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Formation of Σ3{110} incoherent twin boundaries through geometrically necessary boundaries in an Al-8Zn alloy subjected to one pass of equal channel angular pressing

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ABSTRACT

For coarse grained (CG) alloys with high stacking fault energies (SFEs), like aluminum, deformation twins can rarely form. Here, we report that Σ3{110} incoherent twin boundaries (ITBs) could be generated in a CG Al-8Zn alloy by one pass of ECAP. A systematic investigation shows that the Σ3{110} ITBs are formed by gradual evolution from geometrically necessary boundaries (GNBs) delineating deformation bands (DBs) by lattice rotation via <111>-twist CSL boundaries. This is a new deformation mechanism in Al alloys, which has never been reported.

1. Introduction

When a metal is plastically deformed, two types of dislocation structures have been found, i.e., geometrically necessary boundaries (GNBs) and incidental dislocation boundaries (IDBs) [1]. In one grain, GNBs separate regions deformed by different sets of slip systems or by the same sets of slip systems but with different shear strain amplitudes or different strains, forming band structures [2–4]. IDBs are formed by the statistical trapping of glide dislocations and align in a random pattern. In addition to dislocation slip, deformation twinning is also an important deformation mode [5], which has been reviewed by Meyers et al. [6] and by Christian et al. [7]. Compared to nano-crystalline and fine grained materials, nano-twinned metals exhibit extraordinary properties [8], including high yield strength [9–13], enhanced ductility [10,13], high electrical conductivity [9,12] and high strain rate sensitivity [10,11,14].

The twinning tendency of a face centered cubic (FCC) metal is largely determined by its stacking fault energy (SFE). For example, with high SFE, coarse grained FCC metals such as Al and Ni are normally deformed by dislocation slip, while FCC metals with low SFE such as Ag are primarily deformed by twinning. However, for coarse grained FCC metals, deformation twinning has been observed under some extreme deformation conditions, such as at crack tips [15,16] and during high strain rate deformation at cryogenic temperatures [12,17]. Meanwhile, deformation twinning has been shown to be possible in nano-crystalline materials [18–22], which is due to partial dislocation emissions from grain boundaries (GBs) [21,22].

It has been proposed that the SFE of Al-Mg alloys decreases with increasing Mg contents [23]. However, a recent first principles calculation based on density function theory (DFT) shows that an addition of Mg in Al can only slightly decrease the SFE and thus the twinning ability is slightly improved [24]. This has been confirmed by an experimental study on an Al-7Mg alloy subjected to dynamic plastic deformation (DPD) under a strain rate of ~10^2 s^{-1} [25], in which no deformation twins could be found. Instead, a significant fraction of Σ3 incoherent twin boundaries (ITBs) have formed, which was proposed to be gradually evolved from low angle deformation bands (DBs) through <111>-twist CSL boundaries [25]. This has been attributed to the special planar glide deformation mode of the Al-7Mg alloy and the high strain rate introduced by DPD.

Equal channel angular pressing (ECAP) has been one of the most

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important severe plastic deformation (SPD) methods to produce ultrafine grained (UFG) Al and Al alloys. A main focus has been put on the grain refinement mechanism and mechanical properties of the UFG materials [26–29], while the fundamental study on the deformation behavior has been mostly based on single crystals of low alloyed Al alloys [30,31]. In this paper, we report that a significant fraction of Σ3{110} ITBs could form by a special deformation mechanism in an Al-8Zn alloy deformed by conventional ECAP (strain rate, ~2×10^{-2} s^{-1}), which has never been reported in ECAP processed Al alloys.

2. Experimental

The material used in the present work was an Al-8 wt.% Zn alloy produced by melting commercial purity Al and Zn. Note that 0.5 wt.% Al-5Ti-1B grain refiner was added into the melt in order to get fine grains. As illustrated in Fig. 1(a), the as-cast Al-8Zn alloy comprises a granular α-Al grain structure. Fig. 1(b) shows the grain size distribution chart and the average grain size is measured to be ~50 μm.

The as-cast ingots were machined into bars with dimensions of 100 mm × 19.5 mm × 19.5 mm. Before ECAP, the bars were coated with a thin layer of a graphite lubricant to lower the friction during ECAP. Then, these bars were processed by ECAP though a 90° die (Fig. 2(a)) at room temperature (RT), which leads to an imposed equivalent strain of ~1.0 per pass [27,32].

As shown in Fig. 2(b), samples for microstructure observation were cut from the uniformly deformed region of the ECAP processed bars in the longitudinal section. The as-deformed structure was characterized by electron backscatter diffraction (EBSD). Samples for EBSD were electro-polished using a solution of 80% C_2H_5OH + 20% $HClO_4$ at 20 V for 15–17 s at −30 °C. EBSD was performed by using a Hitachi SU-6600 field emission gun SEM (FEG-SEM) equipped with a Nordif EBSD detector and TSL OIM software.

