Heat Treating of Aluminum Alloys

HEAT TREATING in its broadest sense, refers to any of the heating and cooling operations that are performed for the purpose of changing the mechanical properties, the metallurgical structure, or the residual stress state of a metal product. When the term is applied to aluminum alloys, however, its use frequently is restricted to the specific operations employed to increase strength and hardness of the precipitation-hardenable wrought and cast alloys. These usually are referred to as the "heat-treatable" alloys to distinguish them from those alloys in which no significant strengthening can be achieved by heating and cooling. The latter, generally referred to as "non-heat-treatable" alloys, depend primarily on cold work to increase strength. Heating to decrease strength and increase ductility (annealing) is used with alloys of both types; metallurgical reactions may vary with type of alloy and with degree of softening desired. Except for the low-temperature stabilization treatment sometimes given for 5xxx series alloys (which is a mill treatment and not discussed in this article), complete or partial annealing treatments are the only ones used for non-heat-treatable alloys. A general overview of these heat treatments is covered in the article "Principles of Heat Treating of Nonferrous Alloys" in this Volume.

Precipitation from Solid Solution

One essential attribute of a precipitation-hardening alloy system is a temperature-dependent equilibrium solid solubility characterized by increasing solubility with increasing temperature (see, for example, the phase diagrams in Fig 1 and 2). Although this condition is met by most of the binary aluminum alloy systems, many exhibit very little precipitation hardening, and these alloys ordinarily are not considered heat treatable. Alloys of the binary aluminum-silicon and aluminum-manganese systems, for example, exhibit relatively insignificant changes in mechanical properties as a result of heat treatments that produce considerable precipitation. The major aluminum alloy systems with precipitation hardening include:

- Aluminum-copper systems with strengthening from $CuAl_2$

- Aluminum-copper-magnesium systems (magnesium intensifies precipitation)
- Aluminum-magnesium-silicon systems with strengthening from Mg_2Si
- Aluminum-zinc-magnesium systems with strengthening from $MgZn_2$
- Aluminum-zinc-magnesium-copper systems

The general requirement for precipitation strengthening of supersaturated solid solutions involves the formation of finely dispersed precipitates during aging heat treatments (which may include either natural aging or artificial aging). The aging must be accomplished not only below the equilibrium solvus temperature, but below a metastable miscibility gap called the Guinier-Preston (GP) zone solvus line. The supersaturation of vacancies allows diffusion, and thus zone formation, to occur much faster than expected from equilibrium diffusion coefficients. In the precipitation process, the saturated solid solution first develops solute clusters, which then become involved in the formation of transitional (nonequilibrium) precipitates.

The mechanism of strengthening from precipitation involves the formation of coherent clusters of solute atoms (that is, the solute atoms have collected into a cluster but still have the same crystal structure as the solvent phase). This causes a great deal of strain because of mismatch in size between the solvent and solute atoms. Consequently, the presence of the precipitate particles, and even more importantly the strain fields in the matrix surrounding the coherent particles, provide higher strength by obstructing and retarding the movement of dislocations. The characteristic that determines whether a precipitate phase is coherent or noncoherent is the closeness of match or degree of disregistry between atomic spacings on the lattice of the matrix and on that of the precipitate. These changes in properties result from the formation of solute-rich microstructural domains, or GP zones.

The exact size, shape, and distribution of GP zones depend on the alloy in which they form and on the thermal and mechanical history of the specimen. Their shape can
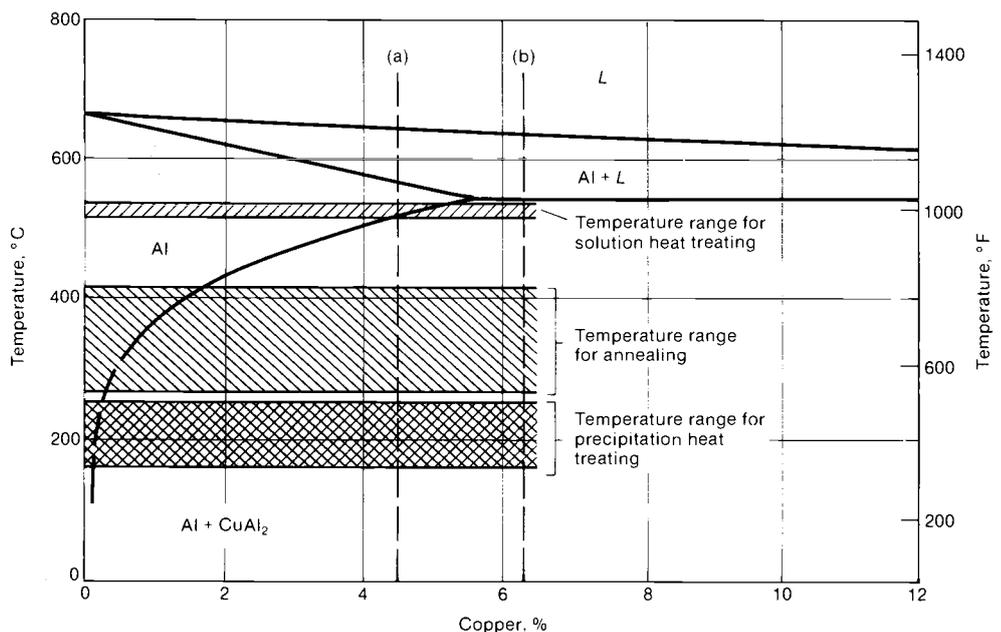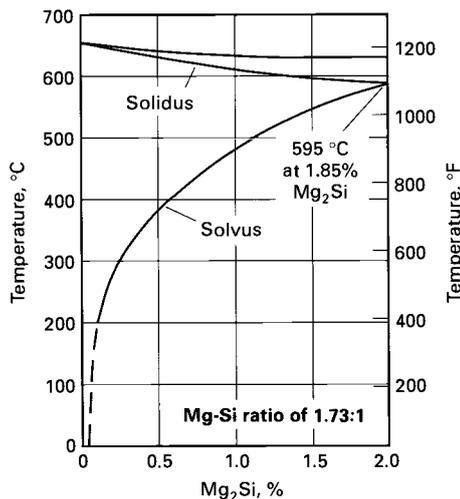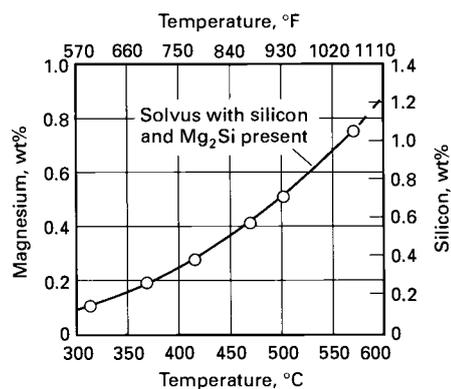

Fig 1 Portion of aluminum-copper binary phase diagram. Temperature ranges for annealing, precipitation heat treating, and solution heat treating are indicated. The range for solution treating is below the eutectic melting point of 548 °C (1018 °F) at 5.65 wt% Cu.

(a)

(b)

Fig 2 Equilibrium solubility as function of temperature for (a) Mg_2Si in aluminum with an Mg-Si ratio of 1.73-to-1 and (b) magnesium and silicon in solid aluminum when both Mg_2Si and silicon are present

sometimes be deduced by refined studies of diffuse x-ray scattering. Under favorable conditions, GP zones can be seen in transmission electron micrographs. Spherical solute-rich zones usually form when the sizes of the solvent and solute atoms are nearly equal, as in the aluminum-silver and aluminum-zinc systems. If there is a large difference in atom sizes, as in the aluminum-copper system, the GP zones usually form as disks whose planes lie parallel with some low-index plane of the matrix lattice. Sometimes, the solute atoms occupy preferred lattice sites within the GP zone, and thus form a small region of lattice order.

The GP zones are of the size range of tens of angstroms in diameter. They are essentially distorted regions of the matrix lattice, rather than discrete particles of a new phase having a different lattice. As such, they are completely coherent with the matrix, imposing local but often large strains on it. These mechanical strains, as well as the presence of a locally solute-rich, sometimes ordered lattice, can account for large changes in mechanical properties of the alloy before any long-range microstructural changes occur.

The GP zones are characteristically metastable and thus dissolve in the presence of a more stable precipitate. This dissolution causes a precipitate-free, visibly denuded region to form around the stable precipitate particles. The final structure consists of equilibrium precipitates, which do not contribute as significantly to hardening. More detailed information about preprecipitation phenomena can be found in the article "Structures Resulting From Precipitation From Solid Solution" in Volume 9 of the 9th Edition of *Metals Handbook*.

Precipitation in Aluminum-Copper Alloys. Figure 1, which illustrates the required solubility-temperature relationship needed in precipitation strengthening, shows the temperature ranges required for solution treatment and subsequent precipitate hardening in the aluminum-copper system. The equilibrium solid solubility of copper in aluminum increases as temperature increases—from about 0.20% at 250 °C (480 °F) to a maximum of 5.65% at the eutectic melting temperature of 548 °C (1018 °F). (It is considerably lower than 0.20% at temperatures below 250 °C.) For aluminum-copper alloys containing from 0.2 to 5.6% Cu, two distinct equilibrium solid states are possible. At temperatures above the lower curve in Fig 1 (solvus), the copper is completely soluble, and when the alloy is held at such temperatures for sufficient time to permit needed diffusion, the copper will be taken completely into solid solution. At temperatures below the solvus, the equilibrium state consists of two solid phases: solid solution, α, plus an intermetallic-compound phase θ (Al_2Cu). When such an alloy is converted to all solid solution by holding above the solvus temperature and then the temperature is decreased to below the solvus, the solid solution becomes supersaturated and the alloy seeks the equilibrium two-phase condition; the second phase tends to form by solid-state precipitation.

The preceding description is a gross oversimplification of the actual changes that occur under different conditions even in simple binary aluminum-copper alloys. A variety of different nonequilibrium precipitate structures is formed at temperatures below solvus. In alloys of the aluminum-copper system, a succession of precipitates is developed from a rapidly cooled supersaturated solid solution (SSS). These precipitates develop sequentially either with increasing temperature or with increasing time at temperature between room temperature and the solvus. The several stages are identified by the following notation:
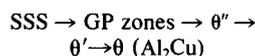
$$SSS \rightarrow GP\ zones \rightarrow \theta'' \rightarrow$$
$$\theta' \rightarrow \theta\ (Al_2Cu)$$

At temperatures in the natural aging range (about −20 to 60 °C, or 0 to 140 °F), the distribution of copper atoms changes with time from random to the disklike planar aggregates (GP zones), which form on particular crystallographic planes of the aluminum matrix. These aggregates create coherency strain fields that increase resistance to deformation, and their formation is responsible for the changes in mechanical properties that occur during natural aging. At higher temperatures, transition forms of approximate composition Al_2Cu develop and further increase strength. In the highest strength condition, both the θ'' and θ' transition precipitates may be present. When time and temperature are increased sufficiently to form high proportions of the equilibrium θ, the alloy softens and is said to be "overaged."

The commercial heat-treatable aluminum alloys are, with few exceptions, based on ternary or quaternary systems with respect to the solutes involved in developing strength by precipitation. Commercial alloys whose strength and hardness can be significantly increased by heat treatment include 2xxx, 6xxx, and 7xxx series wrought alloys (except 7072) and 2xx.0, 3xx.0, and 7xx.0 series casting alloys. Some of these contain only copper, or copper and silicon, as the primary strengthening alloy addition(s). Most of the heat-treatable alloys, however, contain combinations of magnesium with one or more of the elements copper, silicon, and zinc. Characteristically, even small amounts of magnesium in concert with these elements accelerate and accentuate precipitation hardening, while alloys in the 6xxx series contain silicon and magnesium approximately in the proportions required for formulation of magnesium silicide (Mg_2Si). Although not as strong as most 2xxx and 7xxx alloys, 6xxx series alloys have good formability, weldability, machinability, and corrosion resistance, with medium strength.

In the heat-treatable wrought alloys, with some notable exceptions (2024, 2219, and 7178), such solute elements are present in amounts that are within the limits of mutual solid solubility at temperatures below the eutectic temperature (lowest melting temperature). In contrast, some of the casting alloys of the 2xx.0 series and all of the 3xx.0 series alloys contain amounts of soluble elements that far exceed solid-solubility limits. In these alloys, the phase formed by combination of the excess soluble elements with the aluminum will never be dissolved, although the shapes of the undissolved particles may be changed by partial solution.

Most of the heat-treatable aluminum alloy systems exhibit multistage precipitation and undergo accompanying strength changes analogous to those of the aluminum-copper system. Multiple alloying additions of both major solute elements and supplementary elements employed in commercial alloys are strictly functional and serve with different heat treatments to provide the many different combinations of properties—physical,

mechanical, and electrochemical—that are required for different applications. Some alloys, particularly those for foundry production of castings, contain amounts of silicon far in excess of the amount that is soluble or needed for strengthening alone. The function here is chiefly to improve casting soundness and freedom from cracking, but the excess silicon also serves to increase wear resistance, as do other microstructural constituents formed by manganese, nickel, and iron. Parts made of such alloys are commonly used in gasoline and diesel engines (pistons, cylinder blocks, and so forth).

Alloys containing the elements silver, lithium, and germanium are also capable of providing high strength with heat treatment, and in the case of lithium, both increased elastic modulus and lower density, which are highly advantageous—particularly for aerospace applications (see the following section "Aluminum-Lithium Alloys" in this article). Commercial use of alloys containing these elements has been restricted either by cost or by difficulties encountered in producing them. Such alloys are used to some extent, however, and research is being directed toward overcoming their disadvantages.

In the case of alloys having copper as the principal alloying ingredient and no magnesium, strengthening by precipitation can be greatly increased by adding small fractional percentages of tin, cadmium, or indium, or combinations of these elements. Alloys based on these effects have been produced commercially but not in large volumes because of costly special practices and limitations required in processing, and in the case of cadmium, the need for special facilities to avoid health hazards from formation and release of cadmium vapor during alloying. Such alloys, as well as those containing silver, lithium, or other particle-forming elements, may be used on a selective basis in the future.

Aluminum-Lithium Alloys. Like other age-hardened aluminum alloys, aluminum-lithium alloys achieve precipitation strengthening by thermal aging after a solution heat treatment. The precipitate structure is sensitive to a number of processing variables, including, but not limited to, the quenching rate following the solution heat treatment, the degree of cold deformation prior to aging, and the aging time and temperature. Minor alloying elements can also have a significant effect on the aging process by changing the interface energy of the precipitate, by increasing the vacancy concentration, and/or by raising the critical temperature for homogeneous precipitation. Like some other age-hardened 2xxx aluminum alloys, aluminum-lithium-base alloys also gain increased strength and toughness from deformation prior to aging. This unusual phenomenon has given rise to a number of thermomechanical processing steps for alu-

minum-lithium alloys aimed at optimizing mechanical properties after artificial aging.

The age hardening of aluminum-lithium alloys involves the continuous precipitation of δ' (Al_3Li) from a supersaturated solid solution. The aluminum and lithium in the δ' precipitates are positioned at specific locations. The eight shared corner sites are occupied by lithium, and the six shared faces are occupied by aluminum. This gives rise to the aluminum-lithium composition of δ' precipitates. The geometrical similarity between the lattice of the precipitates and the face-centered cubic lattices of the solid solution facilitates the observed cube/cube orientation. The lattice parameters of the precipitate are also closely matched to those of the matrix. Consequently, the microstructure of an aluminum-lithium alloy solution heat treated and aged for short times below the δ' solvus is characterized by a homogeneous distribution of coherent, spherical δ' precipitates.

Aluminum-lithium-base alloys are microstructurally unique. They differ from most of the aluminum alloys in that once the major strengthening precipitate (δ') is homogeneously precipitated, it remains coherent even after extensive aging. In addition, extensive aging at high temperatures (>190 °C, or 375 °F) can result in the precipitation of icosahedral grain-boundary precipitates with five-fold symmetry. Although the quasi-crystalline structure and the composition of these grain-boundary precipitates are not yet exactly known, it has been suggested that both the precipitates and the precipitate-free zones (PFZs) near the grain boundaries might play a major role in the fracture process.

The low ductility and toughness of binary aluminum-lithium alloys can be traced, at least in part, to the inhomogeneous nature of their slip, resulting from coherent-particle hardening of spherical δ' precipitates. The presence of equilibrium δ (aluminum-lithium) precipitates at grain boundaries can also cause PFZs, which can induce further strain localization and promote intergranular failure. Consequently, for the development of commercial alloys, slip has been homogenized by introducing dispersoids (manganese, zirconium) and semicoherent/incoherent precipitates, such as T_1 (Al_2CuLi), θ' (Al_2Cu), or S (Al_2LiMg), through copper or magnesium additions. Magnesium and copper improve the strength of aluminum-lithium alloys through solid-solution and precipitate strengthening, and they can minimize the formation of PFZs near grain boundaries. Zirconium, which forms the cubic Al_3Zr coherent dispersoid, stabilizes the subgrain structure and suppresses recrystallization.

Development of commercially available aluminum-lithium-base alloys was started by adding lithium to aluminum-copper, aluminum-magnesium, and aluminum-copper-

magnesium alloys. These alloys were chosen to superimpose the precipitation-hardening characteristics of aluminum-copper-, aluminum-copper-magnesium-, and aluminum-magnesium-base precipitates to the hardening of lithium-containing precipitates. Proceeding in this manner, alloys 2020 (Al-Cu-Li-Cd), 01429 (Al-Mg-Li), 2090 (Al-Cu-Li), and 2091 and 8090 (Al-Cu-Mg-Li) evolved. Besides these registered alloys, other commercial aluminum-lithium alloys include Weldalite 049 and CP276. Properties and applications of these alloys are discussed in the article "Aluminum-Lithium Alloys" in Volume 2 of the 10th Edition of *Metals Handbook*.

In terms of δ' precipitation, the only effect of magnesium appears to be a reduction in the solubility of lithium. The microstructure of an aluminum-magnesium-lithium alloy in the early stages of aging is similar to that of an aluminum-lithium alloy. Precipitation in the aluminum-copper-lithium system is more complicated than that in either the aluminum-lithium or aluminum-magnesium-lithium systems.

Effects on Physical and Electrochemical Properties. The above description of the precipitation processes in commercial heat-treatable aluminum alloys (as well as the heat-treatable binary alloys, none of which is used commercially in wrought form) affect not only mechanical properties but also physical properties (density and electrical and thermal conductivities) and electrochemical properties (solution potential). On the microstructural and submicroscopic scales, the electrochemical properties develop point-to-point nonuniformities that account for changes in corrosion resistance.

Measurements of changes in physical and electrochemical properties have played an important role in completely describing precipitation reactions and are very useful in analyzing or diagnosing whether heat-treatable products have been properly or improperly heat treated. Although they may be indicative of the strength levels of products, they cannot be relied upon to determine whether or not the product meets specified mechanical-property limits. Since elements in solid solution are always more harmful to electrical conductivity than the same elements combined with others as intermetallic compounds, thermal treatments are applied to ingots used for fabrication of electrical conductor parts. These thermal treatments are intended to precipitate as much as possible of the dissolved impurities. Iron is the principal element involved, and although the amount precipitated is only a few hundredths of a percent, the effect on electrical conductivity of the wire, cable, or other product made from the ingot is of considerable practical importance. These alloys may or may not be heat treatable with respect to mechanical properties. Electrical conductor alloys 6101 and

6201 are heat treatable. These alloys are used in tempers in which their strengthening precipitate, the transition form of Mg_2Si, is largely out of solid solution to optimize both strength and conductivity.

Strengthening by Heat Treatment

Heat treatment to increase strength of aluminum alloys is a three-step process:

- Solution heat treatment: dissolution of soluble phases
- Quenching: development of supersaturation
- Age hardening: precipitation of solute atoms either at room temperature (natural aging) or elevated temperature (artificial aging or precipitation heat treatment)

Each of these steps and the use of quench-factor analysis are described in the following four sections. Typical solution and precipitation heat treatments for mill products are given in Tables 1(a, b, and c) and 2, and treatments for castings are given in Table 3. Temper designations are defined at the end of this article.

Solution Heat Treating

To take advantage of the precipitation-hardening reaction, it is necessary first to produce a solid solution. The process by which this is accomplished is called solution heat treating, and its objective is to take into solid solution the maximum practical amounts of the soluble hardening elements in the alloy. The process consists of soaking the alloy at a temperature sufficiently high and for a time long enough to achieve a nearly homogeneous solid solution.

Nominal commercial solution heat-treating temperature is determined by the composition limits of the alloy and an allowance for unintentional temperature variations. Although ranges normally listed allow variations of ±6 °C (±10 °F) from the nominal, some highly alloyed, controlled-toughness, high-strength alloys require that temperature be controlled within more restrictive limits. Broader ranges may be allowable for alloys with greater intervals of temperature between their solvus and eutectic melting temperatures.

Overheating. Care must be exercised to avoid exceeding the initial eutectic melting temperature. If appreciable eutectic melting occurs as a result of overheating, properties such as tensile strength, ductility, and fracture toughness may be degraded. Materials that exhibit microstructural evidence of overheating are generally categorized as unacceptable by specification. Evidence of grain-boundary melting that occurs above the eutectic melting temperature of the alloy usually is not detectable by either visual examination or nondestructive testing.

Although maximum temperature must be restricted to avoid melting, the lower limit should, when possible, be above the temperature at which complete solution occurs (solvus). In the alloy represented by line (a) in Fig 1, these temperatures would be about 575 and 515 °C (1065 and 960 °F), respectively. However, under production conditions, the temperature interval for solution treatment (shown in Fig 1 for typical 2xxx or 2xx.x) alloys provides a margin to safeguard against eutectic melting and a cushion on the low side for increased solution and diffusion rates.

For alloys containing more than 5.65% Cu, complete solution can never occur. For these alloys, such as alloy 2219 (which has 5.8 to 6.8% Cu), the minimum solution heat-treating temperature is established so that it is as close as practical to the eutectic temperature while providing a margin of safety commensurate with the capability of the equipment. Line (b) in Fig 1 is another example of a composition above 5.65% Cu that does not allow complete dissolution of aluminum-copper precipitates.

For more complex ternary and quaternary systems, solution treatments are modified according to the effect of new elements on the solid solubility and/or the eutectic melting points of the basic binary system. In aluminum-lithium alloys, for example, magnesium reduces the solubility of lithium in aluminum. In the aluminum-copper system, magnesium also lowers the eutectic melting point. The proximity of typical solution-treating temperature ranges to eutectic melting temperatures for three common aluminum-copper-magnesium alloys is shown in the following table:

Alloy	Solution-treating temperature		Eutectic melting temperature	
	°C	°F	°C	°F
2014	496–507	925–945	510	950
2017	496–507	925–945	513	955
2024	488–499	910–930	502	935

Similar considerations apply to other age-hardenable alloy systems such as aluminum-magnesium-silicon alloys. For example, according to Fig 2(a), a 1.08% Mg_2Si alloy would be soaked at a temperature in excess of 500 °C (930 °F) but below the solidus of 595 °C (1100 °F) to avoid incipient melting. However, because some alloy constituents may form complex eutectics that melt at temperatures below the equilibrium eutectic temperature, the upper limit for solution treatment of aluminum-magnesium-silicon alloys is in the range of 515 to 540 °C (960 to 1000 °F). At 540 °C (1000 °F), about 0.6% Mg can be placed in solution (Fig 2b).

Nonequilibrium Melting. When high heating rates are employed, the phenomenon of nonequilibrium melting must be considered. This phenomenon can also be explained with the help of the aluminum-copper phase diagram (Fig 1). The room-temperature microstructure of an F-temper product containing 4% Cu consists of a solid solution of copper in aluminum and particles of Al_2Cu. When this product is heated slowly, the Al_2Cu begins to dissolve, and if heating is slow enough, all of the Al_2Cu is dissolved when temperatures above the solvus (500 °C, or 932 °F) are reached. When the heating rate is high, however, much of the Al_2Cu remains undissolved. If a material with this microstructure is heated at or above the eutectic temperature of 548 °C (1018 °F), melting will begin at the interface between the Al_2Cu and the matrix. With sufficient time above the eutectic temperature, this metastable liquid will dissolve to form a solid solution and will leave no trace provided that hydrogen gas has not condensed at the interface to form a void. If the product is quenched before the liquid has time to equilibrate, however, it will solidify and form fine eutectic rosettes. This nonequilibrium melting should not be confused with true equilibrium melting, which would occur in any alloy containing more than 5.65% Cu. In such an alloy, eutectic melting is equilibrium melting. No matter how long such an alloy is held above the eutectic temperature, the liquid will never solidify. In commercial alloys, which usually are ternaries or quaternaries of the major alloying elements, the situation is more complex. Different phases have different solvus temperatures, and nonequilibrium melting may occur at different temperatures depending on composition, size of precipitates, and rate of heating. When new solution heat-treating equipment (which provides higher heating rates) is employed, careful examination of alloy microstructures should be included as part of the certification process.

Underheating. When the temperatures attained by the parts or pieces being heat treated are appreciably below the normal range, solution is incomplete, and strength somewhat lower than normal is expected. In the aluminum-copper system (Fig 1), the shallow slope of the solvus at its intersection with the composition line indicates that a slight decrease in temperature will result in a large reduction in the concentration of the solid solution and a correspondingly significant decrease in final strength. The effect of solution-treating temperature on the strength of two aluminum alloys is illustrated by the following data:

Solution-treating temperature		Tensile strength		Yield strength	
°C	°F	MPa	ksi	MPa	ksi
6061-T6 sheet 1.6 mm (0.064 in.) thick					
493	920	301	43.7	272	39.4
504	940	316	45.8	288	41.7
516	960	333	48.3	305	44.3
527	980	348	50.5	315	45.7
2024-T4 sheet 0.8 mm (0.032 in.) thick					
488	910	419	60.8	255	37.0
491	915	422	61.2	259	37.5
493	920	433	62.8	269	39.0
496	925	441	63.9	271	39.3

Table 1(a) Typical solution and precipitation heat treatments for commercial heat-treatable aluminum alloy mill products with copper alloying

Alloy	Product form	Solution heat treatment(a)			Precipitation heat treatment			
		Metal temperature(b)		Temper designation	Metal temperature(b)		Time(c), h	Temper designation
		°C	°F		°C	°F		
Al-Cu alloys without magnesium alloying								
2011	Rolled or cold finished rod and bar	525	975	T3(d)	160	320	14	T8(d)
				T4	...	...	...	...
				T451(e)	...	...	...	...
2025	Die forgings	515	960	T4	170	340	10	T6
2219(f)	Flat sheet	535	995	T31(d)	175	350	18	T81(d)
				T37(d)	165	325	24	T87(d)
				T42	190	375	36	T62
	Plate	535	995	T31(d)	175	350	18	T81(d)
				T37(d)	175	350	18	T87(d)
				T351(e)	175	350	18	T851(e)
				T42	190	375	36	T62
	Rolled or cold finished wire, rod, and bar	535	995	T351(e)	190	375	18	T851(e)
2219(f)	Extruded rod, bar, shapes, and tube	535	995	T31(d)	190	375	18	T81(d)
				T3510(e)	190	375	18	T8510(e)
				T3511(e)	190	375	18	T8511(e)
				T42	190	375	36	T62
	Die forgings and rolled rings	535	995	T4	190	375	26	T6
	Hand forgings	535	995	T4	190	375	26	T6
				T352(f)	175	350	18	T852f)
Al-Cu-Mg alloys								
2018	Die forgings	510(g)	950(g)	T4	170	340	10	T61
2024(h)	Flat sheet	495	920	T3(d)	190	375	12	T81(d)
				T361(d)	190	375	8	T861(d)
				T42	190	375	9	T62
					190	375	16	T72
2024(h)	Coiled sheet	495	920	T4	...	...	...	...
				T42	190	375	9	T62
					190	375	16	T72
	Plate	495	920	T351(e)	190	375	12	T851(e)
				T361(d)	190	375	8	T861(d)
				T42	190	375	9	T62
	Rolled or cold finished wire, rod, and bar	495	920	T4	190	375	12	T6
				T351(e)	190	375	12	T851(e)
				T36(d)	190	375	8	T86(d)
				T42	190	375	16	T62
	Extruded rod, bar, shapes, and tube	495	920	T3	190	375	12	T81
				T3510(e)	190	375	12	T8510(e)
				T3511(e)	190	375	12	T8511(e)
				T42	190	375	16	T62
	Drawn tube	495	920	T3(d)	...	...	...	...
				T42	...	...	...	...
2036	Sheet	500	930	T4	...	...	...	...
2038	Sheet	540	1000	T4	205	400	2	T6
2218	Die forgings	510(g)	950(g)	T4	170	340	10	T61
		510(i)	950(i)	T41	240	460	6	T72
Al-Cu-Mg-Si alloys								
2008	Sheet	510	950	T4(d)(j)	205	400	1	T62(e)
2014(h)	Flat sheet	500	935	T3(d)	160	320	18	T62
				T42	160	320	18	T6
	Coiled sheet	500	935	T4	160	320	18	T6
				T42	160	320	18	T62
	Plate	500	935	T42	160	320	18	T62
				T451(e)	160	320	18	T651(e)
	Rolled or cold finished wire, rod, and bar	500	935	T4	160(k)	320(k)	18	T6
				T42	160(k)	320(k)	18	T62
				T451(e)	160(k)	320(k)	18	T651(e)
	Extruded rod, bar, shapes, and tube	500	935	T4	160(k)	320(k)	18	T6
				T42	160(k)	320(k)	18	T62
				T4510(e)	160(k)	320(k)	18	T6510(e)
	Drawn tube	500	935	T4	160(k)	320(k)	18	T6
				T42	160(k)	320(k)	18	T62
	Die forgings	500(l)	935(l)	T4	170	340	10	T6
2017	Rolled or cold finished wire, rod, and bar	500	935	T4	...	...	...	...
				T42	...	...	...	...
2117	Rolled or cold finished wire and rod	500	935	T4	...	...	...	...
				T42	...	...	...	...
2618	Forgings and rolled rings	530	985	T4	200	390	20	T61
4032	Die forgings	510(h)	950(h)	T4	170	340	10	T6
Al-Cu-Li alloys								
2090	Sheet	540	1000	T3(d)	165	325	24	T83(d)
2091	Sheet	530	990	T3(d)	120	250	24	T84(d)
	Extruded bar	530	990	T3(d)	190	375	12	Peak aged(d)
8090	Extruded bar	530	990	T3(d)	190	375	12	Peak aged(d)
CP276	Extruded bar	540	1000	T3(d)	190	375	12–15	Peak aged(d)

(a) Material should be quenched from the solution-treating temperature as rapidly as possible and with minimum delay after removal from the furnace. When material is quenched by total immersion in water, unless otherwise indicated, the water should be at room temperature, and should be suitably cooled so that it remains below 38 °C (100 °F) during the quenching cycle. Use of high-velocity, high-volume jets of cold water also is effective for some materials. (b) The nominal temperatures listed should be attained as rapidly as possible and maintained within ±6 °C (±10 °F) of nominal during the time at temperature. (c) Approximate time at temperature. The specific time will depend on the time required for the load to reach temperature. The times shown are based on rapid heating, with soak time measured from the time the load reaches a temperature within 6 °C (10 °F) of the applicable temperature. (d) Cold working subsequent to solution heat treatment and prior to any precipitation heat treatment is necessary to attain the specified properties for this temper. (e) Stress relieved by stretching to produce a specified amount of permanent set subsequent to solution heat treatment and prior to any precipitation heat treatment. (f) Stress relieved by 1 to 5% cold reduction after solution treatment and prior to precipitation heat treatment. (g) Quenched in water at 100 °C (212 °F). (h) These heat treatments also apply to alclad sheet and plate of these alloys. (i) Quenched with room-temperature air blast. (j) See U.S. Patent 4,840,852. (k) An alternative heat treatment of 8 h at 177 °C (350 °F) may also be used. (l) Quenched in water at 60 to 80 °C (140 to 180 °F).

