

Chapter 10

The light alloys

Introduction

No fewer than 14 pure metals have densities $\leq 4.5 \text{ Mg m}^{-3}$ (see Table 10.1). Of these, titanium, aluminium and magnesium are in common use as structural materials. Beryllium is difficult to work and is toxic, but it is used in moderate quantities for heat shields and structural members in rockets. Lithium is used as an alloying element in aluminium to lower its density and save weight on airframes. Yttrium has an excellent set of properties and, although scarce, may eventually find applications in the nuclear-powered aircraft project. But the majority are unsuitable for structural use because they are chemically reactive or have low melting points.*

Table 10.2 shows that alloys based on aluminium, magnesium and titanium may have better stiffness/weight and strength/weight ratios than steel. Not only that; they

Table 10.1 The light metals

<i>Metal</i>	<i>Density (Mg m⁻³)</i>	<i>T_m(°C)</i>	<i>Comments</i>
Titanium	4.50	1667	High <i>T_m</i> – excellent creep resistance.
Yttrium	4.47	1510	Good strength and ductility; scarce.
Barium	3.50	729	
Scandium	2.99	1538	Scarce.
Aluminium	2.70	660	
Strontium	2.60	770	Reactive in air/water.
Caesium	1.87	28.5	Creeps/melts; very reactive in air/water.
Beryllium	1.85	1287	Difficult to process; very toxic.
Magnesium	1.74	649	
Calcium	1.54	839	Reactive in air/water.
Rubidium	1.53	39	} Creep/melt; very reactive in air/water.
Sodium	0.97	98	
Potassium	0.86	63	
Lithium	0.53	181	

* There are, however, many *non-structural* applications for the light metals. Liquid sodium is used in large quantities for cooling nuclear reactors and in small amounts for cooling the valves of high-performance i.c. engines (it conducts heat 143 times better than water but is less dense, boils at 883°C, and is safe as long as it is kept in a sealed system.) Beryllium is used in windows for X-ray tubes. Magnesium is a catalyst for organic reactions. And the reactivity of calcium, caesium and lithium makes them useful as residual gas scavengers in vacuum systems.

Table 10.2 Mechanical properties of structural light alloys

Alloy	Density ρ (Mg m ⁻³)	Young's modulus E (GPa)	Yield strength σ_y (MPa)	E/ ρ^*	E ^{1/2} / ρ^*	E ^{1/3} / ρ^*	σ_y/ρ^*	Creep temperature (°C)
Al alloys	2.7	71	25–600	26	3.1	1.5	9–220	150–250
Mg alloys	1.7	45	70–270	25	4.0	2.1	41–160	150–250
Ti alloys	4.5	120	170–1280	27	2.4	1.1	38–280	400–600
(Steels)	(7.9)	(210)	(220–1600)	27	1.8	0.75	28–200	(400–600)

* See Chapter 25 and Fig. 25.7 for more information about these groupings.

are also corrosion resistant (with titanium exceptionally so); they are non-toxic; and titanium has good creep properties. So although the light alloys were originally developed for use in the aerospace industry, they are now much more widely used. The dominant use of aluminium alloys is in building and construction: panels, roofs, and frames. The second-largest consumer is the container and packaging industry; after that come transportation systems (the fastest-growing sector, with aluminium replacing steel and cast iron in cars and mass-transit systems); and the use of aluminium as an electrical conductor. Magnesium is lighter but more expensive. Titanium alloys are mostly used in aerospace applications where the temperatures are too high for aluminium or magnesium; but its extreme corrosion resistance makes it attractive in chemical engineering, food processing and bio-engineering. The growth in the use of these alloys is rapid: nearly 7% per year, higher than any other metals, and surpassed only by polymers.

The light alloys derive their strength from *solid solution hardening*, *age* (or *precipitation*) *hardening*, and *work hardening*. We now examine the principles behind each hardening mechanism, and illustrate them by drawing examples from our range of generic alloys.

