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Formation Process of 8º [001] Symmetric Tilt and 65.5º [–110] Symmetric Tilt Grain Boundaries During Annealing of a Cross Rolled Aluminium Sample

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From the viewpoint of grain boundary structure a small-angle symmetric tilt boundary can be considered a boundary that is symmetrically mis-oriented by a small angle from a perfect Σ=1 boundary. The mis-oriention in these boundaries occurs during the re-crystallization period when annealing of the deformed material helps to develop the boundary. The development of the boundary occurs mainly by two methods. One method of development, which is pertinent to the conventional dislocation theory, occurs when the two participating crystals join along a common plane surface. During such unification the participating crystals assume certain arrangements of their atoms in the vicinity of the boundary to match the normal crystal lattice as closely as possible. Dislocations are formed [1] at the positions of poorest fit between the two crystal lattices at the interface. Elastic continuum of the participating crystals allows these dislocations to be separated equally or almost equally from each other along the entire boundary. In the other process, initiation of a boundary occurs when a few closely neighboring mobile edge dislocations in a region of more recrystallized single crystal during their glide on their respective glide plane become immobile after encountering some obstacles [2]. The stained field associated with such entrapped dislocations in conjunction with the elastic continuum of the crystal allows the other mobile edge dislocations of the tilt plane approaching to this region of the single crystal to seize in the place of lowest energy, which lies along a common plane surface. Total effect of such dislocation interactions is the formation of a regular array of dislocations separated by a nearly perfect lattice. Such an arrangement of dislocations has shown to produce a mixed character small angle boundary [2]. The process can only produce a symmetric tilt boundary when sufficient energy, such as that present during re-crystallization, is available in transforming participating non edge dislocations of the tilt plane to edge dislocations by lattice alignment. The reason for this is that a typical single crystal whether it is deformed or annealed always contains neighboring edge, partial, screw and mixed character dislocations. Figure 1 reveals a typical 8º[001] small angle symmetrical tilt boundary in a sample of Al that was prepared by cross rolling and annealing. It possesses a regular array of edge dislocations that run along the common (110) plane. The boundary is symmetrically mis-oriented by 8º from its ideal Σ=1 boundary orientation. In accordance with the process of formations of the boundary discussed above the edge dislocations as marked in the HREM image of Figure 1 are formed at the position of maximum mismatching of cross boundary lattices and assume closely equal spacing.

The processes of formation of boundary described so far are equally applicable to the formation of higher angle tilt boundaries. Figure 2 shows an HREM image of a 65.5º [-110] symmetric tilt grain boundary found in a sample of Al that was prepared by cross rolling and annealing. The boundary can also be considered as a small angle (5º) misoriented [-110]/(-1-11) Σ=3 boundary. The boundary misorientation is suggested to occur by either of two methods of boundary formation discussed so far. According to the first method, the unifications of two crystals occurs across a common plane surface (-1-11) of the participating crystals at a mis-oriention of 5º from an ideal [-110]/(-1-11) Σ=3 boundary. The unification of participating crystals has given an arrangement of atoms in the vicinity of the boundary to match the normal [-110]/(-1-11) Σ=3 lattice arrangement as closely as possible. Such an atomic arrangement along with the elastic continuum of the participating crystals results in a regular array of dislocations at the positions where the maximum mismatching of the normal <110>/(111) Σ=3 lattice arrangement occurs at the interface. Such dislocations are expected to be separated by a nearly perfect Σ=3 lattice. Inspection of the HREM image of Figure 2 reveals that the dislocations as indicated by arrows are present at close to equal spacing on the common (-1-11) twin plane and at the position of maximum lattice mismatching for the normal [-110]/(-1-11) Σ=3 lattice arrangement. The dislocations present at the boundary have not been characterized yet. According to the second method mobile edge dislocations of the participating grains upon glide on their respective glide plane can move as is found recently [3] for a [-110]/(-1-11) Σ=3 twin boundary to form Σ=3 grain boundary dislocations. Such capturing of lattice dislocations along with elastic continuum of the participating crystals can give the present boundary mis-oriention with dislocations to assume equal spacing.

References
[1] M. Bollmann: Crystal Defects and Crystalline Interfaces, (New York, NY., Springer, 1970).
[2] M. Shamsuzzoha, *Microscopy and Microanalysis,* 17 (2011), 1342.
[3] M. Shamsuzzoha, Accepted for publication & in press for the 2014 TMS proceedings.

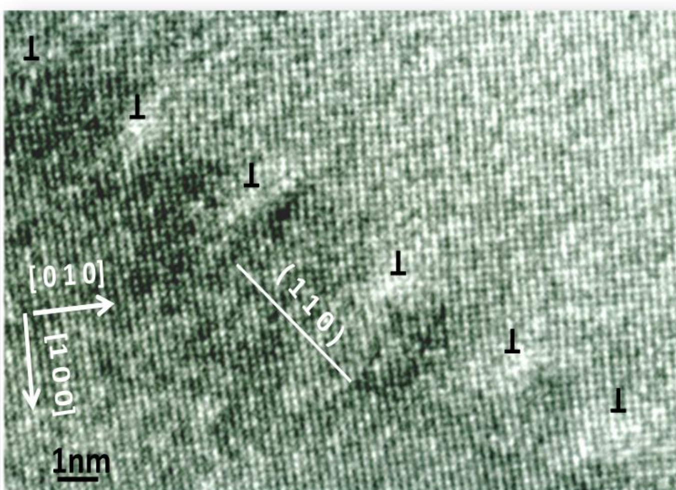

Figure1. HREM micrograph showing an 8º[001] symmetric tilt grain boundary in Al.

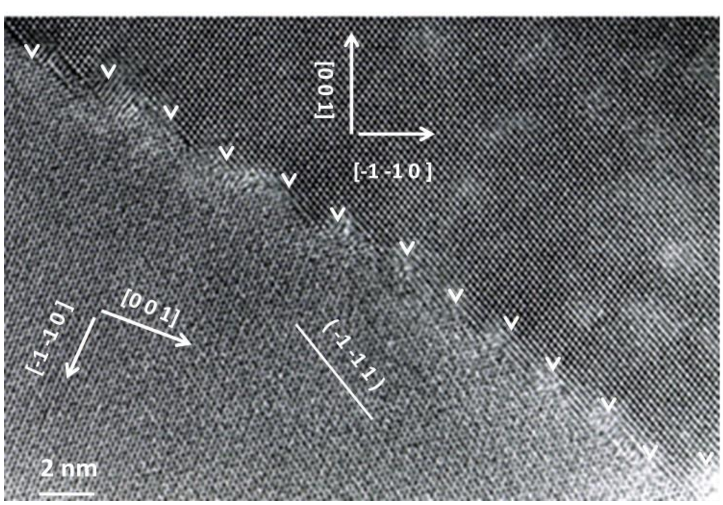

Figure 2. HREM micrograph showing a 65.5º [-110] closely symmetric tilt grain boundary in Al

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