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# Recrystallization Behavior of Commercial Purity Aluminium Alloys

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Additional information is available at the end of the chapter

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## 1. Introduction

Commercial purity aluminium alloys are largely used in the forms of foil for food packaging industries. Aluminium ingots are processed through the multi-operational steps in the consequence of foil preparation. The ingot of 7mm thickness, obtained from either direct chill (DC) casting or continuous casting process, is first cold rolled to reduce the thickness approximately 0.6 mm followed by an intermediate annealing treatment and then final roll pass is given to produce the foils of desired thickness. Therefore, the intermediate annealing treatment and consequent recrystallization behavior are prime factors for controlling microstructures as well as properties of foils. In general, the recrystallization behavior is the resultant of three steps, viz., deformation, recrystallization and grain growth. It is affected by various impurities and the precipitates forming at the operational steps.

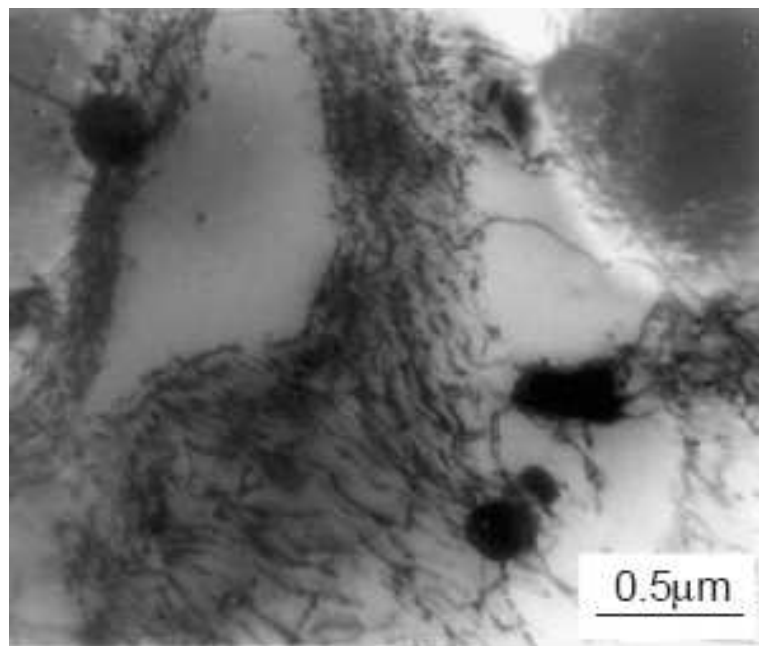
Iron and silicon are the inevitable impurities in commercial purity aluminium alloys. The precipitation reaction due to these impurities easily occurs during heat treatment at the temperature range of 200 – 600°C, affecting cold working, softening, and corrosion resistance of the alloys. The type of intermetallic phases formed during solidification and volume fraction, amount and size of the individual particles can be controlled by changing the silicon (Si) and iron (Fe) content, i.e., by adjusting the Fe/Si ratio [1]. The silicon influences the nature of eutectic constituent, solid solubility of other elements, formation of precipitates and dispersoids, and the transformation characteristics of the precipitates [2]. It is necessary to control the Si content of high quality aluminium foils where material characteristics are determined by Fe/Si ratio. If the material is properly processed, then it can only be assumed that the material properties are not actually deteriorated with increased amount of Fe, Si and/or Mn. In general, composition ratio of the various alloying elements is maintained with respect

to each other, but the absolute value of the single alloying element is not critical as long as primary crystallization of coarse intermetallic phases is avoided.

## 2. Plastic deformation

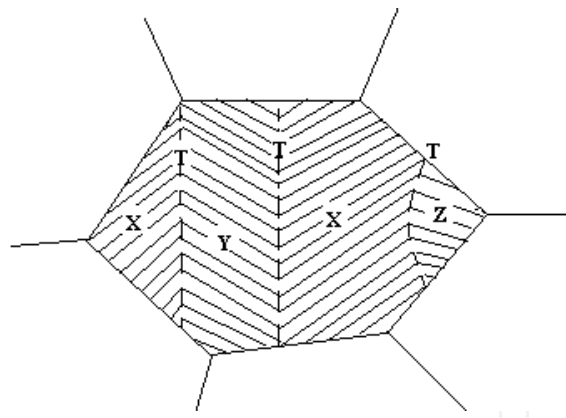
### 2.1. Microstructure of deformed metals

The cold-rolled alloys are consisting high density of dislocations clearly examined through transmission electron microscopy (TEM). During deformation the dislocations are arranged as cell boundaries whereas some cell interiors are free of dislocations (Fig. 1). Due to heavy deformation, the cell boundaries are diffuse in nature instead of simple dislocation.



**Figure 1.** Dislocation structures of 92% cold rolled commercial purity aluminium alloy (AA1235 :0.67 Fe, 0.16Si, 0.01 Ti and rest Al) [3]

However, the orientation of plastically deformed alloy changes from grain to grain or in different regions in a same grain. It happens owing to different rotations during the deformation by use of a different combination of slip systems to achieve the imposed strain. Therefore, the deformation band is nothing but a volume of constant orientation that is significantly different than the orientation of other parts of the grain, which is explained by a schematic diagram (Fig. 2). It shows the variation of orientation at different regions of a same grain. The orientation of one part (X) of a deformed grain changes rapidly to that of a differently oriented part (Y) of the same grain across a thin boundary of a finite width called the transition bands (T) or microbands. Sometimes the orientation of deformation bands is changed twice, such as from X to Y and then Y to X. This special type of deformation band is called kink band.



