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STRENGTHENING MECHANISMS ASSOCIATED WITH T1 PARTICLES IN TWO Al-Li-Cu ALLOYS

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ABSTRACT

The contribution to the yield strength or critical resolved shear stress (CRSS) of T1 precipitates in two aged Al-Li-Cu alloys has been measured experimentally and analyzed theoretically. The tensile yield stress, \( \sigma_y \), of alloys containing 2.30 wt.%Li-2.85 Cu-0.12 Zr (the 2-3 alloy) and 2.90 Li-0.99 Cu-0.12 Zr (the 3-1 alloy) was measured as a function of aging time, \( t \), at 160 and 190 °C after subjecting the samples to reversion heat treatments to dissolve the \( \delta' \) precipitates while leaving the dispersion of T1 precipitates relatively undisturbed. The reversion treatments consisted of reheating the aged samples for 60 s at 265 °C for the 2-3 alloy or 315 °C for the 3-1 alloy, followed by quenching into water. The CRSS was calculated by dividing \( \sigma_y \) by the appropriate Taylor factors for the two alloys. The contribution to the CRSS of the T1 precipitates, \( \Delta \tau_{T1} \), estimated by subtracting the CRSS of the matrix from that of the reverted samples, exhibited normal age hardening behavior. The average values of the diameter, thickness and volume fraction, \( f_{T1} \), were measured as a function of \( t \) using transmission electron microscopy. \( \Delta \tau_{T1} \) was then analyzed quantitatively according to various theories of precipitation hardening. Most of the extant theories are applicable to spherical precipitates and therefore need to be modified for the plate shape of the T1 particles and for the fact that they lie on \{111\}, meaning that only \( 3/4 \) of them are equivalent obstacles to dislocation motion. The data are in fair agreement with two versions of the theory of chemical strengthening, one based on conventional weak obstacle statistics and the other on strong obstacle statistics. The value of the T1/matrix interfacial free energy required to fit either version of the theory is rather large (~6.58 or ~9.27 J/m²). However, the aspect ratio is also very large (40 to 50), implying a significant difference between the energies of the broad and peripheral interfaces of the T1 precipitates and therefore an acceptable value of the energy of the broad interface (between 0.13 and 0.23 J/m²). The Orowan mechanism is also capable of predicting the experimental results with reasonable quantitative accuracy over the entire range of aging times, provided that allowance is made for the dislocation pairs observed in the microstructures of the reverted samples. The pairs are present because ordered \( \delta' \) precipitates nucleate during quenching from the reversion temperature. Calculations suggest that the combination of dislocation pair spacings and T1 particle spacings in this microstructure are such that the paired dislocations can be regarded as single dislocations of Burgers vector 2b. This analysis represents the first indication that the Orowan mechanism can account for a normal age hardening response.

1. Introduction

Al-Li-Cu alloys are among the most promising of Li containing Al-base alloys for aircraft applications. These alloys are age hardenable, the major microstructural constituents in the matrix being metastable spherical, coherent, ordered \( \delta' \) precipitates and thermodynamically stable plate-shaped precipitates of the T1 phase. Compared to the \( \delta' \) precipitates, the role played in strengthening Al-Li-Cu alloys by the plate-shaped T1 precipitates, which have habit planes parallel to \{111\}, has not received equivalent attention; the details of the strengthening mechanisms have been unstudied and are virtually unknown.

Gregson and Flower[1] have suggested that T1 precipitates do not improve the resistance to strain
localization because the crystallographic relationship between $T_1$ and the matrix phase does not enable the precipitates to provide an effective barrier to planar slip. Conversely, Sainfort and Guyot[2] and Sainfort et al.[3] have suggested that $T_1$ precipitates do play an important role in the strengthening of ternary Al-Li-Cu alloys. They concluded that $T_1$ precipitates are unshearable and govern the yield strength of an Al-2.7%Li-1.9%Cu alloy[2].

The results of our previous work[4] tend to support the idea that the $T_1$ precipitates are very potent strengtheners. For example, the critical resolved shear stress (CRSS) in an aged and reverted Al-2.30%Li-2.85%Cu alloy (chemically equivalent to 2090, but heat treated to dissolve the $\delta'$ precipitates produced during normal aging) is nearly 70% of the CRSS of the same alloy isothermally aged and thus containing both $\delta$ and $T_1$ precipitates. This is an interesting result since the volume fraction of $T_1$ is only roughly 0.1 to 0.25 that of $\delta'$. The important role played by $T_1$ precipitates therefore justifies an investigation of the mechanism(s) of strengthening by these particles, and it is this issue with which this paper is concerned. The strengthening effects due to the ordered $\delta'$ precipitates have been considered briefly in a previous paper[4] and are discussed in considerably more detail elsewhere[5].

