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Hardening mechanisms in Al–Sc alloys

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The hardening mechanism in Al–Sc alloys with scandium content of 0.11 and 0.19 at% is studied. Applying theoretical results due to the yield stress and work hardening of two phase alloys as a function of volume fraction and precipitate particle size, it is shown that after ageing at above 300° C the Orowan mechanism operates in these alloys. Using the experimental results, the volume fraction and average radius of the precipitate particles are determined.

1. Introduction

Although the age-hardenable Al–Sc alloys have excellent mechanical properties [1–5], only a few investigations have been reported in the literature connected with the hardening mechanism in these alloys.

Dritz and co-workers [4] investigated the connection between precipitates and mechanical properties in Al–Sc alloys with 0 to 0.8 wt% Sc concentrations. They found that the maximum value of the yield stress increased monotonically with scandium content up to 0.24 at% Sc, while it remains practically constant at higher concentrations. The work-hardening process studied in an Al–0.3 at% Sc alloy [5] was described by the Hollomon equation

$$\sigma = K\varepsilon^n \qquad (1)$$

where σ and ε are the true stress and true strain, respectively. The constants K and n are assumed to be independent of strain. It was found that n decreases with increasing yield stress: after quenching from 635° C the values of $R_{p0.2} = 40$ MPa and $n = 0.163$, while after ageing at 250° C for 50 hours, $R_{p0.2} = 160$ MPa and $n = 0.043$ were obtained. On the basis of these results and theoretical considerations, the authors stated that the mechanism of work hardening depends on the size of the precipitates formed during ageing. If the average particle size is larger than 6.8 nm, then the Orowan mechanism operates. The size and volume fraction of precipitates were measured by transmission electron microscopy.

Equation 1 has the disadvantage that the quantity, n, characteristic of the work hardening, cannot be directly related to the change in the microstructure. Therefore, instead of using this expression, we try to draw conclusions from both the precipitation and the work-hardening characteristics for the properties of the particles and for the hardening mechanism according to the method of Kovács and co-workers [6–8].

In the case of the cutting mechanism, the increase in the tensile yield stress can be given by [7, 9–12].

$$R_{p0.2} = 1.44\, mG \left(\frac{\Gamma}{Gb}\right)^{3/2} \left(\frac{fR}{b}\right)^{1/2} \qquad (2)$$

where the Taylor factor $m = 3$ for a polycrystalline sample, Γ is the specific surface energy of cutting a precipitate, f is the volume fraction, R is the average radius of the particles, G is the shear modulus and $\boldsymbol{b}$ is the Burgers vector.

If the Orowan mechanism is operating, then the increase of the yield stress is as follows [7, 9, 10]

$$R_{p0.2} = 0.85\, m \frac{G}{2\pi(1-\nu)^{1/2}} \left(\frac{f}{2}\right)^{1/2} \times \frac{\boldsymbol{b} \ln (R\pi/2r_0)}{R[1 - (\pi/2)(f/2)^{1/2}]} \qquad (3)$$

where ν is the Poisson number and r_0 is the cut-off radius of the dislocation core. In the case of low volume fraction, $R_{p0.2}$ is proportional to $f^{1/2}$ in both Equations 2 and 3, but $R_{p0.2}/f^{1/2}$ is proportional to $R^{1/2}$ in Equation 2 and to R^{-1} in Equation 3. This different behaviour makes it possible to distinguish between the two mechanisms during the course of precipitation.

The work hardening of polycrystalline face-centred cubic metals can be characterized by Stages II and III, because Stage I does not appear in these materials. The relationship between flow stress, σ, and strain, ε, in these two stages can be expressed as [8]

$$\sigma = \theta\varepsilon + \sigma_1, \quad \text{if } \varepsilon \leqslant \varepsilon_3 \qquad (4)$$

$$\sigma = \chi\varepsilon^{1/2} + \sigma_0, \quad \text{if } \varepsilon \geqslant \varepsilon_3. \qquad (5)$$

Here ε_3 is the strain corresponding to the transition between the two stages. The constants θ, χ and σ_1, σ_0 are independent of strain. From the conditions where the two functions, $\sigma(\varepsilon)$ and $d\sigma/d\varepsilon$ must be continuous at $\varepsilon = \varepsilon_3$, one obtains

$$\chi^2 = 2\theta(\sigma_3 - \sigma_0) \qquad (6)$$

where $\sigma_3 = \sigma(\varepsilon_3)$.