**Figure 2.** Schematic diagram of deformation bands, transition bands and kink bands [4]

After plastic deformation by slip mechanism, the walls of high dislocation density separate the region of low dislocation density. Such types of microstructure have been referred to as cell or subgrain. In the case of cell structures, the boundary consists of a tangled array of dislocations and appears to be a diffuse in nature. For subgrain, the boundary is sharp and consists of a well ordered dislocation array. The cell structure develops in the bulk material during or after the dislocation movement induced by the applied stress. With increasing strain, the walls become quite sharp and interiors become reasonably free of dislocations. At this stage the cell might be called subgrain.

## 2.2. The effect of particles on deformed microstructure

Due to rapid cooling rate in the continuous casting and direct chill casting, the alloying elements remain as supersaturated solid solution. Upon the processing operation of aluminium foils, the increased cold rolling reduction gives rise to a more pronounced dislocation entanglement. It results in the acceleration of both stored energy and precipitation. Therefore, during plastic deformation of two-phase alloys, particles affect the overall dislocation density, inhomogeneity of deformation in the matrix and deformation structure. Subsequently, the recrystallization behaviour of the alloy is also affected due to the influence of driving force as well as nucleation sites for recrystallization. If the particle is strong enough to withstand the applied stress on it, the dislocation then proceeds to encircle the particle and leaves an Orowan loop, otherwise it deforms. When the particle does not deform then extra dislocations are generated at the particle-matrix interface. On the other hand, if the particle deforms either before or after the formation of Orowan loop, no extra dislocations are generated. Generally, smaller particles are weaker than larger particles, which results in weaker slip plane. Movement of subsequent dislocations occurs on the same plane. As a consequence, slip bands are formed.

## 2.3. Deformation texture

The cold rolling texture of aluminium alloy sheets has been characterized as  $\beta$  fiber, which is associated with plane strain deformation [5]. The texture is varied from layer to layer of the

sheet. The surface layer of the rolled sheet is known to have the shear texture, which is different from that of the plane strain deformed center layer [6-10]. According to Grewen and Huber, the ideal orientations in 95% cold rolled aluminium are characterized by  $\{112\}<111>$ ,  $\{110\}<112>$  and  $\{123\}<412>$  [11]. The penetration depth of shear texture increased with increasing friction coefficients and rolling temperature [7]. Above all, material parameters also influence the shear texture formation. The most important factors among material parameters are the yield strength and strain hardening exponent [8,9]. The texture inhomogeneity has often been of considerable importance in rolling of pure aluminium, which has a low yield strength and work hardening exponent and tends to develop the shear texture in the surface layer [12-14]. Aluminium and copper single crystals of the S-orientations develop cube texture after cold-rolling by more than 97.5% and subsequent annealing [15, 16]. This occurs owing to the close correlation between the formation of  $\{001\}<100>$  deformation structure and the development of cube texture. The cube-oriented material formed during rolling is considered to be the preferential nucleation sites of cubically aligned recrystallized grains. The factors which affect the generation of the  $\{001\}<100>$  deformation structure and the equivalent components during rolling are process parameters like friction between roll and sheet, rolling temperature, lubricants and the  $L/d$  value (where  $L$  and  $d$  represent the contact length and the thickness of specimens, respectively). Low viscosity lubricant such as kerosene suppresses the formation of  $\{001\}<100>$  deformation structure, whereas using machine oil as lubricant a cube component is formed. Thereafter, subsequent annealing treatment develops retained rolling and random texture in kerosene treated alloy and sharper cube texture in machine oil treated alloy [17]. Truszkowski et al. [9, 10] and Asbeck and Mecking [8] have reported that the shear texture is developed when  $L/d \geq 5$  or  $L/d \leq 0.5$ .