2. Experimental Methods

The two alloys used in this study were provided by Alcoa in the form of plates 12 mm in thickness. The compositions (in wt.%) are Al-2.30Li-2.85Cu-0.12Zr (the 2-3 alloy) and Al-2.90Li-0.99Cu-0.12Zr (the 3-1 alloy). The grain structures, aging treatments, tensile testing, microstructural characterization, and quantitative metallographic measurements have been described or discussed previously[4,6], while details on the textures are presented elsewhere[5].

To isolate the contributions to strengthening of the $T_1$ and $\delta'$ precipitates the isothermally aged samples were subjected to reversion heat treatments[4] of 60 s at 265 °C for the 2-3 alloy and 315 °C for the 3-1 alloy. The samples were water-quenched after reversion and tensile tested within 24 h. The sizes of the $T_1$ plates in both the aged and reverted samples were measured to account for the slight growth of the $T_1$ particles during reversion.

Very fine $\delta'$ particles were typically observed in the reverted samples, most likely having nucleated during the quench. It is estimated that they could contribute at most 10 to 11 MPa to the observed CRSS of the reverted samples. The more important effect of these $\delta'$ particles is that they force the dislocations to move in pairs, a factor which has to be considered in the analysis of the strengthening caused by the $T_1$ precipitates.

Both the effective average "diameter", $\Delta$, and average thickness, $Z$, of the plate-shaped $T_1$ precipitates were measured from thin foil samples in the transmission electron microscope. When the sizes of the $T_1$ plates were considerably smaller than the foil thickness $\Delta$ was measured from an edge-on variant in bright-field images. For large $T_1$ particles, which are mostly irregular but occasionally hexagonal or rhomboidal in the plane of the plate, the diameters were measured instead from dark-field micrographs of thin foils oriented so that one variant was perpendicular to the electron beam direction. This was done to avoid errors due to truncation of the large precipitates at the foil surfaces. The accuracy in measurements of $Z$ is limited principally by its smaller magnitude ($Z < 3$ nm in the underaged alloys) and the fact that the foil has to be carefully tilted so that one variant of the $T_1$ plates is viewed exactly edge-on. Slight misorientations easily result in an overestimate of $Z$.

The determination of the volume fraction of the $T_1$ plates, $f_{T_1}$, is subject to relatively large uncertainties, due primarily to errors in measuring $Z$ and to the non-uniform spatial distribution of the precipitates. Nucleation of the $T_1$ precipitates occurs preferentially on dislocations, and the presence of dislocation sub-boundaries in the microstructure exacerbates the problem of obtaining reliable values of $f_{T_1}$. This difficulty was overcome by taking measurements in several different regions in different thin foil samples and ignoring the presence of the $T_1$ particles on sub-boundaries. To measure $f_{T_1}$, the number of a single variant of the $T_1$ plates, $N_{T_1}$, within a defined region of the TEM micrographs was counted. The variant nearly parallel to the foil surface was used, since an unknown and possibly large proportion of the edge-on (or nearly edge-on) plates are truncated by the foil surfaces, leading the measured values of $N_{T_1}$ to overestimate the true number of $T_1$ particles that are fully inside the defined volume. After $N_{T_1}$ was measured, $f_{T_1}$ was calculated from the equation

$$f_{T_1} = \frac{N_{T_1}}{A_{T_1}}$$

where $A_{T_1}$ is the area of the defined region.
\[ f_{T_1} = \frac{4\pi N_{T_1} Z(\Delta/2)^2}{V_m} \]  

where \( \pi Z(\Delta/2)^2 \) is the average volume of the precipitates and \( V_m \) is the volume of the sample in the defined region where \( N_{T_1} \) was counted. The factor of 4 results from the assumption that the \( T_1 \) precipitates are distributed equally among all four \{111\} variants. The slight difference in \( Z \) at the center and periphery of the precipitates[7] was ignored. Due to the aforementioned uncertainties, the estimates of \( f_{T_1} \) are regarded as accurate only to within about 25%.