In two-phase alloys, the work-hardening behaviour in the third stage depends on the dislocation–precipitation interaction. If the deformation takes place by the cutting mechanism, the parameter, χ depends only slightly on the precipitation structure [6]. For rigid particles, the calculation of the work hardening is rather difficult. Applying some simplifying assumptions,

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Ashby [10] determined χ for single crystals as

$$\chi_s = CG\left(\frac{bf}{2R}\right)^{1/2} \quad (7)$$

where the constant C depends on the dislocation structure and its value was estimated by Ashby to be $C = 0.25 \pm 0.15$. Equation 7 is valid in the case of spherical particles and large strains. To obtain the parameter χ for polycrystalline materials one has to take into account the connection between tensile and resolved shear stress, as well as between tensile and resolved shear strain [11], which leads to the following relation

$$\chi = 3^{3/2}\chi_s \quad (8)$$

where χ can be determined from the tensile $\sigma(\varepsilon^{1/2})$ stress–strain curve.

If the volume fraction, f, and the size of particles, R, cannot be directly measured, χ gives information about the deformation mechanism.

The aim of this work is to present some results on the hardening mechanism in Al–Sc alloys.

2. Experimental details

The alloys investigated were prepared from high-purity (99.999%) aluminium and (99.9%) scandium using a master alloy with 2 wt % scandium. The scandium content of the alloys was 0.11 and 0.19 at %. Wires 1.1 mm diameter and sheets 0.8 mm thick were produced. After solution heat treatment at 640° C for 2 h and quenching into water, ageing was carried out between 200 and 400° C for different times. An Instron equipment was used to measure the stress–strain relation on 50 mm long wire samples. The error in flow stress was about 2%. σ_0 and χ were obtained by fitting Equation 5 to the linear section of the σ–$\varepsilon^{1/2}$ data by the least squares method. The parabolic, i.e. third, stage was found to be between 3 and 10% strain.

Transmission electron microscopy (TEM) was carried out on sheet samples using a Tesla BS 540 equipment by standard TEM techniques.

3. Results

Fig. 1 shows the effect of isochronal heat treatment on the yield stress. The annealing time was 1 h at each temperature. σ_0 and χ, as determined from Equation 5, are shown in Fig. 2 as a function of ageing tempera-

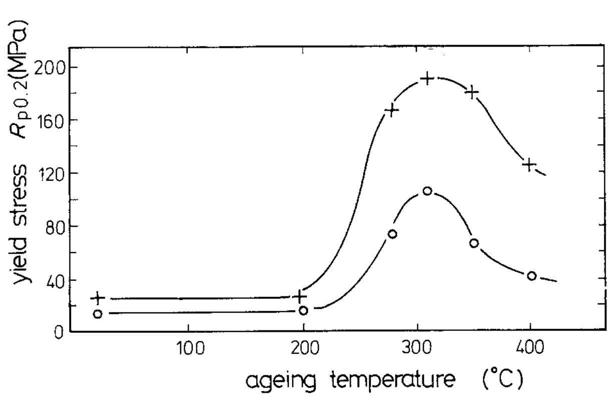

Figure 1 Yield stress of the alloys as a function of ageing temperature. The annealing time was 1 h at each temperature. (+) 0.19 at % Sc, (O) 0.11 at % Sc.

