Misorientation Change of the Grain Boundary in Pure Copper Bicrystals Subjected to One-Pass Equal-Channel Angular Pressing

Yohei Wadamori¹, Kentoku Hirayama¹, Hiroshi Fujiwara, Toshiyuki Ueno and Hiroyuki Miyamoto²

Department of Mechanical Engineering, Doshisha University, Kyotanabe 610-0321, Japan

Keywords: equal-channel angular pressing (ECAP), copper, bicrystal, grain boundary, severe plastic deformation

1. Introduction

Microstructural evolution and the mechanism of grain refining into ultrafine grain (UFG) structures during severe plastic deformation (SPD) have been studied intensively.¹⁻⁴ Most of these reports seem to support, explicitly or implicitly, the grain subdivision as the grain refining mechanism, where the subboundaries emerge inside the grains and increase their misorientation with increasing strain.⁵ In the overall SPD processing, equal-channel angular pressing (ECAP) is unique in that the deformation proceeds by incremental shear restricted to the narrow zone parallel to the intersecting plane of the two channels.¹ On account of this unique deformation mode, microstructural evolution and its efficiency such as grain fragmentation per pass are influenced by deformation routes or strain path, and has been studied with a special interest from the academic and practical perspective.³⁻⁴ Interpretation of the experimental results of the effective deformation route on the UFG formation seems to be divided.⁴⁻⁶ A controversial issue seems to be the facilitating factor, the amount of plastic strain or texture, for UFG formation, especially for the formation of subboundaries, which subdivide grains and are regarded mostly as geometrically necessary boundaries (GNB). In the former case, since straight non-reversed plastic strain was proposed as the factor to facilitate grain fragmentation, redundant strain by reversed shear is a negative factor.⁴⁻⁷ Orientation of an activated slip system with respect to the shear plane was proposed as an alternative factor to initiate the grain subdivision.⁵ However, to the authors’ knowledge, little attention has been paid to the misorientation change of the grain boundaries (GBs) during SPD, which is also an important aspect of UFG formation. During SPD processing, GBs are believed to absorb lattice dislocations and change their misorientation at room temperature, which was widely demonstrated at intermediate to high temperature under moderate deformation such as tensile straining.⁸ Misorientation change of GBs accompanied by the evolution of UFG structure has been mainly evaluated using diffraction images of transmission electron microscopy (TEM) by observing an approximate tendency from the discrete set of spots toward continuously distributed rings.²⁻³ The recent advent of high resolution electron back-scattered diffraction (EBSD) enables one to obtain more quantitative information concerning distributional change of misorientation angles with an increase of strain.⁹ However, the question of how and how much a grain boundary can change its misorientation still remains to be addressed because it is very difficult to track a specified grain boundary with well-defined characteristics. Bicrystals are a very effective approach in this respect. They can be defined with a well-controlled misorientation angle and GB plane orientation, but so far, only a few studies have been reported in the field of SPD¹⁰ and other deformation such as plane strain compressing.¹¹ In this study, pure copper bicrystals have been pressed by ECAP for one pass focusing on the influence of the slip pattern, the geometry of GB plane and the shear plane of ECAP on misorientation change of the GB.

2. Experimental

The experiments were planned with a special interest in the role of slip pattern, especially, the angle between the primary slip plane and the GB plane. In ECAP, shear deformation is quite restricted inside the narrow zone along the intersecting plane of the two channels,¹ so that slip activity tends to be concentrated on one or two slip planes. In this respect, some of the present authors reported several papers concerning unique microstructure of deformed single crystals by ECAP.¹²⁻¹⁵ Copper bicrystals of commercial purity were grown by the Bridgeman method. Crystallographic orientation was controlled so that the GB plane orientation became parallel to (111) with a common (112) orientation on the GB plane. Billets of 50 mm in length with square cross section of 4 mm × 4 mm for ECAP were cut from the bicrystal by