Table 1(b) Typical solution and precipitation heat treatments for Mg-Si aluminum alloys (6xxx series alloys)

Alloy	Product form	Solution heat treatment(a) Metal temperature(b) °C	°F	Temper designation	Precipitation heat treatment Metal temperature(b) °C	°F	Time(c), h	Temper designation
6005	Extruded rod, bar, shapes, and tube	530(d)	985(d)	T1	175	350	8	T5
6009(e)	Sheet	555	1030	T4	205	400	1	T6(e)
6010	Sheet	565	1050	T4	205	400	1	T6(e)
6053	Die forgings	520	970	T4	170	340	10	T6
6061(f)	Sheet	530	985	T4	160	320	18	T6
				T42	160	320	18	T62
	Plate	530	985	T4(g)	160	320	18	T6(g)
				T42	160	320	18	T62
				T451(h)	160	320	18	T651(h)
	Rolled or cold finished wire, rod, and bar	530	985	T4	160(i)	320(i)	18	T6
					160(i)	320(i)	18	T89(j)
					160(i)	320(i)	18	T93(k)
					160(i)	320(i)	18	T913(k)
					160(i)	320(i)	18	T94(k)
				T42	160(i)	320(i)	18	T62
				T451(h)	160(i)	320(i)	18	T651(h)
	Extruded rod, bar, shapes, and tube	530(d)	985(d)	T4	175	350	8	T6
				T4510(h)	175	350	8	T6510(h)
				T4511(h)	175	350	8	T6511(h)
		530	985	T42	175	350	8	T62
6061(f)	Drawn tube	530	985	T4	160(i)	320(i)	18	T6
				T42	160(i)	320(i)	18	T62
	Die and hand forgings	530	985	T4	175	350	8	T6
	Rolled rings	530	985	T4	175	350	8	T6
				T452(l)	175	350	8	T652(l)
6063	Extruded rod, bar, shapes, and tube	(d)	(d)	T1	205(m)	400(m)	1	T5
		520(d)	970(d)	T4	175(n)	350(n)	8	T6
		520	970	T42	175(n)	350(n)	8	T62
	Drawn tube	520	970	T4	175	350	8	T6
					175	350	8	T83(j)(d)
					175	350	8	T831(j)(d)
					175	350	8	T832(j)(d)
				T42	175	350	8	T62
6013(o)	Sheet	570	1055	W(p)	190	375	4	T6
	Plate	570	1055	W(p)	190	375	4	T651
6066	Extruded rod, bar, shapes, and tube	530	990	T4	175	350	8	T6
				T42	175	350	8	T62
				T4510(h)	175	350	8	T6510(h)
				T4511(h)	175	350	8	T6511(h)
	Drawn tube	530	990	T4	175	350	8	T6
				T42	175	350	8	T62
	Die forgings	530	990	T4	175	350	8	T6
6070	Extruded rod, bar, shapes, and tube	545(d)	1015(d)	T4	160	320	18	T6
				T42	160	320	18	T62
6111	Sheet	560	1040	T4	175	350	8	T6(q)
6151	Die forgings	515	960	T4	170	340	10	T6
	Rolled rings	515	960	T4	170	340	10	T6
				T452(l)	170	340	10	T652(l)
6262	Rolled or cold finished wire, rod, and bar	540	1000	T4	170	340	8	T6
					170	340	12	T9(k)
				T451	170	340	8	T651(h)
				T42	170	340	8	T62
6262	Extruded rod, bar, shapes, and tube	540(d)	1000(d)	T4	175	350	12	T6
				T4510(h)	175	350	12	T6510(h)
		540	1000	T42	175	350	12	T62
	Drawn tube	540	1000	T4	170	340	8	T6
					170	340	8	T9(k)
				T42	170	340	8	T62
6463	Extruded rod, bar, shapes, and tube	(d)	(d)	T1	205(m)	400(m)	1	T5
		520(d)	970(d)	T4	175(n)	350(n)	8	T6
		520	970	T42	175(n)	350(n)	8	T62
6951	Sheet	530	985	T4	160	320	18	T6
				T42	160	320	18	T62

(a) Material should be quenched from the solution-treating temperature as rapidly as possible and with minimum delay after removal from the furnace. When material is quenched by total immersion in water, unless otherwise indicated, the water should be at room temperature, and should be suitably cooled so that it remains below 38 °C (100 °F) during the quenching cycle. Use of high-velocity, high-volume jets of cold water also is effective for some materials. (b) The nominal temperatures listed should be attained as rapidly as possible and maintained within ±6 °C (±10 °F) of nominal during the time at temperature. (c) Approximate time at temperature. The specific time will depend on the time required for the load to reach temperature. The times shown are based on rapid heating, with soak time measured from the time the load reaches a temperature within 6 °C (10 °F) of the applicable temperature. (d) By suitable control of extrusion temperature, product may be quenched directly from extrusion press to provide specified properties for this temper. Some products may be adequately quenched in room-temperature air blast. (e) Alternate heat treatments of 4 h at 190 °C (375 °F) or 8 h at 175 °C (350 °F) may also be used. See U.S. Patent 4,082,578. (f) These heat treatments also apply to alclad sheet and plate in these alloys. (g) Applicable to tread plate only. (h) Stress relieved by stretching to produce a specified amount of permanent set prior to precipitation heat treatment. (i) An alternative heat treatment of 8 h at 170 °C (340 °F) also may be used. (j) Cold working after solution treatment is necessary to attain specified properties during precipitation heat treatments. (k) Cold working after precipitation heat treatment is necessary to attain specified properties. (l) Stress relieved by 1 to 5% cold reduction subsequent to solution heat treatment and prior to precipitation heat treatment. (m) An alternative treatment of 3 h at 182 °C (360 °F) also may be used. (n) An alternative treatment of 6 h at 182 °C (360 °F) also may be used. (o) See U.S. Patent 4,589,932. (p) Two weeks of natural aging to a T4 condition. (q) Artificially aged in laboratory from T4 to T6.

Table 1(c) Typical solution and precipitation heat treatments for heat-treatable Zn-Mg aluminum alloys from the 7xxx series

Alloy	Product form	Solution heat treatment(a) Metal temperature(b) °C	°F	Temper designation	Precipitation heat treatment Metal temperature(b) °C	°F	Time(c), h	Temper designation
7001	Extruded rod, bar, shapes, and tube	465	870	W	120	250	24	T6
					120	250	24	T62
				W510(d)	120	250	24	T6510(d)
				W511(d)	120	250	24	T6511(d)
7005	Extruded rod, bar, and shapes	· · ·	· · ·	· · ·	· · ·	· · ·	· · ·	T53(e)
7050	Plate	475	890	W51(d)	(f)	(f)	(f)	T7651(g)
					(h)	(h)	(h)	T7451(g)
	Extrusions	475	890	W510(d)	(f)	(f)	(f)	T76510(g)
				W511(d)	(f)	(f)	(f)	T76511(g)
	Die and hand forgings	475	890	W	(h)	(h)	(h)	T74(g)
				W52(d)	(h)	(h)	(h)	T7452(g)
7075(i)	Sheet	480	900	W	120(j)	250(j)	24	T6
					120(j)	250(j)	24	T62
					(f)	(f)	(f)	T76(g)
					(h)(k)	(h)(k)	(h)(k)	T73(g)
	Plate	480	900	W	120(j)	250(j)	24	T62
				W51(d)	(h)(k)	(h)(k)	(h)(k)	T7351(d)(g)
					120(j)	250(j)	24	T651(d)
					(f)	(f)	(f)	T7651(g)
7075(i)	Rolled or cold finished wire, rod, and bar	490	915	W	120	250	24	T6
					120	250	24	T62
					(h)(k)	(h)(k)	(h)(k)	T73(g)
				W51(d)	120	250	24	T651(d)
					(h)(k)	(h)(k)	(h)(k)	T7351(d)(g)
	Extruded rod, bar, shapes, and tube	465	870	W	120(l)	250(l)	24	T6
					120(l)	250(l)	24	T62
					(h)(k)	(h)(k)	(h)(k)	T73(g)
					(f)	(f)	(f)	T76(g)
				W510(d)	120(l)	250(l)	24	T6510(d)
					(h)(k)	(h)(k)	(h)(k)	T73510(d)(g)
					(f)	(f)	(f)	T76510(g)
				W511(d)	120(l)	250(l)	24	T6511(d)
					(h)(k)	(h)(k)	(h)(k)	T73511(d)(g)
					(f)	(f)	(f)	T76511(g)
	Drawn tube	465	870	W	120	250	24	T6
					120	250	24	T62
					(h)(k)	(h)(k)	(h)(k)	T73(g)
	Die forgings	470(m)	880(h)	W	120	250	24	T6
					(h)	(h)	(h)	T73(g)
				W52(n)	(h)	(h)	(h)	T7352(n)(g)
	Hand forgings	470(m)	880(h)	W	120	250	24	T6
					(h)	(h)	(h)	T73(g)
				W52(n)	120	250	24	T652(n)
					(h)	(h)	(h)	T7352(n)(g)
	Rolled rings	470	880	W	120	250	24	T6
7175	Die forgings	(o)	(o)	W	(o)	(o)	(o)	T66(o)
		(o)	(o)	W	(o)	(o)	(o)	T74(g)(o)
		(o)	(o)	W52(n)	(o)	(o)	(o)	T7452(n)(g)(o)
	Hand forgings	(o)	(o)	W	(o)	(o)	(o)	T74(g)(o)
		(o)	(o)	W52(n)	(o)	(o)	(o)	T7452(n)(g)(o)
7475	Sheet	515(p)	960(p)	W	120	250	3	
					plus 155	315	3	T61(p)
					(f)	(f)	(f)	T761(g)(p)
	Plate	510(p)	950(p)	W51(d)	120	250	24	T651(p)
					(f)	(f)	(f)	T7651(g)(p)
					(h)	(h)	(h)	T7351(g)(p)
Alclad 7475	Sheet	495	920	W	120	250	3	
					plus 155	315	3	T61(p)
					(f)	(f)	(f)	T761(g)(p)

(a) Material should be quenched from the solution-treating temperature as rapidly as possible and with minimum delay after removal from the furnace. When material is quenched by total immersion in water, unless otherwise indicated, the water should be at room temperature, and should be suitably cooled so that it remains below 38 °C (100 °F) during the quenching cycle. Use of high-velocity, high-volume jets of cold water also is effective for some materials. (b) The nominal temperatures listed should be attained as rapidly as possible and maintained within ±6 °C (±10 °F) of nominal during the time at temperature. (c) Approximate time at temperature. The specific time will depend on the time required for the load to reach temperature. The times shown are based on soak heating, with soak time measured from the time the load reaches a temperature within 6 °C (10 °F) of the applicable temperature. (d) Stress relieved by stretching to produce a specified amount of permanent set after solution treatment and prior to precipitation heat treatment. (e) No solution heat treatment; 72 h at room temperature following press quench, followed by two-stage precipitation heat treatment comprised of 8 h at 107 °C (225 °F) plus 16 h at 149 °C (300 °F). (f) Aging practice varies with product, size, nature of equipment, loading procedures, and furnace-control capabilities. The optimum practice for a specific item can be ascertained only by actual trial treatment of the item under specific conditions. Typical procedures involve a two-stage treatment comprised of 3 to 30 h at 121 °C (250 °F) followed by 15 to 18 h at 163 °C (325 °F) for extrusions. An alternative two-stage treatment of 8 h at 99 °C (210 °F) followed by 24 to 28 h at 163 °C (325 °F) also may be used. (g) Aging of aluminum alloys 7050, 7075, 7175, and 7475 from any temper to the T73 or T76 temper series requires closer-than-normal controls on aging variables such as time, temperature, heatup rate, and so forth, for any given item. In addition, when material in a T6-type temper is reaged to a T73- or T76-type temper, the specific condition of the T6 material (such as property levels and other effects of processing variables) is extremely important and will affect the capability of the reaged material to conform to the requirements specified for the applicable T73- or T76-type temper. (h) Two-stage treatment comprised of 6 to 8 h at 107 °C (225 °F) followed by: 24 to 30 h at 163 °C (325 °F) for sheet and plate; 8 to 10 h at 177 °C (350 °F) for rolled or cold finished rod and bar; 6 to 8 h at 177 °C (350 °F) for extrusions and tube; 8 to 10 h at 177 °C (350 °F) for forgings in the T73 temper; and 6 to 8 h at 177 °C (350 °F) for forgings in the T7352 temper. (i) These heat treatments also apply to alclad sheet and plate of these alloys. (j) An alternative two-stage treatment comprised of 4 h at 96 °C (205 °F) followed by 8 h at 157 °C (315 °F) also may be used. (k) For sheet, plate, tube, and extrusions, an alternative two-stage treatment comprised for 6 to 8 h at 107 °C (225 °F) followed by 14 to 18 h at 168 °C (335 °F) may be used, provided that a heatup rate of approximately 14 °C/h (25 °F/h) is employed. For rolled or cold finished rod and bar, the alternative treatment is 10 h at 177 °C (350 °F). (l) An alternative three-stage treatment comprised of 5 h at 99 °C (210 °F), 4 h at 121 °C (250 °F), and then 4 h at 149 °C (300 °F) may also be used. (m) Quenched in water at 60 to 80 °C (140 to 180 °F). (n) Stress relieved by 1 to 5% cold reduction after solution treatment and prior to precipitation heat treatments. (o) 7175-T74 and -T7452 heat treatments are directed to specific results, may vary from supplier to supplier and are either proprietary or patented. (p) Must be preceded by soak at 466 to 477 °C (870 to 890 °F). See U.S. Patent 3,791,880.

Table 2 Soak times and maximum quench delays for solution treatment of wrought aluminum alloys

See Table 1 for solution-treating temperatures.

Thickness(a), mm (in.)	Soak time, minutes				Maximum quench delay, s
	Air furnace(b)		Salt bath(c)		
	min	max(d)	min	max(d)	
≤0.41 (0.016)	20	25	10	15	5
0.51 (0.020)	20	30	10	20	7
0.64 (0.025)	25	35	15	25	7
0.81 (0.032)	25	35	15	25	7
1.02 (0.040)	30	40	20	30	10
1.27 (0.050)	30	40	20	30	10
1.35 (0.053)	30	40	20	30	10
1.80 (0.071)	35	45	25	35	10
2.03 (0.080)	35	45	25	35	10
2.29 (0.090)	35	45	25	35	10
2.54 (0.100)	40	55	30	45	15
3.18 (0.125)	40	55	30	45	15
4.06 (0.160)	50	60	35	45	15
4.57 (0.180)	50	60	35	45	15
6.35 (0.250)	55	65	35	45	15
>6.35 (0.250)–12.7 (0.500)	65	75	45	55	15
For each additional 12.7 (½) or fraction	+30	+30	+20	+20	(e)
Rivets (all)	60	· · ·	30	· · ·	5

(a) Minimum dimension of thickest section. (b) Soak time begins when all pyrometer instruments recover to original operating temperature. (c) Soak time begins at time of immersion except when a heavy charge causes bath temperature to drop below specified minimum, in which case soak time begins when bath regains minimum temperature. (d) Applicable to alclad materials only. (e) Increases in thickness above 12.7 mm (½ in.) do not affect maximum quench delay, which remains constant at 15 s.

In the tabulation above, note especially the effects of small increments of temperature, within the normal range, on the properties of 0.8 mm (0.032 in.) 2024-T4 sheet.

Solution-Treating Time. The time at the nominal solution heat-treating temperature (soak time) required to effect a satisfactory degree of solution of the undissolved or precipitated soluble phase constituents and to achieve good homogeneity of the solid solution is a function of microstructure before heat treatment. This time requirement can vary from less than a minute for thin sheet to as much as 20 h for large sand or plaster-mold castings. Guideline information for soak times required for wrought products of various section thicknesses is given in Table 2. Similar guidelines for castings are presented in Table 3. The time required to heat a load to the treatment temperature in furnace heat treatment also increases with section thickness and furnace loading, and thus total cycle time increases with these factors.

Soak time for alclad sheet and for parts made from alclad sheet must be held to a minimum, because excessive diffusion of alloying elements from the core into the cladding reduces corrosion protection. For the same reason, reheat treatment of alclad sheet less than 0.75 mm (0.030 in.) thick generally is prohibited, and the number of reheat treatments permitted for thicker alclad sheet is limited.

The soak times for wrought alloys take into account the normal thermal lag between furnace and part and the difference between surface and center temperatures for commercial equipment qualified to the standards of MIL-H-6088. The rapid heating rates of salt baths permit all immersion time to be counted as soak time unless the bath temperature drops below the minimum of the range. Even then, soak time begins as soon as the bath temperature returns to the minimum. In air furnaces, soak time does not begin until all furnace instruments return to their original set temperature—that is, the temperature reading before insertion of the load.

In air furnaces, thermocouples may also be attracted to, or buried in, parts located in the load in such a manner as to represent the hottest and coldest temperatures in each zone. In this way, it is possible to ensure that adequate soaking is obtained.

Special consideration is given also to establishing soak times for hand and die forgings; soak time in some specifications is extended to complete solution and homogenization in areas that received marginal reduction during forging. Considerable variation exists in the amount of soak time added; some specifications call for an arbitrary addition, such as one hour, and others require one hour per inch of thickness of the original forging.

In air furnaces, careful attention should be given to arrangement of the load. Air flow and natural temperature distribution within the furnace should be arranged to:

- Offer minimum resistance to air flow
- Produce the least disturbance in the natural temperature distribution
- Afford constant replenishment of the envelope of air around each part

It is common practice to specify a minimum spacing of 50 mm (2 in.) between parts, but large complex shapes may require considerably greater spacing. Many operators have found conservative loading practices to be more economical in the long run than heavier loading, because with lighter loads heating rates are higher and fewer rejections and service failures are encountered.

High-Temperature Oxidation. There is a condition, commonly but erroneously known as HTO or high-temperature oxidation, which can lead to deterioration of properties in aluminum alloys. High-temperature oxidation is a misnamed condition of hydrogen diffusion that affects surface layers during elevated-temperature treatment. This condition can result from moisture contamination in the furnace atmosphere and is sometimes aggravated by sulfur (as in heat-treatment furnaces also used for magnesium alloy castings) or other furnace refractory contamination.

Moisture in contact with aluminum at high temperatures serves as a source of nascent hydrogen, which diffuses into the metal. Foreign materials, such as sulfur compounds, function as decomposers of the natural oxide surface film, eliminating it as a barrier either between the moisture and the aluminum or between the nascent hydrogen and the aluminum. The most common manifestation of high-temperature oxidation is surface blistering, but occasionally the only manifestations are internal discontinuities or voids, which can be detected only by careful ultrasonic inspection or by metallographic techniques.

It is important to recognize that the symptoms of high-temperature oxidation are identical to those of unsoundness or high gas content in the original ingot or of other improper mill practice. Blisters resulting from ingot defects, improper extrusion or improper rolling may be lined up in the direction of working. However, it usually is impossible to distinguish among defect sources, and therefore the possibility that a contaminated atmosphere is the cause of the defects must be checked.

Not all alloys and product forms are equally vulnerable to this type of attack. The 7xxx series alloys are most susceptible, followed by the 2xxx alloys. Extrusions undoubtedly are the most susceptible form; forgings are probably second. Low-strength alloys and alclad sheet and plate are relatively immune to high-temperature oxidation. (Blistering of alclad material as a result of inadequate bonding is not the same as the blistering caused by high-temperature oxidation.)

If the protective oxide film formed during mill operations is removed from the mill product by a subsequent mechanical conditioning operation such as sanding, the conditioned surface will be more susceptible to high-temperature oxidation than those from which the film was not removed.

Moisture can be minimized by thoroughly drying parts and racks before they are charged. Drain holes often are needed in racks of tubular construction to avoid en-

Table 3 Typical heat treatments for aluminum alloy sand and permanent mold castings

Alloy	Temper	Type of casting(a)	Solution heat treatment(b) Temperature(c) °C	°F	Time, h	Aging treatment Temperature(c) °C	°F	Time, h
201.0(d)	T4	S or P	490–500(e)	910–930(e)	2	...	...	...
			+525–530	+980–990	14–20	Minimum of 5 days at room temperature		
	T6	S	510–515(e)	950–960(e)	2	...	...	...
			+525–530	+980–990	14–20	155	310	20
	T7	S	510–515(e)	950–960(e)	2	...	...	...
			+525–530	+980–990	14–20	190	370	5
	T43(f)	...	525	980	20	24 h at room temperature + ½ to 1 h at 160 °C		
	T71	...	490–500(e)	910–930(e)	2	...	...	...
			+525–530	+980–990	14–20	200	390	4
204.0(d)	T4	S or P	530	985	12	Minimum of 5 days at room temperature		
	T4	S or P	520	970	10	...	...	...
	T6(g)	S or P	530	985	12	(g)	(g)	...
206.0(d)	T4	S or P	490–500(e)	910–930(e)	2	...	...	...
			+525–530	+980–990	14–20	Minimum of 5 days at room temperature		
	T6	S or P	490–500(e)	910–930(e)	2	...	...	...
			+525–530	+980–990	14–20	155	310	12–24
	T7	S or P	490–500(e)	910–930(e)	2	...	...	...
			+525–530	+980–990	14–20	200	390	4
	T72	S or P	490–500(e)	910–930(e)	2	...	...	...
			+525–530	+980–990	14–20	243–248	470–480	
208.0	T55	S	...	...	...	155	310	16
222.0	O(h)	S	...	...	...	315	600	3
	T61	S	510	950	12	155	310	11
	T551	P	...	...	...	170	340	16–22
	T65	...	510	950	4–12	170	340	7–9
242.0	O(i)	S	...	...	...	345	650	3
	T571	S	...	...	...	205	400	8
		P	...	...	...	165–170	330–340	22–26
	T77	S	515	960	5(j)	330–355	625–675	2 (minimum)
	T61	S or P	515	960	4–12(j)	205–230	400–450	3–5
295.0	T4	S	515	960	12	...	...	...
	T6	S	515	960	12	155	310	3–6
	T62	S	515	960	12	155	310	12–24
	T7	S	515	960	12	260	500	4–6
296.0	T4	P	510	950	8	...	...	...
	T6	P	510	950	8	155	310	1–8
	T7	P	510	950	8	260	500	4–6
319.0	T5	S	...	...	...	205	400	8
	T6	S	505	940	12	155	310	2–5
		P	505	940	4–12	155	310	2–5
328.0	T6	S	515	960	12	155	310	2–5
332.0	T5	P	...	...	...	205	400	7–9
333.0	T5	P	...	...	...	205	400	7–9
	T6	P	505	950	6–12	155	310	2–5
	T7	P	505	940	6–12	260	500	4–6
336.0	T551	P	...	...	...	205	400	7–9
	T65	P	515	960	8	205	400	7–9
354.0	...	(k)	525–535	980–995	10–12	(h)	(h)	(l)
355.0	T51	S or P	...	...	...	225	440	7–9
	T6	S	525	980	12	155	310	3–5
		P	525	980	4–12	155	310	2–5
	T62	P	525	980	4–12	170	340	14–18
	T7	S	525	980	12	225	440	3–5
		P	525	980	4–12	225	440	3–9
	T71	S	525	980	12	245	475	4–6
		P	525	980	4–12	245	475	3–6
C355.0	T6	S	525	980	12	155	310	3–5
	T61	P	525	980	6–12	Room temperature		8 (minimum)
						155	310	10–12
356.0	T51	S or P	...	...	...	225	440	7–9
	T6	S	540	1000	12	155	310	3–5
		P	540	1000	4–12	155	310	2–5
	T7	S	540	1000	12	205	400	3–5
		P	540	1000	4–12	225	440	7–9
	T71	S	540	1000	10–12	245	475	3
		P	540	1000	4–12	245	475	3–6
A356.0	T6	S	540	1000	12	155	310	3–5
	T61	P	540	1000	6–12	Room temperature		8 (minimum)
						155	310	6–12

(continued)

(a) S, sand; P, permanent mold. (b) Unless otherwise indicated, solution treating is followed by quenching in water at 65–100 °C (150–212 °F). (c) Except where ranges are given, listed temperatures are ±6 °C or ±10 °F. (d) Casting wall thickness, solidification rate, and grain refinement affect the solution heat-treatment cycle in alloys 201.0, 204.0, and 206.0, and care must be taken in approaching the final solution temperature. Too rapid an approach can result in the occurrence of incipient melting. (e) For castings with thick or other slowly solidified sections, a pre-solution heat treatment ranging from about 490 to 515 °C (910 to 960 °F) may be needed to avoid too rapid a temperature rise to the solution temperature and the melting of CuAl₂. (f) Temper T43 for 201.0 was developed for improved impact resistance with some decrease in other mechanical properties. Typical Charpy value is 20 J (15 ft · lb). (g) The French precipitation treatment technology for the heat treatment of 204.0 alloy requires 12 h at temperature. The aging temperatures of 140, 160, or 180 °C (285, 320, or 355 °F) are selected to meet the required combination of properties. (h) Stress relieve for dimensional stability as follows: hold 5 h at 413 ± 14 °C (775 ± 25 °F); furnace cool to 345 °C (650 °F) over a period of 2 h or more; furnace cool to 230 °C (450 °F) over a period of not more than ½ h; furnace cool to 120 °C (250 °F) over a period of approximately 2 h; cool to room temperature in still air outside the furnace. (i) No quench required: cool in still air outside the furnace. (j) Air-blast quench from solution-treating temperature. (k) Casting process varies (sand, permanent mold, or composite) depending on desired mechanical properties. (l) Solution heat treat as indicated, then artificially age by heating uniformly at the temperature and for the time necessary to develop the desired mechanical properties. (m) Quench in water at 65–100 °C (150–212 °F) for 10–20 s only. (n) Cool to room temperature in still air outside the furnace.

Table 3 (continued)

Alloy	Temper	Type of casting(a)	Solution heat treatment(b) Temperature(c) °C	°F	Time, h	Aging treatment Temperature(c) °C	°F	Time, h
357.0	T6	P	540	1000	8	175	350	6
	T61	S	540	1000	10–12	155	310	10–12
A357.0	...	(k)	540	1000	8–12	(h)	(h)	(h)
359.0	...	(k)	540	1000	10–14	(h)	(h)	(h)
A444.0	T4	P	540	1000	8–12	...	...	...
520.0	T4	S	430	810	18(m)	...	...	...
535.0	T5(h)	S	400	750	5	...	...	...
705.0	T5	S	...	...	...	Room temperature		21 days
						100	210	8
		P	...	...	...	Room temperature		21 days
						100	210	10
707.0	T5	S	...	...	...	155	310	3–5
		P	...	...	...	Room temperature, or		21 days
						100	210	8
	T7	S	530	990	8–16	175	350	4–10
		P	530	990	4–8	175	350	4–10
710.0	T5	S	...	...	...	Room temperature		21 days
711.0	T1	P	...	...	...	Room temperature		21 days
712.0	T5	S	...	...	...	Room temperature, or		21 days
						155	315	6–8
713.0	T5	S or P	...	...	...	Room temperature, or		21 days
						120	250	16
771.0	T53(h)	S	415(n)	775(n)	5(n)	180(n)	360(n)	4(n)
	T5	S	...	...	...	180(n)	355(n)	3–5(n)
	T51	S	...	...	...	205	405	6
	T52	S	...	...	...	(h)	(h)	(h)
	T6	S	590(n)	1090(n)	6(n)	130	265	3
	T71	S	590(i)	1090(i)	6(i)	140	285	15
850.0	T5	S or P	...	...	...	220	430	7–9
851.0	T5	S or P	...	...	...	220	430	7–9
	T6	P	480	900	6	220	430	4
852.0	T5	S or P	...	...	...	220	430	7–9

(a) S, sand; P, permanent mold. (b) Unless otherwise indicated, solution treating is followed by quenching in water at 65–100 °C (150–212 °F). (c) Except where ranges are given, listed temperatures are ±6 °C or ±10 °F. (d) Casting wall thickness, solidification rate, and grain refinement affect the solution heat-treatment cycle in alloys 201.0, 204.0, and 206.0, and care must be taken in approaching the final solution temperature. Too rapid an approach can result in the occurrence of incipient melting. (e) For castings with thick or other slowly solidified sections, a pre-solution heat treatment ranging from about 490 to 515 °C (910 to 960 °F) may be needed to avoid too rapid a temperature rise to the solution temperature and the melting of CuAl$_2$. (f) Temper T43 for 201.0 was developed for improved impact resistance with some decrease in other mechanical properties. Typical Charpy value is 20 J (15 ft · lb). (g) The French precipitation treatment technology for the heat treatment of 204.0 alloy requires 12 h at temperature. The aging temperatures of 140, 160, or 180 °C (285, 320, or 355 °F) are selected to meet the required combination of properties. (h) Stress relieve for dimensional stability as follows: hold 5 h at 413 ± 14 °C (775 ± 25 °F); furnace cool to 345 °C (650 °F) over a period of 2 h or more; furnace cool to 230 °C (450 °F) over a period of not more than ½ h; furnace cool to 120 °C (250 °F) over a period of approximately 2 h; cool to room temperature in still air outside the furnace. (i) No quench required; cool in still air outside the furnace. (j) Air-blast quench from solution-treating temperature. (k) Casting process varies (sand, permanent mold, or composite) depending on desired mechanical properties. (j) Solution heat treat as indicated, then artificially age by heating uniformly at the temperature and for the time necessary to develop the desired mechanical properties. (m) Quench in water at 65–100 °C (150–212 °F) for 10–20 s only. (n) Cool to room temperature in still air outside the furnace.

trapment of water. Another common requirement is adjustment of the position of the quench tank with respect to furnace doors and air intake. Because it is unlikely that all moisture can be eliminated from the atmosphere in a production heat-treating furnace, it is extremely important to eliminate all traces of other contaminants from both the parts and the furnace atmosphere.