Solid solution hardening

When other elements dissolve in a metal to form a solid solution they make the metal harder. The solute atoms differ in size, stiffness and charge from the solvent atoms. Because of this the randomly distributed solute atoms interact with dislocations and make it harder for them to move. The theory of solution hardening is rather complicated, but it predicts the following result for the yield strength

$$\sigma_y \propto \epsilon_s^{3/2} C^{1/2}, \quad (10.1)$$

where C is the solute concentration. ϵ_s is a term which represents the “mismatch” between solute and solvent atoms. The form of this result is just what we would expect: badly matched atoms will make it harder for dislocations to move than well-matched atoms; and a large population of solute atoms will obstruct dislocations more than a sparse population.

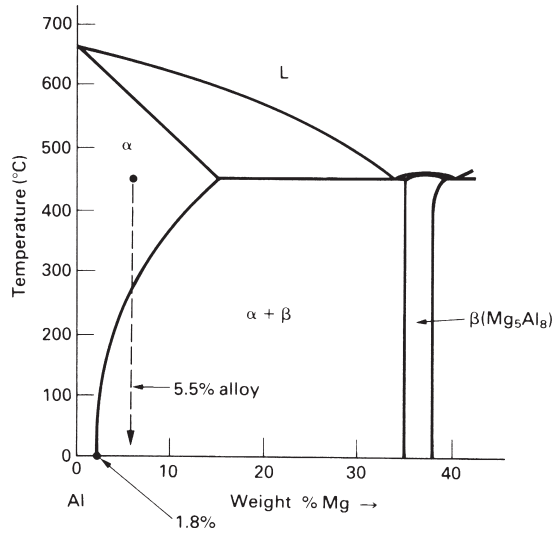


Fig. 10.1. The aluminium end of the Al-Mg phase diagram.

Of the generic aluminium alloys (see Chapter 1, Table 1.4), the 5000 series derives most of its strength from solution hardening. The Al-Mg phase diagram (Fig. 10.1) shows why: at room temperature aluminium can dissolve up to 1.8 wt% magnesium at equilibrium. In practice, Al-Mg alloys can contain as much as 5.5 wt% Mg in solid solution at room temperature – a supersaturation of $5.5 - 1.8 = 3.7$ wt%. In order to get this supersaturation the alloy is given the following schedule of heat treatments.

(a) Hold at 450°C (“solution heat treat”)

This puts the 5.5% alloy into the single phase (α) field and all the Mg will dissolve in the Al to give a random substitutional solid solution.

(b) Cool moderately quickly to room temperature

The phase diagram tells us that, below 275°C, the 5.5% alloy has an *equilibrium* structure that is two-phase, $\alpha + Mg_5Al_8$. If, then, we cool the alloy *slowly* below 275°C, Al and Mg atoms will diffuse together to form precipitates of the intermetallic compound Mg_5Al_8 . However, below 275°C, diffusion is slow and the C-curve for the precipitation reaction is well over to the right (Fig. 10.2). So if we cool the 5.5% alloy moderately quickly we will miss the nose of the C-curve. None of the Mg will be taken out of solution as Mg_5Al_8 , and we will end up with a supersaturated solid solution at room temperature. As Table 10.3 shows, this supersaturated Mg gives a substantial increase in yield strength.

Solution hardening is not confined to 5000 series aluminium alloys. The other alloy series all have elements dissolved in solid solution; and they are all solution strengthened to some degree. But most aluminium alloys owe their strength to fine precipitates of intermetallic compounds, and solution strengthening is not dominant

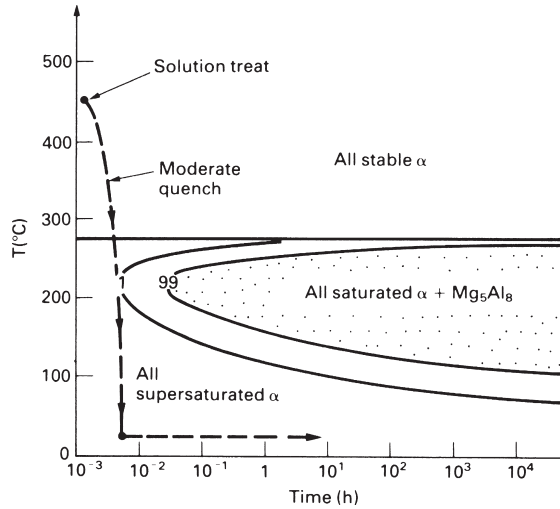


Fig. 10.2. Semi-schematic TTT diagram for the precipitation of Mg_5Al_8 from the Al-5.5 wt% Mg solid solution.