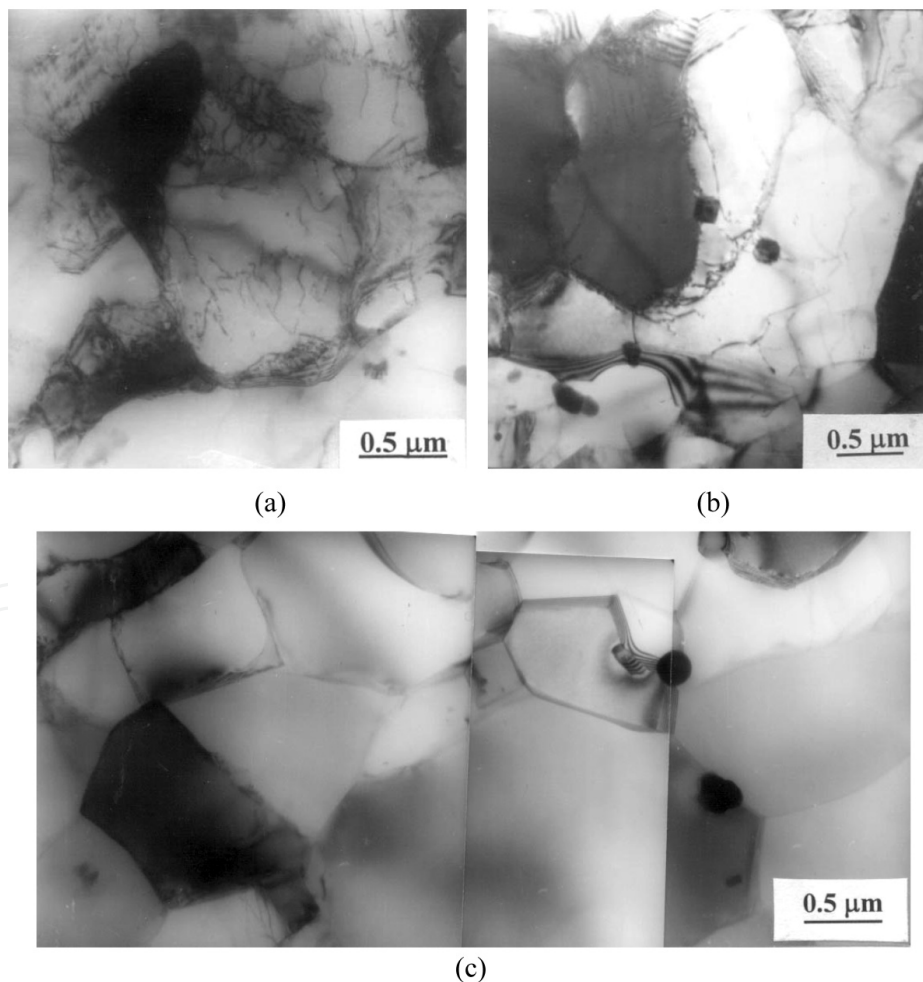
### 3. Recovery

#### 3.1. Subgrain formation and its growth

During plastic deformation of polycrystalline material (specially the alloy of medium and high stacking fault energy), unequal numbers of dislocations of two signs are generated, and the dislocations are rearranged as three-dimensional cell structure with complex dislocation tangles as cell walls. At recovery stage, the dislocations of opposite sign annihilate each other by combination of gliding and climbing mechanisms. The excess dislocations are left in the material at the end of the first stage of recovery. Upon progressing stages, these excess dislocations are arranged in a low energy configuration in the form of regular arrays or low angle grain boundaries (LAGBs). This mechanism is called polygonization [18], and the newly formed cells are called subgrains. In other words, the dislocation movement from the cell interiors to the cell boundary causes low angle grain boundaries, resulting in the subgrain formation. Sometimes, the dynamic recovery also helps to form a well-developed subgrain structure during deformation. In this case, recovery is the only involvement for the coarsening of subgrain structure. Although low stacking fault energy material shows poor subgrain structure, but when fine second-phase particles inhibit recrystallization then at high temper-

ature a well-defined substructure is formed by recovery [19]. Coarsening of substructure takes place by two methods-subgrain boundary migration and subgrain rotation.

The cell structure of a heavily cold rolled alloy is modified as a function of annealing time, examining the subgrain growth. In a commercial purity aluminium alloy (AA1235), after annealing at 250°C for 1 h, the dislocations are examined to rearrange themselves to form a subgrain structure with an average diameter of 0.60  $\mu\text{m}$ . Dislocation networks form the low angle boundary also known as subboundary. The dislocation rearrangement in the regions where the cell structure is not well-developed is described as a “disentanglement of dislocations” (Fig. 3a). Consequently, the dislocation density decreases inside the subgrain and particles are precipitated at subboundary as well as inside of subgrains (Fig. 3b). After 4 h of annealing at same temperature homogeneous subgrain growth takes place with an average diameter of 2  $\mu\text{m}$ . Generally, the sharp delineation of many subboundaries decreases upon annealing and ultimately fades away, in agreement with the proposed coalescence mechanism (Figs. 3c). At this stage, subgrain growth is hindered by the pinning action of  $\text{Al}_3\text{Fe}$  precipitates.

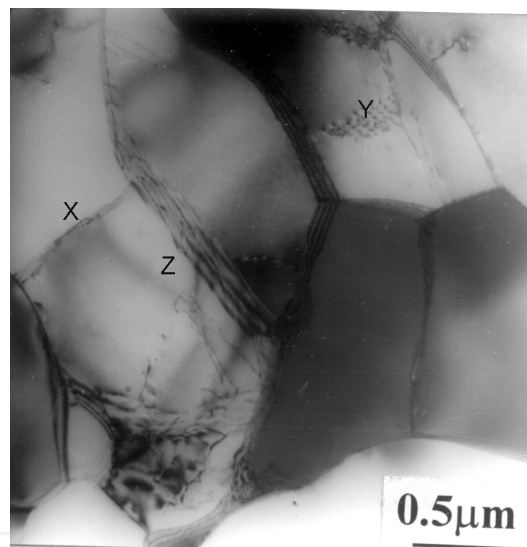


**Figure 3.** TEM micrographs (BF) showing subgrain formation of 92% cold rolled alloy AA1235 annealed at 250°C for (a) & (b) 1h and (c) 4h.



### 3.2. The effect of second phase particles

In two phase alloys, second phase particles inhibit recovery by the pinning of individual dislocation or dislocations of the low angle boundaries. Therefore, the dispersion of fine particles exerts a strong pinning effect on the subgrains [20-22]. Moreover, the particles stabled at high temperature improve the strength and creep properties of the alloy by pinning and stabilizing the recovered substructure. If the particles prevent the subgrain to reach the critical size for the formation of recrystallization nucleus, then recrystallization is also hindered. The evidence of dislocation pinning and restriction of subgrain growth is examined for a commercial purity aluminium alloy (AA8011) after annealing below 400°C [23]. Similarly, the well-defined subgrains are examined in another commercial purity aluminium alloy (AA1235) after 4h annealing at 250°C [24], as shown in Fig.4. The dislocation density change is clearly examined at some portions (X) of boundaries. The distribution of dislocation is also another observation at some points (Y). At this stage, dislocations accumulate at the grain boundary region and sometime they result in the formation of grain boundary lines at Z [25]. It can be said that the effect of particles on subgrain growth is similar to the effect of particles on grain growth.