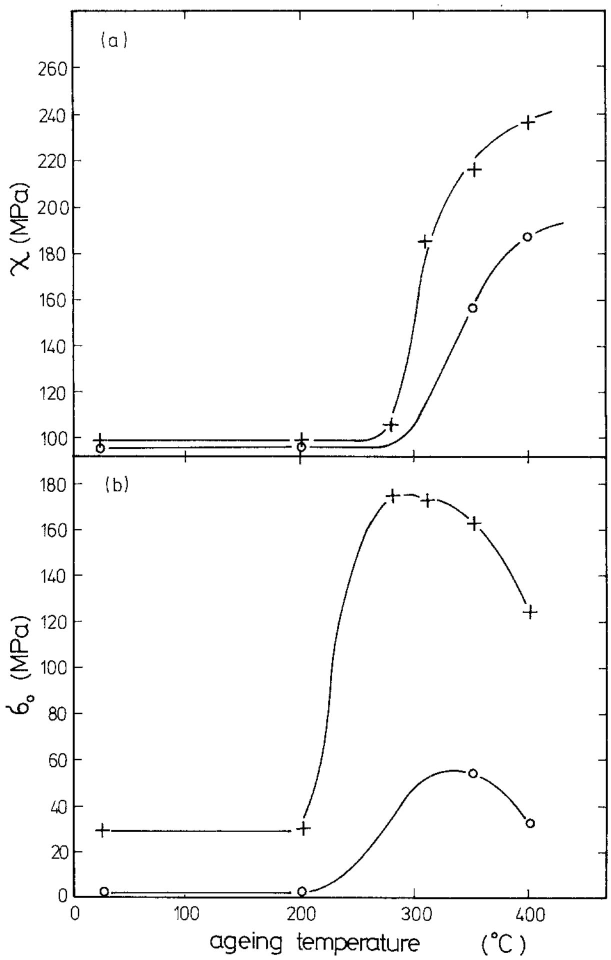

Figure 2 The same as in Fig. 1, for χ and σ_0.

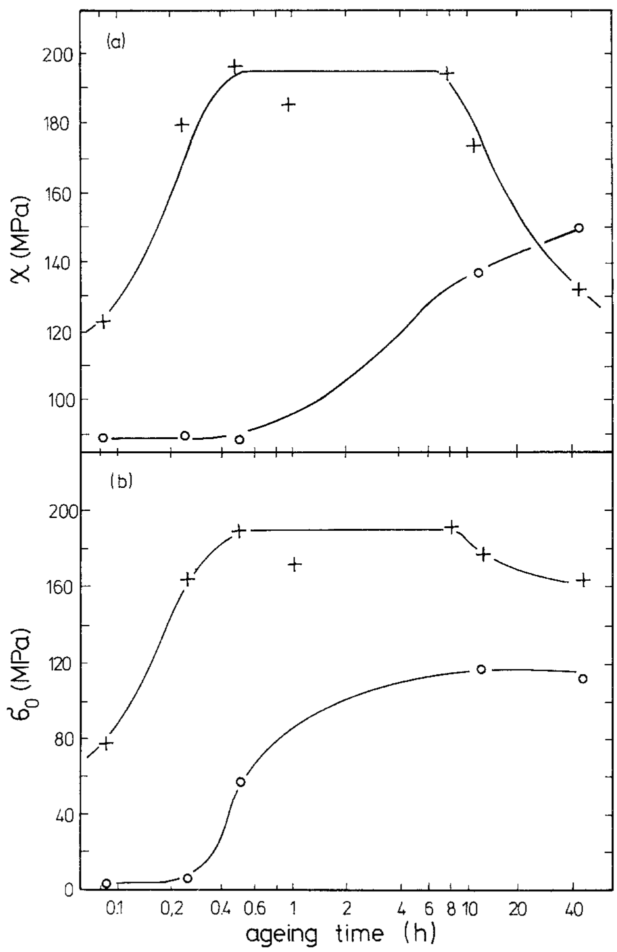

Figure 3 The dependence of χ and σ_0 on the ageing time at 310° C. For key, see Fig. 1.

