Block Models and (011)[100] Recrystallized Orientation in Cold Rolled Single Crystals of α Iron—A Critical Review—

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The block theory for recrystallization nucleation proposed by B. F. Decker and D. Harker (D–H) has been critically reviewed in the light of recent experimental evidence and slip system geometry. Unpublished as well as available data of pole figures of *e.g.* (011)[100]-oriented single crystals of Fe–3%Si have been used for discussion. The experimental results on which the D–H theory was based are found to be incorrect. For example, Goss recrystallized texture is developed in crystals rolled at reductions higher than 60%, while in case of D–H reduction was 40%.

Operative slip systems on rolling have been estimated using Taylor factor, and lattice rotation has been analyzed. On this basis, existing block models have been discussed and classified into “hard” and “soft” block models in view of the relation between local deformation within blocks and total deformation by rolling. The Goss recrystallized orientation was formed from cold rolled {111}⟨211⟩ orientation by a reverse lattice rotation; *i.e.* the rotation direction is opposite to the direction caused by rolling deformation. This fact has been used to discuss the validity of the models. Preference of the hard model has been emphasized. The D–H model belongs to the soft model, but with this it is not possible to give reasonable explanations.

KEY WORDS: recrystallization; texture; silicon steel; lattice rotation; nucleation mechanism; Block hypothesis; modelling; Decker–Harker theory.

1. Introduction

Instead of the classical nucleation theory,¹⁾ so called block hypothesis is accepted as the nucleation mechanism of recrystallization. In the block hypothesis, potential nuclei are believed to be certain local regions (or blocks) already present in deformed material before annealing, and therefore the nucleus orientation is nearly the same as the block orientation.

Recrystallization is known to occur preferentially along deformation bands or highly strained narrow regions associated with certain inhomogeneous deformation inside the grains or along the boundaries. Therefore, the blocks, which will be responsible for the recrystallization nucleation, are considered to be narrow (or small) as well as high energy ones. The models by Burgers and Louwse,²⁾ Cahn³⁾ or those associated with certain deformation banding⁵⁾ will be typical examples, and will be called tentatively as “narrow block” models. On the other hand, like in the model by Decker and Harker,⁴⁾ the blocks are considered to be regions of considerably wide (or large) volume in the deformed crystal, and will be called “wide block” models.

During the deformation, the shapes of materials are modified, and the shape modification as a whole is characteristic of the deformation mode. This will be called “proper deformation” hereafter. In case of rolling, for example, the proper deformation will be described

as reduction of thickness and increase of length. In the wide block model, blocks behave like matrix upon deformation, and the deformation of blocks is essentially identical with the proper deformation. The way of calculation of block orientation change due to slip by Dillamore and Katoh⁶⁾ is an example of this model. In contrast, in case of narrow block model, the proper deformation concept is not applicable to block deformation; the deformation has to be ascribed to a certain kind of strain accommodation in a manner, *e.g.* formulated by Akisue.⁷⁾ As a result, the deformation or lattice rotation in blocks is not directly related to the proper deformation scheme.

The present author hesitates to use the wide block model. This is because the lattice rotation of nuclei will be quite different from that in matrix; the deformation and resulting lattice rotation in matrix occurs as a result of dislocations passing through, while rotation in nuclei blocks is associated with dislocations remaining in the material. In the studies of single crystals, evident fact is that recrystallization is associated with inhomogeneity of deformation, and the orientations of recrystallized grains are remarkably deviated from matrix orientation.^{8–28)}

However, the wide block model has been used to discuss polycrystalline materials. Such confusion will be originated from some incorrect explanations made by Decker and Harker (D–H)⁴⁾; they considered that high

energy wide blocks were established in rolled single crystals of Fe-3%Si as a result of dislocation interaction due to the proper deformation scheme.

Discovery of orientation dependence of recrystallization kinetics, which was most clearly demonstrated by Hibbard and Tully,¹³⁾ seems to enhance misleading. The experiments of D-H have many problems, but the theory has been accepted because of the convenience for application. As will be shown below, the theory as well as experimentals by D-H seems to be almost baseless.

