Crystallographic aspects of nucleation and growth during primary recrystallization in stable single crystals of Al and Al-1%Mn alloy

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Abstract. The early stages of recrystallization have been characterized in stable, single crystals of Al and an Al-1%wt. Mn alloy to rigorously quantify the orientation relations between nuclei and simple deformation structures. Goss\{110\}<001> and brass\{110\}<112> oriented samples were deformed in a channel-die to develop a homogeneous structure composed of two sets of symmetrical microbands and then lightly annealed. SEM/EBSD analyses demonstrate a strong relation between as-deformed orientations and the limited number of recrystallized grain orientations. The misorientation angles across the recrystallization front are mostly grouped in the ranges of 25-55° around axes located near, but not at, the normals of all four \{111\} planes. In some cases the orientation of the growing grain transforms through the formation of a first generation twin with the twinning plane normal lying near the rotation axis. The intensity of new grain nucleation in near-surface areas is significantly greater than those observed in central parts of the sample. Many new grains grow from the surface into the sample centre mostly along the \{111\} planes. However, irrespective of the section new grains possess the same type of orientation relationship with respect to the deformed state.

1. Introduction

The orientations of recrystallized grains are considered to 'originate' from the orientations of the deformed matrix regions from which they grow, e.g. [1, 2]. Particles [3-5], strain heterogeneities [6-10], original grain boundaries (including deformation twin boundaries) [8], etc. are the privileged places of nucleation. However, in the absence of these well-known nucleation sites, recrystallized grains have been shown to appear extensively in the deformed crystals of stable orientations, where high angle boundaries occur only accidentally [11-13]. In these simple cases the conventional mechanism of primary nucleation, based on the presence of pre-existing nuclei in the deformed structure possessing a high angle boundary, cannot explain texture change during recrystallization.

According to the oriented growth theory the growth potential of a given nucleus depends on the orientation difference at the migrating recrystallization front [14] and the fastest growing nuclei contribute to the final recrystallization texture. (However, it is still an open issue to define the most advantageous disorientation across the migrating recrystallization front that is favourable to rapid growth). This assumption leads to the conclusion that in a uniform deformation microstructure with only one misorientation relation, almost equiaxed grains should be formed during recrystallization. However, in the early stages of recrystallization of grains without (significant) deformation gradients a considerable number of recrystallized grains exhibit a strongly elongated shape [11]. This anisotropic
growth suggests that some directions or planes are more privileged for growth, despite a near-uniform disorientation relation in all directions. This leads to the important conclusion that it may not be the disorientation across the recrystallization front but probably the type of migrating grain boundary, *e.g.* their twist or tilt character, which is the most important factor controlling migration velocity, as discussed several times in the past, *e.g.* [1, 2, 4]. Since the movement of pure screw or edge dislocation groups along specific planes leads to a twist or tilt boundary [4], respectively, identification of the boundary type across the recrystallization front is crucial for a description of the dislocation mechanisms responsible for grain growth. Identifying the boundary character requires an exact knowledge of the 3D spatial situation of the boundaries between the new grain and as-deformed neighbourhood, together with the disorientation axis with respect to the boundary plane. This is possible by detailed inspection of three perpendicular sections, *e.g.* ND/ED and ND/TD planes, where ND, ED and TD are the normal, extension and transverse directions, respectively.

Moreover, there are also other aspects of the phenomenon such as the influence of the free surface on the morphological and crystallographic aspects of recrystallization and the role of recrystallization twinning in texture transformation even in high SFE metals. The intensity of recrystallization in near-surface areas is significantly greater than that observed in central parts of the sample and new grains grow from the surface to the sample centre. But it is not clear if the orientation of grains nucleating near the surface and in the centre are similar or not. This leads to the question – is it possible to compare the textures obtained in recrystallization experiments of bulk samples with those obtained during in-situ heating in SEM or TEM, where the influence of free surface is evident. As regards recrystallization twinning in fcc metals an important question is whether all boundaries showing twin-oriented relationship (60°<111>) are in fact twin type boundaries, *i.e.* does the boundary between the twin-related areas coincide with a {111} plane?