3. Results and discussion

After one pass of ECAP (1P), typical microstructures in the longitudinal section are shown in Fig. 3(a). Most of the equiaxed grains are deformed into an elongated shape with an inclination angle of ~30° to the extrusion direction (ED). Interestingly, within some elongated coarse grains, twin-like lamellar grains delimited by high angle boundaries (HABs) can be observed and herein, we name them as deformation bands (DBs). As an example, the grain G1 highlighted by the white oval in Fig. 3(a) was magnified and shown in Fig. 3(b). Surprisingly, there are some Σ3 boundary segments (identified as 60°<111> by the OIM software) coexisting with non-Σ3 general DB boundaries.

Fig. 3(c) shows the {111} and {110} poles of the DB and the matrix in Fig. 3(b). As can be identified, the DB and the matrix share one {111} pole and three sets of {110} poles, confirming that they have a Σ3 twin orientation relationship. However, as indicated in Fig. 3(b) and (c), the traces of DB boundaries are far way from {111} plane traces of both the DB and the matrix. Instead, they are closely parallel to one of the {110} plane traces. It indicates that these Σ3 DB boundary segments are not coherent deformation twin boundaries. The structure of the Σ3{110} ITB has been investigated in Refs. [33,34], where the atomic structure of the two crystals separated by the Σ3{110} boundary is symmetric. By using the single-section trace method [35], fifteen Σ3 boundary segments (in different grains) were investigated and their poles were drawn in the (001) stereographic projection (Fig. 3(d)). Since the trace of each Σ3 boundary segment has two different indices belonging to the DB and the matrix, respectively, thus, 30 pole pionts were drawn. It can be seen that only 2 out of 30 points are positioned far away from the {110} traces, further revealing that most Σ3 boundaries are Σ3{110} ITBs. Furthermore, a distinct character of the DB boundaries is that their misorientation angles vary along their lengths. Based on statistical data, the misorientations of non-Σ3 DB boundaries and Σ3 boundaries concentrate in the range of 50.0–60.3° and 54.1–60.4°, respectively. Although some non-Σ3 DB boundaries are with misorientation angles matching the angle range of Σ3 boundaries (60.0 ± 8.7°) [36], they cannot be identified as Σ3 boundaries because their misorientation axes are not <111> axes. This can be revealed by the inverse pole figure in Fig. 3(e), where the misorientation axes of the DB boundaries within G1 were collected. As can be seen, the misorientation axes of the non-Σ3 DB boundaries are far away from <111>.

Fractions of different CSL boundaries (based on three EBSD maps with the same magnification as Fig. 3(a)) have been calculated and shown in Fig. 3(f). As can be seen, some CSL boundaries are dominating, which have a common feature: <111> rotation axes. These boundaries include Σ3 60°<111>, Σ7 38.2°<111>, Σ13b 27.8°<111>, Σ21a 21.8°<111>, Σ31a 17.9°<111> and Σ43a 15.2°<111>. It implies that during deformation, CSL boundaries

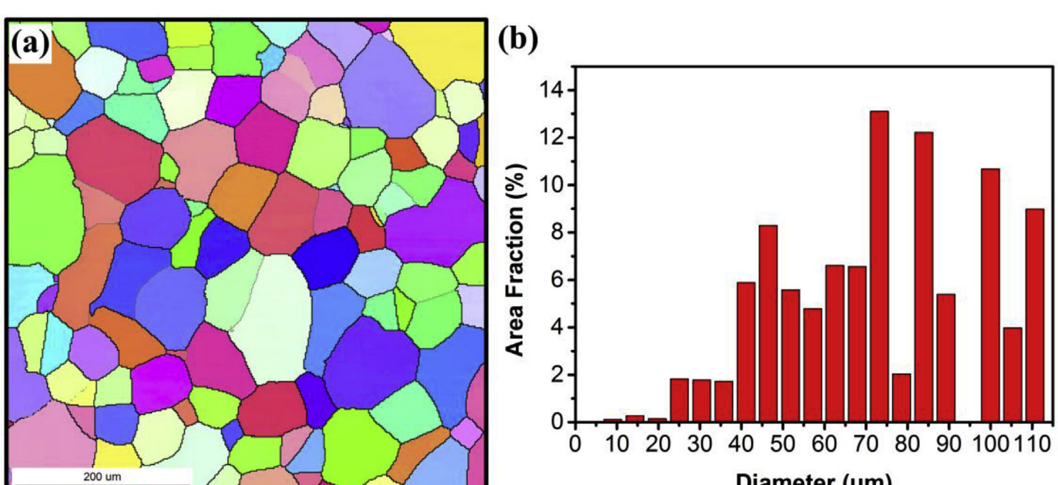

Fig. 1. Microstructures of the as-cast Al-8Zn alloy. (a) EBSD map and (b) grain size distribution chart.