It is the purpose of this report to point out such problems, which are not discussed in previous papers.^{17,24,33)} The origin for the *Goss* recrystallized orientation will be discussed in the light of pole figure evidence and the geometry of slip deformation.

2. High Energy Block Concept and Orientation Dependence of Recrystallization Kinetics

The recrystallized textures in Fe-3%Si or α iron have been examined as the function of initial orientation and reduction in thickness by rolling, using single crystals of various orientations.^{4,8-28,38,39)} Particular interest has been focused on the mechanisms of forming (011)[100]-oriented recrystallized grains.^{4,9,11,12,16,17,21,22,27)} The (011)[100] orientation has also been called as “*Goss*” orientation. In the early study of cold rolled Fe-3%Si single crystals initially with *Goss* orientation, Decker and Harker proposed the wellknown “high energy block” theory for recrystallization nucleation.⁴⁾

The concept proposed by Decker and Harker⁴⁾ is described as the following three items;

(1) The crystallites which are in (011)[100] orientation before rolling will keep that mean orientation during rolling (*i.e.* the *Goss* orientation is stable during rolling).

(2) There exists orientation dependence of stored energy. The crystallites initially oriented to be (011)[100] become more highly strained than others.

(3) In the annealing process the highest energy regions like *Goss*-oriented crystallites will recrystallize first, becoming lowest energy regions which will be able to consume surroundings.

As will be shown later, the above statement is based on incorrect recognition on recrystallized textures from their incomplete pole figures. Firstly, the reduction of 40% is not enough to develop preferred *Goss* recrystallized texture (*PGT*) in *Goss*-oriented single crystal. Commercial silicon steels in 1950 were so brittle that the reductions even by hot rolling with intermediate annealings had to be reduced to 40%. Actually the reduction of around 70% is most convenient for the formation of *PGT*.^{9,10)}

Secondly, the high energy block concept will be correct, but the mechanism of becoming high energy is controversial. Though D-H stated that higher stored energies were attributed to double slip operation in *Goss*-oriented blocks, this cannot be a general explanation for high energy storage. For example, cold rolled (001)[110]-oriented single crystals are deformed by double slip operation but of the lowest stored energy.^{10,23)}

Thirdly, since the blocks having *Goss* orientation have large volume at such low reduction as 40%, the blocks can serve as matrix for recrystallization rather than serve as nuclei, and therefore recrystallized orientations are other than *Goss* orientation, as will be shown in Sec. 4.

Dependence of recrystallization kinetics on the orientation of single crystals was documented by Hibbard and Tully (H-T).¹³⁾ Many investigators, however, seem to have connected the result of H-T to the orientation dependence concept of nucleation by D-H. The data by H-T is concerned with the dependence of initial orientation, and therefore are related with deformed matrix orientation, too. In the cases where the wide block model is applicable, this connective view will be correct. Application of the D-H model to polycrystalline steels by Takechi *et al.*,²⁹⁾ or Hutchinson³⁰⁾ might be based on such recognition. Studies of orientation dependent recovery in nuclei using X-ray diffraction lines width^{31,32)} will be reasonable as far as the wide block model is concerned. But in view of narrow block concept, it is unlikely that diffraction evidence is available from small total volume of nuclei blocks; the data obtained so far will mostly be those from matrix.

3. Conditions Required for the Occurrence of the Preferred *Goss* Recrystallized Texture (*PGT*)

Except the case of $\{223\}\langle 110 \rangle$ initial orientation reported by Hashimoto *et al.*,²¹⁾ the (011)[100] or *Goss* texture of primary recrystallization is formed when the initial (*i.e.* before rolling) orientations of cold rolled single crystals are either of (011)[100], (111)[2 $\bar{1}\bar{1}$], ($\bar{1}\bar{1}\bar{1}$)[211] or those between them. **Table 1** shows the initial orientation, reduction in thickness and annealing temperatures in these crystals. In order for more detailed discussion, marks *SC*, *M*, *A*, *B*, and *C* are used in Table 1 to describe the recrystallized textures; *SC*: much scattered orientations, *M*: (011)[100], *A*: $\{012\}\langle 100 \rangle \sim \{045\}\langle 100 \rangle$, *B*: $\{112\}\langle 201 \rangle \sim \{112\}\langle 311 \rangle$, *C*: $\{111\}\langle 110 \rangle \sim \{111\}\langle 651 \rangle$, respectively. The names *M*, *A*, *B* or *C* is originally due to Dunn.^{8,9)} Marks *G*, *GT*, *DT*, *D1* and *D2* will be used at this time to describe the orientation distribution spectra of cold rolled textures in (011)[100]-oriented single crystals. Schematic spectra shapes for *G*, *GT*, *DT* and *D2* are shown in **Fig. 1**.

