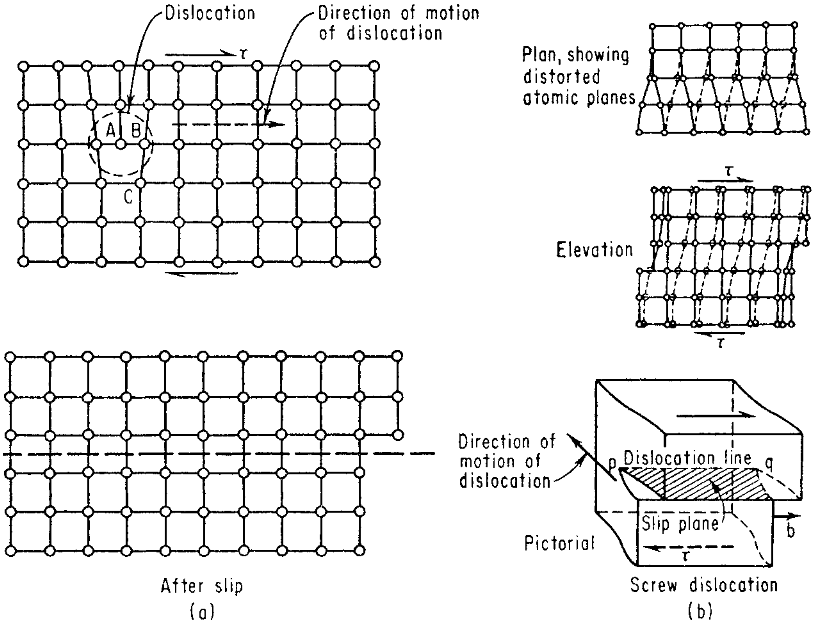


FIG. 6.9 Edge and screw dislocations as types of imperfections in metals.[2] (*a*) Edge dislocation. (*b*) Screw dislocation.

decreased in those metals which are amenable to plastic deformation. Cold working of pure metals and single-phase alloys provides the principal mechanism by which these may be hardened.

The yielding phenomenon is more nonhomogeneous in polycrystalline metals than in single crystals. Plastic deformation in polycrystalline metals initially occurs only in those grains in which the lattice axes are suitably oriented relative to the applied load axis, so that the critically resolved shear stress is exceeded. Other grains rotate and are dependent on the orientation relations of the slip systems and load application; these may deform by differing amounts. As matters of practical considerations the following effects result from plastic deformation:

1. Materials become strain-hardened and the resistance to further strain hardening increases.
2. The tensile and yield strengths increase with increasing deformation, while the ductility properties decrease.
3. Macroscopic internal stresses are induced in which parts of the cross section are in tension while other regions have compressive elastic stresses.
4. Microscopic internal stresses are induced along slip bands and grain boundaries.
5. The grain orientations change with cold work so that some materials may exhibit different mechanical and physical properties in different directions.

The Bauschinger effect in metals is related to the differences in the tensile and compressive yield-strength values, as shown at σ_T and σ_C in Fig. 6.10 when a ductile metal undergoes stress reversal. This change in polycrystalline metals is the result of the nonuniform character of deformation and the different pattern of induced macrostresses. These grains, in which the induced macrostresses are compressive, will yield at lower values upon the application of a reversed compressive stress because they are already part way toward yielding. This effect is encountered in cold-rolled metals where there is lateral contraction together with longitudinal elongation; this accounts for the decreased yield strength in the lateral direction compared with the increased longitudinal yield strength.

The control of metal flow is important in deep-drawing operations performed on sheet metal. It is desirable to achieve a uniform flow in all directions. Cold-rolling

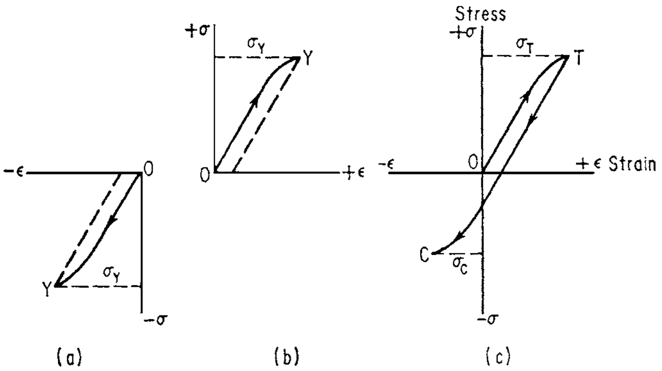


FIG. 6.10 The Bauschinger effect. (*a*) Compression. (*b*) Tension. The application of a compressive stress (*a*) or a tensile stress (*b*) results in the same value of yield strength *y*. (*c*) Stress reversal. A reversal of stress $O \rightarrow T \rightarrow C$ results in different values of tensile and compressive yield strengths; $\sigma_T \neq \sigma_C$.

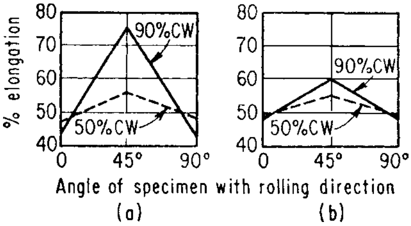


FIG. 6.11 Directionality in ductility in cold-worked and annealed copper sheet.[1] (*a*) Annealed at 1470°F. (*b*) Annealed at 750°F. The variation in ductility with direction for copper sheet is dependent on both the annealing temperature and the amount of cold work (percent CW) prior to annealing.

sheet metal produces a structure in which the grains have a preferred orientation. This characteristic can persist, even though the metal is annealed (recrystallized), resulting in directional properties as shown in Fig. 6.11. A further consequence of this directionality, associated with the deep-drawing operation, is illustrated in Fig. 6.12. The important factors, involved with the control of earing tendencies, are the fabrication practices of the amount of cold work in rolling and duration and temperatures of annealing. When grain textural problems of this kind are encountered, they can be studied by x-ray diffraction techniques and reasonably controlled by the use of optimum cold-working and annealing schedules.

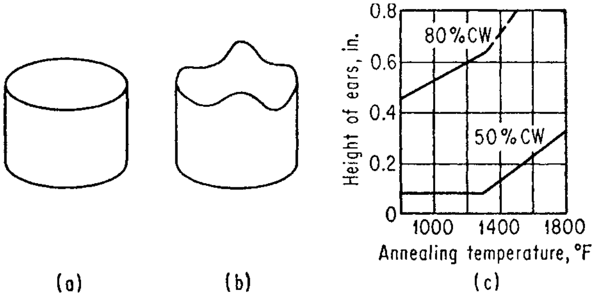


FIG. 6.12 The earing tendencies in cup deep drawn from sheet. (*a*) Uniform flow, nonearing. (*b*) Eared cup, the result of nonuniform flow. (*c*) Height of ears in deep-drawn copper cups related to annealing temperatures and amount of cold work.

6.5 PROPERTY CHANGES RESULTING FROM COLD-WORKING METALS

Cold-working metals by rolling, drawing, swaging, and extrusion is employed to strengthen them and/or to change their shape by plastic deformation. It is used principally on ductile metals which are pure, single-phase alloys and for other alloys which will not crack upon deformation. The increase in tensile strength accompanied by the decrease in ductility characteristic of this process is shown in Fig. 6.13. It is to be noted, especially from the yield-strength curve, that the largest rates of change occur during the initial amounts of cold reduction.

The variations in the macrostresses induced in a cold-drawn bar, illustrated in Fig. 6.14*a*, show that tensile stresses predominate at the surface. The equilibrium state of macrostresses throughout the cross section is altered by removing the surface layers in machining, the result of which may be warping in the machined part. It may be possible, however, to stress-relieve cold-worked metals, which generally have better machinability than softened (annealed) metals, by heating below the recrystallization

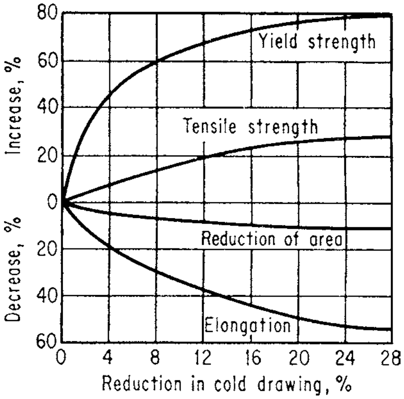


FIG. 6.13 Effect of cold drawing on the tensile properties of steel bars of up to 1-in cross section having tensile strength of 110,000 lb/in^2 or less before cold drawing.[3]

temperature. A typical alteration in the stress distribution, shown in Fig. 6.14*b*, is achieved so that the warping tendencies on machining are reduced, without decreasing the cold-worked strength properties. This stress-relieving treatment may also inhibit season cracking in cold-worked brasses subjected to corrosive environments containing amines. Since stressed regions in a metal are more anodic (i.e., go into solution more readily) than unstressed regions, it is often important to consider the relieving of stresses so that the designed member is not so likely to be subjected to localized corrosive attack.

Changes in electrical resistivity, elastic springback, and thermoelectric force resulting from cold work can be altered by a stress-relieval treatment, in a temperature range from *A* to *B*, as shown in Fig. 6.15. However, the grain flow pattern (preferred orientation) produced by cold working can be changed only by heating the metal to a temperature at which recrystallized stress-free grains will form.

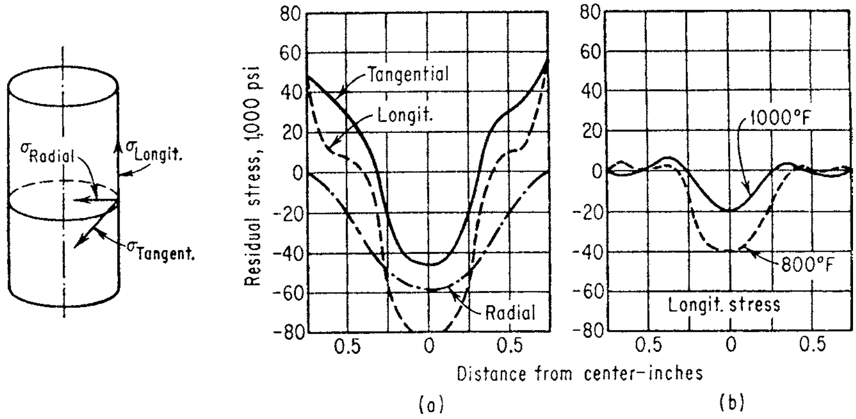


FIG. 6.14 Residual stress.[3] (*a*) In a cold-drawn steel bar 1½ in in diameter 20 percent cold-drawn, 0.45 percent C steel. (*b*) After stress-relieving bar.

Residual tensile stresses at the surface of a metal promote crack nucleation in the fatigue of metal parts. The use of a localized surface deformation treatment by shot peening, which induces compressive stresses in the surface fibers, offers the likelihood of improvement in fatigue and corrosion properties in alloys. Shot-peening a forging flash line in high-strength aluminum alloys used in aircraft may also lessen the tendency toward stress-corrosion cracking. The effectiveness of this localized surface-hardening treatment is dependent on both the nature of surface discontinuities formed by shot-peening and the magnitude of compressive stresses induced at the surface.

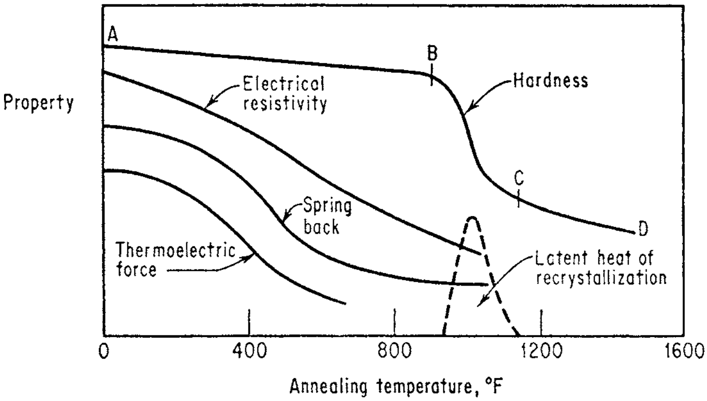


FIG. 6.15 The property changes in 95 percent cold-worked iron with heating temperatures (1 h). The temperature intervals[4]—*A* → *B*, stress relieval; *B* → *C*, recrystallization; and *C* → *D*, grain growth—signify the important phenomena occurring.

6.6 THE ANNEALING PROCESS

Metals are annealed in order to induce softening for further deformation, to relieve residual stresses, to alter the microstructure, and, in some cases after electroplating, to expel by diffusion gases entrapped in the lattice. The process of annealing, the attaining of a strain-free recrystallized grain structure, is dependent mainly on the temperature, time, and the amount of prior cold work. The temperature indicated at C in Fig. 6.15 results in the complete annealing after 1 h of the 95 percent cold-worked iron. Heating beyond this temperature causes grains to grow by coalescence, so that the surface-to-volume ratio of the grains decreases together with decreasing the internal energy of the system. As the amount of cold work (from the originally annealed state) decreases, the recrystallization temperature increases and the recrystallized grain size increases. When a metal is cold-worked slightly (less than 10 percent) and subsequently annealed, an undesirable roughened surface forms because of the abnormally large grain size (orange-peel effect) produced. These aspects of grain-size control in the annealing process enter in material specifications.

The annealing of iron-carbon-base alloys (steels) is accomplished by heating alloys of eutectoid and hypoeutectoid compositions (0.8 percent C and less in plain carbon steels) to the single-phase region; austenite, as shown in Fig. 6.16, above the transition line *GS*; and for hypereutectoid alloys (0.8 to 2.0 percent C) between the transition lines *SK* and *SE* in Fig. 6.16; followed by a furnace cool at a rate of about 25°F per hour to below the eutectoid temperature *SK*. In the annealed condition, a desirable distribution of the equilibrium phases is thereby produced. A control of the microstructure is manifested by this process in steels. Grain-size effects are principally controlled by the high-temperature treatment, grain sizes increasing with increasing temperatures, and in some cases minor impurity additions such as vanadium inhibit grain coarsening to higher temperatures. These factors of grain-size control enter into the considerations of hardening steels by heat treatment.