3. Results

3.1 Kinetics of Precipitation.—Figure 1 shows the variation of \( \Delta, Z, \Delta/Z \) and \( f_{T_1} \) as a function of aging time, \( t \). \( \Delta \) was found to increase until impingement occurred, which is manifested by the sharp curtailment in its increase with \( t \), particularly for the alloys aged at 190 °C (Fig. 1a). The number of \( T_1 \) precipitates per unit volume remained nearly constant (the average values were \( 3 \times 10^{11} \) and \( 4 \times 10^{11} \) mm\(^{-3} \) for the 2-3 alloy aged at 190 °C and 160 °C, and \( 2.5 \times 10^{10} \) and \( 1.7 \times 10^{10} \) mm\(^{-3} \) for the 3-1 alloy aged at 190 °C and 160 °C, respectively), producing the increase in \( f_{T_1} \) with \( t \) seen in figure 1d. These results suggest that the increase in \( \Delta \) and \( Z \) occurs during the growth stage of the \( T_1 \) precipitates, but no attempt was made to confirm this by quantitative analysis of the kinetics of growth.

Fig. 1.—Illustrating the variation of (a) the diameter, \( \Delta \), (b) thickness, \( Z \), (c) aspect ratio, \( \Delta/Z \), and (d) volume fraction, \( f_{T_1} \), of the \( T_1 \) precipitates with aging time, \( t \), in the two ternary Al-Li-Cu alloys studied. Data taken on samples aged at 190 °C for \( t > 50 \) h are not shown.
The values of $\Delta/Z$ appear to pass through a sharp maximum at an aging time between 2 and 4 h (Fig. 1c), suggesting that the Ti precipitates do not acquire their equilibrium shape until well beyond the nucleation stage. The data suggest that the equilibrium value of $\Delta/Z$ lies between 40 and 50. At longer aging times $\Delta/Z$ is smaller still, approaching ~35, but this is probably a consequence of impingement which enables $Z$ to increase while $\Delta$ remains nearly constant. The large value of $\Delta/Z$ indicates that the interfacial energy of the broad faces of the particles (those parallel to $\{111\}$) is substantially lower than that of the peripheral interface. This interpretation is consistent with the low misfit and high degree of coherency of the $(0001)_{\delta'}/\{111\}_{\beta}$ interface[7].

3.2 DISLOCATION MICROSTRUCTURES.—The initial dislocation density in both alloys was approximately $10^8$ mm$^{-2}$; it was introduced during the 5-6% stretch after solution treatment. Typical dislocation microstructures in the underaged unreverted tensile-tested specimens consisted of a minority of pairs of dislocations, with large numbers of single dislocations which are believed to have been generated during the stretch before aging. Long and nearly straight screw dislocation pairs were commonly seen in the deformed samples aged for very short times[5].

The deformation microstructures in the underaged and reverted samples typically consisted of strongly bowed dislocations. They were generally uncoupled, although evidence for dislocation coupling could be found, as shown in figure 2a. The dislocation microstructure in a sample of the 2-3 alloy similarly aged, but not reverted, is shown in figure 2b. The average pair spacing in the reverted samples (Fig. 2a) is quite difficult to measure with assurance because it is so large that pairs are not always easy to identify and recognize as such. This can be attributed to the low volume fraction and very small size of the $\delta'$ particles that precipitate during quenching from the reversion temperature. Despite these uncertainties the spacings of pairs from several micrographs of foils in $[110]$ orientation, such as that in figure 2b, were measured, yielding an average value of 71.3 ± 16.4 nm. A maximum value of the pair spacing, $D$, can be estimated theoretically from the formula[8], valid for screw dislocations,

$$D = \frac{Gb^2}{\pi^2} \left( \frac{32\Gamma}{3\gamma_{apb}f_{\delta'}\langle r \rangle} \right)^{1/2},$$

where $\gamma_{apb}$ is the antiphase boundary energy on $\{111\}$ of the $\delta'$ precipitates, $f_{\delta'}$ and $\langle r \rangle$ are the volume fraction and average radius of the $\delta'$ particles and $\Gamma$ is the line tension of the dislocations of Burgers vector $b$. With $G = 30$ GN/m$^2$[9], $b = 0.2864$ nm, representative values of $\langle r \rangle = 1$ nm and $f_{\delta'} = 0.01$ for the reverted samples, and $\gamma_{apb} = 0.15$ J/m$^2$ and $\Gamma = 8 \times 10^{-10}$ N[5], the value of $D$ from equation (2) is ~125