Such distribution spectra can be described as orientations composed of various angles of lattice rotation, from (011)[100] to either one of (111)[2 $\bar{1}\bar{1}$] or ($\bar{1}\bar{1}\bar{1}$)[211] about a rotation axis of [0 $\bar{1}\bar{1}$] or *TD* (transverse direction). When the initial orientation is close (within 5 deg. or so) to the (011)[100], the rolled textures in high reductions are composed of doublet variants, *i.e.* (111)[2 $\bar{1}\bar{1}$] and ($\bar{1}\bar{1}\bar{1}$)[211].^{9,38,39)} If the initial orientation is deviated more than 5 deg. toward $\{111\}\langle 211 \rangle$, the rolled textures are composed of singlet variant. These properties can be explained by a slip mechanism^{9,17)} (as seen in Fig. 5).

The data by Taoka *et al.*^{17,18)} or present author^{19,24,33)} in Table 1 were determined from surface layers as well as interior by reflection method. As a result, the reduction dependence of lattice rotation in usually

Table 1. Initial orientations, reductions in thickness (below ~75%), cold rolled or recrystallized textures in single crystals of Fe-3%Si having initial orientations of which $[0\bar{1}1]$ axes are parallel to TD .

Initial orientation	Reduction (%)	Cold rolled textures*	Recrystallized textures**	Reference
(011)[100]	23	<i>G</i>	<i>SC</i>	Littmann ³⁹⁾
	41	<i>GT</i>	<i>SC</i>	
	64	<i>DT</i>	strong <i>A</i>	
	70	<i>D2</i> doublet	strong <i>A</i>	
(011)[100]	40	<i>G</i> or <i>GT</i>	diffuse <i>M</i> or <i>SC</i>	Decker <i>et al.</i> ⁴⁾
(011)[100]	70	<i>D2</i> doublet	strong <i>M</i> , weak <i>A</i> , <i>C</i> & <i>B</i>	Dunn ⁸⁻¹⁰⁾
(133)[6 $\bar{1}\bar{1}$]	70	<i>D1</i> singlet	strong <i>M</i> , weak <i>C</i> , <i>A</i> & <i>B</i>	
(233)[3 $\bar{1}\bar{1}$]	70	<i>D1</i> singlet	strong <i>M</i> , medium <i>C</i>	
(111)[2 $\bar{1}\bar{1}$]	70	<i>D1</i> singlet	strong <i>M</i> , weak <i>C</i>	
(011)[100]	30	<i>G</i>	<i>SC</i>	Hu ¹²⁾
	50	<i>GT</i>	<i>SC</i>	
	70	<i>D2</i> doublet	strong <i>M</i>	
(011)[100]	45	<i>GT</i>	diffuse <i>M</i>	Gokyu <i>et al.</i> ¹⁶⁾
	65	<i>D2</i> doublet	strong <i>M</i>	
(155)[10, $\bar{1}$, $\bar{1}$]	45	<i>GT</i>	diffuse <i>M</i>	
	65	<i>D2</i> doublet	strong <i>M</i>	
(011)[100]	8.2	<i>G</i>	no recrystallization	Taoka <i>et al.</i> ^{17,18)} and Furubayashi ^{19,33)}
	21	<i>GT</i>	<i>SC</i>	
	32	<i>GT</i>	<i>SC</i>	
	47	<i>GT</i>	<i>A</i> & <i>B</i>	
	58	<i>D2</i> doublet	strong <i>M</i> , weak <i>A</i>	
	71/75	<i>D2</i> doublet	strong <i>M</i> and/or <i>A</i>	
	52	<i>D1</i> singlet	<i>M</i> & <i>A</i>	
(233)[3 $\bar{1}\bar{1}$]	79	<i>D1</i> singlet	strong <i>M</i> , weak <i>C</i> & <i>A</i>	
(111)[2 $\bar{1}\bar{1}$]	74	<i>D1</i> singlet	strong <i>M</i> , weak <i>A</i>	
(011)[100]	60	<i>D2</i> doublet ^{3*)}	strong <i>M</i> (wet) ^{3*)}	Furubayashi <i>et al.</i> ²⁴⁾
		<i>GT</i> ^{4*)}	<i>A</i> + <i>B</i> (dry) ^{4*)}	
		<i>D2</i> doublet	strong <i>M</i> (interior)	

