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Dislocations, boundaries and slip systems in cube grains of rolled aluminium

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The relationship between the dislocation content of boundaries and the active slip systems is explored by characterisation of Burgers vectors, dislocation lines and relative densities in 11 boundaries in near-cube grains in 10% rolled aluminium. To provide a good basis for comparison, all the boundaries investigated lie in the longitudinal plane. Practically all dislocations have screw character, with Burgers vectors corresponding to the slip systems predicted active. The dislocations arrange in a regular grid and the boundaries are most likely low-energy dislocation structures.

Keywords: Transmission electron microscopy (TEM); Dislocation boundaries; Slip; Burgers vectors; Aluminium

The dislocations responsible for plastic deformation of metals interact, resulting in work hardening and the development of a deformation microstructure as the dislocations assemble in boundaries. The morphology and the crystallographic alignment of these dislocation boundaries depend strongly on the crystallographic orientation of the deforming grain [1,2]. Recent research has shown that the type of dislocation boundaries formed can be predicted based on the slip systems expected to be active [3,4] and that this can improve the prediction of mechanical properties [5]. The aim of the present paper is to take these results one step further towards a detailed understanding of the boundary formation mechanisms through detailed characterisation of the dislocations present in the boundaries, focusing on the relationship between the Burgers vectors in the boundary and the slip systems, as well as on whether the boundaries are low-energy dislocation structures (LEDS) free of long-range stresses [6].

Morphologically, dislocation boundaries fall into two main categories: extended planar boundaries with a specific crystallographic alignment or a three-dimensional arrangement of shorter boundaries forming a fairly equiaxed cell structure [7,8]. In many cases the two types of boundaries coexist, but in grains of certain orientations, e.g. in rolled cube grains and tensile deformed grains of $\langle 1\,0\,0 \rangle$ orientation, only the three-dimensional cells are observed [1].

In this study we investigate the dislocation content and relationship to slip system activity in dislocation boundaries inside near-cube orientation grains in a sample of high-purity Al (99.996%) deformed by rolling to a reduction of 10%. The rolling was carried out at room temperature with roll gap geometries corresponding to plane strain conditions to ensure homogeneous microstructures throughout the sample thickness [9]. Thin foils prepared from the rolled sample by twin jet electropolishing were examined in a JEOL 2000Fx transmission electron microscope operated at 200 kV.

A typical example of the deformed microstructure as seen in the longitudinal plane (containing the rolling and normal directions) is shown in Figure 1. In this study we focus on analysis of boundaries lying in (parallel to) the longitudinal plane, seen as net-like structures in Figure 1

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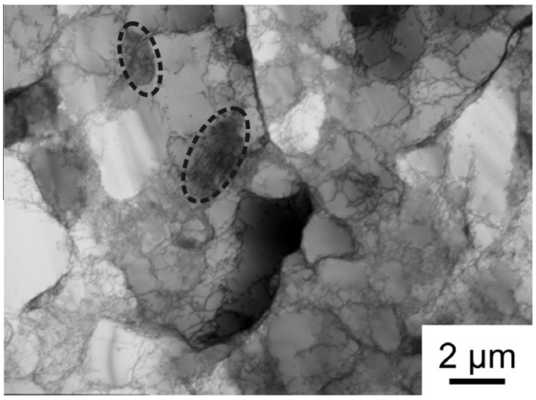

Figure 1. Typical cell structure formed in a grain of cube orientation viewed in the longitudinal plane. Net-like boundaries almost parallel to the longitudinal plane are marked by ovals.

(indicated in the figure by black ovals). To provide a good basis for comparison, the dislocation content was examined in 11 boundaries of this kind, taken from four different near-cube orientation grains and in three different sample locations. Observations were made primarily in the longitudinal plane (eight boundaries), but also in the rolling plane (three boundaries). In the rolling plane, these boundaries are perpendicular to the viewing plane and so appear as narrow string-like features.

The Burgers vectors of the individual dislocations were determined by comparing two-beam diffraction contrast images obtained using different diffraction vectors, **g**. For each diffraction vector, dislocations with a Burgers vector, **b**, that fulfils the condition $\mathbf{g}{\cdot}\mathbf{b} = 0$ become invisible [10]. The weak-beam dark-field technique was employed using (**g**/3**g**) diffraction conditions with a slightly positive deviation parameter to obtain sharp images. The crystallographic direction of each dislocation line was determined by tilting the sample to find crystallographic planes that contain the dislocation line (i.e. satisfy the condition that the dislocation line coincides with the trace of the plane). Once two such planes are known, the dislocation line direction is determined as the intersection line between the two crystallographic planes.

