METAL TREATMENTS

Operations performed on metals previously consolidated by processes such as melting and casting are referred to as metal treatments. Most of these treatments are mechanical and/or thermal. Mechanical treatments involve changes in shape by forming or machining. Metal treatments such as joining and coating of metals are not discussed herein (see Metallic coatings; Welding).

1. Mechanical Forming

Forming processes and techniques that are available for a particular alloy depend on its workability, which is the ability to be plastically deformed.

1.1. Workability Testing

Workability tests measure the amount of deformation that can be tolerated without fracture, or the development of an instability such as buckling or necking. Buckling or wrinkling is shown at the flange of the cup of Figure 1 where the metal was too thin or was insufficiently supported. Local thinning or necking occurred in the walls of the cup. In most workability tests a specimen is deformed to failure at a constant load rate or strain rate in tension, compression, torsion, shear, or bending. The most common technique is a tensile test at a constant strain rate where the load and elongation are measured continuously. Stress and strain are calculated from load and elongation and are presented in the form of an engineering stress–engineering strain diagram (Fig. 2). Yielding represents the transition from elastic deformation, where atomic bonds are being stretched, to plastic or nonrecoverable deformation, where atomic slip is occurring. The yield stress, which specifies this transition, is defined usually as the stress (load per area) that produces a small permanent strain, usually 0.002 (0.2% offset yield stress). Following yielding, the stress required for further strain increases, but unlike conditions in the elastic region, stress and strain are not linearly related. Increasing stress with increasing strain in the plastic region is termed strain hardening. The decrease in stress following the maximum load is a result of necking or localized deformation. Engineering stress is decreasing in Figure 2 only because the area in the necked region is rapidly decreasing. The true stress or actual load per unit area continues to increase with strain all the way to fracture. The onset of necking is defined by the ultimate tensile stress (UTS) which is the maximum load divided by the initial area. Ductility is measured by % elongation (El) and % reduction in area (RA) which are defined as

$$\%\ \mathrm{El} = \left(\frac{L - L_0}{L_0}\right) \times 100$$

where L_0 = initial length; L = final length;

$$\%\ \mathrm{RA} = \left(\frac{A_0 - A}{A_0}\right) \times 100$$

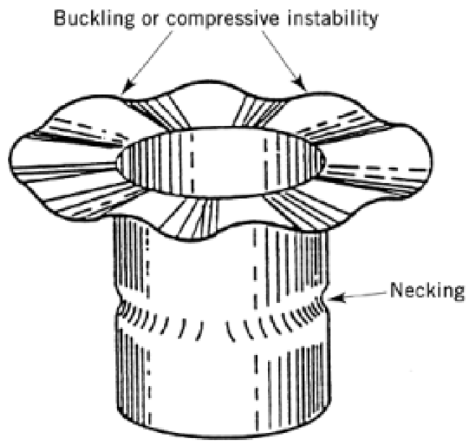

Fig. 1. Examples of instabilities in a deep-drawn cup.

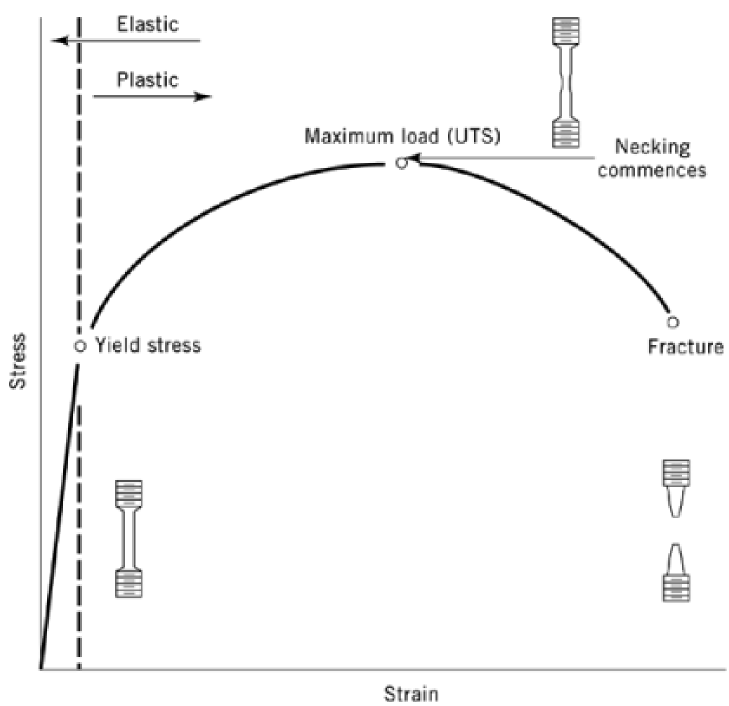

Fig. 2. Schematic stress–strain diagram, where UTS=ultimate tensile stress and (— — —) represents the demarcation between elastic and plastic behavior. See text.

where A_0 = initial area; A = final area.

Both % El and % RA are frequently used as a measure of workability. Workability information also is obtained from parameters such as strain hardening, yield strength, ultimate tensile strength, area under the stress–strain diagram, and strain-rate sensitivity.

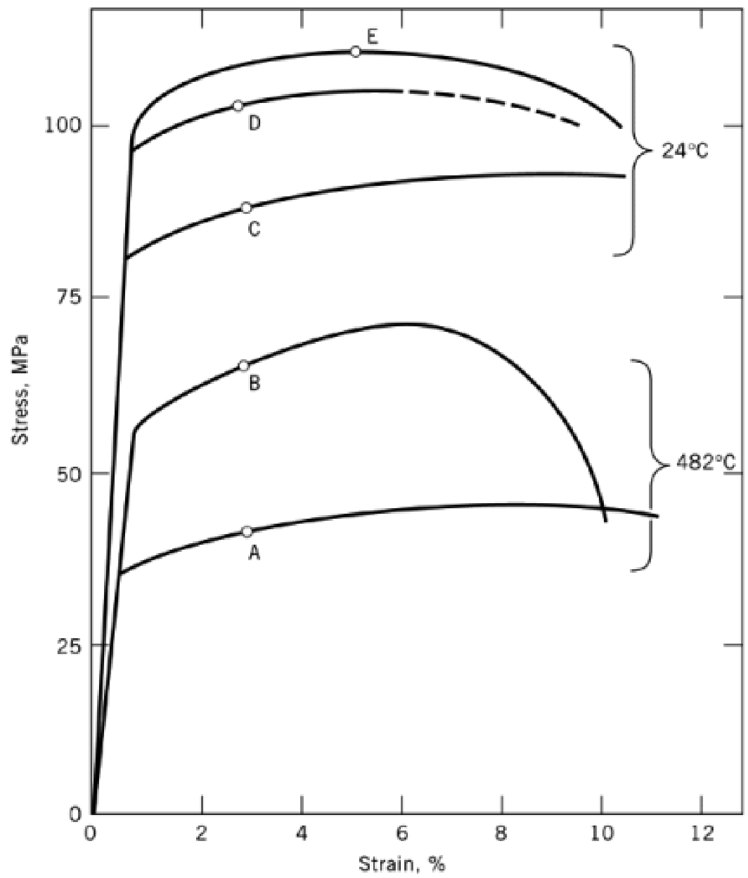

Fig. 3. Effect of temperature and strain rate on stress–strain diagram of Ti–5%Al–2.5%Sn where A–E correspond to the strain rates 1.6×10^{-4}, 5×10^{-1}, 5×10^{-6}, 1.6×10^{-4}, and 5×10 $s-1$, respectively. The dashed line represents an extrapolation of D (1).

Tensile testing frequently is used to assess mechanical properties other than workability. However, the strain rate is usually much faster when workability is being measured in order better to simulate forming processes. Standard testing is done at about 10^{-3} $s-1$, compared to strain rates up to 10^{2} $s-1$ for workability testing. An indication of strain-rate sensitivity is given in Figure 3 for a commercial titanium alloy (1) (see Titanium and titanium alloys). Stress–strain diagrams are not unique to tensile testing. These diagrams are also generated by other testing modes such as compression, torsion, shear, and bending.

Temperature strongly influences stress–strain behavior (Fig. 3). Therefore, evaluating hot workability entails testing over a range of temperatures. Hot-torsion data are presented for two nickel-base superalloys, Nimonic 90 and Nimonic 115, and for a 0.48% C steel over a range of temperatures in Figure 4 (2) (see Nickel and nickel alloys; Steel). Strength is indicated by torque, and ductility is measured by the number of revolutions to failure. It can be seen that the hot ductility of the nickel-base alloys, particularly N-115, is significantly less than that of steel. Furthermore, the required stresses are substantially greater.

When determining the temperature range for hot working, it is usually not sufficient to merely heat directly to various temperatures. Instead, it is also necessary to acquire cooling data on specimens that are tested after cooling from a temperature corresponding to the furnace temperature in a hot-working operation.