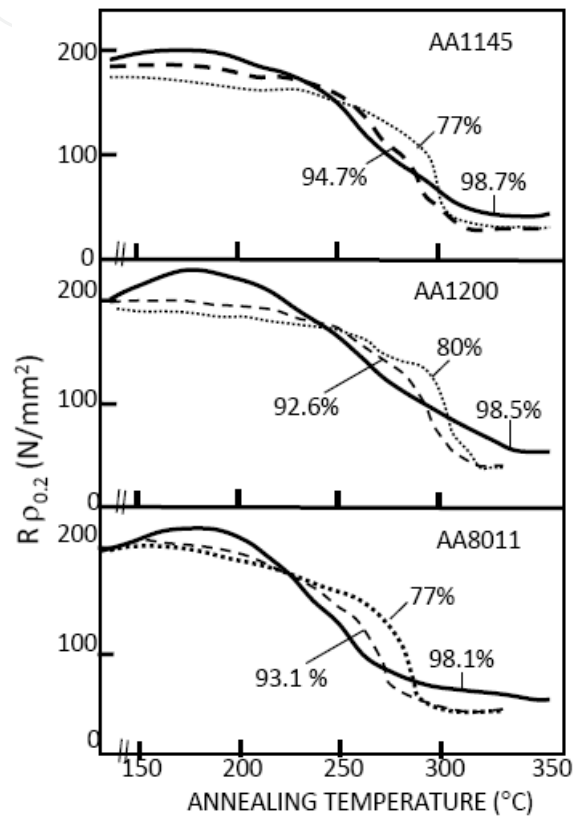
**Figure 4.** TEM micrographs (BF) showing the effect of annealing at 250°C for 4 h on subgrain structure of 92% cold rolled bulk specimen (AA1235).

## 4. Recrystallization

### 4.1. The effect of deformation percentage

Nucleation and growth of new recrystallized grains are closely dependent on the distribution of dislocations within the deformed grains, relating to the deformation percentages. Recrystallization does not happen below a minimum strain and the start temperature of recrystalli-

zation decreases with increasing the minimum strain level. It is known that at low strain (<5%) high activation energy for nucleation results in a low nucleation rate and a relatively high growth rate [26]. With varying cold rolling percentage for different commercial purity aluminium alloys (AA1145, AA1200, AA8011), the degree of softening is identical for low reduction levels (<95%) and less steep for reduction levels above 95% (Fig.5) [27]. It corresponds to the gradual coarsening of microstructure with increasing temperature for high reduction levels (>95%), relating to the continuous recrystallization.



**Figure 5.** Yield stress variation with annealing temperature for three commercial purity aluminium alloys at different percentage of cold rolling [27]

#### 4.2. Nucleation of recrystallization

The recrystallization nuclei originate at preferred sites such as prior grain boundaries, transition bands, and shear bands of deformed microstructure. The nucleation process at the grain boundary region by two different mechanisms:

1. **Strain induced boundary migration**, where a nucleus forms when an original grain boundary bulges out [28].
2. **Subgrain coarsening by coalescence at the original grain boundary**, where a nucleus forms in the consequence of subgrain growth [29-31].



With an increasing degree of deformation, the number of potential nucleation sites such as deformation bands and grain boundary bands increases significantly and the number of such sites is larger in fine grain than in coarse grain specimens [32]. The effectiveness of nucleation sites is enhanced by the presence of intermetallic particles ( $\text{Al}_3\text{Fe}$ ) with increasing degree of deformation. Potential nucleation sites are more important factor than number of nuclei. If the same numbers of nuclei form at the grain boundaries of both fine grain material and coarse grain material, then the fine grain material shows more homogeneous recrystallization. In case of heterogeneous recrystallization, all grains are not recrystallized at the same rate. It is known that crystallographic orientation affects the slip systems and strain path during deformation. Therefore, distribution and density of dislocations, large scale microstructural inhomogeneities, availability of nucleation sites, and growth rate of recrystallized grains are also dependent on crystallographic orientation.

#### 4.3. Sequence of precipitation and recrystallization

It is well established that the recrystallization behavior of a deformed and supersaturated alloy is largely dependent on whether or not precipitation of the second phase particles take place simultaneously with recrystallization [33]. Hornbogen and Köster have suggested that recrystallization occurs prior to precipitation at high temperatures, whereas precipitation takes place prior to recrystallization at low temperatures [34]. If particles are precipitated during recrystallization they may hinder both the formation and migration of recrystallization fronts. Alternatively, if the second phase particles are precipitated in the matrix prior to cold rolling, the recrystallization behaviour will depend on the size and dispersion of the second phase particles.

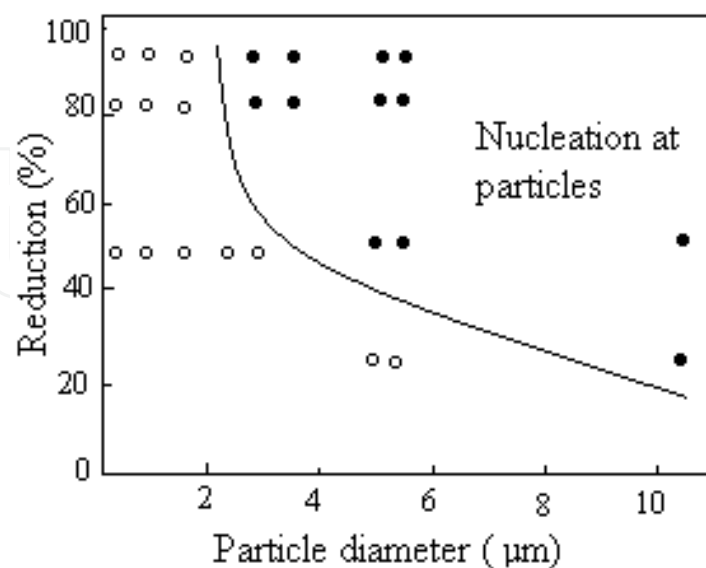
#### 4.4. The effect of inter particle spacing

