Shear Band Formation in Al-2mass%Cu Single Crystals Containing ϑ Precipitates

By K. Morii*, H. Terada** and Y. Nakayama*

The effect of ϑ precipitates and deformation at elevated temperatures on the shear band formation has been investigated in rolled single crystals of an Al-2%Cu alloy based on observations of slip bands and dislocation structures. Irrespective of heat treatments to give either a nominal solid solution or ϑ precipitates, shear bands were not formed in (011)[$\bar{1}00$] crystals but in (211)[$\bar{1}11$] crystals during rolling at room temperature. Increase in the temperature up to 473 K resulted in rather homogeneous slip and developed no shear bands even in the (211)[$\bar{1}11$] crystals. A simple correlation between the shear band formation and dislocation structures was found; the shear bands occurred in the crystals in which aligned micro-bands (layered dislocation walls) evolved.

(Received June 9, 1986)

Keywords: shear bands, localized slip, aluminum-alloy, ϑ -precipitates, rolling deformation

I. Introduction

Shear bands have recently aroused increasing interests, since they are recognized as an important mode of deformation or failure⁽¹⁾⁽²⁾. Many works have shown shear bands formed in different kinds of materials under various deformation conditions⁽³⁾⁽⁴⁾. On account of recent studies on transmission electron microscopy, the formation of shear bands has well been understood and documented for single phase FCC metals and alloys deformed mainly by rolling⁽⁵⁾⁻⁽⁸⁾. In these materials two types of shear bands are usually found depending on stacking fault energy (SFE) and deformation temperature. Shear bands of the first type are derived from a microstructure with aligned mechanical twin lamellae. The bands can be observed in materials with low SFE such as copper alloys with high solute contents, silver and copper rolled at 77 K, and are referred to as those of alloy type. Shear bands of the second type are formed in a dislocation structure of aligned micro-bands⁽⁵⁾ (layered

dislocation walls⁽⁸⁾). Such shear bands are observed in copper and some aluminum alloys after rolling at room temperature, and may be called those of pure metal type.

Based on such observations, a microstructural view for shear band formation has recently been presented by Morii, Mecking and Nakayama⁽⁹⁾; the presence of aligned lamellar obstacles such as mechanical twins and layered dislocation walls is essential for the nucleation of shear bands in single phase materials during rolling deformation.

In this respect, the shear band formation in materials containing a second phase seems to be attractive, since a microstructure evolved during deformation can be influenced by the presence of the second phase, as well as by SFE and the geometry of slip systems. The aim of the present work is to investigate the effects of precipitates such as G.P. zones, ϑ' platelets and ϑ particles on the shear band formation in rolled single crystals of Al-Cu alloys. This report describes the results on single crystals of an Al-2%Cu alloy containing the ϑ (CuAl_2) phase, while those as to G.P. zones and ϑ' precipitates will be treated separately⁽¹⁰⁾.

II. Experimental

An ingot of an Al-2%Cu alloy prepared from high purity Al and Cu (99.99%) was

* College of Engineering, University of Osaka Prefecture, Mozu-Umemachi, Sakai 591, Japan.

** Graduate Student, University of Osaka Prefecture, Mozu-Umemachi, Sakai 591, Japan. Now with Hitachi Microcomputer Engineering, Kodaira, Tokyo 187, Japan.

Table 1 Chemical composition (mass%).

Cu	Fe	Si	Mn	Al
1.80	Tr.	0.02	Tr.	Bal.

cold-rolled to a sheet with 3 mm thickness. The chemical composition of the alloy is given in Table 1. Single crystal bars with (211) and (011) planes parallel to their surfaces were grown in vacuum ($<1.3 \times 10^{-3}$ Pa) by the modified Bridgman method. After homogenization annealing for 6 h (21.6 ks) at 873 K, they were spark cut into small blocks 10 mm(W) $\times$ 8 mm(L) $\times$ 2 mm(T) in size having two different orientations (211)[$\bar{1}11$] and (011)[$\bar{1}00$]. For convenience, an expected slip geometry of these orientations is illustrated in Fig. 1.

These blocks were solution treated for 48 h (172.8 ks) at 773 K followed by water quenching. Subsequent aging was done at 623 K for 48 h (172.8 ks) to produce the ϑ phase precipitates⁽¹⁾. Both as solution treated and aged specimens were subjected to the following experiments.

