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Terrace-like morphology of the boundary created through basal-prismatic transformation in magnesium

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Boundaries created through basal-prismatic transformation in submicron-sized single crystal magnesium have been investigated systematically using in situ transmission electron microscopy. We found that these boundaries not only deviated significantly from the twin plane associated with $\{10\bar{1}2\}$ twin, but also possessed a non-planar morphology. After the sample was thinned to be less than 90 nm, aberration-corrected scanning transmission electron microscopy observation found that the basic components of these boundaries are actually terrace-like basal-prismatic interfaces.

Keywords: Basal-prismatic interface; Grain boundaries; Magnesium; TEM

At room temperature, dislocation slip [1] and deformation twinning [2,3] are known to be the main mechanisms of plastic deformation for crystalline metals. In these displacive processes, once lattice gliding dislocations (or twinning dislocations) run across the material, steps will be generated on the free surfaces with the magnitude being multiples of Burgers vector of the dislocation [4,5]. For magnesium having the hexagonal close packed (hcp) structure, it has been well accepted that $\{10\bar{1}2\}$ deformation twinning will be a major contributor to the plasticity [6,7], when the crystal is subjected to an effective tensile stress along the $[0001]$ direction. Recently, we demonstrated that when submicron sized single crystal magnesium is deformed under compression along the $[1\bar{1}00]$ orientation or tension along the $[0001]$ orientation, plastic strains can be accommodated by basal-prismatic (BP) transformation [8] which results in the effective reorientation of the hcp unit cell by about 90° around the $[11\bar{2}0]$ direction (*a*-axis). In this paper, for the sake of simplicity, we term this novel deformation mechanism as unit cell reconstruction (UCR).

Compared with deformation twinning, UCR in magnesium has several distinct features. First of all, it is the BP transformation at the BP interface that dominates the boundary migration instead of the gliding of twinning dislocations on twinning planes. Second, the boundary that separates the reoriented crystal from its parent crystal is not a crystallographic mirror plane. Third, UCR produces tetragonal deformation instead of simple shear [9,10]. These new features suggest that the boundary generated through UCR must be different from the boundaries produced by deformation twinning (twin boundaries). However, information regarding the morphology of such boundaries remains very limited [11–13]. In this work, we focus on the atomic-scale configurations of the boundaries generated through UCR, while their effects on the mechanical response will be discussed in future.

Submicron-sized single-crystal magnesium samples used in this work were fabricated with exactly the same method reported in our previous work, employing focused ion beam (FIB, Helios 600, FEI) micromachining [8]. Figure 1 shows the typical SEM images of a pillar sample (Fig. 1a) and a dog-bone shaped sample (Fig. 1b). Compression tests were conducted both on pillar samples and dog-bone samples. Tension tests were applied on dog-bone samples only. The axial direction of the compressive samples was designed to be $[1\bar{1}00]$ and that for tensile samples was $[0001]$. The viewing direction for pillar samples was designed to be $[11\bar{2}0]$ or $[0001]$, and for dog-bone samples it was $[11\bar{2}0]$. The cross-section of these samples was close to square shape. The nominal sample size, defined as the square root of the cross-sectional area of the samples, ranged from 100 nm to 500 nm. The aspect ratio of the pillar samples ranged from 3:1 to 4:1, and that of dog-bone samples ranged from 4:1 to 10:1. In tension tests, the entire gauge volume tends to be converted into the new grain

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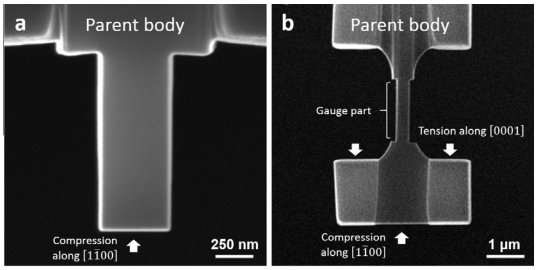

Figure 1. SEM images showing the typical sample geometry and loading conditions. (a) A pillar sample. (b) A dog-bone sample. In compression tests the external load was applied to the bottom surface via a diamond probe. In tension tests the pulling force was applied to the two shoulders (indicated by the two white arrows) by a specially designed diamond gripper.