*¹⁾ *G*: pole distribution in low reductions in which the highest density still occurs near initial (011)[100] orientation. *GT*: continuous pole distribution with nearly equal density from the (011)[100] toward (111)[2 $\bar{1}\bar{1}$] and (111)[211]. *DT*: rotational distribution similar to *GT* but having the tendency to form peak(s) at (111)[2 $\bar{1}\bar{1}$] and/or (111)[211]. *D1*: the singlet (111)[2 $\bar{1}\bar{1}$] texture without large rotational distribution. *D2*: the doublet {111}<211> texture, i.e. strong peaks at (111)[2 $\bar{1}\bar{1}$] and (111)[211] with weak distribution between them. See Fig. 1.

**¹⁾ *SC* means much scattered orientations without strong ones. *M* = (011)[100], *A* = {012}<100> ~ {045}<100>, *B* = {112}<201> ~ {112}<311>, *C* = {111}<110> ~ {111}<651>.

^{3*)} Surface textures rolled with a wet surface condition.

^{4*)} Surface textures rolled with a dry surface condition.

lubricated surfaces was found to be very similar to that in the interior, except for the case of rolling at unusually dry surface condition.²⁴⁾ On the dry surfaces the lattice rotation was greatly suppressed so that specific rolled or recrystallized texture, *GT* or *A* + *B* respectively, is formed even at 60% reduction, as shown in Table 1.

In Fig. 1, the X-ray pole density distribution of recrystallized (011)[100] (*Goss*) texture component is also shown for the single crystals of two initial orientations; (011)[100] in (a) and (111)[2 $\bar{1}\bar{1}$] in (b). Minor components other than the predominant (011)[100] of recrystallized texture are omitted in Fig. 1. From Table 1 and Fig. 1, one will see the conditions required for the formation of preferred *Goss* recrystallized texture (*PGT*) is as follows;

- 1) Strong {111}<211> rolled texture has to be present, either singlet or doublet.
- 2) Reduction of rolling has to be larger than a critical

value, R_c .

From Table 1, R_c seems to be in a range between 47 and 58% for (011)[100] crystals.

Few exceptions to this are the case of Decker *et al.*⁴⁾ or Gokyu *et al.*¹⁶⁾ where *Goss* texture (*M*) was assigned from their pole figures at reduction of 40 or 45%, respectively.

For (111)[2 $\bar{1}\bar{1}$] crystals, R_c seems to be lower but not much different from the case of (011)[100] crystals.^{17,19)} The texture *A* or *B* is developed even when *PGT* is not formed. The formation mechanism for *A* and *B* will not be discussed here.

4. Orientations of Recrystallization Nuclei by Pole Figure Analysis

Textures in a cold rolled 47% and recrystallized single crystal of (011)[100] orientation (see Table 1) are shown

Fig. 1. Schematic drawings of orientation distribution maps of cold rolled textures and recrystallized (011)[100] textures in single crystals of Fe-3%Si as a function of reduction of rolling. Ordinate axes indicate orientation or rotation angle from (011)[100] orientation. Small solid circles indicate initial orientations. (a): in case of (011)[100]-oriented single crystals, and (b): (111)[211] single crystals.

by (110) pole figures in **Fig. 2**.^{17,19)} The crystal surfaces were in a wet condition, *i.e.* lubricated with oil, during rolling. Pole figures were obtained from the surface of rolled crystal by Decker-Asp-Harker reflection method. The recrystallized textures can be described as composed of doublet {025}<100> or quadruplet {125}<921> orientations, but *PGT* is not developed.