The control of the atmosphere in the annealing furnace is desirable in order to prevent gas-metal attack. Moisture-free neutral atmospheres are used for steels which oxidize readily. When copper and its alloys contain oxygen, as oxide, it is necessary to keep the hydrogen content in the atmosphere to a minimum. At temperatures lower

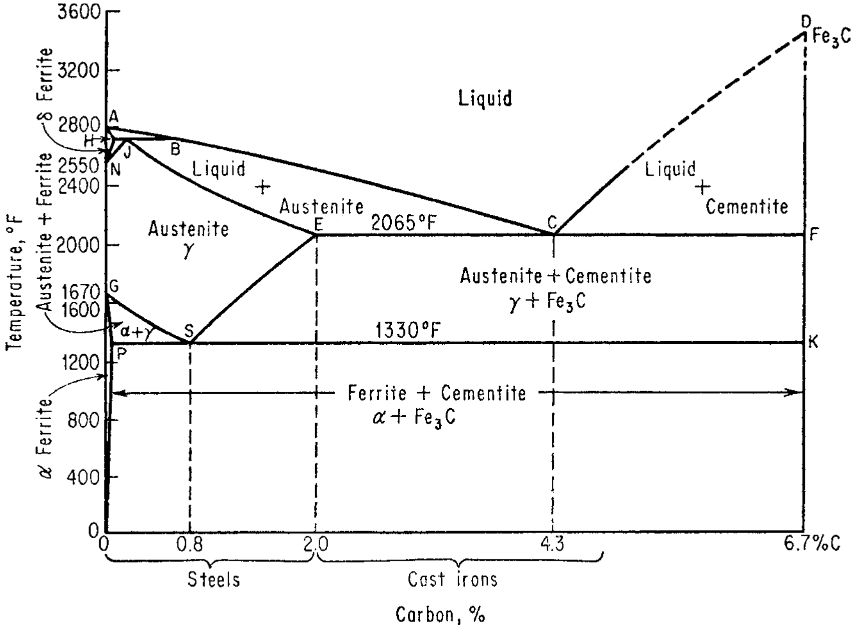


FIG. 6.16 The iron-carbon phase diagram.[4]

than 900°F, the hydrogen should not exceed 1 percent, and as the temperature is increased the hydrogen content should be reduced in order to prevent hydrogen embrittlement. In nickel and its alloys the atmosphere must be free from sulfur and slightly reducing by containing 2 percent or more of CO. Some aluminum alloys containing magnesium are affected by high-temperature oxidation in annealing (and heat treatment) and therefore require atmosphere control.

It is a characteristic property that strengths in all metals decrease with increasing temperatures. The coalescence of precipitate particles is one factor involved, so that material specifications for high-temperature use are concerned with alloy compositions that form particles having lower solubility and lower mobility. A second factor is concerned with the mobility of dislocations which increases at higher temperatures. Since strain hardening is reasoned to be due to the interaction of dislocations, then by the proper additions of solid-solution alloying elements that impede dislocations, resistance to softening will increase at the high temperature. The recrystallization temperature of iron is raised by the addition of 1 atomic percent of Mn, Cr, V, W, Cb, Ta in the same order in which the atomic size of the alloying elements differs from that of iron. The practical implications of these basic atomic considerations are important in selecting metals for high-temperature service.

6.7 THE PHASE DIAGRAM AS AN AID TO ALLOY SELECTION

Phase diagrams, which are determined experimentally and are based upon thermodynamic principles, are temperature-composition representations of slowly cooled alloys (annealed state). They are useful for predicting property changes with composition and selecting feasible fabrication processes. Phase diagrams also indicate the possible response of alloys to hardening by heat treatment. Shown on these diagrams are first-order phase transitions and the phases present. In two-phase regions, the compositions of each phase are shown on the phase boundary lines and the relative amounts of each phase present can be determined by a simple lever relation at a given temperature.

The particular phases that are formed in a system are governed principally by the physical interactions of valence electrons in the atoms and secondarily by atomic-size factors. When two different atoms in the solid state exist on, or where one is in, an atomic lattice, the phase is a solid solution (e.g., γ austenite phase in Fig. 6.16) analogous to a miscible liquid solution. When the atoms are strongly electropositive and strongly electronegative to one another, an intermetallic compound is formed (e.g., Fe_3C, cementite). The two atoms are electronically indifferent to one another and a phase mixture issues (e.g., $\alpha + Fe_3C$) analogous to the immiscibility of water and oil.

The thermodynamic criteria for a first-order phase change, indicated by the solid lines on the phase diagram, are that, at the transition temperature, (1) the change in Gibbs's free energy for the system is zero, (2) there is a discontinuity in entropy (a latent heat of transformation and a discontinuous change in specific heat), and (3) there is a discontinuous change in volume (a dilational effect).

In the selection of alloys for sound castings, particular attention is given that part of the system where the liquidus line (*ABCD*) goes through a minimum. For alloys between the composition limits of $E \rightarrow F$, a eutectic reaction occurs at 2065°F such that

$$\text{liq } C = \gamma + \text{Fe}_3\text{C}$$

It is for this reason in the iron-carbon system that cast irons are classified as having carbon contents greater than 2 percent. For purposes of controlling grain size, obtaining sound castings free from internal porosities (blowholes) and internal shrinkage cavities, and possessing good mold-filling characteristics, alloys and low-melting solutions are chosen near the eutectic composition (i.e., at *C*). Aluminum-silicon die-casting alloys have a composition of about 11 percent silicon near the eutectic composition. Special considerations need be given to the properties and structures in cast irons because the Fe_3C phase is thermodynamically unstable and decomposition to graphite (in gray cast irons) may result.

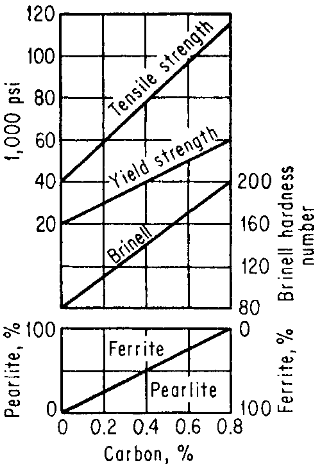


FIG. 6.17 Relation of mechanical properties and structure to carbon content of slowly cooled carbon steels.[4]

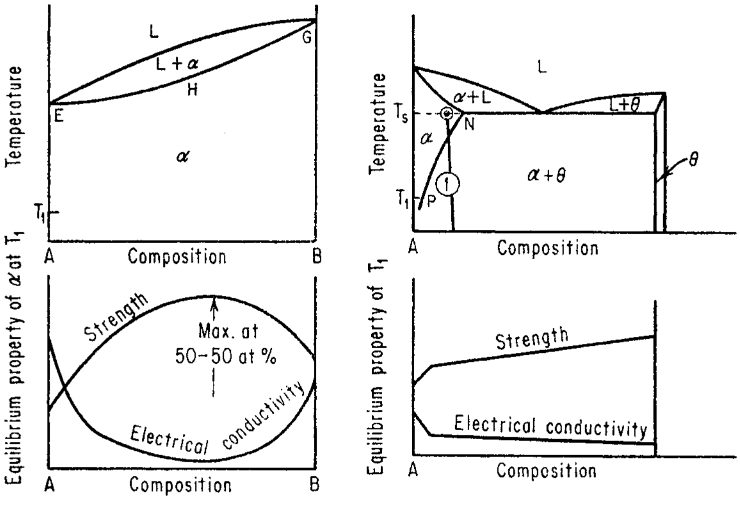
The predominant phase-diagram characteristic in steels is the eutectoid reaction, in the solid state, along *GSE* where $\gamma_s = \alpha + Fe_3C$ (pearlite) at 1330°F. Steels are therefore classified as alloys in the Fe-C system having a carbon content less than 2.0 percent C; and furthermore, according to their applications, compositions are designated as hypoeutectoid ($C < 0.8$ percent), eutectoid ($C = 0.8$ percent), and hypereutectoid ($C < 2$ percent > 0.8 percent). Since the slowly cooled room-temperature structures of steels contain a mechanical aggregate of the ferrite and Fe_3C-cementite phases, the property relations vary linearly as shown in Fig. 6.17. The ductility decreases with increasing carbon contents.

Some important characteristics of the equilibrium phases in steels are listed below:

Phase	Characteristics
α ferrite	Low C solubility (less than 0.03%) bcc, ductile, and ferromagnetic below 1440°F
Fe_3C, cementite	Intermetallic compound, orthorhombic, hard, brittle, and fixed composition at 6.7% C
γ austenite	Can dissolve up to 2% C in solid solution, fcc, nonmagnetic, and in this region annealing, hardening, forging, normalizing, and carburizing processes take place

Low-carbon alloys can be readily worked by rolling, drawing, and stamping because of the predominant ductility of the ferrite. Wires for suspension cables having a carbon content of about 0.7 percent are drawn at about 1100°F (patenting) because of the greater difficulty, in room-temperature deformation, caused by the presence of a relatively large amount of the brittle Fe_3C phase.

Extensive substitutional solid-solution alloys form in binary systems when they have similar chemical characteristics and atomic diameters in addition to having the same lattice structure. Such alloys include copper-nickel (monel metal being a commercially useful one), chromium-molybdenum, copper-gold, and silver-gold (jewelry alloys). The phase diagram and the equilibrium-property changes for this system are shown in Fig. 6.18*a*. Each pure element is strengthened by the addition of the other, whereby the strongest alloy is at an equal atom concentration. There are no first-order phase changes up to the start of melting (the solidus line *EHG*), so that these are not hardened by heat treatment but only by cold work. The electrical conductivity decreases



(a) (b)

FIG. 6.18 Binary systems.[4] (*a*) Complete solid-solubility phase diagram. (*b*) Partial solid-solubility part of phase diagram. α is a substitutional solid solution, a phase with two different atoms on the same lattice. In the AlCu system θ is an intermediate phase (precipitant) having a composition nominally of $CuAl_2$.

from each end of the composition axis. Because of the presence of but one phase, these alloys are selected for their resistance to electrochemical corrosion. High-temperature-service metals are alloys which have essentially a single-phase solid solution with minor additions of other elements to achieve specific effects.

Another important system is one in which there are present regions of partial solid solubility as shown in Fig. 6.18*b* together with equilibrium-property changes. An important consideration in the selection of alloys containing two or more phases is that galvanic-corrosion attack may occur when there exists a difference in the electromotive potential between the phases in the environmental electrolyte. Sacrificial galvanic protection of the base metal in which the coating is more anodic than the base metal is used in zinc-plating iron-base alloys (galvanizing alloys). The intimate mechanical mixture of phases which are electrochemically different may result in pitting corrosion, or even more seriously, intergranular corrosion may result if the alloy is improperly treated by causing localized precipitation at grain boundaries.

Heat treatment by a precipitation-hardening process is indeed an important strengthening mechanism in particular alloys such as the aircraft aluminum-base, copper-beryllium, magnesium-aluminum, and alpha-beta titanium alloys (Ti, Al, and V). In these alloys a distinctive feature is that the solvus line *NP* in Fig. 6.18*b* shows decreasing solid solubility with decreasing temperature. This in general is a necessary, but not necessarily sufficient, condition for hardening by precipitation since other thermodynamic conditions as well as coherency relations between the precipitated phases must prevail. The sequence of steps for this process is as follows: An alloy is solution heat-treated to a temperature T_s, rapidly quenched so that a metastable supersaturated solid solution is attained, and then aged at experimentally determined temperature-time aging treatments to achieve desired mechanical properties. This is the principal hardening process for those particular nonferrous alloys (including Inconels) which can respond to a precipitation-hardening process.

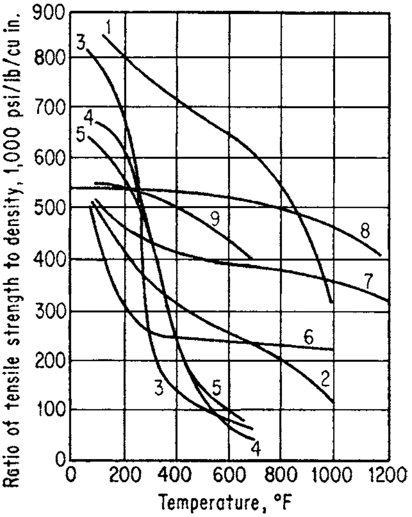


FIG. 6.19 Approximate comparison of materials on a strength-weight basis from room temperature to 1000°F.[6] (1) T_1, 8Mn; (2) 9990 T_1; (3) 75S, T6Al; (4) 24S, T4Al; (5) AZ31A Mg; (6) annealed stainless steel; (7) half-hard stainless steel; (8) Inconel X; (9) glass-cloth laminate.

The engineer is frequently concerned with the strength-to-density ratio of materials and its variation with temperature. A number of materials are compared on this basis in Fig. 6.19 in which the alloys designated by curves 1, 3, 4, 5, and 8 are heat-treatable nonferrous alloys.