At wide inter particle spacing, when only a few particles are present, there is basically no difference between the recrystallization behaviour of two-phase and single phase alloys. In this case, nucleation generally occurs at the original matrix boundaries and the ultimate recrystallized grain size may vary owing to inhomogeneous distribution of nucleation sites. With an increase in particle content of the alloy (i.e., decrease in interparticle spacing) nucleation occurs more rapidly at the lattice curvature of particle-matrix (for the particles  $>1\text{ }\mu\text{m}$ ) interface than at the grain boundary region [35]. In addition, due to the increase in quantity and uniform distribution of particle-matrix nucleation sites, the final recrystallized grains become more uniform and finer. However, the trend towards increased nucleation due to decrease in interparticle spacing occurs until particles maintain a critical spacing ( $C_1$ ) to allow the nucleation to occur simultaneously and independently at each particle. After reaching the critical spacing ( $C_1$ ), a nucleus attached to one particle would be viable itself, if it left to develop by itself. It would be non-viable if nuclei start to form simultaneously at neighbouring particles. They would then interfere with each other before reaching a viable size. Thus, due to the formation of fewer amount of viable nuclei at particle-matrix interfaces during increase in particle content (i.e., decrease in interparticle spacing) the overall nucleation rate of recrystallization decreases. Further increase in particle content leads to a second critical spacing ( $C_2$ ) at

which viable nuclei formation is inhibited owing to the proximity of particles. In such circumstances nucleation is likely to occur predominantly at the original matrix grain boundaries. As a consequence, the nucleation as well as recrystallization rate drastically get reduced if the inter particle spacing falls below  $C_2$  [36].

#### 4.5. The effect of particle size

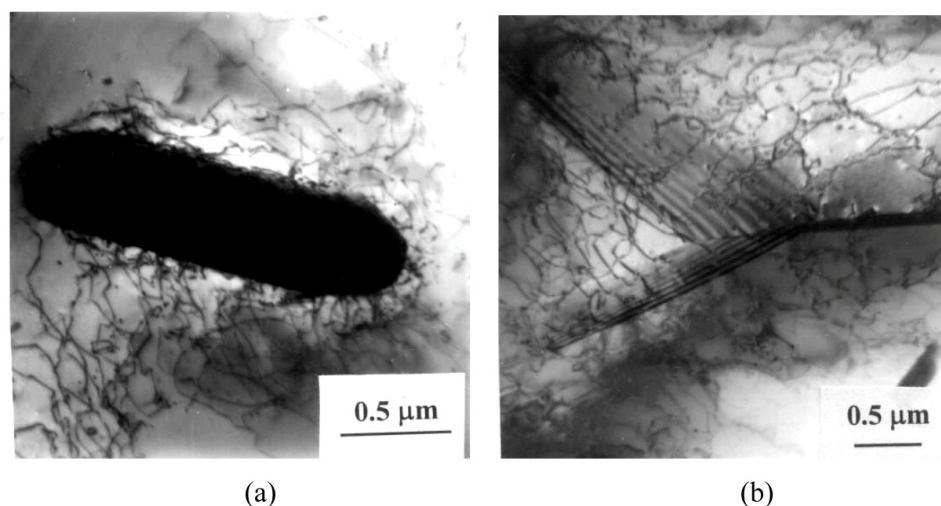
At the time of deformation, large particles ( $> 1 \mu\text{m}$ ) lead to a heterogeneous distribution of dislocations, whereas fine particles ( $< 0.1 \mu\text{m}$ ) give rise to a homogeneous distribution of dislocations. As a consequence, the number of possible recrystallization nucleation sites increases for large particles and decreases for fine particles. This can be related to the degree of deformation (Fig. 6), explaining the nucleation of recrystallized grains at the particles greater than  $2 \mu\text{m}$  for a highly deformed metal [37]. The fine particles also inhibit sub-boundary migration and thus the nucleation process is retarded. As the particles become more finely spaced, nearly all cell boundaries will be pinned by the particles at the end of the deformation stage. In this case, when recrystallization occurs, the mechanism of nucleation is not well understood, but the kinetics of the process is certainly very slow [38]. Closely spaced, thermally stable particles preserve the deformed/ recovered microstructure up to the melting point of the matrix. Kim et al. have revealed that the retention of dislocation substructure at high temperature provides an additional strengthening mechanism to the dispersion hardening of the alloys used at high temperature structural applications [39]. With the decrease in the particle size and spacing between particles, recrystallization kinetics is retarded, but final grain size becomes quite large [37, 40]. Several researchers have studied the relationship between dispersion characteristics, deformation substructure, and recrystallization [41-43].



**Figure 6.** The conditions of deformation and particle size for which nucleation of recrystallized grains is observed to occur at particles [37].

