COMMENTS ON THE “DISLOCATION INTERACTION WITH SEMICOHERENT PRECIPITATES (Ω PHASE) IN DEFORMED Al-Cu-Mg-Ag ALLOY”

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Introduction

In a recent paper, Li and Wawner [1] reported results of a study of the interaction between dislocations and \( \{111\}_\alpha \) precipitate plates of \( \Omega \) (Al\(_2\)Cu) phase in a plastically deformed Al-Cu-Mg-Ag alloy using both conventional and high resolution transmission electron microscopy (HRTEM). Based on experimental observations that \( \Omega \) phase was sheared in plastically-deformed samples, they propose a strengthening model for aluminium alloys containing \( \{111\}_\alpha \) precipitate plates. This model is based essentially on the order strengthening mechanism, currently accepted for spherical particles [2,3], and includes increments in strength arising from both order strengthening and interfacial (chemical) strengthening for sheared precipitates. However, it does not take into account the plate-like shape and orientation of the \( \Omega \) precipitates. Despite this major deficiency, the model is claimed to be applicable not only to \( \Omega \) precipitate plates in Al-Cu-Mg-Ag alloys, but also to other \( \{111\}_\alpha \) precipitate plates, such as \( T_1 \) (Al\(_2\)CuLi) phase in Al-Cu-Li alloys [4,5]. The aim of this comment is to demonstrate that this model is an oversimplification and inevitably incorrect.

Critical Thickness of \( \{111\}_\alpha \) Precipitate Plates

Li and Wawner [1] claim that there exists a critical thickness of \( \{111\}_\alpha \) precipitate plates, below which the plates are sheared during plastic deformation and above which a transition from dislocation shearing to dislocation (Orowan) looping occurs. They suggest that it is a reduction in the critical thickness of \( \{111\}_\alpha \) plates that is a major factor in strengthening the alloy. This notion of a critical size for dispersed strengthening particles was originally proposed for spherical particles [6,3]. If \( \Delta \tau_s \) represents the increment in critical resolved shear stress (CRSS) required for particle shearing and \( \Delta \tau_o \) is the CRSS increment produced by Orowan looping, then, for a given volume fraction of spherical particles, it is proposed [3] that there is a critical particle diameter \( d_c \) such that

\[
\frac{\Delta \tau_s}{\Delta \tau_o} < 1 \quad \text{when particle diameter } d < d_c, \quad \text{and} \\
\frac{\Delta \tau_s}{\Delta \tau_o} > 1 \quad \text{when } d > d_c.
\]

A transition from dislocation shearing to dislocation looping thus occurs when the diameter of spherical particles increases above the critical size [3].
In the case of plate-shaped particles, Li and Wawner [1] claim that the critical particle size corresponds simply to the thickness of the precipitates. However, such an assumption ignores the plate-like form and the effect that the aspect ratio of the precipitate plates may have on any strengthening increment [5,7–11]. For example, for a face-centred cubic matrix containing shearable \{100\}_\alpha \text{ precipitate plates, it has been calculated [9,10] that the contribution of order strengthening to CRSS is given by an equation of the form:}

$$\Delta \tau_o = \left( \frac{\gamma_{\text{apb}}}{2b} \right) \left( \frac{2f \gamma_{\text{apb}}}{\pi T} \left( \frac{A \sqrt{6}}{3} \right)^{1/2} \right)^{1/2} - C \Gamma,$$

(1)

where \(t\) and \(A\) are the thickness and aspect ratio respectively of the plates, \(b\) is the Burgers vector of glide dislocations in the matrix phase, \(f\) is the volume fraction of precipitates, \(\Gamma\) is the dislocation line tension in the matrix phase, \(\gamma_{\text{apb}}\) is the specific anti-phase boundary energy on the slip plane of the precipitate phase, and \(C\) is a constant. Similarly, it has been proposed [11] that the contribution of shear-resistant \{100\}_\alpha \text{ precipitate plates to CRSS via Orowan looping is given by:}

$$\Delta \tau_o = \frac{G_m b}{2 \pi (1 - \nu)} \left( - \ln \frac{1.225}{\frac{D}{r_0}} \right),$$

(2)

where \(G_m\) is the shear modulus of the matrix phase, \(\nu\) is Poisson’s ratio, and \(r_0\) is the core radius of the dislocations. For given values of \(f\) and \(\gamma_{\text{apb}}\), the ratio \(\Delta \tau_o/\Delta \tau_o\) must thus be a function of the thickness and the aspect ratio (diameter) of the precipitate plates. A similar analysis is undoubtedly applicable to \{111\}_\alpha \text{ plates.}

It is thus clearly an oversimplification to propose [1] that there exists a simple critical thickness for precipitate plates, above which the ratio \(\Delta \tau_o/\Delta \tau_o\) >1 and there is a transition from dislocation shearing to Orowan looping. If indeed such a transition occurs, then it will be at a value of resolved shear stress that is determined by both plate thickness and diameter (aspect ratio). It is perhaps to be emphasised that Li and Wawner [1] produce no direct evidence that such a transition does occur for plates of \(\Omega\) phase; their experimental evidence focuses on establishing that such plates are sheared. In unpublished research, the present authors have observed no such transition in the deformation behaviour of \(T_1\) plates in Al-Cu-Li alloys; such plates are sheared in both underaged and overaged microstructures.