6.8 HEAT-TREATMENT CONSIDERATIONS FOR STEEL PARTS

The heat-treating process for steel involves heating to the austenite region where the carbon is soluble, cooling at specific rates, and tempering to relieve some of the stress which results from the transformation. Some important considerations involved in specifying heat-treated parts are strength properties, warping tendencies, mass effects (hardenability), fatigue and impact properties, induced transformation stresses, and the use of surface-hardening processes for enhanced wear resistance. Temperature and time factors affect the structures issuing from the decomposition of austenite; for a eutectoid steel (0.8 percent C) they are as follows:

Decomposition product from γ	Structure	Mechanism	Temperature range, °F
Pearlite	Equilibrium ferrite + Fe_3C	Nucleation; growth	1300–1000
Bainite	Nonequilibrium ferrite + carbide	Nucleation; growth	1000–450
Martensite	Supersaturated tetragonal lattice	Diffusionless	M_s (≤450)

The tensile strength of a slowly cooled (annealed) eutectoid steel containing a coarse pearlite structure is about 120,000 lb/in^2. To form bainite, the steel must be cooled rapidly enough to escape pearlite transformation and must be kept at an intermediate temperature range to completion, from which a product having a tensile strength of about 250,000 lb/in^2 can be formed. Martensite, the hardest and most brittle product, forms independently of time by quenching rapidly enough to escape higher-temperature transformation products. The carbon atoms are trapped in the martensite, causing its lattice to be highly strained internally; its tensile strength is in excess of 300,000 lb/in^2. Isothermal transformation characteristics of all steels show the temperature-time and transformation products as in Fig. 6.20, where the lines indicate the start and end of transformation. On the temperature-time coordinates, involved cooling curves can be superimposed which show that, for a 1-in round water-quenched specimen, mixed products will be present. The outside will be martensite and the middle sections will contain pearlite. Alloying elements are added to steels principally to retard pearlite transformation either so that less drastic quenching media can be used or to ensure more uniform hardness throughout. This retardation is shown in Fig. 6.20*b* for an SAE 4340 steel containing alloying additions of Ni, Cr, or Mo and 0.4 percent C.

The carbon content in steels is the most significant element upon which selection for the maximum attainable hardness of the martensite is based. This relation is shown in Fig. 6.21. Since the atomic rearrangements involved in the transformation from the fcc austenite to the body-centered-tetragonal martensite result in a volumetric expansion, on cooling, of about 1 percent (for a eutectoid steel) nonuniform stress patterns can be induced on transformation. As cooling starts at the surface, by the normal process of heat transfer, parts of a member can be expanding, because of transformation, while further inward normal contraction occurs on the cooling austenite. The danger of cracking and distortion (warping) as a consequence of the steep thermal gradients and the transformation involved in hardening steels can be eliminated by using good design and the selection of the proper alloys. Where section size, time factors, and alloy content (as it affects transformation curves) permit, improved practices by martempering shown by *EFGH* in Fig. 6.20, followed by tempering or austempering shown by *EFK*, may be feasible and are worthy of investigation for the particular alloy used.

Uniform mass distribution and the elimination of sharp corners (potential stress raisers) by the use of generous fillets are recommended. Some design features pertinent to the elimination of quench cracks and the minimization of distortion by warping are illustrated by Fig. 6.22 in which *a*, *c*, and *e* represent poor designs in comparison with the suggested improvements apparent in *b, d,* and *f.*

Steels are tempered to relieve stresses, to impart ductility, and to produce a desirable microstructure by a reheating process of the quenched member. The tempering

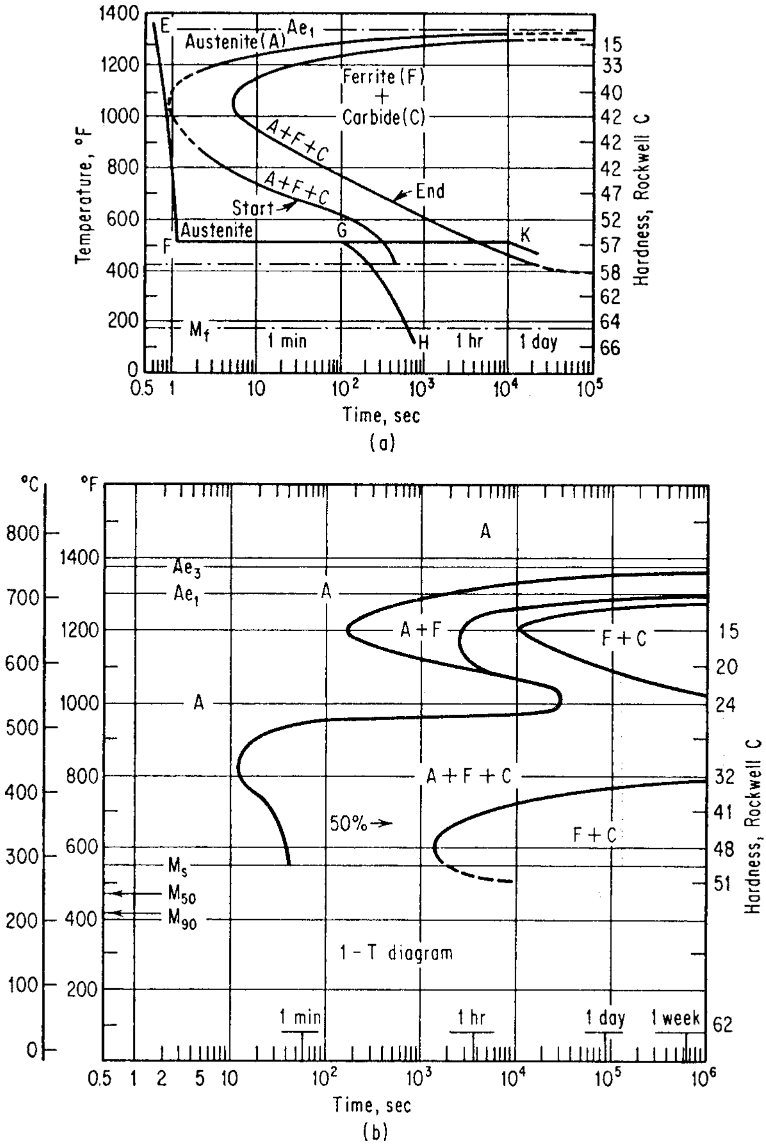


FIG. 6.20 Isothermal transformation diagram.[5] (*a*) Eutectoid carbon steel. (*b*) SAE 4340 steel. M_s = start of martensite temperature; M_f = finish of martensite temperature; *EFGH* = martempering (follow by tempering); *EFK* = austempering. *A*, austenite; F, ferrite; *C*, carbide; M_s and M_f, temperatures for start and finish of martensite transformation; M_{50} and M_{90}, temperatures for 50% and 90% of martensite transformation.

process is dependent on the temperature, time, and alloy content of the steel. Different alloys soften at different rates according to the constitutionally dependent diffusional structure. The response to tempering for 1 h for three different steels of the same carbon content is shown in Fig. 6.23. In addition, the tempering characteristics of a high-speed

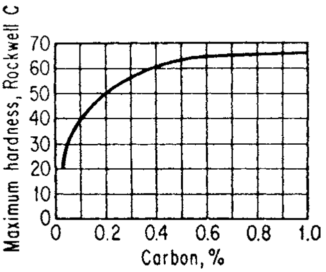


FIG. 6.21 Relation of maximum attainable hardness of quenched steels to carbon content.[7]

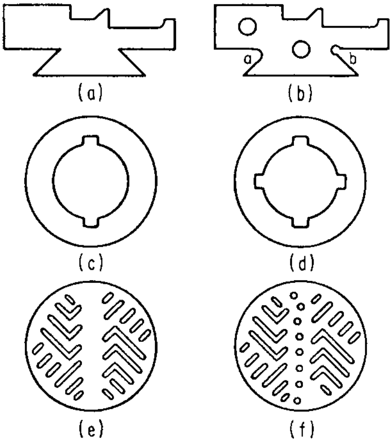


FIG. 6.22 Examples of good (*b*, *d*, and *f*) and bad (*a*, *c*, and *e*) designs for heat-treated parts.[8] (1) *b* is better than *a* because of fillets and more uniform mass distribution. (2) In *c* cracks may form at keyways. (3) Warping may be more pronounced in *e* than in *f*, which are blanking dies.[8]

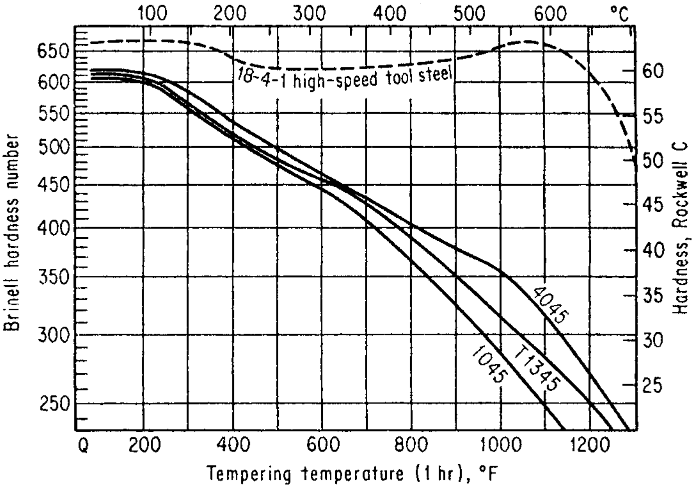


FIG. 6.23 Effect of tempering temperature on the hardnesses of SAE 1045, T1345, and 4045 steels. In the high-speed tool steel 18-4-1, secondary hardening occurs at about 1050°F.[9]

tool steel, 18 percent W, 4 percent Cr, 1 percent V, and 0.9 percent C, are shown to illustrate the secondary hardening at about 1050°F. The pronounced tendency for high-carbon steels to retain austenite on transformation normally has deleterious effects on dimensional stability and fatigue performance. In high-speed tool steel, the secondary hardening is due to the transformation of part of the retained austenite to newly transformed martensite. The structure contains tempered and untempered martensite with perhaps some retained austenite. Multiple tempering treatments on this type of steel produce a more uniform product.

In low-alloy steels where the carbon content is above 0.25 percent, there may be a tempering-temperature interval at about 450 to 650°F, during which the notch impact strength goes through a minimum. This is shown in Fig. 6.24 and is associated with the formation of an embrittling carbide network (ϵ carbide) about the martensite subgrain boundaries. Tempering is therefore carried out up to 400°F where the parts are to be used principally for wear resistance, or in the range of 800 to 1100°F where greater toughness is required. In the nomenclature of structural steels adopted by the Society of Automotive Engineers and the American Iron and Steel Institute, the first two numbers designate the type of steel according to the principal alloying elements and the last two numbers designate the carbon content:

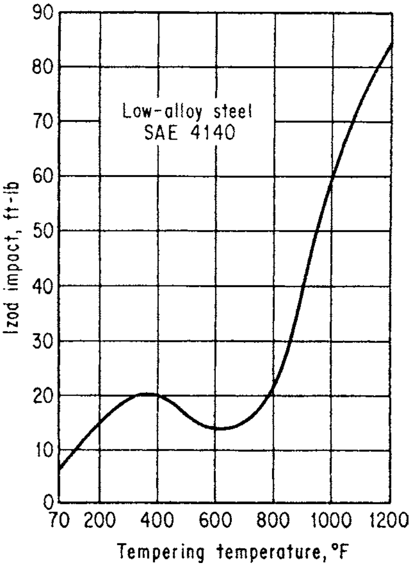


FIG. 6.24 In the tempering of this 4140 steel the notched-bar impact properties decrease in the range of 450 to 650°F.[9]

Series designation	Types
10xx	Nonsulfurized carbon steels
11xx	Resulfurized carbon steels (free-machining)
12xx	Rephosphorized and resulfurized carbon steels (free-machining)
13xx	Manganese 1.75%
23xx*	Nickel 3.50%
25xx*	Nickel 5.00%
31xx	Nickel 1.25%, chromium 0.65%
33xx	Nickel 3.50%, chromium 1.55%
40xx	Molybdenum 0.20 or 0.25%
41xx	Chromium 0.50 or 0.95%, molybdenum 0.12 or 0.20%
43xx	Nickel 1.80%, chromium 0.50 or 0.80%, molybdenum 0.25%
44xx	Molybdenum 0.40%
45xx	Molybdenum 0.52%
46xx	Nickel 1.80%, molybdenum 0.25%
47xx	Nickel 1.05%, chromium 0.45%, molybdenum 0.20 or 0.35%
48xx	Nickel 3.50%, molybdenum 0.25%
50xx	Chromium 0.25, 0.40, or 0.50%
50xxx	Carbon 1.00%, chromium 0.50%
51xx	Chromium 0.80, 0.90, 0.95, or 1.00%
51xxx	Carbon 1.00%, chromium 1.05%
52xxx	Carbon 1.00%, chromium 1.45%
61xx	Chromium 0.60, 0.80, or 0.95%, vanadium 0.12%, 0.10% min, or 0.15% min

(*Continued*)

Series designation	Types
81xx	Nickel 0.30%, chromium 0.40%, molybdenum 0.12%
86xx	Nickel 0.55%, chromium 0.50%, molybdenum 0.20%
87xx	Nickel 0.55%, chromium 0.50%, molybdenum 0.25%
88xx	Nickel 0.55%, chromium 0.50%, molybdenum 0.35%
92xx	Manganese 0.85%, silicon 2.00%, chromium 0 or 0.35%
93xx	Nickel 3.25%, chromium 1.20%, molybdenum 0.12%
94xx	Nickel 0.45%, chromium 0.40%, molybdenum 0.12%
98xx	Nickel 1.00%, chromium 0.80%, molybdenum 0.25%

*Not included in the current list of standard steels.

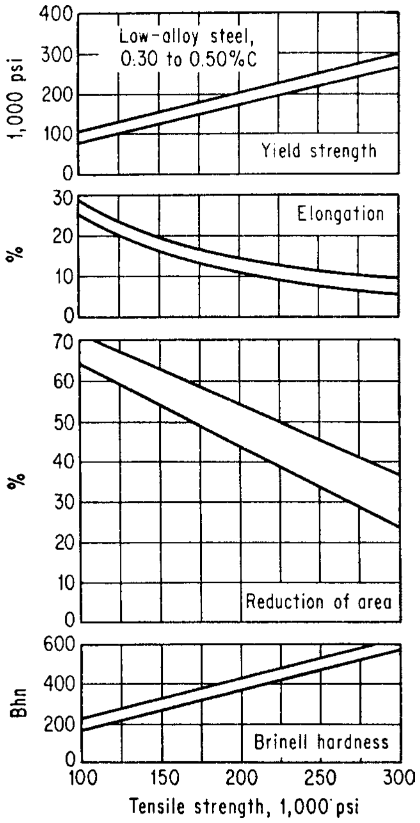


FIG. 6.25 The most probable properties of tempered martensite for a variety of low-alloy steels.[4]

The most probable properties of tempered martensite for low-alloy steels fall within narrow bands even though there are differences in sources and treatments. The relations for these shown in Fig. 6.25 are useful in predicting properties to within approximately 10 percent.