Fig. 2.—Dark field, weak beam transmission electron micrographs of the deformation microstructures in the 2-3 alloy aged at 160 °C for 1 h: (a) Reverted, illustrating dislocation pairs indicated by the arrows; (b) Unreverted, illustrating the complex dislocation arrays typically observed.
nm, which is only in fair agreement with the measured spacings, but is considerably larger than the pair spacings typical of dislocations in the underaged unreverted 3-1 and 2-3 alloys (35 to 40 nm[4]). A possible source of the discrepancy between the calculated and measured spacings is that equation (2) is valid only when the trailing dislocation in the pair avoids the precipitates[8], which may be difficult when the δ' particles are so small. Intersection of the trailing dislocation with the δ' particles adds a term to the denominator of equation (2), thereby increasing it and lowering D.

The dislocations in figure 2a appear to have interacted strongly with the T1 particles, judging from the considerable bowing and tangling observed. In regions comparatively devoid of T1 precipitates the dislocation density was smaller than average, suggesting that the T1 particles are major sources of pinning. No convincing evidence of shearing of the T1 precipitates was ever found in any of the samples examined, even in samples deformed to fracture. The microstructures of the overaged reverted alloys contained significant amounts of Orowan loops. These were elliptical in shape and most likely associated with the T1 plates, although a one-to-one correspondence could not be established.

3.3 CRITICAL RESOLVED SHEAR STRESS.—Figure 3 shows the variation of the CRSS of the aged samples, \( \tau_t \) (which includes mainly the contributions from the matrix and the δ' and T1 precipitates), and also the CRSS of the reverted samples, \( \tau_r \) (which primarily includes the contributions from the matrix and the T1 precipitates), as a function of aging time. The other two quantities, \( \tau_{ma} \) and \( \tau_{mr} \), are the CRSS of the matrix of the aged and reverted specimens, respectively. Meaningful estimates of these quantities are very

![Graph](image_url)
important to the analysis, but are rather difficult to obtain. The curves in figure 3 differ from those published previously[4] due to revised estimates of $\tau_{ma}$ and $\tau_{mr}[5]$, and include the contribution from the $\delta$ precipitates that nucleate during quenching from the reversion temperature. The magnitudes of $\tau_{ma}$ and $\tau_{mr}$ for the overaged alloys are subject to greater uncertainty than for the underaged and peak aged alloys because they are calculated using estimates of the Cu and Li removed from solid solution by the precipitate phases that form. In the overaged samples the extent of $\delta$ and $T_2$ phase formation is increasingly important, but was ignored in the calculation. The values in figure 3 are therefore overestimates and represent upper limits of $\tau_{ma}$ and $\tau_{mr}$.

The values of $\tau$ were calculated by dividing the tensile yield stress, $\sigma_y$, by the Taylor factor, $M$, using the measured values $M=3.331$ and 2.991 for the 2-3 and 3-1 alloys, respectively. From these data it is seen that the 2-3 alloy is 20-35% stronger than the 3-1 alloy. The maximum values of $\tau$ at 190 °C correlate reasonably well with the onset of impingement of the $T_1$ plates (cf. Fig. 1a). In the overaged alloys $\tau_r$ approaches $\tau_o$, indicating that the role played by the $T_1$ phase is increasingly important to the overall strength. The peak values of both $\tau$ and $\tau_r$ are reached almost at the same aging time.

4. Analysis of Strengthening Mechanisms

The contribution to strengthening of the $T_1$ particles, $\Delta\tau_{T_1}$, can be obtained directly from the curves in figure 3 assuming that the contributions of the precipitates and matrix add linearly, i.e.

$$\Delta\tau_{T_1} = \tau_r - \tau_{mr}. \quad (3)$$

Whereas it is not possible to conclude from the electron micrographs (Fig. 2) whether the $T_1$ particles in the underaged samples are sheared or bypassed, it is nevertheless possible to eliminate all mechanisms of strengthening by particle shearing other than chemical strengthening, for reasons discussed by Huang[10]. For example, with essentially 25 at.% Li in the $T_1$ phase, the difference between the shear moduli of the $T_1$ precipitates and the matrix should be large, and the contribution of modulus strengthening correspondingly significant. However, the modulus hardening theory of Melander and Persson[11] predicts a stress increment for the two alloys which is much smaller than the values actually observed[10].

