Cube texture formation during the early stages of recrystallization of Al-1%wt.Mn and AA1050 aluminium alloys

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Abstract. The cube texture formation during primary recrystallization was analysed in plane strain deformed samples of a commercial AA1050 alloy and an Al-1%wt.Mn model alloy single crystal of the Goss{110}<001> orientation. The textures were measured with the use of X-ray diffraction and scanning electron microscopy equipped with an electron backscattered diffraction facility. After recrystallization of the Al-1%wt.Mn single crystal, the texture of the recrystallized grains was dominated by four variants of the S{123}<634> orientation. The cube grains were only sporadically detected by the SEM/EBSD system. Nevertheless, an increased density of <111> poles corresponding to the cube orientation was observed. The latter was connected with the superposition of four variants of the S{123}<634> orientation. This indicates that the cube texture after the recrystallization was a 'compromise texture'. In the case of the recrystallized AA1050 alloy, the strong cube texture results from both the increased density of the particular <111> poles of the four variants of the S orientation and the ~40°(<111>)-type rotation. The first mechanism transforms the S<sub>def</sub>-oriented areas into S<sub>rex</sub> ones, whereas the second the near S-oriented, as-deformed areas into near cube-oriented grains.

1. Introduction
The crystallographic aspects of the new orientation generation during recrystallization have been the subject of discussion for several years. This concerns both the description of the mechanisms 'controlling' the texture transformation as well as the relations between the 'texture images' of the deformed and recrystallized states. It is widely accepted that the starting point of new orientations are the components found in the deformed state, e.g. [1-6]. It means that the orientation of each new grain originates from the 'family' of orientations in the deformed state. However, a clearly observable change of the nuclei orientation with respect to the as-deformed orientations, e.g. [7,8] suggests a connection with the dislocation removal from a given region, e.g. [7-13]. The literature on recrystallization pays a lot of attention to the role of the high-angle boundaries (HABs) and the conditions of their migration. Nevertheless, literature is unable to provide a comprehensive answer to the question about the mechanisms controlling the formation of nuclei inside the deformed structure and especially to explain the different nuclei orientation with respect to the deformed matrix. It is clear that, in deformed polycrystalline materials, the fraction of the HABs is generally very large, and these boundaries play a key role in nucleation. However, nucleation by way of migration of the deformation-induced HABs has not been reported as unique or universal. For instance, in the case of face centred cubic (fcc) crystallites with stable orientations in plane strain compression (\{110\}<112>
or \{110\}<001\}, HABs are not present or are only very rare, yet nucleation still occurs, e.g. [8,14,16,17]. Therefore, in deformed materials, where the fraction of low-angle boundaries (LABs) is still large, the mechanisms based on the migration of the LABs are also needed for the early stages of recrystallization. It is assumed that the thermally activated migration of the LABs combined with the movement of the dislocations stored inside the (sub)grains and/or the diffusion of vacancies, e.g. [18, 19] may play the key role in the orientation change of a ‘chosen area’ with respect to the deformed matrix, and in the formation of HABs (in fact, it is the ‘transformation’ of the LABs into high-angle ones). Still, the nature of this process is not clear.

The understanding of the nucleation mechanisms is critically important for the description of the origin of the cube texture in fcc metals. In particular, the recrystallization texture of aluminium alloys is often represented by a strong cube{100}\<001\> orientation with some scattering in the rolling direction towards the Goss\{110\}<001\> orientation. Some authors stated that the cube-oriented grains in highly deformed aluminium alloys are nucleated from deformation, transition or shear bands preferentially located in the Cu or ND-rotated Cu-texture components, e.g. [20]. Yet the details of the transformation that leads to the extreme sharpness of the cube texture are not understood.

The aim of this study is to examine crystallographic aspects of the strong cube texture formation during the early stages of recrystallization of two aluminium alloys. Particular attention was paid to the description of the misorientations across the recrystallization front and on possible dislocation mechanisms during the primary recrystallization. The use of aluminium alloys with a high stacking fault energy (SFE) practically excludes mechanical and recrystallization twinning, and it facilitates identifying the mechanisms responsible for the texture transformation during the initial annealing stages. The strong cube texture formation were analysed by means of a scanning electron microscopy equipped with an electron backscattered diffraction (SEM/EBSD) facility as well as (for more global scale analyses) by X-ray diffraction.