¹Graduate Student, Doshisha University
²Corresponding author, E-mail: hmiyamot@mail.doshisha.ac.jp
a spark-erosion machine so that the GB was inclined by 45 degrees in relation to the longitudinal axis. ECAP was carried out for one billet with the GB plane parallel to the shear plane (Bicrystal A) and (112) parallel to shear direction (SD), while the other bicrystal was pressed with the GB plane perpendicular to the shear plane (Bicrystal B) and (111) parallel to SD as shown in Fig. 1. Bicrystal B is obtained by reversing around the ID axis from Bicrystal A. The primary slip plane and a degree of slip concentration among the slip systems can be predicted by the shear factor, \( \cos \phi \cos \lambda \), where \( \phi \) is the angle between the two normals of the slip and shear planes, and \( \lambda \) is the angle between a slip direction and SD. Accordingly, in Bicrystal A, the primary slip plane (PSP) and direction (PSD), the GB and the SP were all supposed to be parallel while the PSP was encountered with the GB by an angle of about 70° in Bicrystal B (Fig. 1). Note that the shear factor of the PSP are 0.92 and 0.73 followed by the secondary slip plane (SSP) of 0.34 and 0.30 in crystal 1 while 0.88 and 0.86 followed by 0.23 and 0.33 in crystal 2 in Bicrystal A. On the other hand, in Bicrystal B, 0.83 and 0.80 of PSP followed by 0.46 and 0.50 of SSP in crystal 1 while 0.81 and 0.72 of PSP followed by the 0.37 and 0.41 in crystal 2. Therefore slips are supposed to be distributed dominantly in the two slip systems in PSP. ECAP was conducted at a pressing speed of 10 mm/s at room temperature using MoS2 lubricant. After ECAP for one pass, microstructures were examined using an optical microscope, and field-emission scanning electron microscope (FE-SEM, JEOL JSM 7000F) equipped with EBSD (Oxford instruments). EBSD measurements were carried out on the central part of the billets to avoid the frictional effect between the billets and the ECAP die.

3. Results and Discussion

Figure 2 shows the macroscopic appearance and orientation image maps (OIM) around the GBs after one-pass ECAP. In Bicrystal A, the GB is observed to be inclined by about 45° (Fig. 2(a)) from ED axis. The deformation is homogeneous in crystal 1, a number of shear bands (SBs) were observed in crystal 2. Such dense SBs occur as a result of plastic instability, and are one of the unique microstructures observed in deformed single crystals by ECAP, and accompany orientation splitting.12) Shear strain tends to concentrate locally within the SBs.12) As shown in Fig. 2(b), the SBs with a different orientation can be recognized in crystal 2. In our previous studies using single
crystals, microstructure inside SBs consisted of several micro-SBs with a cluster of microbands, whereas matrix is equiaxed cellular structures. Since our primary objective is to examine the misorientation change by stable homogeneous deformation, an attention is paid to the orientation of the matrix in crystal 2. Figure 2(c) shows the orientation of the GB and its vicinity of crystal 1 and matrix in crystal 2. It is evident that GB remains the single boundary carrying most of the misorientation between two grains.

In Bicrystal B, the GB was rotated from the initial inclination toward the ED (Fig. 2(d)) with an inclination of 20°. The result can be explained again by the flow rule as shown in Fig. 3(c).10) Deformation of both the grains seems to be homogeneous, and SBs were not observed in Bicrystal B. Macroscopically straight and single GB can be recognized in OIM (Fig. 2(e)). Even at higher magnification (Fig. 2(f)), the GB seems to be a single boundary, and it carries most of the misorientation between the two crystals. If one observes the morphology of GBs in Figs. 2(c) and 2(f), both the GBs are microscopically winding or wavy. This wavy morphology suggests that it is in nonequilibrium state and has absorbed an appreciable amount of lattice dislocations. In Fig. 4, {100} pole-figures indicate the orientations of crystal 1 and 2 in the immediate vicinity of grain boundaries before and after ECAP. The solid line and the dotted line indicate the orientation of crystal 1 and 2, respectively.

Fig. 2 Macroscopic appearances observed by optical microscopy and OIM in the vicinity of grain boundaries in Bicrystal A (a)-(c) and Bicrystal B (d)-(f) after one-pass ECAP. “M” and “S” denote matrix and shear band respectively. The square mark in (a), (b), (d) and (e) correspond to the areas of (b), (c), (e) and (f), respectively.