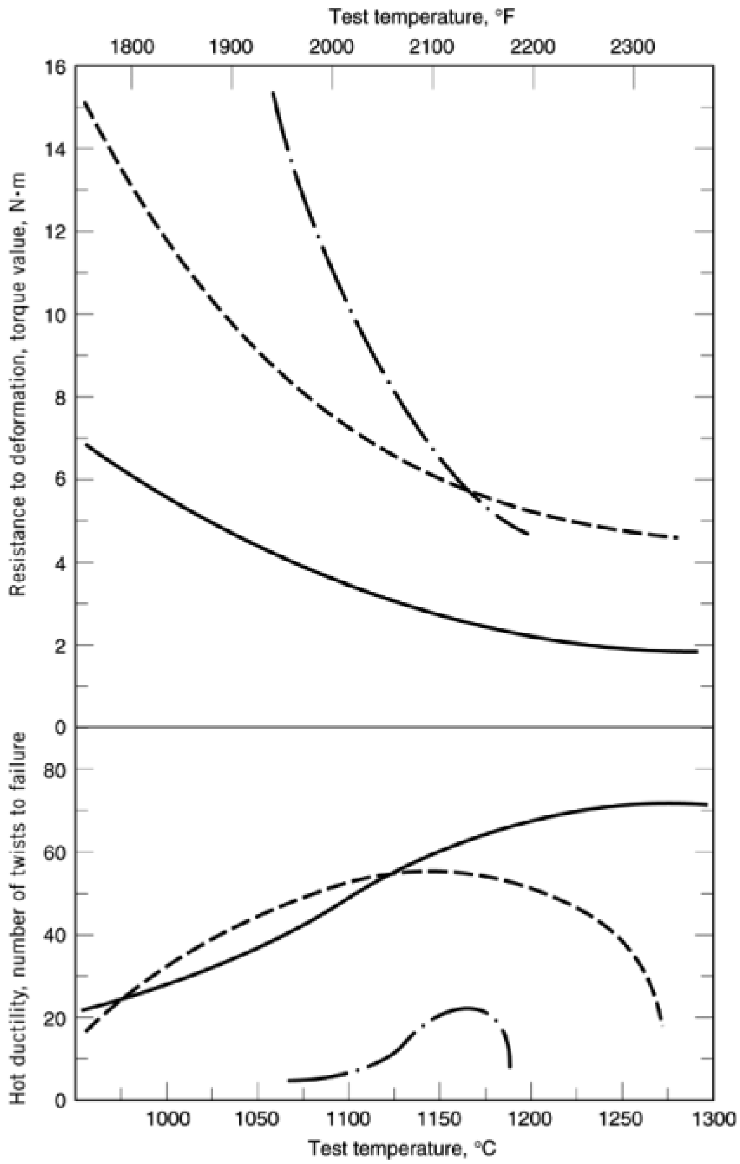

Fig. 4. Torsion properties versus temperature for the nickel alloys (— • —) Nimonic 115 and (— — —) Nimonic 90 and for (—) 0.48%C steel (2). To convert N·m to ft·lbf, divide by 1.35.

Both heating and cooling tensile data are shown in Figure 5 for a Nimonic 115 ingot. The ductility on testing is lower after cooling from 1105 and 1135°C compared to that determined on heating. These tensile tests were performed on a Gleeble machine, where the specimen is resistance-heated.

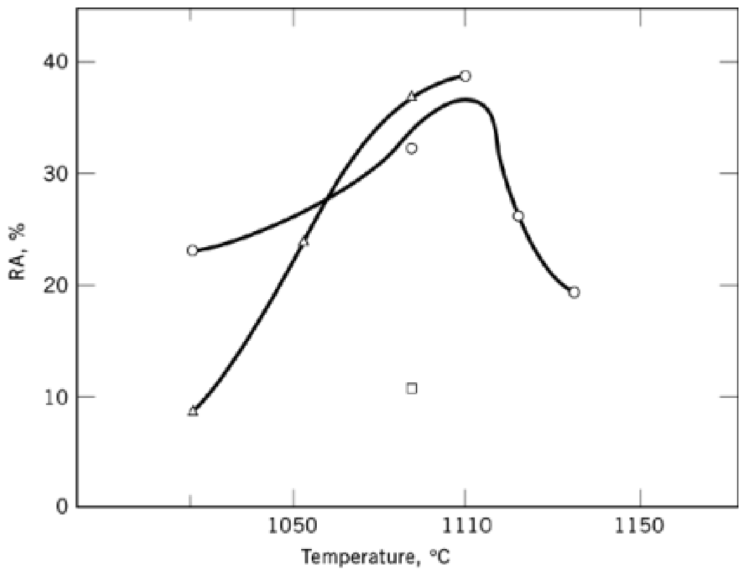

Fig. 5. Hot workability of cast Nimonic 115 as determined by tensile testing using a Gleeble machine: (○), heating; (□), cooling 1135°C; (△), cooling 1105°C. RA=reduction in area.

2. Plastic Deformation

When plastic deformation occurs, crystallographic planes slip past each other. Slip is facilitated by the unique atomic structure of metals, which consists of an electron cloud surrounding positive nuclei. This structure permits shifting of atomic position without separation of atomic planes and resultant fracture. The stress required to slip an atomic plane past an adjacent plane is extremely high if the entire plane moves at the same time. Therefore, the plane moves locally, which gives rise to line defects called dislocations. These dislocations explain strain hardening and many other phenomena.

Dislocation may be either edge or screw type. In slipping under the application of the shear stress shown in Figure 6, two distinctively different intermediate stages are possible. The stage shown in Figure 6**b** represents an edge dislocation, whereas in Figure 6**c** a screw dislocation is shown. An edge dislocation contains an extra half-plane of atoms at the slip plane as shown in Figure 6**e**, the front view of Figure 6**b**. The term screw dislocation is derived from the fact that following this line defect from above and below the slip plane generates a screw pattern as shown in Figure 6**f**. A given dislocation line can be pure edge, pure screw, or any combination of edge and screw components.

An analogy to slip dislocation is the movement of a caterpillar where a hump started at one end moves toward the other end until the entire caterpillar moves forward. Another analogy is the displacement of a rug by forming a hump at one end and moving it toward the other end. Strain hardening occurs because the dislocation density increases from about 10^7 dislocations/cm^2 to as high as 10^{13}/cm^2. This makes dislocation motion more difficult because dislocations interact with each other and become entangled. Slip tends to occur on more closely packed planes in close-packed directions.

2.1. Hot Working

Plastic deformation at temperatures sufficiently high that strain hardening does not result is termed hot working. The temperature range for successful hot working depends on composition and other factors such

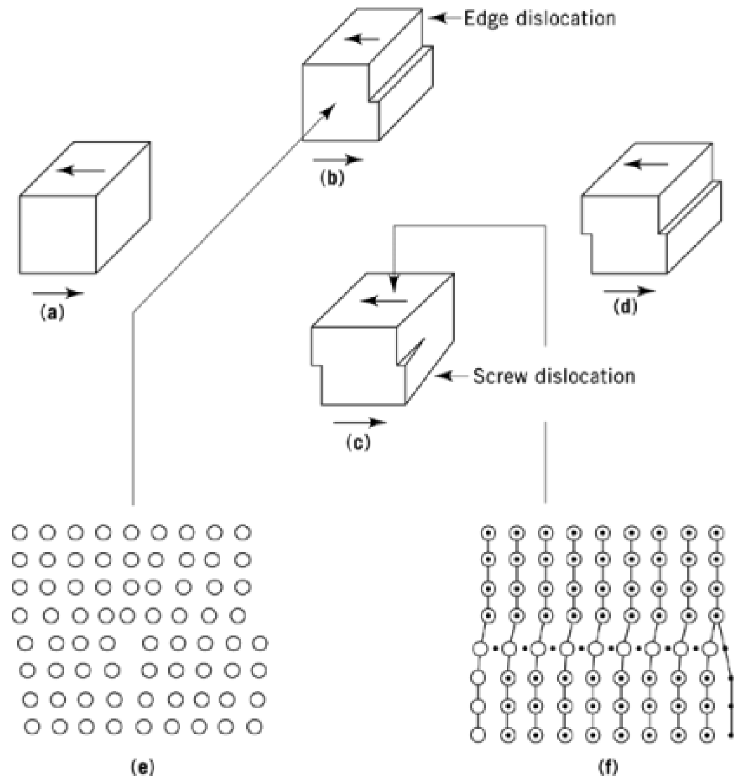

Fig. 6. Continuum (**a–d**) and atomic (**e–f**) representations of edge and screw dislocations. For the screw dislocations, open circles and dots correspond to atoms above and below the slip plane, respectively.

as grain size, previous cold working, reduction, and strain rate. Typical hot-working ranges are presented in Figure 7 for various alloys.

The lack of strain-hardening results from sufficient thermal energy for recrystallization, which refers to the formation of new grains. Because the new grains have relatively low dislocation densities, strength is not increased. The driving force for recrystallization is the large strain energy associated with the high dislocation density generated during deformation. Recrystallization for hot rolling is shown schematically in Figure 8. The recrystallized grain size is small, but growth occurs with time at a temperature dependent on the alloy.

Hot working permits forming of relatively brittle materials that cannot readily be cold worked. Other advantages are grain refinement, reduction of segregation, healing of defects, such as porosity, and dispersion of inclusions. Disadvantages are the formation of oxide surface scales and the requirement of heating facilities.

2.2. Cold Working

Cold working involves plastic deformation well below the recrystallization temperature. Required stresses for cold working are greater than for hot working and the amount of strain without heat treatment is limited. Advantages are close dimension control, good surface finish, and increased low temperature strength because of strain hardening. Grain refinement can be achieved by annealing, which entails heating after cold working to temperatures at which recrystallization occurs. The effect of cold working on tensile properties and grain structure and subsequent annealing are shown schematically in Figure 9.