2. Experimental
The materials were a commercial polycrystalline AA1050 alloy and a (single phase) Al-1\%wt.\(\text{Mn}\) single crystals of the Goss\{110\}\<001\> initial orientation. Samples of 10 x 10 x 10 mm\(^3\) were plane strain compressed (PSC) in a channel-die. The multi-stage deformation was performed at ambient temperature at the initial strain rate of 0.02 s\(^{-1}\) to the logarithmic strain of 0.51 (40%). In order to limit the friction between the sample, the punch and the walls of the channel-die, each sample was wrapped in 0.2 mm thick Teflon\textsuperscript{TM} tape. The deformed samples were annealed for 1h at temperatures ranging between 100\(^{\circ}\)C and 450\(^{\circ}\)C with the purpose to obtain different stages of recrystallization. The deformed and recrystallized microstructures were analyzed in the ND/ED section (where: ND and ED denote, the normal and the extension directions, respectively) with the use of SEM - QUANTA 3D FEG, equipped with a field emission gun and an EBSD facility. The SEM was working in the automatic beam scanning mode and the maps were created with step sizes ranging between 100 nm and 700 nm. The numbers of correctly indexed points was always above 97\%. In each case, the orientation maps covered the area of at least 100 \(\mu\)m x 100 \(\mu\)m. In the \{111\} pole figures corresponding to the SEM orientation maps, the indices \(\langle hkl \rangle <uvw \rangle\) represent the texture component which has the \(\langle hkl \rangle\) plane parallel to the ED/TD plane (where: TD denote the transverse direction) and the \(\langle uvw \rangle\) direction parallel to ED. At the sample scale, the textures were measured on the ED/TD plane with the use of the Philips X Pert PW-1830 X-ray diffractometer.

3. Results and discussion
SEM/EBSD and X-ray diffraction were applied to reveal the inherent characteristics of the microstructure and texture development during the PSC and subsequent annealing. The use of single crystals of a stable orientation enables one to precisely define the orientation relationship developed when the nuclei are formed in the as-deformed structures free from HABs. Those results are related to the mechanism responsible for the new orientation formation observed in the commercial AA1050
alloy. In the two materials, the orientations of the nuclei were similar, despite the fact that the deformation textures were quite different.

3.1. Deformation and recrystallization behaviour of an Al-1%wt.Mn single crystal of Goss\{110\}<001> orientation.

3.1.1. Microstructure and texture of the as-deformed state. At a logarithmic strain of 0.51, the microstructure of the Goss-oriented crystal consist of two complementary sets of elongated microbands delineated by dislocation walls (Fig. 1a). Both sets of dislocation walls were inclined very close to the expected traces of the active \{111\} slip planes (Fig. 1b). The adjacent elongated cells displayed opposite rotation senses, as reported earlier [16, 17, 21]. Most of the rotations occurred around the axes lying close to TD, both away from and almost back to, the initial crystal orientation. The misorientation angles were significantly below 10°. Consequently, the textures of the areas corresponding to the above microstructures confirm the stability of the initial orientations at a logarithmic strain of 0.51, as shown by the \{111\} pole figures of the mapped areas.

Figure 1. Microstructures and corresponding \{111\} pole figures observed in a Goss(110)[00-1] oriented single crystal of Al-1%wt.Mn alloy, (a, b) as-deformed state and (c, d) after recrystallization at 688°C for 480 s. In (d) the black lines mark the four positions of S\{123\}<634> orientations, whereas bold blue (dashed) line - cube orientation. For better visualized the microstructures details in the deformed and recrystallized states the ‘cubic’ and IPF (||Y) color codes were applied in (a) and (c), respectively.