Fig. 3 Materials flow rule, showing that (a) the points A and B flows to A’ and B’, via M and N, respectively, following the rule AM + MA’ = BN + NB’. Inclination change of the grain boundaries in (b) Bicrystal A, and (c) Bicrystal B, predicted by the rule.

Fig. 4 {100} pole-figures obtained by EBSD indicating orientations of crystal 1 and 2 in the immediate vicinity of grain boundaries before and after ECAP. The solid line and the dotted line indicate the orientation of crystal 1 and 2, respectively.
change in the immediate vicinity of the GBs was not recognized in OIM. Figure 5 shows the misorientation at several points along the GBs. The measured points both in crystal 1 and 2 are within one micron from the GB. It is evident that the change in misorientation before and after deformation is negligibly small in Bicrystal A, whereas the misorientation in Bicrystal B decreased by 15° after ECAP.

GBs are the source and sink for lattice dislocations at moderate to high temperature. Lattice dislocations encountered and trapped in the boundary become extrinsic GB dislocations (GBD). Such dislocations are considered to increase the GB energy, and dissociate into several dislocations with smaller burgers vectors defined by a discrete site complete (DSC) lattice and cause misorientation change. These trapped lattice dislocation, and subsequent dissociations were observed by transmission electron microscopy. In the SPD process, a considerable amount of lattice dislocations can be trapped in GBs at room temperature, leaving GB in the nonequilibrium state. According to the Frank-Bilby relation, the misorientation of a GB in an equilibrium state is proportional to both the GBD density and the directional component of their burgers vector perpendicular to both the GB plane and the misorientation axis. Based on the relation, Pantleon proposed a statistical model of dislocation boundaries with low-energy dislocation configuration, and predicted the evolution of average misorientation angles of deformation structures with increasing strain. In the present study, GBs are supposed to be in the nonequilibrium state with a large amount of trapped lattice dislocations (TLDs) as suggested by the very winding GB, but one can predict, in a first rough approximation, the misorientation change in terms of the relation between the primary slip system and GB plane.

In Bicrystal A, where the misorientation change across the GB was negligibly small, the primary slip plane that is supposed to generate most of the shear strain, is almost parallel to the GB. Therefore, lattice dislocations do not encounter the GB, resulting in little misorientation change. It should be pointed out that shear strain and lattice dislocations in crystal 2 were supplied by SBs, which were nearly parallel to the GB. On the other hand, in Bicrystal B, lattice dislocations of primary slip system have a burgers vector with a larger angle with the GB plane. Thus, these lattice dislocations trapped in the GBs should cause the misorientation change.

It should be emphasized that misorientation change is influenced appreciably by the initial orientation in ECAP where the unique slip pattern of the predominant slip by the primary slip system occurs. In other deformation modes, such as rolling and torsion, several slip systems may be activated in most orientations, so that effect of the primary slip system is not as obvious as ECAP. Misorientation change of as much as 10° in bicrystals deformed by ECAP was reported by Han, but it was not analyzed in detail. There is a discrepancy with our results about the influence of the relation between SP and GB plane on a degree of misorientation changes. This discrepancy may derive from different crystal orientations of component crystals and the GB plane orientation.

4. Conclusions

The dependence of the misorientation change of the grain boundaries in pure copper bicrystals on the initial crystallographic orientation was examined in terms of slip geometry, grain boundary plane orientation and the shear plane of ECAP. Direct evidence of the misorientation change of the grain boundary during severe plastic deformation was obtained. Marked dependence of misorientation change on the initial orientation was revealed, and can be associated with the unique slip patterns of ECAP, where the shear deformation is restricted to the narrow zone parallel to the intersecting plane of the two channels. When the primary slip system is closely parallel to the shear plane and grain boundary plane, the misorientation change is negligibly small. On the other hand, when the primary slip plane is perpendicular to the grain boundary plane, the misorientation change is rather large. The degree of change can be related to slip geometry in terms of grain boundary plane orientation.

REFERENCES