Structural steels may be specified by hardenability requirements, the H designation, rather than stringent specification of the chemistry. Hardenability, determined by the standardized Jominy end-quench test, is a measurement related to the variation in hardness with mass, in quenched steels. Since different structures are formed as a function of the cooling rate and the transformation is affected by the nature of the alloying elements, it is necessary to know whether the particular steel is shallow (A) or deeply hardenable (C), as in Fig. 6.26. The hardenability of a particular steel is a useful criterion in selection because it is related to the mechanical properties pertinent to the section size.

The selection of through-hardened steel based upon carbon content is indicated on the next page for some typical applications.

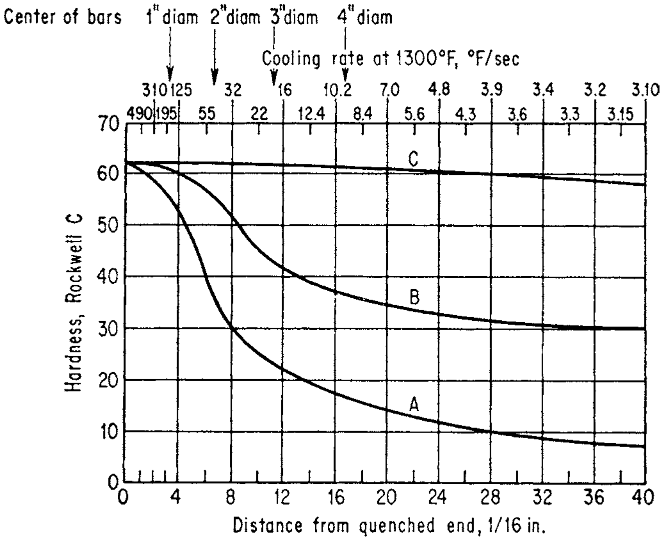


FIG. 6.26 Hardenability curves for different steels with the same carbon content.[7]

Carbon range	Requirement	Approx. tensile strength level, lb/in^2	Applications
Medium, 0.3 to 0.5%	Strength and toughness	150,000	Shafts, bolts, forgings, nuts
Intermediate, 0.5 to 0.7%	Strength	225,000	Springs
High, 0.8 to 1.0%	Wear resistance	300,000	Bearings, rollers, bushings

6.9 SURFACE-HARDENING TREATMENTS

The combination of high surface wear resistance and a tough-ductile core is particularly desirable in gears, shafts, and bearings. Various types of surface-hardening treatments and processes can achieve these characteristics in steels; the most important of these are the following:

Base metal	Process
Low C, to 0.3%	Carburization: A carbon diffusion in the γ-phase region, with controlled hydrocarbon atmosphere or in a box filled with carbon. The case depth is dependent on temperature-time factors. Heat treatment follows process.
Medium C, 0.4 to 0.5%	Localized surface heating by induction or a controlled flame to above the Ac_3 temperature; quenched and tempered.

(*Continued*)

Base metal	Process
Nitriding (nitralloys, stainless steels)	Formation of nitrides (in heat-treated parts) in ammonia atmosphere at 950 to 1000°F, held for long times. A thin and very hard surface forms and there may be dimensional changes.
Low C, 0.2%	Cyaniding: Parts placed in molten salt baths at heat-treating temperatures; some limited carburization and nitriding occur for cases not exceeding 0.020 in. Parts are quenched and tempered.

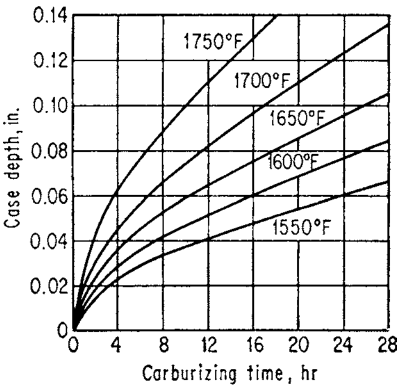


FIG. 6.27 Relation of time and temperature to carbon penetration in gas carburizing.[10]

The carbon penetration in carburization is determined by temperature-time-distance relations issuing from the solution of diffusional equations where *D*, the diffusion coefficient, is independent of concentrations. These relations, shown in Fig. 6.27, permit the selection of a treatment to provide specific case depths. Typical applications are as follows:

Case depth, in	Applications (automotive)
0.020 or more	Push rods, light-load gears, water pump shafts
0.020–0.040	Valve rocker arms, steering-arm bushings, brake and clutch pedal shifts
0.040–0.060	Ring gears, transmission gears, piston pins, roller bearings
0.060 or more	Camshafts

The heat treatments used on carburized parts depend upon grain-size requirements, minimization of retained austenite in the microstructure, amount of undissolved carbide network, and core-strength requirements. As a result of carburization, the surface fiber stresses are compressive. This leads to better fatigue properties. This treatment, which alters the surface chemistry by diffusion of up to 1 percent carbon in a low-carbon steel, gives better wear resistance because the surface hardness is treated for values above Rockwell C 60, while the low-carbon-content core has ductile properties to be capable of the transmission of torsional or bending loads.

Selection of the nitriding process requires careful consideration of cost because of the long times involved in the case formation. A very hard case having a hardness of about Rockwell C 70 ensures excellent wear resistance. Nitrided parts have good corrosion resistance and improved fatigue properties. Nitriding follows the finish-machining and -grinding operations, and many parts can be nitrided without great likelihood of distortion. Long service (several hundred hours) at 500°F has been attained in nitrided gears made from a chromium-base hot-work steel H11. Some typical nitrided steels and their applications are as follows:

Steel	Nitriding treatment		Case hardness, in	Case depth, Rockwell C	Applications
	h	°F			
4140	48	975	0.025–0.035	53–58	Gears, shafts, splines
4340	48	975	0.025–0.035	50–55	Gears, drive shafts
Nitralloy 135M	48	975	0.020–0.025	65–70	Valve stems, seals, dynamic faceplates
H11	70	960 + 980	0.015–0.020	67–72	High-temperature power gears, shafts, pistons

6.10 PRESTRESSING

Concrete has a tensile strength which is a small fraction of its compressive strength and it can be prestressed (pretensioned) with high carbon, 0.7 to 0.85 percent C, steel wire, or strands. These have a tensile strength of about 260,000 lb/in^2, a yield strength of at least 80 percent of the tensile strength, and an adequate ductility. This wire or strand is pretensioned and when the concrete is cured the tensioning is relaxed. As a result the concrete is in compression, having more favorable load-carrying characteristics. When the steel is put in tension after concrete is cured, by different techniques, this is called "posttensioning."

When SAE 1010 steel was prestressed by roll-induced tension, different effects were observed. By prestraining this steel to 0.9 of its true fracture strain the fatigue life was increased. Prestraining to 0.95 of the true fracture strain, on the other hand, resulted in a decreased fatigue life. Generalizations on the effects of prestressing can create misleading results. Involved are the nature and amount of prestressing, type of materials, and effects on property changes. Plain carbon steels, when prestressed within narrow limits, may have their torsional fracture strains increased. On the other hand, prestressing 304 stainless steel resulted in stress corrosion cracking in a corrosive environment. Transition temperatures on impact can be increased (an undesirable effect) in prestressed low-carbon steels. Shot-peening gear teeth, which can result in improved fatigue properties because of the induced surface compressive stresses, can be a beneficial effect of prestressing.

6.11 SOME PRACTICAL CONSIDERATIONS OF INDUCED RESIDUAL STRESSES IN ALLOYS

The presence of residual stresses, especially at surfaces, may affect the fatigue and bending properties significantly. These stresses can be induced in varying degrees by processes involving thermal changes as in heat treatments, welding, and exposures to steep temperature gradients, as well as mechanical effects such as plastic deformation, machining, grinding, and pretensioning. Effects of these when superimposed on externally applied stresses may reduce fatigue properties, or these properties may be improved by specific surface treatments. Comprehensive studies have been made to quantitatively measure the type—tensile or compressive—and the magnitude of residual surface elastic stresses so that the design engineer and user of structural members can optimize such material behavior.

When surface elastic stresses are present in crystalline material, they may be measured nondestructively by x-ray diffraction techniques, a review of which is given by D. P. Koistenen et al., in the SAE publication "TR-182." A two-exposure method for locating the shifts in x-ray diffraction peaks at high Bragg angles θ can result in stress measurements accurate to within ±5000 lb/in^2, using elasticity relations and the fact that elastic atom displacements enable x-ray measurements to act as sensitive nondestructive strain gauges. The relation which is used is

$$\sigma_\phi = \frac{E(\cot\theta)(2\theta_n - 2\theta_i)}{2(1+\nu)\sin^2\psi}\,\frac{1}{57.3}\ \text{lb/in}^2$$

where σ_ϕ = surface elastic stress, lb/in^2
E = Young's modulus, lb/in^2
θ_n, θ_i = measured Bragg angle at normal (n) and inclined (i) incidence
ν = Poisson's ratio
ψ = angle between normal and inclined incident x-ray beam

Typically the back-reflection x-rays used will penetrate the surface by about 0.0002 in. Subsurface stresses may be measured in thick specimens by removing surface layers by electropolishing. These measurements are used for design of gears, bearings, welded plates, aircraft structures, and generally where fatigue and bending properties in use are important. Furthermore, improvements in surface properties can be enhanced by surface treatments which will induce residual compressive stresses.

The effect of localized heating (to about 1300°F) in spot welding a constrained steel specimen had a residual stress pattern as shown in Fig. 6.28.

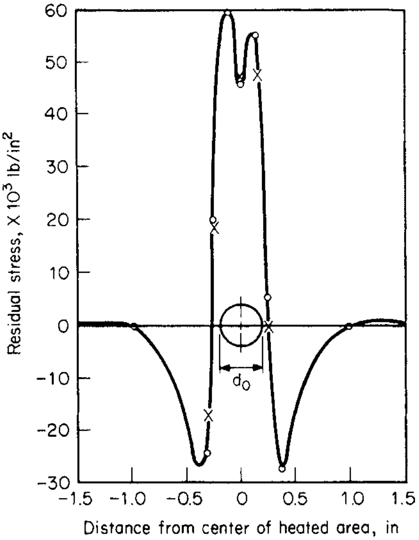


FIG. 6.28 Induced residual stress pattern by localized spot welding.[33]

Surface finishing in heat-treated steel gears, bearings, and dies often entails grinding in the hardened condition because of the greater dimensional tolerances that are desired. Metal removal by the abrasive grinding wheels can cause different degrees of surface residual stresses in the metals because of the frictional heat generated. Very abusive grinding may cause cracks. Examples of the stress patterns and fatigue properties in grinding a hardened aircraft alloy AISI 4340 are shown in Figs. 6.29 and 6.30. The significant decrease in fatigue strength of over 40 percent focuses special attention on the importance of controlling and specifying grinding and metal removal practices in hardened alloys. In addition to this, residual stress patterns may be induced due to the presence of retained austenite in hardened high-carbon steels.

Surface treatments which will induce compressive stresses in surface layers can improve fatigue properties substantially. These include mechanical deformation of surfaces by shot peening and roller burnishing, as well as microstructural effects in carburizing or nitriding steels. The beneficial effects of surface treatments by hardened 4340 steel are shown in Fig. 6.31.

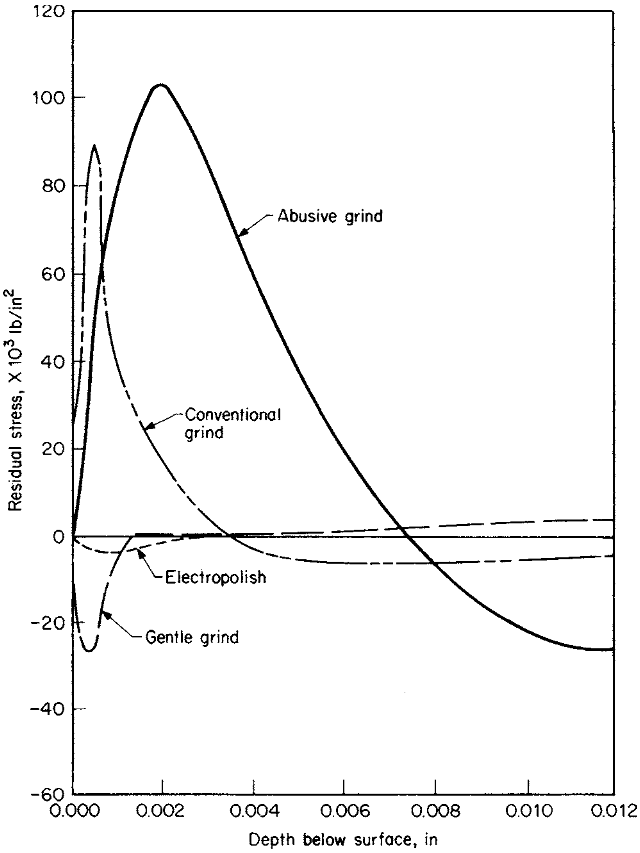


FIG. 6.29 Induced stresses in grinding ASAI 4340 steel heat-treated to Rockwell C 50.

To recover some fatigue properties in decarburized steel surfaces it may be helpful to consider shot peening these. Skin rolling of sheet or tubular products may also induce compressive surface stresses.