In order to clarify those problems, it is advisable to perform experimental research under conditions where the number of 'free parameters' affecting this transformation is minimized. Therefore, the current experimental investigations have focused on an analysis of transformations which occur in model single crystals, with stable orientations, of a face centered cubic (fcc) metal (Al and Al-1%Mn). This approach to the problem makes its description simpler and clearer since the stability of the orientation during deformation enables one to precisely define the orientation relationship which appears when the 'primary nucleus' is formed. Another important aspect is that the deformation of stable orientation crystallites (up to large strains) practically does not lead to the appearance of high-angle boundaries. In the absence of high-angle boundaries, the conventional mechanism of primary nucleation, based on the presence of pre-existing nuclei in the deformed structure, cannot be accepted as valid for explaining a texture change during recrystallization. The basic research technique employed local orientation measurements based on scanning electron microscopy (SEM/EBSD). Since the ‘tilt or twist’ nature of a grain boundary can only be precisely defined by measurements in three sections the disorientation characterization across a recrystallization front is complemented by a spatial description of boundaries of grains showing anisotropic growth.

2. Experimental

The single crystal bars of high purity Al and Al-1%wt.Mn alloys were grown by directional solidification. The Goss (110)[00-1] and brass (110)[1-1-2] oriented samples were carefully cut from the bars using a wire saw to dimensions of 10 x 10 x 10 mm³ (height x width x length). The deformation was carried out by plane strain compression (PSC) in a channel-die. Figure 1 shows the configuration of the most active slip systems for ideal Goss and brass orientations. The multi-stage test was performed at 293 K at the initial strain rate of 0.02 s⁻¹ to a logarithmic strain of 0.51 (ε=ln(h_initial/h_final)). In order to limit friction between the sample, the punch and the walls of the channel-die, each sample was wrapped in 0.2 mm thick PTFE films. To investigate the interrelation between as-deformed and recrystallization textures, the deformed samples were annealed to the first stages of primary recrystallization, typically in the range of 400 - 700 K for less than 60 s. Annealing was carried out in an air furnace. All specimens were water quenched post-annealing. The new grain
orientations were mostly analysed on the ND/ED plane in the centre section of the annealed sample. However, for a 3D analysis of the grain boundaries additional orientation maps were made on two other sections, i.e. ND/TD and ED/TD. The influence of the section depth was investigated on the free surface as well as on $\frac{1}{4}$ and $\frac{1}{2}$ sample thickness. Prior to annealing all the samples were mechanically ground on all faces to 4000 grid SiC paper and electropolished. The samples were characterized by scanning electron microscopy (SEM) using a JEOL 6500F or Zeiss Supra 55VP. The SEM's were equipped with an electron backscattered diffraction (EBSD) system, working in automatic beam scanning mode.

**Figure 1.** Stereographic projections showing configuration of the most active slip systems in ideal (a) Goss (110)[00-1] and (b) brass (110)[1-1-2] orientations [12].

3. Results and discussion

3.1. Deformed state

3.1.1. Microband structure and orientation stability. At a logarithmic strain of 0.51, the microstructures of Goss and brass oriented crystals consist of two complementary sets of elongated microbands delineated by dislocation walls, as presented in Fig. 2 for Al and Al-1%Mn alloy of Goss orientation and also discussed many times in the past, e.g. [15-19]. Based on SEM/EBSD measurements, rectangular cells were formed as a result of microband intersections. Both sets of dislocation walls were inclined very close to the expected traces of the active {111} slip planes; they were always within the scatter limit of normals to the {111} planes. The orientations of areas corresponding to the above microstructures confirm the stability of the initial orientations at logarithmic strains of 0.51. The samples revealed a relatively small orientation spread, as shown for Al-1%Mn alloy single crystals of brass orientation (Fig. 3), even if the initial orientation deviated a few degrees from the ideal position. However, in the case of initial samples showing ‘larger deviation’ from ideal positions cyclical orientation changes were superimposed on a strong, continuous lattice rotation. The majority of adjacent microbands displayed opposite rotation senses, as reported earlier for an Al-0.3%Mn alloy, e.g. [20].