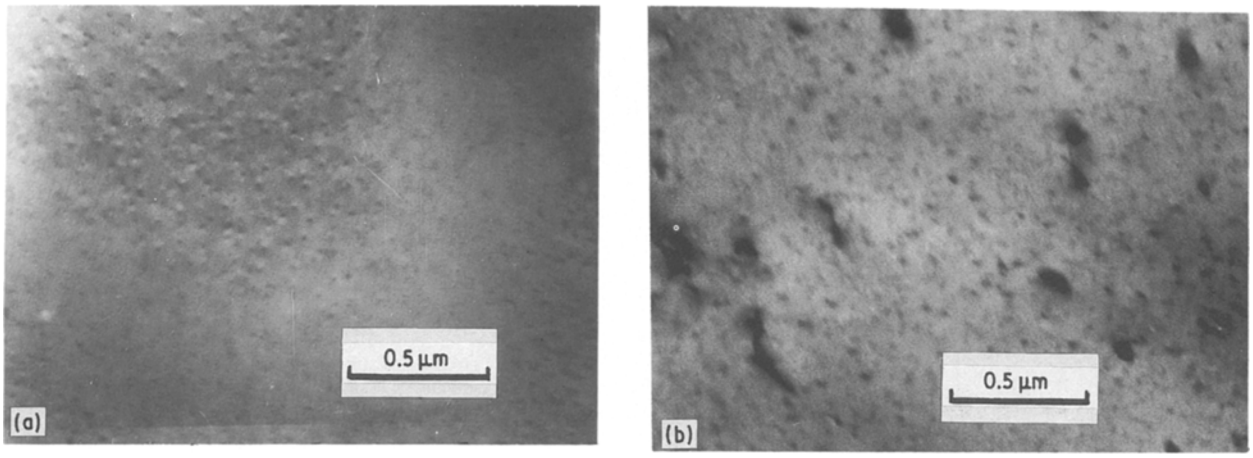

Figure 4 The precipitation structure in the alloy with 0.19 at % Sc after ageing at 310° C for (a) 1 h and (b) 12 h.

ture. These parameters only change after ageing above 200° C. σ_0 reaches its maximum between 280 and 320° C in a similar manner to $R_{p0.2}$ in Fig. 1. χ remains practically constant below 300° C and its value is approximately the same for the two alloys in this temperature range. This means that in these states of the alloys, χ depends only on the matrix properties and not on the precipitation structure.

Fig. 3 shows the change of σ_0 and χ during ageing at 310° C as a function of ageing time. The behaviour of the two alloys is different. χ changes later in the dilute alloy than in the more concentrated one. The increase in χ is caused by precipitation of scandium. This finding is in accordance with the results obtained by isochronal ageing. In the case of the higher concentration alloy, the plateau of the χ, σ_0 curves shows that after an initial change the precipitation structure, i.e. the volume fraction and size of the particles, remains practically constant up to about 8 h ageing time. Overageing in this alloy appears only after about 10 h heat treatment and it causes the decrease of both σ_0 and χ. Transmission electron micrographs were taken of the alloy with 0.19 at % Sc after ageing at 310° C for 1 h (Fig. 4a) and 12 h (Fig. 4b). It can be seen that the size of the precipitates is very fine and it hardly depends on the ageing time investigated.

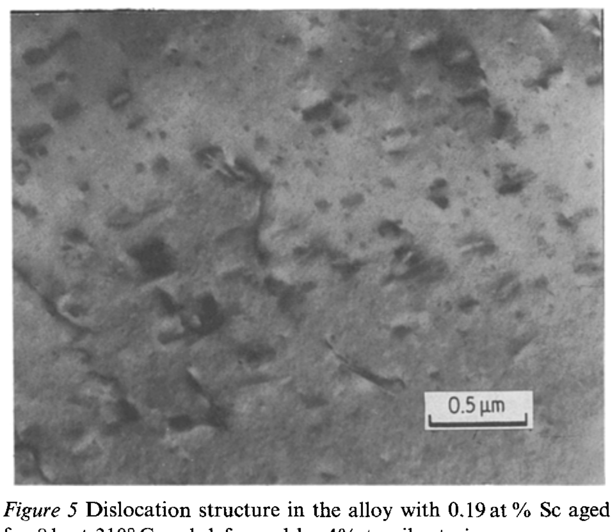

Figure 5 Dislocation structure in the alloy with 0.19 at % Sc aged for 8 h at 310° C and deformed by 4% tensile strain.

4. Discussion and conclusions

In the early period of ageing, the work-hardening parameter, χ, is independent of the precipitation process indicating that at this stage the cutting mechanism is operating. The abrupt change in χ with increasing ageing time in the more concentrated alloy indicates that after precipitation the Orowan mechanism operates. This statement is supported by Fig. 5. This transmission electron micrograph was taken on a sample aged for 8 h at 310° C and deformed by about 4% tensile strain. The sets of dislocation loops show clearly the operation of the Orowan mechanism.