The most virulent contaminants in attacking aluminum are sulfur compounds. Residues from forming or machining lubricants, or from a sulfur dioxide protective atmosphere used in prior heat treatment of magnesium, are potential sources of sulfur contamination. In one plant, surface contamination resulted from sulfur-containing materials in tote boxes used to transport parts. In another, an epidemic of blistering was cured by rectifying a "sour" degreaser. In a third instance, it was found that a vapor-degreasing operation was not completely removing a thin, hard waxy residue, and an alkaline cleaning operation was added.

Very often, the source of contamination is obscure and difficult to detect, and the problem must be combated in another way. The most common of the alternative meth-

ods is use of a protective fluoborate compound in the furnace. Such a compound usually is effective in minimizing the harmful effects of moisture and other undesirable contaminants because it forms a barrier layer or film on the aluminum surface. The additive is not a universal solution; in some applications, high-temperature oxidation has occurred even though a fluoborate compound was employed. Also, the use of such compounds, particularly ammonium fluoborate, may present a hazard to personnel if used in poorly sealed furnaces or in furnaces that discharge their atmospheres into enclosed areas.

Protective fluoborate compounds accentuate staining or darkening of the parts being treated. (At times, this attack, particularly on parts located near the protective-compound container during heat treatment, has been severe enough to be termed "corrosion.") Although this minor nuisance might be considered a small price to pay for solution of a problem of high-temperature oxidation, the residual compound in the furnace dissipates slowly. Therefore, subsequent loads of alloys and product forms whose end uses require bright surfaces, and

that are not susceptible to high-temperature oxidation, may be detrimentally affected.

Successful use of fluoborate protective compounds appears to depend on specifying the right amount for each furnace; this must be established on a trial-and-error basis. One aircraft manufacturer adds 4 g/m^3 (0.004 oz/ft^3) of furnace chamber to each load. Another adds 0.45 kg (1 lb) per shift to a metal container hung on the furnace chamber wall, thus avoiding loss of the compound during quenching.

A second method of combating high-temperature oxidation is to anodize the work before it is heat treated. The resultant aluminum oxide film prevents attack by contaminants in the furnace atmosphere. The only deterrents to the use of anodizing are its cost (in money and time) and the slight surface frostiness which results from the subsequent stripping operation.

The usual objection to the blistered surface produced by high-temperature oxidation is its unsightly appearance. This often can be improved (for salvage purposes) by applying local pressure to flatten each blister and then finishing by a mechanical process such as polishing, buffing, sanding, or

abrasive blasting. In general, the effect of HTO on static properties and fatigue strength is slight. However, if a void resulting from HTO is located close to another stress concentration, such as a hole, much greater degradation of fatigue strength is likely. In critical aluminum alloy forgings, any blistering must be evaluated carefully for its effect on the integrity of the part. Any "cosmetic" salvage should be performed only after it has been established that the blisters are superficial and will not remain in the finished product.

Precipitation Heat Treating without Prior Solution Heat Treatment. Certain alloys that are relatively insensitive to cooling rate during quenching can be either air cooled or water quenched directly from a final hot-working operation. In either condition, these alloys respond strongly to precipitation heat treatment. This practice is widely used in producing thin extruded shapes of alloys 6061, 6063, 6463, and 7005. Upon precipitation heat treating after quenching at the extrusion press, these alloys develop strengths nearly equal to those obtained by adding a separate solution heat treating operation. Changes in properties occurring during the precipitation treatment follow the principles outlined in the discussion of solution heat–treated alloys.

Quenching

Quenching is in many ways the most critical step in the sequence of heat-treating operations. The objective of quenching is to preserve the solid solution formed at the solution heat-treating temperature, by rapidly cooling to some lower temperature, usually near room temperature. From the preceding general discussion, this statement applies not only to retaining solute atoms in solution, but also to maintaining a certain minimum number of vacant lattice sites to assist in promoting the low-temperature diffusion required for zone formation. The solute atoms that precipitate either on grain boundaries, dispersoids, or other particles, as well as the vacancies that migrate (with extreme rapidity) to disordered regions, are irretrievably lost for practical purposes and fail to contribute to the subsequent strengthening.

In most instances, to avoid those types of precipitation that are detrimental to mechanical properties or to corrosion resistance, the solid solution formed during solution heat treatment must be quenched rapidly enough (and without interruption) to produce a supersaturated solution at room temperature—the optimum condition for precipitation hardening. The resistance to stress-corrosion cracking of certain copper-free aluminum-zinc-magnesium alloys, however, is improved by slow quenching. Most frequently, parts are quenched by immersion in cold water or, in continuous heat treating of sheet, plate, or extrusions in

primary fabricating mills, by progressive flooding or high-velocity spraying with cold water. However, parts of complex shape, often with both thin and thick sections (such as die forgings, most castings, impact extrusions, and components formed from sheet) are commonly quenched in a medium that provides somewhat slower cooling. This medium may be water at 65 to 80 °C (150 to 180 °F), boiling water, an aqueous solution of polyalkylene glycol, or some other fluid medium such as forced air or mist.

If appreciable precipitation during cooling is to be avoided, two requirements must be satisfied. First, the time required for transfer of the load from the furnace to the quenching medium must be short enough to preclude slow precooling into the temperature range where very rapid precipitation takes place. For alloy 7075, this range was determined to be 400 to 290 °C (750 to 550 °F), and some sources quote this range (or a slightly different range) as the most critical range for quenching of any aluminum alloy. Later work has shown that the most critical range is alloy-dependent, and as will be discussed in detail under "Quench-Factor Analysis," significant errors can result from the assumption that precipitation is negligible outside of a so-called "critical range."

The second requirement for avoidance of appreciable precipitation during quenching is that the volume, heat-absorption capacity, and rate of flow of the quenching medium be such that little or no precipitation occurs during cooling. Any interruption of the quench that might allow reheating into a temperature range where rapid precipitation can occur must be prohibited.

For maximum dimensional stability, some forgings and castings are fan cooled or still-air cooled. In such instances, precipitation-hardening response is limited, but satisfactory values of strength and hardness are obtained. Extrusions produced without separate solution heat treatment can be air or mist quenched, but thicker sections may require water quenching by immersion or spraying. Alloys that are relatively dilute, such as 6063 and 7005, are particularly well suited to air quenching, and their mechanical properties are not greatly affected by its low cooling rate. Lower quenching rates are also employed for forgings, castings, and complex shapes to minimize warpage or other distortion and the magnitude of residual stresses developed as a consequence of temperature nonuniformity from surface to interior.

Effect of Quench Rate on Properties. As a broad generalization, the highest strengths attainable and the best combinations of strength and toughness are those associated with the most rapid quenching rates. Resistance to corrosion and stress-corrosion cracking are other characteristics that are generally improved by maximum rapidity of quenching. Some of the alloys used in arti-

ficially aged tempers, and in particular the copper-free 7xxx alloys, are exceptions to this rule. The effect of quench rate on mechanical properties may also depend on the desired temper. In the underaged condition, for example, a slow quench rate is more detrimental on ductility and fracture toughness. Strength would be more affected after near-to-peak aging.

Because of these effects, much work has been done over the years to understand and predict how quenching conditions and product form influence properties. The relative effects of quench methods can be compared in terms of average quench rates. In Fig 3, for example, the effects of quenching on the yield strength of four alloys are compared in terms of average quenching rates through the range from 400 to 290 °C (750 to 550 °F). For alloys relatively high in sensitivity to quenching rate, such as 7075, rates of about 300 °C/s (540 °F/s) or higher are required in order to obtain near-maximum strength after precipitation heat treatment. The other alloys in Fig 3 maintain their strengths at cooling rates as low as about 100 °C/s (180 °F/s). Similar comparisons in terms of average quench rates are shown in Tables 4 and 5.

Average quench rates are useful in comparing experimental results from various quench methods. In Table 4, for example, a severe reduction in strength occurred at the average quench rate of 36 °C/s (65 °F/s). However, average quench rates only compare results in a "critical" temperature range, where precipitation is most likely to occur. This method is not entirely accurate, because significant precipitation can also occur outside the specified critical temperature range of average quench rates. Moreover, for high-strength alloys, toughness and corrosion resistance may be impaired without significant loss of tensile strength.

Therefore, a more sophisticated comparison, known as quench-factor analysis, is needed for quantitative property prediction or property optimization. Quench-factor analysis, as discussed in a later section, is useful when cooling rates are nonuniform.

Delay in Quenching. Whether the transfer of parts from the furnace to the quench is performed manually or mechanically, it must be completed in less than the specified maximum time. The maximum allowable transfer time or "quench delay" varies with the temperature and velocity of the ambient air and the mass and emissivity of the parts. From cooling curves such as those illustrated in Fig 4, maximum quench delays (see table accompanying Fig 4) can be determined that will ensure complete immersion before the parts cool below 400 °C (750 °F). MIL-H-6088 specifies maximum quench delays for high-strength alloys of 5, 7, 10, and 15 s for thickness ranges of up to 0.016 in. (0.41 mm), 0.017 to 0.031 in. (0.43 to 0.79 mm), 0.032 to 0.090 in. (0.81 to 2.29 mm),
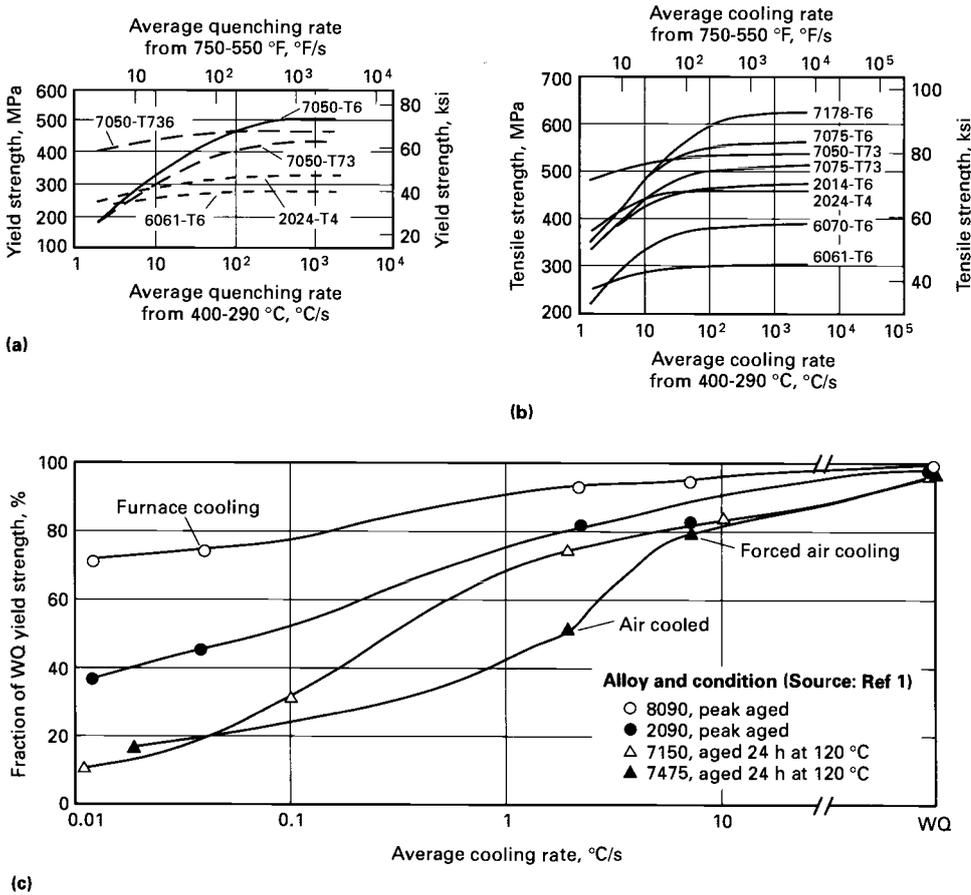
Fig 3 Quench sensitivity of various aluminum alloys as a function of average quench rates. (a) Yield strength after aging of four wrought alloys. (b) Tensile strength after aging of eight wrought alloys. (c) Relative quench sensitivity of two aluminum-lithium alloys (2090 and 8090, both solution treated for 1 h at 520 °C, or 970 °F) and two Zn-Mg-Cu aluminum alloys (7150 and 7475, both solution treated for 40 min at 480 °C, or 895 °F)

and over 0.090 in., respectively. Quench delay is conservatively defined as commencing "when the furnace door begins to open or the first corner of a load emerges from a salt bath" and ending "when the last corner of the load is immersed in the water quench tank." Recommended maximum quench-delay times are listed in Table 2. However, exceeding the

maximum delay time is permitted if temperature measurements of the load prove that all parts are above 415 °C (775 °F) when quenched. The C-curves used in quench-factor analysis can also assist in determining a maximum allowable delay.

It is relatively easy to control quench delay in day-to-day operations by using a

stopwatch or, if necessary, by attaching thermocouples to parts. However, although the cooling rate between 400 and 260 °C (750 and 500 °F) is most critical and must be extremely high for many high-strength alloys, it cannot be directly measured in production operations. It is usual to rely on standardized practices, augmented by results of tension tests and tests of susceptibility to intergranular corrosion.

Water-immersion quenching normally is controlled in practice by stipulating maximum quench-delay time and maximum water temperature. The first requirement controls the cooling rate during transfer and, for high-strength alloys, often is based on the criterion of complete immersion before the metal cools below 415 °C (775 °F). This specification of 415 °C (775 °F) is based on a critical temperature for alloy 7075, which has one of the more severe C-curves (Fig 5). Therefore, the criterion for complete immersion of other alloys might be based on a temperature lower than the 415 °C (775 °F) specification, depending on the characteristics of the particular C-curve.

The second requirement controls the cooling rate during immersion. MIL-H-6088 specifies that for water-immersion quenching, except quenching of forgings and castings, the temperature of the water shall not exceed 38 °C (100 °F) upon completion of quenching. This requirement controls both the temperature of the quench water prior to immersion and the ratio of the combined mass of load and rack to the volume of water. However, to ensure adequate quenching effectiveness, it is necessary also that the cooling fluid flow past all surfaces of each part during the first few seconds after immersion. Before parts enter the furnace, their placement in racks or baskets should be compatible with this requirement. During the first few seconds of quenching, agitation of the parts or the water should be sufficient to prevent local increases in temperature due to the formation of steam pockets.

In one application, it was found that 2024-T4 plates 13 by 760 by 760 mm (½ by 30 by 30 in.), quenched singly into a large volume of still water, were quite susceptible to intergranular corrosion. This susceptibility disappeared completely when the quenching practice was modified by adding sufficient agitation to break up the insulating blanket of steam that formed on the surface of the hot metal. Quenching practices for small parts such as fasteners and hydraulic fittings have been modified for the same reason. Dumping in bulk from baskets has been replaced by methods, such as the use of shaker hearth furnaces or special racking, which permit parts to be quenched singly.

Spray Quenching. For spray quenching, the quench rate is controlled by the velocity of the water and by volume of water per unit

Table 4 Effect of average quench rate on tensile properties of aluminum-lithium alloy 2090

Average quench rate at center of plate	Condition	Yield strength(a)		Tensile strength(a)		Elongation(a), %
		MPa	ksi	MPa	ksi	
0.5 °C/s (13 mm plate, air cooled)	As-quenched	162		334		2
	6% stretch + aged 8 h at 190 °C	448		513		5
36 °C/s (38 mm plate, quenched in room-temperature water)	As-quenched	128		312		12
	6% stretch + aged 8 h at 190 °C	338		476		6
46 °C/s (13 mm plate, quenched in boiling water)	As-quenched	138		331		16
	6% stretch + aged 8 h at 190 °C	530		570		9
48 °C/s (13 mm plate, quenched in room-temperature water)	As-quenched	139		331		17
	6% stretch + aged 8 h at 190 °C	526		570		7
85 °C/s (13 mm plate, quenched in ice brine)	As-quenched	135		349		19
	6% stretch + aged 8 h at 190 °C	535		575		7

(a) Data are averages from 4 specimens.

Table 5 The effect of quench rate on the mechanical properties of age-hardened aluminum-lithium alloy 8090

Alloy composition	Cooling from solution treatment(a)	Stretch, %	Aging treatment	Yield strength(b) MPa	Yield strength(b) ksi	Ultimate tensile strength(b) MPa	Ultimate tensile strength(b) ksi	Elongation in 50 mm (2 in.)(b), %
Al-2.28Li-0.86Cu-0.90Mg-0.13Zr-0.13Fe-0.06Si	Air cool (~0.25 °C/s)	2	190 °C for 16 h	380	55	446	64.5	7.7
		4	170 °C for 24 h	401	58	465	67.5	6.0
	Polymer quench (~18 °C/s)	2	190 °C for 16 h	415	60	481	70	8.0
		4	170 °C for 24 h	415	60	481	70	7.2
	Water quench (~120 °C)	2	190 °C for 16 h	428	62	492	71.4	8.1
Al-2.58Li-1.36Cu-0.89Mg-0.13Zr-0.17Fe-0.04Si	Air cool (~0.25 °C/s)	4	170 °C for 24 h	417	60	483	70	7.5
		2	190 °C for 16 h	417	60	485	70.3	6.5
		4	170 °C for 24 h	442	64	503	73	4.5
	Polymer quench (~18 °C/s)	2	190 °C for 16 h	448	65	524	76	6.8
		4	170 °C for 24 h	448	65	519	75	5.0
	Water quench (~120 °C/s)	2	190 °C for 16 h	464	67	535	77.5	8.2
		4	170 °C for 24 h	448	65	517	75	6.3

(a) Solution treatment of 550 °C (1020 °F) for 1 h. (b) Data are averages from two specimens.

area per unit time of impingement of the water on the workpiece. Rate of travel of the workpiece through the sprays is an important variable.

Local increases in temperature that occur within the first few seconds of quenching, caused by a phenomenon such as plugged spray nozzles, are particularly deleterious. The remaining "internal heat" may be sufficient to reheat the surface region. When this happens, a large loss in strength occurs at the previously quenched surface. The loss of strength in the affected area of a heavy part is much more severe than that caused by an inadequate quenching rate alone. This is illustrated for 75 mm (3 in.) thick 7075-T62 plate in Fig 6, which compares, at various depths, the properties of a plate for which quenching was interrupted on one side after 3 s with those of a plate that was quenched from one side only.

Quench Severity and Quenchant Selection. Quench severity is commonly expressed in terms of an H-value (or Grossmann number), where the H-value is related to the thermal conductivity (k) of the part(s) and the coefficient of heat transfer (C) be-

tween the quenchant and the part. These quantities are related by the equation $H = C/2k$, where the coefficient of heat transfer (C) is affected by the quenchant velocity at the surface of the part and several inherent characteristics of the quenchant (such as quenchant boiling point, viscosity, density, thermal conductivity, and specific heat).

Water, which is the most widely used and effective quenching medium, can obtain cooling rates up to about 200 °C/s (400 °F/s) at the midplane of 25 mm (1 in.) thick aluminum alloy plate (see the dashed line in Fig 7). No rates higher than those defined by this line have been observed, although rates approaching them were measured with impinging spray quenches. Lower cooling rates are achieved by immersion in heated water (Fig 7) or by reducing the velocity of the quenchant around the part (Table 6). Cooling rates can also be reduced by lowering surface tension or by increasing the stability of the vapor film around the part.

Polymer quenchants, which retard cooling rates by the formation of films around the part, are compared with water in Table 6. The effective film coefficient is essential-

ly the heat transfer coefficient (C), which is related to the Grossmann number (H). The application of polymer quenchants is covered in AMS specifications 3025 and 2770, although many aluminum and aerospace companies have developed internal specifications that differ from AMS-2770. Typical parameters for quenching wrought products (other than forgings) in glycol-water solutions are presented in Table 7.

Other Factors Affecting Quench Rate. Quenching rates are very sensitive to the surface condition of the parts. Lowest rates are observed with products having freshly machined or bright-etched, clean surfaces, or products that have been coated with materials that decrease heat transfer. The presence of oxide films or stains increases cooling rates. Further marked changes can be effected through the application of nonreflective coatings, which also accelerate heating (Fig 8). Surface roughness exerts a similar effect; this appears related to vapor film stability. The manner in which complex products, such as engineered castings and die forgings, enter the quenching medium can significantly alter the relative cooling

Thickness		Maximum quench delay, s	
mm	in.	Alclad	Nonclad
0.41	0.016	6.4	4.4
0.51	0.020	8.0	5.5
0.64	0.025	10.0	6.8
0.81	0.032	12.8	8.8
1.02	0.040	20.0	11.0
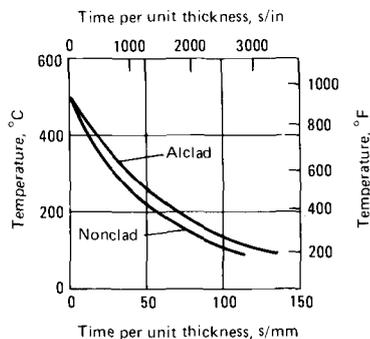
Fig 4 Cooling curves for alclad and nonclad aluminum products cooled from 495 °C (920 °F) in forced air. Air temperature, 25 °C (80 °F); air velocity, 2.3 m/s (450 ft/min). Tabulated values of quench delay (maximum delay before the material being quenched has cooled below 400 °C, or 750 °F) were determined from cooling curves shown.
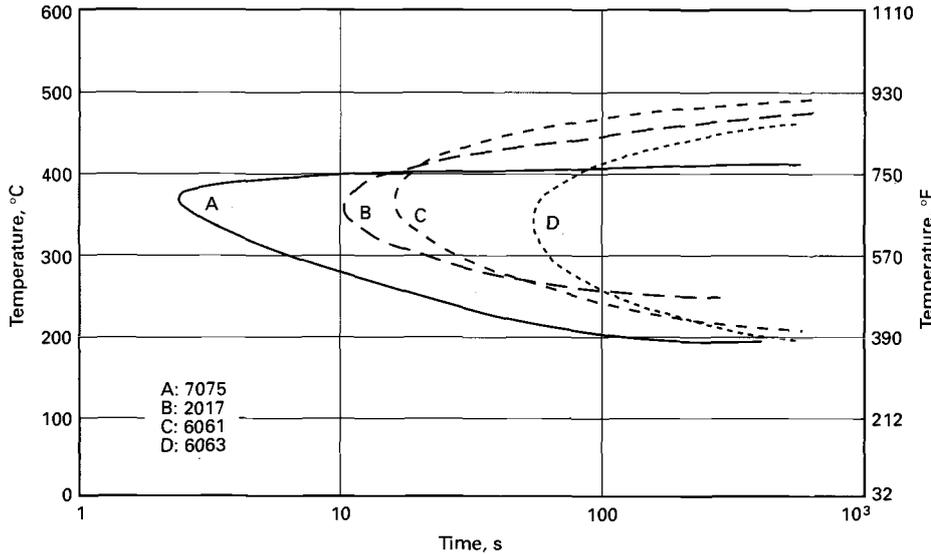
Fig 5 Time-temperature-property curves at 95% of maximum tensile stress for various alloys. See the section "Quench-Factor Analysis" for discussion. Source: Ref 2

rates at various points, thereby affecting mechanical properties and residual stresses established during quenching. Similarly, quenching complex extruded shapes whose wall thicknesses differ widely poses special problems if distortion and stresses are to be minimized. In batch heat-treating opera-tions, placement and spacing of parts on the racks can be a major factor in determining the quenching rates. In immersion quench-ing, adequate volumes of the quenching medium must be provided to prevent an excessive temperature rise in the medium. When jet agitation is used to induce water flow between parts, jets should not impinge directly and cause rapid localized cooling.

Quenching to Minimize Residual Stress and Warpage. Although cold-water immer-sion or flushing is most common, because it produces the most effective quench (and has been required by MIL-H-6088 for 2014, 2017, 2024, 2117, 7075, and 7178 alloys except forgings), it presents problems in-volving residual stress and warpage.

Residual stresses in heavy sections of aluminum alloys originate from differential thermal expansion during quenching —that is, the still-warm central material contracts, pulling in the already cooled outer shell. The magnitude of stresses increases with section size, as shown in Fig 9.

The distribution pattern of residual stresses in as-quenched parts (compression in the outer layers and tension in the central portion) is usually desirable in service. Compressive stresses inhibit failure by fa-tigue and stress corrosion—two mecha-nisms that initiate in the outer fibers. Un-fortunately, metal-removal operations required after heat treating often expose material that is stressed in tension. Also, metal-removal operations that are asym-metrical (with respect to residual stresses) cause distortion by redistributing residual stresses. When close-tolerance parts are being fabricated, the resulting warpage can be costly and difficult to correct.

Although service performance is some-times a factor, the major incentive for re-ducing residual stress differentials has been a reduction in warpage during machining or an improvement in shape before machining.

One approach to reducing the cooling-rate differential between surface and center is the use of a milder quenching medium— water that is hotter than that normally used or water-glycol solutions. Boiling water, which is the slowest quenching medium used for thick sections, is sometimes em-ployed for quenching wrought products even though it lowers mechanical properties and corrosion resistance. Quenching of castings in boiling water, however, is stan-dard practice, and is reflected in design allowables.

Another approach to the minimization of residual stresses that is generally successful consists of rough machining to within 3.2 mm (0.125 in.) or less of finish dimensions, heat treating, and then finish machining. This procedure is intended to reduce the cooling-rate differential between surface and center by reducing thickness; other benefits that accrue if this technique is used to reduce or reverse surface tension stress-es in finished parts are improvements in strength, fatigue life, corrosion resistance, and reduced probability of stress-corrosion cracking.

Several factors (especially quenching warpage) sometimes preclude general use of this procedure. The thinner and less sym-

Table 6 Grossmann numbers and heat transfer coefficients (C) of quenchant-to-part films

Type	Quenchant				Grossmann Number ($H = C/2k$)	Effective film heat transfer coefficient (C)	
	Temperature		Velocity				
	°C	°F	m/s	ft/min		W/cm² · K	Btu/ft² · h · °F
Water	27	80	0.00	0	1.07	3.55	2460
			0.25	50	1.35	4.78	3105
			0.50	100	1.55	5.14	3565
Water	38	100	0.00	0	0.99	3.28	2275
			0.25	50	1.21	4.01	2785
			0.50	100	1.48	4.91	3400
Water	49	120	0.00	0	1.10	3.65	2530
			0.25	50	1.29	4.29	2970
			0.50	100	1.60	5.31	3680
Water	60	140	0.00	0	0.86	2.85	1980
			0.25	50	1.09	3.62	2510
			0.50	100	1.33	4.41	3060
Water	71	160	0.00	0	0.21	0.70	485
			0.25	50	0.57	1.89	1310
			0.50	100	0.79	2.62	1815
Water	82	180	0.00	0	0.11	0.36	255
			0.25	50	0.21	0.69	485
			0.50	100	0.27	0.89	620
Water	93	200	0.00	0	0.06	0.20	138
			0.25	50	0.08	0.27	184
			0.50	100	0.09	0.30	207
Water	100	212	0.00	0	0.04	0.13	92
			0.25	50	0.04	0.13	92
			0.50	100	0.04	0.13	92
Polyalkylene glycol (UCON A)(a)	30	85	0.00	0	0.19	0.63	429
			0.25	50	0.21	0.70	475
			0.50	100	0.23	0.77	529
Polyvinyl pyrrolidone (PVP90)(a)	30	85	0.00	0	0.44	1.49	1012
			0.25	50	0.40	1.34	912
			0.50	100	0.42	1.41	966

(a) Polymer quenchants with concentrations of 25%. K is equal to the thermal conductivity of the aluminum alloy (7075). Source: Ref 4

Fig 6 Through-thickness property variations due to quench rate and temperature-rise effects in 7075-T62 plate 75 mm (3 in.) thick

Because of the difficulties encountered with quenching in cold water, milder quenchants have been employed. Indiscriminate use of milder quenchants can have catastrophic effects; however, when their use is based on sound engineering judgment and a metallurgical knowledge of the effects on the specific alloy, significant cost savings or performance improvements can be realized.

The most frequent advantage is the reduction in costly straightening operations and in resultant uncontrolled residual stresses. For example, one aircraft manufacturer utilizes water-spray and air-blast quenching for weldments and complex formed parts made from 6061, an alloy whose corrosion resistance is insensitive to quenching rate. Straightening requirements are negligible and, through careful control of racking and coolant flow, the decrease in mechanical properties is minimized, as shown by the data in Fig 10.

Another development for reducing straightening costs is quenching in water-polymer solutions. Quenching of formed sheet-metal parts in aqueous solutions of polyalkylene glycol or in similar inversely soluble media has significantly reduced the cost of straightening these parts after quenching. The SAE heat-treatment specification AMS-2770 recommends, for several alloys, maximum thicknesses that can be quenched in solutions of specific concentrations while maintaining acceptable property levels. Typical parameters for quenching wrought products (other than forgings) in glycol-water solutions are presented in Table 7. Additional information on polymer quenchants for aluminum alloys can be found in Ref 5.

Forming and Straightening after Quenching. Immediately after being quenched, most aluminum alloys are nearly as ductile as they are in the annealed condition. Consequently, it is often advantageous to form or straighten parts in this temper. Moreover, at the mill level, controlled mechanical deformation is the most common method of reducing residual quenching stresses. Because precipitation hardening will occur at room temperature, forming or straightening usually follows as soon after quenching as possible. In addition, maximum effectiveness in stress relief is obtained by working the metal immediately after quenching.

Forming and straightening operations vary in degree from minor corrections of warpage to complete forming of complex parts from solution-treated flat blanks. Particular value is gained when enough forming can be done at this stage of processing to eliminate the distortion caused by quenching. However, production operations must be adjusted so that most of the plastic deformation is accomplished before an appreciable amount of precipitation hardening takes place.