3.1.2. Recrystallization behaviour. Recrystallized grains were analysed inspects large sample areas (Fig. 1c). Only a small fraction of the grains was observed to be twinned. For the statistical analysis of the misorientation relation between the deformed material and the recrystallized grains, the few twinned grains were ignored. The misorientation relation was calculated between the ‘average’ (stable) orientation of the deformed state and the orientations of the recrystallized grains. Figures 1b and d show the orientations of the as-deformed crystal and the recrystallized grains. The grain orientations form groups, which are symmetric with respect to the external directions. Those groups of orientations can be described, in an idealized form, by the four symmetric S\{123\}<634> orientations. This symmetry results from the positive and negative rotations of the new grain
orientations around the axes grouped near the normals of the \{111\} planes, highly active during deformation. A comparison of the pole figures of the deformed and the recrystallized states shows that the <111> poles of the recrystallized grains do not coincide with the <111> poles of the deformed areas or, more precisely, overlapping of the <111> rotation axes was only rarely observed (Fig. 1d). It is important to note that no new grains of the near Goss orientation were observed.

Figure 2. The distribution of misorientation axes and angles between the average deformed orientation and recrystallized grains (taken from the orientation map presented in Figure 1c in the Goss(110)[00-1] oriented single crystal of Al-1\%wt.Mn alloy). Red circles mark the orientation in the deformed state.

The frequency distribution of the misorientation angles and axes between the new grains and the ‘average’ as-deformed orientations is presented in Fig. 2. The plot shows that most of the angles across the migrating recrystallization front are within the range of 25-55°. The scattering of the misorientation axes is larger than that of the orientations in the deformed state. The spread of the \{111\} poles of the as-deformed state is schematically marked by the red areas. Basically, the majority of the misorientation axes across the recrystallization front are concentrated around the <111> poles of the two highly stressed slip planes during the deformation. A ‘grouping’ of rotation axes near the two remaining the <111> poles is weaker. Only occasionally, other misorientation axes were observed.

3.1.3. Cube component. It is clear that the Goss\(\{110\}\langle001\rangle\)-oriented areas in the as-deformed state are transformed during the recrystallization into four variants of \(S^{(IV)}\)-oriented grains. The transformation, which occurs during the formation of the new grain could be described as a result of a (+/-) \(\pm30-40°\) rotation of the crystal lattice of the as-deformed areas around the selected <112> axes near the normal of the most active \{111\} slip planes, and can be schematically presented as:

\[
\text{Goss}\{110\}\langle001\rangle_{\text{deformed}} \rightarrow (\pm30-40°<112>\text{ rotation}) \rightarrow \sim S\{123\}<634>^{(IV)}_{\text{recrystallized}}
\]

The description of the mechanisms responsible for the formation of recrystallization grains with new orientation, given by Misyczek et al. [8], combines the rapid growth along specific planes with the rotation of the crystal lattice of new grains around the specific axes. The thermally activated movement of the dislocations can lead to the creation of a twist or tilt grain boundary, due to the screw or edge character of the dislocations, respectively. In an idealized form, if the dislocations stored along the plane of growth are pure screw (and their Burgers vectors are orthogonal), their movement leads to a pure twist boundary with a rotation axis perpendicular to the boundary plane. If the dislocations are pure edge, their movements leads to a pure twist boundary with a rotation axis parallel to the boundary plane, as discussed in [8]. It is assumed here that the rotation around the <112> direction may result from the thermally activated (pure edge) dislocations movement along the \{111\}<011> ‘systems’. This mechanism leads to tilt boundary formation and can be responsible for rapid growth. A similar explanation of the phenomenon was given in [8-11], based on the analysis of the grains showing growth anisotropy. The cube grains (within 20°) are observed very rarely after the recrystallization. However, an increased density of <111> poles in the positions typically occupied by the cube
orientation was observed. The explanation of this phenomenon can be based on the 30-40°<112> crystal lattice rotation during the recrystallization from the Goss to the S orientation and superposition of four variants of the S orientation. It can be clearly seen that one of the <111> poles of each of the four variants of the S orientation marked the position of the cube orientation. This indicates that the cube texture after the recrystallization was a 'compromise texture'. This 'artefact', was even more clearly observed in the case of the polycrystalline AA1050 alloy (to be discussed in the next section).