Figure 3 shows the relation of texture intensity between recrystallized *Goss* texture and cold rolled texture, which is shown by the fraction of {111}<211> or (011)[100], in (011)[100]-oriented single crystals. The original texture data are due to Taoka *et al.*^{17,18)} The *PGT* is developed at higher reductions of rolling like 71%, where the intensity of initial (011)[100] (hereafter called as "initial *Goss*") texture already diminishes. The *PGT* fraction is rather inversely proportional to that of cold rolled (011)[100] orientation, but it increases nearly proportionally with the fraction of cold rolled {111}<211> orientation. In particular, *PGT* is developed also in (111)[211]- or (233)[311]-oriented single crystals, in spite of the fact that they both had no initial *Goss* texture, as seen in **Fig. 1**.

These facts will naturally be understandable if one assumes that recrystallization nuclei with *Goss* orientation are formed within the increasing {111}<211> regions rather than in the diminishing (011)[100] (*i.e.* initial *Goss*) regions. Except for the case of twinning or martensitic nucleation mechanism, the recrystallization nuclei are considered to have orientations similar to those of nucleation sites in the deformed state. This type of concept is known as the block hypothesis. In pole figure examinations, however, intensity peaks at poles of

Fig. 2. (110) pole figures of 47% cold rolled or/and recrystallized textures in (011)[100]-oriented single crystal.¹⁷⁻¹⁹⁾ The textures were obtained from the surface of rolled crystal by Decker-Asp-Harker reflection method. Dotted lines: cold rolled texture, solid lines: recrystallized texture for annealing at 1173 K for 900 s, and dot-and-dashed lines: recrystallized texture for annealing at 873 K for 14.4 ks and at 1173 K for 900 s. Numerals indicate the intensity multiples of random specimen. *M*, *A* and *B* are {110} poles of following orientations; *M*: (011)[100], *A*: {025}<100>, *B*: {125}<921>.

Fig. 3. Correlation of orientation strength between recrystallized *Goss* textures and cold rolled textures in (011)[100]-oriented single crystals. Abscissa or ordinate axis indicates the fraction of components in cold rolled or recrystallized texture, respectively. "47%", "58%" and "71%" mean reductions of rolling.

recrystallized orientations are not usually found in as rolled specimens. Hu⁴⁰⁾ and Walter *et al.*⁴¹⁾ have discussed this problem, and one of the conclusions is that

undetectable diffraction intensities do not always stand for the lack of nuclei. In case of Fig. 2, detectable pole intensities for recrystallized orientations *A* and *B* are not found in the cold rolled state. This seems to be almost always true for low reductions of rolling. But in high reductions (as seen in Fig. 4), weak but apparent intensity distributions (a little bit higher than 0.5 times random level) are often observed across the poles of recrystallized orientations (as seen in Fig. 4). From this, it may be said that the volume fraction of recrystallization nuclei are very low (*e.g.* less than 1 %) even in high reductions.

Figure 4 shows such examples in a (011)[100] crystal

(a) (011)[100]-oriented Single Crystal

(b) (111)[211]-oriented Single Crystal

Fig. 4. (110) pole figures (inner parts) showing relation between cold rolled texture (dotted lines) and recrystallized texture (solid lines) in single crystals of Fe-3%Si. (a): (011)[100]-oriented crystal, rolled 75 % and annealed for 2 h at 873 K,¹⁷⁾ and (b): (111)[211]-oriented crystal rolled 73 % and annealed for 4 h at 873 K.³³⁾ *M*, *C1* or *C2*: {110} poles of ideal (011)[100], (111)[511] or (111)[511] recrystallized orientation, respectively. *D*, *DA* and *DB*: ideal {111}⟨211⟩ type rolled orientations. Solid squares in (a): initial (*Goss*) orientation.

in (a) and a (111)[211] crystal in (b) both heavily cold rolled and recrystallized. Either in (a) or (b), a weak intensity distribution across the poles of recrystallized *Goss* orientation are found in cold rolled states.