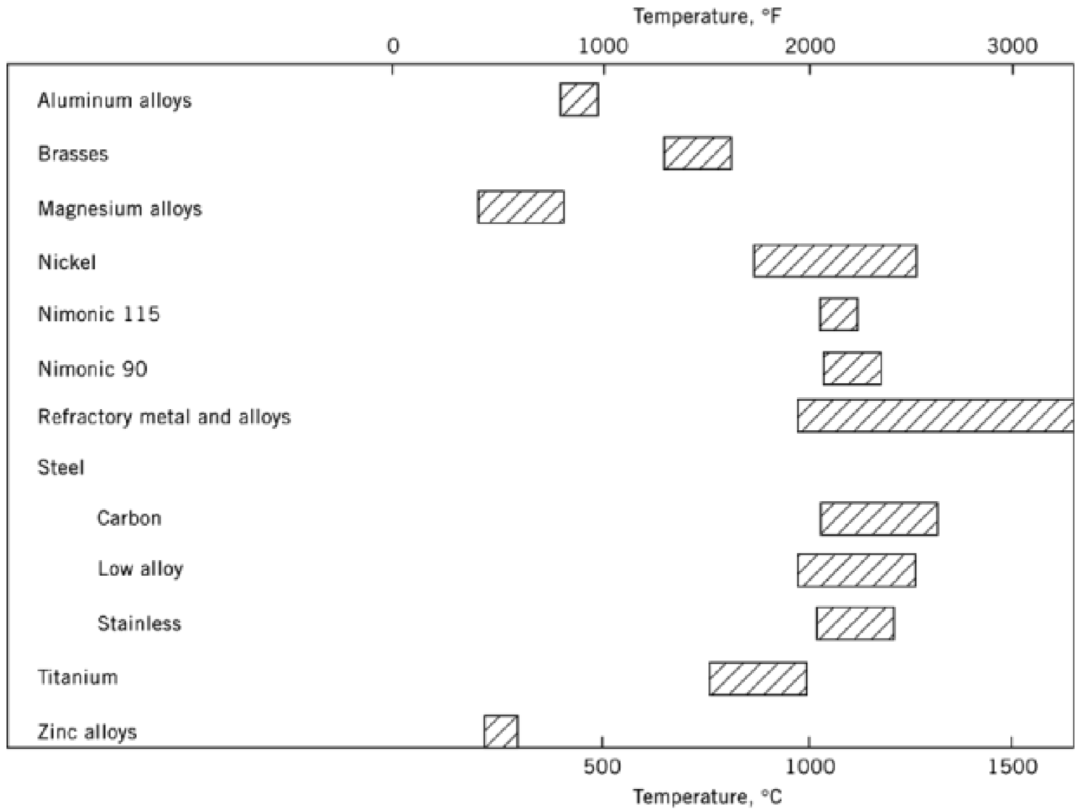

Fig. 7. Hot-working ranges of various metals and alloys.

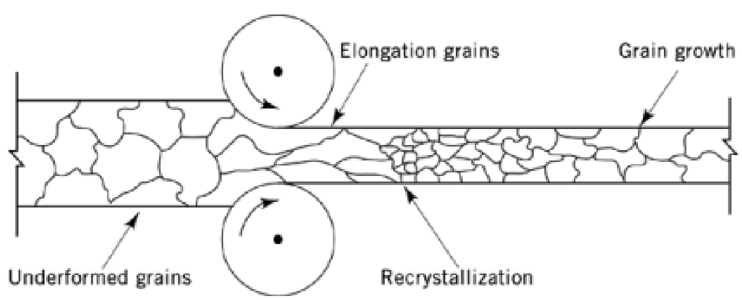

Fig. 8. Recrystallization and grain growth during hot rolling.

2.3. Primary Forming Processes

Primary forming operations are usually hot-working operations directed toward converting cast ingots into wrought blooms, billets, bars, or slabs (Fig. 10). In primary working operations the large grains typical of cast structures are refined, porosity is reduced, segregation is reduced, inclusions are more favorably distributed, and a shape desirable for subsequent operations is produced. Figure 11 illustrates the principal operations used for ingot breakdown, ie, forging, extruding, and rolling. Extrusion differs from forging and rolling in that more deformation occurs in one pass. Forging and rolling include many passes and some reheating. In addition,

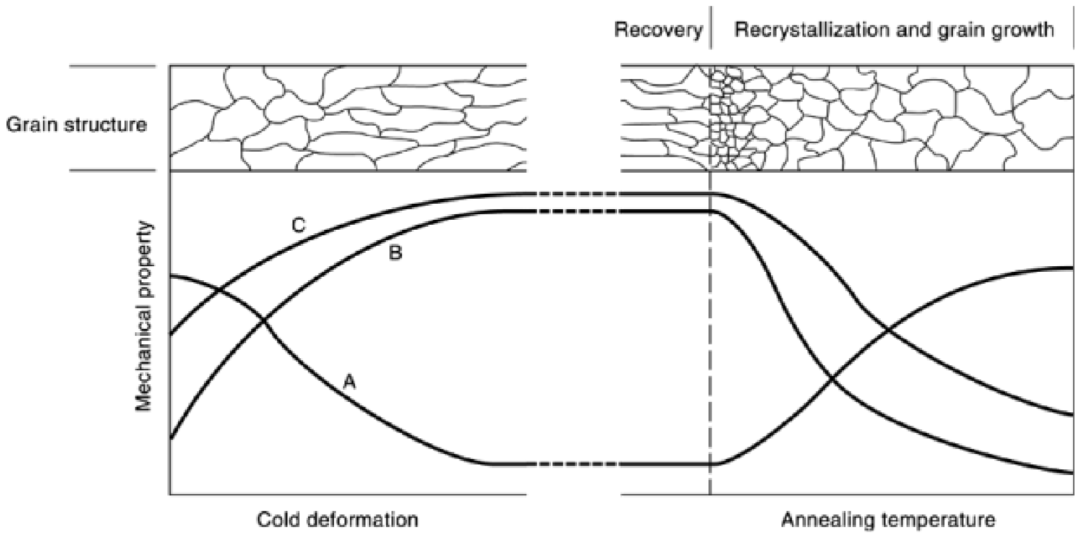

Fig. 9. Variation of tensile properties and grain structure with cold working and annealing: A, elongation; B, yield stress; and C, ultimate tensile stress.

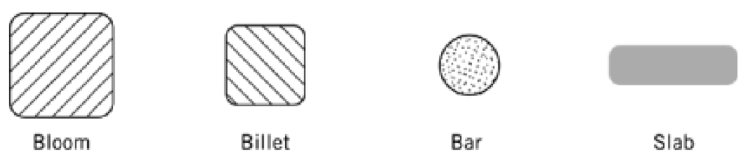

Fig. 10. Distinctions between various intermediate wrought shapes.

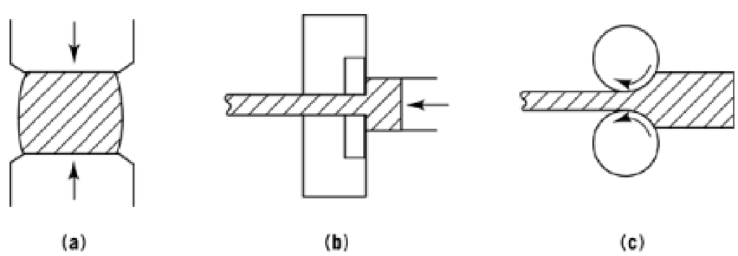

Fig. 11. Primary hot-working processes for ingot breakdown: (**a**) forging; (**b**) extruding; and (**c**) rolling.

intermediate conditioning is sometimes necessary. This makes extrusion attractive but it is an expensive operation.

2.4. Secondary Forming Processes

The objective of secondary forming processes, either cold- or hot-working, is to form a shape. Such processes include rolling, open- and closed-die forging, upset forging, extruding, roll forging, ring rolling, deep drawing, spinning, bending, stretching, stamping, drawing, and high velocity forming.

Sheet and semifinished products such as round, rectangular, and shaped bars are produced by rolling. Flat, V-shaped, and swaging dies are used for open-die forging. In closed-die forging, the metal is forced to flow

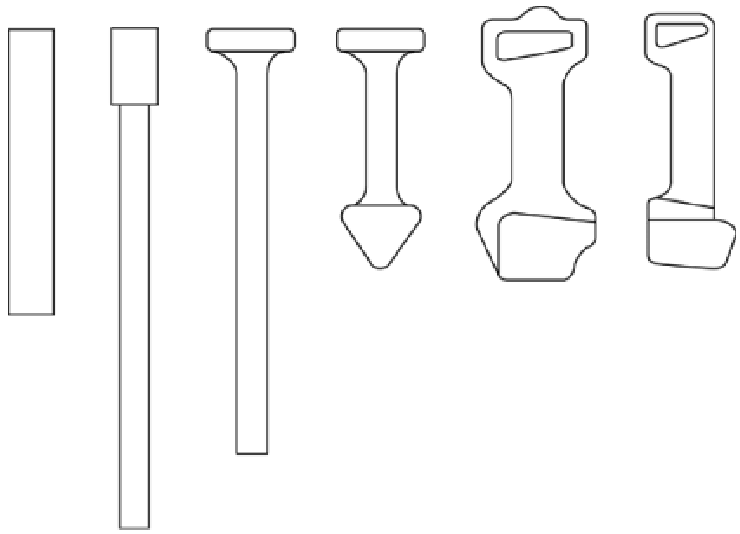

Fig. 12. Typical steps in forging a turbine blade.

into die cavities to form the impressions of dies attached to the anvil and ram. Forging is performed on both hammers and presses. Hammers have a strain rate of $1 - 100\ \mathrm{s} - 1$ as compared to $0.05 - 5\ \mathrm{s} - 1$ for presses. The energy or stress required for deformation is greater for hammers because of the faster strain rate. (Stress dependance on strain rate is illustrated in Fig. 3). Die contact time and, therefore, die chilling are shorter for hammers. However, the faster strain rate can more readily cause an excessively increased temperature locally, resulting in localized grain growth and possibly incipient melting. The propensity for cracking caused by brittle second-phase particles is less for the slower strain rates of presses.

In a typical process used to manufacture a forged turbine blade, the first three operations involve gathering material for subsequent closed-die forging (Fig. 12). Gathering is accomplished by one extrusion and two upsetting operations. Initial forging results in a perform that is of the correct volume to produce the finished forging in the final operation. A lubricant must be used during closed-die forging to minimize sticking to the die and to promote metal flow. Forging operations must be designed to provide adequate metal flow to prevent critical grain growth.

Extrusion for gathering and producing shapes can entail significantly more than direct forward extrusion. An example of backward extrusion is given in Figure 13**a**.

Roll forging differs from rolling, producing a short length of varying cross section, as opposed to the long, uniform cross sections produced by rolling. In ring rolling, ring-shaped forgings are produced. A seamless ring is produced by reducing the cross section and increasing the circumference of a heated, doughnut-shaped blank between two rotating rolls.

