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Mechanism of cube grain nucleation during recrystallization of deformed commercial purity aluminium

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Abstract. **Cube texture is a sharp recrystallization texture component in *fcc* metals like aluminium, copper, etc. It is described by an ideal orientation i.e. (100) ⟨100⟩. The subject of cube texture nucleation i.e. cube grain nucleation, from the deformed state of aluminium and copper is of scientific curiosity with concurrent technological implications. There are essentially two models currently in dispute over the mechanism of cube grain nucleation i.e. the differential stored energy model founded on the hypothesis proposed by Ridha and Hutchinson and the micro-growth selection model of Duggan *et al*. In this paper, calculations are made on the proposal of Ridha and Hutchinson model and the results are obtained in favour of the differential stored energy model. It is also shown that there is no need for the micro-growth model.**

Keywords. **Recrystallization; cube texture; commercial purity aluminium; differential stored energy model.**

1. Introduction

When a polycrystalline metal is deformed, due to the choice of slip systems in various grains, a deformation texture results (Grewen and Huber 1978). When the same material is recrystallized, a sharp recrystallization texture develops of which cube texture, (100) ⟨001⟩, is a classic example.

It is well known that cube texture develops from cube oriented deformation bands in *fcc* metals like aluminium and copper (Kashyap 1991). Cube textures arise from the cube grains, which nucleate from cube oriented bands in deformed aluminium.

Doherty *et al* (1993) and Hjelen *et al* (1991) resolved the dispute between oriented nucleation and oriented growth hypothesis in favour of oriented nucleation by selecting cube grain nucleation and growth as a testing ground in commercial purity aluminium. However, the problem of nucleation of cube grains in deformed aluminium is still under dispute.

The classic mechanism of cube grain nucleation in deformed aluminium was that of Dillamore and Katoh (1974) wherein the transition bands in the grains of aluminium were supposed to nucleate cube grains. The theory was put to trouble with direct observations of cube bands in deformed aluminium by Kashyap (1991) and Hjelen *et al* (1991). Ridha and Hutchinson (1982) proposed that a cube grain has a special characteristic in that during deformation two perpendicular slip systems operate in which they hypothesized that edge dislocations with perpendicular burgers vectors have low elastic interaction with a low Taylor factor and hence subgrains of cube orientations can recover faster and grow bigger in size (size advantage) when compared to subgrains of other orientations.

Duggan *et al* (1993) in their classic experiment on copper proposed the micro growth selection model wherein they proposed that cube subgrain grew to a larger size when they had S–(123) $\langle 63\bar{4} \rangle$ subgrain where they used the classic oriented growth model of Liebmann *et al* (1956) where a 40° ⟨111⟩ type CSL boundary existed between cube subgrain and S subgrain and hence could migrate faster (due to high mobility) and cube subgrain grows to a larger size to get the size advantage.

Samajdhar and Doherty (1995) measured directly by electron microscopy the dislocation content of cube subgrains and neighbouring S-subgrains. They found cube subgrains to be larger with less free dislocations than S-subgrains without taking the dislocation origin, they proposed the differential stored energy model wherein the cube subgrains had a lower stored energy as compared to S-subgrains which led to the growth of cube subgrains and ultimately to get the required size advantage. In this paper, an attempt has been made to calculate and justify from the hypothesis of Ridha and Hutchinson (1982) and observations of Samajdhar and Doherty (1995), of the dislocation origin of the differential stored energy model.

2. Results and discussion

Ridha and Hutchinson (1982) proposed that (figure 1) a cube grain has two perpendicular slip systems. The two perpendicular edge dislocations are shown in figure 2.

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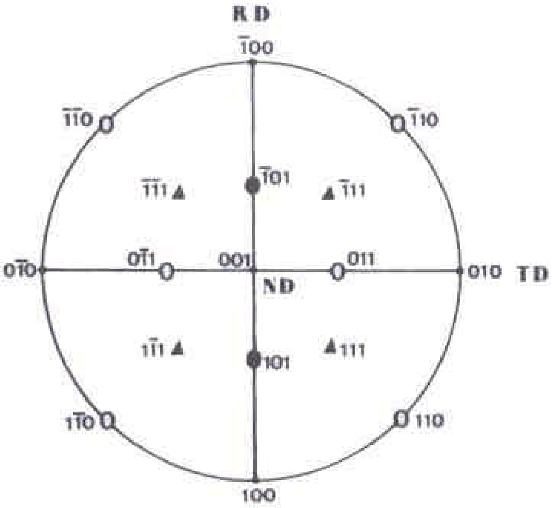

Figure 1. Stereographic projection showing all possible planes {111} and directions ⟨110⟩ for *fcc* metals. Systems active during rolling deformation of the cube orientation are indicated by solid symbols.