Although the most severe forming operations may have to be arranged to avoid

metrical a section, the more it will warp during quenching, and the residual stresses resulting from straightening of warped parts (plus straightening costs) often are less desirable than the quenching stresses. Holding fixtures and die quenching may be helpful, but precautions must be taken to ensure that they do not retard quenching rates excessively. Other factors that must be considered are the availability of heat-treating facilities and whether or not the advantages of such a manufacturing sequence offset the delay and cost entailed in a double-machining setup.

Warpage of thin sections during quenching is also a problem. Even in the same load, symmetry of cooling usually varies significantly among identical parts and the

resultant inconsistent warpage usually requires costly hand straightening. Consequently, a significant amount of effort has been devoted to reducing or eliminating warpage by changing racking positions to achieve symmetry of cooling.

For sheet-metal parts, one manufacturer uses a double screen floor in the quenching rack to reduce the force of initial contact between water and parts. Others allow parts to "free fall" from rack to quench tank. Spacing and positioning on the rack are carefully controlled so that parts will enter the water with minimum impact. With this technique, water turbulences must be avoided, because it will often cause parts to float for a few seconds, greatly reducing their cooling rate.

Fig 7 Effects of thickness and quenching medium on average cooling rates at midplane of aluminum alloy sheet and plate quenched from solution temperatures. The dashed line delineates the maximum cooling rates theoretically obtainable at the midplane of plate, assuming an infinite heat transfer coefficient (C) and a diffusivity factor of 1400 cm²/s. Source: Ref 3

natural aging, it often is desirable to allow some natural aging to occur and thus avoid formation of Lüders lines. This condition of nonuniform deformation is most likely to occur shortly after quenching and diminishes significantly after a few hours of natural aging. Complete freedom from Lüders lines, however, may require one or two days of natural aging prior to forming. Thus, the forming operation may have to be timed so as to obtain the most appropriate trade-off of these characteristics for the specific parts involved. Lüders lines also can be reduced by employing low strain rates or by forming at temperatures of 150 to 175 °C (300 to 350 °F).

Residual stresses in sheet-metal parts formed in the quenched condition are higher than those in parts formed in the annealed condition. Consequently, forming in the quenched condition should be selected judi-

ciously for parts that are critical in fatigue (Fig 11) or stress corrosion.

Re-solution heat treatment of parts formed after quenching often causes excessive grain growth in critically strained regions and thus is not recommended.

Quench-Factor Analysis

During the quenching of alloys from a solid-solution temperature condition, the rate of precipitation during quenching is maximized in a so-called "critical" temperature range, because the diffusion of dissolved species and the subsequent nucleation of precipitates exhibit opposite behavior as a function of temperature. At high temperatures, nucleation rates are small because of the low degree of supersaturation, and so precipitation rates are low despite the high diffusion rates. At low temperatures, diffusion rate is low, and thus

precipitation rate is low despite the high degree of supersaturation. At intermediate temperatures, precipitation rate is highest. Consequently, times to produce equal amounts of precipitation follow a C-shape pattern.

Using isothermal quenching techniques, Fink and Willey pioneered the attempts to describe the effects of quench rates with the use of C-curves (Ref 6). The C-curves plot the time required at different temperatures to precipitate a sufficient amount of solute to: reduce strength by a certain amount (Fig 5); cause a change in the corrosion behavior from pitting to intergranular (Fig 12); produce a given electrical conductivity (Fig 13); or relate other properties, such as fracture toughness, to isothermal quench conditions. The nose of the C-curves identifies the critical temperature range (the region of highest precipitation rates). Investigators use critical temperature ranges in conjunction with properties of samples quenched continuously from the solution temperature to compare relative sensitivities of alloys to quenching condition.

Although average quench rates through a critical temperature range can provide reasonable property predictions if cooling rates are fairly uniform, average quench rates cannot provide quantitative predictions when cooling rates vary considerably during the quench. For such instances, a procedure known as "quench-factor analysis" uses information from the entire C-curve to predict how any quench curve affects properties. Quench-factor analysis is useful in designing suitable limits for quench delays, or when it is not sufficient just to ensure that the cooling curve misses the nose of the C-curve.

The method of quench-factor analysis, as outlined by Evancho and Staley (Ref 7), is based on the determination of a quench factor (τ), which is the major variable in the following equation for precipitation kinetics during continuous cooling:

$$\zeta = 1 - \exp{(k\tau)} \qquad \text{(Eq 1)}$$

where ζ is the fraction transformed and k is a constant related to the transformation fraction of a given C-curve. The quench factor (τ) is defined as:

$$\tau = \int \frac{dt}{C_t} \qquad \text{(Eq 2)}$$

where t is time and C_t is critical time as a function of temperature to transform a specified fraction (x). The locus of critical times for a given transformation fraction x (or a percentage of mechanical properties from precipitation) is the C-curve, and the value of k is related to x as follows: $k = \ln(1 - x)$, or $e^k = 1 - x$. Therefore, when $\tau = 1$, the fraction transformed, ζ, equals the fraction value designated by the C-curve. Equation 2 is based on the assumption that the reaction rate is a function only of the amount transformed and temperature.

Table 7 Limits for quenching in glycol-water solutions

Data are for wrought aluminum alloy products other than forgings.

Glycol concentration, vol%	Alloys	Maximum thickness	
		mm	in.
12–16	2014, 2017, 2117, 2024, 2219	2.03	0.080
	7075, 7175	25.4	1.000
17–22	2014, 2017, 2117, 2024, 2219	1.80	0.071
	7075, 7079, 7175, 7178, 6061	12.7	0.500
23–28	2014, 2017, 2117, 2024, 2219	1.60	0.063
	7075, 7079, 7175, 7178, 6061	9.53	0.375
29–34	2014, 2017, 2117, 2024, 2219	1.02	0.040
	7075, 7079, 7175, 7178, 6061	6.35	0.250
35–40	7075, 7079, 7175, 7178, 6061	2.03	0.080
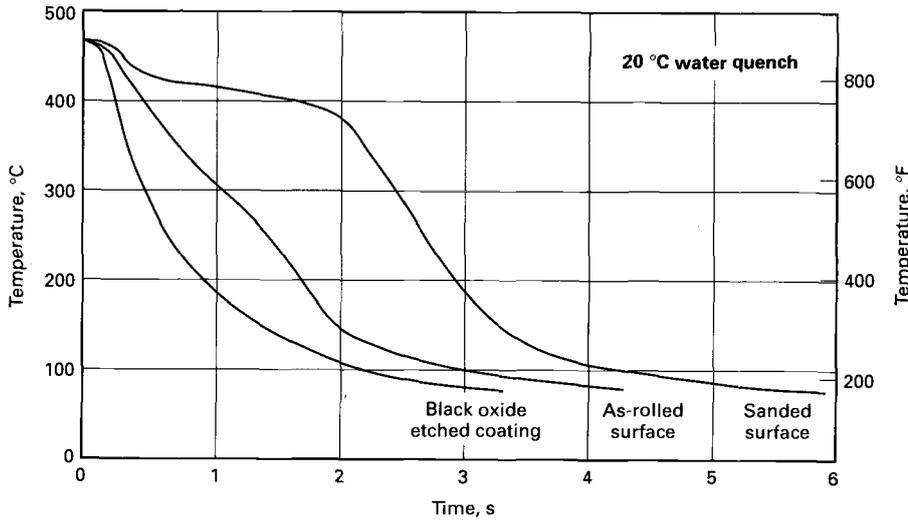
(a)
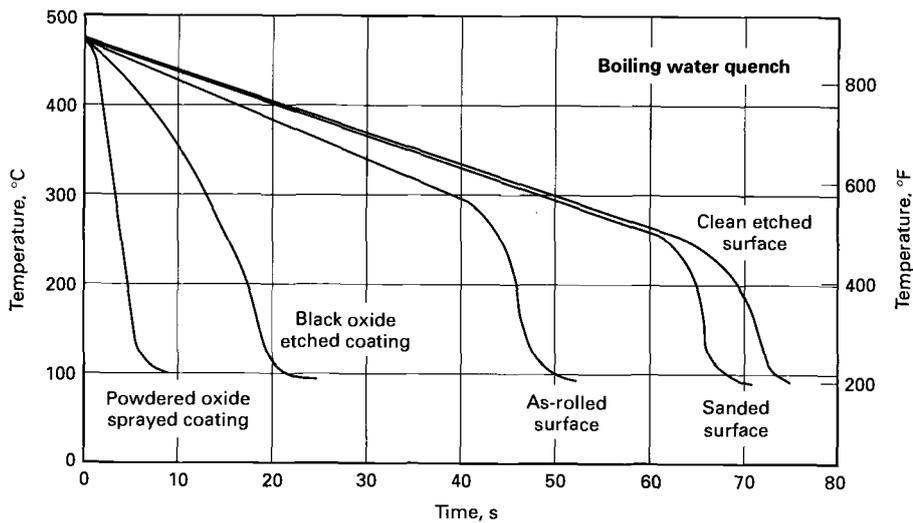
(b)

Fig 8 Effect of surface conditions on the midplane cooling of a 13 mm (0.5 in.) thick plate of 7075 from quenching in (a) 20 °C (70 °F) water and (b) boiling water. Source: Ref 5

Fig 9 Effect of quenching from 540 °C (1000 °F) on residual stresses in solid cylinders of alloy 6151

The numerical evaluation of the quench factor involves the integration of Eq 2. This integral can be graphically integrated using the method illustrated in Fig 14. Examples of the way to use the quench factor (τ) in the analysis of quench methods are described below. Neither the average quenching rate through a critical temperature range nor quench-factor analysis can predict strength when the temperature increases during quenching after it is cooled below some critical temperature. Under this condition, strength in the affected areas can be significantly lower than in other areas of the material. The most likely way for this phenomenon to occur is during spray quenching, when the surface cools rapidly by the impinging spray, but reheats by heat flow from the hotter interior when the spray is interrupted.

Predicting Strengths of Thick Products. Effects of the quenching rate on alloy strengths can be represented on a generalized graph of the type shown in Fig 3, and the expected quenching rates of products having various dimensions can be determined from Fig 7. Nevertheless, combining these two kinds of information to predict mechanical properties must be done with caution. Inconsistencies were encountered, for example, in correlating properties of thick sections quenched in high-cooling-rate media with properties of thinner sections quenched in media affording milder quenching action. One of the reasons for the inconsistencies is believed to be the different shapes of the cooling curves. This difficulty can be overcome by using quench-factor analysis. The other reason is that the degree of recrystallization and texture of the thick and thin sections may be different.

Predicting Corrosion Behavior. Alloy 2024-T4, for example, is susceptible to intergranular corrosion when a critical amount of solute is precipitated during quenching, but will corrode in the less severe pitting mode when lesser amounts are precipitated. For predicting the effects of proposed quenching conditions on the corrosion characteristics of 2024-T4, the postulated quench curve is drawn and the quench factor is calculated using the C-curve in Fig 12. Corrosion characteristics are predicted from the plot in Fig 15. When the quench factor (τ) is less than 1.0, continuously quenched 2024-T4 will corrode by pitting.

These relationships are applied to studies of effects of proposed changes in quench practice on design of new quenching systems. For example, consider that the goal of a proposed quenching system for 2024-T4 sheet products is to minimize warpage while preventing susceptibility to intergranular corrosion. Warpage occurs when the stresses imposed by temperature differences across the parts exceed the flow stress. As quenching rate decreases, the tendency for large differences in temperature to occur
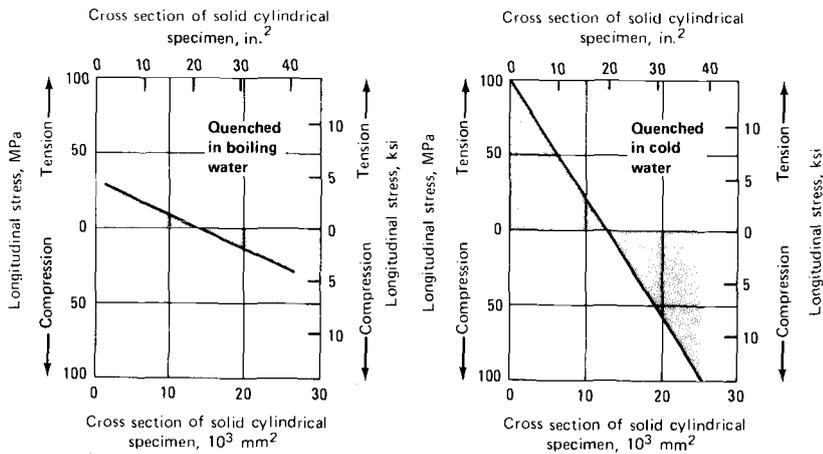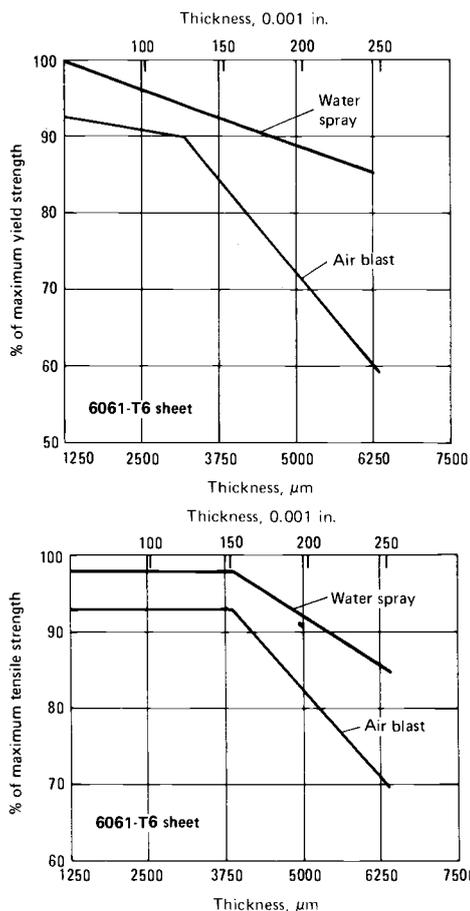
Fig 10 Effect of quenching medium on strength of 6061-T6 sheet. Water-immersion quench equals 100%. Control of coolant flow will minimize decrease in mechanical properties.

Fig 11 Fatigue characteristics of 1 mm (0.04 in.) alclad 2024-T4 sheet after 90° bending in the annealed condition and subsequent flattening as indicated. Flattening (unbending) was done either in the annealed condition (curve 2), or in the solution-treated and quenched condition (curves 3, 4, 5) with indicated storage times at −18 to −12 °C (0 to 10 °F).

decreases but the tendency for intergranular corrosion to occur increases.

The C-curve in Fig 12 indicates that quenching rate can be decreased near the solution heat-treating temperature and near room temperature without greatly sacrificing corrosion characteristics, but this information does not provide a quantitative answer. Simple calculations, however, can reveal a multitude of hypothetical cooling curves that provide slow quenching during a large portion of the quench cycle but sufficiently rapid quenching where critical times are short so that desirable corrosion characteristics are obtained.

As an example, one-, two-, and three-step quench curves that would ensure accept-

able corrosion behavior in 2024-T4 sheet (quench factor, 0.99) were calculated. Some of these curves are plotted in Fig 16. This illustration shows that 2024 can be quenched at a rate of 470 °C/s (850 °F/s) or higher and still develop acceptable corrosion characteristics if the quenching rate is linear from the solution temperature to 150 °C (300 °F). If sheet 3.2 mm (0.125 in.) thick is air-blast quenched (rate of heat removal, 5.68 W/m^2 · °C) to 395 °C (740 °F), however, the quenching rate from 395 to 150 °C must be at least 945 °C/s (1700 °F/s) to maintain the acceptable corrosion behavior. It may also be air-blast quenched to 395 °C (740 °F), spray quenched at 3300 °C/s (6000 °F/s) to 250 °C (480 °F), then air-blast quenched to 150 °C (300 °F).

Other curves could be drawn, of course, but the important points are that air-blast quenching cannot be continued to more than a few degrees below 395 °C (740 °F) and cannot be initiated at more than a few degrees above 270 °C (520 °F) even if infinite quenching rates are attained from 395 to 270 °C (740 to 520 °F).

Predicting yield strength is more complex than predicting corrosion behavior and requires some knowledge of the relationship between extent of precipitation and loss in ability to develop property. Because attainable strength of precipitation-hardening aluminum alloys is a function of the amount of solute remaining in solid solution after quenching, relationships between strength (σ_x) attainable after continuous cooling and

Fig 12 C-curve indicating type of corrosion attack on 2024-T4 sheet
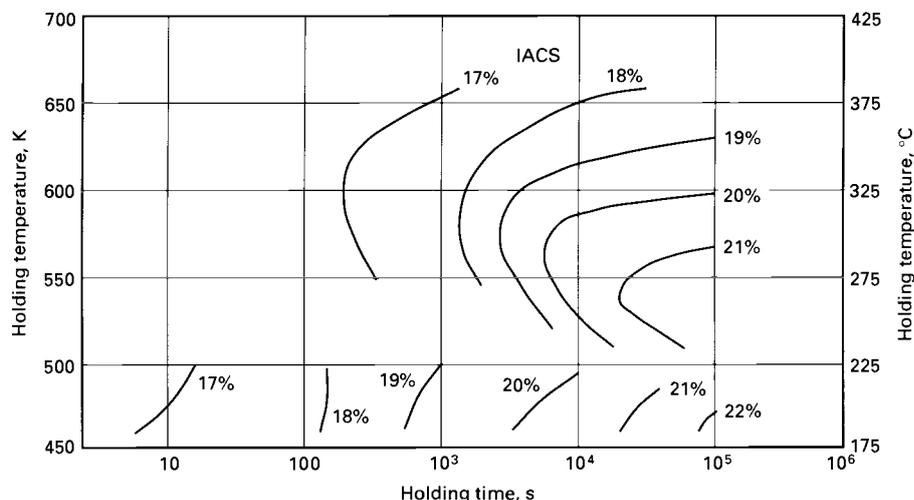
Fig 13 Change in electrical conductivity of an Al-2.5% Li binary alloy after the following: solution treated at 540 °C (1000 °F) for ½ h, immersed into an adjacent salt or oil bath for the appropriate isothermal holding temperature and time, then quenched into water. Source: Ref 1
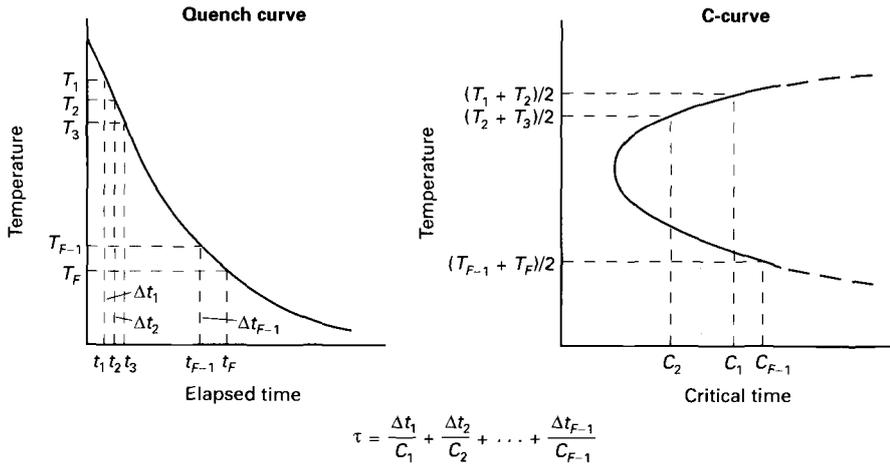
Fig 14 Method of determining quench factor, τ, using a cooling curve and a C-curve

$$\tau = \frac{\Delta t_1}{C_1} + \frac{\Delta t_2}{C_2} + \ldots + \frac{\Delta t_{F-1}}{C_{F-1}}$$

Fig 16 Quench curves for 2024-T4 sheet, to eliminate susceptibility to intergranular corrosion

Fig 15 Type and depth of attack on 2024-T4 sheet versus quench factor

quench factor (τ) can be expressed as follows:

$$\sigma_x = \sigma_{max} \exp (k_1 \tau) \qquad \text{(Eq 3)}$$

where σ_{max} is the strength attainable with an infinite quenching rate and:

$$\tau = \int \frac{dt}{C_x} \qquad \text{(Eq 4)}$$

where t is time and C_x is the C-curve for σ_x—that is, critical time as a function of temperature to reduce attainable strength to x of σ_{max}. The constant k_1 is related to the natural logarithm of x. For example, if τ is based on the C-curve for 99.5% of maximum yield strength, then $k_1 = -0.005013 = \ln (0.995)$.

The advantage of predicting yield strength from quench factor instead of from average quenching rate is illustrated by the

following comparison. Four specimens of alloy 7075-T6 quenched by various means (see Fig 17) were selected. Yield strengths were predicted both from average quenching rate between 400 and 290 °C (750 and 550 °F) and from quench factor. Quench factor was calculated using the C-curve for 99.5% maximum yield strength for 7075-T6 (Fig 18), and yield strength was estimated from the above equation defining the quench factor (τ) (see Fig 19).

A comparison of predicted yield strength with actual yield strength is given in Table 8. Yield strengths predicted from quench factor agree very well with measured yield strengths for all specimens, the maximum error being 19.3 MPa (2.8 ksi). Yield strengths predicted from average quenching rates, however, differ from measured values by as much as 226 MPa (32.8 ksi).

The advantage of using the quench factor for predicting yield strength from cooling curves is apparent. Cooling curves that have long holding times either above or below the critical temperature range from 400 to 290 °C (750 to 550 °F) cannot be used to predict yield strength from average quenching rate. In such instances, prediction of yield strength on the basis of quench factor is particularly advantageous.

With the use of finite-element analysis, quench factors can also be plotted as a function of Grossmann quench severity values (H) or the heat transfer coefficients (C) between the quenchant and a particular part (Fig 20). However, an underlying assumption of both quench-factor analysis and average-cooling-rate estimation is that the only effect of temperature is on the kinetics of precipitation. This assumption is not valid, however, when portions of the metal are quenched locally but reheated significantly before quenching is complete.

Age Hardening

After solution treatment and quenching, hardening is achieved either at room temperature (natural aging) or with a precipitation heat treatment (artificial aging). In some alloys, sufficient precipitation occurs in a few days at room temperature to yield stable products with properties that are adequate for many applications. These alloys sometimes are precipitation heat treated to provide increased strength and hardness in wrought or cast products. Other alloys with slow precipitation reactions at room temperature are always precipitation heat treated before being used.

In some alloys, notably those of the 2xxx series, cold working of freshly quenched material greatly increases its response to later precipitation heat treatment. Mills take advantage of this phenomenon by applying a controlled amount of rolling (sheet and plate) or stretching (extrusion, bar, and plate) to produce higher mechanical properties. However, if the higher properties are used in design, reheat treatment must be avoided.
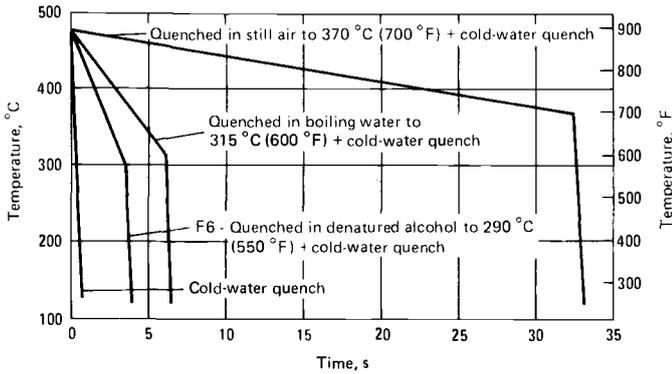
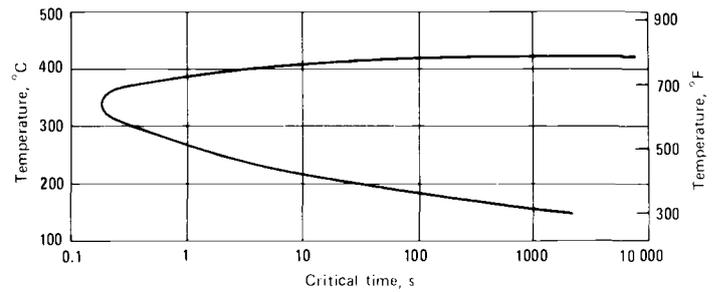**Fig 17** Cooling curves for 7075-T6 sheet

Fig 18 C-curve for 99.5% maximum yield strength of 7075-T6 sheet

Natural Aging. The more highly alloyed members of the 6xxx wrought series, the copper-containing alloys of the 7xxx group, and all of the 2xxx alloys are almost always solution heat treated and quenched. For some of these alloys—particularly the 2xxx alloys—the precipitation hardening that results from natural aging alone produces useful tempers (T3 and T4 types) that are characterized by high ratios of tensile to yield strength and high fracture toughness and resistance to fatigue. For the alloys that are used in these tempers, the relatively high supersaturation of atoms and vacancies retained by rapid quenching causes rapid formation of GP zones, and strength increases rapidly, attaining nearly maximum stable values in four or five days. Tensile-property specifications for products in T3- and T4-type tempers are based on a nominal natural aging time of four days. In alloys for which T3- or T4-type tempers are standard, the changes that occur on further natural aging are of relatively minor magnitude, and products of these combinations of alloy and temper are regarded as essentially stable after about one week.

In contrast to the relatively stable condition reached in a few days by 2xxx alloys that are used in T3- or T4-type tempers, the 6xxx alloys and to an even greater degree the 7xxx alloys are considerably less stable at room temperature and continue to exhibit significant changes in mechanical properties for many years. The differences in rate and

duration of changes in tensile yield strength of representative alloys of the three types are illustrated in Fig 21. Because of the relative instability of the 7xxx alloys, the naturally aged temper (after solution heat treatment and quenching) is designated by the suffix letter W. For a specific description of this condition, the time of natural aging should be included (example: 7075-W, 1 month).

Aging characteristics vary from alloy to alloy with respect to both time to initial change in mechanical properties and rate of change, but aging effects always are lessened by reductions in aging temperature (see Fig 21). With some alloys, aging can be suppressed or delayed for several days by holding at a temperature of $-18\ ^\circ C$ (0 $^\circ F$) or lower. It is usual practice to complete forming and straightening before aging changes mechanical properties appreciably. When scheduling makes this impractical, aging may be avoided in some alloys by refrigerating prior to forming. It is conventional practice to refrigerate alloy 2024-T4 rivets to maintain good driving characteristics. Full-size wing plates for current-generation jet aircraft have been solution heat treated and quenched at the primary fabricating mill, packed in dry ice in specially designed insulated shipping containers and transported by rail about 2000 miles to the aircraft manufacturer's plant for forming.

Unanticipated difficulties may arise as a result of failure to control refrigerator or part temperature closely enough. If opening of the cold box to insert or remove parts is done too frequently, the cooling capacity of the refrigerator may be exceeded. At times, the rate at which heavy-gage parts can be cooled in a still-air cold box has been found to be insufficient. This problem has been solved in one plant by immersing parts in a solvent at $-40\ ^\circ C$ ($-40\ ^\circ F$) before placing them in the refrigerator.

The T3-type tempers are distinguished from T4-type tempers by significant mechanical-property differences resulting from cold work strain hardening associated with certain mechanical operations performed after quenching. Roller or stretcher leveling to achieve flatness or straightness introduces modest strains (on the order of 1 to 4%) that cause changes in mechanical properties (primarily, increases in strength). Further increases in strength can be obtained by cold rolling, additional stretching, combinations of these operations, or for products such as hand forgings, compressive deformation. The tempers produced by these operations followed by natural aging alone (no precipitation heat treatment) are classified as T3-type tempers, and an additional digit is used to indicate a variation in strain hardening that results in significant changes in properties. In the most recently intro-
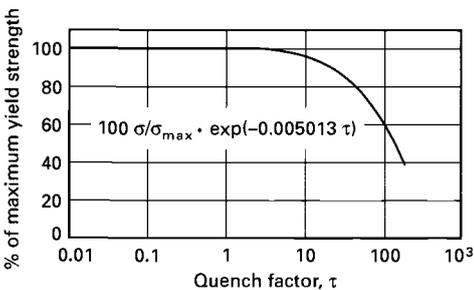

Fig 19 Yield strength versus quench factor

Table 8 Yield-strength values for 7075-T6 sheet predicted from cooling curves using average quench rate and quench factor

| Quench	Average quench rate from 400 to 290 °C (750 to 550 °F)		Quench factor, τ	Measured yield strength		Yield strength predicted from average quench rate		Yield strength predicted from quench factor	
	°C/s	°F/s		MPa	ksi	MPa	ksi	MPa	ksi
Cold water	935	1680	0.464	506	73.4	499	72.4	498	72.3
Denatured alcohol to 290 °C (550 °F), then cold water	50	90	8.539	476	69.1	463	67.2	478	69.4
Boiling water to 315 °C (600 °F), then cold water	30	55	15.327	458	66.4	443	64.2	463	67.1
Still air to 370 °C (700 °F), then cold water	5	9	21.334	468	67.9	242	35.1	449	65.1
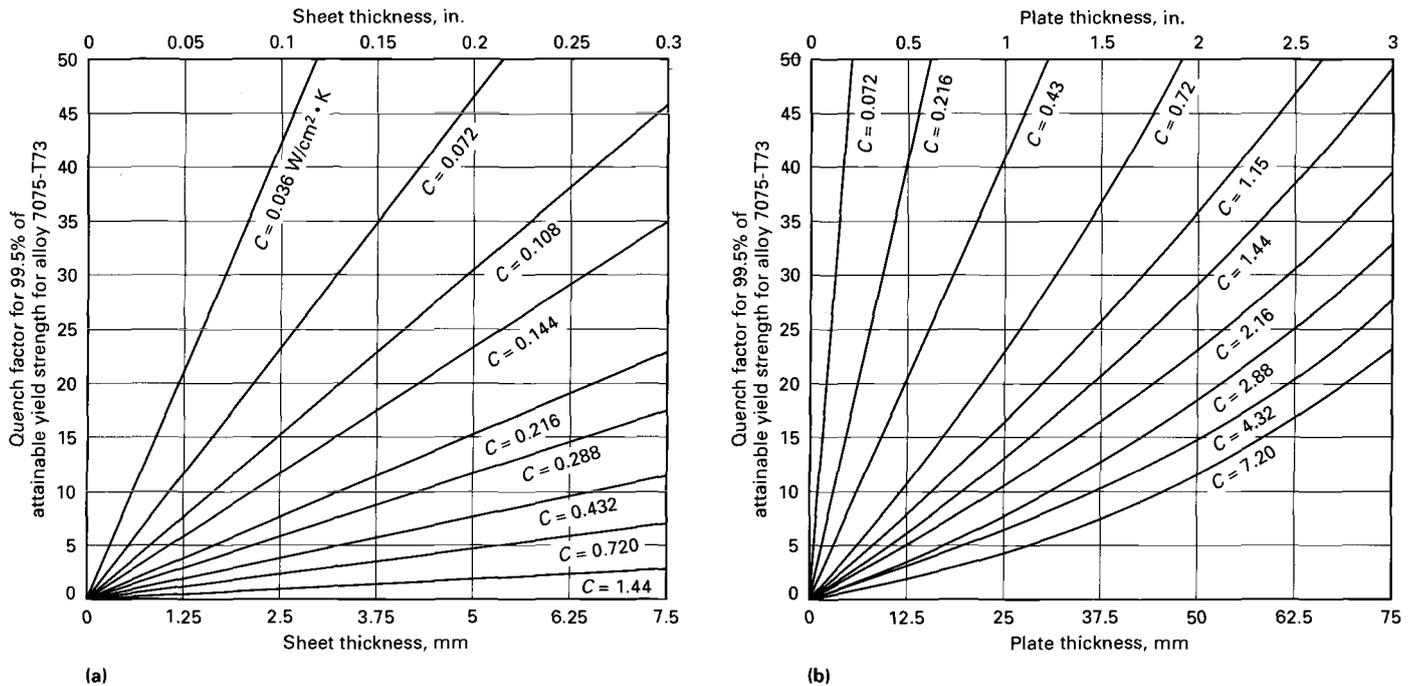
Fig 20 Plot of quench factors derived from finite element analysis with given product sizes and film (heat transfer) coefficients (C). Heat transfer coefficients between the quenchant and part are expressed in W/cm² · K. Source: Ref 4

duced 2*xxx* aircraft alloy, 2324, high strength is achieved by cold rolling plate to a T39 temper.