Deformation was imposed by rolling with a laboratory mill with 53 mm diameter rolls at a speed of 2.0 mm/s at room temperature (RT) and 473 K. For rolling at an elevated temperature specimens were re-heated for 100 s at the temperature between successive rolling passes, and the rolls were heated at the temperature by small furnaces. The orientation of specimens was determined before and after the deformation by X-ray pole figures. Slip band patterns were revealed by polishing the cross section of a pre-rolled specimen and imposing small amounts of rolling strains, and were observed

by an optical microscope. Transmission electron microscopy was also carried out at 100 kV in some foils prepared parallel to the longitudinal section of rolled specimens.

III. Results

The measurements of X-ray pole figures indicated that the initial orientations, (211)[$\bar{1}11$] and (011)[$\bar{1}11$], were retained almost stably up to about 80% reduction for the specimens examined irrespective of the heat treatments and rolling temperatures. The present aging treatment corresponded to "over aging"; the precipitates were considered to be mostly the ϑ phase, though a limited amount of ϑ' may also be present. An example is shown in Fig. 2(a); a mean separation of the precipitates was about 20 μ m.

1. Observation of slip bands

(1) Rolling at RT

At low rolling reductions $<35\%$, crystallographic slip bands expected from Fig. 1

Fig. 1 Schematic illustration of slip geometry of (a) (211)[$\bar{1}11$] and (b) (011)[$\bar{1}00$] specimens.

Fig. 2 Optical micrographs of (a) precipitates after aging for 48 h (172.8 ks) at 623 K, and slip bands of aged specimens; (b) (211)[$\bar{1}11$] rolled 35% and (c) (011)[$\bar{1}00$] rolled 28% at RT, longitudinal section.

were found to occur in every specimen examined, as shown in Fig. 2. With increasing rolling strain, the (011)[$\bar{1}00$] specimens were deformed by slip on two symmetrical planes (Fig. 1(b)), and no remarkable inhomogenieties of slip appeared in both as solution treated and aged specimens (Fig. 3). In contrast to the (011)[$\bar{1}00$], the (211)[$\bar{1}11$] specimens showed anomalous slip bands after rolling of about 50% reduction. Typical examples are shown in Fig. 4 for aged and in Fig. 5 for as solution treated specimens. The anomalous slip bands indicated by arrows in Fig. 4(a) occurred deviating from their slip planes. Such bands grew to form shear bands as the rolling strain

Fig. 3 Slip bands in (011)[$\bar{1}00$] specimens rolled at RT; (a) aged specimen rolled 77% and (b) solution treated and rolled 70%, longitudinal section.

Fig. 5 Shear bands in solution treated (211)[$\bar{1}11$] specimens rolled at RT to (a) 50% and (b) 67%, longitudinal section.

increased (Fig. 4(b) and (c)). An indentation mark produced on the band by a Knoop hardness tester was completely sheared after additional rolling of about 15%, indicating highly localized shear taking place in the bands. Such shear bands were also formed in the solution treated (211)[$\bar{1}11$] specimens after about 50% rolling reduction (Fig. 5). No significant differences in the appearance of the bands were observed between the aged and solution treated specimens. In addition, the configuration of the shear bands in the present alloy is quite similar to that in Cu⁽⁶⁾ and an Al-3%Mg alloy⁽⁸⁾.

Fig. 4 Shear bands in aged (211)[$\bar{1}11$] specimens rolled at RT to (a) 50%, (b) 68% and (c) 73%, longitudinal section.

(2) Rolling at 473 K

Since the $(011)[\bar{1}00]$ specimens did not exhibit any shear bands at RT rolling, only the $(211)[\bar{1}11]$ specimens were rolled up to 80% reduction at 473 K. In this case, crystallographic slip occurred on the planes expected from Fig. 1(a), and no shear bands were formed in both as solution treated and aged specimens even at 75% rolling reduction, as seen in Fig. 6. Most of slip bands showed rather wavy appearance, suggesting high thermal activation for slip during rolling at 473 K.