6.12 NOTCHED IMPACT PROPERTIES: CRITERIA FOR MATERIAL SELECTION

When materials are subject to high deformation rates and are particularly sensitive to stress concentrations at sharp notches, criteria must be established to indicate safe operating-temperature ranges. The impact test (Izod or Charpy V-notch) performed on notched specimens conducted over a prescribed temperature range indicates the likelihood of ductile (shear-type) or brittle (cleavage-type) failure. In this test the velocity of the striking head at the instant of impact is about 18 fps, so the strain rates are several orders of magnitude greater than in a tensile test. The energy absorbed in fracturing a standard notched specimen is measured by the differences in potential energy from free fall of the hammer to the elevation after fracturing it. The typical effect of temperature upon impact energy for a metal which shows ductile and brittle characteristics is shown

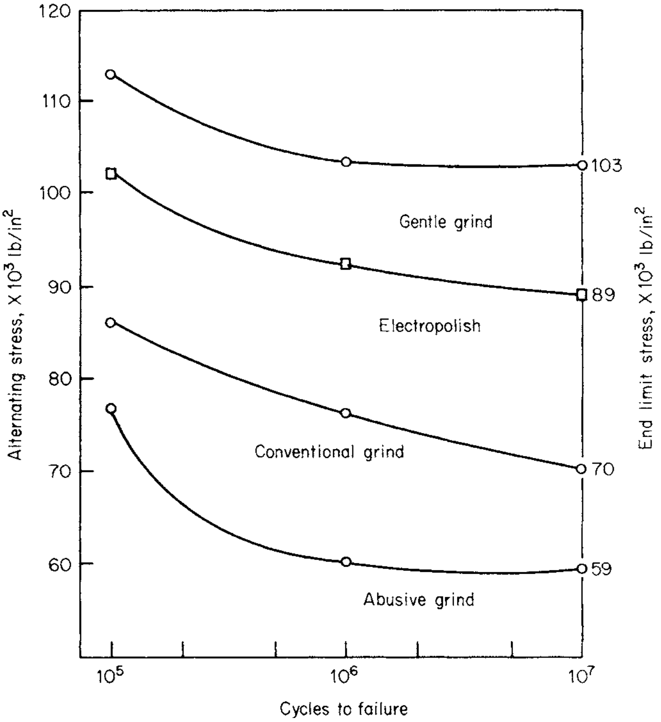


FIG. 6.30 Effect of severity of surface grinding of hardened 4340 Rockwell C (50) steel.[34,35]

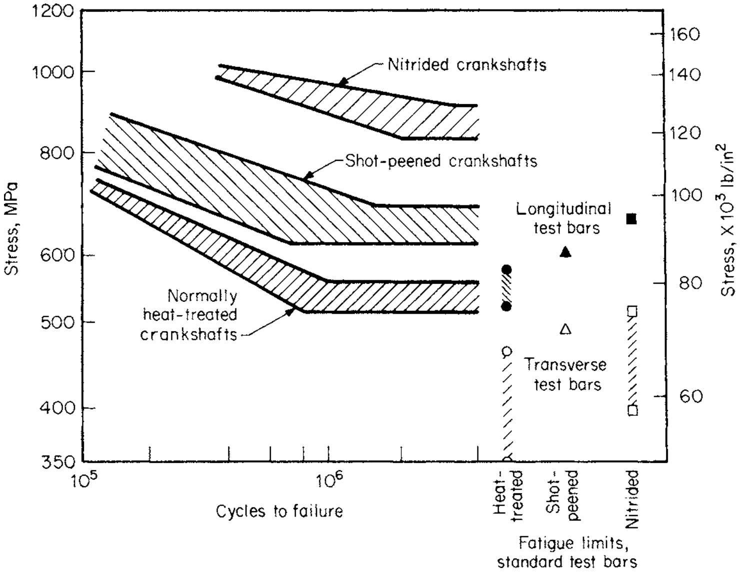


FIG. 6.31 Improved fatigue strength by shot-peening and nitriding 4340 hardened crankshafts.[36]

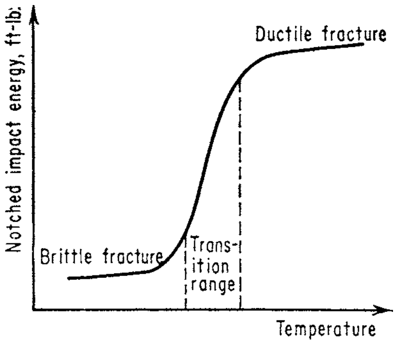


FIG. 6.32 The ductile-to-brittle transition in impact.

in Fig. 6.32. Interest centers on the transition temperature range and the material-sensitive factors, such as composition, microstructure, and embrittling treatments. ASTM specifications for structural steel for ship plates specify a minimum impact energy at a given temperature, as, for example, 15 ft·lb at 40°F. It is desirable to use materials at impact energy levels and at service temperatures where crack propagation does not proceed. Some impact characteristics for construction steels are shown in Fig. 6.33.

It is generally characteristic of pure metals which are fcc in lattice structure to possess toughness (have no brittle fracture tendencies) at very low temperatures. Body-centered-cubic pure metals, as well as hexagonal metals, do show ductile-to-brittle behavior. Tantalum, a bcc metal, is a possible exception and is ductile in impact even at cryogenic temperatures. Alloyed metals do not follow any general pattern of behavior; some specific impact values for these are as follows:

Material	Charpy V notch at 80°F	Impact strength at −100°F, ft·lb
Cu-Be (2%) HT	5.4	5.5
Phosphor bronze (5% Sn):		
Annealed	167	193
Spring temper	46	44
Nilvar Fe-Ni (36%):		
Annealed	218	162
Hard	97	77
2024-T6 aluminum aircraft alloy, HT aged	12	12
7079-T6 aluminum aircraft alloy, HT aged	4.5	3.5
Mg-Al-Zn extruded	7.0	3.0

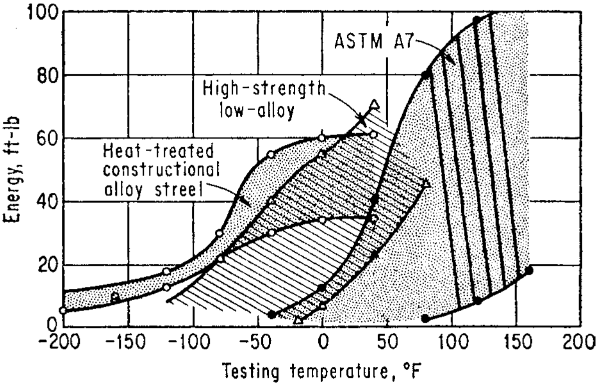


FIG. 6.33 Influence of testing temperature on notch toughness, comparing carbon steel of structural quality (ASTM A7) with high-strength, low-alloy and heat-treated constructional alloy steel.[4] Charpy V notch.

Austenitic stainless steels are ductile and do not exhibit transition in impact down to very low temperatures.

Some brittle service failures in steel structures have occurred in welded ships, gas-transmission pipes, pressure vessels, bridges, turbine generator rotors, and storage tanks. Serious consideration must then be given the effect of stress raisers, service temperatures, tempering embrittling structures in steels, grain-size effects, as well as the effects of minor impurity elements in materials.

6.13 FATIGUE CHARACTERISTICS FOR MATERIALS SPECIFICATIONS

Most fatigue failures observed in service as well as under controlled laboratory tests are principally the result of poor design and machining practice. The introduction of potential stress raisers by inadequate fillets, sharp undercuts, and toolmarks at the surfaces of critically cyclically stressed parts may give rise to crack nucleation and propagation so that ultimate failure occurs. Particular attention should be given to material fatigue properties where rotating and vibrating members experience surface fibers under reversals of stress.

In a fatigue test, a highly polished round standard specimen is subjected to cyclic loading; the number of stress reversals to failure is recorded. For sheets, the standard specimen is cantilever supported. Failure due to tensile stresses usually starts at the surface. Typical of a fatigue fracture is its conchoidal appearance, where there is a smooth region in which the severed sections rubbed against each other and where the crack progressed to a depth where the load could no longer be sustained. From the fatigue data a curve of stress vs. number of cycles to failure is plotted. Note that there can be considerable statistical fluctuation in the results (about ±15 percent variations in stress).

Two characteristics can be observed in fatigue curves with respect to the endurance limit shown by E_a and E_b in Fig. 6.34:

E_a, curve a, the asymptotic stress value, typical in most materials

E_b, curve b, a stress value taken at an arbitrary number of cycles; e.g., 500,000,000, typical in Al and Mg alloys

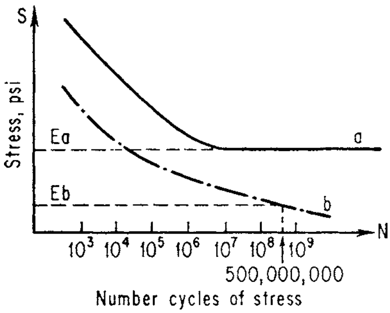


FIG. 6.34 Fatigue curves. (a) Most materials have an endurance limit E_a (asymptotic stress). (b) Endurance limit E_b (nonasymptotic set at arbitrary value of N).

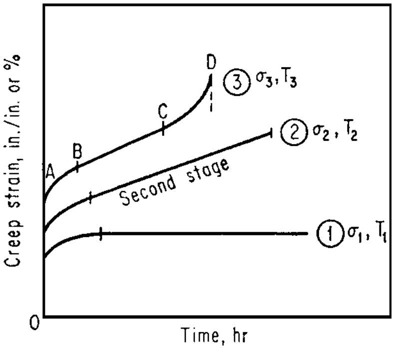


FIG. 6.35 Typical creep curves. At constant temperature, $\sigma_3 > \sigma_2 > \sigma_1$. At constant stress $T_3 > T_2 > T_1$.

For design specifications, the endurance limit represents a safe working stress for fatigue. The endurance ratio is defined as the ratio of the endurance limit to the ultimate tensile strength. These values are strongly dependent upon the presence of notches on the surface and a corrosive environment, and on surface-hardening treatments. In corrosion, the pits formed act as stress raisers leading generally to greatly reduced endurance ratios. References 15 through 20 provide useful information on corrosion. A poorly machined surface or a rolled sheet with surface scratches evidences low endurance ratios, as do parts in service with sharp undercuts and insufficiently filleted changes in section.

Improvements in fatigue properties are brought about by those surface-hardening treatments which produce induced compressive stresses as in steels, nitrided (about 160,000 lb/in^2 compressive stress) or carburized (about 35,000 lb/in^2 compressive stress).

Metallurgical factors related to poorer fatigue properties are the presence of retained austenite in hardened steels, the presence of flakes or sharp inclusions in the microstructure, and treatments which induce preferential corrosive grain-boundary attack. When parts are quenched or formed so that surface tensile stresses are present, stress-relieval treatments are advisable.

6.14 SHEAR AND TORSIONAL PROPERTIES

When rivets, bolts, and fastening pins are used in structural components, their shear and torsional properties are evaluated mainly for ductile materials. The stress which produces shear fracture is reported as the ultimate shear strength or sometimes as shear strength. Some typical and averaged ratios of strength are given below:

Material	Brinell hardness	Ultimate shear strength/ ultimate tensile strength
Ductile cast irons	170–300	0.9–1.0
Aluminum aircraft alloys	120	0.65
Brasses	100	0.7

In the elastic region the shear or torsional properties are related by

$$G = \tau/\gamma$$

where G = modulus of rigidity in shear
τ = shear stress
γ = shear strain

Furthermore, one can calculate with reasonably good accuracy the relation of elastic moduli by

$$G = E/2(1 + \nu)$$

where E = Young's modulus of rigidity in tension
ν = Poisson's ratio

This relation presumes isotropic elastic behavior.

The elastic shear strength, determined at a fixed offset on the shear stress-strain diagram (i.e., 0.04 percent offset), has been related to the tensile yield strength

(0.2 percent offset) such that its ratio is about 0.6 for ductile cast irons and 0.55 for aluminum alloys.

6.15 MATERIALS FOR HIGH-TEMPERATURE APPLICATIONS

6.15.1 Introduction

Selection of materials to withstand stress at high temperatures is based upon experimentally determined temperature stress-time properties. Some useful engineering design criteria follow.

1. Dimensional change, occurring by plastic flow, when metals are stressed at high temperatures for prolonged periods of time, as measured by creep tests
2. Stresses that lead to fracture, after certain set time periods, as determined by stress-rupture tests, where the stresses and deformation rates are higher than in a creep test
3. The effect of environmental exposure on the oxidation or scaling tendencies
4. Considerations of such properties as density, melting point, emissivity, ability to be coated and laminated, elastic modulus, and the temperature dependence on thermal conductivity and thermal expansion

Furthermore, the microstructural changes occurring in alloys used at high temperatures are correlated with property changes in order to account for the significant discontinuities which occur with exposure time. As a result of these evaluations, special alloys that have been (or are being) developed are recommended for use in different temperature ranges extending to about 2800°F (refractory range). Vacuum or electron-beam melting and special welding techniques are of special interest here in fabricating parts.

6.15.2 Creep and Stress: Rupture Properties

In a creep test, the specimen is heated in a temperature-controlled furnace, an axial load is applied, and the deformation is recorded as a function of time, for periods of 1000 to 3000 h. Typical changes in creep strain with time, for different conditions of stress and temperature, are shown in Fig. 6.35. Plastic flow creep, associated with the movement of dislocations by climb sliding of grain boundaries and the diffusion of vacancies, is characterized by:

1. *OA*, elastic extension on application of load
2. *AB*, first stage of creep with changing rate of creep strain
3. *BC*, second stage of creep, in which strain rate is linear and essentially constant
4. *CD*, third stage of increasing creep rate leading to fracture

Increasing stress at a constant temperature or increasing temperature at constant stress results in the transfers from curve 1 to curves 2 and 3 in Fig. 6.35.