3.2. Recrystallization behaviour of commercial AA1050 alloy

3.2.1. Microstructure and texture of the as-deformed state. Figure 3a shows the microstructure of the sample deformed to 40%. The microstructure has a large fraction of HABs (misorientations greater than 15°) and LABs (misorientations ranged between 2° and 15°). The microstructure is composed of elongated (in fact, flat) grains subdivided by a network of LABs. The width of the grains (measured in the ND/ED section) range between 15 µm and 30 µm, whereas the length is between 50 µm and 100 µm. The longer axis of the grains are nearly parallel to ED. The development a typical copper-type rolling texture composed of four variants of the S{123}<634> orientation with scattering towards the brass{110}<112> and C{112}<111> orientations was observed, as presented in Fig. 3b. Based on the X-ray diffraction and the SEM/EBSD measurements, it is clear that the cube component is not present in the deformed state (Fig. 4a). The orientations of the dominant fraction of the deformed grains have the <111> direction, nearly parallel to TD. Very often, neighbouring layers were observed to be strongly misoriented and a near-twin orientation relationship sometimes occurred between those areas.

![Figure 3.](image)

3.2.2. Microstructure and texture after annealing. In our recrystallization experiments, clear separation was observed between the deformed (or recovered) and the recrystallized areas. The nuclei were observed as single isolated ones or grouped into parallel chains reflecting the 'directionality' of the deformed microstructure. It is not clear, at the moment, why some layers of a given variant of the S orientation undergo recrystallization, whereas others are still only recovered.

The textures of the as-deformed and annealed states (in sample scale) were analysed by X-ray diffraction. For annealing at temperatures up to 250°C the {111} pole figures show that, the intensity and the exact position of the main texture components after annealing were similar to those observed for the as-deformed state (see Fig. 4a). At temperatures of 300°C and above, the development of a strong cube texture component is observed, as presented in Fig. 4b. Figure 5a shows an orientation map of the sample partly recrystallized at 300°C for 1h. The four variants of the S orientation are
marked in different colours: S1 - (123)[6-43] – maroon, S2 - (231)[3-46] – navy blue, S3 - (213)[-3-6-4] – lime green and S4 - (231)[-34-6] – fuchsia. Additionally, the cube-oriented areas are marked in yellow. In all the cases, a scattering of 20° from the ideal position was applied. Basically, the nuclei were largely confined to particular layer(s). However, for longer annealing times, the grains grew also in the ND direction to widths larger than those of the layers (Fig. 5a), respect from just before. A majority of the new grain orientations was classified as cube, but some of the grains were identified as S-oriented. Also a small number of grains were classified as ‘other orientations’ (Fig. 5b). Possible mechanisms leading to nucleation of the near S-, cube- and Goss-oriented grains are discussed below.

![Figure 4](image1.png)

**Figure 4.** The {111} pole figures showing global textures measured by X-ray diffraction of: (a) the deformed sample of AA1050 and (b) after annealing at 250°C for 480 s. Thin black lines mark the four positions of S orientations, whereas bold black (dashed) line – the cube orientation.

![Figure 5](image2.png)

**Figure 5.** The early stages of recrystallization observed in the AA1050 sample deformed to logarithmic stain of 0.51 and subsequently annealed at 300°C for 1h. (a) SEM/EBSD orientation map and corresponding {111} pole figure. The four variants of the S orientation are marked in different colors: S1 - (123)[6-43] – maroon, S2 - (231)[3-46] – navy blue, S3 - (213)[-3-6-4] – lime green and S4 - (231)[-34-6] – fuchsia, whereas the cube(100)[001] oriented areas are marked in yellow. The step size was 700 nm.

*S-oriented grains.* Figure 6a shows a typical case of the S-oriented grain growing along the structure of the deformed/recovered grains. The new grain orientation is similar to one of the deformed/recovered grain (Figs. 6b and c). Therefore, the misorientation calculated between the deformed/recovered areas and the new grain is low (misorientations <10°). It is clear that the growth along ND is hindered. The misorientation is relatively large with respect to the other two S-oriented
grains (two other variants of the S orientation) where, a HABs of the ~30-40° <111>-type is formed across the recrystallization front. This enables a rapid growth along ED. One may assume that the new, recrystallized grain is growing from the as-deformed areas of nearly the same orientation which is often discussed in literature, e.g. [20]. However, if the results of the recrystallization experiments on single crystals of a stable orientation (§ 3.1.2. and, e.g. [7,8,23]) where a new grain orientation similar to one in the deformed state has never been observed also applies for the AA1050 polycrystalline sample, other explanation have to be analysed. One may therefore speculate that the new grains nucleate from one of the two other ‘S-oriented’ areas of the deformed state and grew quickly into those S-oriented bands. In this way, the formation of the new grain of the S orientation (rotation around the normal of the most active {111} slip plane - HA) can be schematically presented as:

\[ -S\{123\}<634>^{(I-V)} \text{deformed} \rightarrow (-30-40° <111>^{HA} \text{rotation}) \rightarrow -S\{123\}<634>^{(I-V)} \text{recrystallized} \]

This means that the S-oriented grains of the given variant nucleate inside the as-deformed areas of the other S variant. In this case, again, a twist boundary is formed at the head of the recrystallization front.