It should also be noted that Equation 2 in [1] is incorrect as reproduced. It was developed from the order strengthening equation for alloys containing spherical particles [2]. Assuming that the shear force required to disorder an ordered precipitate phase across the slip plane is equal to \(G_m b^2\) [12], the appropriate form of this approximate relationship for the critical diameter is:

$$d_c = \frac{4G_m b^2}{\pi \gamma_{\text{apb}}}.\tag{3}$$

To take an additional contribution of interfacial strengthening into account, Li and Wawner [1] suggest that Equation 2 in their paper can be modified simply by replacing \(\gamma_{\text{apb}}\) by the sum of \((\gamma_{\text{apb}} + \gamma_i)\), where \(\gamma_i\) is the specific interfacial energy between precipitate and matrix phases for the new interface created by shearing. However, this is again incorrect for both spherical and plate-shaped particles. For spherical particles, the force required to create additional precipitate-matrix interface and simultaneous disorder of an ordered precipitate phase has been shown [2] to be given approximately by an equation of the form:

$$F = 2b \gamma_i + \frac{\pi}{4} d \gamma_{\text{apb}}.\tag{4}$$
Assuming that this shear force is equal to $G_m b^2$ [12], the critical diameter of such ordered spherical particles is then given as:

$$d_c = \frac{4b(G_m b - 2\gamma_d)}{\pi Y_{sb}}.$$  \hfill (5)

**Interface Structure and Precipitate Shearing**

Li and Wawner [1] consistently describe the $\Omega$ phase (and the $T_1$ phase) precipitates as semicoherent and particular examples of a class of semicoherent particles. However, this description represents an oversimplification which is misleading. Both of these precipitate phases form as thin, hexagonal-shaped plates of large aspect ratio (typically 40:1) on $\{111\}_m$ planes [13–16]. Each is nearly fully coherent with the aluminium matrix phase across the broad faces of the plates and, in the case of $T_1$ phase, the plates are near perfectly coherent with the matrix phase normal to the habit plane. In the case of the $\Omega$ phase, there is a misfit of $\sim 9.3\%$ normal to the habit plane [13] and thus those interfaces that define the rim of the plates will become partially coherent and include misfit-compensating dislocations at some finite thickness [14,15]. When a gliding dislocation interacts with a plate of $\Omega$ phase (or $T_1$) phase in the manner depicted in Figure 1 of [1], the dislocation encounters the plate across a broad planar interface that is near fully coherent.

Li and Wawner [1] suggest that, when a coherent precipitate is sheared by a gliding dislocation, the resulting precipitate-matrix interface that is created is inevitably coherent. They fail to recognise that this is only generally the case when the crystal structures of precipitate and matrix are similar. They further suggest that the situation is different for a “semicoherent precipitate with a different crystal structure from matrix,” and that the new interface created upon shearing of the precipitate is inevitably a high energy interface. But this is not necessarily the case. For an $\Omega$ phase plate, the interface that is encountered by a gliding dislocation is itself coherent. The structure of the segments of new interface that are created by shearing of such a plate will depend on the Burgers vector of the shearing dislocation and the plane and direction of shear, i.e. on the crystallography of the displacement and the orientation of the precipitate-matrix interface that emerges.

Li and Wawner [1] claim that shearing of $\Omega$ precipitates occurs in the $[001]_\Omega$ direction on the $(110)_\Omega$ plane, as in [17]. The slip system reported in [17] is for monolithic $\theta$ phase ($\text{Al}_2\text{Cu}$), which is the
equilibrium intermetallic phase in binary Al-Cu alloys. Recent studies indicate that the $\Omega$ phase is metastable, with an orthorhombic (or tetragonal) structure [13,18,19], and that it transforms [20] to the $\theta$ phase only after prolonged ageing at elevated temperatures. Since the $\Omega$ phase is structurally distinguishable from the phase $\theta$, it is possible that the deformation behaviour of the $\Omega$ phase will be different from that of the $\theta$ phase. Furthermore, when constrained in a matrix of aluminium, the deformation modes of both intermetallic phases could be very different from those in monolithic phases without such constraint.