A typical result from the longitudinal plane is shown in Figures 2 and 3, which show the same boundary

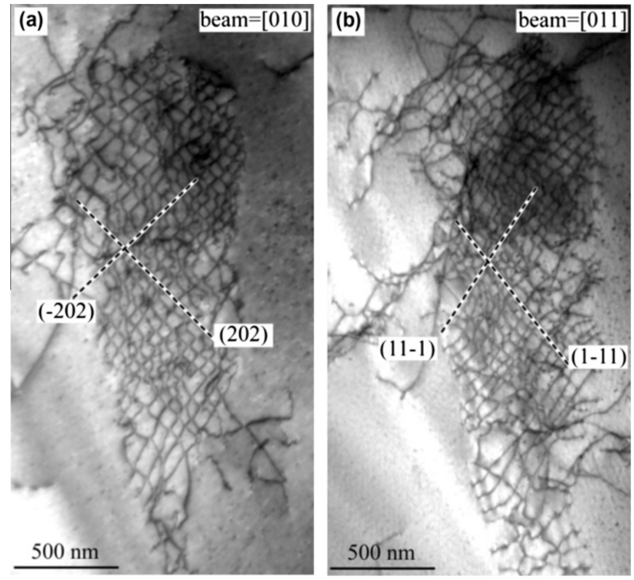

Figure 3. Determination of dislocation line directions. (a) Image obtained in the [0 1 0] beam direction. The dislocation lines are parallel to the traces of $(\bar{1}\,0\,1)$ and $(1\,0\,1)$. (b) Image obtained in the [0 1 1] beam direction. The dislocation lines are parallel to the marked traces of $(1\,1\,\bar{1})$ and $(1\,\bar{1}\,1)$. The scale bar refers to the projection plane, which is close to the boundary plane in (a).

under different viewing conditions. The planar net-like structure consists of two sets of dislocations, each set being composed of a number of parallel straight dislocation lines. The images in Figure 2 were obtained under two beam conditions using diffraction vectors of $[\bar{2}\,0\,0]$, $[2\,0\,2]$ and $[0\,0\,2]$. As illustrated in Table 1, all six Burgers vectors of the $\langle 1\ 1\ 0\rangle$ type should be visible for at least two of the three employed diffraction vectors. However, only two ($\mathbf{b} \parallel a/2[1\,0\,1]$ and $\mathbf{b} \parallel a/2[1\,0\,\bar{1}]$) can be positively identified, assuming that all dislocations are of the $\langle 1\ 1\ 0\rangle$ type. One set of dislocations is visible in all of the images in Figure 2(a)–(c), which, according to Table 1, determines the Burgers vector to be $\mathbf{b1} \parallel a/2[1\,0\,1]$. The other set of dislocations is visible in Figure 2(a) and (c) but invisible in Figure 2(b), and the Burgers vector is therefore $\mathbf{b2} \parallel a/2[1\,0\,\bar{1}]$. It should be noted that for some of the other boundaries investigated other **g**s were employed, including $[\bar{1}\,1\,1]$, $[1\,1\,1]$, $[\bar{2}\,0\,0]$ and

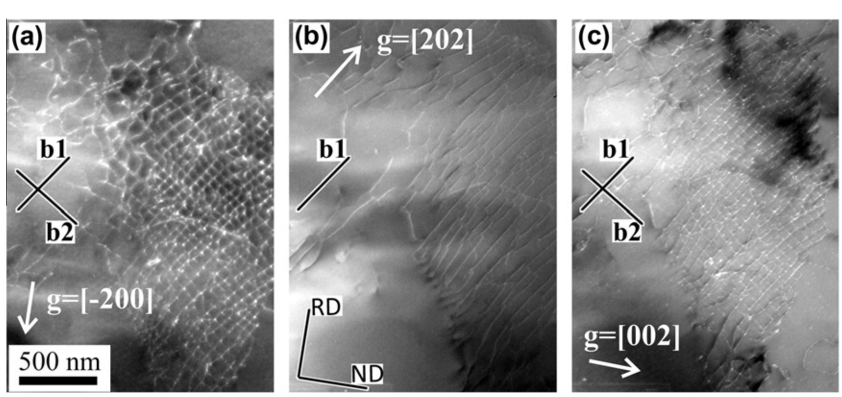

Figure 2. Dislocation configuration of a cell boundary parallel to the longitudinal plane. The local grain orientation is ND/RD = (−0.11 −0.04 0.99)[0.99 0.01 0.12]. As marked on the images, the diffraction vectors are (a) $\mathbf{g} = [\bar{2}\,0\,0]$, (b) $\mathbf{g} = [2\,0\,2]$ and (c) $\mathbf{g} = [0\,0\,2]$.