Formation of Recrystallization Cube Texture in High Purity Face-Centered Cubic Metal Sheets

W. Mao

(Submitted 19 May 1999; in revised form 19 May 1999)

An investigation on recrystallization textures in high purity face-centered cubic (fcc) aluminum, copper, and nickel indicated that the cube texture is a unique dominant final texture. In a macroview of rolling deformation, a balanced activation of four slip systems can result in certain stability of some substructure with cube orientation in the deformed matrix. In the stable substructure the dislocation density is very low, and the dislocation configuration is rather simple in comparison to other orientations so that the cube substructure can easily be transformed into cube recrystallization nuclei by a recovery process. A high orientation gradient and correspondingly high angle boundaries to the deformed matrix are usually expected around the cube nuclei, which, therefore, grow rapidly. After the primary recrystallization, the size of cube grains is much larger than the grains with other orientations, which will be expensed as the cube grains grow further, so that the cube texture can finally become a dominant texture component.

Keywords cube texture, face-centered cubic metals, rolling deformation

1. Introduction

The cube texture was first observed in sheet metal more than 70 years ago (Ref 1), and it subsequently was found in different face-centered cubic (fcc) metal sheets after a recrystallization annealing. The corresponding research on the formation mechanisms of cube texture has been conducted intensively since then. Many engineering applications require the cube texture for special uses, for example, high voltage aluminum capacitor foil (Ref 2), nickel superconducting tapes (Ref 3), and new silicon steel sheets (body-centered cubic metal) (Ref 4). Therefore, it has become more important to reveal the formation mechanisms of cube texture completely.

Recrystallization is a process of nucleation and grain growth involving the migration of high angle boundaries (Ref 5). All factors that effect the nucleation and migration of high angle boundary will influence the formation of cube texture. Ibe et al. (Ref 6) found in their early work that the boundaries between recrystallization grains and deformed matrix will move very fast if they have a special orientation relationship (Ref 6). A theory of growth selection was proposed (Ref 6), which explains the mechanism of cube texture formation (Ref 7, 8) because the cube texture and main rolling texture component have just the special orientation relationship. Dillamore and Katoh (Ref 9) investigated the formation process of certain substructures in transition bands during cold rolling of cubic metals and its important influence on recrystallization nucleation (Ref 9); afterward they proposed a theory of oriented nucleation. Some experimental observations supported the theory that could also explain the formation of cube texture (Ref 10). The theories, giving different explanations of the cube texture formation, are mainly based on the investigations that concentrated on microstructure analysis.

W. Mao, University of Science and Technology Beijing, Department of Materials Science and Engineering, Xue Yuan Lu 30, 100083 Beijing, China. Contact e-mail: wmmao@public.bta.net.cn.

As observed in a macroview, the cube texture or cube orientation concerns a rolling coordinate system, that is, an origin-rolling direction-transverse direction-normal direction O-RD-TD-ND system. Therefore, even though the cube texture is formed directly in the recrystallization annealing, it should be induced initially by the geometry of rolling deformation. From this point of view, this study provides a complete explanation of cube texture formation in fcc metal sheets. The rolling inhomogeneity (Ref 11), impurity atoms, and annealing conditions (Ref 12), which may have reduced the volume fraction of cube texture, will be discussed to simplify the analysis.

2. Experimental Procedures

High purity aluminum (>99.99% Al), copper (99.995% Cu), and nickel (99.988% Ni) were homogeneously rolled (Ref 11). The rolling reductions were 95% (0.3 mm) and 98.5% (0.09 mm) for aluminum, 95% (0.5 mm) and 99% (0.1 mm) for copper, and 88.4% (0.58 mm) and 96.8% (0.16 mm) for nickel. The rolling sheets were annealed properly in a salt bath or vacuum furnace. The {111} pole figures of the annealed samples were measured using x-ray diffraction techniques (copper tube). Figure 1 shows the pole figures.

3. Results

A very strong cube texture and a weak {124} ⟨211⟩ texture were obtained after the recrystallization of 95% rolled aluminum sheet (Fig. 1a), in which both textures were observed very frequently in recrystallized aluminum sheets. The {124} ⟨211⟩ texture is commonly called the *R*-texture (Ref 7). It was demonstrated that the grains with cube orientation, or cube grains, have a much larger size than the *R*-grains (Ref 7). The orientation relationship of both grains was 40° ⟨111⟩ (Ref 6) (Fig. 1a). A unique cube texture was obtained when the rolling reduction reached 98.5% (Fig. 1b). Because of the large size of cube grains (Ref 7) and possibly the surface effect (Ref 2), the *R*-grains could not grow further and finally disappeared.