Table 10.3 Yield strengths of 5000 series (Al-Mg) alloys

Alloy	wt% Mg	σ_y (MPa) (annealed condition)
5005	0.8	40
5050	1.5	55
5052	2.5	90
5454	2.7	120
5083	4.5	145
5456	5.1	160

} supersaturated

as it is in the 5000 series. Turning to the other light alloys, the most widely used titanium alloy (Ti-6 Al 4V) is dominated by solution hardening (Ti effectively dissolves about 7 wt% Al, and has complete solubility for V). Finally, magnesium alloys can be solution strengthened with Li, Al, Ag and Zn, which dissolve in Mg by between 2 and 5 wt%.

Age (precipitation) hardening

When the phase diagram for an alloy has the shape shown in Fig. 10.3 (a solid solubility that decreases markedly as the temperature falls), then the potential for *age* (or *precipitation*) *hardening* exists. The classic example is the Duralumins, or 2000 series aluminium alloys, which contain about 4% copper.

The Al-Cu phase diagram tells us that, between 500°C and 580°C, the 4% Cu alloy is single phase: the Cu dissolves in the Al to give the random substitutional solid

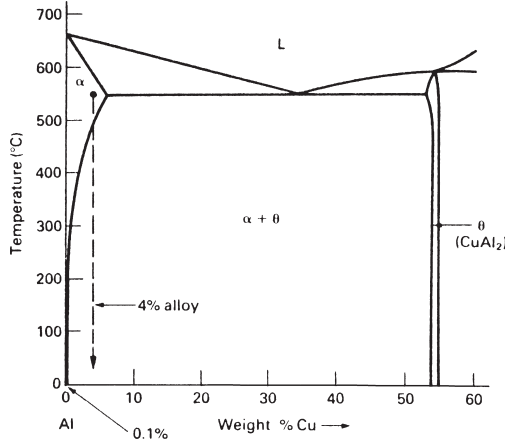


Fig. 10.3. The aluminium end of the Al–Cu phase diagram.

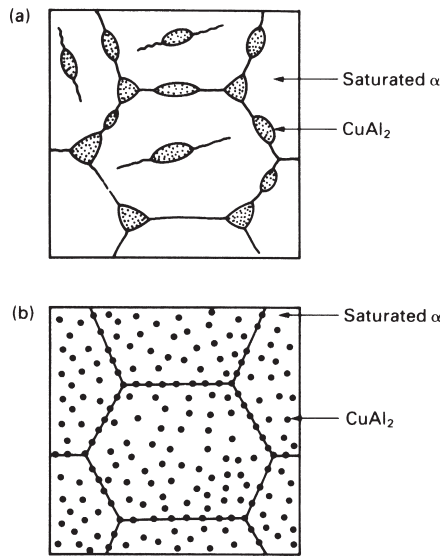


Fig. 10.4. Room temperature microstructures in the Al + 4 wt% Cu alloy. **(a)** Produced by slow cooling from 550°C. **(b)** Produced by moderately fast cooling from 550°C. The precipitates in **(a)** are large and far apart. The precipitates in **(b)** are small and close together.

solution α . Below 500°C the alloy enters the two-phase field of $\alpha + \text{CuAl}_2$. As the temperature decreases the amount of CuAl_2 increases, and at room temperature the equilibrium mixture is 93 wt% $\alpha + 7$ wt% CuAl_2 . Figure 10.4(a) shows the microstructure that we would get by cooling an Al–4 wt% Cu alloy *slowly* from 550°C to room temperature. In slow cooling the driving force for the precipitation of CuAl_2 is small and the nucleation rate is low (see Fig. 8.3). In order to accommodate the equilibrium

