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Formation of large scaled zero-strain deformation twins in coarse-grained copper

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ABSTRACT

Most deformation twins in nanocrystalline metals were reported to produce zero macroscopic strain via either random activation of partials or cooperative slip of three partials. In their coarse-grained counterparts, deformation twins produce a net macroscopic strain and change the grain shape since a twin forms via the glide of a single twinning partial dislocation on successive {111} planes driven by Peach-Koehler force. Here, we report the formation of large scaled zero-macrostrain deformation twins in coarse-grained copper with medium stacking fault energy. The responsible formation mechanism is proposed and its effect on twinning or detwinning processes is discussed.

Twinning is a typical plastic deformation mode, accomplished by the glide of twinning dislocations (TDs) to generate macroscopic strain [1, 2]. Deformation twinning has been widely observed in face-centered-cubic (fcc) [3–5] materials and its tendency is mainly determined by stacking fault energy (SFE) [6]. For example, dislocation slip is the preferred plastic deformation mechanism for coarse-grained Cu and Al with medium-to-high SFEs, while plastic deformation is accomplished by a combination of dislocation slip and deformation twinning for Ag with low SFE [7].

Normally deformation twinning in coarse-grained materials produces a net macroscopic strain and changes the grain shape since a twin forms via the glide of a single twinning partial dislocation on successive {111} planes driven by Peach-Koehler force [1]. On the other hand, deformation twinning in nanocrystalline materials can produce much less and often zero net macroscopic strain [8,9], attributed to either random activation of partials (RAP) or cooperative slip of three partials (CSTP). Wu et al. proposed the RAP twinning mechanism in which the participating Burgers vectors of the partials sum to zero [9]. This mechanism leads to zero net macrostrain during a twinning process in nanocrystalline metals. Wang et al. [10–13] proposed the CSTP twinning mechanism, which is accomplished via the collective glide of Σ3{112} incoherent twin boundaries (ITBs). ITBs can be presented as a set of Shockley partial dislocations on successive {111} planes with a repeatable sequence of b_2:b_1:b_3. The Burgers vector b_1 represents a pure edge partial ($b_1 = 1/6[11\bar{2}]$), and the other two are mixed partials with opposite screw components ($b_2 = 1/6[\bar{2}11]$ and $b_3 = 1/6[1\bar{2}1]$) [10–12]. Therefore, the net Burgers vector of the three partials in one unit equals zero. The ITBs can readily move, leading to, depending on the moving direction, twinning or detwinning under a small net Peach-Koehler force via the stop-start and move-drag partial dislocation mechanism [11].

Most ITBs were observed at the end of a grown twin or as steps at coherent twin boundaries (CTBs) in nanocrystalline materials [11,12,14–17], nanolamellar materials [18] and columnar-grained nanotwinned materials [10,19,20]. For example, deformation induced ITBs were observed in nanocrystalline Cu—Ni [16], Cu—Al alloys [17] and columnar-grained nanotwinned Cu [20] recently, and the moving of them led to twinning or detwinning of the nanotwins depending on the moving direction. Such deformation induced zero-strain twins were seldom observed in coarse-grained materials. The only examples were in Ni-Co-based superalloy and Ag with low SFEs (23.4 ± 3.1 mJ·m^{-2} and 19 ± 3 mJ·m^{-2}, respectively) [7,13,21]. The zero-strain twins in the coarse-grained Ni-Co-based superalloy result from the cooperative dissociation of three full dislocations [21], and the ITB nucleation in the coarse-grained Ag occurs via the collective glide of Σ3{112} ITBs [13]. In this paper, we report the formation of large scaled zero-strain deformation twins in coarse-grained Cu with a medium SFE (45 mJ·m^{-2}) [2]. The responsible twinning/detwinning mechanism is proposed, and the effect on twinning or detwinning processes is discussed.