A very strong cube texture was also obtained after the recrystallization of 95% rolled copper sheet (Fig. 1c), and a twinning texture {122} ⟨212⟩ of cube orientation was observed. The orientation relationship of both grains was 60° ⟨111⟩ (Fig. 1c). Copper has a lower stacking fault energy than aluminum; therefore, cube twinning grains will form when the normal growth of cube grains is interrupted or hindered during annealing. A very sharp cube texture was obtained when the rolling reduction reached 99% (Fig. 1d), but the surface effect of the thin samples obviously reduced the possibility of twin formation.

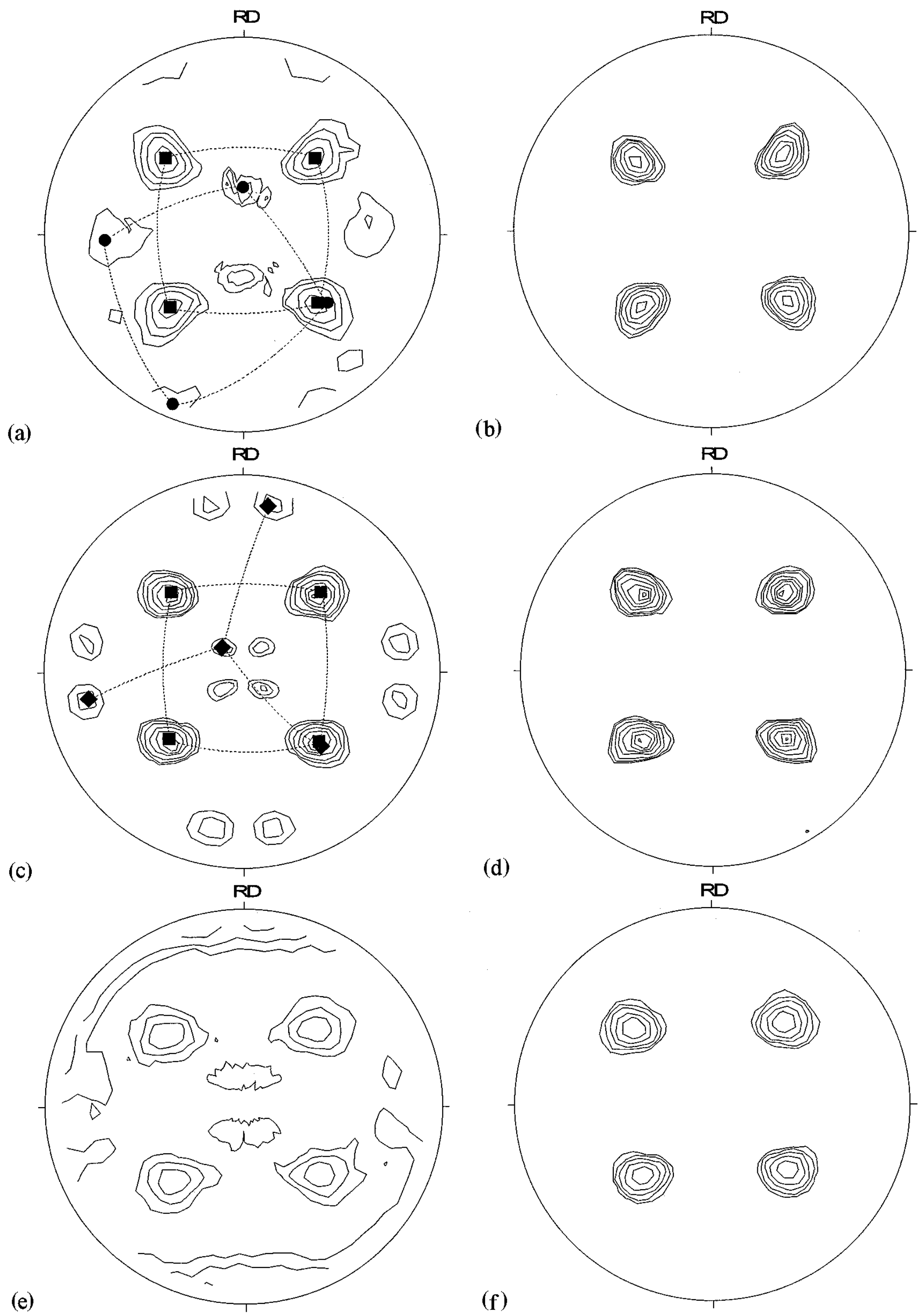

Fig. 1 The recrystallization texture in cold rolled face-centered cubic pure metal sheets ({111} pole figures: ■, {001} ⟨100⟩ ; ◆, {122} ⟨212⟩ ; and ●, {124} ⟨211⟩. Density levels: 1, 2, 4, 8, 14, 22, 32, and 44. (a) Al 95% red, maximum 18.5. (b) Al 98.5% red, maximum 30.8. (c) Cu 95% red, maximum 35.0. (d) Cu 99% red, maximum 50.5. (e) Ni 88.4% red, maximum 7.2. (f) Ni 96.8% red, maximum 21.1

The stacking fault energy of nickel is very sensitive to the impurity atoms; therefore, annealing twins can often be observed in pure nickel (Ref 13). Like copper and aluminum, a cube texture, corresponding twinning texture, and the *R*-texture were obtained after the recrystallization of 88.4% rolled nickel sheet (Fig. 1e). Nevertheless, only a strong cube texture was obtained when the rolling reduction reached 98.6% (Fig. 1f).