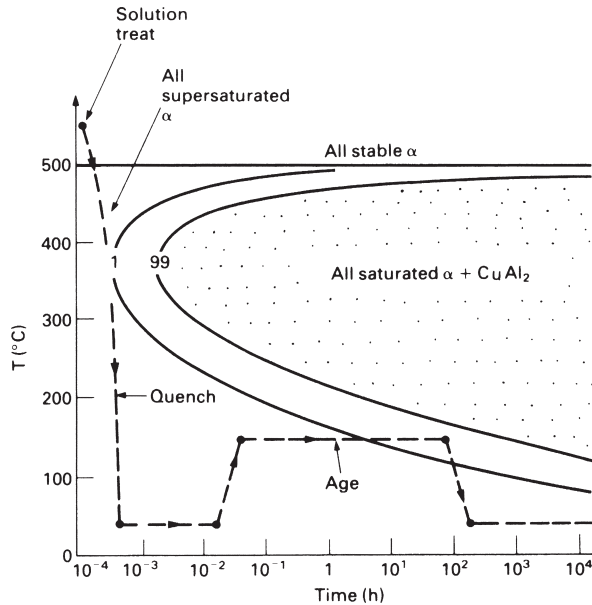


Fig. 10.5. TTT diagram for the precipitation of CuAl_2 from the Al + 4 wt% Cu solid solution. Note that the equilibrium solubility of Cu in Al at room temperature is only 0.1 wt% (see Fig. 10.3). The quenched solution is therefore carrying $4/0.1 = 40$ times as much Cu as it wants to.

amount of CuAl_2 the few nuclei that do form grow into large precipitates of CuAl_2 spaced well apart. Moving dislocations find it easy to avoid the precipitates and the alloy is rather soft. If, on the other hand, we cool the alloy rather *quickly*, we produce a much finer structure (Fig. 10.4b). Because the driving force is large the nucleation rate is high (see Fig. 8.3). The precipitates, although small, are closely spaced: they get in the way of moving dislocations and make the alloy harder.

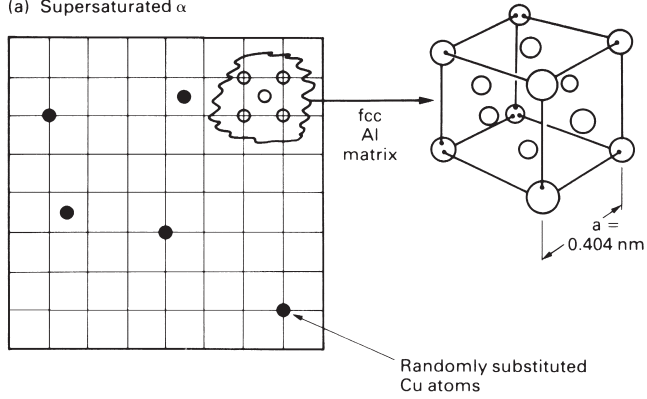
There are limits to the precipitation hardening that can be produced by direct cooling: if the cooling rate is too high we will miss the nose of the C-curve for the precipitation reaction and will not get any precipitates at all! But large increases in yield strength *are* possible if we *age harden* the alloy.

To age harden our Al-4 wt% Cu alloy we use the following schedule of heat treatments.