The engineering design considerations for dimensional stability are based upon

1. Stresses resulting in a second-stage creep rate of 0.0001 percent per hour (1 percent per 10,000 h or 1 percent per 1.1 years)

2. A second-stage creep rate of 0.00001 percent per hour (1 percent per 100,000 h or 1 percent per 11 years), where weight is of secondary importance relative to long service life, as in stationary turbines

The time at which a stress can be sustained to failure is measured in a stress-rupture test and is normally reported as rupture values for 10, 100, 1000, and 10,000 h or more. Because of the higher stresses applied in stress-rupture tests, shown in Fig. 6.36, some extrapolation of data may be possible and some degree of uncertainty may ensue. Discontinuous changes at points *a* in the stress-rupture data shown in Fig. 6.37 are associated with a change from transgranular to intergranular fracture, and further microstructural changes can occur at increasing times. A composite picture of various high-temperature test results is given in Fig. 6.38 for a type 316 austenitic stainless steel.

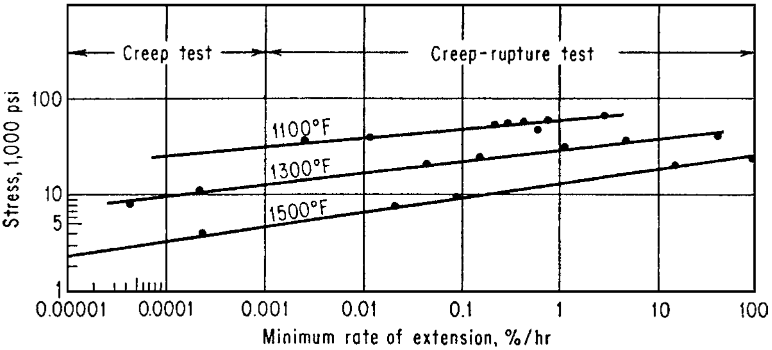


FIG. 6.36 Correlation of creep and rupture test data for type 316 stainless steel (18 Cr, 8 Ni, and Mo).[12]

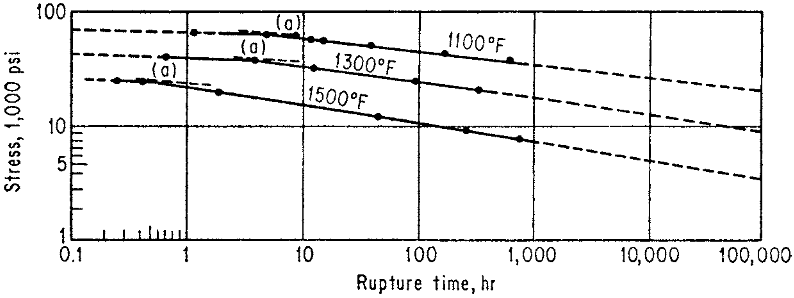


FIG. 6.37 Stress vs. rupture time for type 316 stainless steel.[12] The structural character associated with point (*a*), on each of the three relations, is that the mode of fracture changes from transgranular to intergranular.

6.15.3 Heat-Resistant Superalloys: Thermal Fatigue

The materials especially developed to function at high temperatures for sustained stress application in aircraft (jet engines) are referred to as "superalloys." These are nickel-base alloys designated commercially as Nimonic Hastelloy, Inconel Waspaloy, and René or cobalt-base alloys designated as S-816, HS25, and L605. Significant

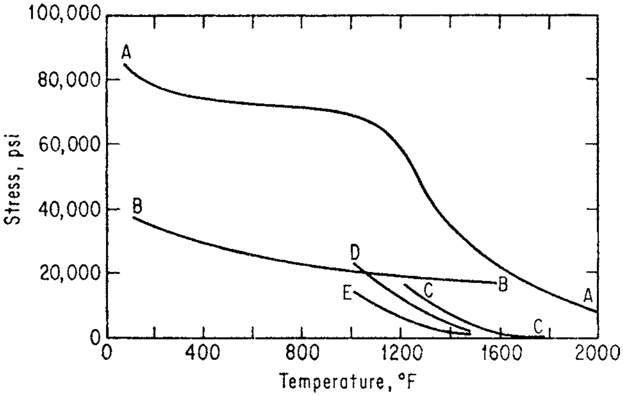


FIG. 6.38 Properties of type 316 stainless steel (18 Cr, 8 Ni, and Mo).[12] *A* = short-time tensile strength; *B* = short-time yield strength, 0.2 percent offset; *C* = stress of rupture, 10,000 h; *D* = stress for creep rate, 0.0001 percent per hour; *E* = stress for creep rate, 0.00001 percent per hour.

improvements have been made in engineering design of gas turbine blades (coolant vents) and material processing techniques (directional solidification, single crystal, and protective coating). The developed progress in the past several decades[40] led to increased engine efficiencies and increased time between overhauls from 100 h to more than 10,000 h.

Thermal fatigue is an important high-temperature design property which is related to fracture occurring with cyclical stress applications. This type of fatigue failure, resulting from thermal stresses (constrained expansion and metal structural changes), is also termed low-cycle fatigue. At sustained temperatures above 800°F the *S–N* curves generally do not have an endurance limit (asymptotic stress) and the fatigue stress continually decreases with cycles to failure.

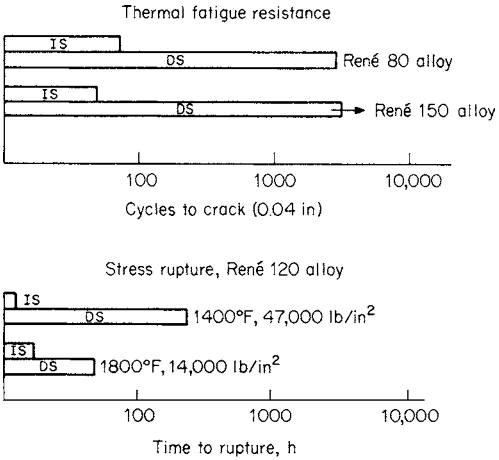


FIG. 6.39 Improved thermal fatigue resistance and stress-rupture properties of directionally solidified (DS) and isotropic superalloy (IS).[37,38]

Processing some superalloy turbine blades by directional solidification (DS) or by eliminating grain-boundary effects in single crystals results in significant improvements in thermal fatigue resistance and stress-rupture lives, as shown in Fig. 6.39. DS is controlled by casting with stationary and movable heat sinks so that all grain boundaries are made to be parallel to the applied stress direction. Crystal growth directions are controlled to take advantage of the favorable anisotropic high-strength property values as well as minimizing deleterious grain-boundary reactions.

Materials selections for aircraft gas turbine systems are typically as follows:

Component	Materials used
Turbofin compressor blades	Titanium alloys: Ti-6Al-4V, Ti-6Al-2Sn-4Mo
Combustor (burner)	Sheet alloys, shaped and joined Hastelloy X, Inconel 617
Gas turbine blades	Cast Waspaloy, René, PNA 1422 (special processes—directional or single-crystal solidification)

Thermal shock failures, as in engine valves, can occur as a result of steep temperature gradients, leading to constrained thermal dilational changes when encountering high stress. These restrained stresses can exceed the breaking point of the material at the operating temperatures. Some superalloy properties useful for design are given in Table 6.3.

TABLE 6.3 Superalloy Properties[39]

Alloy	Rupture strength, after 1000 h, lb/in²			Characteristics and applications
	1200°F	1500°F	1800°F	
Hastelloy	34,000	10,000	2,100	Jet engine sheet parts, good oxidation resistance
Inconel 617	47,000	14,000	3,600	Gas turbine aircraft engine parts
René 41	10,200	29,000	—	Jet engine blade, parts
Waspaloy	88,000	25,500	—	Jet engine blades
In 102	52,000	7,400	—	Superheater, jet engine part
TAZ 8A	12,000	61,000	14,000	Good thermal fatigue and oxidation resistance
René	—	5,700	10,000	Compressor disk alloy
S816	46,000	18,000	—	Gas turbine blades, bolts, springs
Haynes 150	—	5,800	1,700	Resists thermal shock, high-temperature corrosion

6.16 MATERIALS FOR LOW-TEMPERATURE APPLICATION

Materials for low-temperature application are of increasing importance because of the technological advances in cryogenics. The most important mechanical properties are usually strength and stiffness, which generally increase as the temperature is decreased. The temperature dependence on ductility is a particularly important criterion in design, because some materials exhibit a transition from ductile to brittle behavior with decreasing temperature. Factors related to this transition are microstructure, stress concentrations present in notches, and the effects of rapidly applied strain rates in materials. Mechanical design can also influence the tendency for brittle failure at low temperature, and for this reason, it is essential that sharp notches (which can result from surface-finishing operations) be eliminated and that corners at changes of section be adequately filleted.

Low-temperature tests on metals are made by measuring the tensile and fatigue properties on unnotched and notched specimens and the notched impact strength. Metals exhibiting brittle characteristics at room temperature, by having low values of percent elongation and percent reduction in area in a tensile test as well as low impact strength, can be expected to be brittle at low temperatures also. Magnesium alloys, some high-strength aluminum alloys in the heat-treated condition, copper-beryllium heat-treated alloys, and tungsten and its alloys all exhibit this behavior. At best, applications of these at low temperatures can be made only provided that they adequately fulfill design requirements at room temperature.

When metals exhibit transitions in ductile-to-brittle behavior, low-temperature applications should be limited to the ductile region, or where experience based on field tests is reliable, a minimum value of impact strength should be specified. The failure, by breaking in two, of 19 out of 250 welded transport ships in World War II, caused by the brittleness of ship plates at ambient temperatures, focused considerable attention on this property. It was further revealed in tests that these materials had Charpy V-notch impact strengths of about 11 ft·lb at this temperature. Design specifications for applications of these materials are now based on higher impact values. For temperatures extending from subatmospheric temperatures to liquid-nitrogen temperatures (−320°F), transitions are reported for ferritic and martensitic steels, cast steels, some titanium alloys, and some copper alloys.

Design for low-temperature applications of metals need not be particularly concerned with the Charpy V-notch impact values provided they can sustain some shear deformation and that tensile or torsion loads are slowly applied. Many parts are used successfully in polar regions, being based on material design considerations within the elastic limit. When severe service requirements are expected in use, relative to rapid rates of applied strain on notch-sensitive metals, particular attention is placed on selecting materials which have transition temperatures below that of the environment. Some important factors related to the ductile-to-brittle transition in impact are the composition, microstructure, and changes occurring by heat treatment, preferred directions of grain orientation, grain size, and surface condition.

The transition temperatures in steels are generally raised by increasing carbon content, by the presence of more than 0.05 percent sulfur, and significantly by phosphorus at a rate of 13°F per 0.01 percent P. Manganese up to 1.5 percent decreases the transition temperature and high nickel additions are effective, so that in the austenitic stainless steels the behavior is ductile down to liquid-nitrogen temperatures. In high-strength medium-alloy steels it is desirable, from the standpoint of lowering the transition range, that the structure be composed of a uniformly tempered martensite, rather than containing mixed products of martensite and bainite or martensite and pearlite. This can be controlled by heat treatment. The preferred orientation that can be induced in rolled and forged metals can affect notched impact properties, so that specimens made from the longitudinal or rolling direction have higher impact strengths than those taken from the transverse direction. Transgranular fracture is normally characteristic of low-temperature behavior of metals. The metallurgical factors leading to intergranular fracture, due to the segregation of embrittling constituents at grain boundaries, cause concern in design for low-temperature applications. In addition to the control of these factors for enhanced low-temperature use, it is important to minimize or eliminate notch-producing effects and stress concentration, by specifying proper fabrication methods and providing adequate controls on these, as by surface inspection.

Examples of some low-temperature properties of the refractory metals, all of which have bcc structures, are shown in Fig. 6.40. The high ductility of tantalum at very low temperatures is a distinctive feature in this class that makes it attractive for use as a cryogenic (as well as a high-temperature) material. Based on the increase of yield-strength-to-density ratio with decreasing temperatures shown in Fig. 6.41, the

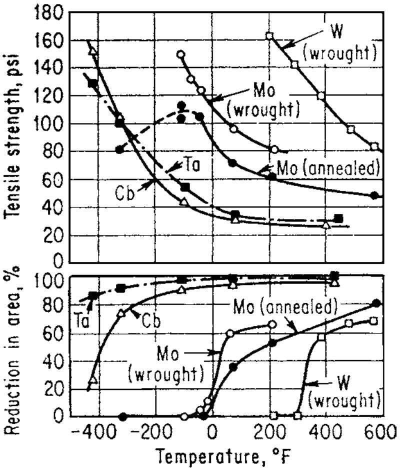


FIG. 6.40 Strength and ductility of refractory metals at low temperatures.[13]

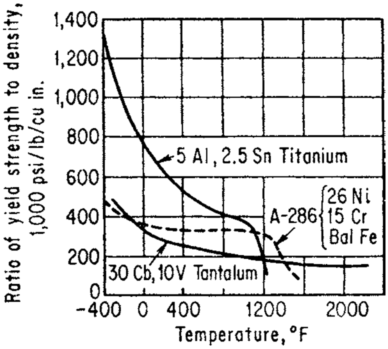


FIG. 6.41 Yield-strength-to-density ratios related to temperature for some alloys of interest in cryogenic applications.[13]

three alloys of a titanium-base Al (5) Sn (2.5), an austenitic iron-base Ni (26), Cr (15) alloy A286, as well as the tantalum-base Cb (30), 10 V alloy, are also useful for cryogenic use.

Comparisons of the magnitude of property changes obtained by testing at room temperature and −100°F for some materials of commercial interest are shown in Table 6.4.

6.17 RADIATION DAMAGE[31]

A close relationship exists between the structure and the properties of materials. Modification and control of these properties are available through the use of various metallurgical processes, among them nuclear radiation. Nuclear radiation is a process whereby an atomic nucleus undergoes a change in its properties brought about by interatomic collisions.

The energy transfer which occurs when neutrons enter a metal may be estimated by simple mechanics, the quantity of energy transferred being dependent upon the atomic mass. The initial atomic collision, or primary "knock-on" as it is called, has enough energy to displace approximately 1000 further atoms, or so-called secondary knock-ons. Each primary or secondary knock-on must leave behind it a resulting vacancy in the lattice. The primaries make very frequent collisions because of their slower movement, and the faster neutrons produce clusters of "damage," in the order of 100 to 1000 Å in size, which are well separated from one another.