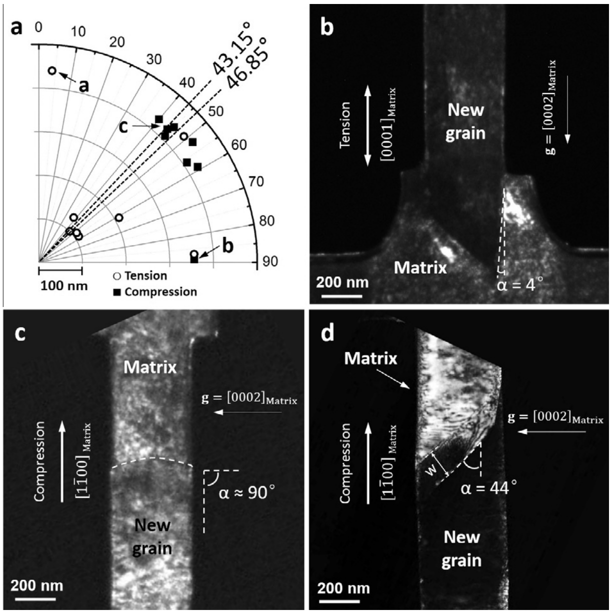

Figure 2. Statistic measurement of the angle α between the trace line of the boundary and the loading direction. (a) A protractor diagram shows the distribution of the angle α measured from 16 boundaries. The radial axis represents the diameter of samples. The theoretical angle α for $\{10\bar{1}2\}$ twinning are marked using dashed lines. (b), (c) and (d) are dark field TEM images that correspond to the angle α marked in (a) respectively.

within a single strain burst when the aspect ratio is small [8]. Therefore, longer dog-bone samples allowed us to monitor the boundary migration process within the gauge part during tensile loading. All the mechanical tests reported in this paper were carried out using Hysitron PI95 Picoindenter under displacement control mode inside a transmission electron microscope (TEM, JEOL 2100F, 200 keV) [14]. The probe velocity was set to be 5 nm/s which corresponded to a strain rate in the order of 10^{-3}/s.

In order to reveal the 3D morphology of the boundaries created by UCR, we not only examined the sample from two orientations, but also purposely thinned a typical sample using FIB from 500 nm down to 90 nm. This sectioning allowed us to investigate the change of the projected boundary profile. However, FIB machining will introduce surface damage layer with a thickness of tens of nanometers, which makes it difficult to thin a 90 nm sample. We therefore employed a Nano Mill (M1040, Fischione) for the final thinning process, where the damaged layer can be as thin as 2 nm. After that, an ARM200F spherical aberration-corrected TEM (200 keV) was used under the scanning TEM (STEM) mode, which avoids the delocalization effects in imaging.

Regardless of compression or tension, a new grain always forms in our single crystal magnesium samples, creating a boundary between the new grain and the matrix. Selected area diffraction analysis found that for all the samples reported in this work, the basal plane of the new grain and that of the matrix is always perpendicular to each other. Given the observation direction along $[11\bar{2}0]$, if we define the angle between the trace line of the boundary and the loading direction as α, then the α expected for $\{10\bar{1}2\}$ deformation twinning will be 43.15° for compression and 46.85° for tension. However, measurements for 16 boundaries indicate that the α is statistically inconsistent with that expected from $\{10\bar{1}2\}$ deformation twinning (Fig. 2a). This conclusion is supported by the morphology of the newly formed boundaries: some of them are even parallel with or perpendicular to the loading direction, and some of them exhibit considerable width. Typical examples corresponding to these three scenarios are shown in Figure 2b–d.

Figure 2b is a typical example showing a newly formed boundary parallel to the loading axis. The postmortem TEM dark field image was taken at the root part of this sample. The new grain was formed during a strain burst. As shown in Figure 2b, one of its ends penetrated into the matrix like a dagger, instead of stopping at the end of the gauge part, with a boundary almost aligned with the extrapolation of the right surface of the gauge part, as delineated by the dashed white lines. Another extreme case is shown in Figure 2c. This image was a snapshot extracted from a movie recorded during the compression test of a pillar sample. The migrating boundary appeared to be in an arched shape and almost perpendicular to the loading direction, as outlined by the white dashed line in Figure 2c. In addition, for the given viewing direction, the sample was observed to widen toward both sides symmetrically along with the propagation of the boundary and no simple shear was observed. Conventional plastic deformation mechanisms, either an ordinary dislocation slip or deformation twining, appear to be inadequate to explain these phenomena [8].

Besides the two extreme cases described above, another very interesting feature of the boundaries created by UCR is that they may exhibit considerable width when viewed along $[11\bar{2}0]$. This is unexpected because both the twin plane and the BP interface should appear edge on for the given observation direction. One typical example is shown in Figure 2d. This pillar sample has a nominal size of 424 nm. Even though the $\alpha = 44°$ is close to the theoretical value for $\{10\bar{1}2\}$ deformation twinning, two nearly parallel trace lines instead of one can be seen clearly in between the new grain and its matrix, with the width (w) of about 120 nm. In addition, the contrast in the area framed by these two trace lines is not uniform, with its left lower part close to that of the new grain (dark) and its upper right part close to that of the matrix (bright). On the one hand, it indicates that the volume fraction of the new grain along the thickness direction is not uniform inside the band; on the other hand, it simply rules out the possibility that the band