#### 4.6. Type of particles in commercial purity aluminium

In commercial purity aluminium alloys, the main impurity Fe combines with both Al and Si to form a large variety of phases during solidification or during subsequent thermomechanical processing [44, 45]. Shoji and Fujikura have identified three types of precipitates Si,  $\alpha$ -AlFeSi, and  $\text{Al}_3\text{Fe}$  in cold rolled commercial alloy AlFe0.6Si0.16 (all in wt%) after annealing in salt bath at different temperatures 300, 400 and 500°C, respectively [3, 46]. Some researchers have identified only  $\text{Al}_3\text{Fe}$  precipitates during recrystallization of commercial Al-Fe-Si alloy [47]. In AA8011 alloy,  $\text{Al}_3\text{Fe}$  and Si precipitates are observed in cold rolled and annealed conditions [48]. After heavy plastic deformation (~92%), very small particles of 0.17  $\mu\text{m}$  size are distributed in the dislocations [3]. Upon progress of annealing, the particle size increases, and both smaller and larger, spherical and plate-shaped  $\text{Al}_3\text{Fe}$  particles are inhomogeneously distributed and situated at subboundaries and subgrain interiors. Since the dislocations cannot cross the grain boundary, during annealing a large variation of dislocation density occurs from grain to grain owing to the pinning effect of precipitate particles. After completion of recrystallization, particle also inhibits the grain boundary migration along with pinning of the dislocations. Even after a long time of high temperature annealing for heavily deformed alloy, the  $\text{Al}_3\text{Fe}$  particles may pin down the dislocations, resulting in the presence of a large amount dislocations inside the grains (Fig. 7) [3, 24]. During the intersection of grain boundary by particles, a Zener drag force is generated to restrain the boundary migration [49]. The coherent particle generally loses the coherency when a high angle grain boundary moves past a coherent particle. The coherent particles are twice as effective in pinning a grain boundary as incoherent particles of the same size [4]. There are many alternative situations during particle-boundary interaction. The particle may dissolve during passage of the boundary and re-precipitate in a coherent orientation, it may reorient itself to a coherent orientation, or the boundary may cut through the particle [50].



**Figure 7.** TEM micrographs (BF) of 92% cold rolled alloy AA1235 after annealing at 480°C for 8 h showing (a) pinning of dislocations by a particle, and (b) presence of dislocations even after completion of recrystallization

#### 4.7. Particle distributions in bimodal alloys

Many commercial alloys are bimodal alloys which contain both large ( $>1 \mu\text{m}$ ) and small particles. Large particles act as nucleation sites for recrystallization and small particles hinder the grain boundary migration, i.e., retard the recrystallization. Therefore, recrystallization behaviour of bimodal alloys is affected by particle distribution. During deformation, the fine dispersion of small particles does not affect significantly the degree of lattice curvature generated at the large particle-matrix interface. Later, by altering the dispersion parameters of particles, recrystallization kinetics and microstructures are controlled. Chan and Humphreys have reported that in the bimodal alloy, nucleation of recrystallization takes place at the large particles, but fine particles determine the time for completion of recrystallization [51]. They have observed a large number of small island grains and coarse irregular grains in the microstructure of bimodal alloys. With increasing coarsening of the fine precipitates, the grain shapes become more regular and the mean size decreases. Nes has taken the help of the parameter  $f/r$  (where  $f$  and  $r$  are volume fraction and radius of small particles, respectively) to describe a model to account for the grain size of bimodal alloys [52]. In bimodal alloys, critical particle size for the growth of a nucleus is

$$d = \frac{4\gamma}{P_d - P_z} = \frac{4\gamma}{\frac{\rho G b^2}{2} - \frac{3f\gamma}{2r}} \quad (1)$$

where  $P_d$  is driving force (stored energy in the subgrain boundaries),

$\rho$  is dislocation density,

$G$  is shear modulus,

$b$  is burgers vector,

$P_z$  is Zener pinning force,

$\gamma$  is grain boundary energy,

$f$  is volume fraction of small particles,

$r$  is radius of small particles

As  $f/r$  of fine dispersion increases, the large particles act as the nucleation sites, but growth of the grains is slowed down at the early stage of annealing. However, the earlier developed grains either consume the smaller grains, or form island grains.

#### 4.8. Recrystallization texture

Annealing of heavily deformed material leads to a wide range of recrystallization textures. Different types of recrystallization textures can be produced in similar alloys each having almost identical and very strong deformation textures [11]. In f.c.c. metals, sometimes recrystallization texture contains cube texture along with retained rolling texture and in some cases

nearly random textures [11]. It is reported that with increasing deformation, the volume fraction of the cube texture decreases preferably for low annealing temperatures and that of the rolling texture increases for all temperatures, whereas the random part of the texture strongly decreases [53]. The possible origin of recrystallization texture has been reviewed several times [54, 55]. It is notable that “oriented nucleation” and “oriented growth” are two possible theories for the development of recrystallization texture [56, 57]. The recrystallization texture develops from the competition between cube oriented grains which nucleate at the cube bands, R-oriented grains (i.e., retained rolling texture) which stem from the grain boundaries between the former deformed grains, and randomly oriented grains due to particle stimulated nucleation (PSN).

4.8.1. The effect of particles

It is already discussed in previous section (§ 4.4) that widely spaced coarse particles (>1 μm) enhance nucleation and the rate of recrystallization if it is present before deformation. The finely dispersed particles normally reduce recrystallization kinetics having a greater retarding influence on nucleation than on growth. Therefore, the texture changes can be rationalized on the basis of the orientation dependent nucleation. If nucleation of all components is retarded in equal proportions, then the time available for growth of the first formed grains is increased. The recrystallization texture will therefore show increased selectivity and will become even more strongly biased towards the most favoured orientation nuclei. The effect of particle size on the formation of major components of recrystallization texture is shown in Table 1 [58].