Deep drawing, spinning, bending, stretching, and stamping are cold-working processes applicable to the forming of shapes from sheet and strip (see Fig. 13**b**–**f**). Deep drawing uses a shaped punch to force sheet metal into a die or through a die opening. The drawing metal must have excellent ductility and several draws may be required, with annealing between draws. In spinning, a tool is pressed against one side of a sheet-metal die which is rotated at high speed. The tool gradually makes the disk conform to the shape of the forging instrument that it is forced against. Spinning is used in place of deep drawing if production does not justify the high cost of deep-drawing punches and dies. Stamping is used for cutting blanks from sheet and strip, and usually precedes deep drawing and spinning. Embossing and coining are also stamping operations. In

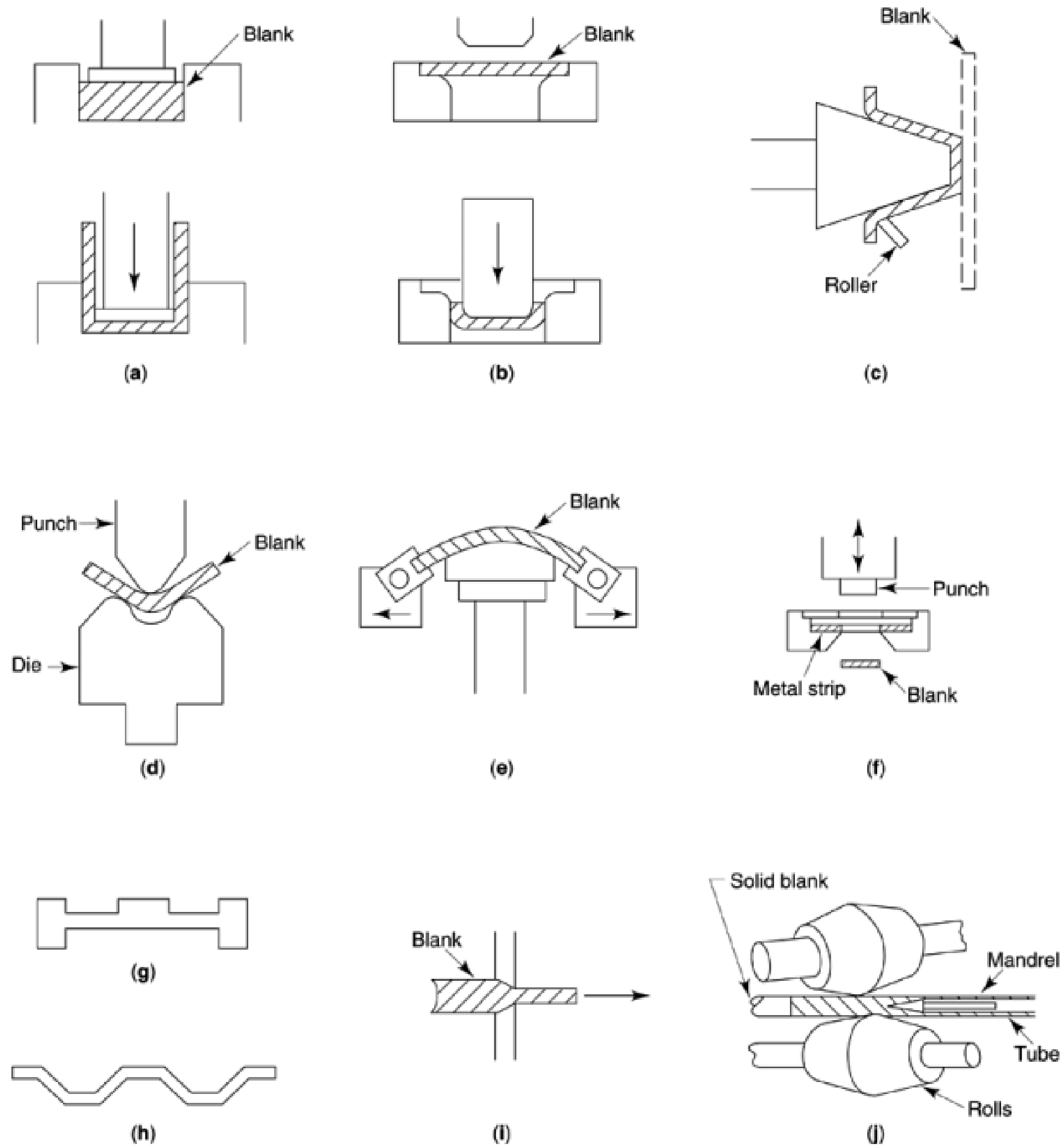

Fig. 13. Various forming operations: (**a**) backward extrusion; (**b**) deep drawing; (**c**) spinning; (**d**) bending; (**e**) stretching; (**f**) stamping; (**g**) coining; (**h**) embossing; (**i**) drawing; (**j**) rotary piercing.

embossing, the impressions of punch and die match each other, whereas in coining they do not, as shown in Figure 13**g** and **h**. Therefore, embossing results in more bending and less flow of metal than coining.

Drawing is a method of reducing the diameter of wire, rod, and tubing. It is similar to extrusion, except that the metal is pulled through the die instead of pushed through it, as shown in Figure 13**i**.

Tubes may be seamed or seamless. Seamed tubes are produced by bending plate, sheet, or strip into the appropriate shape and welding longitudinally. Seamless tubes are manufactured by opening the center of an ingot or billet and working the resulting shell or by working a cast hollow ingot. This may involve extrusion and subsequent drawing. Another method for producing a seamless tube is rotary piercing where metal is rolled over a mandrel, as shown in Figure 13**j**.

High velocity forming has become successful for alloys having poor workability. Examples of such processes are explosive forming, electromagnetic forming, and electrohydraulic forming. In explosive forming, an explosive charge is detonated in a water tank containing the workpiece and die. Shock waves from the explosion propagate throughout the liquid and impact the workpiece with sufficient energy to force it into the die (see Metallic coatings, explosively clad metals).

3. Thermal Treatments

3.1. Annealing

In annealing, a cold-worked material is heated to soften it and improve its ductility. The three stages of annealing are recovery, recrystallization, and grain growth (see Fig. 9). Recovery occurs at relatively low temperature and may result in some softening caused mainly by the arrangement of dislocations into a more favorable distribution. Recrystallization is the formation of new grains with a relatively low dislocation density and little internal strain, which replaces strained grains with high dislocation densities. At increasing temperature, the newly formed grains exhibit grain growth. Prolonged exposure at a given temperature also tends to promote grain growth.

3.2. Precipitation Hardening

In precipitation hardening, also called age hardening, fine particles are precipitated from a supersaturated solid solution. These particles impede the movement of dislocations, thereby making the alloy stronger and less ductile. In order for an alloy to exhibit precipitation hardening, it must exhibit partial solid solubility and decreasing solid solubility with decreasing temperatures.

An example of the many alloy systems satisfying these requirements is the aluminum–copper system. The diagram in Figure 14 shows a portion of the equilibrium phase diagram for the binary Al–Cu system, including the phases existing under equilibrium conditions at various temperatures as Cu is added to Al. At about 500–600°C, an alloy containing 4.5% Cu consists only of alpha, a solid solution of Cu in Al. Below 500°C, the phase θ ($CuAl_2$) exists in addition to alpha. The objective of precipitation hardening is to distribute the second phase as fine particles which are effective in blocking dislocation motion.

Precipitation hardening consists of solutioning, quenching, and aging. Solutioning entails heating above the solvus temperature in order to form a homogeneous solid solution.

Rapidly quenching to room temperature retains a maximum amount of alloying element (Cu) in solid solution. The cooling rate required varies considerably with different alloys. For some alloys, air cooling is sufficiently rapid, whereas other alloys require water-quenching. After cooling, the alloy is in a relatively soft metastable condition referred to as the solution-treated condition.

In aging, the alloy is heated below the solvus to permit precipitation of fine particles of a second phase θ ($CuAl_2$). The solvus represents the boundary on a phase diagram between the solid-solution region and a region consisting of a second phase in addition to the solid solution.

The precipitation-hardening process and resulting structures are shown in Figure 15. Particles formed initially during aging tend to fit into the lattice of the matrix solid solution, which distorts the lattice at the particle–matrix interface. This accommodation to the matrix phase is termed coherency, and contributes significantly to dislocation blockage, and, therefore, strengthening.

An optimum combination of temperatures and time generates particle size and spacing for the best combination of properties. If aging temperatures are too high, and times too long, particles coalesce and lose

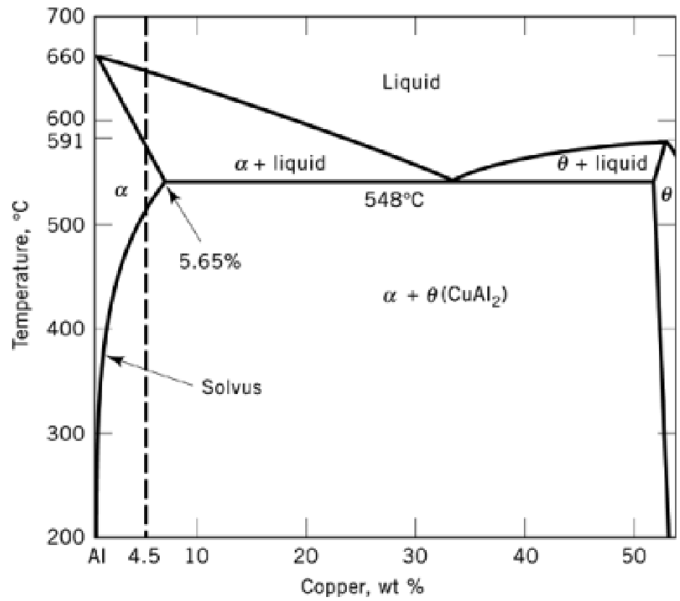

Fig. 14. Aluminum-rich portion of aluminum–copper-phase diagram.