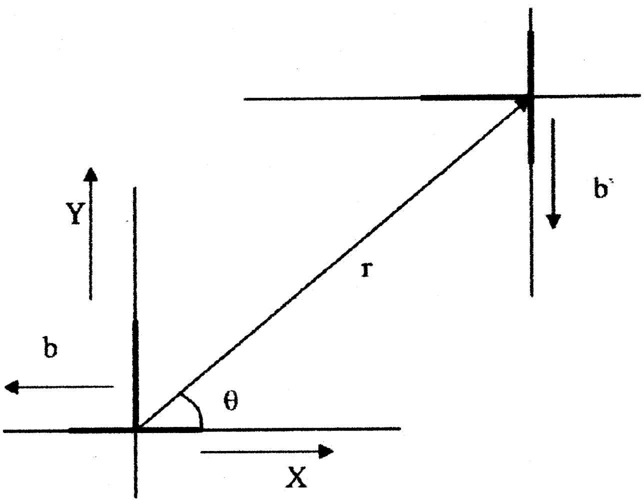

Figure 2. Two perpendicular edge dislocations.

The force of interaction between two dislocations parallel to the *z*-axis with burgers vectors

$$\vec{b}_1 = b_x i + b_y j + b_z k, \tag{1}$$

$$\vec{b}_1 = b'_x i + b'_y j + b'_z k, \tag{2}$$

is given by Weertman and Weertman (1992).

$$F = \frac{n}{2p(1-n)r}[\{\cos q(1-n)b_z b'_z + b_x b'_x (1-2\sin^2 q)$$
$$+ b_y b'_y (1+2\sin^2 q) + ...$$
$$\sin q(b_x b'_x + b_y b'_y)(1+2\sin^2 q)i\} + \sin q\{(1-n)b_z b'_z$$
$$+ b_x b'_x (1+2\cos^2 q) - ...$$
$$b_y b'_y (1+2\sin^2 q) - \cos q(b_y b'_y + b_x b'_x)(1-2\sin^2 q)\} j]. \tag{3}$$

Let $b_x = b$, $b_y = b'$,

then

$b_x = b_y = b_z = b_x = 0$.

Substituting into (3) and simplifying,

$$F = \frac{n}{2p(1-n)r}[\sin q(b_x b'_y)(1-2\sin^2 q)i$$
$$- \cos q(b_x b'_y)(1-2\sin^2 q)j], \tag{4}$$

where n is the shear modulus, GPa and r the Poisson's ratio.

$$F = \frac{n}{2p(1-n)r}[\sin q(bb')(1-2\sin^2 q)i - \cos q(bb')$$
$$(1-2\sin^2 q)j], \tag{5}$$

$$F = \frac{nbb'}{2p(1-n)r}(1-2\sin^2 q)[\sin q\, i - \cos q\, j], \tag{6}$$

$$F = \frac{nbb'}{2p(1-n)r}\cos 2q[\sin q\, i - \cos q\, j], \tag{7}$$

$$F = \frac{nbb'}{2p(1-n)r}\cos 2q[\sin q\, i - \cos q\, j], \tag{8}$$

$$F = \frac{nbb'}{2p(1-n)r}[\cos 2q \sin q\, i - \cos 2q \cos q\, j], \tag{9}$$

$$\vec{F} = F_x i + F_y j, \tag{10}$$

where F_x is the glide force, and F_y the climb force between dislocations.

Table 1 shows the calculations for the glide force, F_x and climb force, F_y, as a function of angle, q, between dislocations.

Figure 3 shows F_x *a* cos 2q sin$q$ and F_y *a* cos 2q cos$q$ vs q for a range of angles.

It can be seen from the above calculations that $F_x < 0$ and $F_y = 0$ for $q = 90°$ or $q = -90°$, which means that the glide force is attractive and climb force is zero. It also means that the two perpendicular edge dislocations will form a tilt boundary as shown in figure 4.

Hull and Bacon (1997) have given the geometry of this type of unsymmetrical tilt boundary. This type of boundary