A discrepancy is evident between the claimed [001](110)$_\Omega$ slip system and the apparent [001](100)$_\Omega$ slip system shown schematically in Figures 10–12 in [1]. These schematic diagrams are again not self-consistent and, more importantly, they are not supported by experimental evidence. According to Figure 10 in [1], two adjacent segments of a sheared $\Omega$ plate are displaced by $c/3[001](100)_\Omega$ after the plate is sheared by a single dislocation, and further shearing by two dislocations at the same location will yield a displacement of one unit cell height between the two segments of the $\Omega$ plate. This is contrary to the displacement of $5c/6[001](100)_\Omega$ that is shown in Figure 11 [1]. From the HRTEM image shown in Figure 8 in [1], it is difficult to be convinced that the shear plane in the $\Omega$ phase is parallel to (100)$_\Omega$. As shown schematically in the present Figure 1, the shear plane of an $\Omega$ plate will be revealed unambiguously in HRTEM images only when the habit plane and the shear plane of the $\Omega$ phase are simultaneously parallel to the electron beam direction. If the intersection of the habit plane and the slip plane of a (111)$_\alpha$ plate is assumed to be parallel to [101]$_\alpha$, then, when viewed in [110]$_\alpha$ or [011]$_\alpha$ directions, the shear plane in the (111)$_\alpha$ plate in Figure 1 will be inclined with respect to the electron beam direction. In the latter two imaging conditions, it becomes difficult to determine reliably the shear plane in the precipitates, although the existence of shearing of the precipitates can be confirmed unambiguously.

**Specific Interfacial Energy**

Li and Wawner [1] note that the displacements between any two adjacent segments of sheared $\Omega$ plates are commonly small and infer that, once sheared, the plates become more difficult to shear on the same slip plane. They interpret this increased resistance to shearing in terms of a reduction in critical thickness and attribute this reduction to a continuous increase in the specific interfacial energy $\gamma_i$ of the new interface between precipitate and matrix phases created by shearing. Whether one uses the modified form of equation 2 in [1] or the corrected equation 5 above, an increase in $\gamma_i$ is seen to lead to a decrease in $d_c$. To rationalise this model, Li and Wawner [1] treat the displacements created at the coherent broad faces of the plates by a single shearing dislocation as misfit-compensating defects (dislocations). They argue that, with repeated shearing, the Burgers vector of each such dislocation will increase in magnitude continuously and that this means a continuous increase in the structural (misfit) component of the interfacial energy (Fig. 13 in [1]).

Various aspects of this model are of questionable foundation. As demonstrated clearly above, it is an oversimplification to propose that there exists a simple critical thickness for $\Omega$ plates. Furthermore, even if it is accepted that the shear displacements at the initially coherent interfaces be modelled as misfit-compensating defects, continued shearing of a precipitate plate will not change the magnitude of the Burgers vector of such misfit-compensating dislocations; it will only increase the number of such dislocations. The mismatch between the precipitate and matrix lattices at the newly-created precipitate-matrix interface could be accommodated either by an appropriate array of dislocations or by a local rotation of the matrix lattice. Li and Wawner [1] ignore observations [21] that, during plastic deformation of a superalloy strengthened by $\{111\}$ plates, a rotation of the matrix lattice occurs in regions close to the new interface created by dislocation shearing to accommodate the mismatch at that
interface. Regardless of the form of accommodation, it is to be expected that, once the repeat structure of this new interface is established, the specific interfacial energy of this interface will be defined and remain a constant. Repeated shearing will increase the interfacial area and thus the total interfacial energy, but it will not increase the interfacial energy per unit area of interface.

Regardless of the precise identity of the slip system in the $\Omega$ phase, there seems no doubt that the slip plane and direction are not parallel to those in the matrix phase. If such is the case, a misfit strain will inevitably be generated at the sheared interface after an $\Omega$ plate is sheared by a single dislocation. With repeated shearing, the apparent mismatch between the lattices across the new interface will increase, but the misfit strain will remain constant. The accumulating mismatch will be periodically accommodated by misfit-compensating dislocations. Since the structural component of the specific interfacial energy is a function of misfit strain [22], it is to be expected that by definition the specific interfacial energy will remain constant.

It should also be noted that Equation 3 in [1] was originally developed [23] to estimate the specific interfacial energy across the habit plane of truly semi-coherent martensite plates. Even if this equation is applicable to the $\Omega$ phase, the calculated specific interfacial energy would represent that on the habit plane, rather than that on the newly-created precipitate-matrix interface.

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References