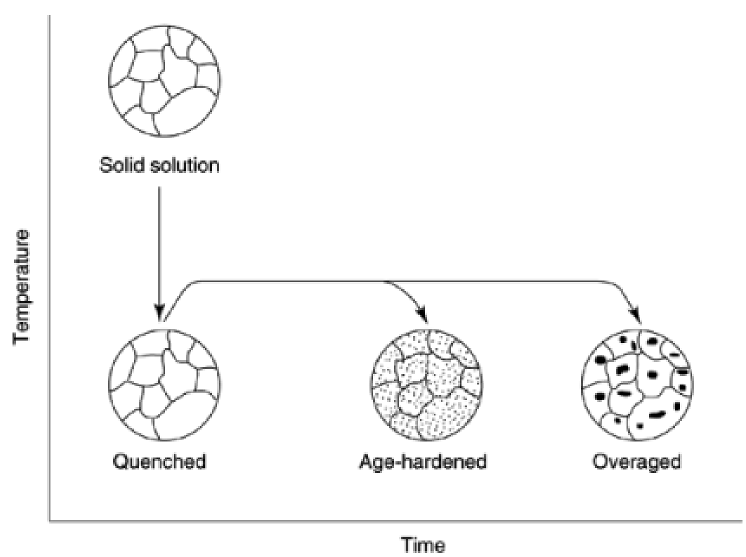

Fig. 15. Microstructures during precipitation hardening.

coherency, resulting in decreased strength, as shown in Figure 16. Table 1 gives the effect of precipitation hardening on a 95.5% Al–4.5% Cu alloy.

An example of a precipitation-hardened microstructure is presented in Figure 17 for Udimet 700, which is a precipitation-strengthened nickel-base superalloy. The precipitated phase is $Ni_3(Ti,Al)$, called gamma prime, and is the primary strengthening phase in many commercial superalloys (see High temperature alloys). Both coarse and fine gamma prime are present owing to high and low aging temperatures.

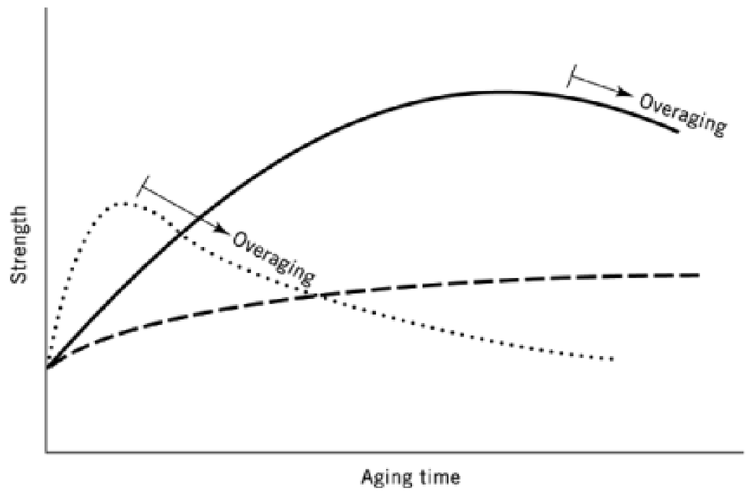

Fig. 16. Influence of temperature and time on strength during aging where (—) corresponds to the optimum temperature, and (— — —) and (...) below and above optimum, respectively.

Table 1. Effect of Heat Treatment on Tensile Properties of Al–4.5% Cu[a]

Treatment	Yield strength, MPa[b]	Ultimate tensile strength, MPa[b]	Elongation, %
solution-treated	103	241	30
age-hardened	331	414	20
overaged	69	117	20

[a] Heat treatment consisted of a solution treatment at 515°C, a water quench, followed by aging at 155°C for 10 h.
[b] To convert MPa to psi, multiply by 145.

3.3. Heat Treatment of Steel

Steels are alloys having up to about 2% carbon in iron plus other alloying elements. The vast application of steels is mainly owing to their ability to be heat treated to produce a wide spectrum of properties. This occurs because of a crystallographic or allotropic transformation which takes place upon quenching. This transformation and its role in heat treatment can be explained by the crystal structure of iron and by the appropriate phase diagram for steels (see Steel).

Iron (qv) exists in three allotropic modifications, each of which is stable over a certain range of temperatures. When pure iron freezes at 1538°C, the body-centered cubic (bcc) δ-modification forms, and is stable to 1394°C. Between 1394 and 912°C, the face-centered cubic (fcc) γ-modification exists. At 912°C, bcc α-iron forms and prevails at all lower temperatures. These various allotropic forms of iron have different capacities for dissolving carbon. γ-Iron can contain up to 2% carbon, whereas α-iron can contain a maximum of only about 0.02% C. This difference in solubility of carbon in iron is responsible for the unique heat-treating capabilities of steel. The solid solutions of carbon and other elements in γ-iron and α-iron are called austenite and ferrite, respectively.

The phase diagram for the iron-rich side of the iron–carbon system is shown in Figure 18. This diagram does not truly represent equilibrium because cementite [12169-32-3], Fe_3C, is a metastable phase. The stable phase for carbon is graphite, but the decomposition of the metastable Fe_3C to graphite and iron is so sluggish that Fe_3C must be treated as the stable phase for most practical purposes. At a composition of 0.77% C and a temperature of 727°C, a eutectoid reaction ($\text{Solid 1} \longrightarrow \text{Solid 2} + \text{Solid 3}$) occurs. The product of this reaction is

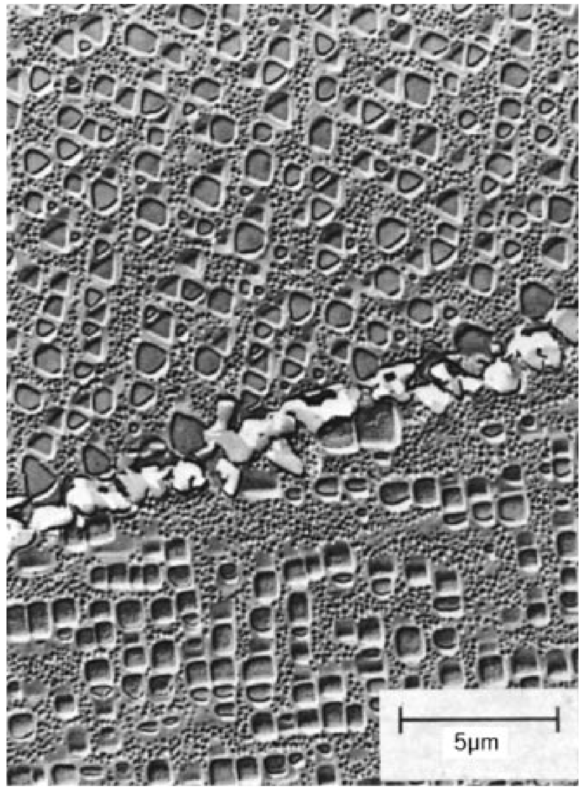

Fig. 17. Structure of U-700 after precipitation hardening temperature of 1168°C/4 h+1079°C/4 h+843°C/24 h+760°C/16 h with air cooling from each temperature. A grain boundary with precipitated carbides is passing through the center of the electron micrograph. Matrix precipitates are γ'-$Ni_3(TiAl)$.

pearlite, which has a lamellar structure consisting of alternate plates of ferrite and cementite. As the cooling rate increases, the interlamellar spacing decreases and the pearlite becomes finer, as indicated in Figure 19. Pearlite is not a phase in the thermodynamic sense but rather a constituent consisting of two phases, ie, ferrite and cementite.

If the quenching rate is so rapid that pearlite does not form because diffusion cannot occur, another phase, termed martensite, forms, as shown in Figure 19. Martensite is a supersaturated solid solution of carbon in α-iron. It has a body-centered tetragonal crystal structure. Carbon retained in solution distorts the lattice in one edge direction. The strains generated by the carbon in solution impede the movement of dislocations, which results in tremendous strengthening. As shown in Table 2, the strength can increase by almost a factor of 3 by quenching rapidly to form martensite, as compared to cooling slowly to form coarse pearlite. Because martensitic structures have such low ductilities, these are usually tempered following quenching. Tempering entails heating above 100°C to precipitate Fe_3C.

Other common heat-treatment processes are annealing and normalizing. Annealing is usually applied to produce softening and involves heating and cooling. Normalizing is a process in which a steel is heated into the austenite region and then air-cooled. The objective is to obliterate the effects of any previous heat treatment or cold working and to ensure a homogeneous austenite on reheating for hardening or full annealing.

Structures that form as a function of temperature and time on cooling for a steel of a given composition are usually represented graphically by continuous-cooling and isothermal-transformation diagrams. Another

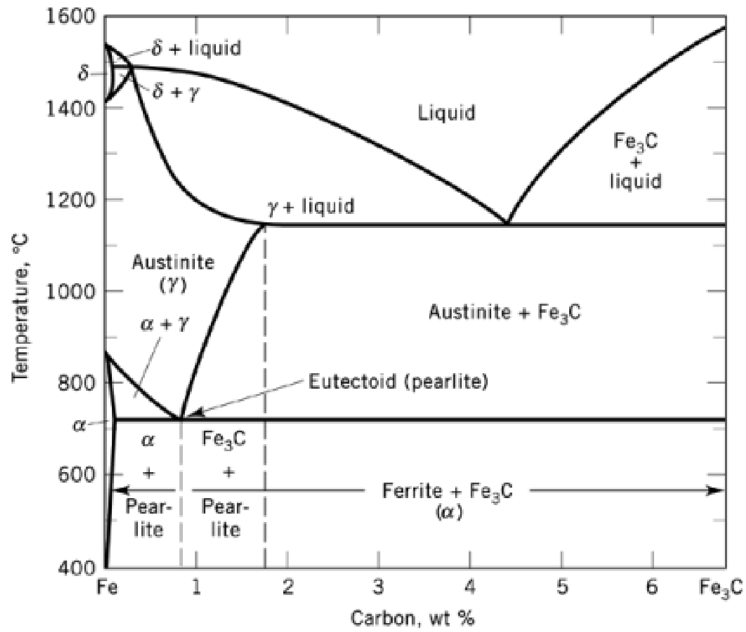

Fig. 18. Iron–iron carbide phase diagram.