Material (% of reduction)	Particles		Recrystallization Texture
	Volume fraction (f)	Diameter (μm)	
Al (99.9965) (90%)	0	-	Cube
Fe-AlN (70%)	0.06	0.017	Rolling
Al-Al <sub>2</sub> O <sub>3</sub> (30-90%)	0.4	0.1	Deformation
Cu-SiO <sub>2</sub> (70%)	0.5	0.23	Rolling + Twins
Al-FeSi (90%)	0.5	0.2- 7	Cube, Rolling, Random
Al-Si (90%)	0.8	2	Cube, Rolling, Random
Al- Ni (80%)	10.0	1	Rolling, Random
Al- SiC (70%)	20.0	10	Random

**Table 1.** Major components of the recrystallization texture for some particle containing alloys [58]

The effect of PSN on recrystallization texture is as follows [59]:

1. PSN results in a sharp recrystallization texture for low strained material or in case of particle containing single crystals. When PSN originates in several deformation zones at



a single particle of a lightly deformed polycrystals, a spread of orientation is observed around the particles in the deformed matrix.

2. In the case of heavily deformed material, the presence of different deformation zones of different grains or different deformation regions of the same grain results in a single grain with a wide range of orientations. As a consequence, either a weak texture or randomly oriented grains form.

The recrystallization texture is dependent on the particle size, strength and spacing. When the particles precipitate during annealing, the recrystallization texture developed is similar to the deformation texture. There is no clear reason for the retention of the rolling texture in the recrystallized alloys. It appears that a strong cube texture is rarely formed and particles may be responsible for suppressing the formation or viability of cube sites. This enables other components including the retained rolling components to dominate the texture.

Although PSN nuclei are oriented randomly in the heavily rolled polycrystal, the final texture of this alloy is not random. Therefore, it is concluded that grains from other sites with different orientation affect the final texture. From Table 2.2 it is observed that for the alloys (e.g., Al-Fe-Si or Al-Si) containing low volume fraction of particles, cube and rolling components are developed, whereas rolling and random components develop for high volume fraction of particles containing alloys (e.g., Al-Ni). The cube grains are larger than randomly oriented grains owing to the faster growth rate of former [59]. It results in formation of small island grains inside the large cube grains. Some researchers have shown that the strength of the cube component decreases with an increase in the number of supercritical sized particles in the case of hot-rolled aluminium alloy AA3004 [60]. On the other hand, volume fraction of randomly oriented grains increases with an increase in the number of supercritical sized particles. Hornbogen et al. have distinguished between 'discontinuous' (conventional recrystallization) and 'continuous' recrystallization [61]. The former cases give rise to normal recrystallization textures, such as the cube texture, while the latter cause retention of the rolling texture.

#### 4.8.2. Effect of Fe on the commercial purity aluminium

A typical recrystallization texture of highest purity aluminium is the cube-orientation {001}<100> [5, 11]. In commercial purity aluminium the iron and silicon contents have been found to be important factors in controlling the recrystallization texture [62, 63]. It is reported that a small amount of Fe may cause a change almost from pure cube to retained rolling texture [4]. During recrystallization, precipitation of  $\text{Al}_3\text{Fe}$  particles inhibits the growth of the early formed cube nuclei, thereby forcing nucleation in the abundant rolling texture components. Thus, the recrystallization textures of most commercial aluminium alloys are composed of three texture types, namely the cube texture, the retained rolling texture and the random texture [64, 65]. Cube-oriented subgrains are known to rapidly recover either dynamically during deformation or statically during the early stage of recrystallization, which gives rise to a size advantage of cube nuclei [66]. Growth of the dominant cube grains is lowered by solute drag and/or precipitation, which leads to the development of rolling component and reduces the strength of the cube component. In a sheet, a mixture of cube and rolling components is



desired to prohibit earing during deep drawing of that sheet [54]. It is interesting to note from the research work of Hirsch and Lücke that 95% cold-rolled alloy Al-0.007% Fe shows the retained rolling texture with a small amount of cube texture at 360°C compared to strong cube texture at 280 and 520°C [67]. It is due to the formation of iron-rich phase at 360°C, which restricts cube grain growth. When precipitation occurs before or after completion of recrystallization, the effect on boundary migration is less drastic and a strong cube texture develops. It has been observed that the growing grains of cube component can consume the recrystallized grains of non-cube component along with the surrounding deformed material [68]. It is reported that coarse iron rich particles inhibit the formation of cube texture in cold rolled aluminium alloys [11]. It is evident that increasing iron content decreases the strength of the cube texture, especially when it is present in solid solution form before rolling [11]. Even the addition of 0.1% iron to aluminium may change the recrystallization texture [69].

#### *4.8.3. Effect of recrystallization texture on microstructure and mechanical properties*

Grain size and mechanical properties can be controlled by maintaining proper texture in recrystallized microstructure. It is already discussed in §4.8.2 that Al<sub>3</sub>Fe particles suppress the cube texture formation. Cube grain growth is also inhibited by “orientation pinning” [59]. Orientation pinning occurs when a recrystallized grain of a given orientation grows into deformed material of its own orientation. It results in a low angle boundary and the growth of the recrystallized grain will practically stop due to the low mobility of low angle boundaries. The case where orientation pinning is expected to be important is for the growth of grains with the same orientation as a main component in the deformation texture, like retained rolling components. Vatne et al. have also reported a case where the growth of cube grains has been most likely reduced due to orientation pinning. This may be due to an extremely high as-deformed cube fraction of 35% [70]. Therefore, fine grains can be generated in the foil by controlling recrystallization texture during intermediate annealing treatment, which improves the mechanical properties of the foil. Blade has demonstrated that earing tendency of aluminium alloy increases with increase in cube texture component [71]. It is shown that earing tendency becomes zero when 25% cube texture is present.

