

THE ORIENTATION DISTRIBUTION IN CHANNEL DIE COMPRESSED Al SINGLE CRYSTALS AS STUDIED BY MICRO-DIFFRACTION TECHNIQUES IN SEM AND TEM

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INTRODUCTION

An aspect still missing in the description of deformation textures concerns the nature of the orientation distribution on a local scale. The way in which the subboundary misorientation develops during cold deformation is well documented, but to which extent neighbouring subgrains are either narrowly scattered around a mean orientation or accumulate into forming medium or long range orientation gradients has only quite recently become a topic of scientific interest. By applying the EBSP/SEM-technique in the analysis of cold rolled commercial purity aluminium, Ørsund et al.¹ clearly demonstrated that a local accumulation of subboundary misorientation was a characteristic aspect associated with any deformation texture component. Working with high purity aluminium. Hjelen et al.² found that within the various stable texture components the local spread in orientation were strongly scattered, typically in the range $\pm 7-10^\circ$ on average. Further, "misorientation-peaks", 30° or more away from the mean matrix orientation, were a typical feature. The main objective of the present work is to investigate the evolution of such "local-orientation-features" in more detail. For this purpose the microtextural nature of aluminium single crystals of different starting orientations and deformed to various strain levels has been investigated.

EXPERIMENTAL

High purity (99.99) aluminium single crystals have been channel die compressed at room temperature to strains of 0.2, 1 and 2 (for details see Ref. 3). In order to examine the effect of starting orientation on the local texture evolution both the stable $\{112\}\langle 111 \rangle$ and the highly unstable $\{001\}\langle 110 \rangle$ orientations have been investigated. Two micro-diffraction techniques have been employed, i.e. analysis of electron back scattering patterns in SEM^{1 2} and convergent beam diffraction patterns in TEM⁴.

RESULTS AND DISCUSSION

Typical microstructures at strains of 0.2 and 1 are given by the TEM micrographs in Fig. 1. As can be seen from the micrographs the substructure is built up of well defined subboundaries even at strains as low as 0.2. In the case of the stable orientation (Fig. 1a) these boundaries are parallel to the trace of the most active primary (111) -slip planes, while in the case of the unstable orientation the subboundaries coincide with the traces of the symmetrically oriented $(\bar{1}11)$ and $(1\bar{1}1)$ slip planes. At large strains the slip pattern remain generally crystallographic in the stable orientation case. At a strain of $\epsilon = 1$ activity also on secondary slip systems are clearly visible in Fig. 1b. In the case of the unstable orientation the subboundary pattern becomes less

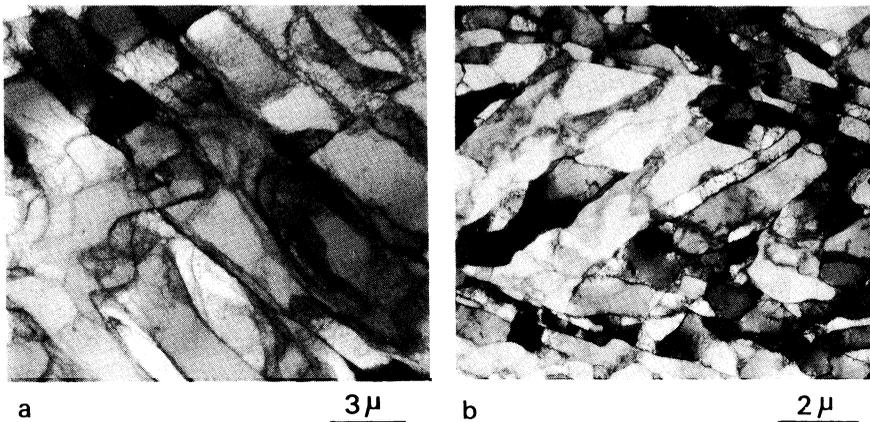


Fig. 1. TEM micrographs, stable orientation. a) $\epsilon=0.2$ and b) $\epsilon=1$.

crystallographic with increasing strains. At large strains ($\epsilon > 1$) the substructure develops a laminated appearance with the subboundaries generally aligned parallel to the deformation direction. (For a more detailed description of the substructure evolution see Ref. 3).