3.1.2. The as-deformed rotation axes and angles. The distribution of disorientation axes between adjacent pixels of mapped areas of the deformed single crystals is visualized by stereographic projections for the ED/TD reference system, as presented in Fig. 4a and b for Al single crystals of Goss and brass orientations, respectively. The maximum density of disorientation axes occurred in the vicinity of TD but with some systematic variations. In the case of Goss-oriented crystals (for which
the intersection of the active (111) and (11-1) planes coincides with TD, there is a small spread away from TD so that the most frequent axes were near the closest <112> axes to TD. Moreover, the spread is sufficiently wide to observe disorientation axes near [-221]. In the case of brass-oriented crystals, the maximum disorientation axis is located between TD and the intersection trace of the active (111) and (11-1) planes. The disorientation axes display a systematic departure from TD with a wide spectrum of disorientation axes between [-11-1] (||TD), [-22-1] (location of maximum density), [-110] (intersection trace of slip planes). For both orientations the distributions of the disorientation angles between particular pixels of the orientation maps were significantly below 10° (see Fig. 3).

**Figure 2.** Microstructures and corresponding {111} pole figures observed in Goss oriented single crystals of (a) Al and (b) Al-1%Mn alloy. SEM/EBSD local orientation measurements in ND/ED section with a step size of 100nm. IPF colour code was applied [12].

**Figure 3.** Misorientation line scan along ED and misorientation axes distribution in misorientation relation between neighbouring pixels of the deformed state. Al-1%wt.Mn alloy single crystal of brass orientation deformed to a strain of 0.51.

3.2. Early stages of recrystallization
After short annealing only discontinuous recrystallization was observed and these partially recrystallized structures give rise to new orientations against a background of well-defined microtexture components of the as-deformed state. The nucleation characteristics observed at the SEM scale (in the middle of the bulk sample) turn out to be the same for both metals and orientations. The new grains were uniformly distributed in the deformed matrix. Only a small fraction of the grains was twinned. The microstructures were composed predominantly of single, isolated grains that show an elongated shape and some quantity of grains forming compact chains. The longer axes of some elongated grains in the longitudinal section were closely parallel to the traces of the active \{111\} slip planes.

Figure 4. The distributions of misorientations in the as-deformed state. Stereographic projections showing the \langle112\rangle and \langle110\rangle directions for Al of (a) Goss and (b) brass orientations. White and black circles marked positions of \langle011\rangle and \langle112\rangle directions, respectively.

In the following sections the different aspects of the early stages of recrystallization will be discussed. They include: (i) a description of preferential grouping of new grain orientations, (ii) a determination of the grain boundary type between the as-deformed areas and new grains (analysis of anisotropic grain growth on three perpendicular sections), (iii) recrystallization twinning during the early stages of recrystallization, and, (iv) the influence of the free surface on the intensity and crystallography of recrystallization.

3.2.1. The orientations of recrystallized grains. Orientations of single isolated nuclei/grain. As can be seen in Fig. 5 the orientations of recrystallized grains are different from those of the deformed state. A general rule is that the \langle111\rangle poles of the deformed crystal do not coincide with the poles of the recrystallized grains, or more precisely coincide very rarely [11]. Despite the scatter of the recrystallized grain orientations, it can be claimed with certainty that they are not random. Only a finite number of groups of recrystallized grain orientations were observed, as presented in Fig. 5 for Al-1%Mn single crystals of Goss and brass orientations.

In the case of both orientations the ‘texture image’ of the first recrystallized grains is quite symmetrical with respect to the external coordinate directions. A primary nucleus of homogeneous orientation is related to the adjacent deformation orientation by a rotation around an axis located near the pole of one of the four \{111\} planes of its deformed environment. It was noticed that if the formation of a grain orientation occurs by a positive rotation around a certain axis in a given region, grains characterized by a negative rotation appear in neighbouring regions. (Note that if those grains
meet a pseudo-twin boundary is formed, i.e. the grains possess a twin orientation relationship but the common \{111\} plane does not coincide with the boundary plane). This phenomenon of opposite local rotations is observed clearly when recrystallization twins occur and the twinning plane normal is placed near the disorientation axes between each part of the new grain and its nearest as-deformed neighbourhood (Fig. 6).

Figure 5. \{111\} pole figures showing the as-deformed (a,c) and recrystallized (b,d) grain orientations in samples of Goss (a,b) and brass (c,d) orientations for the Al-1\%Mn alloy. Recrystallization anneals: Goss – 618 K/25 s, brass – 696 K/25 s [12].