## **5. Grain growth**

### **5.1. The effect of second-phase particles and orientation gradient**

During grain growth, boundary migration is retarded by a Zener drag effect of the second-phase particles. As the driving pressure of grain growth is extremely low, particles may have a very large influence on both the kinetics of grain growth and the resultant microstructure. Humphreys et al. have elaborately discussed the growth related formulation, i.e., driving and pinning pressures for growth, grain growth rate, limiting grain size due to particles, etc. [4]. In the case of planar grain boundary, Zener limiting grain size ( $D_{Zener}$ ) which equals to the critical radius of island grains, arises from the balance between the driving pressure for grain growth ( $P$ ) and Zener pinning pressure ( $P_z$ ).

$$D_{Zener} = \frac{4r}{3f} \quad (2)$$

Equation 2 is only applicable to those materials where the second-phase particles are stable during grain growth. Therefore, stability of the particles with temperature is one of the important factors for growth mechanism. The instability of the second-phase particles, i.e., precipitation after the formation of grain and subgrain structure or coarsening of particles during grain growth, affects the grain growth by a different way and it has been discussed elsewhere [4].

Orientation gradient has a greater effect on the grain boundary mobility. High stacking fault energy alloys (Al, Ni and  $\alpha$ -iron etc.) readily form a cellular or subgrain structure during deformation. This structure is often not uniform and orientation gradients are usually developed here. In addition, many alloys produced by thermomechanical processing develop a preferred orientation or texture by recrystallization and grain growth. A model developed by Ferry illustrates that the grain coarsening is rapid for large orientation gradient and particle-free systems, and it is markedly reduced in a system containing a large volume fraction of fine particles, despite the presence of the orientation gradient [72]. The latter case occurs due to retardation of the onset of recrystallization.

## 5.2. The competition between normal and abnormal grain growth

Grain growth may be divided into two types, normal grain growth and abnormal grain growth. When the grains grow uniformly with a narrow range of size and shapes and the grain size distribution is independent of time, then, this type of grain growth is called “normal” or “continuous” grain growth. In “abnormal” or “discontinuous” grain growth, few grains are large compared to the rest and grain size distribution is bimodal. Since this discontinuous growth of selected grains has similar kinetics to primary recrystallization and has some microstructural similarities, abnormal grain growth is sometime called secondary recrystallization. The normal grain growth theory is based on the grain boundary interfacial free energy as the driving force. If the initial grain size distribution is too wide, a fraction of large grains will grow in an abnormal manner until all the other grains have been consumed. When completed, this process has resulted in a more narrow size distribution and at longer times the steady state may be approached asymptotically. It appears convenient to define this state as a normal grain growth. Abnormal grain growth may sometime be a necessary initial stage toward find normal grain growth [73].

In spite of considerable research efforts, the origin of abnormal grain growth is yet to be fully understood. However, by Monte Carlo simulation it has been observed that two different conditions may initiate abnormal grain growth [74]. Firstly, anisotropy in the grain boundary energy may lead to rapid growth of grains having boundary energies much lower than the average [75]. This is often the case when the material exhibits a strong primary recrystallization

texture [76]. Secondly, abnormal grain growth may arise from anisotropy in the grain boundary mobility [74, 77].

Hillert has deduced three conditions for the development of abnormal grain growth in a material as follows [73]:

1. Normal grain growth cannot take place due to particle pinning,
2. The average grain size has a value below the limit  $1/2z$  (where  $z = 3f/4r$ ,  $f$  and  $r$  are the volume fraction and the radius of particles, respectively).
3. There is at least one grain much larger than the average.

Whether these conditions are automatically fulfilled in a material where the normal grain growth has stopped growing due to the presence of second phase particles, is a question of considerable practical importance. Again, the growth of a very large grain in a material where the normal grain growth has stopped depends upon the value of the final grain size reached by the normal grain growth. The mathematical analysis, on the other hand, predicts that normal grain growth should proceed up to the limit  $1/2z$ . This causes abnormal grain growth to be impossible for a limited time period when the normal grain growth has started to slow down but has not yet reached the limit. Therefore, initiation of abnormal grain growth is possible by a continuous decrease in the  $z$  value. This effect can be accomplished increasing the particle size  $r$  through coalescence or decreasing the volume fraction  $f$  by dissolving the second phase. Any process that leads to a slow increase in the grain size limit may initiate the development of abnormal grain growth. Most cases of abnormal grain growth, met within practice, seem to be connected with the dissolution of the second phase rather than the coalescence [73].

The occurrence of abnormal grain growth may be limited by “nucleation” rather than growth considerations. It is evident from a large number of publications that abnormal grain growth is likely to occur as the annealing temperature is raised and as the particle dispersion becomes unstable [76, 78-79]. If a single strong texture component is present in a fine-grained recrystallized material, then abnormal grain growth commonly occurs on further annealing at high temperatures [80, 81]. This is due to a low misorientation and hence low energy and mobility of grain boundaries in highly textured materials. Due to the presence of another texture component, higher energy and mobility are introduced in the boundaries to migrate preferentially by a process, which is closely related to primary recrystallization. The avoidance of abnormal grain growth at elevated temperatures is an important aspect of grain size control in steels and other alloys. A safe method of avoiding abnormal grain growth would be to form an average grain size so much larger than the particle-limited grain size [73]. Another method is by choosing a material with a larger volume fraction of second phase particles. In other words, it can be said that abnormal grain growth is not likely to occur where most of the particles are observed to be situated at the grain boundaries.

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