2. Microstructural observation

Dislocation structure developed during rolling was examined mainly for the specimens rolled to 40–60% reductions, since the onset of shear banding was found at about 50% reduction (Figs. 4 and 5). It was revealed that no extreme differences in the dislocation structure were observed whether the specimen was as solution treated or aged. However, the differences in the initial orientation and rolling temperature yielded quite different dislocation structures, as indicated in Fig. 7 for the aged specimens. The dislocation arrangement of the $(211)[\bar{1}11]$ specimen rolled at RT was characterized by micro-bands, which consisted of layered dislocation walls nearly parallel to operative slip planes (Fig. 7(a)). However, after rolling at 473 K, the specimen with the identical orientation showed rather a homogeneous pattern of dislocation distribution (Fig. 7(b)). A similar dislocation pattern was also observed in the $(011)[\bar{1}00]$ specimen rolled at RT, though weak tendency existed toward the formation of dislocation walls parallel to active slip planes (Fig. 7(c)).

Fig. 6 Slip bands of $(211)[\bar{1}11]$ specimens rolled at 473 K, (a) solution treated specimen rolled 75% and (b) aged specimen rolled 72%, longitudinal section.

IV. Discussion

A salient feature of the present observation on the Al-2%Cu single crystals is that regardless of the heat treatments employed for obtaining either a nominal solid solution or θ precipitates, the shear bands formed only during rolling at RT in the $(211)[\bar{1}11]$ specimens in which the aligned dislocation walls developed (Figs. 4, 5 and 7). Also the shear band formation was suppressed by increasing the rolling temperature to 473 K and by applying an alternating initial orientation $(011)[\bar{1}00]$, whereby a more homogeneous pattern of dislocation

Fig. 7 Dislocation structure of rolled Al-2%Cu single crystals, (a) aged $(211)[\bar{1}11]$ specimen rolled at RT to 56%, (b) solution treated $(211)[\bar{1}11]$ specimen rolled at 473 K to 63% and (c) aged $(011)[\bar{1}00]$ specimen rolled at RT to 50%, longitudinal section.

structure was produced (Figs. 3, 6 and 7). These experimental findings are consistent with the view⁽⁹⁾ noted in the Introduction; the occurrence of shear bands appears to be closely connected with the dislocation structure built up by deformation.

Next, the feature of microstructural evolution in the rolled Al-2%Cu single crystals is considered in terms of (i) ϑ precipitates, (ii) initial orientation and (iii) rolling temperature.

Optical and electron microscopy showed that the slip band pattern and dislocation structure of the solution treated and rolled specimen were nearly similar to those in the aged and rolled specimen. Two possible reasons for such behavior may be pointed out. First, the initial microstructure of the as aged specimen consisted of coarse ϑ precipitates which were well separated from one another (mean spacing $l \approx 20 \mu\text{m}$) owing to the present aging condition. In such a circumstance, the contribution of precipitates to the slip mode of the aged alloy is estimated to be relatively low at large strains, according to a deformation model proposed by Ashby⁽¹²⁾. This is because the geometric slip distance λ^G which is dependent on the shape and spacing of precipitates is compatible with or larger than the slip distance λ^S which depends on strain, and then the statistically stored dislocation density ρ^S may dominate the total dislocation density ρ^T of the alloy. Accordingly, the slip band pattern and dislocation structure seem to be unaffected by the presence of ϑ , but specified by the matrix property such as given below. Secondly, an equilibrium concentration of the solute element in the matrix after the present aging treatment is found to be rather high: Since the α solid solution of Al can dissolve about 0.9 mass% Cu at 623 K⁽¹¹⁾, only a half of the total amount of Cu of the Al-2%Cu alloy will precipitate after aging at the temperature. The presence of Cu atoms in the matrix obviously hinders dynamic and static recovery by pinning dislocations. Hence, it is possible that the heavily over-aged alloy exhibits a similar deformation pattern to that of the nominal solid solution after large strains, apart from the deformation only in the vicinity of precipitates.