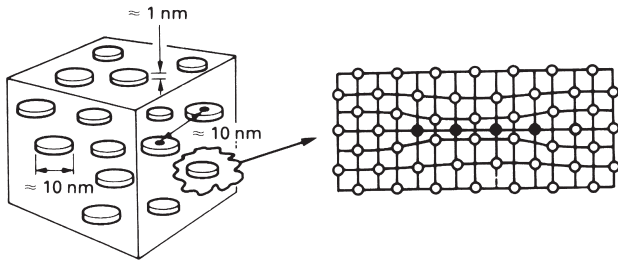
- Solution heat treat at 550°C. This gets all the Cu into solid solution.
- Cool rapidly to room temperature by quenching into water or oil (“quench”).* We will miss the nose of the C-curve and will end up with a highly supersaturated solid solution at room temperature (Fig. 10.5).
- Hold at 150°C for 100 hours (“age”). As Fig. 10.5 shows, the supersaturated α will transform to the equilibrium mixture of saturated $\alpha + \text{CuAl}_2$. But it will do so under a very high driving force and will give a very fine (and very strong) structure.

* The C-curve nose is $\approx 150^\circ\text{C}$ higher for Al-4 Cu than for Al-5.5 Mg (compare Figs 10.5 and 10.2). Diffusion is faster, and a more rapid quench is needed to miss the nose.

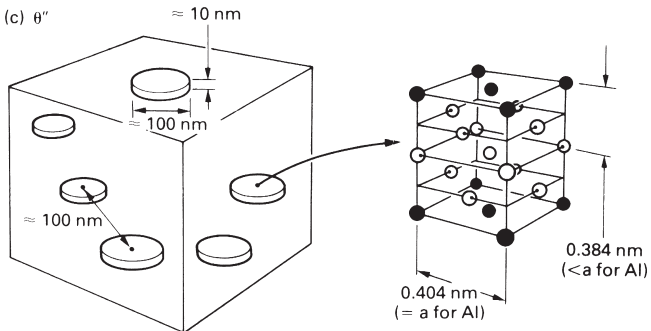
(a) Supersaturated α



(b) Guinier – Preston (GP) zones



(c) θ''