4.1 CHEMICAL STRENGTHENING.—Of all the possible strengthening mechanisms involving the shearing of precipitates the contribution to the CRSS from chemical strengthening, $\tau_{ch}$, cannot be readily dismissed because of the high surface to volume ratio of the $T_1$ precipitates. The customary theoretical treatment of chemical strengthening, for example that of Brown and Ham[12], invokes Friedel statistics to evaluate the spacing of the precipitates along the dislocation line in the critical cutting configuration. An alternative treatment is possible, however, for strong shearable precipitates. It makes use of the results of the computer simulation experiments of Foreman and Makin[13] and the empirical equation for these results proposed by Brown and Ham[12]. Both of these are considered below.

4.1.1 Weak obstacle statistics.—An expression for $\tau_{ch}$ can be obtained by starting with the equation derived by Brown and Ham[12], i.e.,

$$\tau_{ch} = \left(\frac{4\gamma_i \frac{b}{L_T}}{L_s^2 b} \right)^{1/2}, \quad (4)$$

where $\gamma_i$ is the free energyof the interface created by the passage of a glide dislocation through a $T_1$ particle and $L_s$ is the square lattice spacing defined by $L_s = \frac{n_s^{1/2}}{\gamma_i}$, where $n_s$ is the number of particles per unit area intersecting the slip plane. For parallel plate shape particles intersected by a random plane $n_s$ is given by the equation[14]

$$n_s = \frac{f}{ZA}, \quad (5)$$

which is valid for randomly spaced parallel circular plates. Only three of the four possible variants of $T_1$
precipitates on \{111\} are equivalent obstacles to dislocation motion since they all intersect a specific slip plane at the same angle (70.53°). The fourth variant parallel to the glide plane is a different type of obstacle with a much larger value of \(L_s\) because the probability of intersection of a fairly thin plate by a plane parallel to the plate is far smaller than when the plane and plate are inclined to each other. In the following calculations the strength contribution from the fourth variant is neglected, which appears to be a reasonable assumption. \(L_s\) can then be computed assuming that \(f\) in equation (5) is simply \(3fT_1/4\), producing the result

\[
L_s = \left(\frac{4Z\Delta}{3fT_1}\right)^{1/2}.
\]

Equation (4) thus becomes

\[
\tau_{ch} = \left(\frac{3\gamma_2^2bfT_1}{\Gamma Z\Delta}\right)^{1/2}.
\]

If strengthening from the \(T_1\) precipitates is governed by the chemical strengthening mechanism, then a plot of \(\Delta T_{T_1}\) vs. \(\left(\frac{fT_1}{\Gamma Z\Delta}\right)^{1/2}\) should be linear. The slope of the curve should be equal to \(\left(3\gamma_2^2b\right)^{1/2}\) independently of whether the dislocations are coupled by the fine \(\delta\) particles present. Figure 4a shows such a plot for the data on all the samples aged up to the peak strength condition. The data in figure 1 were used to compute the abscissa, with \(\Gamma\) calculated for screw dislocations in recognition of the observations\[15,16\] that screw dislocations control the flow stress in \(Al\) alloys containing \(\delta\) precipitates. With Poisson's ratio, \(v\), equalling 0.339\[15,17\], it is easy to show that \(\Gamma\) is given by the expression\[12\]

\[
\Gamma = 2.026\frac{Gb^2}{4\pi} \ln \frac{\Lambda}{r_0} ,
\]

where \(\Lambda\) and \(r_0\) (= 2b) are outer and inner cut-off radii. Calculation of \(\Lambda\) is complicated somewhat by the presence of the fine \(\delta\) precipitates, and as a compromise the geometric mean of \(L_s\) for the \(\delta\) particles \((\bar{r}) = 1\) nm and \(f_\delta = 0.01\) and the average "diameter" of the geometric figure created by the intersection of a slip plane and a \(T_1\) precipitate, \((0.84\bar{r})_{0.5}\), was used.