The orientation dependence of the shear band formation has commonly been observed in rolled FCC single crystals such as Cu⁽⁶⁾, Al-3%Mg⁽⁸⁾ and Cu base alloys⁽⁵⁾. As in the present case, the shear bands usually do not occur in $\{011\}\langle 100 \rangle$, but in $\{211\}\langle 111 \rangle$ crystals. It has been pointed out that such behavior is attributable to differences in dislocation structure formed during deformation, since the slip geometry which is predetermined by the initial orientations can control the development of dislocation structures⁽⁹⁾: Symmetric slip on two $\{111\}$ planes in the $(011)[\bar{1}00]$ specimen (Fig. 1(b)) may result in a rather homogeneous pattern of dislocation arrangements. In contrast, the slip geometry of the $(211)[\bar{1}11]$ specimen is asymmetric (Fig. 1(a)), so that the amount of slip on active slip planes differs from one another⁽¹³⁾. This implies that more dislocations are apt to be accumulated on the most active slip planes so as to evolve aligned dislocation walls.

The dislocation structure observed in the $(211)[\bar{1}11]$ specimen rolled at 473 K, however, does not accord with that outlined above, suggesting the importance of deformation temperature as a structure affecting factor. The 473 K rolled specimens were re-heated repeatedly for 100 s at 473 K between successive rolling passes. Thereby, three different processes, work hardening, dynamic and static recovery and aging, are considered to be taking place⁽¹⁴⁾. In this case, two remarkable effects on the development of dislocation structure can be suggested: The first effect is that the accumulation of dislocations will be retarded due to dynamic as well as static recovery, which lends itself to form a subgrain structure, such as found in pure aluminum⁽⁸⁾⁽¹⁵⁾. The second effect is due to the precipitation of a metastable ϑ' phase during the re-heating time. Plastic deformation prior to aging is believed to increase the rate of precipitation of the intermediate phase at dislocations⁽¹⁴⁾⁽¹⁶⁾. If this is true, a dislocation arrangement after rolling must be affected by nonhomogeneous slip caused by the presence of the ϑ' particles⁽¹²⁾, since the geometrically necessary dislocation density ρ^G may increase depending on the size and spacing of ϑ' . Thus, the two effects noted above are

able to contribute to the formation of a different dislocation structure from that built up according to the slip geometry of specimens, and will yield a plausible reason why the aligned dislocation walls did not form in the 473 K rolled specimen (Fig. 7(b)).

V. Summary

Different dislocation structures evolved in the rolled Al-2%Cu single crystals are able to be interpreted based on their deformation behavior by taking into account the slip geometry and precipitation- and recovery characteristics of the alloy. The occurrence of the shear bands is simply correlated with the dislocation structure, provided that a lamellar arrangement of dislocations such as aligned dislocation walls is responsible for the initiation of localized shear banding.

REFERENCES

- (1) J. W. Hutchinson: *Scripta Met.*, **18** (1984), 421.
- (2) M. Hatherly: *Proceedings of 6th Int. Conf. on Strength of Metals and Alloys*, ed. by R. C. Gifkins, Pergamon Press, Oxford, (1982), p. 1181.
- (3) D. Peirce, R. J. Asaro and A. Needlmann: *Acta Met.*, **30** (1982), 1087.
- (4) M. Hatherly and A. S. Malin: *Scripta Met.*, **18** (1984), 449.
- (5) P. T. Wakefield and M. Hatherly: *Metal Sci.*, **15** (1981), 109.
- (6) K. Morii and Y. Nakayama: *Trans. JIM*, **22** (1981), 857.
- (7) Y. Nakayama and K. Morii: *Trans. JIM*, **23** (1982), 422.
- (8) K. Morii, Y. Nakayama and H. Mecking: *Proceedings of 7th Int. Conf. on Textures of Materials*, ed. by C. M. Brakmann *et al.*, Netherlands Society for Materials Sci., (1984), p. 117.
- (9) K. Morii, H. Mecking and Y. Nakayama: *Acta Met.*, **33** (1985), 379.
- (10) K. Morii and Y. Nakayama: to be published.
- (11) E. Hornborgen: *Metall*, **43** (1967), 163.
- (12) M. F. Ashby: *Phil. Mag.*, **21** (1970), 399.
- (13) U. F. Kocks and H. Chandra: *Acta Met.*, **30** (1982), 695.
- (14) M. V. Heimendahl: *Acta Met.*, **15** (1967), 417.
- (15) K. Morii and Y. Nakayama: *Scripta Met.*, **19** (1985), 185.
- (16) A. Kelly and R. B. Nicholson: *Progress in Materials Sci.* **10** (1963), 149.