First recrystallization twins. Once a nucleus appears, it can transform through the formation of a first generation recrystallization twin. Annealing twins are frequently observed in fcc metals of medium-low SFE. Little significance was attached to annealing twins in Al and its alloys until the publication of results in Göttingen based on the application of high voltage TEM to recrystallization [21] produced evidence of extensive twin formation during the early stages of recrystallization of several fcc metals including pure Al. The orientation maps made for both metals and orientations indicate that some of the new grains undergo twinning. However, the quantity of recrystallization twins strongly depends on the analysed surface; the intensity is the highest on the free sample surface and significantly decreases as the sample centre approaches [22]. The most frequent situation occurs when the twinning plane normal is situated near the rotation axis (near of all the four \{111\} planes).

It was also observed that the normals of the twinning planes lie among the rotation axes of the disorientation relations between the recrystallized grains and their as-deformed neighbourhood. This situation is shown in Fig. 6. The area described as R1 is rotated by an angle of about 30-35°, whereas the region of R2 represents a 25-30° rotation of opposite sense around axes lying near the [11-1] pole of the deformed state. Since both twin-related regions represent nearly the same disorientations with respect to the as-deformed neighbourhood it is difficult to indicate which region was formed first. Another case of twinning occurs if the normal to the twinning plane is situated well away from the...
rotation axis, i.e. the twinning in the grain occurs on a plane other than that of the \(\langle111\rangle\) pole lying near the rotation axis, as observed earlier for medium to low SFE metals [11]. This usually leads to different disorientations of both recrystallized areas with respect to the deformed state. Nevertheless it is not clear at the moment if recrystallization twining starts when normal grain growth is stagnated or if twinned areas appear simultaneously and they grow together, or whether both these mechanisms are valid.

Figure 6. Recrystallization twinning in pure Al. (a) SEM/EBSD orientation map and (b) corresponding \{111\} pole figure showing new grain and as-deformed state orientations. (c) Misorientation axis (in sample coordinates) distribution across the recrystallization front. Misorientation angle and axis distribution (presented in crystallite coordinates) for (d) R1 and (e) R2 areas. Sample recrystallized at 618 K for 25 s.

‘Pseudo-twin’ boundaries between growing grains. The two cases described above concern the twin boundaries that represent the 'ideal twin disorientation' (~60°<111>) or the boundaries that were only slightly deviated (<2°) from this ideal relation. However, the SEM/EBSD system detects boundaries, as twin boundaries, if they satisfy the Brandon criterion [23]. It is in fact possible to find boundaries with disorientation angles and axes more deviated from the ideal position, but within the limit of the Brandon criterion. Most of these boundaries are ‘pseudo-twin boundaries’; they appear between the growing grains. Very often it was observed that if, within a given area, a new grain forms as a result of a positive ~25-35° rotation around a near <111> axis, in the neighbouring area a new grain appears as result of a negative ~25-35° rotation around an axis close to the previous one. The combination of these rotations leads to the formation of a grain boundary between impinging grains.
that fulfils Brandon’s criterion for twin boundaries. But their boundary is not coherent since their common \( \{111\} \) plane does not coincide with the boundary plane. For a precise identification of coherent twin boundaries the position of the common the \( \{111\} \) plane is needed.

A new algorithm to identify (coherent) twin boundaries was applied. In the first step the grain boundaries with the twin orientation relationship, i.e. \( 60^\circ <111> \), were found on the orientation maps using Brandon’s criterion (for the maximum deviation from the ideal axis and angle positions) [23]. In the second step the boundary direction was determined. Then for each boundary (line) with a twin orientation relationship the common \( \{111\} \) plane was identified; their normal (in fact their projection on the analysed plane) was marked by black arrows. Using this method, it is possible to reject boundaries which are clearly incoherent. But, it is important to note that (on the basis of two dimensional orientation maps) this procedure cannot absolutely guarantee a twin boundary. Ambiguous cases occur if the intersection of the common \( \{111\} \) plane and the boundary plane lies on the analysed section. Figure 7a shows a grain boundary map with marked boundaries between areas showing twin type orientation relationship (red lines) and coherent twin boundaries (yellow lines). The number of neighbouring grains showing twin-type orientation relationships was higher in Goss-oriented samples then in brass ones (Fig. 7b). Moreover, in both crystal orientations the intensity of twinning was stronger in near-surface volumes then in the sample centre.