A dog-bone-shaped commercial pure Cu (99.5%) specimen with a gauge size of Φ6 mm × 20 mm was prepared and treated by free-end torsional deformation. The torsional treatment was carried out at a

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constant rate of 1 rpm for 1 revolution (MTS 858 Mini Bionix II). After torsional treatment, the specimen was cut using a low speed diamond saw for microstructural characterization. The microstructures of the specimen were examined from the diametrical cross-section by electron backscattering diffraction (EBSD, EDAX TSL) in a FEI Quanta FEG 250 field emission gun scanning electron microscope (FEG-SEM), transmission electron microscopy (TEM) on a JEOL 2100F transmission electron microscope with an accelerating voltage of 200 kV and high resolution TEM (HRTEM) on an FEI Titan G2 60-300 TEM with a spherical aberration corrector under an accelerating voltage of 300 kV. Both EBSD and TEM investigations were taken at regions of ~1.5 mm from the center. The TEM samples were electropolished with an electrolyte of nitric acid (33%) and methyl alcohol (67%) at a temperature of 243 K.

Fig. 1 shows the EBSD images of the specimens before and after torsional treatment. It can be seen from Fig. 1a that the average grain size of the as-received specimen is ~150 μm. From Fig. 1b, obvious orientation fluctuation can be observed, which is originated from the profuse dislocation activities during torsional treatment [22]. Besides, many deformation twins can be observed in Fig. 1b (arrowed).

Fig. 2 shows the bright field TEM images and the corresponding selected area electron diffraction (SAED) patterns of the sample after torsional treatment. The torsional processing leads to the formation of abundant dislocations and dislocation cells, as shown in Fig. 2a. In addition, deformation induced twins can also be observed, as shown in Fig. 2b. Fig. 2c shows the step-like twin boundary, and the SAED patterns of the circled areas from two directions of the steps (indicated by red arrows) were shown in Fig. 2d and e. Fig. 2d is the SAED pattern from direction 1, a typical CTB pattern of fcc metals. Fig. 2e shows a typical SAED pattern from direction 2, and extra diffraction spots are observed, corresponding to a periodicity of three times the interplanar spacing of {111}. Both experimental observations and molecular dynamic simulations [10,12,13] revealed that the ITBs were formed by the cooperative slip of three different partials b_1, b_2 and b_3 on three adjacent {111} planes and the Σ3{112} ITBs can dissociate into two tilt walls bounding a 9R phase [23,24]. The SAED pattern in Fig. 2e agrees well with the fast Fourier transform (FFT) of the 9R phase in previous investigations [16, 24].

Fig. 3a and b show the magnified images of a CTB and a typical equilibrium state of an ITB, respectively. It is well known that in the absence of stress the initial compact Σ3{112} ITBs can dissociate spontaneously into two tilt boundaries, bonding a 9R phase, to decrease the elastic energy by emitting Shockley partials on every three {111} planes that results in a repeated pattern with a period of 3 times the interplanar spacing of {111} planes [23,24], as revealed in Figs. 2e and 3c. The equilibrium separation width between two boundaries is of the order of ~1 nm (one to a few dislocation core sizes) and the width is inversely proportional to the SFE [10–12]. Fig. 3c is a magnified image of the 9R phase, showing a repeated pattern with a period of 3 times the interplanar spacing of {111} planes. Fig. 3d shows the juncture area of the CTB and ITB, with the phase boundaries (PB1 and PB2) being marked by asterisks. The width of the 9R phase is from several nanometers to ~50 nm, and the maximum width is larger than 50 nm. Previous investigations [10,12,16] showed that the width of the observed 9R phase is ~5 nm in nanocrystalline Cu. Recently, Ma et al. [25] observed a much longer 9R phase (18.3 nm) in nanocrystalline Cu—Zn alloy, which has a much lower SFE than pure copper. Liu et al. [13] also observed a 9R phase (~12.5 nm in width) in coarse-grained (~1 μm) Ag with a much lower SFE. In present study, the width of the 9R phase reaches 50 nm or even more, far beyond the width in previous observations [10,12,13,16,25]. It is worth mentioning that the length of the 9R phase in Fig. 3d increases with decreasing the distance to the CTB. Fig. 3e shows HRTEM images of another 9R phase, indicating the formation of large scaled 9R phase. The magnified image of the rectangle area inserted in Fig. 3e shows that PB2 was pinned by dislocations.