It is evident that linearity is not very well obeyed by the data in figure 4a, as demanded by equation (7), although a straight line can be easily fitted, as shown. The value of \(\gamma_1\) extracted from the slope of the line in figure 4a is 6.58 J/m², and the values obtained by analyzing the data on the individual alloys separately are 4.83 and 4.65 J/m² for the 3-1 alloy aged at 160 and 190°C, and 12.97 and 7.95 J/m² for the 2-3 alloy aged at 160 and 190 °C. These values of \(\gamma_1\) refer to the interface generated by a glide dislocation that shears a \(T_1\) precipitate, and it is implicitly assumed that this interface and the peripheral particle/matrix interface are energetically equivalent. Although the linear behavior predicted by equation (7) is obeyed reasonably well by the individual data sets, the values of \(\gamma_1\) extracted from the slopes of the curves vary significantly for the two alloys, \(\gamma_1\) for the 2-3 alloy far exceeding \(\gamma_1\) for the 3-1 alloy, especially for the samples aged at 160 °C. There is no physical reason why this should obtain, since the equilibrium chemistry of the \(T_1\) and matrix phases are expected to be similar in both of the aged alloys.

There are no reported data on \(\gamma_1\) for the peripheral \(T_1\)/matrix interfaces, but if the equilibrium aspect ratio of the \(T_1\) precipitates is between 40 and 50, as suggested by the data in figure 1c, the average value derived from figure 4a (6.58 J/m²) and most of the individual values may not be as outlandishly high as they seem. The reason for this is that the magnitude of \(\Delta Z\) implies that the coherent or semicoherent (0001)\(T_1\){111}Al interfaces have energies between 0.132 and 0.165 J/m², which is quite a reasonable result.

4.1.2 Strong obstacle statistics.—This approach has not been previously applied to the problem of chemical strengthening because it is usually assumed that the precipitates in question are weak. Since the \(T_1\) precipitates are strong, however, it seems sensible to apply strong obstacle statistics to estimate the strengthening by this mechanism. Brown and Ham\[12\] have shown that the results of the computer simulation experiments of Foreman and Makin\[13\] in the regime of strong obstacles (when the arms of the bowing dislocation subtend an angle less than ~120°) obey the equation
\[
\tau_c = 0.81 \frac{F_m}{b L_s},
\]

where \( \tau_c \) is the CRSS of the array of point obstacles and \( F_m \) is the maximum resisting force exerted by the obstacle on the dislocation. Since \( F_m \) is simply \( 2 \gamma b \sqrt{L_s} \) and \( L_s \) is given by equation (6), the desired equation for the chemical strengthening by strong obstacles assumes the particularly simple form

\[
\tau_{ch} = 1.4 \gamma \left( \frac{f_{T_1}}{Z \Delta} \right)^{1/2}.
\]

Equation (10) predicts that a plot of \( \Delta \tau_{T_1} \) vs. \( (f_{T_1}/Z \Delta)^{1/2} \) should be linear, with a slope equal to \( 1.4 \gamma \). The data in figure 3 are plotted in this way in figure 4b. The straight line shown fits these data better than in figure 4a, but the scatter is nevertheless considerable. The average value of \( \gamma \) obtained from the slope is 9.27 J/m²; those obtained from individual analyses of the data are 6.24 and 5.28 J/m² for the 3-1 alloy aged at 160 and 190°C, and 19.98 and 8.69 J/m² for the 2-3 alloy aged at 160 and 190 °C. As in the previous case the values of \( \gamma \) for the 2-3 alloy exceed those for the 3-1 alloy, with that for the 2-3 alloy aged at 160 °C far larger than the others. Despite the even larger average magnitude of \( \gamma \) (9.27 J/m²) required by the mechanism of chemical strengthening by strong particles, the aspect ratio is consistent with an interfacial energy of the broad interfaces between 0.185 and 0.232 J/m², which is still not at all unreasonable.

4.2 STRENGTHENING BY THE OROWAN MECHANISM.—Although the possibility of chemical strengthening cannot be ruled out by the previous analysis, it is not convincing enough to preclude other possibilities. In particular, if the \( T_1 \) plates are actually non-shearable the observed values of \( \Delta \tau_{T_1} \) could conceivably be characterized by the Orowan equation, even in the earliest stages of aging. Calculating the strengthening due to the Orowan mechanism for plate-like precipitates is not as straightforward as that for spherical particles. The only published equation for an aligned array of plate-shaped particles is due to Kelly [18], whose formula for the CRSS was derived for the case in which only one set of randomly spaced parallel plates is intersected by the slip plane. An alternative derivation is presented here which circumvents the difficulty of calculating values of \( L_s \) using specific, regular geometric arrays of plate-shaped precipitates.