It can be seen in Fig. 1 that all the recrystallization textures have a rather perfect orthorhombic symmetry 222 in the O-RD-TD-ND coordinate system, and the cube texture is the unique final stable recrystallization texture.

4. Discussion

The cube texture can become dominant only under the following conditions according to the basic characteristics of metal recrystallization:

- A cube substructure should be present in the deformed matrix after heavy rolling.
- The cube substructure should become recrystallization nuclei.
- The cube nuclei should grow rapidly by means of migration of high angle boundaries.
- The cube grains could grow at the expense of the other small grains because of size effect, allowing the cube texture to become dominant and unique.

The cold-rolled high purity fcc metals appear to meet all of these conditions.

4.1 Cube Substructure in Deformed Matrix

The cube orientation in fcc metals is not a stable orientation during rolling, and it will rotate commonly toward {124} ⟨211⟩ or {123} ⟨634⟩ (Ref 7,14). The cube orientation can also rotate to {011} ⟨100⟩, if the metal flow during rolling is three dimensional, for example, in the case where widening occurs (Ref 15). The instability of the cube orientation does not mean that all cube grains will leave the cube orientation completely after rolling.

There will be two activated slip systems with same orientation factor in the cube grain in the case where the neighboring grains have no effect on a deforming cube grain. Supposing that 1, 2 and 3 represent RD, TD, and ND in the O-RD-TD-ND coordinate system, respectively, the strains produced by the two slip systems are: $\varepsilon_{11} > 0$, $\varepsilon_{33} < 0$, $\varepsilon_{12} = \varepsilon_{31} = 0$, and $\varepsilon_{23} \neq 0$. Figure 2 gives the ε_{23} value under the deformation condition. The Φ and φ_1 are the orientation angles (Ref 16), of which $\Phi = 0°$ for {001} ⟨100⟩ and $\Phi = 45°$ for {011} ⟨100⟩ are valid against $\varepsilon_{23}(\Phi)$. It can be seen that a large shear strain, ε_{23}, will be induced around cube orientation. However, because of the hindering effects of neighboring grains in polycrystalline aggregate, the shear strain, ε_{23}, could be hardly fulfilled and will be commonly reduced by activation of additional slip systems, which will result in a rotation of cube orientation toward {123} ⟨634⟩ on one hand and also lead to stability of few cube substructures on the other hand. Figure 3 demonstrates the possibly activated slip systems during rolling deformation of a cube grain. They are $(1\bar{1}1)$ [011], $(11\bar{1})$ [011], $(\bar{1}11)$ $[01\bar{1}]$, and (111) $[01\bar{1}]$, which have the same orientation factor. The ε_{23} will become zero when the neighboring grains interrupt the activation of two slip systems and the other two slip systems are, therefore, activated. In this case all the shear strains are zero, and the plastic deformation could be conducted further. At the same time the activation of the four slip systems made no orientation changes, and the cube orientation will therefore stay stable during rolling, which should be the important reason that the cube substructures may survive after a heavy rolling. Ridha and Hutchinson (Ref 10) have observed this cube substructure in cold-rolled copper sheet.

There are still another two orientations, that is, {001} ⟨110⟩ and {011} ⟨100⟩, with which the grains might be deformed by activation of four slip systems similar to the case for {001} ⟨100⟩, and therefore no shear strains or orientation changes would be induced. The two orientations are not stable during rolling. By means of activation of two slip systems, the {001} ⟨110⟩ will be rotated usually to {112} ⟨111⟩, or {225} ⟨554⟩, (Ref 17) and produce shear strain, ε_{31}. Similarly the {011} ⟨100⟩ will be rotated to {011} ⟨211⟩ (Ref 17) and produce shear strain, ε_{12}. Figure 2 gives the corresponding calculated value, ε_{31} and ε_{12}, either, in which $\Phi = 0°$ for {001} ⟨100⟩ and $\Phi = 35°$ for {112} ⟨111⟩, against $\varepsilon_{13}(\Phi)$. Furthermore, $\varphi_1 = 0°$ for {011} ⟨100⟩ and $\varphi_1 = 35°$ for {011} ⟨211⟩ against $\varepsilon_{12}(\varphi_1)$ are valid. It is clearly seen that the ε_{31} and ε_{12} are lower than ε_{23} around cube orientation, which means that the {001} ⟨110⟩ and {011}⟨100⟩ substructures should encounter less obstruction while they leave their original orientations by activation of two slip systems during rolling. It can therefore be deduced that the {001} ⟨100⟩ substructure would have higher rolling stability than those of {001} ⟨110⟩,and {011} ⟨100⟩. It has been found that the orientations located between {001} ⟨100⟩ and {001} ⟨110⟩ will move over cube orientation before they reach the final stable orientations (Ref 18), and that will increase the retained cube substructures in the deformed matrix.

