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# Temperature dependent slip mode modification in Cu–Al solid solution alloy single crystals

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ABSTRACT

Slip mode modification of planar slip Cu–Al single crystals with low stacking fault energy was studied at various temperatures. The changes of surface slip appearance and dislocation substructure with temperature was observed and explained in terms of the roles of stacking fault energy and dynamic strain aging.

## 1. Introduction

The effects of the stacking fault energy on the slip mode and dislocation structure were studied extensively and stacking fault energy has been used as the main criterion to predict the slip mode [1–8]. As the stacking fault energy decreases, the separation between partial dislocations increases and cross slip becomes more difficult. In materials with low stacking fault energy, the planar array of dislocations confined in a slip plane increases [3] and the distribution of dislocations on individual slip planes is more uniform. In metals with high stacking fault energy, dislocations are arranged in dislocation tangles mainly as a result of extensive cross slip [1–3]. In metals with high stacking fault energy, the motion of dislocations in the three-dimensional structure is relatively free and they tend to cluster together to lower the strain energy of individual dislocation. Therefore, dislocations usually form dislocation tangles and/or cell structure and the distribution of dislocations is very uneven.

Hong and Laird [5] developed a model of planar slip behavior based on the idea that the joining of partials is resisted by the frictional effects due to alloying as well as the stacking fault energy. It was suggested that planarity of slip is promoted not only by low stacking fault energy but by increase in shear modulus, atomic size misfit and solute content. The critical solute concentration predicted by their model [5] for the transition from wavy slip to planar slip was in good agreement with the observations for copper base solid solution alloys and more complex alloys. Despite the extensive studies of the effect of stacking fault energy and alloying on the slip mode [1–10], their effects on the slip mode at elevated temperatures are not well understood. In some alloys, dynamic strain aging (DSA) plays an important role in deformation at some intermediate temperatures [11–15], making it more complicated to predict the slip mode and dislocation substructure. In this paper, the changes of slip mode and the deformation behavior were studied as a function of temperature in Cu–16 at% Al alloy crystals with a low stacking fault energy.

## 2. Experimental procedures

Cu–16 at% Al single crystals with a low stacking fault energy were grown by the Bridgman method, annealed to remove defects, checked for orientation by Laue technique. Cu–16 at% crystals used in this study was observed to have a single slip orientation ($[\bar{1}\,3\,5]$ parallel to the stress axis) [6]. They were electro-machined carefully with a gage section of about 6 mm × 6 mm × 10 mm, electro-polished, and compressed using a UTS testing machine at a strain rate of $4\times10^{-4}$ and at temperatures from room temperature to 550 °C. The surface slip morphology was observed by SEM (Hitachi S-4800) and dislocation structure of the electro-polished TEM samples from the deformed specimens was observed by a TEM (JEOL-ARM200F).

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## 3. Results and discussion

One interesting feature in Cu–Al alloy is that the dynamic solute-dislocation interaction [5,6,12] plays an important role at intermediate temperatures, exhibiting a yield stress plateau. Therefore, no pronounced change of the yield stress is expected in the temperature of dynamic solute-dislocation interaction. Fig. 1 displays the schematic plot of flow stress versus temperature as observed in many alloys [11–12]. In the region of yield stress plateau, the flow stress consists of three stress components, athermal stress component $\tau_G$, thermal stress component $\tau^*$, and stress component associated with dynamic solute-dislocation interaction (i.e., dynamic strain aging), $\tau_D$. The plateau corresponds to the superposition of the frictional stress due to DSA increasing with temperature and the other components of flow stress monotonically decreasing with temperature mainly due to the decline in the thermal stress component.

At temperature $T_D$ where the frictional stress due to DSA reaches the maximum, the thermal stress component $\tau^*$ is very small. On the other hand, at room temperature, the frictional stress $\tau_D$ due to DSA is negligible because DSA is just beginning to work for most alloys [12]. It is interesting to note that the magnitude of thermal stress component, $\tau^*$ at room temperature in Cu–Al is close to the magnitude of the maximum frictional stress due to DSA, $\tau_D$ (Fig. 1) [5–8, 12]. The major strengthening, therefore, is attributed to the athermal and thermal stress components at room temperature and to the athermal and DSA component at the temperature of maximum DSA. Since the major strengthening components at low temperatures are different from those at the temperature of significant DSA, the different strengthening and slip modes are expected to work at room temperature and at the temperature of significant DSA.