It should also be noted that the recrystallized *Goss* orientation in Fig. 4(a) is not far but usually deviated from initial *Goss* orientation. In (111)[211] crystal in (b) there is no initial *Goss* component. These results are inconsistent with Decker and Harker whose statement can be described as that initial *Goss* texture is the origin for the recrystallized texture.⁴⁾ Therefore, the crystallites with the (011)[100] orientation in cold rolled material (which is not present before rolling and will be called as “deformed *Goss*” orientation hereafter), is considered to form during rolling in the {111}⟨211⟩ region, and become recrystallization nuclei with “deformed *Goss*” texture.

5. Estimation of Operative Slip Systems and Lattice Rotations

5.1. Operative Slip Systems in Rolling

In view of the experimental results described above, one has to prepare certain successful explanations for the following facts;

- (1) the minimum critical reduction of rolling, *R_c*, exists in order for the development of *P_{GT}*,
- (2) the preferred recrystallized *Goss* orientations coincide with the deformed *Goss* orientations,
- (3) the intensity of deformed *Goss* texture becomes detectable when the reduction nearly arrives at *R_c*.

The last fact will suggest that the deformed *Goss* texture is absent at the beginning of rolling, but it is newly formed from matrix regions with (111)[211] orientation.

In order to explain these results in terms of some dislocation mechanisms, slip systems operative will be discussed first. The lattice rotations in rolling can be calculated by assuming operative slip systems. The number of necessary slip systems can be reduced to two or three if the rolling is assumed as a plane strain deformation. Slip systems were defined for (011)[100] crystal^{4,9,16,17)} in this way. Planes and edges of dodecahedrons in Fig. 5(a) indicate the slip planes and directions of {111}⟨011⟩ type slip system. Experimentally, operation of {011} slip planes was most clearly identified by surface trace analysis in rolled single crystals, but fine slip traces along {112} planes were also observed.¹⁷⁾

Figure 5(b) demonstrates the change in the operative slip systems with the orientation or the lattice rotation angle from (011)[100] around *TD* axis using Taylor factor. Taylor factor for each slip system is calculated for the biaxial stress⁴²⁾ in rolling. The crystal orientation is rotated around *TD*, but the direction of rotation is unique for each slip system, as indicated by arrows along curves in Fig. 5(b). Operation of the systems with slip direction 2 (hereafter called “slip 2”) makes the crystal orientation to rotate clockwise, while that with slip direction 1 or (3 and 4) (hereafter called “slip 1, slip 3 or slip 4”) does counterclockwise. At (011)[100] orientation (*i.e.* 0 deg.), slip 1 (*a1*, *F1*, *A1*) and slip 2 (*f2*,

Fig. 5. Slip systems which are operative most in rolling for body centred cubic crystals as a function of orientation; the figure is shown only for orientations between (011)[100] and (111)[211]. Dodecahedrons in (a) are composed of {011} type slip planes, and they are drawn as viewed from TD. Edge lines are parallel to the four $\langle 111 \rangle$ slip directions indicated by 1, 2, 3 and 4 (after Taoka *et al.*¹⁷). Capital letters A, B, ... and F or small letters a, b etc. show the slip planes of {011} or {112} type, respectively. The a or f is the {112} plane between A and F or C and E, respectively. Curves in (b) show change in Taylor factor with orientation.

E2 and C2) are most and equally operative. Therefore this orientation is unstable with respect to the operation of slip 1 and slip 2, in agreement with observed results.

Once the orientation is rotated *e.g.* clockwise a little bit, activity of slip 2 becomes dominant geometrically, and the orientation is rotated clockwise toward 35 deg. In this paper, the more operative system has been called as slip 2. At intermediate reductions, therefore, more rotated (single slip) regions and less rotated (double slip) regions are expected to form, in consistent with actual wide distribution in cold rolled orientation in Figs. 1 and 2. However, the presence of the critical reduction described before will mean that the less rotated regions of double slip in the early or intermediate reductions will not be responsible for the nucleation of Goss recrystallized grains.