![Figure 7](image-url)

**Figure 7.** (a) Grain boundary map with marked boundaries between areas showing a twin orientation relationship (in red) and coherent twin boundaries (in yellow), where the normal to a common \( \{111\} \) plane is perpendicular to the boundary, with the 20° tolerance of the orthogonality (b). Fraction of twin boundaries on the orientation map in relation to tolerance angle of the orthogonality grain boundary direction and the direction of projection of \( <111> \) direction. Al-1%Mn alloy recrystallized at 680 K for 70 s and at 723 K for 565 s of Goss and brass orientations, respectively.

### 3.2.2. Disorientation axes distributions.

Figures 8 show the disorientation axes distributions in the ED/TD sample coordinate system. The disorientations were calculated separately for the neighbouring pixels of the deformed/recovered areas (Fig. 8a) and across the recrystallization front, i.e. between the neighbouring pixels of new grains and the nearest deformed/recovered neighbourhood (Fig. 8b). A comparison of the preferred disorientation axes situation shows that there is a lack of correspondence between the distributions for the deformed crystal and the one identified across the recrystallization front. This leads to the important conclusion that the mechanisms responsible for crystal lattice rotations during deformation do not exactly match those controlling the formation of the recrystallized grains.
Recovered state. The disorientation axes between particular pixels of the recovered state were generally similar to those observed in the deformed state. Most of the disorientation axes coincide with $<111>$ and $<011>$ directions for the brass and Goss oriented samples, respectively. However, a discernible scattering towards some other axes (also near TD) was also observed, as discussed earlier for the deformed state.

Disorientation across the recrystallization front. For the orientation relationships of boundaries between isolated single grains of homogeneous orientation (surrounded by high-angle boundaries) and deformed/recovered regions the numbers of disorientation angles are presented in Fig. 9. The range of potential disorientation angles was binned by intervals of $1.25^\circ$ and the number of disorientations across the recrystallization front corresponding to a given angular range was counted. The most populated disorientation angles in the disorientation relation between the new grain and the deformed areas varied between $25^\circ$ and $55^\circ$ with a wide maximum between $30^\circ$ and $45^\circ$. Most of the single, isolated, nuclei are misoriented by $25$-$30^\circ$ with respect to the deformed state, so that the special role of near-$40^\circ$ disorientations is not confirmed, in contradiction to much previous literature, e.g. [1, 2, 24, 25].

Figure 8. Interrelation between the distribution of misorientation axes at early stages of recrystallization observed for the Al-1%Mn alloy with the brass $(110)[1-1-2]$ orientation. (a) The distribution of misorientation axes for the deformed/recovered state between particular pixels of the orientation map and (b) the misorientation axes between deformed/recovered and recrystallized areas. Sample recrystallized at 680 K for 25 s.

It is evident that the formation of new grain orientations, which is not random, leads to some preferences in the distribution of the disorientation axes between the deformed/recovered and recrystallized areas. For both the Goss and brass orientations (Fig. 9), a definite majority of disorientations axes are located around the poles of the $(111)$ and $(1-1-1)$ planes, i.e. those of the active slip systems during deformation. Nevertheless, the misorientation axes do not coincide exactly with the poles of these planes. Distributions of misorientation axes show that only a relatively small fraction of the disorientation axes coincide with poles of the $(111)$ planes (Fig. 10). This means that the orientation relationship, usually described in the literature by $a<111>$ relations, is in fact very rarely observed. The most populated disorientation axes were usually shifted by about $20^\circ$-$25^\circ$ towards the $<221>$, $<012>$, $<123>$ or $<112>$ directions, lying close to selected normals of the $(111)$ slip planes. This shift was significantly greater than the orientation spread typically observed inside the deformed state (usually below $10^\circ$). It should be mentioned that in some rare cases, disorientation axes were found quite distant from the $(111)$ plane poles of the active systems (including the vicinity of the two remaining $<111>$ poles).
3.2.3. **The directionality of grain growth. Analysis in the ND/ED section.** The orientation maps measured on the ND-ED plane for both metals and orientations show that a considerable number of recrystallized grains are of elongated shape. The straight high angle grain boundaries were observed along traces of \{111\} planes. These grains exhibit a well-defined lattice rotation with respect to the deformed/recovered areas. An example of the crystal lattice rotation, related to the appearance of heavily elongated grains, was carried out on Al-1%Mn alloy single crystals of Goss\{110\}<001> orientation (Fig. 11). To a first approximation, this aspect of nucleation and grain growth leads to the following conclusions:

- The longest grain direction runs mostly along the traces of (111) and (11-1) planes, i.e. the most active planes during strain, and which are symmetrically inclined to the ND-TD plane.
- The orientations of the elongated grains are related to those of the group of the deformed crystals. The crystal lattice of each of the grain groups is rotated (positively or negatively) around diverse groups of disorientation axes.
- The rotation axes are situated near the poles of (111) and (11-1) planes. This may suggest a mechanism for the formation of a new grain by thermally activation movement of dislocation families, 'travelling' in pairs on one of the above-mentioned planes, as proposed in [11]. Although the rotation axes approach the normal vector of the preferred slip planes they only rarely coincide with the exact location of the <111> directions.

**Figure 9.** Misorientation relation at early stages of recrystallization between recrystallized area and nearest as-deformed neighbourhood. Distribution of misorientation angles (plots uses 1.25° bins) and axes for (a, b) Al and (c, d) Al-1%Mn with (a, c) Goss) and (b, d) brass orientations. Red points marked misorientation relation for twinned areas. Recrystallization anneals: (a) - 618 K/25 s, (b) – 696 K/25 s, (c) - 686 K/25 s and (d) - 696 K/30 s.
• The crystal lattice rotation of elongated grains takes place around axes near the <111> pole, corresponding to the {111} plane across which preferred grain growth occurs. This is shown by stereographic projections of the rotation axis distributions between deformed (recovered) and recrystallized regions. The crystal lattice of the grains elongated along a (111) plane undergoes rotation around axes near the [111] direction. By analogy, the crystal lattice of the grains elongated along the (11-1) plane undergoes a rotation around the near [11-1] direction.

3.2.4. ‘Tilt’ or ‘twist’ grain boundaries. Analysis on three perpendicular sections. The anisotropic growth of new grains in a relatively homogeneous structure of the deformed state, as observed in the middle of the longitudinal section of Goss{110}<001> oriented samples, enables one to elucidate the mechanisms responsible for rapid grain growth. However, it is crucial to determine the grain boundary plane with respect to the rotation axis vector [1]. This opens up the possibility of determining the ‘twist’ or ‘tilt’ character of the grain boundary. But the tilt or twist nature of a grain boundary formed between new grains and the as-deformed state can only be precisely defined by SEM/EBSD measurements in three perpendicular sections.

Figure 10. The scattering of misorientation axes around the [111] pole for Al-1%Mn of brass orientation. Detail from Fig. 9d.

The orientation maps in the ND/TD and ED/TD sections confirm that the majority of grains are characterized by a strongly elongated shape. The longer axes of grains run horizontally or vertically, along the traces of the {111} planes in both sections. In the case of the ND/TD section the traces of the highly active slip planes are situated horizontally, whereas the less active are vertical (Fig. 12a). In the case of the ED-TD section the traces of the highly active slip planes are situated vertically, whereas the less active are inclined at ~35° to ED (Fig. 12b). In all sections the elongated grains of both families are growing from the surface into the deformed structure. It is clear that the traces of elongated grain boundaries are not random and coincide with traces of the {111} planes. Therefore, it is concluded that the majority of flat grains are situated along {111} planes (Fig. 12c).

Based on the results of 3D analysis it is concluded that the long straight boundaries perpendicular to the <111> axis (which is also the rotation axis in the disorientation relation between as-deformed state and new grains) were twist boundaries. This idea, based on the coordinated movement of dislocations at the head of the recrystallization front leading to twist boundary formation with a disorientation axis parallel to the [111] direction, is linked to the explanation of Ridha and Hutchinson.
[26] for the rotations around <111> axes. Consequently, the boundaries at the head of the growing grain were tilt boundaries. Since grain growth is due to rapid motion of the tilt boundary at the head of a growing grain this description is consistent with the results of Kohara et al. [27] and Parthasarathi and Beck [28] who observed that tilt boundaries move faster than twist boundaries in fcc metals.