Several uncertainties exist about these clusters of damage, and because of this it is more logical to speak of radiation damage than of point defects, although much of the damage in metals consists of point defects. Aside from displacement collisions, replacement collisions are also possible in which moving atoms replace lattice atoms. The latter type of collision consumes less energy than the former.

Another effect, important to the life of the material, is that of transmutation, or the conversion of one element into another. Due to the behavior of complex alloys, the cumulative effect of transmutation over long periods of time will quite often be of

TABLE 6.4 Low-Temperature Test Properties

Tensile Test Properties

Material	Tensile strength, 1,000 psi		Yield strength, 1,000 psi		Elongation, %		Modulus of elasticity, 100 psi	
	80°F	−100°F	80°F	−100°F	80°F	−100°F	80°F	−100°F
Beryllium copper 2% Be—heat-treated sheet	189	194	154	170	3	3	19.1	19.1
Phosphor bronze 5% Sn; spring temp	98	107	90	97	7	11	16.5	17.1
Molybdenum sheet partly recrystallized	97	141	75	130	19	3.5	48.7	47.5
Tungsten wire, drawn	214	. . .	196	. . .	2.6	. . . .	56	
Tantalum sheet, annealed	55	71	36	66	27	25	28.2	28.3
Nilvar sheet 36 Ni; 74 Fe, rolled	76	101	62	76	27	30	21.8	19.7

Sheet Fatigue Properties

Material	Fatigue strength at 2 × 10^7 cycles 1,000 psi		Endurance ratios	
	80°F	−100°F	80°F	−100°F
Beryllium copper	31	45	0.16	0.23
Molybdenum	46	65	0.59	0.47
Tantalum	35	41	0.64	0.58
Nilvar	26	30	0.33	0.30

Charpy V-notch Impact Strength

Material	Impact strength, ft-lb	
	80°F	−100°F
Beryllium copper	5.4	5.4
Phosphor bronze	46	44
Nilvar	97	77

importance. U^{235}, the outstanding example of this phenomenon, has enough energy after the capture of a slow neutron to displace one or more atoms.

Moving charged particles may also donate energy to the valence electrons. In metals this energy degenerates into heat, while in nonconductors the electrons remain in excited states and will sometimes produce changes in properties.

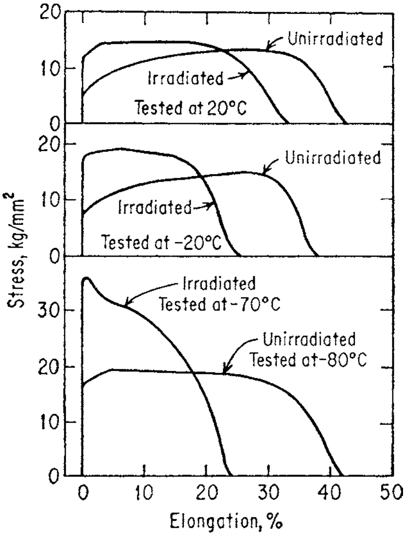


FIG. 6.42 Effect of irradiation on stress-strain curves of Fe single crystals tested at different temperatures. Irradiation dose 8×10^{17} thermal n/cm^2. (*Courtesy of D. McLean.*[31])

Figure 6.42 illustrates the effect or irradiation on the stress-strain curve of iron crystals at various temperatures.[32] In metals other than iron irradiation tends to produce a ferrous-type yield point and has the effect of hardening a metal. This hardening may be classified as a friction force and a locking force on the dislocations. Some factors of irradiation hardening are:

1. It differs from the usual alloy hardening in that it is less marked in cold-worked than annealed metals.
2. Annealing at intermediate temperatures may increase the hardening.
3. Alloys may exhibit additional effects, due, for example, to accelerated phase changes and aging.

The most noticeable effect of irradiation is the rise in transition temperatures of metals which are susceptible to cold brittleness. Yet another consequence of irradiation is the development of internal cracks produced by growth stresses. At high enough temperatures gas atoms can be diffused and may set up large pressures within the cracks.

Some other effects of irradiation are swelling, phase changes which may result in greater stability, radiation growth, and creep. The reader is referred to Ref. 31 for an analysis of these phenomena.

6.18 PRACTICAL REFERENCE DATA

Tables 6.5 through 6.9 give various properties of commonly used materials. Figure 6.43 provides a hardness conversion graph for steel. References 21 through 30 yield more information.

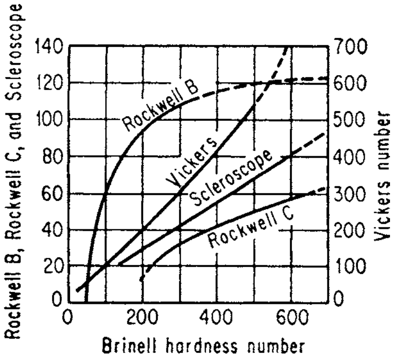


FIG. 6.43 Hardness conversion curves for steel.

TABLE 6.5 Physical Properties of Metallic Elements*

Element	Symbol	Melting point, °F	Boiling point, °F	Specific heat,[a] cal/g/°C	Thermal conductivity,[a] Btu/hr/sq ft/°F/ft	Density,[a] g/cm^3	Modulus of elasticity in tension, million psi	Coefficient of linear thermal expansion,[a] μ in./in./°F	Electrical resistivity, microhm-cm	Crystal structure
Aluminum	Al	1220	4442	0.215	128.	2.70	9	13.1	2.65	f.c.c.
Antimony	Sb	1167	2516	0.049	10.8	6.62	11.3	4.7	39	Rhomb.
Arsenic	As	1503 (28 atm)	1135[b]	0.082	.....	5.72	.....	2.6	33.3	Rhomb.
Barium	Ba	1317	2980	0.068	.....	3.5	.....	....	......	b.c.c.
Beryllium	Be	2332	5020	0.45	84.4	1.85	42	6.4	4	h.c.p.
Bismuth	Bi	520	2840	0.029	4.8	9.80	4.6	7.4	107	Rhomb.
Boron	B	3690	......	0.309	.....	2.34	.....	4.6	10^{12}	Orthorhomb.
Cadmium	Cd	610	1409	0.055	53.	8.65	8	16.55	6.83	h.c.p.
Calcium	Ca	1540	2625	0.149	72.3	1.55	3.5	12.4	3.91	f.c.c.
Carbon (graphite)	C	6740[b]	8730	0.165	13.8	2.25	0.7	0.3 to 2.4	1375	Hexag.
Cerium	Ce	1479	6280	0.045	6.6	6.77	6	4.4	75	f.c.c.
Cesium	Cs	84	1273	0.048	.....	1.90	.....	54	20	b.c.c.
Chromium	Cr	3407	4829	0.11	40.3	7.19	36	3.4	12.9	b.c.c.
Cobalt	Co	2723	5250	0.099	41.5	8.85	30	7.66	6.24	h.c.p.
Columbium	Cb	4474	8901	0.065	31.5	8.57	.....	4.06	12.5	b.c.c.
Copper	Cu	1981	4703	0.092	226.	8.96	16	9.2	1.67	f.c.c.
Gallium	Ga	86	4059	0.079	19.4	5.91	.....	10	17.4	Orthorhomb.
Germanium[c]	Ge	1719	5125	0.073	33.7	5.32	.....	3.19	46 × 10^6	Diam. cubic
Gold	Au	1954	5380	0.031	171.	19.32	11.6	7.9	2.35	f.c.c.
Indium	In	313	3632	0.057	13.8	7.31	1.57	18	8.37	f.c.tetr.
Iridium	Ir	4449	9570	0.031	33.7	22.50	76	3.8	5.3	f.c.c.
Iron	Fe	2798	5430	0.11	43.3	7.87	28.5	6.53	9.71	b.c.c.
Lanthanum	La	1688	6280	0.048	8.	6.19	10.5	2.77	57	Hexag.
Lead	Pb	621	3137	0.031	20.	11.36	2	16.3	20.6	f.c.c.
Lithium	Li	357	2426	0.79	41.	0.534	.....	31	8.55	b.c.c.
Magnesium	Mg	1202	2025	0.245	88.5	1.74	6.35	15.05	4.45	h.c.p.
Manganese	Mn	2273	3900	0.115	.....	7.43	23	12.22	185	Complex cubic
Mercury	Hg	−37	675	0.033	4.7	13.55	.....	....	98.4	Rhomb.

Element	Symbol	Melting point, °F	Boiling point, °F	Specific heat,[a] cal/g/°C	Thermal conductivity,[a] Btu/hr/sq ft/°F/ft	Density,[a] g/cm³	Modulus of elasticity in tension, million psi	Coefficient of linear thermal expansion,[a] μ in./in./°F	Electrical resistivity, microhm-cm	Crystal structure
Molybdenum	Mo	4730	10040	0.066	82.	10.22	47	2.7	5.2	b.c.c.
Nickel	Ni	2647	4950	0.105	53.	8.90	30	7.39	6.84	f.c.c.
Osmium	Os	4900	9950	0.031	.....	22.57	81	2.6	9.5	h.c.p.
Palladium	Pd	2826	7200	0.058	40.5	12.02	16.3	6.53	10.8	f.c.c.
Phosphorus (white)	P	112	536	0.177	.....	1.83	.....	70	10^{17}	Cubic
Platinum	Pt	3217	8185	0.0314	39.8	21.45	21.3	4.9	10.6	f.c.c.
Plutonium	Pu	1184	6000	0.033	4.8	19.00	14	30.55	141.4	Monoclinic
Potassium	K	147	1400	0.177	58.	0.86	.....	46	6.15	b.c.c.
Rhenium	Re	5755	10650	0.033	41.	21.04	66.7	3.7	19.3	h.c.p.
Rhodium	Rh	3571	8130	0.059	50.6	12.44	42.5	4.6	4.51	f.c.c.
Rubidium	Rb	102	1270	0.080	.....	1.53	.....	50	12.5	b.c.c.
Ruthenium	Ru	4530	8850	0.057	.....	12.20	60	5.1	7.6	h.c.p.
Selenium	Se	423	1265	0.084	.....	4.79	8.4	21	12	Hexag.
Silicon[c]	Si	2570	4860	0.162	48.2	2.33	16.35	1.6 to 4.1	10^{5}	Diam. cubic
Silver	Ag	1761	4010	0.056	242.	10.49	11	10.9	1.59	f.c.c.
Sodium	Na	208	1638	0.295	77.2	0.971	.....	39	4.2	b.c.c.
Strontium	Sr	1414	2520	0.176	.....	2.60	.....	.....	23	f.c.c.
Tantalum	Ta	5425	9800	0.034	31.3	16.60	27	3.6	12.45	b.c.c.
Tellurium	Te	841	1814	0.047	3.3	6.24	6	9.3	4.4×10^{5}	Hexag.
Thallium	Tl	577	2655	0.031	22.5	11.85	.....	16	18	h.c.p.
Thorium	Th	3182	7000	0.034	21.7	11.66	.....	6.9	13	f.c.c.
Tin	Sn	449	4120	0.054	36.2	7.30	6.3	13	11	Tetrag.
Titanium	Ti	3035	5900	0.124	9.8	4.51	16.8	4.67	42	h.c.p.
Tungsten	W	6170	10706	0.033	96.	19.30	50	2.55	5.65	b.c.c.
Uranium	U	2070	6904	0.028	17.1	19.07	24	3.8 to 7.8	30	Orthorhomb.
Vanadium	V	3450	6150	0.119	16.9	6.1	19	4.6	26	b.c.c.
Yttrium	Y	2748	5490	0.071	8.5	4.47	17	....	57	h.c.p.
Zinc	Zn	787	1663	0.092	65.	7.13	.....	22	5.92	h.c.p.
Zirconium	Zr	3366	6470	0.067	9.6	6.49	13.7	3.2	40	h.c.p.

*Courtesy of "Metals Handbook," vol. 1, 8th ed., American Society of Metals, Cleveland, 1961.

[a] Near 68°F (20°C).

[b] Sublimes; triple point at 2028 atm.

[c] Semiconductor.