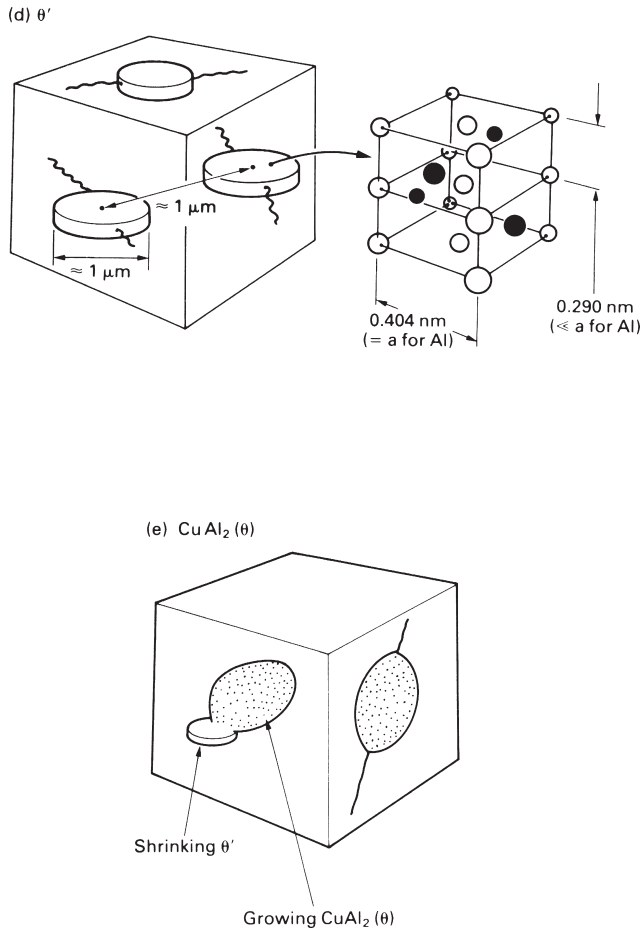


Fig. 10.6. Stages in the precipitation of CuAl_2 . Disc-shaped GP zones (**b**) nucleate homogeneously from supersaturated solid solution (**a**). The disc faces are perfectly coherent with the matrix. The disc edges are also coherent, but with a large *coherency strain*. (**c**) Some of the GP zones grow to form precipitates called θ'' . (The remaining GP zones dissolve and transfer Cu to the growing θ'' by diffusion through the matrix.) Disc faces are perfectly coherent. Disc edges are coherent, but the mismatch of lattice parameters between the θ'' and the Al matrix generates coherency strain. (**d**) Precipitates called θ' nucleate at matrix dislocations. The θ'' precipitates all dissolve and transfer Cu to the growing θ' . Disc faces are still perfectly coherent with the matrix. But disc edges are now *incoherent*. Neither faces nor edges show coherency strain, but for different reasons. (**e**) Equilibrium $\text{CuAl}_2 (\theta)$ nucleates at grain boundaries and at θ' -matrix interfaces. The θ' precipitates all dissolve and transfer Cu to the growing θ . The CuAl_2 is completely *incoherent* with the matrix (see structure in Fig. 2.3). Because of this it grows as *rounded* rather than disc-shaped particles.

Figure 10.5, as we have drawn it, is oversimplified. Because the transformation is taking place at a low temperature, where the atoms are not very mobile, it is not easy for the CuAl_2 to separate out in one go. Instead, the transformation takes place in four distinct stages. These are shown in Figs 10.6(a)–(e). The progression may appear rather involved but it is a good illustration of much of the material in the earlier chapters. More importantly, each stage of the transformation has a direct effect on the yield strength.

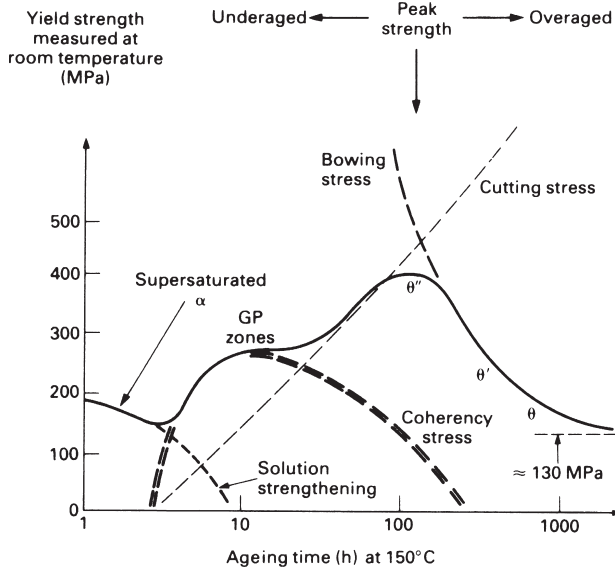


Fig. 10.7. The yield strength of quenched Al-4 wt% Cu changes dramatically during ageing at 150°C

Four separate hardening mechanisms are at work during the ageing process:

(a) Solid solution hardening

At the start of ageing the alloy is mostly strengthened by the 4 wt% of copper that is trapped in the supersaturated α . But when the GP zones form, almost all of the Cu is removed from solution and the solution strengthening virtually disappears (Fig. 10.7).

(b) Coherency stress hardening

The coherency strains around the GP zones and θ'' precipitates generate stresses that help prevent dislocation movement. The GP zones give the larger hardening effect (Fig. 10.7).

(c) Precipitation hardening

The precipitates can obstruct the dislocations directly. But their effectiveness is limited by two things: dislocations can either *cut through* the precipitates, or they can *bow around* them (Fig. 10.8).

Resistance to cutting depends on a number of factors, of which the shearing resistance of the precipitate lattice is only one. In fact the cutting stress *increases* with ageing time (Fig. 10.7).

Bowing is easier when the precipitates are far apart. During ageing the precipitate spacing increases from 10 nm to 1 μm and beyond (Fig. 10.9). The bowing stress therefore decreases with ageing time (Fig. 10.7).