Fig. 2 exhibits the stress–strain responses of Cu–16 at% Al single crystals. As expected, the yield stress does not change appreciably with temperature from room temperature up to 450 °C. The yield stress decreased appreciably at 550 °C. For Cu–Al, the plateau in the flow stress versus temperature diagram starts from room temperature [5,6,12]. It should be noted that the totally different strengthening behaviors were observed despite the almost the same yield stress level at various temperatures. The stress–strain responses at room temperature and 350 °C exhibit the typical easy glide behavior of planar slip alloys in stage I followed by stage II with a higher strain hardening. The stage I at low temperatures exhibits almost the zero slope associated with the spreading of a Lüders band along the crystal [12]. The hardening behaviors at 400 °C–450 °C are different from that at room temperature despite the similarity of the yield stress level. The stress serrations were noticeable at 400 °C–450 °C compared to the rather smooth stress–strain curves at room temperature and 550 °C. As explained, at intermediate temperatures (400 °C–450 °C), the contribution of thermal stress component decreased significantly and the DSA stress component increased noticeably. The different deformation and hardening behaviors at low and intermediate temperatures can be attributed to the difference of the constituent stress components.

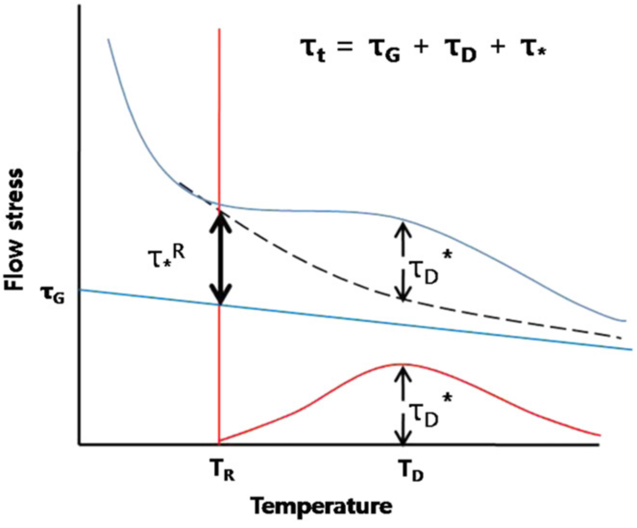

**Fig. 1.** Schematic plot of flow stress versus temperature as affected by various hardening contributions. The flow stress expected in the absence of dynamic strain aging appears as the broken line.

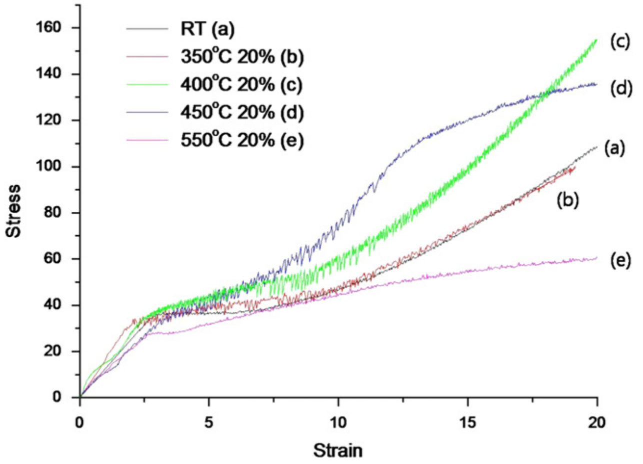

**Fig. 2.** Stress–Strain responses of Cu–16 at%Al single crystals at various temperatures.

The higher hardening rates for the deformation at 400 °C and 450 °C in stage I and stage II can be associated with the greater contribution of the dynamic solute-dislocation interaction [12]. It is well known that solute atoms become mobile above a certain temperature and will tend to interact with moving dislocations, which not only increases the dislocation drag stress but also the dislocation accumulation rate [5,10–13]. At 450 °C, stage III hardening due to the thermally activated cross slip sets in at 12.5% strain. At higher temperatures (550 °C), the mobility of solute atoms become too high to exert a drag force, resulting in the decrease of the DSA stress component as shown in Fig. 1 [11–13], resulting in the lower hardening rate. The increasing contribution of dynamic recovery associated with cross slip at 550 °C is thought to further decrease the hardening rate [15].

The dislocation substructures and the slip modes of alloys can also be linked to the hardening behaviors [5]. The slip morphology of the specimens deformed to 20% strain at various temperatures were observed by SEM. Figs. 3((a)–(d)) display the surface slip morphology of the specimen deformed at room temperature (a), 350 °C (b), 450 °C (c) and 550 °C (d). At room temperature, the surface slip appearance was observed to be coarse as in the typical alloys with low stacking fault energy [5]. The slip line density appears to increase and become more homogeneous with increasing temperature to 350 °C, which is reasonable because of easier cross slip due to thermal activation at a high temperature. At 450 °C, however, the slip lines were shown to bunch together and the slip line density decreased, suggesting the slip is more localized in the existing slip bands at the same total strain. The change of slip behaviors with temperature can be explained in relation to the change of stress–strain responses. Hong [11] suggested that dynamic solute-dislocation interaction tend to enhance the slip inhomogeneity and dislocation clustering in the region of flow stress plateau. The localized slip morphology