Precipitation heat treatments generally are low-temperature, long-term processes. Temperatures range from 115 to 190 °C (240 to 375 °F); times vary from 5 to 48 h.

Choice of time-temperature cycles for precipitation heat treatment should receive careful consideration. Larger particles of precipitate result from longer times and higher temperatures; however, the larger particles must, of necessity, be fewer in number with greater distances between them. The objective is to select the cycle that produces optimum precipitate size and distribution pattern. Unfortunately, the cycle required to maximize one property, such as tensile strength, is usually different from that required to maximize others, such as yield strength and corrosion resistance. Consequently, the cycles used represent compromises that provide the best combinations of properties.

Production of material in T5- through T10-type tempers (see the section on temper designations near the end of this article) necessitates precipitation heat treating at elevated temperatures (artificial aging). Although the hardening precipitate developed by this operation is submicroscopic, structures before and after precipitation heat treatment often can be distinguished by etching metallographic specimens. In aluminum alloys in the solution heat treated and quenched condition, coloration contrast between grains of differing orientation is relatively high, particularly in 2*xxx* series

wrought alloys and 2*xx*.0 series casting alloys. This contrast is noticeably decreased by precipitation heat treatment.

Differences in type, volume fraction, size, and distribution of the precipitated particles govern properties as well as the changes observed with time and temperature, and these are all affected by the initial state of the structure. The initial structure may vary in wrought products from unrecrystallized to recrystallized and may exhibit only modest strain from quenching or additional strain from cold working after solution heat treatment. These conditions, as well as the time and temperature of precipitation heat treatment, affect the final structure and the resulting mechanical properties.

Because mechanical properties and other characteristics change continuously with time and with temperature, as shown in Fig 22(a), (b), and (c) by typical curves for three wrought alloys, treatment to produce a combination of properties corresponding to a specific alloy-temper combination requires one or more rather specific and coordinated combinations of time and temperature, with both parameters being subject to practical limitations. Recommended commercial treatments often are compromises between time and cost factors and the probability of obtaining the intended properties, with consideration of allowances for variables such as composition within specified range and temperature variations within the furnace and load. Use of higher temperatures may reduce treatment time; but if the temperature is too high, characteristic fea-

tures of the precipitation-hardening process reduce the probability of obtaining the required properties.

T6 and T7 Tempers. Precipitation heat treatment following solution heat treatment and quenching produces T6- and T7-type tempers. Alloys in T6-type tempers generally have the highest strengths practical without sacrifice of the minimum levels of other properties and characteristics found by experience to be satisfactory and useful for engineering applications. Alloys in T7 tempers are overaged, which means that some degree of strength has been sacrificed or "traded off" to improve one or more other characteristics. Strength may be sacrificed to improve dimensional stability, particularly in products intended for service at elevated temperatures, or to lower residual stresses in order to reduce warpage or distortion in machining. T7-type tempers frequently are specified for cast or forged engine parts. Precipitation heat-treating temperatures used to produce these tempers generally are higher than those used to produce T6-type tempers in the same alloys.

Two important groups of T7-type tempers—the T73 and T76 types—have been developed for the wrought alloys of the 7*xxx* series, which contain more than about 1.25% copper. These tempers are intended to improve resistance to exfoliation corrosion and stress-corrosion cracking, but as a result of overaging, they also increase fracture toughness and, under some conditions, reduce rates of fatigue-crack propagation. The T73-type temper
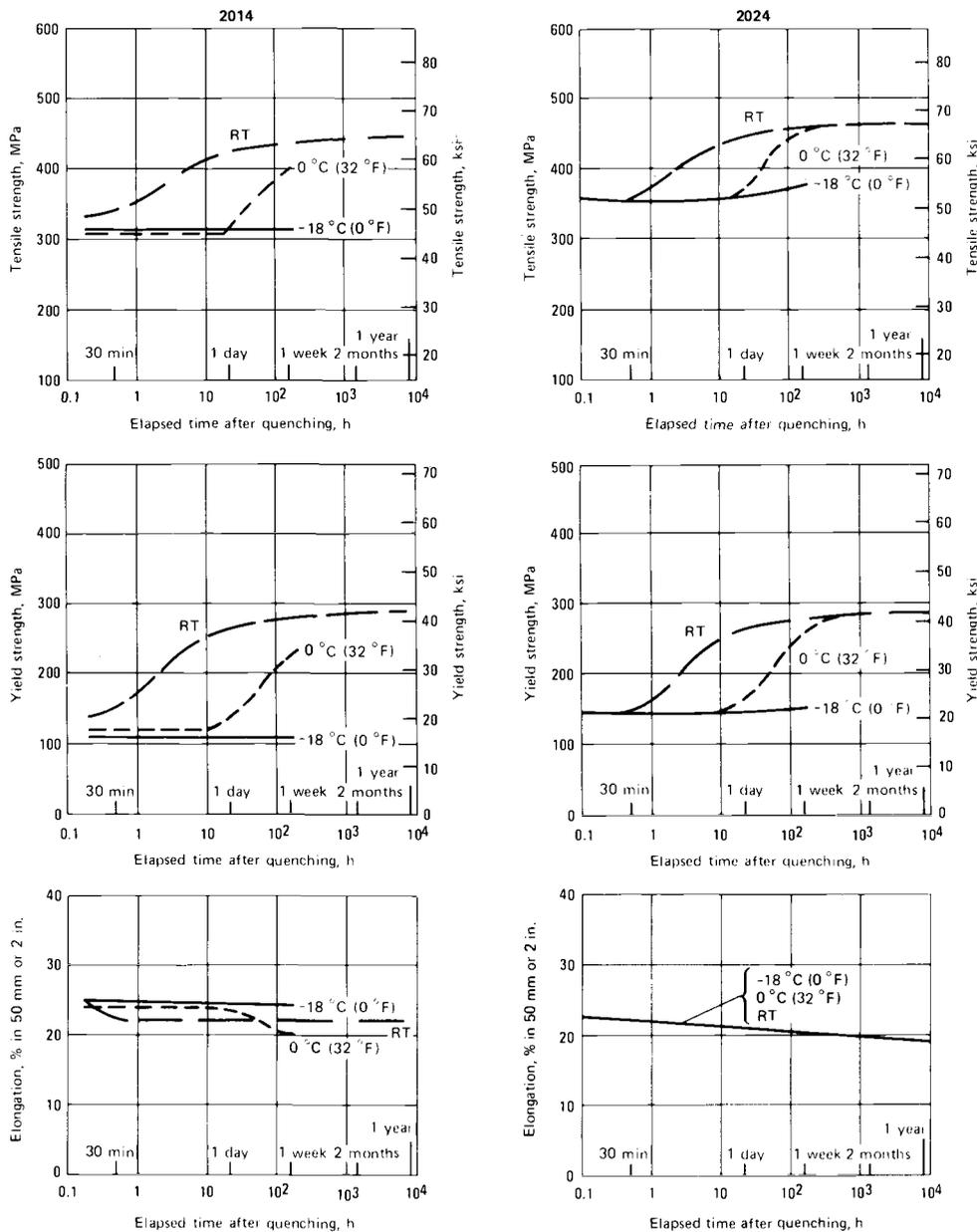
Fig 21 Aging characteristics of aluminum sheet alloys at room temperature, at 0 °C (32 °F), and at −18 °C (0 °F)

has greatly minimized stress-corrosion cracking of large and complex machined parts made of these alloys, which occasionally occurred with T6-type tempers. The precipitation heat treatment used to produce the T73- and T76-type tempers consist either of a two-stage isothermal precipitation heat treatment or of heating at a controlled rate to a single treatment temperature. The microstructural/ electrochemical relationships that are required in order to achieve the desired corrosion-resisting characteristics can be developed by using only a single-stage precipitation heat treatment above about 150 °C (300 °F), but higher strength is obtained by preceding this with a lower-temperature stage or with a slow-con-

trolled heatup. Extended natural aging can provide the same results, but the times required at room temperature are impractical. Either during the preliminary stage or during slow heatup, a fine, high-density dispersion of GP zones is nucleated. Either the time and temperature of the first step or the rate of heating must be controlled to produce GP zones that will not dissolve but will transform to the η' precipitate when heated to the aging temperature above 150 °C (300 °F). The aging practice that produces the results in the shortest time depends on the GP-zone solvus temperature. This temperature, in turn, depends on vacancy concentration, a factor influenced by solution heat-treating temperature and quench rate, and on compo-

sition. If first-step aging time is too short, if first-step aging temperature is too far below the GP-zone solvus, or if heating rates are too high, the GP zones will dissolve above 150 °C (300 °F), and the resultant coarse and widely distributed precipitate will provide lower strength. The T76-type treatments have the same operational sequence but employ second-stage heating only long enough to develop a resistance to exfoliation corrosion higher than that provided by the T6-type tempers. Materials in the T73-type temper also have high resistance to exfoliation corrosion.

Recommended treatments to produce T5- and T6-type tempers, and those of the T7-type employed for dimensional and property stabilization, provide adequate tolerance for normal variations encountered with good operating practices. On the other hand, the T73, T74 (formerly T736), and T76 tempers for alloys 7049, 7050, 7075, 7175, and 7475 involve changes in strength that occur significantly more rapidly at the temperatures employed in the second stage of the T7x precipitation heat-treatment cycle compared to the changes occurring at the temperatures employed to produce the T6 temper.

As illustrated in Fig 23, variations in soak time of several hours, and variations in soak temperature of up to 11 °C (20 °F) from the nominal aging practice of 24 h at 120 °C (250 °F) affect the strength of 7075-T6 by as much as 28 MPa (4 ksi). In contrast, similar variations in second-step soak time and temperature for 7075-T73—that is, variations for 24 h at 165 °C (325 °F)—affect strength by up to 150 MPa (22 ksi).

Consequently, control of both temperature and time to achieve the mechanical properties and corrosion resistance specified for these tempers is more critical than the control required in producing the T6 temper. Moreover, rate of heating from the first to the second aging step must be considered, because precipitation occurs during this period.

Heat treaters attempt to adjust these new problems by empirically modifying soak times to compensate for precipitation during heating and for effects of soaking at temperatures above or below the nominal. A method has been developed (Ref 8) that permits quantitative compensation for the effects of precipitation during heating and of soaking either above or below the recommended temperature. For overaging, these effects can be described by the following equation:

$$YS = Y \exp - \left(\frac{t_c}{F_{YS}} + \theta \right) \qquad \text{(Eq 5)}$$

where YS is yield strength; Y is a term having units of strength that is dependent on alloy, fabrication, and test direction; t_c is time at soak temperature; F_{YS} is a temperature-dependent term; and
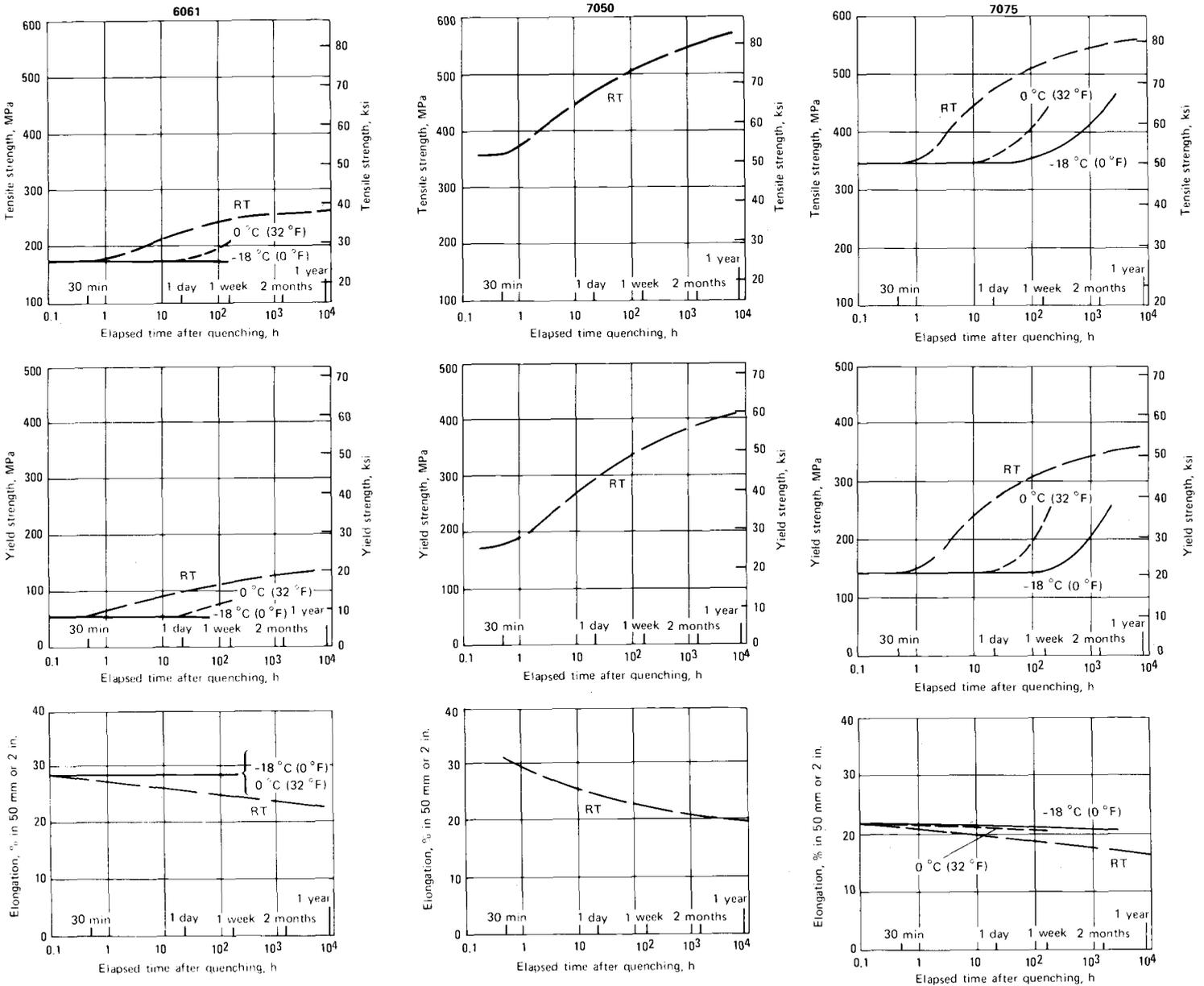
Fig 21 (continued)

$$\theta = \int \frac{dt}{F_{YS}} \qquad \text{(Eq 6)}$$

where t is time during heating.

Equation 5 provides the basis for selection of a nominal aging time that will result in the desired yield strength and gives the furnace operator a method of compensating for heating rate and for differences between desired and attained soak temperatures.

Specifics will be illustrated using data for alloy 7050. The value of F_{YS} (in units of hours) for 7050 can be calculated by the following equation:

$$F_{YS} = 1.45 \times 10^{-16} \exp\left(\frac{32\,562}{T_F + 460}\right) \qquad \text{(Eq 7a)}$$

where T_F is temperature in °F, or

$$F_{YS} = 1.45 \times 10^{-16} \exp\left(\frac{18\,090}{T_K}\right) \qquad \text{(Eq 7b)}$$

where T_K is temperature in K.

In one experiment, lengths of 7050-W (4 days) extrusions were aged at 24 h at 120 °C (250 °F) plus the equivalent of 3 to 42 h at 165 °C (325 °F). For the second step, a logarithmic heatup was used in which 10 h were required for the load to reach 155 °C (315 °F), and nominal soak temperature was 165 °C (325 °F). Figure 24 indicates that yield strength generally agreed with values predicted using Eq 5. The deviation of the curve for short-transverse strength at the short aging times indicates that the method is inadequate for predicting strength on the underaging side of the aging curve.

The effects of neglecting to compensate for soaking at temperatures other than the nominal can be large (Fig 25). For example, the calculated difference in strength between alloy 7050 extrusions soaked 29 h at 160 °C (320 °F) and at 165 °C (325 °F) is about 50 MPa (7 ksi), and the calculated difference in strength between 7050 extrusions soaked 29 h at 155 °C (315 °F) and at 170 °C (335 °F) is about 100 MPa (14 ksi).

Neglecting to compensate for time spent heating the work to the soak temperature will increase the variability. Strength loss attributed to heatup was 14 MPa (2 ksi).

These kinetic relationships also can assist in selection of equivalent aging times for alternate second-step aging temperatures. Equations 5 and 7 can be rearranged
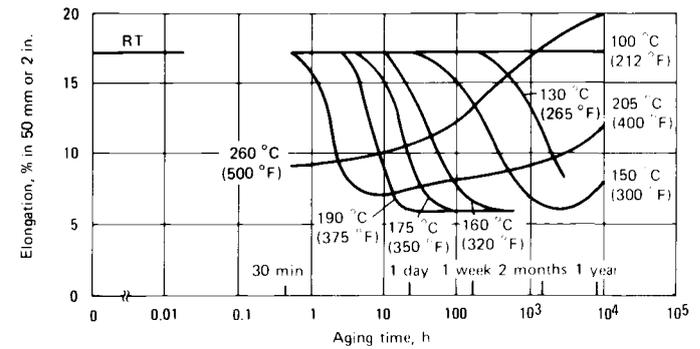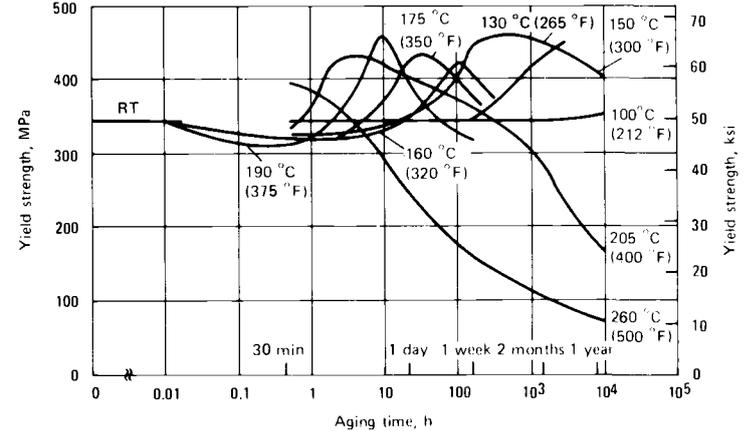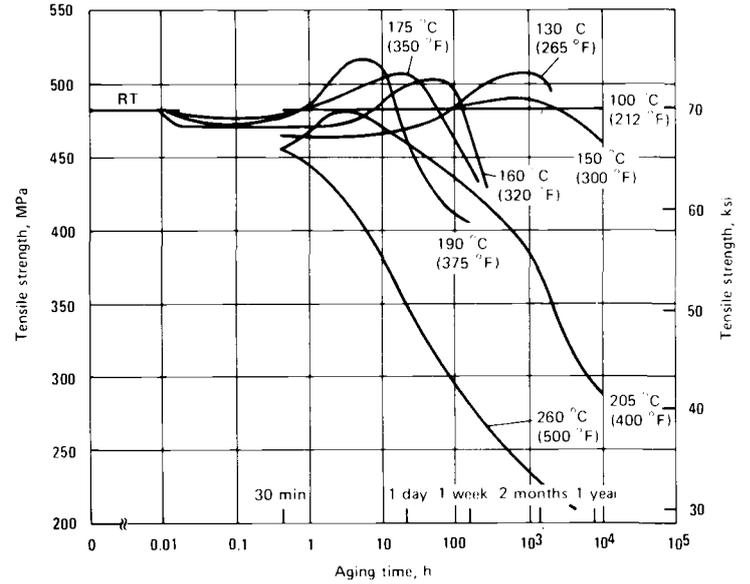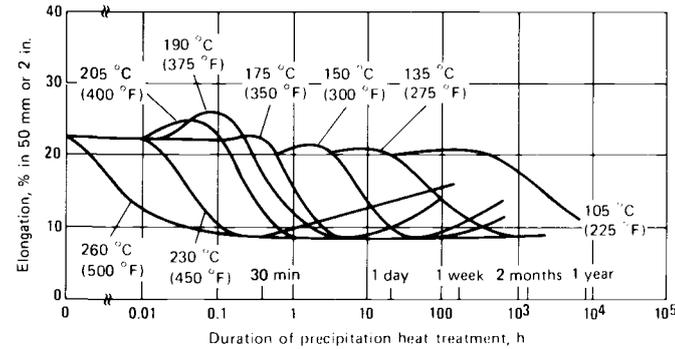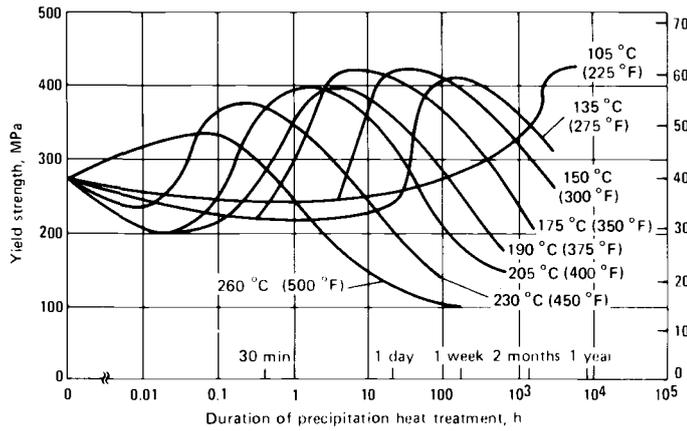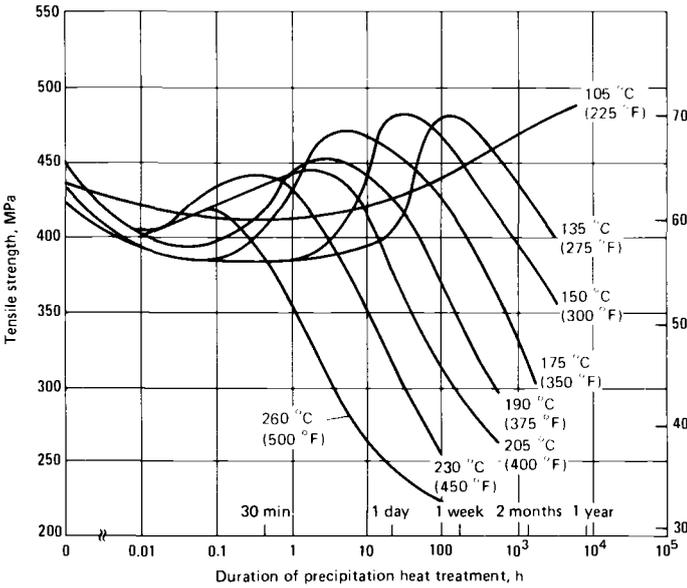
Fig 22(a) Aging characteristics of alloy 2014 sheet

Fig 22(b) Aging characteristics of alloy 2024 sheet (see also Fig 22d)

to yield the following equation:

$$t_2=t_1 \exp \frac{32\,562}{T_1+460} - \frac{32\,562}{T_2+460}$$ (Eq 8)

where t_1 is aging time at temperature T_1, t_2 is aging time at temperature T_2 that will provide equivalent yield strength, and T_1 and T_2 are in °F. For example, the time at 175 °C (350 °F) equivalent to aging alloy 7050 for 29 h at 165 °C (325 °F) is calculated as follows:

$t_{350}=29/\exp\,(1.28)=29/3.6=8$ h

Thermomechanical effects on aging occur from deformation after solution heat treatment. The deformation step may be warm or cold and before, after, or during aging. The simplest thermomechanical practices are those of the conventional T3, T8, or T9 tempers. The rate and extent of precipitation strengthening are distinctly increased in some alloys by cold working after

quenching, whereas other alloys show little or no added strengthening when treated by this sequence of operations.

Alloys of the 2xxx series such as 2014, 2124, and 2219 are particularly responsive to cold work between quenching and aging, and this characteristic is the basis for the higher-strength T8 tempers. The strength improvement accruing from the combination of cold working and precipitation heat treating is a result of nucleation of addition-

Fig 22(c) Aging characteristics of alloy 6061 sheet

type of approach results in strengths similar to those obtained with T8 processing but with the better toughness and fatigue characteristics of T3 products. Alloys 2024, 2124, and 2219 in T8-type tempers are particularly well suited for supersonic and military aircraft; alloy 2219 in such tempers, and alloy 2014-T65, were the principal materials for the fuel and oxidizer tanks (which also served as the primary structure) of the Saturn V space vehicles. Re-solution heat treatment of mill products supplied in these tempers can result in grain growth and in substantially lower strength than is normal for the original temper. Such reheat treatment is not recommended.

Alloys of the 7xxx series do not respond favorably to the sequence of operations used to produce T8-type tempers, and no such tempers are standard for these alloys. The strains associated with stretching or compressing of 7xxx alloys have relatively little effect on the mechanical properties of material precipitation heat treated to T6-type tempers. On the other hand, these operations have measurable detrimental effects on final strength when T73-, T736, or T76-type tempers are produced, particularly in the direction opposite the direction of cold work. Accordingly, specification properties are somewhat lower for the stress-relieved versions of these tempers. Decreasing the overaging time to compensate for the loss in strength is not advisable, because this would impair development of the desired corrosion characteristics.

Temperature control and uniformity present essentially the same problems in precipitation heat treating as they do in solution heat treating.

Good temperature control and uniformity throughout the furnace and load are required for all precipitation heat treating. Recommended temperatures are generally those that are least critical and that can be used with practical time cycles. Except for 7xxx alloys in T7x tempers, these temperatures generally allow some latitude and should have a high probability of meeting property specification requirements. Furnace radiation effects seldom are troublesome except in those few furnaces that are used for both solution and precipitation heat treating. Generally, such situations should be avoided, because the high heat capacity needed for the higher temperatures may be difficult to control at normal aging temperatures.

Soak time in precipitation heat treating is not difficult to control; the specified times carry rather broad tolerances. Heavier loads with parts racked closer together, and even nested, are not abnormal. The principal hazard is undersoaking due to gross excesses in loading practices. Some regions of the load may reach soak temperature long after soak
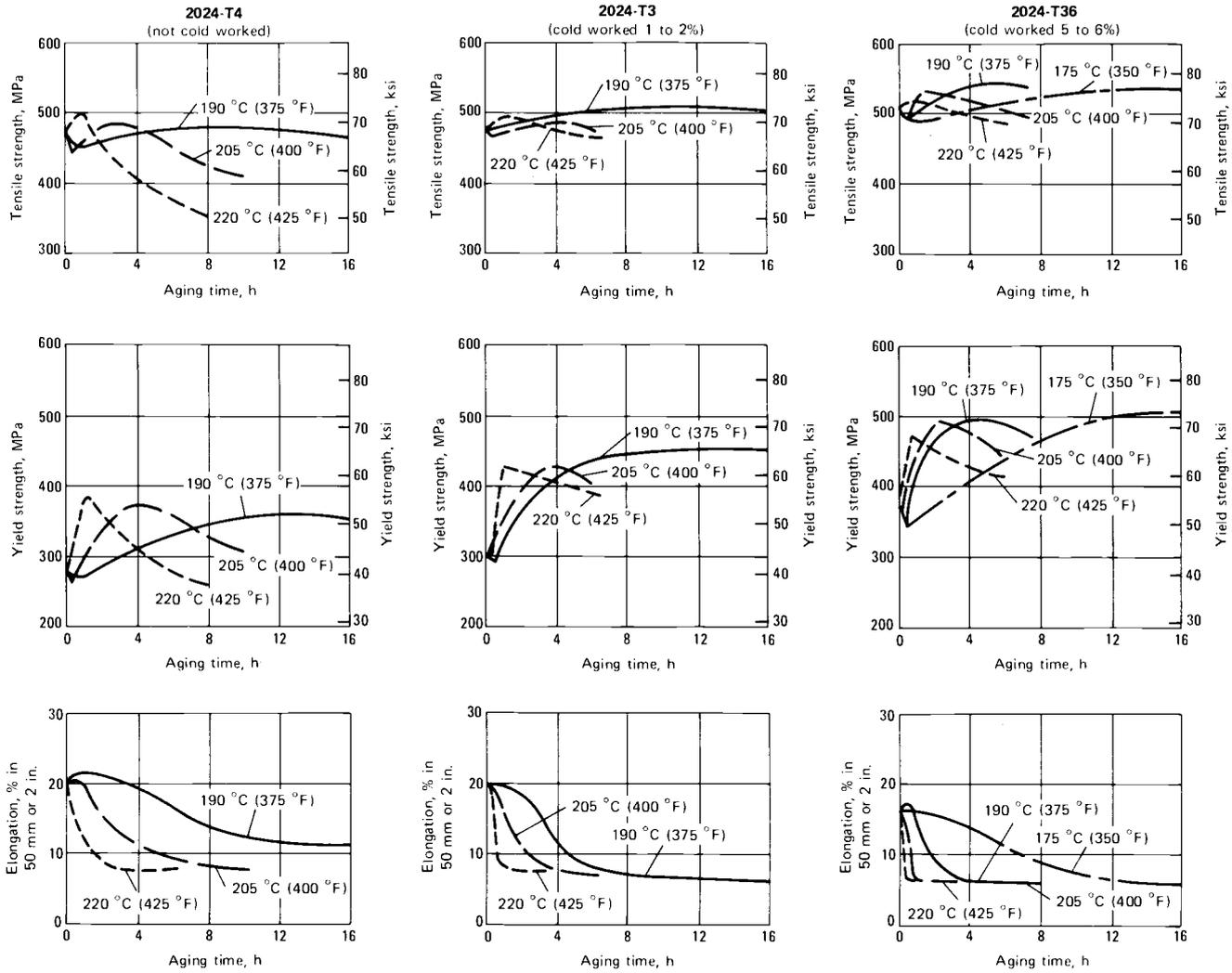
al precipitate particles by the increased strain. In some alloys of the 2xxx series, strain introduced by cold working after solution heat treatment and quenching also induces nucleation of a finer precipitate dispersion that increases strength. Depending on the aging temper, however, toughness may be adversely affected, as illustrated in Fig 26 for 2024 sheet.