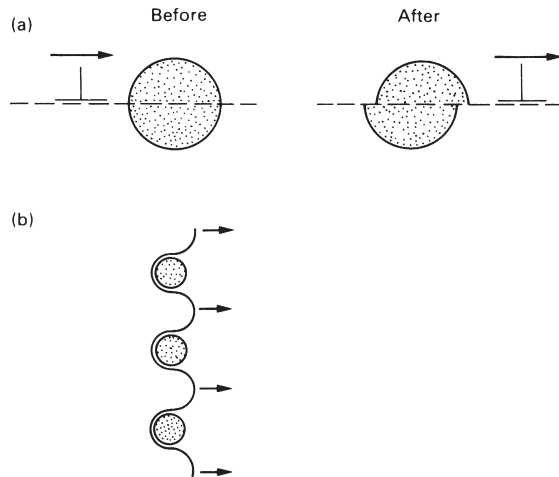


Fig. 10.8. Dislocations can get past precipitates by (a) cutting or (b) bowing.

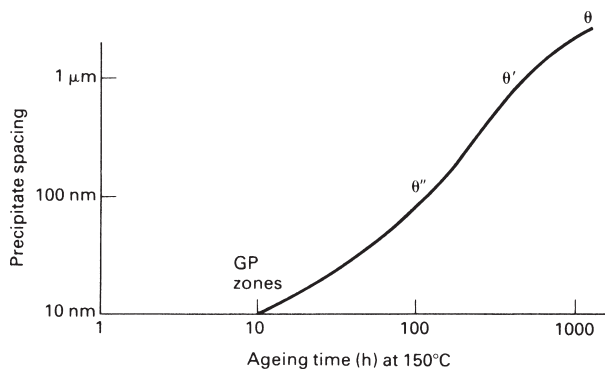


Fig. 10.9. The gradual increase of particle spacing with ageing time.

The four hardening mechanisms add up to give the overall variation of yield strength shown in Fig. 10.7. *Peak strength is reached if the transformation is stopped at θ'' .* If the alloy is aged some more the strength will *decrease*; and the only way of recovering the strength of an overaged alloy is to solution-treat it at 550°C , quench, and start again! If the alloy is not aged for long enough, then it will not reach peak strength; but this can be put right by more ageing.

Although we have chosen to age our alloy at 150°C , we could, in fact, have aged it at any temperature below 180°C (see Fig. 10.10). The lower the ageing temperature, the longer the time required to get peak hardness. In practice, the ageing time should be long enough to give good control of the heat treatment operation without being too long (and expensive).

Finally, Table 10.4 shows that copper is not the only alloying element that can age-harden aluminium. Magnesium and titanium can be age hardened too, but not as much as aluminium.

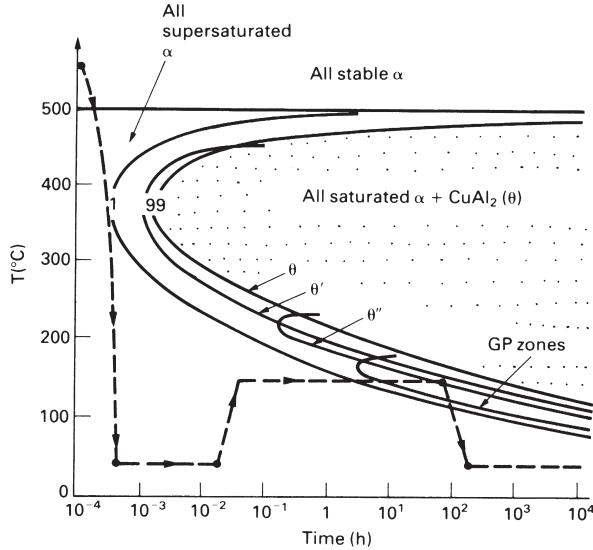


Fig. 10.10. Detailed TTT diagram for the Al-4 wt% Cu alloy. We get peak strength by ageing to give θ'' . The lower the ageing temperature, the longer the ageing time. Note that GP zones do not form above 180°C: if we age above this temperature we will fail to get the peak value of yield strength.