As the angle approaches 35 deg., slip 3 and slip 4 will come to operation. D3 and D4 have to operate equally to keep the plane strain condition. Experimentally observed cold rolled orientations have been described as ones centered at near (111)[211] (*i.e.* 35 deg.) but orientation distribution around TD still exists.

5.2. Shear Strain Estimation

Total shear strain (TSS) of slip for a computational

crystal of (011)[100] orientation to achieve required rolling reductions and lattice rotations has been numerically calculated by assuming the slip 2 only is operative. This assumption enables one to explain the rolled orientation of maximum rotation in real crystals.¹⁷ TSS is found to be 1.38 or 1.60 for single slip of $f/2$ or $C2/E2$ respectively to rotate the orientation from 0 to 35 deg., which corresponds to the reduction of 58 %, just above the critical reduction.

In case of (111)[211] computational crystal, slip 2 and slip 3/4 are assumed operative from the beginning. The shear strain for $f/2$ has to be $2\sqrt{3}$ (≈ 3.5) times larger than that for D3 or D4 so as to keep the initial orientation unchanged.^{17,33} As a result, TSS for $f/2$ or $C2/E2$ necessary to attain 58% reduction are found to be 1.16 or 1.34 respectively, which is comparable with TSS for corresponding slip in the (011)[100] case. As a reference, TSS for D3/D4 at 58% reduction is 0 and 0.23 in (011)[100] and (111)[211] crystals, respectively.

Since the critical reductions for (111)[211] and (011)[100] real crystals was nearly the same, the similarity of TSS for the slip 2 between (011)[100] and (111)[211] will suggest that there exists certain characteristic shear strain of slip (1.2–1.6) above which Goss texture is developed. Strain accumulation by way of predominant activation of single or quasi-single slip system (slip 2) will account for the preferred formation of Goss-oriented nuclei.³³

5.3. Requisites for Constructing Working Hypotheses

From above estimation, it will be necessary to construct such a working model that can explain the formation of local regions having “deformed Goss” orientation; in which the lattice rotation has to be from (111)[211] to (011)[100] and operative slip system is essentially of slip 2 only.

In case of single crystals with (111)[211] initial orientation, (011)[100] texture is not existent apparently in as rolled state. Therefore from this fact alone, the statement (1) of Decker and Harker described in Sec. 2 is evidently not correct. At the same time, the presence of deformation induced boundaries between (111)[211] and $(\bar{1}\bar{1}1)[211]$ are not always necessary for the occurrence of preferred Goss recrystallized texture. “Compromise orientation” concept³⁶ is also invalid in Goss texture development because of singlet component in rolled (111)[211] crystals.

The author has proposed a dislocation model to explain the formation of Goss-oriented nuclei from cold rolled orientation of (111)[211] during rolling.³³ The essence of the model is as follows. Dislocations of the slip 2 will be accumulated along narrow deformation bands after passing through the matrix. In (011)[100]-oriented single crystals, the passing the dislocations of slip 2 makes the matrix orientation to rotate clockwise around TD axis, though in (111)[211]-oriented single crystals the rotation is cancelled from the beginning by the reverse rotation due to the operation of slip 3 and slip 4. In both cases, great majority of remaining dislocations in the rolled crystals will be those belonging to the most active slip system; slip 2. Accumulation of

dislocations makes the narrow block orientation to rotate into opposite direction (*i.e.* counterclockwise), and thus the (011)[100] orientation is newly formed.³³⁾

Besides the pole figure analysis, direct diffraction evidence for the presence of (011)[100]-oriented regions in as rolled states has been presented so far; in which microscale textures of as rolled (reduction: $\sim 70\%$) or partially recrystallized single crystals of Fe-3%Si³³⁻³⁵⁾ or Fe-19%Cr²⁷⁾ with initial $\{111\}\langle 211\rangle$ orientation have been examined by electron microscopy and electron diffraction. From these studies, the only regions, in which the (011)[100] orientation was found, are narrow deformation bands within the matrixes of the $\{111\}\langle 211\rangle$ orientation. However, there is some discrepancy in understanding the nature of the deformation bands among the authors; the present author considered the bands as a kind of “kink bands”³³⁾ similar to fcc metals,⁵⁾ but Haratani *et al.*,³⁴⁾ Ushioda *et al.*³⁵⁾ or Tsuzaki²⁷⁾ as “shear bands”.