Strengthening from thermomechanical processing is the basis for the higher-strength T8-type tempers of alloys 2011, 2024, 2124, 2219, and 2419, which are produced by applying controlled amounts of cold rolling, stretching, or combinations of these operations. Normally, cold work is introduced by stretching; however, other methods such as cold rolling can be used. Recently, 2324-T39 was developed. The T39 temper is obtained by cold rolling approximately 10% after quenching followed by stretching to stress relieve. This

Fig 22(d) Effects of cold work after quenching and before aging on tensile properties of alloy 2024 sheet

time has been called. Placement of load thermocouples is critical, and limiting the size and spacing of a load may be necessary for aging to the T73 and T76 tempers. As discussed above, soak time is not as critical for peak-aged (T6 and T8) tempers.
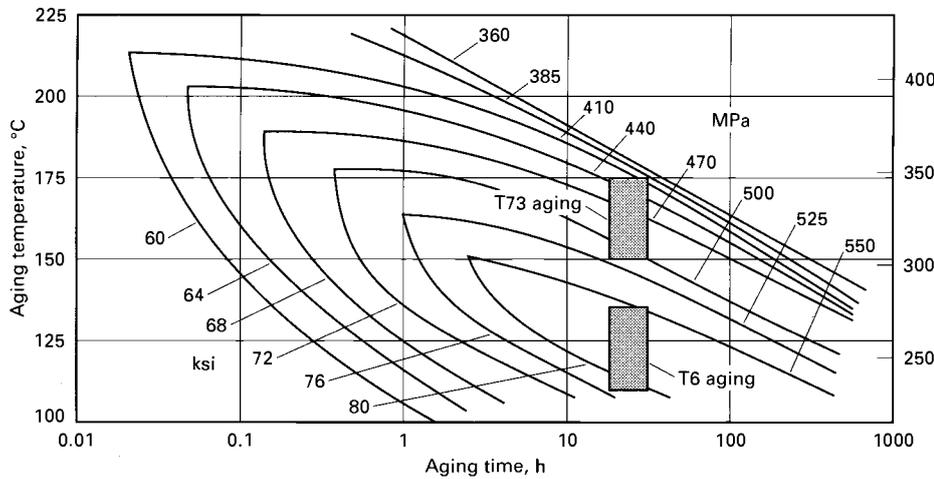
Hardening of Cast Alloys

In general, the principles and procedures for heat treating wrought and cast alloys are similar. The major differences between solution-treating conditions for castings and those for wrought products are found in soak times and quenching media. Solution of the relatively large microconstituents present in castings requires longer soaking

Fig 23 Iso-yield-strength curves for alloy 7075
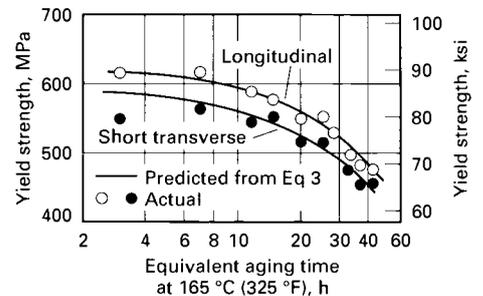
Fig 24 Actual versus predicted yield strengths for alloy 7050 extrusions
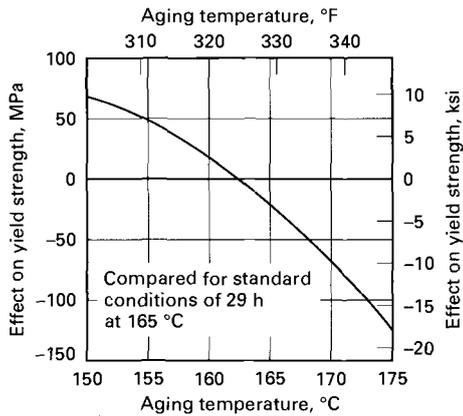
Fig 25 Effect of aging temperature on yield strength of alloy 7050-T736

Fig 26 Effect of stretching and aging on the toughness and yield strength of 2024 sheet

periods than those used for wrought products (Table 3). When heat treatment of castings must be repeated, solution times become similar to those for wrought products, because the gross solution and homogenization has been accomplished and is irreversible under normal conditions. Reduction of stresses and distortion from quenching are also important, because castings generally are complex shapes with variations in section thickness.

Different casting processes and foundry practices also result in microstructural differences with relevance to heat-treatment practice, because the coarser microstructures associated with slow solidification rates require a longer solution heat treatment exposure. Therefore, the time required at temperature to achieve solution is progressively shorter for investment, sand, and permanent mold castings. Foundry practice (chills, gating, type of mold) also plays an important role in the response of a casting, or a portion of a casting, to heat treatment. For example, thin-wall sand castings produced with extensive use of chills can often display finer microstructures than heavy-section permanent mold parts produced in such a way that process advantages are not exploited.

For these reasons, solution heat-treatment practices can be optimized for any specific part to achieve solution with the shortest reasonable cycle once production practice is finalized, even though most foundries and heat treaters will standardize a practice with a large margin of safety. There also exists a fundamental difference between unmodified and modified alloys in which the size and shape of silicon crystals are modified with additions of elements such as calcium, sodium, strontium, or antimony. Modified alloys undergo rapid spheroidization while complete spheroidization is not achieved in unmodified alloys even after very long times. The practical implication is that sh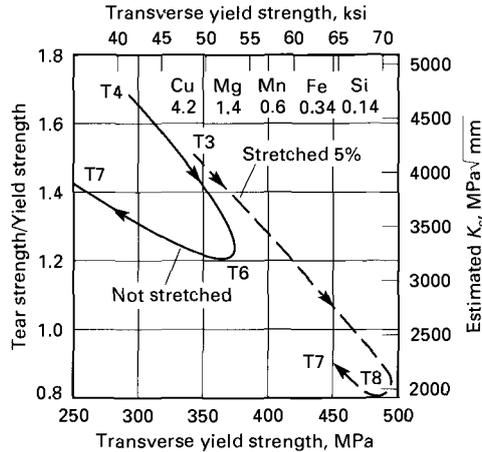orter solution heat treatment could be employed in fully modified castings. The microsegregation of silicon and magnesium is not severe in the aluminum-silicon-magnesium casting alloys, and hence it takes only a short time to homogenize the alloy and to place the Mg_2Si into solution.

Quenchants. Quenching of aluminum castings is often done in boiling water or a milder medium to reduce quenching stresses in complex shapes. A commercially important variety is a mixture of polyalkylene glycol and water, which has no detrimental effect on properties for thicknesses under approximately 3.2 mm (0.125 in.). Quenchant additions can be made for the following purposes:

- To promote stable vapor film boiling by the deposition of compounds on the surface of parts as they are submerged in the quench solution
- To suppress variations in heat flux by increasing vapor film boiling stability through chemically decreased quench solution surface tension
- To moderate quench rate for a given water temperature

The key to the compromise between goals involving property development and the physical consequences of quenching is uniformity of heat extraction, which is in turn a complex function of the operable heat extraction mechanism. Nucleate, vapor film, and convective boiling occur with dramatically different heat extraction rates at different intervals. Differences in section thickness, load density, positioning, racking methods, surface condition, and casting geometry also influence the results.

Property Development. Yield strength is largely controlled by the limiting hardening-element level, and tensile strength (in a general sense) is related to the ductility at a given yield strength. Ductility, however, is controlled for a given yield strength by soundness and microstructural fineness, and is thus determined in the foundry and not by the heat treater. This effect of casting methods on property development is shown in Fig 27. Because of the finer cast structure and higher supersaturation of the more rapidly solidified permanent mold castings, their tensile properties are superior to those of sand castings of the same composition similarly heat treated.

Tempers. Cast products of heat-treatable aluminum alloys have the highest combinations of strength, ductility, and toughness when produced in T6-type tempers. Developing T6-type tempers in cast products requires the same sequence of operations employed in developing tempers of the same type in wrought products—solution heat treating, quenching, and precipitation heat treating. Premium-quality casting specifications such as MIL-A-21180 can require different strengths and ductility levels in the same casting.

Among precipitation treatments unique to castings are those resulting in the T5 and T7 tempers. The T5 temper is produced merely by applying a precipitation treatment to the as-cast casting, without previous solution treatment. A moderate increase in strength is achieved without warpage and subsequent straightening. High hardness and dimensional and strength stability at elevated temperatures account for the almost universal use of materials in T5 tempers for pistons and other engine parts. Some applications demand combinations of strength, toughness, and dimensional stability that cannot be met by heat treating to T5-, T6-, or T8-type tempers. For these applications, T7-type tempers are developed by solution heat treating, quenching in a medium that provides a moderate cooling rate, and then precipitation heat treating at a temperature higher than those used to develop T5-, T6-, and T8-type tempers. Heat treating to T7-type tempers results in lower strength than that of material in T6- or T8-type tempers, develops high ductility and toughness, and carries precipitation far enough to minimize further precipitation during service.

Stress Relief

Immediately after being quenched, most aluminum alloys are nearly as ductile as they are in the annealed condition. Consequently, it is often advantageous to stress relieve parts by working the metal immediately after quenching. Numerous attempts also have been made to develop a thermal treatment that will remove, or appreciably reduce, quenching stresses. Normal precipitation heat-treating temperatures are generally too low to provide appreciable stress relief. Exposure to higher temperatures (at which stresses are relieved more effectively) results in lower properties. However, such treatments are sometimes utilized when even moderate reduction of residual stress levels is important enough so that
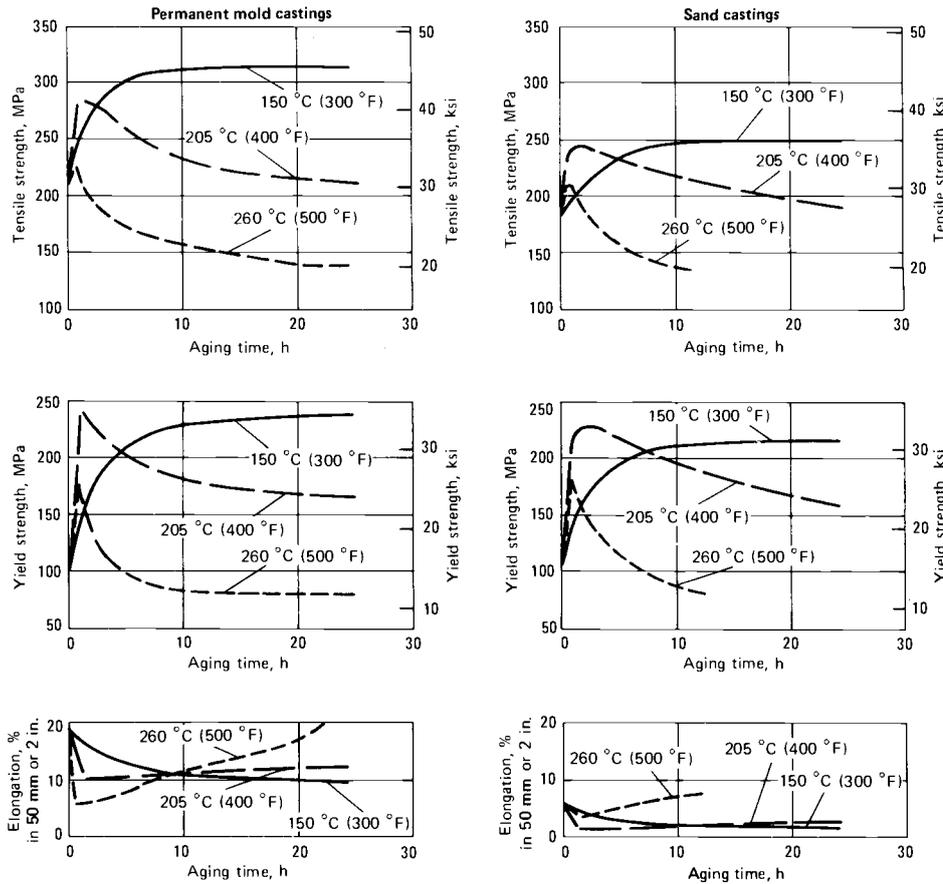
Fig 27 Comparison of the precipitation-hardening characteristics of 356.0-T4 sand and permanent mold castings
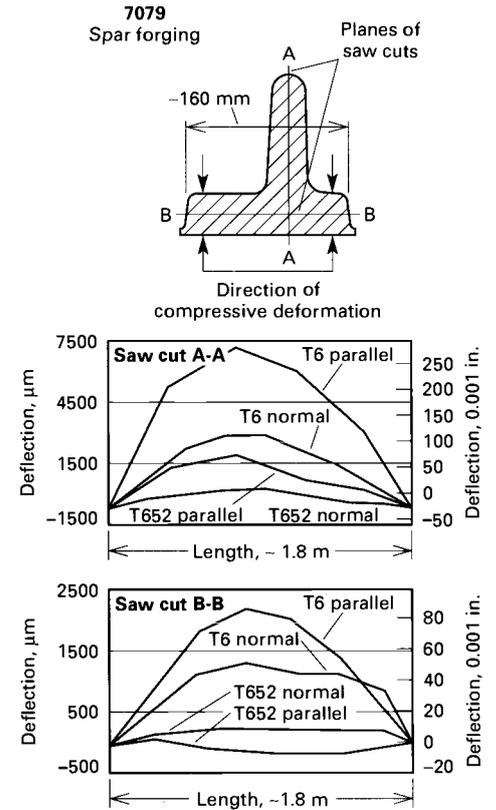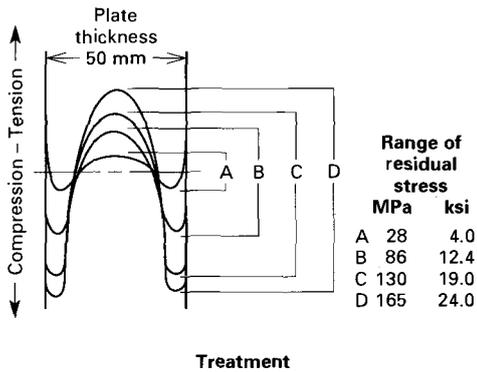

Fig 28 Effect of 3% permanent deformation in compression (T652 treatment) on distribution of stress in a large forging. Parallel and normal refer to warpage directions with respect to the plane of the saw cut.

some sacrifice in mechanical properties can be accepted. The T7 temper for castings is a typical example of this kind of treatment.

Mechanical Stress Relief. Deformation consists of stretching (bar, extrusions, and plate) or compressing (forgings) the product sufficiently to achieve a small but controlled amount (1 to 3%) of plastic deformation. If the benefits of mechanical stress relieving are needed, the user should refrain from reheat treating.

Figure 28 illustrates the beneficial effect of 3% permanent deformation in compression on a large forging.

These methods are most readily adaptable to mill and forge shop products and require equipment of greater capacity than that found in most manufacturing plants. Application of these methods to die forgings and extrusions usually requires construction of special dies and jaws. Stretching generally is limited to material of uniform cross section; however, it has been applied successfully to stepped extrusions and to a 3 by 14 m (10 by 47 ft) aircraft wing skin roll-tapered to a thickness range of 7.1 to 3.2 mm (0.280 to 0.125 in.).

Specific combinations of the supplemental digits are used to denote the tempers produced when mechanical deformation is used primarily to relieve residual stresses

induced during the quenching operation. For products stress relieved by stretching, the digits 51 follow the basic T*x* designation (T451, for example). For products stress relieved by compressive deformation, the supplementary digits are 52.

An additional digit is added to designations for extrusions: an added zero specifies that the product has not been straightened after final stretching; an added one indicates that straightening may have been performed after final stretching.

Effect of Precipitation Heat Treating on Residual Stress. The stresses developed during quenching from solution heat treatment are reduced during subsequent precipitation heat treatment. The degree of relaxation of stresses is highly dependent upon the time and temperature of the precipitation treatment and the alloy composition. In general, the precipitation treatments used to obtain the T6 tempers provide only modest reduction in stresses, ranging from about 10 to 35%. To achieve a substantial lowering of quenching stresses by thermal stress relaxation, higher-temperature treatments of the T7 type are required. These treatments are used when the lower strengths resulting from overaging are acceptable.

Other thermal stress-relief treatments, known as subzero treatment and cold stabi-

lization, involve cycling of parts above and below room temperature. The temperatures chosen are those that can be readily obtained with boiling water and mixtures of dry ice and alcohol—namely, 100 and −73 °C (212 and −100 °F)—and the number of cycles ranges from one to five. The maximum reduction in residual stress that can be effected by these techniques is about 25%. The maximum effect can be obtained only if the subzero step is performed first, and immediately after quenching from the solution-treating temperature while yield strength is low. No benefit is gained from more than one cycle.

A 25% reduction in residual stress is sometimes sufficient to permit fabrication of a part that could not be made without this reduction. However, if a general reduction is needed, as much as 83% relief of residual stress is possible by increasing the severity of the uphill quench—that is, more closely approximating the reverse of the cooling-rate differential during the original quench. This may be accomplished by a patented process that involves extending the subzero step to −195 °C (−320 °F) and then very rapidly uphill quenching in a blast of live steam (Fig 29). The rate of reheating is extremely critical, and therefore, to ensure proper application of the steam blast, a

Fig 29 Effectiveness of various uphill quenching treatments in reducing residual quenching stresses in 2014 plate. Note: uphill quenching treatments (single-cycle only) were applied from ½ to 1½ h after quenching from the recommended solution-treating temperature. All specimens were aged to the T6 temper after uphill quenching.

Treatment

A: Cooled to -195 °C, then uphill quenched in a steam blast
B: Cooled to -75 ° C, then uphill quenched in a steam blast
C: Cooled to -75 or -195 ° C, then uphill quenched in boiling water
D: Standard specimen, quenched and aged to T6 temper in conventional manner with no further treatment

Range of residual stress

	MPa	ksi
A	28	4.0
B	86	12.4
C	130	19.0
D	165	24.0

special fixture usually is required for each part.

This process will not solve all problems of warpage in machining. It may reduce warpage internally but increase warpage of the extreme outer layers, although in the opposite direction (Fig 30). Also, the effect of the altered residual-stress pattern on performance must be evaluated carefully for each part. This is particularly important for parts subjected to cyclic loading or exposed to corrosive environments such as marine atmospheres, especially if the process is introduced after the start of production and original performance tests are not repeated. Further disadvantages are the cost and hazard involved in handling liquid nitrogen and live steam.

Effects of Reheating

The precipitation characteristics of aluminum alloys must be considered frequently during evaluation of the effects of reheating on mechanical properties and corrosion resistance. Such evaluations are necessary for determining standard practices for manufacturing operations, such as hot forming and straightening, adhesive bonding, and paint and dry-film lubricant curing, and for evaluating the effects of both short-term and long-term exposure in elevated temperatures in service.

The stage of precipitation that exists in an alloy at the time of reheating plays a significant role in the effects of reheating. Consequently, it is extremely dangerous to reheat material in a solution heat-treated

Table 9 Reheating schedules for wrought aluminum alloys

The schedules given in this table normally will not decrease strength more than 5%.

Alloy and temper	Reheating time at a temperature of:						
	150 °C (300 °F)	165 °C (325 °F)	175 °C (350 °F)	190 °C (375 °F)	205 °C (400 °F)	220 °C (425 °F)	230 °C (450 °F)
2014-T4	(a)	(a)	(a)	(a)	(a)	(a)	(a)
2014-T6	2–50 h	8–10 h	2–4 h	½–1 h	5–15 min	(b)	(b)
2024-T3, 2024-T4	(a)	(a)	(a)	(a)	(a)	(a)	(a)
2024-T81, 2024-T86	20–40 h	· · ·	2–4 h	1 h	½ h	15 min	5 min
6061-T6, 6062-T6, 6063-T6	100–200 h	50–100 h	8–10 h	1–2 h	½ h	15 min	5 min
7075-T6, 7178-T6	10–12 h	1–2 h	1–2 h	½–1 h	5–10 min	(b)	(a)

(a) Reheating not recommended. (b) Bring to temperature

temper without first carefully testing the effects of such reheating. In one such test, 2024-T4 sheet was found to be very susceptible to intergranular corrosion when subjected to a 15-min drying operation at 150 °C (300 °F) during the first 8 h after quenching; no susceptibility was evident when the same drying operation was performed more than 16 h after quenching. In another test, 7075-W (0.2 to 600 h) bar and plate were reheated for hot forming at 175 °C (350 °F) for 20 min. Strengths after aging to the T6 temper were 10 to 15% lower than those for standard 7075-T6. In contrast, similar reheating of T6 material for up to 1 h at 175 °C (350 °F) produced no detrimental effect.

If reheating is performed on material in the W or T4 condition, its effect can be estimated from families of precipitation heat-treating curves such as those presented in Fig 22. Such curves can also be used for reheating of precipitation heat-treated material at the precipitation heat-treating temperature. For reheating at other temperatures, other data may be needed (Fig 31). The heat-treating and reheating curves may be used as the bases for limitations on reheating (Table 9).

Annealing

Annealing treatments employed for aluminum alloys are of several types that differ in objective. Annealing times and temperatures depend on alloy type as well as on initial structure and temper.

Full Annealing. The softest, most ductile, and most workable condition of both non-heat-treatable and heat-treatable wrought alloys is produced by full annealing to the temper designated "O." Strain-hardened products in this temper normally become recrystallized, but hot-worked products may remain unrecrystallized. In the case of heat-treatable alloys, the solutes are sufficiently thoroughly precipitated to prevent natural age hardening. A higher maximum temperature than that used for stress-relief annealing, controlled cooling to a lower temperature, and additional holding time at the lower temperature generally are employed.

For both heat-treatable and non-heat-treatable aluminum alloys, reduction or elimination of the strengthening effects of cold working is accomplished by heating at a temperature from about 260 to about 440 °C (500 to 825 °F). The rate of softening is strongly temperature-dependent; the time required to soften a given material by a given amount can vary from hours at low temperatures to seconds at high temperatures.

If the purpose of annealing is merely to remove the effects of strain hardening, heating to about 345 °C (650 °F) will usually suffice. If it is necessary to remove the hardening effects of a heat treatment or of cooling from hot-working temperatures, a treatment designed to produce a coarse, widely spaced precipitate is employed. This usually consists of soaking at 415 to 440 °C (775 to 825 °F) followed by slow cooling (28 °C/h, or 50 °F/h, max) to about 260 °C (500 °F). The high diffusion rates that exist during soaking and slow cooling permit maximum coalescence of precipitate particles and result in minimum hardness.

As a result of this treatment, only partial precipitation occurs in 7xxx alloys, and a second treatment (soaking at 230 ± 6 °C, or 450 ± 10 °F, for 2 h) is required. When the need arises for small additional improvements in formability, cooling at 28 °C/h (50 °F/h) should be extended to 230 °C (450 °F), and the material should be soaked at 230 °C for 6 h. The effects of eliminating or prolonging the 230 °C second step on the ductility of 7075-O sheet are compared with the standard treatment in Table 10.

In annealing, it is important to ensure that the proper temperature is reached in all portions of the load; therefore, it is common to specify a soaking period of at least 1 h. The maximum annealing temperature is moderately critical; it is advisable not to exceed 415 °C (775 °F), because of oxidation and grain growth. The heating rate can be critical, especially for alloy 3003, which usually requires rapid heating for prevention of grain growth. Relatively slow cooling, in still air or in the furnace, is recommended for all alloys to minimize distortion. Typical annealing conditions used for some alloys in common use are listed in Table 11.
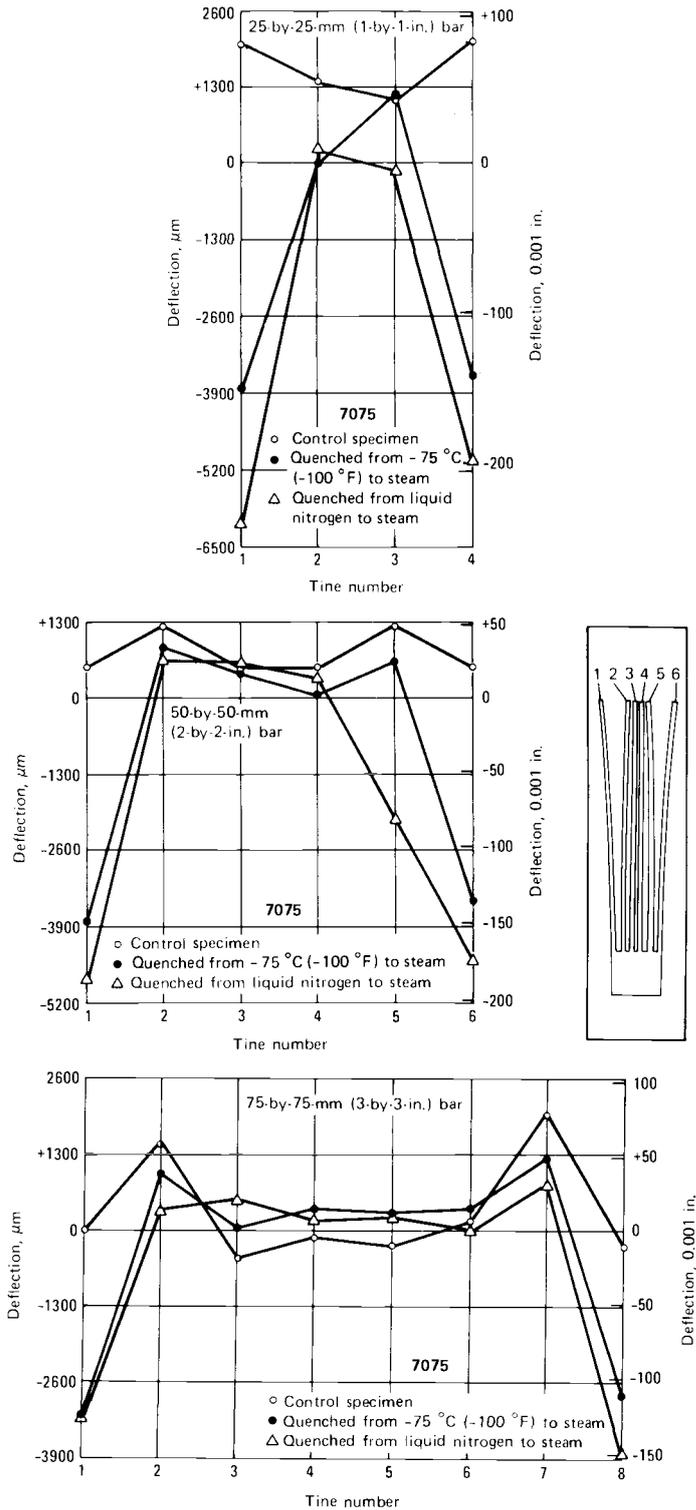
Fig 31 Effects of reheating on tensile properties of alclad 2024-T81 sheet

Fig 30 Effect of uphill quenching on deflection of tines. Six-tine specimen was machined from 50 by 50 mm (2 by 2 in.) bar. Similar specimens machined from 25 by 25 mm (1 by 1 in.) and 75 by 75 mm (3 by 3 in.) bars had four and eight tines, respectively.

Products that can be heated and cooled very rapidly, such as wire, are annealed by continuous processes that require a total heating and cooling time of only a few seconds. Continuous annealing of coiled sheet is accomplished in a total time of a few minutes. For these extremely rapid operations, maximum temperature may exceed 440 °C (825 °F).

Although material annealed from the precipitation-hardened condition usually has sufficient ductility for most forming opera-
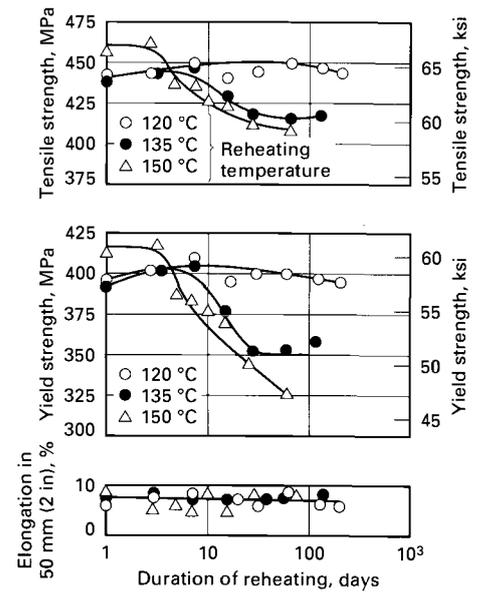
tions, this ductility often is slightly lower than that of material that has not been subjected to prior heat treatment—that is, material annealed at the producing source. Therefore, when maximum ductility is required, annealing of a previously heat-treated product is sometimes unsuccessful.

Partial Annealing. Annealing of cold-worked non-heat-treatable wrought alloys to obtain intermediate mechanical properties (H2-type tempers) is referred to as partial annealing or recovery annealing. Temperatures used are below those that produce extensive recrystallization, and incomplete softening is accomplished by substructural changes in dislocation density and rearrangement into cellular patterns (polygonization). Bendability and formability of an alloy annealed to an H2-type temper generally are significantly higher than those of the same alloy in which an equal strength level is developed by a final cold-working operation (H1-type temper). Treatments to produce H2-type tempers require close control of temperature to achieve uniform and consistent mechanical properties.