Table 2. Properties of Steel Structures for a Eutectoid Steel

Structure	Yield strength, MPa[a]	Ultimate tensile strength, MPa[a]	Elongation, %
coarse pearlite	372	621	24
fine pearlite	524	1010	20
martensite		1724	low

[a] To convert MPa to psi, multiply by 145.

constituent that sometimes forms at temperatures below that for pearlite is bainite, which consists of ferrite and Fe_3C, but in a less well-defined arrangement than pearlite. There is not sufficient temperature and time for carbon atoms to diffuse long distances, and a rather poorly defined acicular or feathery structure results.

3.4. Homogenization

When alloys solidify, substantial segregation occurs. Therefore, ingots are sometimes given a high temperature heat treatment to generate a more homogeneous structure by diffusion in the solid state. Also, homogenization may eliminate undesirable phases that are present in the segregated cast structure.

3.5. Thermomechanical Processing

It is possible to develop desirable structures and, therefore, properties by uniquely combining thermal treatments and forming operations. An example of such thermomechanical processing is Minigrain processing (3) to produce fine grains in alloys such as Inconel 718 and Incoloy 901. These alloys are representative of iron–nickel base alloys that precipitate a second phase in addition to that primarily responsible for precipitation strengthening. Figure 20 shows an example of Minigrain processing for Incoloy 901. The primary strengthening

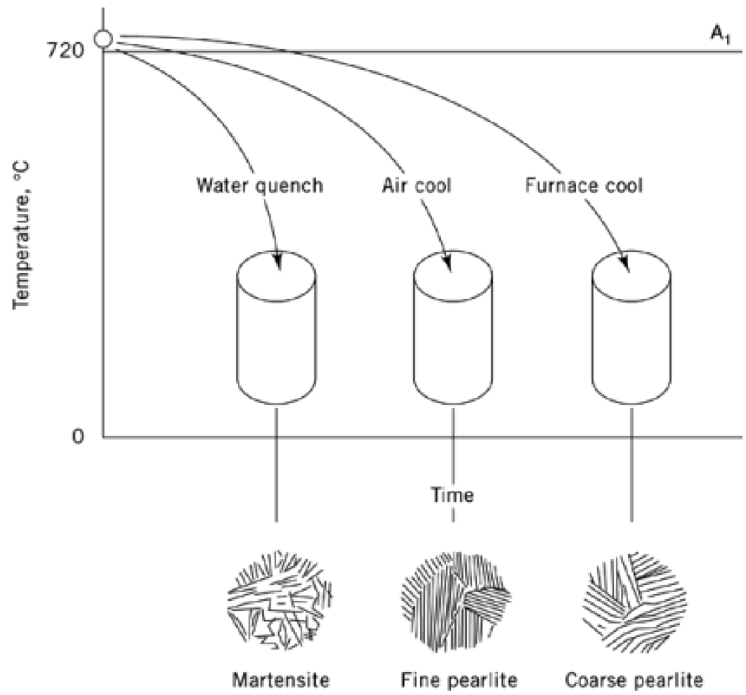

Fig. 19. Effect of cooling rate on structure of a eutectoid steel. A_1=austenitizing temperature; $>A_1$=austenite phase.

phase for Incoloy 901 is a γ' (fcc Ni_3Ti), whereas the phase used for grain-size control is η (hexagonal Ni_3Ti). A conditioning heat treatment is applied which precipitates η in a needle-like registered pattern. This heat treatment is on the order of 8 h at 899°C. Working is carried out at about 954°C, which is below the η-solvus temperature. Final deformation occurs below the recrystallization temperature. A fine-grained structure is generated by subsequent recrystallization below the solvus. The needlelike η-phase has become spherical and restricts grain growth. Aging follows standard procedure to precipitate γ'. The resulting fine-grained structure has unique properties, such as excellent low cycle fatigue properties.

4. Machining

The term machinability is used to indicate the ease or difficulty with which a material can be machined to the size, shape, and desired surface finish. Machining parameters that affect machinability include feed, speed, depth of cut, cutting fluid, cutting-tool material, and cutting-tool geometry (4) (see Tool materials). The relative ease of machining materials also depends on the particular machining operation and corresponding material removal characteristics. Machinability index and machinability rating are used as qualitative measures of machinability under specified conditions. Machinability ratings have been based on one or more of the following criteria: tool life, cutting speed, and power consumption.

Tool life tests are used to evaluate the effects of changes in tool materials, cutting variables, processing history, workpiece composition, or workpiece microstructure. Tool life, T, and cutting speed, V_c, can be related by the following equation:

$$V_c T^n = C_t$$

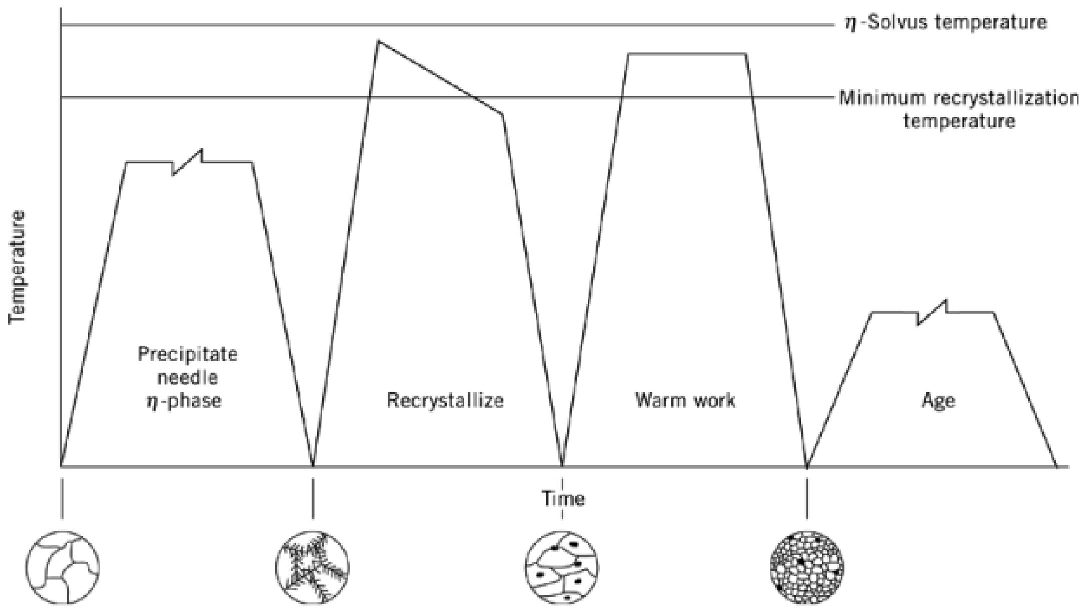

Fig. 20. Minigrain thermomechanical processing of Incoloy 901.

where n and C_t are empirical constants that reflect the cutting conditions under which the tests were made and the machinability of the material, respectively. It is more practical to measure the machinability, C_t, as the cutting speed necessary to cause tool failure within a specified period, than as tool life, T, at a specified speed. To determine C_t, tool life is measured for each of several cutting speeds using standardized cutting conditions and tool shape. These data yield values of n and C_t and the cutting speed that corresponds exactly to a specified tool life can be calculated.

Power consumption during metal cutting can be estimated from forces acting on a tool during the cutting. This can be measured on a dynamometer. In cutting operations the power consumption, P, at the spindle, in watts, is approximately equal to

$$P = V_c \cdot F_c$$

where V_c is the cutting speed in units of m/s and F_c is the force, in newtons, parallel to the cutting direction.

The unit power consumption, P_u, or power required to cut a given quantity of a particular material, can be determined by

$$P_u = P/Q$$

in cutting operations, where Q is the material removal rate, which can be expressed as

$$Q \approx d \cdot f \cdot V_c$$

where d is depth of cut and f is feed rate. Therefore, P_u can be expressed as follows:

$$P_u = F_c/(d \cdot f)$$

where P_u is in units of J/cm^3. The factors that affect the unit power consumption are principally the inherent machinability of the material and friction. The choice of cutting-tool shape, material, and application of coolant have comparatively little effect except for altering the amount of power expended in friction.

Metallurgical factors that affect machinability include chemical composition, previous processing treatments, and physical characteristics. The influence of chemical composition is evident in steels. Carbon and carbide-forming elements such as chromium, tungsten, molybdenum, and vanadium tend to decrease machinability by increasing hardness (qv). Elements which dissolve in ferrite, such as nickel, also usually reduce machinability because of the increased toughness and hardness imparted to the steel. Elements which form hard abrasive particles, such as alumina or silica, are also detrimental to machinability. Elements which form soft inclusions have a beneficial effect. Examples for steels are sulfur, lead, phosphorus, calcium, selenium, and tellurium. In resulfurized steels, sulfur along with manganese is added for the sole purpose of decreasing machining costs through higher machining speeds, improved tool life, or eliminating secondary operations. The manganese sulfide, MnS, inclusions that are formed improve the machining by producing a broken chip compared to a continuous chip. The improved chip formation is a result of the MnS particles that create an interface allowing the chip to separate from the workpiece. In addition, the sulfides prevent the chips from sticking to the tool and undermining the cutting edge.

Processing treatments affecting machinability include hot working, cold working, and heat treatment. The finishing temperatures during hot working influence grain size. For steels, larger grain sizes are preferred for most machining operations. Machining characteristics of most steels and many soft nonferrous metals can be greatly improved by cold working. The lower ductility, caused by cold working, promotes clean shearing and chip breakage. The machining characteristics of most metals are markedly affected by heat treatment. For example, high strength precipitation-strengthened alloys such as nickel-base superalloys are usually machined in the annealed condition prior to final aging.

Physical characteristics of metals have a significant impact on machinability. These include microstructural features such as grain size, mechanical properties such as tensile properties, and physical properties such as thermal conductivity.