The derivation starts with the Orowan equation for point obstacles[8,12], i.e.

\[
\tau_0 = \frac{0.81(2 \Gamma)}{bL_s},
\]

\[\text{Fig. 4.—Analysis of strengthening by the } T_1 \text{ precipitates, } \Delta \tau_{T_1}, \text{ according to two versions of the theory of chemical strengthening: (a) Weak obstacle statistics, equation (7); (b) Strong obstacle statistics, equation (10).}\]
where $\tau_o$ is the Orowan stress. Comparison with the equilibrium equation for the critical stress on an impenetrable obstacle, $2T/bL$, where $L$ is the obstacle spacing along the dislocation, establishes that the average spacing of the obstacles in the computer simulation experiments is simply $L_s/0.81$, which will be referred to here as the Orowan spacing. To account for the finite size of the $T_1$ particles the "effective" Orowan spacing becomes

$$L_o = \frac{L_s}{0.81} - (Z\Delta)^{1/2},$$

where $Z\Delta$ is the average area intersected by a random plane through an array of parallel plate-shaped precipitates of diameter $\Delta$ and thickness $Z[14]$. In keeping with previous estimates of the Orowan stress[12] the line tension for impenetrable particles is taken as

$$\Gamma = \frac{Gb^2}{4\pi(1-\nu)^{1/2}}\ln\frac{\Delta}{\pi b},$$

where $\Lambda = 2\Delta/\pi$, as suggested by Kelly[18]. Using equations (11) to (13), in conjunction with (6), the expression for the Orowan stress for plate-shaped precipitates on $\{111\}$ becomes

$$\tau_o = \frac{Gb}{2\pi(1-\nu)^{1/2}}\left(\frac{f_{T_1}}{Z\Delta}\right)^{1/2} \ln \frac{(\Delta/\pi b)}{1.43 - \sqrt{\frac{f_{T_1}}{Z\Delta}}}.$$  

Figure 5a shows a plot of $\Delta \tau_{T_1}$ vs. $(f_{T_1}/Z\Delta)^{1/2} \ln (\Delta/\pi b)$, excluding the data on the overaged samples. In the analysis that follows $(f_{T_1})^{1/2}$ has been neglected in the denominator of equation (14) since it is relatively small for all aging conditions. The slope of the line in figure 5a is $2.615 \times 10^{-6}$ MPa·m (the individual values are $1.639 \times 10^{-6}$ and $1.377 \times 10^{-6}$ MPa·m for the 3-1 alloy aged at 160 and 190 °C, and $4.671 \times 10^{-6}$ and $2.522 \times 10^{-6}$ MPa·m for the 2-3 alloy aged at 160 and 190 °C). The theoretically predicted value of the slope is $2.352 \times 10^{-6}$ MPa·m, provided that $b$ is replaced by $2b$ to allow for the dislocations travelling in pairs[19], as seen in figure 2b. Under these circumstances the agreement must be regarded as excellent.

If the Orowan mechanism is valid for the underaged alloys it should be equally applicable to the data on the overaged alloys. Figure 5b displays all the data, plotted in the same manner as the data in figure 5a.
The scatter is clearly significantly greater than when just the data on the underaged samples are considered, but when a linear fit is forced the resulting slope is 1.954 x 10^-6 MPa m, still in excellent agreement with the theoretically predicted value.

5. Discussion

The analyses presented in the preceding section are realistic evaluations of the strengthening contributions of the T1 precipitates to the strength of the two ternary Al-Li-Cu alloys studied. Unfortunately, there remain uncertainties that prevent a definitive choice between the two most promising candidate mechanisms. If the aspect ratio of 40 to 50 indeed reflects the equilibrium shape of the T1 particles then the very large values of \( \gamma \) (6.58 J/m\(^2\) from figure 4a or 9.27 J/m\(^2\) from figure 4b) derived from consideration of the chemical strengthening mechanism become more reasonable than one would be inclined to suspect on first encounter. There is some evidence that \( \Delta Z \) decreases with increasing aging time (Fig. 1c), but this is probably an effect of impingement rather than the approach to equilibrium shape, as already noted. It seems unlikely that \( \Delta Z \) would diminish so significantly from the measured values that the values of \( \gamma \) derived from the data would be substantially reduced. In this respect the large values of \( \gamma \) are not so unreasonable, and a more compelling argument against the operation of either version of the chemical strengthening mechanism are provided by the observations that the data in figure 4 simply do not obey equations (7) or (10) particularly well.