Figure 32 shows changes in yield strength as functions of temperature and time for sheet of two non-heat-treatable alloys (1100 and 5052) initially in the highly cold-worked condition (H18 temper). From these curves, it is apparent that, by selection of appropriate combinations of time and temperature, mechanical properties intermediate to those of cold-worked and fully annealed material can be obtained. It is also evident that yield strength depends much more strongly on temperature than on time of heating.

Stress-Relief Annealing. For cold-worked wrought alloys, annealing merely to remove the effects of strain hardening is referred to as stress-relief annealing. Such treatments

Table 10 Effects of annealing treatments on ductility of 7075-O sheet

Annealing treatment	Elongation in tension(a), % in 50 mm (2 in.) for thickness of:			Bend angle(b), degrees, for thickness of:		Elongation in bending(c), % in 50 mm (2 in.) for thickness of:	
	0.5 mm (0.020 in.)	1.6 mm (0.064 in.)	2.6 mm (0.102 in.)	1.6 mm (0.064 in.)	2.6 mm (0.102 in.)	1.6 mm (0.064 in.)	2.6 mm (0.102 in.)
Treatment 1(d)	12	12	12	82	73	48	50
Treatment 2(e)	14	14	14	91	76	58	57
Treatment 3(f)	16	16	···	92.5	84	56	60

(a) Uniform elongation of gridded tension specimens. (b) Bend angle at first fracture. (c) Elongation in bend test for 1.3 mm (0.05 in.) gage spanning fracture. (d) Soak 2 h at 415 ± 14 °C (775 ± 25 °F); furnace cool to 260 °C (500 °F) at 30 °C/h (50 °F/h); air cool. (e) Soak 2 h at 425 °C (800 °F), air cool; soak 2 h at 230 °C (450 °F), air cool. (f) Soak 1 h at 425 °C (800 °F); furnace cool to 230 °C (450 °F) at 30 °C/h (50 °F/h); soak 6 h at 230 °C (450 °F), air cool

employ temperatures up to about 345 °C (650 °F), or up to 400 ± 8 °C (750 ± 15 °F) for 3003 alloy, and cooling to room temperature. No appreciable holding time is required. Such treatment may result in simple recovery, partial recrystallization, or full recrystallization. Age hardening may follow stress-relief annealing of heat-treatable alloys, however, because a concentration of soluble alloying elements sufficient to cause natural aging remains in solid solution after such treatments.

A special form of stress-relief temper is used for heat-treatable alloy products that subsequently will be inspected ultrasonically. The product is heated to its normal solution heat-treating temperature, then cooled in still air to room temperature. This temper is referred to as the O1 temper.

Controlled-Atmosphere Annealing and Stabilizing. Aluminum alloys that contain even very small amounts of magnesium will form a surface magnesium oxide unless the atmosphere in the annealing furnace is free of moisture and oxygen. Examples include alloy 3004, which is used for cooking utensils, and alloys of the 5xxx series.

Another problem that control of the annealing atmosphere helps to overcome or avoid is oil staining by oil-base roll lubricants that do not burn off at lower annealing temperatures. If the oxygen content of the furnace atmosphere is kept very low during such annealing, the oil will not oxidize and stain the work.

Temperature control for full and partial annealing is somewhat more critical than for stress-relief annealing; the temperatures and times specified are selected to produce recrystallization and, in the case of heat-treatable alloys, a precipitate of maximum size; for this the cooling rate must be close-

ly controlled. Even allowing the load to cool in the furnace may result in an excessively high rate. Similarly, lowering the furnace-control instrument by 28 °C (50 °F) each hour may produce stepped cooling, which is not satisfactory for severe forming operations. For maximum softening, a continuous cooling rate of not more than 28 °C/h (50 °F/h) is recommended.

Annealing of castings for 2 to 4 h at temperatures from 315 to 345 °C (600 to 650 °F) provides the most complete relief of residual stresses and precipitation of the phases formed by the excess solute retained in solid solution in the as-cast condition. Such annealing treatments provide maximum dimensional stability for service at elevated temperatures. The annealed temper is designated "O." (This temper was designated "T2" prior to 1975.)

Grain Growth

Many of the aluminum alloys in common use are subject to grain growth during solution treatment or annealing. This phenomenon can occur during or after recrystallization of material that has been subjected to a small critical amount of prior cold work. It is usually manifested by surface roughening during subsequent fabrication operations and frequently results in rejections for appearance or functional reasons. Less frequently, some deterioration of mechanical properties is encountered, and this is undesirable regardless of surface-roughening effects.

Degree of susceptibility to grain growth varies with alloy, structure, and chemical-composition variation, and from one product form to another. The critical range of cold work is ordinarily about 5 to 15%. Usually, temperatures of 400 °C (750 °F) and above must be reached before grain growth occurs, but some growth has been encountered at temperatures as low as 345 °C (650 °F). Grain growth that occurs during initial recrystallization is more a function of composition, structure, and degree of cold work than of temperature per se; temperatures in excess of 455 °C (850 °F) in common alloys can lead to secondary-recrystallization grain-growth problems. The common symptom indicating moderately large-grain material is roughening or "orange peel" on the external surfaces of bends. Severe growth of grains to fingernail size and larger sometimes is evident in parts made from annealed (O temper) material by stretch forming and then thermal treating or similar operations. This type of grain growth often is detected during subsequent anodizing, etching, and chemical milling operations.

Cracking during welding or brazing is another characteristic which may indicate that severe grain growth has occurred. In such instances, cracks propagate along grain boundaries that provide little obstruction to their progress.

Table 11 Typical full annealing treatments for some common wrought aluminum alloys

These treatments, which anneal the material to the O temper, are typical for various sizes and methods of manufacture and may not exactly describe optimum treatments for specific items.

Alloy	Metal temperature °C	°F	Approximate time at temperature, h	Alloy	Metal temperature °C	°F	Approximate time at temperature, h
1060	345	650	(a)	5457	345	650	(a)
1100	345	650	(a)	5652	345	650	(a)
1350	345	650	(a)	6005	415(b)	775(b)	2–3
2014	415(b)	775(b)	2–3	6009	415(b)	775(b)	2–3
2017	415(b)	775(b)	2–3	6010	415(b)	775(b)	2–3
2024	415(b)	775(b)	2–3	6053	415(b)	775(b)	2–3
2036	385(b)	725(b)	2–3	6061	415(b)	775(b)	2–3
2117	415(b)	775(b)	2–3	6063	415(b)	775(b)	2–3
2124	415(b)	775(b)	2–3	6066	415(b)	775(b)	2–3
2219	415(b)	775(b)	2–3	7001	415(c)	775(c)	2–3
3003	415	775	(a)	7005	345(d)	650(d)	2–3
3004	345	650	(a)	7049	415(c)	775(c)	2–3
3105	345	650	(a)	7050	415(c)	775(c)	2–3
5005	345	650	(a)	7075	415(c)	775(c)	2–3
5050	345	650	(a)	7079	415(c)	775(c)	2–3
5052	345	650	(a)	7178	415(c)	775(c)	2–3
5056	345	650	(a)	7475	415(c)	775(c)	2–3
5083	345	650	(a)				
5086	345	650	(a)	**Brazing sheet**			
5154	345	650	(a)				
5182	345	650	(a)				
5254	345	650	(a)	No. 11 and 12	345	650	(a)
5454	345	650	(a)	No. 21 and 22	345	650	(a)
5456	345	650	(a)	No. 23 and 24	345	650	(a)

(a) Time in the furnace need not be longer than necessary to bring all parts of the load to appealing temperature. Cooling rate is unimportant. (b) These treatments are intended to remove the effects of solution treatment and include cooling at a rate of about 30 °C/h (50 °F/h) from the annealing temperature to 260 °C (500 °F). Rate of subsequent cooling is unimportant. Treatment at 345 °C (650 °F), followed by uncontrolled cooling, may be used to remove the effects of cold work or to partly remove the effects of heat treatment. (c) These treatments are intended to remove the effects of solution treatment and include cooling at an uncontrolled rate to 205 °C (400 °F) or less, followed by reheating to 230 °C (450 °F) for 4 h. Treatment at 345 °C (650 °F), followed by uncontrolled cooling, may be used to remove the effects of cold work or to partly remove the effects of heat treatment. (d) Cooling rate to 205 °C (400 °F) or below is less than or equal to 30 °C/h (50 °F/h).
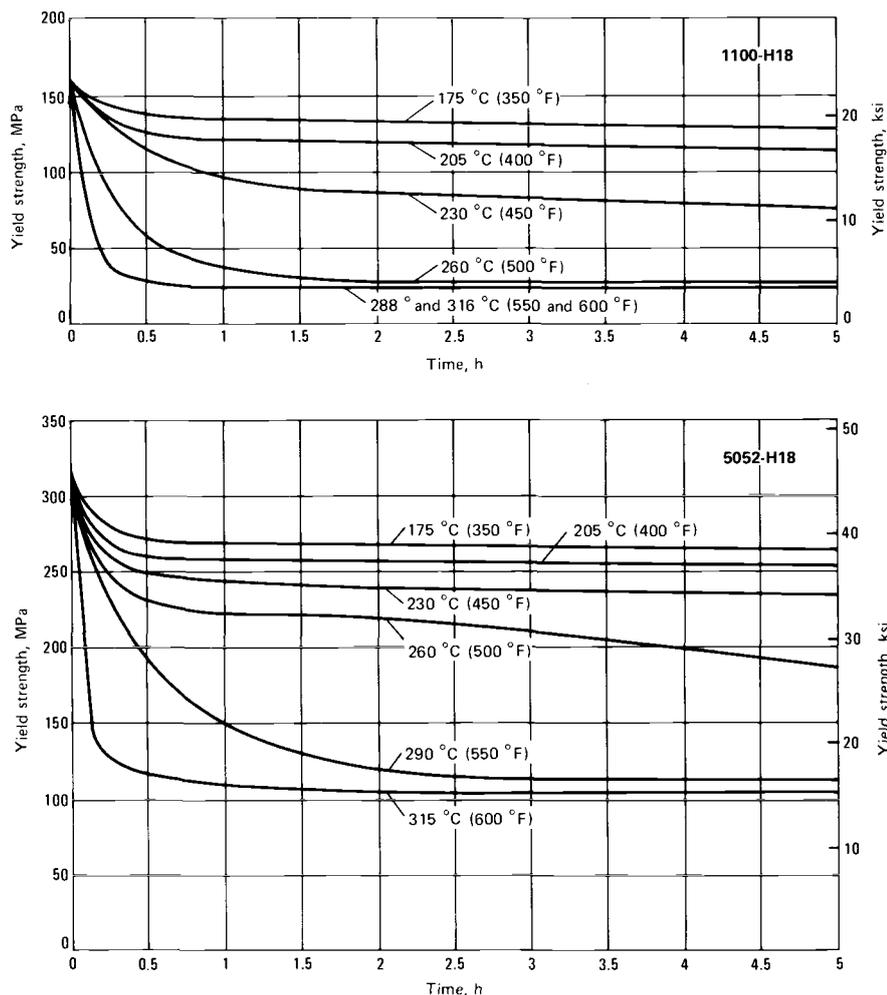
Fig 32 Representative isothermal annealing curves for alloys 1100-H18 and 5052-H18

If the surface roughening is objectionable from either an appearance or a functional aspect, the desirability of surface-smoothing operations, such as sanding or buffing, must be evaluated. If reductions in mechanical properties are suspected, these must be established by test and evaluated in relation to the anticipated service.

In one application, a part that had been made by stretch forming O-temper 2 mm (0.080 in.) sheet and heat treating exhibited significantly lower tensile and yield strengths in portions where severe grain growth had occurred than in portions having normal grain size:

Test	Grain structure	Tensile strength		Yield strength	
		MPa	ksi	MPa	ksi
Transverse					
1	Coarse	265	38.5	247	35.8
2	Coarse	263	38.2	241	35.0
3	Fine	311	45.1	261	37.8
Longitudinal					
1	Coarse	259	37.6	243	35.3
2	Coarse	269	39.0	245	35.6
3	Fine	305	44.2	270	39.1

In other similar investigations, no detrimental effects have been discovered, and in many cases such parts have served satisfactorily in critical applications.

When a grain-growth problem is discovered, it is too late to change the condition of the parts in question, but several possible methods are available for preventing recurrence of the difficulty. The simplest of these is relieving the causative stress by interjecting a stress-relief anneal into the manufacturing sequence immediately prior to the solution-treating or full-annealing cycle in which the grain growth occurred. This approach is usually successful and practical. Another possibility is to adjust the amount of stress present in the part immediately prior to the critical heat treatment so that the stress level is outside the critical range. This may be done by adding a cold-working operation before forming, such as pre-stretching of blanks, or by forming in multiple stages with a stress-relief anneal before each stage.

A third method that is sometimes successful consists of increasing the heating rate during the critical heat treatment by

reducing the size of furnace loads or by changing from an air furnace to a salt bath. In one application, severe grain growth was found during bending of alloy 1100 rectangular tubing. The roughening of the inside surfaces of the parts, which occurred during forming of the large-grain material, impaired their functioning as radar waveguides. Investigation disclosed that, to minimize handling marks, the material was procured in the strain-hardened (H14) temper and was stress-relief annealed at 345 °C (650 °F) immediately prior to forming. Grain growth occurred during annealing as a result of the moderate amount of cold work introduced at the mill. The problem was eliminated by changing the stress-relieving operation to a 5-min heating period in an air furnace operating at 540 °C (1000 °F). The explanation advanced for the success of this treatment was that, due to the rapid heating rate, the temperature of the material was raised through the recrystallization range for the less severely cold-worked grains before the critically cold-worked grains had time to grow appreciably.

Heating Equipment and Accessories

The general methods for heat treating aluminum alloys include the use of molten salt baths, air-chamber furnaces, and induction heaters. The choice of heating equipment depends largely on the alloy and the configuration of the parts to be processed. The type of heat treatment can also influence the choice of heating equipment. For example, both molten salt baths and air-chamber furnaces are suitable for solution treating of aluminum alloys, while induction heating requires additional analysis to define the proper temperature range for solution treatment. Induction methods can provide high heating rates, which affect transformation behavior (see, for example, the section "Nonequilibrium Melting" in this article).

Molten salt baths and air-chamber furnaces both have advantages and disadvantages in solution heat treatments, as discussed below. Oil- and gas-fired furnaces, in designs that allow the products of combustion to come in contact with the work, are usually unsatisfactory because they promote high-temperature oxidation.

Salt baths heat the work faster (see Table 2) than air furnaces, provided that the amount of work introduced at any one time is controlled to prevent the temperature from falling below the desired range. If the temperature is permitted to fall below the minimum limit, much of the advantage of the salt bath is lost, because of the necessity for reheating the large mass of salt.

Salt baths are also more readily adapted to the introduction, at any time, of small amounts of work requiring different soaking

Table 12 Frequency selection for induction heating with a longitudinal-flux coil and a transverse-flux coil

Material	Minimum part thickness, mm (in.), for a frequency of:				
	60 Hz	200 Hz	1 kHz	3 kHz	10 kHz
Solenoid (longitudinal-flux) coil					
Steel below Curie temperature	>38 (1.5)	13 (0.5)	5 (0.2)	2.3 (0.09)	1 (0.04)
Steel above Curie temperature	>175 (7.0)	100 (4.0)	43 (1.7)	25 (1.0)	13 (0.5)
Brass	>50 (2.0)	28 (1.1)	13 (0.5)	7 (0.28)	4 (0.16)
Aluminum	>38 (1.5)	22 (0.85)	9 (0.375)	5 (0.2)	3 (0.12)
Transverse-flux coil					
Aluminum	>5 (0.2)	1.3 (0.05)	0.25 (0.01)	0.08 (0.003)	0.04 (0.0016)
Brass	>10 (0.4)	2.5 (0.1)	0.5 (0.02)	0.15 (0.006)	0.08 (0.0032)
Steel	>50 (2.0)	13 (0.5)	2.5 (0.1)	0.9 (0.035)	0.5 (0.020)

Fig 33 Power requirement for transverse-flux induction heating of aluminum strip 1 mm (0.04 in.) thick and 1270 mm (50 in.) wide. Source: Ref 9

periods. (Economical utilization of air furnaces usually dictates accumulation of a large load of parts of similar thickness before charging.) Also, the buoyant effect of the salt reduces distortion during heating, and the large reservoir of heat facilitates temperature control and uniformity.

Salt bath operation entails special housekeeping requirements. Dragout is costly and unsightly. Because residual salt on parts may result in corrosion, all salt must be completely removed, including that from crevices and blind holes. In addition, salt residue from the quench water must be kept to a minimum by a constant water overflow or by providing a fresh-water rinse for all parts after quenching. When these provisions are impractical, corrosion can be inhibited by adding 14 g (½ oz) of sodium or potassium dichromate to each 45 kg (100 lb) of the molten salt.

Precautions. Molten salt baths are potentially hazardous and require special precautions. Operators must be protected from splashing and dripping of the hot salt. Because heated nitrates are powerful oxidizing agents, they must never by allowed to come in contact with combustibles and reducing agents, such as magnesium and cyanides. Most authorities advise against inserting aluminum alloys containing more than a few percent of magnesium into molten nitrate. To avoid exposure of personnel to nitrous fumes produced during decomposition of nitrates, good ventilation is essential.

When molten nitrates are being used, the possibilities of explosions resulting from both physical and chemical reactions must be avoided. The former result from rapid expansion of gases entrapped beneath the surface of the bath. Hence, parts entering the bath must be clean and dry; they must also be free of pockets or cavities that contain air or other gases. Chemical-reaction explosions result from rapid breakdown of the nitrates due to overheating or reaction with the pot material. Stainless steel pots (preferably of type 321 or 347) are more resistant to scaling than those made of carbon steel or cast iron and therefore present a lower probability of local overheating. Sludge or sediment accumulations in

bottom-heated pots can also lead to local overheating. Overheat controls are essential to ensure against temperatures exceeding 595 °C (1100 °F).
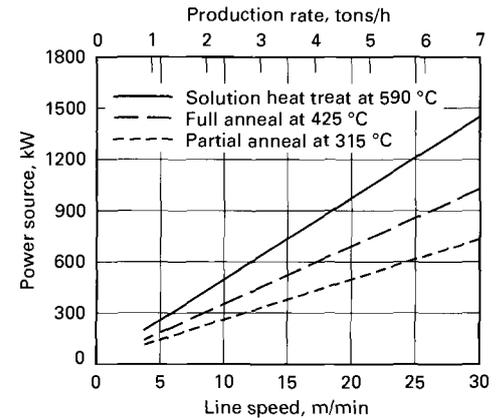
It is vitally important that water be kept away from a nitrate tank. In controlling a nitrate fire, do not use water or any fire extinguisher containing water. The best extinguisher is dry sand, a supply of which should be kept near the tank.

Extra sacks of salt should be stored in a dry place, distant from the tank. If the fresh salt being added to the bath is even slightly damp, it should be added very slowly or when the bath is frozen.

Air furnaces are used more widely than salt baths because they permit greater flexibility in operating temperature. When production schedules and the variety of alloys requiring heat treatment necessitate frequent changes in temperature, the time and cost of adjusting the temperature of a large mass of salt makes the use of an air furnace almost mandatory. However, waiting periods are often required to allow the walls of air furnaces to stabilize at the new temperature before parts are introduced. Otherwise, parts may radiate heat to colder walls or absorb radiant heat from hotter walls, and the temperature indicated by the control instrument will not reflect actual metal temperature in the usual manner. Air furnaces are also more economical when the product mix includes a few rather large parts; holding the temperature of a large volume of salt in readiness for an occasional large part is far more expensive than heating an equal volume of air.

Induction heating with either solenoid (longitudinal-flux) coils or transverse-flux coils provides an efficient method for in-line heating of flat-rolled products (particularly strip). Solenoid coils create a longitudinal flux, which can produce efficient heating for thicker and/or lower resistivity materials. Solenoid coils can also be used efficiently in the heating of thinner magnetic material (see, for example, steel below the Curie temperature in Table 12).

For several nonferrous materials (aluminum, copper, brass), however, efficiency and power factors with solenoid coils are significantly lower than for ferrous materi-

als. Therefore, transverse-flux coils are ideally suited for heating nonferrous materials, because transverse-flux lines do not exhibit the degree of current cancellation associated with longitudinal flux lines. This aspect of transverse-flux coils improves efficiency and also permits the use of lower frequencies (Table 12). This reduces the capital equipment costs, and where it shifts from requiring RF frequencies, the power source conversion efficiency is also significantly improved. Aluminum, brass, copper, and austenitic stainless steel strip lines are ideally suited for transverse-flux heating. Each of these materials often requires in-line processes like partial or full annealing or solution heat treating to provide necessary mechanical properties for subsequent finishing operations.

Transverse-flux induction heating offers several benefits for in-line strip heating and has been known for many years. However, it requires specially designed iron-cored laminated inductor coils and tighter control of the power, strip handling, and process parameters. Frequency selection is dictated by the resistivity and thickness of the material, while power requirements depend on the production rates, the specific heat, and the processing temperatures for a given material. Figure 33 shows typical power source requirements for transverse-flux heating of aluminum strip.

Furnace Temperature Control

The importance of close temperature control in solution treating has been noted in the previous section on solution treating. Each control zone of each furnace should contain at least two thermocouples. One thermocouple, with its instrument, should act as a controller, regulating the heat input; the other should act independently as a safety cutoff, requiring manual reset if its set temperature (usually the maximum of

the specified range) is exceeded during the solution-treating cycle.

Safety cutoffs are mandatory for salt baths to guard against explosions and often have paid for themselves in air furnaces by saving a load of parts or even the furnace itself. It is important, however, that they be tested periodically (by deliberately overshooting the empty furnace) to guard against "frozen" corroded contacts resulting from prolonged periods of idleness.

At least one of the instruments for each zone should be of the recording type, and both instruments should have restricted scales—for instance, 400 to 600 °C (750 to 1110 °F), rather than 0 to 600 °C (32 to 1110 °F). This is required for maximum accuracy because manufacturers' guarantees are specified in terms of percent of scale.

In the placement of instruments, exposure to extremes in ambient temperature, humidity, vibration, dust, and corrosive fumes should be avoided. Ambient temperatures between 5 and 50 °C (40 and 120 °F) are satisfactory, but temperature changes of 6 °C/h (10 °F/h) or more should be avoided. It is also essential that instruments and thermocouple circuits be shielded from electromagnetic fields commonly associated with the leads of high-amperage furnace heating elements.

Temperature-sensing elements must be capable of responding more rapidly to temperature changes than the materials being processed. Therefore, thermocouple wire diameter should not exceed 1½ times the thickness of the minimum-gage material to be heat treated, and should in no case exceed 14 gage. Thermocouples for salt baths should be enclosed in suitable protection tubes. Air-furnace thermocouples should be installed in open-end protection tubes, with the thermocouple junction extending sufficiently beyond the tube to prevent any loss in sensitivity.

Temperature-sensing elements should be located in the furnace work chamber, not in ducts and plenums, and should be as close as possible to the working zone. Specification MIL-H-6088C restricts distance between the sensing element and the working zone to a maximum of 100 mm (4 in.). The safety-cutoff thermocouple should be located to reflect the highest temperature in the working zone. The control thermocouple should be located in a position where it will read a temperature approximately halfway between the hottest and coldest temperatures.

Probe Checks. After the temperature-measurement equipment is properly installed, it must be checked frequently for accuracy. This is accomplished by inserting a calibrated probe thermocouple into the furnace adjacent to each furnace thermocouple and comparing its reading on a calibrated test potentiometer with that indicated by the furnace instrument. Correction

factors should be applied after each probe check, but if the correction required exceeds ±3 °C (±5 °F), the source of the deviation should be corrected. MIL-H-6088 recommends that this check be made weekly, but many operators make the check as frequently as once each shift.

Temperature-Uniformity Surveys. In controlling the temperature of parts that are being heat treated it must first be determined that the temperature indicated by the furnace instruments truly represents the temperature of the nearby air or salt. Second, the uniformity of temperature within the working zone must be shown to be within a range of 11 °C, or 20 °F (6 °C, or 10 °F, for precipitation heat treatment of alloy 2024). This is accomplished by measuring the temperature at several test locations, using calibrated test thermocouples and a calibrated test potentiometer, and reading furnace instruments nearly simultaneously. MIL-H-6088 recommends monthly surveys with one test location per 1.1 m³, or 40 ft³ (0.7 m³, or 25 ft³, for air furnaces on initial survey), but with a minimum of nine test locations distributed as shown in Fig 34. Despite the large size of some furnaces, rather surprising temperature uniformities have been reported. In one instance the initial survey of an air furnace measuring 12.5 by 1.2 by 3.0 m (41 by 4 by 10 ft) showed maximum temperature variations of +1.7, -1.1 °C (+3, -2 °F). When a partition 0.3 m (1 ft) thick was lowered, converting the furnace to two chambers 6.1 by 1.2 by 3.0 m (20 by 4 by 10 ft) each, the spread was +1.1, -0.6 °C (+2, -1 °F) in one section and +0.6, -1.1 °C (+1, -2 °F) in the other.

For each furnace load, one thermocouple (the "cold" couple) should be placed in the coldest area of the furnace and another (the "hot" couple) in the hottest area. In addition to these two thermocouples, a load thermocouple should be installed. The load couple should be of approximately the same gage as the sheet or other product being heat treated. If heavy plate, forgings, or castings are being heat treated, a similar discarded item should be used at the controlling load couple. The thermocouple should be placed in a drilled hole and packed to hold it firmly in place during the heat-treating cycle. In some instances, the items being heat treated can be used as the load couples. The thermocouples can be placed in holes drilled in areas that will be removed in making the finished article.

It is important that items of different thicknesses—1 mm (0.040 in.) sheet and 25 mm (1 in.) plate, for example—not be heat treated in the same furnace load.

In salt baths, uniformity surveys usually are made by holding a probe thermocouple in each location until thermal equilibrium is reached; in air furnaces, a mock heat-treating cycle is required. First, the air furnace is
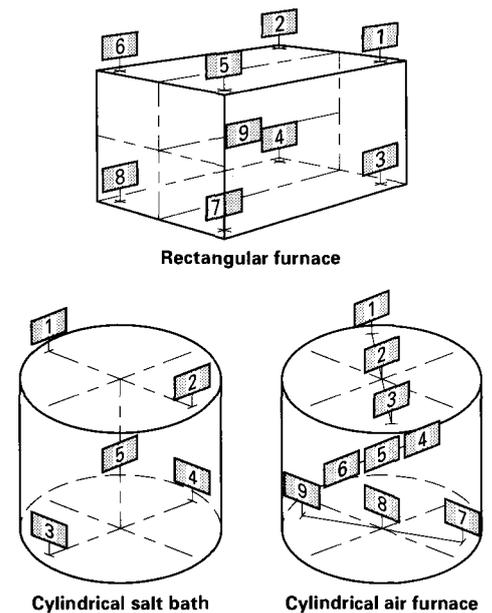

Rectangular furnace

Cylindrical salt bath   Cylindrical air furnace

Fig 34 Location of thermocouples for surveying temperature uniformity in the working zones of air furnaces and salt baths

stabilized at the test temperature. Then a rack containing the test thermocouples is inserted into the furnace. By using multiple switches or a multipoint recording instrument, all test thermocouples and furnace instruments are read every 5 min. As the temperature approaches the test range, it is advisable to increase the frequency of readings to detect possible overshooting. After thermal equilibrium is reached, readings should be continued until the recurrent temperature pattern is established.

Surveys of salt baths generally are considered acceptable whether they are made while the bath is empty or filled with work. It is controversial whether surveys of air furnaces should be made with or without a load. Undoubtedly, recovery overshoots are most likely to occur with a very light load and would not be detected if a heavier load were used. Certainly, if all loads are essentially alike, surveys should be made with typical loads. With widely varying loads, the optimum approach is to make several surveys initially, including one with an empty furnace, and then to make succeeding surveys with an empty furnace to ensure against changes in furnace characteristics. If any changes are made in the furnace that might affect temperature distribution, such as repair of vanes or louvers, several surveys should be repeated.

Another aspect of the problem of temperature control in air furnaces is the necessity of ensuring that the temperature of the parts is the same as that of the surrounding air. Furnace components whose temperature differs from the air temperature must be suitably shielded to prevent radiation to or

from the parts being heat treated. In a furnace used for solution heat treating of rivets, unshielded heating elements have been known to produce part temperatures as much as 20 °C (35 °F) higher than the control temperature, resulting in eutectic melting and cracking. In two other instances, reradiation through inadequate shielding produced a radiation effect of as much as 11 °C (20 °F). One of these problems was solved by painting the shield with reflective aluminum paint and the other by adding a 13 mm (½ in.) thick layer of asbestos to the 1.6 mm (¹⁄₁₆ in.) stainless steel shield.

Furnace-wall temperatures that differ appreciably from the temperature of the parts also must be avoided. Consequently, when the operating temperature of an air furnace is changed, waiting periods are required after the furnace instrument indicates stability, to allow the furnace walls to stabilize at the new temperature. The magnitude of this limitation is directly proportional to the efficiency of the furnace as an insulated chamber, but possibilities of such radiation should be recognized even in thin-wall furnaces.

Radiation effects are potentially dangerous because they often cannot be detected by ordinary thermocouples. Specially prepared radiation panels with thermocouples attached are used, and their readings are compared with adjacent free thermocouples. These panels normally are made of material of the same gage as the thinnest parts to be heat treated and should have a single surface area of about 650 cm² (100 in.²). A thermocouple is attached to the center of the panel by welding or peening. In order to detect the maximum effect, panel surfaces should be darkened so that their emissivity is at least as high as that of any material to be processed. During the test, the panel surfaces should be parallel to the suspected source or recipient of radiation. As an example of the number of panels required, several aerospace companies specify one panel for every 1.5 linear meters (5 linear feet) of furnace wall.