For the Orowan mechanism we have two equations with the measured parameters $R_{p0.2}$ and χ, which contain three unknown quantities characteristic of the precipitation structure. These are the volume fraction, f, average particle radius, R, and C in Equation 7. To determine these quantities, we need a third quantitative

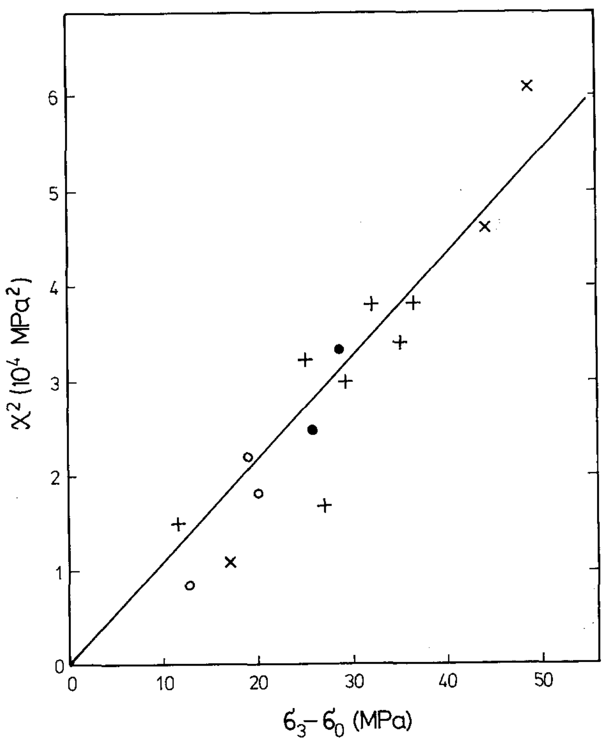

Figure 6 Plot of χ^2 against $\sigma_3 = \sigma_0$. (○, ●) 0.11 at % Sc, (+, x) 0.19 at % Sc. (○) 0.5 to 45 h at 310° C, (●) 1 h at 340 to 400° C, (+) 0.083 to 45 h at 310° C, (x) 1 h at 280 to 400° C. θ = 1090 MPa.

condition. For this purpose we suppose that in the plateau region during ageing at 310° C, Al_3Sc precipitates are in equilibrium with the remaining solid solution; therefore, if the solute concentration is known, then the volume fraction of the precipitates can be determined. By electrical resistivity measurements the equilibrium solute concentration of scandium at 640° C and at 310° C was found to be 0.150 and 0.0103 at %, respectively [13]. Using these values the volume fraction of precipitates formed at 310° C can be calculated as $f = 0.59\%$. Substituting this and the measured mean value of the yield stress, $R_{p0.2} = 170$ MPa, valid in the plateau region, into Equation 3, we obtain the average particle radius $R = 4.3$ nm, which is in good agreement with previous measurements [5]. Finally from Equations 7 and 8 with the measured mean value, $\chi = 191$ MPa, also valid in the plateau region, C can also be calculated as $C = 0.0971$.

Considering the results obtained for the alloy with 0.11 at % Sc, the yield stress in this case is smaller than in the previous case, which is due to the smaller volume fraction of precipitate particles. The equilibrium volume fraction for this alloy again calculated by the use of the solid solubility at 310° C, was found to be 0.43%. Substituting this value, together with the measured yield stress, $R_{p0.2} = 110$ MPa, into Equation 3, the average particle size in this case is found to be 5.9 nm. The application of Equation 9 leads now to $C = 0.0979$ in good agreement with the previous result.

The obtained $C \simeq 0.1$ value can be considered to be comparable with the theoretical one, $C = 0.25 \pm 0.15$, although the agreement is not too good. It is worth mentioning, however, that C depends only slightly on the values of the f and R pairs which belong to a given yield stress; therefore, the error of the present value is probably rather low.

The model applied can also be controlled by the validity of Equation 6. Fig. 6 shows the values of χ^2 as a function of $(\sigma_3 - \sigma_0)$ for all the measurements. It can be seen that Equation 6 is fulfilled, i.e. a linear connection exists between these quantities, independent of the composition and the state of the alloys. This means that the slope, θ, does not depend on the precipitation structure of the alloy [8]. Its value was found to be $\theta = 1090$ MPa.

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