Unlike the growth ITBs formed during pre-deposition, the Σ3{112} ITBs exhibited in Fig. 2c were formed during torsional deformation. Since the ITBs are originated from the grain boundaries (GBs) and terminated inside the grains, their nucleation is closely related to the GBs. Previous studies [16,26] proposed that extensively migrating GBs can emit Shockley partials, while non-equilibrium GBs facilitate the nucleation of ITBs. An et al. [16] indicated that under a high stress and a proper stress orientation a GB can emit synchronously three partials (b_1:b_2:b_3) as "zonal" twinning dislocations, leading to the nucleation of ITBs, and this is more likely to happen in non-equilibrium GBs to release the high local stress concentration [26–28]. The required high stress and a proper stress orientation can be readily provided by torsional processing [29,30].

Unlike nanocrystalline metals, in which most deformation twins were reported to produce zero macroscopic strain via either the RAP mechanism or the CSTP mechanism [31], twins in coarse-grained counterparts produce a net macroscopic strain and change the grain shape since a twin forms via the glide of a single twinning partial dislocation on successive {111} plane driven by Peach-Koehler force [1]. Recently, Zhang et al. [14] investigated the grain-size effect on the zero

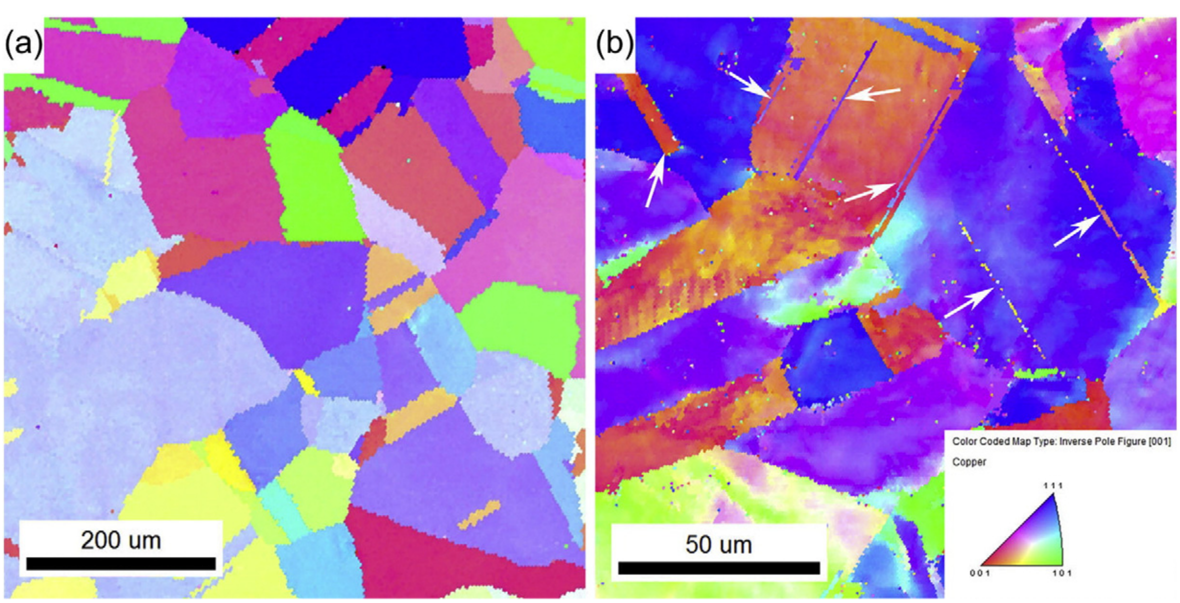

Fig. 1. The EBSD images of (a) the as-received sample and (b) the sample after torsional treatment.