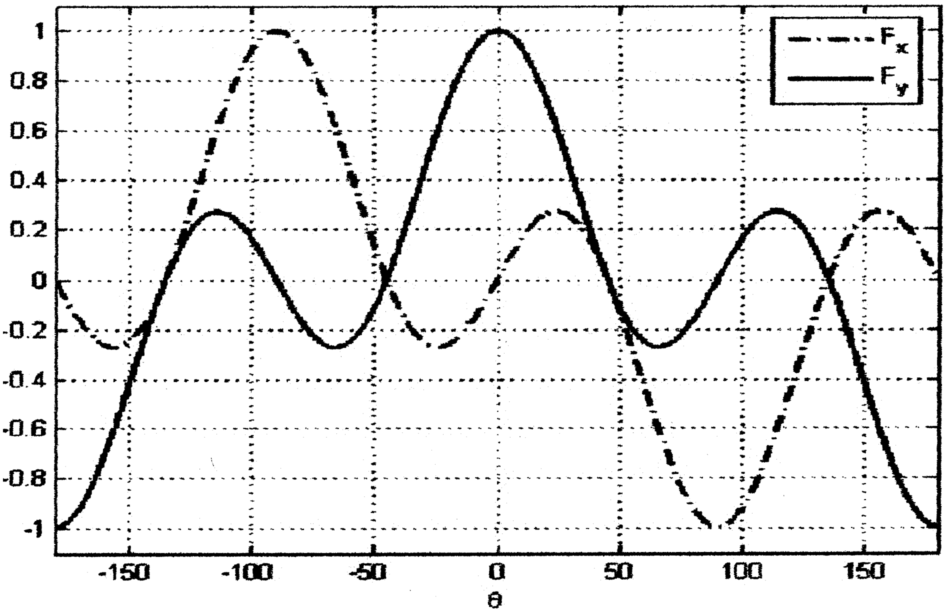

Figure 3. Plot of F_x and F_y vs q.

Figure 4. Unsymmetrical tilt boundary.

Table 1. Variation of F and F with q - der E 70 GPa, $b =$ $^{-9}$ m, $r =$ $^{-5}$ m and $n = 0{\cdot}33$.

q	F_x	F_y
180	1·	1 061
135	5·	– ·129
–90	3·	1 676
45	·383	–1 471
	0	6·
45	–2 383	1·
90	–3 345	·676
135	·438	–6 129
180	1·421	1·061

is in accordance with the low energy dislocation network (LEDS) as proposed by Wilsdorf (2004).

Nes and Vatne (1996) have given the criterion for growth of a nucleus in recrystallization i.e. radius R given by the Gibbs–Thomson relationship, n,

$$R > R_c = \frac{2n}{P_D}, \tag{11}$$

where R_c is the critical radius, n the surface energy, P_D the driving pressure for recrystallization, G the growth rate, $= M\,(P_D - 2r/R)$, where M = mobility.

They argued that 40° ⟨111⟩ type boundary (Σ7 coincident site lattice CSL boundary) has a low surface energy, n, due to which it grows to a large size when it is associated with a subgrain.

Nes and Vatne (1996) have worked out the growth of 40° ⟨111⟩ boundary as

$$G \quad = G \;+ \Delta G \quad = 0\;6m\frac{r_R}{R}, \tag{12}$$

where G_R is the growth rate of random boundaries and ΔG_{111} the increment in the growth rate of 40° ⟨111⟩ boundary.

Note that increment is large in the early stages due to a reduction in surface energy, ΔG_{111} is large, and mobility, M, also could be large.

The evidence for the high mobility of 40° ⟨111⟩ boundary is given by Aust and Rutter (1965) in their classic experiments of lead doped by tin.

But it is presently argued that the unsymmetrical tilt boundary proposed by the present authors with perpendicular edge dislocations leads to a lower surface energy of the tilt boundary [LEDS]. This leads to a higher growth rate of the cube subgrain without the necessity of having a S-subgrain ((123) ⟨63$\bar{4}$⟩) i.e. without a 40° ⟨111⟩ mobility advantage as proposed by Duggan *et al* (1993) in their microgrowth model.

The unsymmetrical tilt boundary as proposed should have a lower surface energy as opposed to the high mobility of 40° ⟨111⟩ CSL boundary. Therefore, from Nes and Vatne (1996) argument that a lower surface energy boundary can move faster especially in the early stages of nucleation when associated with a cube subgrain should give rise to the size advantage of cube subgrain. This enables the cube subgrain to grow faster and become a viable cube nucleus in the cube deformation band in deformed commercial purity aluminium.

From the direct observations of Samajdhar and Doherty (1995) that cube subgrains have larger subgrains size with little interior dislocations as compared to S-subgrains corroborates with the idea as originally proposed by Ridha and Hutchinson (1982) that the dislocation interactions are minimal with two perpendicular edge dislocations. The present authors propose that the unsymmetrical tilt boundary formed by perpendicular edge dislocations with the formation of LEDS as proposed by Wilsdorf (2004) has a low surface energy with which the growth rate of the cube subgrain is large and hence gets a size advantage and leads to the nucleation of the cube grain in the cube band in deformed commercial purity aluminium.

3. Conclusions

The elastic interactions between two perpendicular dislocations are calculated and a LEDS of unsymmetrical tilt boundary for the cube subgrain, is proposed which leads to a size advantage of cube subgrain because of the low surface energy and turns to the nucleation of cube grains from cube bands in deformed commercial purity aluminium.

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