3.3. The influence of the free surface on the morphology and crystallography of new grains

The fraction of recrystallization strongly depends on the sample section location. It is clear that for the applied recrystallization anneal the surface is almost completely covered by new grains, whereas in the ¼ and ½ sample sections the density of new grains significantly decreases. The shape of the recrystallized grains on the surface is nearly equiaxed and the average grain size is smaller than that observed on the other two sections. Moreover, the large grains in the ¼ and ½ sample sections reveal a strong tendency to grow within volumes of slightly larger lattice disturbances.

![Figure 11](image_url)

**Figure 11.** ‘Directionality’ of grain growth at the early stages of recrystallization observed in Al single crystal of Goss orientation. (a) Orientation map. (b) The distribution of the misorientation axes between deformed/recovered and recrystallized areas. (c) and (d) {111} pole figures showing the orientations of recrystallized grains growing along traces of the {111} planes marked as red and blue, respectively.

The observed positive or negative ~30-35° rotations of a new grain orientation around axes lying close to the normal of the most active {111} planes lead to four variants of near S[123]<634> components. Irrespective of the surface the pole figures of new grains revealed a quite symmetrical
character. Since nearly the same groups of new grain orientations were detected on each section, (Fig. 13), this led to the same type of disorientation with respect to the components of the as-deformed state. The positive and negative rotations around axes situated near the normals of the most active \{111\} planes are able to explain the strong cube \{100\}<001> texture component formation during recrystallization of many Al alloys. The \{111\} poles of four variants of S orientation ‘marked’ the position of poles characteristic of the cube orientation. (Four variants of S orientation are formed as a result of positive and negative \( \pm 30^\circ \) rotations of the Goss orientation around axes near the \{112\} direction). The orientations maps do not show cube-oriented grains (in the 20° scattering from ideal position) or more precisely they are detected very rarely. However, a strong cube texture is observed on pole figures especially those presented in ‘continuous’ form. This phenomenon is discussed in more detail in another work [29].

**Figure 12.** Directionality of grain growth observed on (a) ND/TD and (b) ED/TD planes. Dashed lines marked the directions of the privileged growth of new grains; they corresponds to the position of traces of the \{111\} planes. (c) Schematic presentation of the situation of the \{111\} planes.
4. Conclusions
Based on local EBSD orientation measurement the orientation relations which appear at the initial stages of recrystallization between a new grain and the as-deformed areas of stable single crystals have been analyzed. The results allows us to draw the following conclusions:

- The orientations of new grains during primary recrystallization are not random. For Al and Al-1%Mn and both Goss and Brass orientations at the early stages of recrystallization, the appearance of a limited number of new grain orientation groups was demonstrated. The orientation relation across the recrystallization front defines the final rotation by angles in the range 25-55° around axes mostly grouped about the <122>, <012>, <112> and <111> directions. For a single isolated nucleus of uniform orientation, the most preferred rotation angles were near 25-30°, whereas the rotation axes are usually grouped around one of the normal of all four {111} planes but only rarely coincide with them.

- The growth direction of new grain indicates the privileged role of the {111} planes in the initial stages of recrystallization. In most cases, the new grains were strongly elongated with planar facets close to those of dislocation slip planes.

- The orientation of the growing new grain in some cases transforms through the formation of a first generation twin. The most frequent situation occurs when the normal of the twinning face plane is situated near the rotation axis, around which the crystal lattice of the ‘primary nuclei’ rotates.

- A large fraction of new grains, regardless of the metal, develop a strongly elongated shape. For both initial orientations, preferred grain growth occurred mostly along {111}-type planes. However, the most active planes during strain, located symmetrically to the ND-TD plane, play a decisive role in this process. The orientations of the elongated grains were directly related to those of the group of as-deformed orientations. The crystal lattice of the grains elongated along a {111} plane undergoes a general rotation around an axis near to their normal.

- It is proposed that the thermally activated, coordinated movement of dislocation groups can lead to the creation of a twist or tilt grain boundary according to their screw or edge character.

Figure 13. The texture image observed in different sections of the Goss-oriented single crystal of Al-1%Mn alloy recrystallized at 696 K for 70 s.

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