5. Surface Treatments

In some metal-forming operations such as rod and wire drawing, various surface treatments are applied to the workpiece. These include descaling, cleaning the application of lubricant carriers, and the use of lubricants (see Lubrication and lubricants). Descaling can be mechanical or chemical (pickling). Lubricant carriers are applied by dipping the workpiece in hot solution or slurries such as lime or phosphate coatings. Details of surface treatments are available (5).

5.1. Shot Peening

A surface treatment that can be used to enhance the surface performance of parts is shot peening. Shot peening is a cold-working process in which the surface of a part is bombarded with small spherical media called shot. Each piece of shot striking the surface of the material imparts a small indentation. To create the indentation, the material surface must yield in tension. Material below the surface tends to restore the surface to its original shape, producing a hemisphere of cold-worked material highly stressed in compression. Overlapping indentations develop an even layer of metal in residual compressive stress. The maximum compressive residual stress produced at or under the surface by shot peening is at least as great as half the yield strength of the material being peened. In addition, the surface hardness of many materials increases as a result of cold working.

Cracks neither initiate nor propagate in a compressively stressed zone. Because nearly all fatigue and stress corrosion failures originate at the surface, compressive residual stresses induced by shot peening provide considerable increases in the resistance to fatigue failures, corrosion fatigue, stress-corrosion cracking, hydrogen-assisted cracking, fretting, galling, and erosion caused by cavitation. Cold working associated with shot peening also provides benefits such as work hardening, intergranular corrosion resistance, surface texturing, closing of porosity, and testing the bond of coatings. Shot peening is often used in the post-machining treatment of superalloy disks in jet engines.

6. Toxicity in Metal Treatments

Workers in the metals treatment industry are exposed to fumes, dusts, and mists containing metals and metal compounds, as well as to various chemicals from sources such as grinding wheels and lubricants. Exposure can be by inhalation, ingestion, or skin contact. Historically, metal toxicology was concerned with overt effects such as abdominal colic from lead toxicity. Because of the occupational health and safety standards of the 1990s such effects are rare. Subtle, chronic, or long-term effects of metals treatment exposure are under study. An index to safety precautions for various metal treatment processes is available (6). As additional information is gained, standards are adjusted.

Regulation of occupational toxicity is under the supervision of OSHA (7). Industrial employers must provide a written Hazard Communication Program (see Hazard analysis and risk assessment). Elements of this program must include labels and other forms of warning, material safety and data sheets (MSDS), training, a list of hazardous chemicals, a list of hazards of nonroutine tasks, and methods to inform on-site contractors of hazards. Standards are specified as threshold limit values (TLVs) and permissible exposure limits (PELs). The former refer to airborne concentration of substances and represent conditions to which nearly all workers may be repeatedly exposed day after day without known adverse effects. The latter are acceptable airborne concentrations of contaminants adopted by OSHA as standards under U.S. federal laws (29 CFR 1910.1000). Both TLVs and PELs can be specified as time-weighted average (TWA), short-term exposure limit (STEL), or ceiling (C) values. The TWA is the average concentration for a normal 8-h workday. A 15-min TWA exposure which should not be exceeded at any time during the workday is defined as STEL; C represents a concentration that should not be exceeded during any part of the working exposure. Standards vary significantly for the various metals. For example, the TLV–TWA for beryllium is 0.002 mg/m^3, the value for aluminum is 10 mg/m^3 (8).

7. Developments and Outlook

The principal challenges facing the metals industry as of this writing (ca 1994) are the continuing need for improvements in efficiency in order to meet global competition and environmental requirements; the cost and availability of raw materials; and development of alloys and processing methods to meet the demands of new technologies. The drive for continuous improvement in efficiency has led to development of such processes as strip rolling and strip casting, where sheet and strip are produced directly from the melt, and all hot-rolling steps are eliminated. Operations have also been eliminated by employing continuous casting. About 80% of U.S. crude steel (qv) production comes from continuous casting.

The metals industry relies heavily on raw materials imports. Many strategic elements, such as chromium, nickel, manganese, cobalt, niobium, and tungsten, are imported. For example, no significant chromium deposits exist in the United States. Chromium is used in stainless and specialty steels and superalloys. Efforts to alleviate raw material shortages include stockpiling, extensive recycling (qv), improved melting and processing methods, alloy development and substitution, and use of less expensive raw materials. The widespread use of duplex melting, which utilizes methods such as argon–oxygen decarburization (AOD), has helped greatly in these efforts.

Advanced materials and processes directed toward enhancing the performance and efficiency of aircraft turbines are examples of developments to meet technology demands. These include powder metallurgy techniques, directionally solidified eutectics and single crystals, and composites such as tungsten-reinforced superalloys (see Metallurgy, POWDER; Metal-matrix composites).

7.1. Powder Metallurgy of Superalloys

In addition to providing improved yields and efficiencies, powder metallurgy allows higher alloying additions in wrought products because of reduced segregation, and therefore superplastic structures are readily achieved. This is a significant factor in enabling superalloys which are heavily alloyed to provide superior mechanical properties and oxidation and corrosion resistance for rotating parts exposed to high operating temperatures in aircraft turbine engines.

Superalloy powders are produced by argon atomization, the rotating electrode process, and soluble gas-vacuum atomization (9) (see Coating processes, powder technology). Consolidation of the powder is accomplished by vacuum hot pressing, forging, extruding, or hot isostatic pressing. In vacuum hot pressing, loose powder is compacted in a cylindrical die with end plugs. Because of the long times involved and the small relative movement between particles, this process is seldom used. In forging, extruding, and hot isostatic pressing, the loose powder is put in a can, usually stainless steel, which is evacuated and sealed. Forging and extrusion are similar to conventional processing. In hot isostatic pressing (HIP) the can is placed in a resistance furnace located inside a water-cooled pressure vessel. Pressures are usually around 103 MPa (15,000 psi), but may be as high as 200 MPa (29,000 psi), and temperatures are up to 1260°C. The unique advantage of this process is its ability to produce near-net shapes or complex parts. Novel canning techniques to achieve these shapes are also available.

Figure 21 compares conventional processing for a superalloy disk to the various processing routes available by powder metallurgy. Figure 21**b** shows the preferred method for powder metallurgy turbine engine disks. This method ensures adequate material consolidation and distribution of micrometer-sized oxide particles. However, for some applications alternative routes are preferred because high quality parts can be produced using less material and fewer processing steps. The isothermal forging process (Fig. 21**b**) utilizes superplasticity, which refers to the ability of some alloys to exhibit extensive ductility. Elongations $>2000\%$ have been seen in tensile testing. The main prerequisite for superplasticity is an extremely fine and stable grain size, which is readily produced by powder metallurgy technicians. The dies are heated to the temperature of the workpiece. Dies are usually made of molybdenum in order to withstand elevated temperatures and, therefore, must be protected from oxidation by forging under a vacuum or inert gas atmosphere. Compared with conventional forging, strain rates are very low. Isothermal forging has an advantage over conventional forging in that it permits the formation of complex shapes in one operation.

A further development in superalloy powder metallurgy is oxide dispersion-strengthened (ODS) material, which contains a finely dispersed oxide that is stable at elevated temperatures. The dispersed oxide provides strength at elevated temperatures where precipitated phases such as γ' are dissolved. Thoria is the dispersed phase in TD nickel and TD nichrome, but emphasis is on dispersion of yttria in alloys such as MA 754 (a Ni–Cr alloy). Oxide dispersion-strengthened materials are produced by mechanical mixing of the oxide and appropriate metal powders in an attritor. Consolidation of these alloys is challenging because it involves thermomechanical processing to achieve large recrystallized grains that are elongated and of a specific crystallographic texture.

Another interesting development which utilizes powder-making techniques and further improves yields is spray forming (Fig. 22) (10). In this process a stream of liquid metal is extracted from a melting operation and passed through an atomizer. The resulting liquid droplets are then sprayed directly onto a collector, and