![Image](image.png)

**Figure 6.** Crystallography of grain growth at the early stages of recrystallization observed in a sample of AA1050 alloy deformed to a logarithmic stain of 0.5 and then recrystallized at 300°C for 1h. Details from map presented in Fig. 5. (a) and (d) Orientation maps and corresponding (b, c) and (e, f) {111} pole figure. R^S – S oriented new grain, R^C – cube-oriented new grain. (‘Arbitrary’ color codes was applied).

**Cube-oriented grains.** Figure 6d shows a cube-oriented grain growing into S-oriented areas of the as-deformed state. A new grain orientation is presented in Fig. 6e. The bottom part of the grain is in contact with a layer in which the size of the cells forming the internal structure is significantly larger than the one observed in the other as-deformed areas. This suggests a more advanced stage of recovery here. Moreover, those cell orientations differ from the orientations typically observed in the deformed state (Fig. 6f). The spread of the orientations of strongly recovered cells mark the transition from the near S to the cube orientations and suggests the possible rotation mechanism (rotation around the normal of the less active {111} slip plane - LA) that can be described schematically as:
\[ -S\{123\}<634> \text{ deformed} \rightarrow (-40-50^\circ<111>^\text{LA} \text{ rotation}) \rightarrow -\text{cube}\{100\}<001> \text{ recrystallized} \]

The new grain is elongated along the boundaries of the as-deformed flat grains. Therefore, it is likely that the preferred growth is connected with rapid migration of the HAB along ED, i.e. the new grain grows preferentially along the boundaries of the flat grains.

**Other recrystallized grain orientations.** There is also a small quantity of grains characterized by other orientations. Those grain orientations are not random and a majority of them represents the ED- or ND-rotated cube\{100\}<001> orientation. However, the grouping some of the new grain orientations near the Goss position can suggest a mechanism as:

\[ -S\{123\}<634> \text{ (I-IV) deformed} \rightarrow (40-45^\circ<112>^\text{LA} \text{ rotation}) \rightarrow -\text{Goss}\{110\}<001> \text{ recrystallized} \]

Which is in a way the reverse transformation type to that seen in the single crystal. To sum up, the present analysis suggest that, despite the fact that the new grains have a 40°<111> misorientation relation with respect to the deformed areas, no direct evidence was found supporting the idea that possible 40°<111> relations occurs at a higher rate than relations with other misorientation relations.

### 4. Conclusions

Based on local orientation measurements and X-ray diffraction, the present results provide detailed information about the development of the recrystallization microstructures and textures in a commercial AA1050 alloy and a single crystal of a model Al-1%wt.Mn alloy it was found that:

(i) Although the deformation texture of the two materials were different, in both materials nuclei of new grain orientations developed during recrystallization are similar, but different from those in the deformed matrix.

(ii) After recrystallization of the Al-1%wt.Mn single crystal of the Goss orientation, the texture of the recrystallized grains was dominated by four variants of the S\{123\}<634> orientation. Cube grains were only sporadically detected by the SEM/EBSD system. Nevertheless, an increased density of the <111> poles corresponding to the cube orientation was observed. This effect was connected with the formation of nuclei with orientations near the four variants of the S\{123\}<634> orientation.

(iii) In the case of the recrystallized AA1050 alloy, the strong cube texture after recrystallization results from both the increased density of the particular <111> poles of the four variants of the S orientation and the ~40°(~<111>)-type rotation of the as-deformed orientations. The first mechanism transforms the S\text{def}-oriented areas into (another variant) S\text{rec} ones, whereas the second, the near S-oriented, as-deformed areas into near cube-oriented grains.

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