The nature of the orientation distribution on a local scale has been investigated by systematically step scanning the beam along lines selected both parallel and normal to the rolling direction (in transverse section foils (convergent beam) or electro polished transverse section surfaces (EBSP)). By this approach the local variation in orientation over relatively large distances can be mapped with a high degree of accuracy. Typical results from such scans are presented in terms of "continuous-orientation-change-curves" and pole figures in Fig. 2 where a) and b) are due to convergent beam examination and c) are obtained by use of the EBSP-technique. These "zig-zag" curves represent recordings of the orientation changes from subgrain to subgrain, with reference to a pre-selected orientation. The angular deviation of a specific subgrain given on the vertical axis in Fig. 2 represents the minimum angle of rotation which this subgrain in question will have to be rotated to coincide with the reference orientation. With respect to the reference orientation the curve in Fig. 2 simply represent a one-parameter-misorientation-angle-plot. For a complete representation of the spatial variation in subgrain misorientation, also the orientation of the axes of rotation between neighbour pair needs to be included, an attempt of such an analysis is given in Ref. 5. However, the one-parameter misorientation representation in Fig. 2 is useful in order to illustrate the turbulent nature of crystalline orientations on a local scale. Note that at large strains the magnitude of the local turbulence is not significantly affected by the initial orientations (except for occasional transition boundaries in the unstable case). It is further noted that the same type of misorientation scattering is also found in cold rolled high purity aluminium polycrystals².

Another useful one-parameter representation is given in Fig. 3. In these plots the statistical average of the misorientation angle between a randomly selected subgrain and each of its neighbours along a step-scan is recorded. At low strain (Fig. 3a, $\epsilon = 0.2$) this misorientation angle does not vary significantly as going from the next

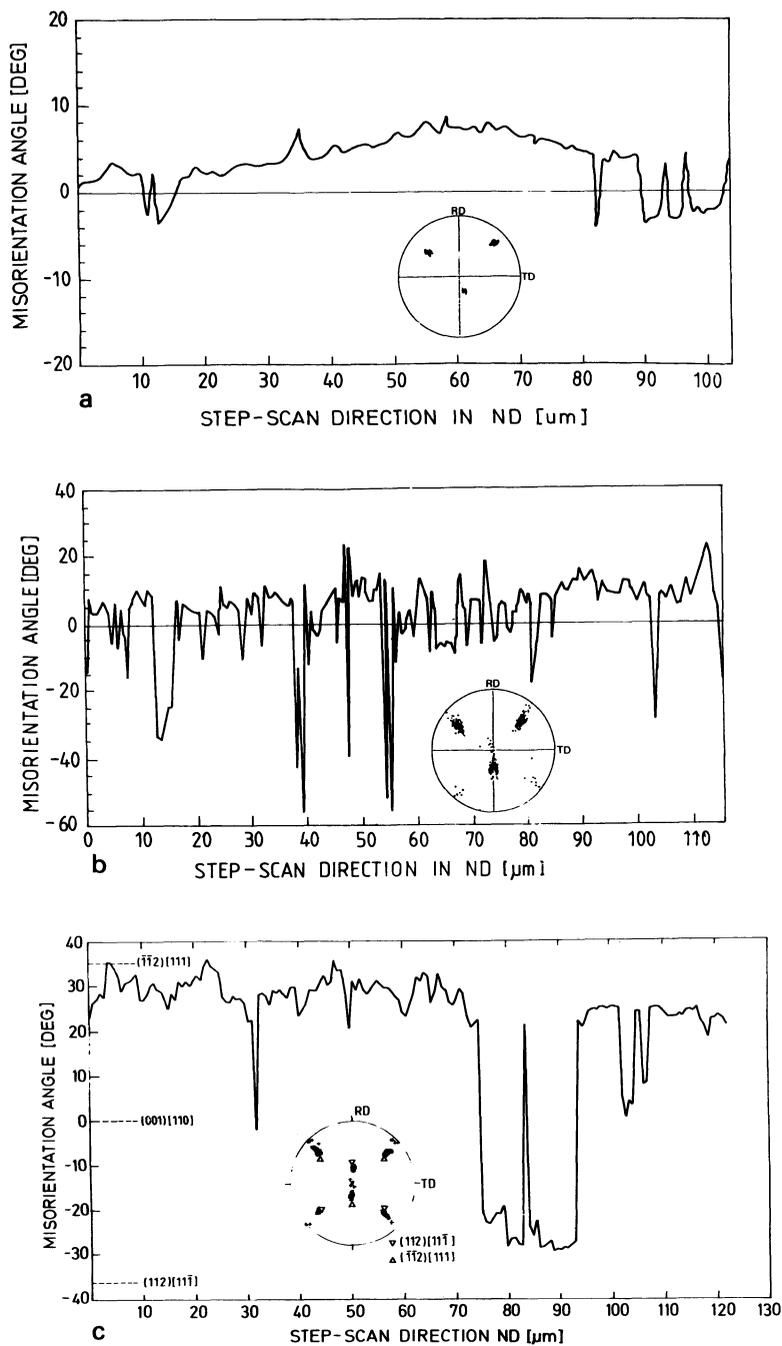


Fig. 2. Step-scan illustrating the change in the subgrain-to-subgrain orientation. a) $\epsilon=0.2$ and b) $\epsilon=1$, both stable orientation, analysed by convergent beam. c) $\epsilon=1$, unstable orientation, EBSD-analysis.