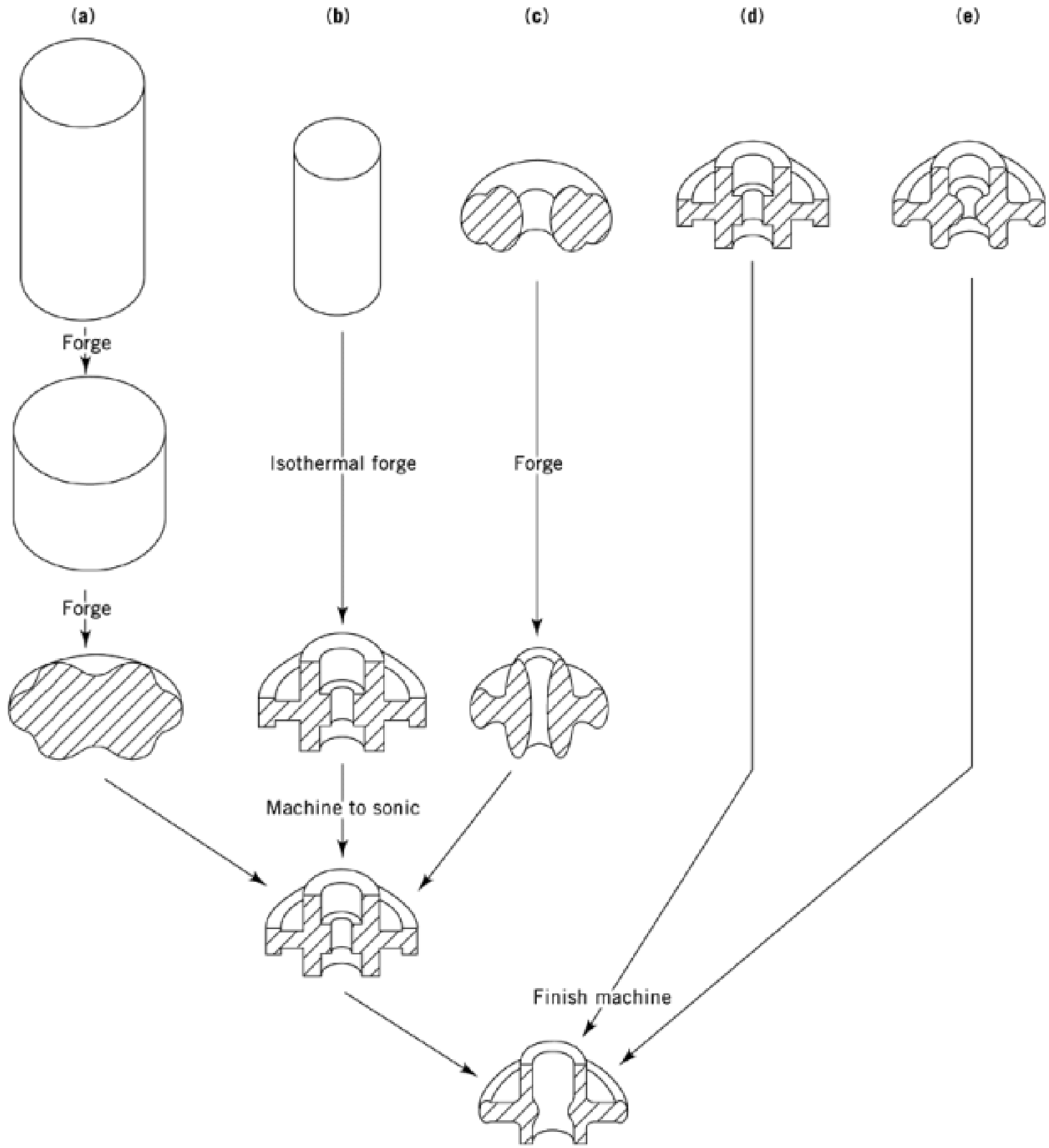

Fig. 21. Processing of superalloys: (**a**) conventional forging; (**b**) isothermal forging using a powder metallurgy billet; (**c**) forging preform using a powder metallurgy preform; (**d**) hot isotactic processing using a powder metallurgy sonic shape; and (**e**) hot isotactic processing of a near-net powder metallurgy shape.

flattened into a layer which builds up a preform. This technique was first explored using aluminum- and iron-based alloys and has some level of commercial production for stainless steel tubes, rolling mill rolls, and aluminum compressor rotors.

Among the challenges of applying this technology to aircraft-quality metal are the need for reducing porosity, removing ceramic inclusions, and lowering cost, respectively. Studies of atomizing utilizing argon or nitrogen gases have shown that nitrogen atomization lowers porosity to acceptable levels. However, the

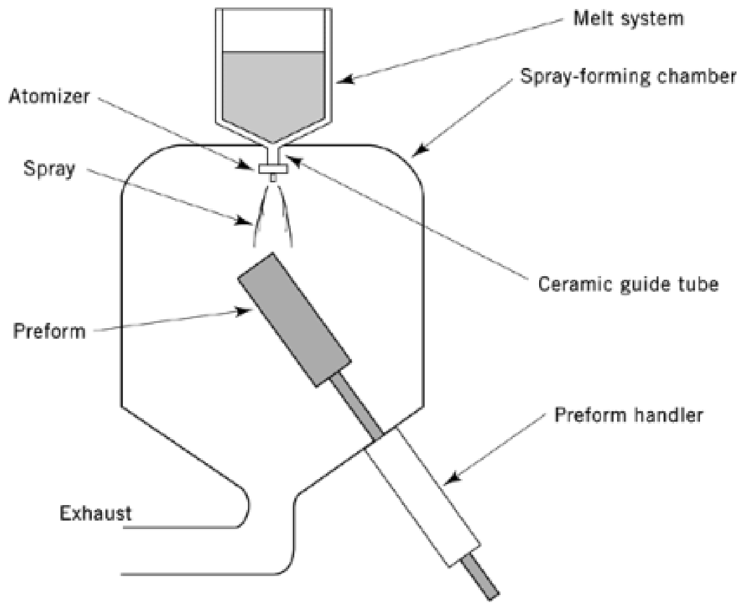

Fig. 22. Spray-forming system (10).

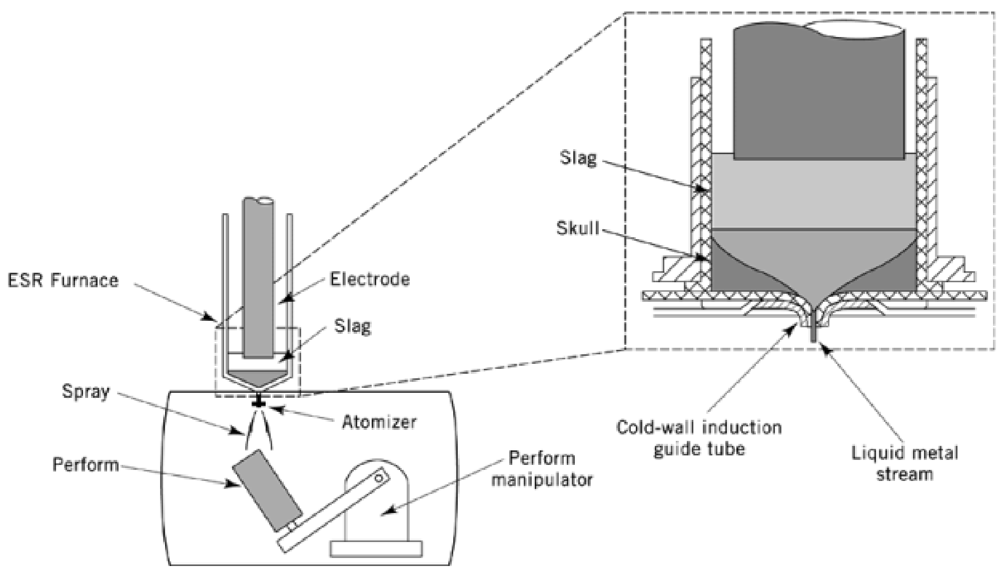

Fig. 23. ESR melting combined with spray forging (11).

resulting increases in the nitrogen content of the metal may have detrimental effects. To address ceramic inclusion removal and cost concerns, it has been proposed that the liquid stream be extracted from an electroslag remelting (ESR) furnace through a nonceramic nozzle, as shown in Figure 23 (11, 12).

BIBLIOGRAPHY

"Metal Treatments" in *ECT* 2nd ed., Vol. 13, pp. 315–331, by H. Herman, University of Pennsylvania, and J. G. Byrne, University of Utah; in *ECT* 3rd ed., Vol. 15, pp. 325–345, by L. A. Jackman, Special Metals Corp.

Cited Publications

1. D. K. Allen, *Metallurgy Theory and Practice*, American Technical Society, Chicago, Ill., 1969, p. 503.
2. W. Betteridge and J. Heslop, *The Nimonic Alloys and Other Nickel-Base High-Temperature Alloys*, Crane, Russak & Co., Inc., New York, 1974, p. 130.
3. E. E. Brown, R. C. Boettner, and D. L. Ruckle, *Superalloys—Processing, Proceedings of the Second International Conference*, TMS, Warrendale, Ohio, 1972, pp. L-1–L-12.
4. F. W. Boulger, "Machinability of Steels," *Metals Handbook*, 10th ed., Vol. **1**, 1990, p. 591.
5. K. Lang, ed., *Handbook of Metal Forming*, McGraw-Hill Book Co., Inc., New York, 1985, pp. 26.18–26.23.
6. Comprehensive Index for *Metals Handbook*, 10th ed., ASM International, Materials Park, Ohio, 1994.
7. *Occupational Safety and Health Standards for General Industry* (29 CFR Part 1910), Commerce Clearing House, Inc., Washington, D.C., 1991.
8. R. A. Goyer, "Toxicity of Metals", *Metals Handbook*, 10th ed., Vol. 2, ASM International, Materials Park, Ohio, 1990.
9. C. T. Sims and W. C. Hagel, *The Superalloys II*, John Wiley & Sons, Inc. New York, 1987.
10. M. G. Benz and co-workers, *Superalloys 718, 625, 706 and Various Derivatives*, TMS, Warrendale, Ohio, 1994, p. 99.
11. *Ibid.*, p. 106.
12. U.S. Pat. 5,160,532 (Nov. 3, 1992), M. G. Benz and T. F. Sawyer (to General Electric Corp.).

General References

13. J. Datski, *Material Properties and Manufacturing Processes*, John Wiley & Sons, Inc., New York, 1966.
14. L. E. Doyle and co-workers, *Manufacturing Processes and Materials for Engineers*, Prentice-Hall, Inc., Englewood Cliffs, N.J., 1985.
15. M. M. Eisenstadt, *Introduction to Mechanical Properties of Materials*, The Macmillan Co., New York, 1971.
16. A. G. Guy, *Introduction to Materials Science*, McGraw-Hill Book Co., Inc., New York, 1972.
17. R. W. Hanks, *Materials Engineering Science*, Harcourt, Brace & World, Inc., New York, 1970.
18. C. A. Keyser, *Materials Science in Engineering*, Charles E. Merrill Co., Columbus, Ohio, 1968.
19. F. A. McClintock and A. S. Argon, *Mechanical Behavior of Materials*, Addison-Wesley Publishing Co., Inc., Reading, Pa., 1966.
20. W. J. Patton, *Materials in Industry*, 3rd ed., Prentice-Hall, Inc., Englewood Cliffs, N.J., 1986.
21. R. E. Reed-Hill, *Physical Metallurgy Principles*, D. Van Nostrand Co., Inc., Princeton, N.J., 1964.
22. L. H. Van Vlack, *Materials Science for Engineers*, Addison-Wesley Publishing Co., Reading, Pa., 1970.
23. *Shot Peening Applications*, 7th ed., Metal Improvement Co., Inc., Paramus, N.J.

LAURENCE A. JACKMAN
Teledyne Allvac
CARLOS N. RUIZ
Teledyne Allvac

Related Articles

Metallic coatings; Welding; Powder metallurgy; Metal-matrix composites