TABLE 6.6 Typical Mechanical Properties of Cast Iron, Cast Steel, and Other Metals (Room Temperature)*

Metal	Modulus of elasticity E, million psi	Ultimate tensile strength σ_u, thousand psi	Yield strength σ_y, thousand psi	Endurance limit σ_{end}, thousand psi	Hardness, Brinell
Gray cast iron, ASTM 20, med. sec.	12	22	....	10	180
Gray cast iron, ASTM 50, med. sec.	19	53	....	25	240
Nodular ductile cast iron:					
Type 60-45-10	22–25	60–80	45–60	35	140–190
Type 120-90-02	22–25	120–150	90–125	52	240–325
Austenitic	18.5	58–68	32–38	32	140–200
Malleable cast iron, ferritic 32510	25	50	32.5	28	110–156
Malleable cast iron, pearlitic 60003	28	80–100	60–80	39–40	197–269
Ingot iron, hot rolled	29.8	44	23	28	83
Ingot iron, cold drawn	29.8	73	69	33	142
Wrought iron, hot rolled longit	29.5	48	27	23	97–105
Cast carbon steel, normalized 70000	30	70	38	31	140
Cast steel, low alloy, 100,000 norm. and temp	29–30	100	68	45	209
Cast steel, low alloy, 200,000 quench. and temp	29–30	200	170	85	400
Wrought plain C steel:					
C1020 hot rolled	29–30	66	44	32	143
C1045 hard. and temp. 1000°F	29–30	118	88	......	277
C1095 hard. and temp. 700°F	29–30	180	118	......	375
Low-alloy steels:					
Wrought 1330, HT and temp. 1000°F	29–30	122	100	......	248
Wrought 2317, HT and temp. 1000°F	29–30	107	72	......	222
Wrought 4340, HT and temp. 800°F	29–30	220	200	......	445
Wrought 6150, HT and temp. 1000°F	29–30	187	179	......	444
Wrought 8750, HT and temp. 800°F	29–30	214	194	......	423
Ultra high strength steel H11, HT	30	295–311	241–247	132	
300M HT and temp. 500°F	....	289	242	116	
4340 HT and temp. 400°F	30	287	270	107	
25 Ni Maraging	24	319	284		
Austenitic stainless steel 302, cold worked	28	110	75	34	240
Ferritic Stainless Steel 430, cold worked	29	75–90	45–80		
Martensitic Stainless Steel 410, HT	29	90–190	60–145	40	180–390
Martensitic Stainless Steel 440A, HT	29	260	240	......	510
Nitriding Steels, 135 Mod., hard. and temp. (core properties)	29–30	145–159	125–141	45–90	285–320
Nitiding Steels 5Ni-2A, hard. and temp	29–30	206	202	90	
Structural Steel	30	50–65	30–40	......	120
Aluminum Alloys, cast:					
195 SHT and aged	10.1	36	24	8	75
220 SHT	9.5	48	26	8	75
142 SHT and aged	10.3	28–47	25–42	9.5	75–110
355 SHT and aged	10.2	35–42	25–27	9–10	80–90
A13 as cast	10.3	39	21	19	

TABLE 6.6 Typical Mechanical Properties of Cast Iron, Cast Steel, and Other Metals (Room Temperature)* (*Continued*)

Metal	Modulus of elasticity E, million psi	Ultimate tensile strength σ_u, thousand psi	Yield strength σ_y, thousand psi	Endurance limit σ_{end}, thousand psi	Hardness, Brinell
Aluminum Alloys, wrought:					
EC ann	10	12	4		
ECH 19, hard	10	27	24	7	
3003 H 18, hard	10	29	27	10	55
2024 H T (T3)	10.6	70	50	20	120
5052 H 38, hard	10.2	42	37	20	77
7075 HT (T6)	10.4	83	73	23	150
7079 HT (T6)	10.3	78	68	23	145
Copper alloys, cast:					
Leaded red brass BB11-4A	9–14.8	33–46	17–24	......	55
Leaded tin bronze BB11-2A	12–16	36–48	16–21	......	60–72
Yellow brass BB11-7A	12–14	60–78	25–40	......	80–95
Aluminum bronze BB11-9BHT	15	90	40	......	180
Copper alloys, wrought:					
Oxygen-free 102 ann	17	32–35	10	11	
Hard	....	50–55	45	13	
Beryllium copper, 172 HT	19	165–183	150–170	35–40	
Cartridge Brass, 260 hard	16	76	63	21	
Muntz metal, 280 ann	15	54	21		
Admiralty, 442 ann	16	53	22		
Manganese bronze, 675 hard	15	84	60		
Phosphor bronze, 521 spring	16	112	70		
Silicon bronze, 647 HT	18	100	88		
Cupro-Nickel, 715 hard	22	80	73		
Magnesium Alloys, cast:					
AZ63A, aged	6.5	30–40	14–19	11–15	55–73
AZ92A, aged	6.5	26–40	16–21	11–15	80–84
HK31A, T6	6.5	31	16	9–11	55
Magnesium Alloys, wrought:					
AZ61AF, forged	6.5	43	26	17–22	55
ZE-10A-H24	6.5	34–38	19–28	20–24	
HM31A-T5	6.5	42	33	12–14	
Nickel-alloy castings 210	21.5	45–60	20–30	......	80–125
Monel 411 cast	19	65–90	32–45	......	125–150
Inconel 610 cast	23	70–95	30–45	......	190
Nickel alloys, wrought:					
200 Spring	30	90–130	70–115		
Duranickel, 301 Spring	30	155–190			
K Monel, K500 Spring	26	145–165	130–180		
Titanium alloys, wrought, unalloyed	15–16	60–110	40–95	60–70	
5Al-2.5 Sn	16–17	115–140	110–135	95	
13V-11 Cr-3Al HT	14.5–16	190–240	170–220	50–55	
Zinc, wrought, comm. rolled	....	25–31	....	4.1	
Zirconium, wrought:	14	64	53		
Reactor grade	14	49	29		
Zircaloy 2	13.8	68	61		
Pure metals, wrought:					
Beryllium, ann	44	60–90	45–55		
Hafnium, ann	20	77	32		
Thorium, ann	10	34	26		
Vanadium, ann	20	72	64		
Uranium, ann	30	90	25		

TABLE 6.6 Typical Mechanical Properties of Cast Iron, Cast Steel, and Other Metals (Room Temperature)* (*Continued*)

Metal	Modulus of elasticity E, million psi	Ultimate tensile strength σ_u, thousand psi	Yield strength σ_y, thousand psi	Endurance limit σ_{end}, thousand psi	Hardness, Brinell
Precious metals:					
Gold, ann	12	19	....	46	25
Silver, ann	11	22	8	....	25–35
Platinum, ann	21	17–26	2–5.5	....	38–52
Palladium, ann	17	30	5	....	46
Rhodium, ann	42	73	....	....	55–156
Osmium, cast	80	....	....	....	350
Iridium, ann	74	....	....	....	170

Babbitt has a compressive elastic limit of 1.3 to 2.5 × 10^3 lb/in^2 and a Brinell hardness of 20.

Compressive yield strength of all metals, except those cold-worked = tensile yield strength.

Poisson's ratio is in the range 0.25 to 0.35 for metals.

Yield strength is determined at 0.2 percent permanent deformation.

Modulus of elasticity in shear for metals is approximately 0.4 of the modulus of elasticity in tension E.

Compressive yield strength of cast iron 80,000 to 150,000 × 10^3 lb/in^2.

*From *Materials in Design Engineering*, Materials Selector Issue, vol. 56, no. 5, 1962; courtesy of Reinhold Publishing Corporation, New York.

TABLE 6.7 Mechanical Properties and Applications of Steels

Carbon steels*

AISI	Condition	Tensile strength, lb/in^2	Yield strength, lb/in^2	Elongation in 2 in, %	Brinell hardness	Machinability	Comments and uses
1020	Annealed	55,000	30,000	25	111	65	Weldable, structural beams, column: case hardener, gears, camshafts, kingpins (hard surface and soft core)
1040	Annealed	76,000	42,000	18	149	60	Not readily weldable. Hardened bolts (shallow), case-fused axles, shafts, chain pins, hobbing splints
1070	Annealed	102,000	56,000	12	212	55	Hardened springs, hand tools, saw material
1095	Annealed	120,000	66,000	10	248	45	Hardened springs, knives, ore screens

Alloy steels (quenched and tempered)†

AISI	Tempering temperature, °F	Tensile strength, lb/in^2	Yield strength, lb/in^2	Elongation, % in 2 in	Brinell hardness	Machinability (soft)	Comments and uses
1141	400 1000	237,000 130,000	176,000 111,000	6 18	461 262	70	Free-machining (sulfur); gears, transmission, and spline shafts works
1345	400 1000	270,000 140,000	240,000 124,000	10 16	530 310	45	Steering knuckle forgings
4140	400 1000	257,000 138,000	238,000 121,000	8 18	510 285	65	Can be nitrided, transmission shafts, tractor pins, pinions
4340	400 1000	272,000 170,000	243,000 156,000	10 13	520 360	50	Drums for helicoptor transmission (nitrided)
6150	400 1000	280,000 168,000	245,000 155,000	8 13	538 345	55	High-stress, high-temperature springs, die-casting dies, plastic molds
8650	400 1000	281,000 170,000	243,000 153,000	10 15	525 340	60	Springs, gears
9310	400	174,000	150,000	15	390	50	Carburized aircraft gears, high shock resistance, surface C not to exceed 0.9 percent to avoid retained austenite (poorer wear)

TABLE 6.7 Mechanical Properties and Applications of Steels (*Continued*)

Stainless and heat-resistant steels‡

	Typical composition			Mechanical properties at 70°F, annealed			Maximum continuous service temperature, °F (air)	
AISI	Cr	Ni	C	Tensile strength, lb/in^2	Yield strength, lb/in^2	Elongation, %		Characteristics and applications
205	17	1.8	0.20	120,000	69,000	58	—	Spinning and drawing operations, nonmagnetic cryogenic parts
304	18	8	0.08	85,000	35,000	60	1650	Chemical and food processing, cryogenic hypodermic needles
316	17	12	0.08	84,000	32,000	50	1650	Higher corrosion resistance, high creep strength, chemical and pulp equipment
317	18§	10§	0.08§	90,000	35,000	45	1650	Stabilized for weldments, exhaust manifold aircraft, fire walls, pressure vessels
403	12	—	0.15	70,000	45,000	25	1300	Martensitic, turbine quality, jet engine rings
440C	17	—	1.2	110,000	65,000	14	1400	Martensitic, highest hardness, balls, bearings, cases, and nozzles
S15500	17	3.5	0.07	160,000	145,000	15	—	Martensitic, maraging high strength, cams, gears, shafting

Tool Steels¶

		Typical composition					
AISI	Type	C	W	Nb	Cr	V	Applications
W1	Water hardening	0.6 1.4	— —	— —	— —	— —	Low C, blanking tools, forging dies; high C, lathe tools, reamers, taps
S1	Shock resisting	0.5	2.5	—	1.5	—	Pneumatic tools, swaging dies, shear blades, master hobs
O7	Oil hardening	1.2	1.75	—	0.75	—	Mandrels, slitters, drills, taps, blanking dies
A2	Age hardening	1.0	—	1.0	5.0	—	Thread rolling dies, extrusion dies, coining dies
D2	High C–high Cr	1.5	—	1.0	12.0	—	Blanking dies, roll thread dies, shear blades, slitters

AISI	Type	Typical composition					Applications
		C	W	Nb	Cr	V	
D3	High C–high Cr	2.25	—	—	12.0	—	More wear-resistant than D2
H12	Hot work	0.35	1.5	1.5	5.0	0.4	Extrusion dies, forging die inserts, punches
T1	High speed	0.75	18	—	4	1	Drills, taps, tool cutters, high hot hardness
M4	High speed	1.3	5.5	4.5	4	4	Lathe and planer tools, brooches, swaging dies
P2	Mold dies	0.07	—	0.2	2	0.5 Ni	Plastic molding dies (hobbed and carburized)

*From "Metals Handbook," vol. 1, 8th ed., American Society for Metals, 1961. Machinability values from "SAE Handbook", 1979; refer to a rating of 100 for AISI 1212 (a resulfurized and rephosphorized) free-machining steel in cold-drawn condition. These plain carbon steels have a low or shallow hardenability when heat-treated.

†From "Metal Progress Databook," 1981; "ASM Metals Handbook," vol. 1, 8th ed., American Society for Metals, 1961. These alloyed steels have greater hardenability than the plain carbon steels when heat-treated.

‡"Metal Progress Databook," 1981.

§Plus Ti (5 × C).

¶From "Metal Progress Databook," 1981.

TABLE 6.8 Typical Properties of Refractory Ceramics and Cermets and Other Materials

Type	Specific gravity	Melting point, °F	Mean specific heat, Btu/lb/°F	Mean coefficient of thermal expansion, μ in./in./°F	Mean thermal conductivity, Btu/hr/sq ft/°F/ft	Hardness, Mohs scale	Maximum service temperature (oxidiz.), °F
Alumina (99+) (Al_2O_3)	3.85	3725	0.23	4.3	10.7	9	3540
Beryllia (BeO)	3.0	4620	0.29	5.3	9.5	9	4350
Magnesia (MgO)	3.6	5070	0.26	7.8	1.47	6	4350
Thoria (ThO_2)	....	6000	....	5.28	0.0	7	4890
Zirconia (ZrO_2)	5.5–6	4710	0.16	3.06	0.53	7–8	4530
Quartzite (SiO_2)	2.65	2552	0.26	0.28	0.8		
Silicon carbide (Dens)(SiC)	3.2	......	0.33	2.17	25	9–10	3000
Boron carbide	2.5	......	....	1.73	16	....	1000
Titanium carbide (TiC)	6.5	......	....	4.3–7.5			
Tungsten carbide (WC)	14.3	......	....	2.5–3.9	26–50		
Boron nitride	2.05–2.15	4930	....	5.5	10–20		
Graphite	2.25	......	0.18	1.0–1.3	70–120	1.2	

TABLE 6.9 Typical Properties of Plastics at Room Temperature

Type	Specific gravity	Coefficient of thermal expansion, 10^{-5}/°F	Thermal conductivity, Btu/hr/sq ft/°F/ft	Volume resistivity, ohm-cm	Dialectric strength $^{(a)}$, volts/mil	Modulus of elasticity in tension, 10^5 psi	Tensile strength, 10^3 psi
Acrylic, general purpose, type I	1.17–1.19	4.5	0.12	$>10^{15}$	450–530	3.5–4.5	6–9
Cellulose acetate, type I (med.)	1.24–1.34	4.4–9.0	0.1–0.19	10^{12}	250–600	....	2.7–6.5
Epoxy, general purpose	1.12–2.4	1.7–5.0	0.1–0.8	10^{13}	350–550	....	2–12
Nylon 6	1.13–1.14	4.6–5.4	0.1–0.14	10^{14}	420–485	2.5–3.4	10.2–12
Phenolic, type I (mech.)	1.31	3.3–4.4	....	1.7×10^{12}	350–400	4–5	6–9
Polyester, Allyl type	1.30–1.45	2.8–5.6	0.12	$>10^{13}$	330–500	2–3	4.5–7
Silicone, general (mineral)	1.80–2.0	2.8–3.2	0.09	$>10^{13}$	350–400	....	4.2
Polystyrene, general purpose	1.04–1.07	3.3–4.8	0.06–0.09	10^{18}	>500	4–5	5–8
Polyethylene, low density	0.92	8.9–11	0.19	10^{18}	480	0.22	1.4–2
Polyethylene, medium density	0.93	8.3–16.7	0.19	$>10^{15}$	480	....	2
Polyethylene, high density	0.96	8.3–16.7	0.19	$>10^{15}$	480	....	4.4
Polypropylene	0.89–0.91	6.2	0.08	10^{16}	769–820	1.4–1.7	5

$^{(a)}$ Short time.

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