The viability of the Orowan mechanism depends on the coupled motion of the dislocations that attack the T1 precipitates. The microstructural evidence for this is not widespread, which is not surprising considering that the alloys were stretched 5 to 6% following solution treatment. The dislocations introduced thereby stimulate the nucleation of the T1 particles, but also make the microstructure more messy than that encountered in an unworked precipitation hardened alloy. Notwithstanding this difficulty, the reverted samples always contain a dispersion of very fine \( \delta' \) precipitates which nucleate on cooling from the reversion temperature, and these are undoubtedly responsible for the dislocation pairs observed in figure 2b. The question that then remains is whether the coupling produced by this dispersion is strong enough to withstand the turbulent disruption caused by interaction with the T1 precipitates.

To address this issue we consider the criterion of Ashby[19], who treated the problem of Orowan strengthening of an ordered matrix, in which the dislocations typically travel in closely coupled pairs. He demonstrated that so long as the separation of the dislocations was small compared to the planar particle spacing, the dislocation pair could be regarded as a single dislocation of Burgers vector \( 2b \). The Orowan spacings are readily estimated from equations (6) and (12); they vary from 270 to 600 nm for the 2-3 alloy and from 740 to 3500 nm for the 3-1 alloy over the range of aging times studied. These far exceed the value of \( D \) calculated using equation (2) (~125 nm) and those measured experimentally, leaving little doubt that the criterion of Ashby is satisfied.

The sizeable scatter apparent in figure 5b undoubtedly detracts from an interpretation favoring the validity of the Orowan mechanism. It is important to reiterate, however, that the values of \( \Delta T_{T1} \) used in the analysis were derived using calculated values of \( \tau_{mr} \) estimated without taking into account the additional depletion of Li and, more importantly, Cu associated with the precipitation of continuously increasing quantities of the T2 and \( \delta \) phases. The growth of the Cu-rich T2 phase, in particular, depletes the matrix of copper, weakening it far more than an equivalent loss of lithium because of the greater solid solution strengthening effect of Cu[5]. Consequently the values of \( \tau_{mr} \) would be smaller at long aging times than those depicted in figure 3, and the values of \( \Delta T_{T1} \) for the overaged samples correspondingly larger, bringing them into better agreement with the other data. Unfortunately, quantitative information on the relative proportions of the T2 and \( \delta \) phases as a function of aging time is unavailable, so the foregoing suggestion cannot be tested quantitatively. Nevertheless, the T2 and \( \delta \) phases are known to be present in increasing amounts when the alloys are in the overaged condition[10], lending additional credibility to the analysis of the data in terms of the Orowan mechanism. A definitive choice between the two viable mechanisms must, however, await additional experimentation or a more refined analysis.

6. Summary

The results of this investigation demonstrate that the contribution to strengthening by the T1 precipitates is
potent, particularly in the 2-3 alloy, considering that it is present in minor amounts compared to the 6' precipitates. In discussing the attempts to evaluate the contribution of the T1 precipitates quantitatively, however, two major sources of uncertainty should be kept in mind. One is that the strengthening contribution of the T1 variant parallel to the slip planes has been ignored, and the other is that the contribution of the 6' precipitates that nucleate during the quench from the reversion temperature has been incorporated into the estimates of $\tau_{ma}$ and $\tau_{mr}$, implicitly assuming that it is small enough to be regarded as linearly additive[12] to the other contributions.

Analysis of the data indicates that the chemical strengthening and Orowan strengthening mechanisms are both viable candidates for the dominant mechanism of strengthening by the T1 precipitates. The former requires very large values of the incoherent interfacial free energy of the peripheral interface of the plates, either $-6.58$ or $-9.27$ J/m$^2$ depending on whether the theory of chemical strengthening is formulated in terms of weak or strong obstacle statistics. Either of these values would probably be completely unrealistic were it not for the extremely large aspect ratio (40 to 50) of the T1 precipitates. If this aspect ratio is indicative of the equilibrium shape of the T1 precipitates, which seems quite likely, then the broad faces of the precipitates have interfacial energies in the range 0.13 to 0.23 J/m$^2$. These are realistic values for the coherent or semicoherent (0001)$_T$/\{111\}_A1 interfaces.

The success of the analysis in terms of the Orowan mechanism requires that the dislocations travel in pairs. This is likely in the reverted samples because of the ease with which 6' precipitates nucleate during the quench from the reversion temperature, and microstructural evidence has been found for its occurrence. Computation of the expected pair spacing indicates that D is much smaller than $b$ in calculating the Burgers vector $2b$ in calculating the Orowan stress.

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References

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