6. Soft vs. Hard Models for Recrystallization Nuclei

The two different views of block hypothesis, narrow block model (NBM) and wide block model (WBM), have been mentioned in the introduction. In case of the Goss texture, developed in α iron single crystals having cold rolled (111)[2 $\bar{1}\bar{1}$] texture, the WBM is not acceptable, because the shear strain directions of most operative slip systems (slip 2) to cause the rotation of orientation from (111)[2 $\bar{1}\bar{1}$] to (011)[100] are completely reverse to the proper deformation of rolling, as seen in Fig. 5(b). That is, the rotation is toward the direction from (011)[100] to (111)[2 $\bar{1}\bar{1}$] in case of proper deformation. If one accepts the way of Dillamore *et al.*,⁶⁾ slip 3 and slip 4 in Fig. 5 have to be activated much more than slip 2, in order to explain the rotation from (111)[2 $\bar{1}\bar{1}$] to (011)[100], but it seems impossible. The Goss oriented recrystallized grains are formed preferably along narrow deformation bands. This fact is another reason why the NBM is more acceptable.

The deformation bands formed in rolling of single crystals initially oriented (011)[100] or $\{111\}\langle 211\rangle$ are preferential sites for recrystallization nucleation, and will correspond to NBM. The lattice rotation due to deformation in the block is seemingly reverse to that due to proper deformation. Such situations occur if the block is harder to deform than matrix, or harder for dislocations to pass through, as in the model by the present author³³⁾ described in Sec. 5.3. As an example, the narrow blocks will be similar to, or behave like, a certain deformation induced (or geometrically necessary³⁷⁾) boundaries like the kink bands. In contrast, WBM will be characterized by the ease of deformation or of dislocation passing, and will be soft region like matrix on rolling. NBM and WBM will therefore be called also as “Hard (block) Model” and “Soft (block) Model”, respectively. The two models are compared schematically in Fig. 6.

In order for the nucleation to occur, it is necessary to form sites of large misorientation even in the Soft Model. Since uniform lattice rotation does not provide

Fig. 6. Schematic drawing of (a) Soft Model and (b) Hard Model. Slip lines inside the wide block (WB) in (a) or matrix (MX) indicate those due to the “proper deformation” (see the text). Deformation and lattice rotation in narrow block (NB) in (b) will be described as an accommodation of deformation strain in neighboring MX rather than as the proper deformation.

misorientation, only the “transition bands” between regions of different slip modes are responsible for misorientation, and become nucleation sites actually. Typical example has been found in cold rolled (001)[100] single crystals of α iron.^{11,25,26)} However, the recrystallization tendency in the (001)[100] crystals is known to be much lower than rolled (011)[100] or $\{111\}\langle 211\rangle$ crystals.¹³⁾ Transition bands in rolled (011)[100] crystals was also ineffective nucleation sites of recrystallization.^{18,19)} These facts will suggest that the Soft Model approach to the analysis of block deformation is not fruitful in predicting nuclei orientations.

7. Concluding Remark

Experimental results and explanations by Decker and Harker⁴⁾ have been suggested incorrect in the light of later experimental results. The reduction of 40% adopted by them is too low to develop preferred Goss recrystallized texture (PGT). PGT appears only after $\{111\}\langle 211\rangle$ orientation is formed by rolling, and therefore Goss-oriented nuclei will be formed during rolling from matrix with $\{111\}\langle 211\rangle$ orientation and this will be called deformed Goss orientations. The theory proposed by Decker and Harker on such experimental basis stated that potential nuclei were those regions which were of the highest stored energy by the proper deformation. With this type of construction named Soft Model, however, it seems difficult to explain the formation of crystallites with Goss orientation by rotation from (111)[2 $\bar{1}\bar{1}$] to (011)[100], which is reverse to that due to proper deformation. The development of PGT will be understandable in terms of the Hard Model or NBM (Narrow band model) in which the reverse rotation is justified by the accumulation of dislocations associated with the proper deformation.

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