Table 10.4 Yield strengths of heat-treatable alloys

Alloy series	Typical composition (wt%)	σ_y (MPa)	
		Slowly cooled	Quenched and aged
2000	Al + 4 Cu + Mg, Si, Mn	130	465
6000	Al + 0.5 Mg 0.5 Si	85	210
7000	Al + 6 Zn + Mg, Cu, Mn	300	570

Work hardening

Commercially pure aluminium (1000 series) and the non-heat-treatable aluminium alloys (3000 and 5000 series) are usually work hardened. The work hardening superimposes on any solution hardening, to give considerable extra strength (Table 10.5).

Work hardening is achieved by cold rolling. The yield strength increases with strain (reduction in thickness) according to

$$\sigma_y = A\epsilon^n, \tag{10.2}$$

where A and n are constants. For aluminium alloys, n lies between 1/6 and 1/3.

Table 10.5 Yield strengths of work-hardened aluminium alloys

Alloy number	σ_y (MPa)		
	Annealed	"Half hard"	"Hard"
1100	35	115	145
3005	65	140	185
5456	140	300	370

Thermal stability

Aluminium and magnesium melt at just over 900 K. Room temperature is $0.3 T_m$, and 100°C is $0.4 T_m$. Substantial diffusion can take place in these alloys if they are used for long periods at temperatures approaching $80\text{--}100^\circ\text{C}$. Several processes can occur to reduce the yield strength: loss of solutes from supersaturated solid solution, over-ageing of precipitates and recrystallisation of cold-worked microstructures.

This lack of *thermal stability* has some interesting consequences. During supersonic flight frictional heating can warm the skin of an aircraft to 150°C . Because of this, Rolls-Royce had to develop a special age-hardened aluminium alloy (RR58) which would not over-age during the lifetime of the Concorde supersonic airliner. When aluminium cables are fastened to copper busbars in power circuits contact resistance heating at the junction leads to interdiffusion of Cu and Al. Massive, brittle plates of CuAl_2 form, which can lead to joint failures; and when light alloys are welded, the properties of the heat-affected zone are usually well below those of the parent metal.

Background reading

M. F. Ashby and D. R. H. Jones, *Engineering Materials I*, 2nd edition, Butterworth-Heinemann, 1996, Chapters 7 (Case study 2), 10, 12 (Case study 2), 27.

Further reading

I. J. Polmear, *Light Alloys*, 3rd edition, Arnold, 1995.

R. W. K. Honeycombe, *The Plastic Deformation of Metals*, Arnold, 1968.

D. A. Porter and K. E. Easterling, *Phase Transformations in Metals and Alloys*, 2nd edition, Chapman and Hall, 1992.

Problems

10.1 An alloy of Al–4 weight% Cu was heated to 550°C for a few minutes and was then quenched into water. Samples of the quenched alloy were aged at 150°C for

various times before being quenched again. Hardness measurements taken from the re-quenched samples gave the following data:

Ageing time (h)	0	10	100	200	1000
Hardness (MPa)	650	950	1200	1150	1000

Account briefly for this behaviour.

Peak hardness is obtained after 100 h at 150°C. Estimate how long it would take to get peak hardness at (a) 130°C, (b) 170°C.

[Hint: use Fig. 10.10.]

Answers: (a) 10^3 h; (b) 10 h.

- 10.2** A batch of 7000 series aluminium alloy rivets for an aircraft wing was inadvertently over-aged. What steps can be taken to reclaim this batch of rivets?
- 10.3** Two pieces of work-hardened 5000 series aluminium alloy plate were butt welded together by arc welding. After the weld had cooled to room temperature, a series of hardness measurements was made on the surface of the fabrication. Sketch the variation in hardness as the position of the hardness indenter passes across the weld from one plate to the other. Account for the form of the hardness profile, and indicate its practical consequences.
- 10.4** One of the major uses of aluminium is for making beverage cans. The body is cold-drawn from a single slug of 3000 series non-heat treatable alloy because this has the large ductility required for the drawing operation. However, the top of the can must have a much lower ductility in order to allow the ring-pull to work (the top must tear easily). Which alloy would you select for the top from Table 10.5? Explain the reasoning behind your choice. Why are non-heat treatable alloys used for can manufacture?