Instrument Calibration. All instruments and thermocouples must be accurately calibrated, and it is essential that the calibrations be traceable directly to the National Bureau of Standards. The chain of traceability should consist of not more than four links for sensing elements and three links for measuring elements. To illustrate, if the article calibrated by the National Bureau of Standards is called a primary standard, then the chain of traceability of measuring elements should consist of primary standard, test potentiometer, and furnace instrument. Similarly, the chain for sensing elements should consist of primary standard, secondary standard, test thermocouple, and furnace thermocouple. Every effort should be made to ensure that the temperature indicated by the furnace instruments is as close

as possible to the actual temperature. To achieve this, it is necessary to apply correction factors obtained during calibration to the next lower echelon of accuracy. Even then, if all errors inherent in the chain are in the same direction, a considerable difference will exist between the measured and actual temperatures. Therefore, it is advisable to operate as close to the mean of the desired range as possible.

Dimensional Changes during Heat Treatment

In addition to the completely reversible changes in dimensions that are simple functions of temperature change and are caused by thermal expansion and contraction, dimensional changes of more permanent character are encountered during heat treatment. These changes are of several types, some of mechanical origin and others caused by changes in metallurgical structure. Changes of mechanical origin include those arising from stresses developed by gravitational or other applied forces, from thermally induced stresses or from relaxation of residual stresses. Dimensional changes also accompany recrystallization, solution, and precipitation of alloying elements.

Solution Heat Treatment. Distortion as a result of creep during solution heat treatment should be avoided by proper loading of parts in baskets, racks, or fixtures, or by provision of adequate support for long pieces of plate, rod, bar, and extrusions heat treated in horizontal roller hearth furnaces. Sheet is provided with air-pressure support in continuous heat-treating furnaces to avoid scratching, gouging, and distortion. If parts are to be solution heat treated in fixtures or racks made of materials (such as steel) with coefficients of thermal expansion lower than that of the aluminum being treated, allowance should be made for this differential expansion to ensure that expansion of the aluminum is not restricted. Straightening immediately after solution heat treating may be preferable to fixturing.

Solution of phases formed by major alloying elements causes volumetric expansion or contraction, depending on the alloy system, and this may have to be taken into account in heat treatment of long pieces. For example, solution heat treatment and quenching of alloy 2219 causes lengthwise contraction of about 2 mm/m (0.002 in./in.). Solution heat treatment and quenching of alloys of the 7xxx series is accompanied by lengthwise expansion—about 0.6 mm/m (0.0006 in./in.) for alloy 7075 rod or plate.

Quenching. The most troublesome changes in dimensions and shape are those that occur during quenching or that result from stresses induced by quenching. Due to its nonuniform cooling, quenching may produce warpage or distortion, particularly in

Average cooling rate at center of cylinders, °F/s

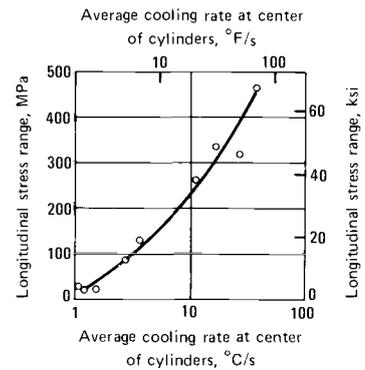Average cooling rate at center of cylinders, °C/s

Fig 35 Effect of quenching rate on longitudinal stress ranges in alloy 2014-T4 cylinders quenched in various media. Cylinders were 75 mm (3 in.) in diameter by 230 mm (9 in.) long. Cooling rate was measured from 400 to 290 °C (750 to 555 °F). Stress range is maximum tensile stress plus maximum compressive stress.

thin material and in thin sections of parts that contain variations in thickness. For thick-section products or parts, changes in external shape may be small because of rigidity, but the interior-to-surface temperature gradients that form with rapid cooling create residual stresses; these stresses normally are compressive at the surfaces and tensile in the interior.

As previously discussed, warpage or distortion of thin-section material can be reduced by using a quenching medium that provides slower cooling; however, cooling must be sufficient to produce the required properties. Slower quenching can also reduce the magnitude of residual stresses in thicker parts or pieces, as shown in Fig 9 for cylindrical specimens of alloy 6151 quenched in cold or boiling water. Stress range (maximum tensile stress plus maximum compressive stress) for a cylinder with a radius of 89 mm (3.5 in.) is about 205 MPa (30 ksi) when the cylinder is quenched in cold water but less than 70 MPa (10 ksi) when it is quenched in boiling water. The effects of average cooling rate through the temperature range from 400 to 290 °C (750 to 550 °F) on longitudinal stress ranges developed in alloy 2014 cylinders 75 mm (3 in.) in diameter are shown in Fig 35.

High stresses induced by rapid quenching generally are reduced only modestly by the precipitation heat treatments used to produce T6- or T8-type tempers. Consequently, for the alloys that require rapid cooling to develop the properties of these tempers, those incorporating mechanical stress relief (Tx51, Tx52) usually are specified when substantial metal must be removed to produce final shapes. Other T8-type tempers, such as T86 and T87, also have low residual stress as a result of the stretching required to produce them.

Heat Treatments for Precipitation and Stabilization. The most significant dimensional changes associated with precipitation heat

treatments and stabilizing heat treatments arise from concurrent dilution of the solid solution (which changes lattice parameter) and formation of precipitate. Changes in density and specific volume resulting from these changes in metallurgical structure are the reverse of those caused by solution of the alloy phases. However, because the strongest tempers are those in which the precipitate is present in nonequilibrium transition forms, the amount of change during precipitation heat treatment does not totally compensate for the previous (and opposite) change that occurred during solution heat treatment. Most of the heat-treatable alloys expand (grow) during precipitation heat treatment. Exceptions are alloys of the 7xxx wrought series and the 7xx.0 casting series, which exhibit contraction.

In alloys of the 2xxx series, the amount of growth decreases with increasing magnesium content. Thus, growth of about 1.5 mm/m (0.0015 in./in.) can be expected during precipitation heat treatment of alloy 2219-T87, about 0.5 mm/m (0.005 in./in.) for treatment of alloy 2014-T6 and less than 0.1 mm/m (0.0001 in./in.) for treatment of alloy 2024-T851. Alloys 7050 and 7075, on the other hand, contract about 0.3 mm/m (0.0003 in./in.) on precipitation heat treating from the W temper to the T6 temper and about 0.7 mm/m (0.0007 in./in.) on treating from the W temper to the T73 temper.

Stabilizing T7-type treatments cause greater amounts of growth than the T5-, T6-, or T8-type treatments for the same alloys. This increased growth is associated either with formation of increased amounts of transition precipitates or with transformation of transition precipitates to equilibrium phases.

Dimensional Stability in Service

Dimensional stability of heat-treated parts in service depends on alloy, temper, and service conditions. Of the latter, excluding mechanical conditions such as applied loads, the most important is service temperature range relative to the range in which precipitation occurs. Residual stresses constitute another source of dimensional changes. Stress relief minimizes changes due to residual stresses, and most mill products usually are supplied in tempers that include stress relief. Potential dimensional change as a result of further precipitation in parts that operate at elevated temperatures is minimized for wrought products by use of T7-type stabilizing treatments and for castings by use of T5-type treatments. However, components of high-precision equipment, such as instruments for aerospace guidance systems and optical and telescopic devices, may require special supplementary treatments during manufacture to further reduce stresses or subsequent precipitation. (These treatments are discussed below, under "Stability of Precision Equipment.")

The T3- and T4-type tempers are the least stable dimensionally because of possible precipitation in service. Alloys 2024 and its variants have the smallest dimensional change in aging; the total change from the quenched to the average state is of the order of 0.06 mm/m (0.00006 in./in.), less than the change due to a temperature variation of 3 °C (5 °F). These alloys therefore can be used in the T3- and T4-type tempers, except for precision equipment. For all other alloys, T6- or T8-type tempers should be used, because in these tempers all the alloys have good dimensional stability.

Stability of Precision Equipment. Proper maintenance of high-precision devices, such as gyros, accelerometers, and optical systems, requires use of materials in which dimensional changes from metallurgical instability are limited from 10 μm/m (10 μin./in.). Several laboratory investigations and considerable practical experience have shown that wrought alloys 2024 and 6061 and casting alloy 356.0 are well suited and generally preferred for such applications. Dimensional changes were no greater than 10 μm/m when alloys 2024-T851 and -T62, 6061-T651 and -T62, and 356.0-T51, -T6, and -T7 were tested for more than a year at room temperature and for several months at 70 °C (160 °F), and then the same alloys were tested with repeated thermal cycling between 20 and −70 °C (68 and −94 °F).

Because stresses applied or induced by acceleration in such devices generally are not high, strength levels lower than those of the highest-strength tempers frequently are satisfactory. To increase precision of machining to intended dimensions, as well as to promote maximum stability, it is common practice to apply additional thermal treatments for stress relief and precipitation of 1 to 2 h at temperatures of 175 to 205 °C (350 to 400 °F) after rough machining. These additional treatments sometimes are repeated at successive stages of processing, and even after final machining. In addition, it has been claimed that one or two cyclic treatments consisting of cooling to −100 °C (−150 °F), holding for 2 h, heating to 232 to 240 °C (450 to 465 °F) and again holding for 2 h can improve dimensional stability of 356-T6 castings.

Quality Assurance

Quality-assurance criteria that heat-treated materials must meet always include minimum tensile properties and, for certain alloys and tempers, adequate fracture toughness and resistance to detrimental forms of corrosion (such as intergranular or exfoliation attack) or to stress-corrosion cracking. All processing steps through heat treatment must be carefully controlled to ensure high and reliable performance.

Tensile Tests. In general, the relatively constant relationships among various prop-
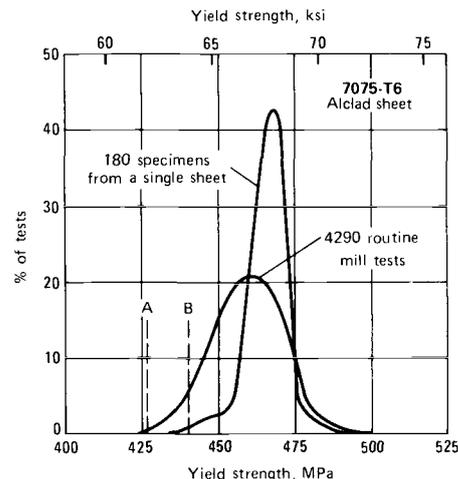

Fig 36 Comparison of distribution of yield strength in heat-treated 7075-T6 clad sheet product with distribution in a single sheet. A is 95% probability that not more than 1% of all material will fall below this value; B is 95% probability that not more than 10% of all material will fall below this value. (A and B refer only to curve representing 4290 routine mill tests.)

erties allow the use of tensile properties alone as acceptance criteria. The minimum guaranteed strength is ordinarily that value above which it has been statistically predicted with 95% probability that 99% or more of the material will pass. The inherent variability within lots and among specimens from a given piece is shown in Fig 36. Testing provides a check for evidence of conformance; process capability and process control are the foundations for guaranteed values.

Published minimum guaranteed values are applicable only to specimens cut from a specific location in the product, with their axes oriented at a specific angle to the direction of working as defined in the applicable procurement specification. In thick plate, for example, the guaranteed values apply to specimens taken from a plane midway between the center and the surface, and their axes parallel to the width dimension (long transverse). Different properties should be expected in specimens taken from other locations, or in specimens whose axes were parallel to thickness dimension (short transverse). However, the specified "referee" locations and orientations do provide a useful basis for lot-to-lot comparisons, and constitute a valuable adjunct to other process-control measures.

Tensile tests can be used to evaluate the effects of changes in the process, provided specimens are carefully selected. A variation in process that produces above-minimum properties on test specimens, however, is not necessarily satisfactory. Its acceptability can be judged only by comparing the resulting properties with those developed by the standard process on similarly located specimens. Finally, variations in heat-treating procedure are likely to affect

Table 13 Typical acceptable hardness values for wrought aluminum alloys

Acceptable hardness does not guarantee acceptable properties; acceptance should be based on acceptable hardness plus written evidence of compliance with specified heat-treating procedures. Hardness values higher than the listed maximums are acceptable provided that the material is positively identified as the correct alloy.

Alloy and temper	Product form(a)	Hardness			
		HRB	HRE	HRH	HR15T
2014-T3, -T4, -T42	All	65–70	87–95	...	...
2014-T6, -T62, -T65	Sheet(b)	80–90	103–110	...	...
	All others	81–90	104–110	...	...
2014-T61	All	...	100–109	...	...
2024-T3	Not clad(c)	69–83	97–106	111–118	82.5–87.5
	Clad, ≤1.60 mm (0.063 in.)	52–71	91–100	109–116	80–84.5
	Clad, >1.60 mm (0.063 in.)	52–71	93–102	109–116	...
2024-T36	All	76–90	100–110	...	85–90
2024-T4, -T42(d)	Not clad	69–83	97–106	111–118	82.5–87.5
	Clad, ≤1.60 mm (0.063 in.)	52–71	91–100	109–116	80–84.5
	Clad, >1.60 mm (0.063 in.)	52–71	93–102	109–116	...
2024-T6, -T62	All	74.5–83.5	99–106	...	84–88
2024-T81	Not clad	74.5–83.5	99–106	...	84–88
	Clad	...	99–106	...	...
2024-T86	All	83–90	105–110	...	87.5–90
6053-T6	All	...	79–87	...	74.5–78.5
6061-T4(d)	Sheet	...	60–75	88–100	64–75
	Extrusions; bar	...	70–81	82–103	67–78
6061-T6	Not clad, 0.41 mm (0.016 in.)	...	...	...	75–84
	Not clad, ≥0.51 mm (0.020 in.)	47–72	85–97	...	78–84
	Clad	...	84–96	...	...
6063-T5	All	...	55–70	89–97	62.5–70
6063-T6	All	...	70–85	...	...
6151-T6	All	...	91–102	...	...
7075-T6, -T65	Not clad(e)	85–94	106–114	...	87.5–92
	Clad:				
	≤0.91 mm (0.036 in.)	...	102–110	...	86–90
	>0.91 ≤ 1.27 mm (>0.036 ≤ 0.050 in.)	78–90	104–110	...	...
	>1.27 ≤ 1.57 mm (>0.050 ≤ 0.062 in.)	76–90	104–110	...	...
	>1.57 ≤ 1.78 mm (>0.062≤ 0.070 in.)	76–90	102–110	...	...
	>1.78 mm (0.070 in.)	73–90	102–110	...	...
7079-T6, -T65	All(e)	81–93	104–114	...	87.5–92
7178-T6	Not clad(f)	85 min	105 min	...	88 min
	Clad:				
	≤0.91 mm (0.036 in.)	...	102 min	...	86 min
	>0.91 ≤ 1.57 mm (>0.036≤ 0.062 in.)	85 min	...	...	...
	>1.57 mm (0.062 in.)	88 min	...	...	...

(a) Minimum hardness values shown for clad products are valid for thicknesses up to and including 2.31 mm (0.091 in.); for heavier-gage material, cladding should be locally removed for hardness testing or test should be performed on edge of sheet. (b) 126 to 158 HB (10 mm ball, 500 kg load). (c) 100 to 130 HB (10 mm ball, 500 kg load). (d) Alloys 2024-T4, 2024-T42 and 6061-T4 should not be rejected for low hardness until they have remained at room temperature for at least three days following solution treatment. (e) 136 to 164 HB (10 mm ball, 500 kg load). (f) 136 HB min (10 mm ball, 500 kg load)

Fig 37 Tensile strength versus hardness for various aluminum alloys and tempers

- Use a specimen that has at least 19 cm² (3 in.²) of surface area
- Remove any cladding by filing or etching
- Clean the specimen by immersing it for 1 min in a solution containing 5% concentrated nitric acid and 0.5% hydrofluoric acid at a temperature of 95 °C (200 °F); rinse in distilled water. Immerse for 1 min in concentrated nitric acid at room temperature; rinse in distilled water
- Immerse the specimen for 6 h in a freshly prepared solution containing 57 g of sodium chloride and 10 mL of 30% hydrogen peroxide per liter of water at a temperature of 30 ± 5 °C (86 ± 9 °F). More than one specimen may be corroded in the same container provided that at least 4.6 mL of solution is used for each square centimeter (30 mL/in.²) of specimen surface and that the specimens are electrically insulated from each other
- After the immersion period, wash the specimen with a soft-bristle brush to remove any loose corrosion product. Cut a cross-sectional specimen at least 19 mm (¾ in.) long through the most severely corroded area; mount and metallographically polish this specimen
- Examine the cross-sectional specimen microscopically at magnifications of 100× and 500× both before and after etching with Keller's reagent
- Describe the results of the microscopic examination in terms of the five degrees of severity of intergranular attack illustrated in Fig 38

Electrical Conductivity. For control of the corrosion and stress-corrosion characteristics of certain tempers, notably the T73 and T76 types, the materials must meet combination criteria of yield strength plus electrical conductivity. Although these criteria are based on indirect measurements of properties, their validity for ensuring the intended corrosion and stress-corrosion resistance

the relationships among tensile properties and other mechanical properties. In applications where other properties are more important than tensile properties, the other properties should be checked also.

Hardness tests are less valuable for acceptance and rejection of heat-treated aluminum alloys than they are for steel. Nevertheless, hardness tests have some utility for process control. Typical hardness values for various alloys and tempers are given in Table 13. Figure 37 shows the general relationship between longitudinal tensile strength and hardness for aluminum alloys.

Intergranular-Corrosion Test. The extent of precipitation during elevated-temperature aging of alloys 2014, 2219, and 2024 markedly

influences the type of corrosion attack and the corrosion resistance. With thin-section products quenched at rates sufficiently rapid to prevent precipitation in the grain boundaries during the quench, short periods of precipitation heat treating produce localized grain boundary precipitates adjacent to the depleted areas, producing susceptibility to intergranular corrosion. Additional heating, however, induces extensive general precipitation within the grains, lowering the corrosion potential differences between the grains and the boundary areas, thus removing the cause of the selective corrosion.

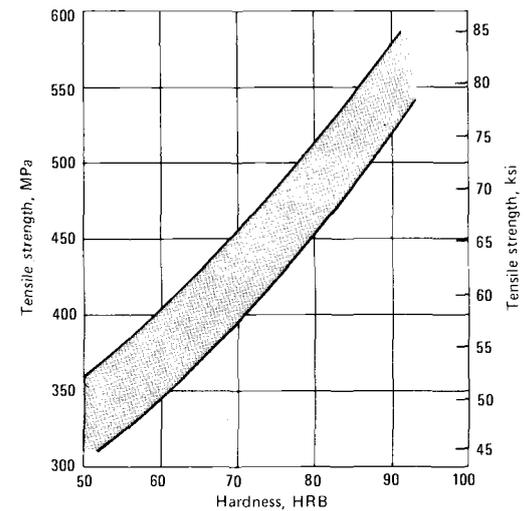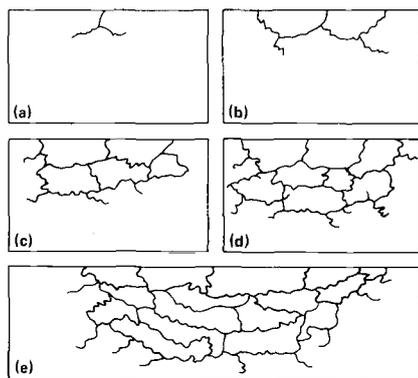The most common test for susceptibility to intergranular corrosion is carried out as follows:

Fig 38 Five degrees of severity of intergranular attack. Severity of intergranular attack (schematic), as observed microscopically in transverse sections after test for susceptibility to intergranular corrosion. Top of each area shown in surface exposed to corrosive solution

Alloy	Condition	Product form
2048	T8	Sheet and plate
2124	T3, T8	Sheet and plate
2419	T8	Sheet, plate, extrusions, and forgings
7049	T7	Plate, forgings, and extrusions
7050	T7	Sheet, plate, forgings, and extrusions
7150	T6	Sheet and plate
7175	T6, T7	Sheet, plate, forgings, and extrusions
7475	T6, T7	Sheet and plate

has been firmly established by extensive correlation and testing.

Low tensile strengths may be accompanied by high levels of electrical conductivity, so electrical conductivity is sometimes used as a quality-assurance diagnostic tool. However, because the correlation between strength and electrical conductivity is strongly a function of chemical composition and fabricating practice, use of electrical conductivity is not recommended except for rough screening. This screening must be followed by hardness testing, and then by tensile testing if the hardness tests indicate that the heat treatment was suspect.

Fracture Toughness Indices. Fracture toughness is rarely, if ever, a design consideration in the 1000, 3000, 4000, 5000, and 6000 series alloys. The fracture toughness of these alloys is sufficiently high that thicknesses beyond those commonly produced would be required to obtain a valid test.

Fracture toughness is a meaningful design-related parameter for some conventional high-strength alloys and all the controlled-toughness, high-strength alloys. Conventional aerospace alloys for which fracture toughness minimums may be useful in design include 2014, 2024, 2219, 7075, and 7079. These alloys have toughness levels that are inferior to those of their controlled-toughness counterparts. Consequently, these products are not used in fracture-critical applications, although fracture toughness can be a meaningful design parameter. Fracture toughness is not guaranteed in conventional high-strength alloys.

Fracture toughness quality control and material procurement minimums are appropriate for controlled-toughness, high-strength alloys. The alloys and tempers currently identified as controlled-toughness, high-strength products include:

The fracture toughness of these alloys and tempers range in measured K_{Ic} values from about 20 MPa$\sqrt{m}$ (18 ksi$\sqrt{in.}$) upward. Controlled-toughness alloys are often derivatives of conventional alloys. For example, 7475 alloy is a derivative of 7075 with maximum compositional limits on some elements that were found to decrease toughness.

In products of the newer controlled-toughness high-strength alloys 2090, 2091, 2124, 2224, 2324, 7050, 7149, 7150, 7175, 7475, and 8090, which provide guaranteed levels of fracture toughness, minimum values of the applicable indices, K_{IC} or K_c, are established by accumulation of statistical data from production lots as a basis for guaranteed minimum values. If the minimum specified fracture toughness value is not attained, the material is not acceptable. Some specifications allow use of less-expensive screening tests (such as the notch tensile or chevron-notched short bar) as a basis for release of high-toughness alloy products. In these instances, correlations between K_{Ic} and the screening test result is used to establish the appropriate notch-yield ratio as a lot-release criterion.

Temper Designations for Heat-Treatable Aluminum Alloys

The temper designations used in the United States for heat-treatable aluminum alloys are part of the system that has been adopted as an American National Standard (ANSI H35.1). Used for all wrought and cast product forms except ingot, the system is based on the sequences of mechanical or thermal treatments, or both, used to produce the various tempers. The temper designation follows the alloy designation and is separated from it by a hyphen. Basic temper designations consist of individual capital letters. Major subdivisions of basic tempers, where required, are indicated by one or more digits following the letter. These digits designate specific sequences of treatments that produce specific combinations of characteristics in the product. Variations in treatment conditions within major subdivisions are identified by additional digits. The conditions during heat treatment (such as time, temperature, and quenching rate) used to produce a given temper in one alloy

may differ from those employed to produce the same temper in another alloy.

Designations for the common heat-treated tempers, and descriptions of the sequences of operations used to produce those tempers, are given in the following paragraphs. (For the entire aluminum alloy temper designation system, including designations for non-heat-treatable alloys, see Volume 2 of this *Metals Handbook* series.)

Basic temper designations for heat-treated conditions include the codes O, W, and T. Other basic temper designations are F (as fabricated) and H (strain hardened).

O, annealed. Applies to wrought products that are annealed to obtain lowest strength temper and to cast products that are annealed to improve ductility and dimensional stability. The O may be followed by a digit other than zero.

W, solution heat treated. An unstable temper applicable to any alloy that naturally ages (spontaneously ages at room temperature) after solution heat treatment. This designation is specific only when the period of natural aging is indicated—for example, W ½ h. (See also the discussion of the Tx51, Tx52, and Tx54 tempers, in the section below on subdivision of the T temper.)

T, heat treated to produce stable tempers other than O. Applies to products that are thermally treated, with or without supplementary strain hardening, to produce stable tempers. The T is always followed by one or more digits, as discussed below.

Major Subdivisions of T Temper. In T-type designations, the T is followed by a number from 1 to 10; each number denotes a specific sequence of basic treatments, as described below.

T1, cooled from an elevated-temperature shaping process and naturally aged to a substantially stable condition. Applies to products that are not cold worked after an elevated-temperature shaping process such as casting or extrusion, and for which mechanical properties have been stabilized by room-temperature aging. If the products are flattened or straightened after cooling from the shaping process, the effects of the cold work imparted by flattening or straightening are not recognized in specified property limits.

T2, cooled from an elevated-temperature shaping process, cold worked, and naturally aged to a substantially stable condition. Applies to products that are cold worked specifically to improve strength after cooling from a hot-working process such as rolling or extrusion, and for which mechanical properties have been stabilized by room-temperature aging. The effects of cold work, including any cold work imparted by flattening or straightening, are recognized in specified property limits.

T3, solution heat treated, cold worked, and naturally aged to a substantially stable condition. Applies to products that are cold

worked specifically to improve strength after solution heat treatment, and for which mechanical properties have been stabilized by room-temperature aging. The effects of cold work, including any cold work imparted by flattening or straightening, are recognized in specified property limits.

T4, solution heat treated and naturally aged to a substantially stable condition. Applies to products that are not cold worked after solution heat treatment, and for which mechanical properties have been stabilized by room-temperature aging. If the products are flattened or straightened, the effects of the cold work imparted by flattening or straightening are not recognized in specified property limits.

T5, cooled from an elevated-temperature shaping process and artificially aged. Applies to products that are not cold worked after an elevated-temperature shaping process such as casting or extrusion, and for which mechanical properties or dimensional stability, or both, have been substantially improved by precipitation heat treatment. If the products are flattened or straightened after cooling from the shaping process, the effects of the cold work imparted by flattening or straightening are not recognized in specified property limits.

T6, solution heat treated and artificially aged. Applies to products that are not cold worked after solution heat treatment, and for which mechanical properties or dimensional stability, or both, have been substantially improved by precipitation heat treatment. If the products are flattened or straightened, the effects of the cold work imparted by flattening or straightening are not recognized in specified property limits.

T7, solution heat treated and stabilized. Applies to products that have been precipitation heat treated to the extent that they are overaged. Stabilization heat treatment carries the mechanical properties beyond the point of maximum strength to provide some special characteristic, such as enhanced resistance to stress-corrosion cracking or to exfoliation corrosion.

T8, solution heat treated, cold worked, and artificially aged. Applies to products that are cold worked specifically to improve strength after solution heat treatment, and for which mechanical properties or dimensional stability, or both, have been substantially improved by precipitation heat treatment. The effects of cold work, including any cold work imparted by flattening or straightening, are recognized in specified property limits.

T9, solution heat treated, artificially aged, and cold worked. Applies to products that are cold worked specifically to improve strength after they have been precipitation heat treated.

T10, cooled from an elevated-temperature shaping process, cold worked, and artificially aged. Applies to products that are cold worked specifically to improve strength after cooling from a hot-working process such as rolling or extrusion, and for which mechanical properties or dimensional stability, or both, have been substantially improved by precipitation heat treatment. The effects of cold work, including any cold work imparted by flattening or straightening, are recognized in specified property limits.

Other Subdivisions of T Temper Codes for Stress-Relieved Products. When it is desirable to identify a variation of one of the ten major T tempers described above, additional digits, the first (x) of which cannot be zero, may be added to the designation.

The following specific sets of additional digits have been assigned to stress-relieved wrought products.

Tx51, stress relieved by stretching. Applies to the following products when stretched to the indicated amounts after solution heat treatment or after cooling from an elevated-temperature shaping process:

Product form	Permanent set, %
Plate	1½–3
Rod, bar, shapes, extruded tube	1–3
Drawn tube	½–3

Tx51 applies directly to plate and to rolled or cold finished rod and bar. These products receive no further straightening after stretching. Tx51 also applies to extruded rod, bar, shapes, and tubing, and to drawn tubing, when designated as follows:

• Tx510. Products that receive no further straightening after stretching
• Tx511. Products that may receive minor straightening after stretching to comply with standard tolerances
• Tx52. Stress relieved by compressing. Applies to products that are stress relieved by compressing after solution heat treatment, or after cooling from a hot-working process to produce a permanent set of 1 to 5%
• Tx54. Stress relieved by combining stretching and compressing. Applies to die forgings that are stress relieved by restriking cold in the finish die. (These same digits—and 51, 52, and 54—may be added to the designation W to indicate unstable solution heat-treated and stress-relieved tempers)

Temper designations T42 and T62 have been assigned to wrought products heat treated from the O or the F temper to demonstrate response from the heat treatment described below. Temper designations T42 and T62 also may be applied to wrought products heat treated from any temper by the user when such heat treatment results in the mechanical properties applicable to these tempers.

• *T42.* Solution heat treated from the O or the F temper to demonstrate response to heat treatment and naturally aged to a substantially stable condition
• *T62.* Solution heat treated from the O or the F temper to demonstrate response to heat treatment and artificially aged

Subdivision of the O Temper. In temper designations for annealed products, a digit following the O indicates special characteristics. For example, O1 denotes that a product has been heat treated according to a time/temperature schedule approximately the same as that used for solution heat treatment, and then air cooled to room temperature, to accentuate ultrasonic response and provide dimensional stability; this designation applies to products that are to be machined prior to solution heat treatment by the user.

REFERENCES

1. S. Hirano *et al.*, Quench Sensitivity in Al-Li Based Alloys, *Proceedings of Conference on Aluminum-Lithium Alloys (Vol 1)*, Materials and Component Engineering Publications, 1989, p 335-344
2. T. Sheppard, *Mater. Sci. Technol.*, Vol 4, July 1988, p 636
3. J.E. Hatch, in *Aluminum Properties and Physical Metallurgy*, American Society for Metals, 1984, p 165-166
4. C.E. Bates, Selecting Quenchants to Maximize Tensile Properties and Minimize Distortion in Aluminum Parts, *J. Heat Treat.*, Vol 5 (No. 1), 1987, p 27-40
5. T. Croucher, Critical Parameters for Evaluating Polymer Quenching of Aluminum, *Heat Treat.*, Vol 19 (No. 12), Dec 1987, p 21-25
6. W.L. Fink and L.A. Willey, Quenching of 75S Aluminum Alloy, *Trans. AIME*, Vol 175, 1948, p 414-427
7. J.W. Evancho and J.T. Staley, Kinetics of Precipitation in Aluminum Alloys during Continuous Cooling, *Metall. Trans. A*, Vol 5A, Jan 1974, p 43-47
8. J.T. Staley, *Industrial Heating XLIV*, Oct 1977, p 6-9
9. G.F. Bobart, *J. Heat Treat.*, Vol 6 (No. 1), 1988, p 47-52