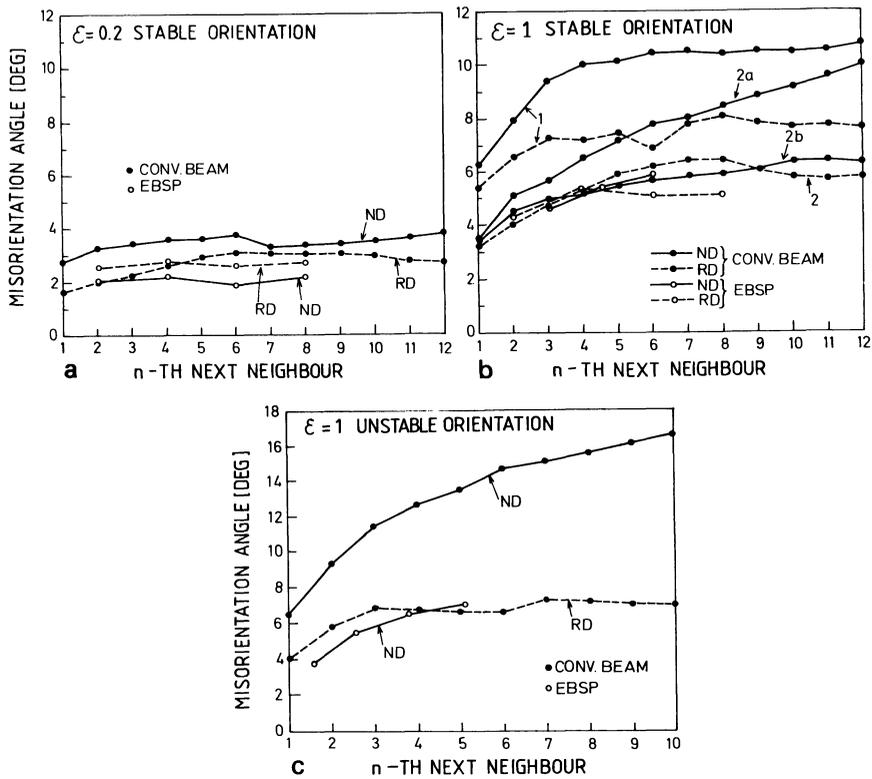


Fig. 3. Variation in misorientation angle between a given subgrain and its neighbours. a) $\epsilon = 0,2$, b) and c) $\epsilon = 1$, stable and unstable orientations respectively.

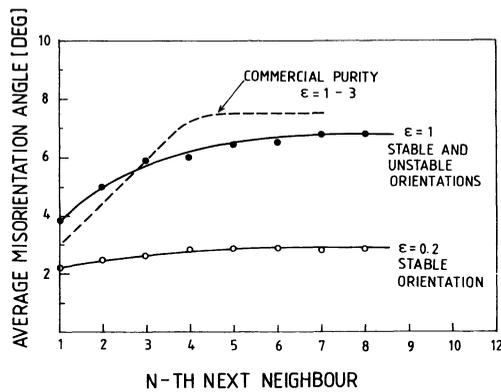


Fig. 4. Average misorientation angle between a given subgrain and its neighbours. (Broken line, Ref. 6).

neighbour towards higher order neighbours. However, at large strains (Fig. 3b and c, $\epsilon = 1$) this picture changes, the misorientation angle as a function of "order of neighbour" increases with a slope which is steepest in the beginning and then flattens out after the 5th-7th order neighbour. This pattern illustrates an accumulation of misorientation on a local scale. However, the magnitude of this local "lattice curvature" varies from region to region of the investigated single crystals as illustrated by the different curves in Fig. 3b and c, representing step-scans from different foils or EBSP-sections. By averaging the results from many such scans (each scan typically 50 - 100 μm in length), approximately the same misorientation-accumulation-pattern is obtained irrespectively of starting single crystal orientation. A typical plot, representative for the present single crystals are shown in Fig. 4. This figure also include results obtained from polycrystalline aluminium⁶ and it is interesting to note how similar poly- and single crystal data turned out to be.

As a final comment it is pointed out that in highly deformed aluminium (plane strain, $\epsilon > 1$) the local orientation distribution is (within any texture component) both highly turbulent and at the same time spatially organized in such a way as to produce a local accumulation of misorientation. None of these observations are predicted by any existing work hardening or crystalline plasticity theories.

REFERENCES

1. R. Ørsund, J. Hjelen and E. Nes, Scripta Met., 23, 1193(1989).
2. J. Hjelen, R. Ørsund and E. Nes, Submitted to Acta Met.
3. J.F. Butler, Jr., Ph.D. Thesis, University of Pittsburgh (1990).
4. H. Weiland and R. Schwarzer, Experimental Techniques of Texture Analysis, ed. H.J. Bunge, DGM Verlag, Oberursel 1985, 301-313.
5. H. Weiland, J. Liu and E. Nes, this conference.
6. R. Ørsund, T. Furu and E. Nes, to be published