4.2 Formation of Cube Nuclei

Dislocation slip is the dominant plastic deformation mechanism for fcc metals. The strains induced by a dislocation slip γ should be (Ref 19):

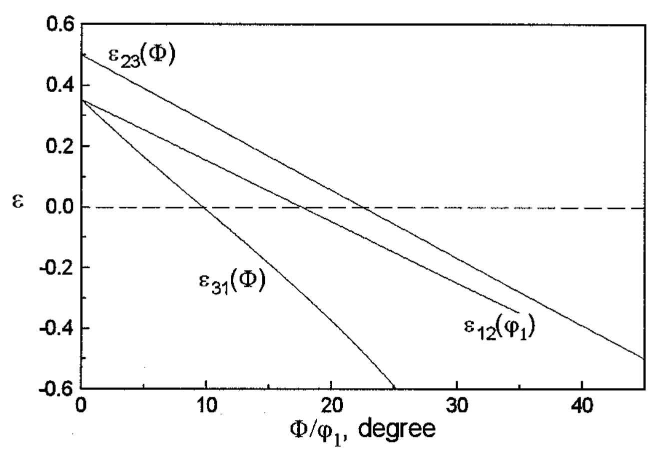

Fig. 2 Different shear strains induced by two slip systems starting from orientation {001} ⟨100⟩ , {001} ⟨110⟩ , or {011} ⟨100⟩

$$\{\varepsilon_{ij}\} = \begin{bmatrix} \varepsilon_{11} & \varepsilon_{12} & \varepsilon_{13} \\ \varepsilon_{21} & \varepsilon_{22} & \varepsilon_{23} \\ \varepsilon_{31} & \varepsilon_{32} & \varepsilon_{33} \end{bmatrix} =$$

$$\gamma \begin{bmatrix} b_1n_1 & \frac{1}{2}(b_1n_2 + b_2n_1) & \frac{1}{2}(b_1n_3 + b_3n_1) \\ \frac{1}{2}(b_2n_1 + b_1n_2) & b_2n_2 & \frac{1}{2}(b_2n_3 + b_3n_3) \\ \frac{1}{2}(b_3n_1 + b_1n_3) & \frac{1}{2}(b_3n_2 + b_2n_3) & b_3n_3 \end{bmatrix} \quad \text{(Eq a)}$$

in which the Burgers vector is $b = b_1 + b_2 + b_3$, and the normal vector of slip plane is $n = n_1 + n_2 + n_3$. For common cases there will be $\varepsilon_{ij} \neq 0$; therefore, the simple single slip will result in strain discontinuity among differently oriented grains. In order to coordinate the strain continuity of polycrystalline aggregate, which should be generally maintained during plastic deformation, multislips of dislocations are usually activated in rather complicated ways under the reaction stresses induced by neighboring grains (Ref 15, 20). That will produce interactions, twists, and blockages of dislocations, which increase drastically the dislocation density and complicate the dislocation configuration. However, if the cube substructure is deformed by activation of the four slip systems in a balanced way so that the initial orientation remains, the strains induced by the slip systems should be:

$$\{\varepsilon_{ij}\} = \begin{bmatrix} \varepsilon_{11} > 0 & 0 & 0 \\ 0 & 0 & 0 \\ 0 & 0 & -\varepsilon_{11} \end{bmatrix} \quad \text{(Eq b)}$$

The cube substructure deformed in this way will undergo very limited reaction stresses from the neighboring grains, and the complicated multislip as well as crossslip will rarely appear. The two Burgers vectors concerning the four slip systems are perpendicular mutually, according to Fig. 3; therefore, the possibilities of dislocation reactions and corresponding twists and blockages are decreased. The most dislocations will slip out of the deformed cube substructure, in which the dislocation density is much lower than that in other substructures and the dislocation configuration, becomes rather simple. The differences in dislocation density and configuration lead to a clear tendency that the cube substructure would be transformed into recrystallization nuclei very preferentially and easily during annealing, and the nucleation rate for cube recrystallization grains is therefore drastically increased. The advantages of cube oriented nucleation were also observed in cold-rolled copper sheet (Ref 10).

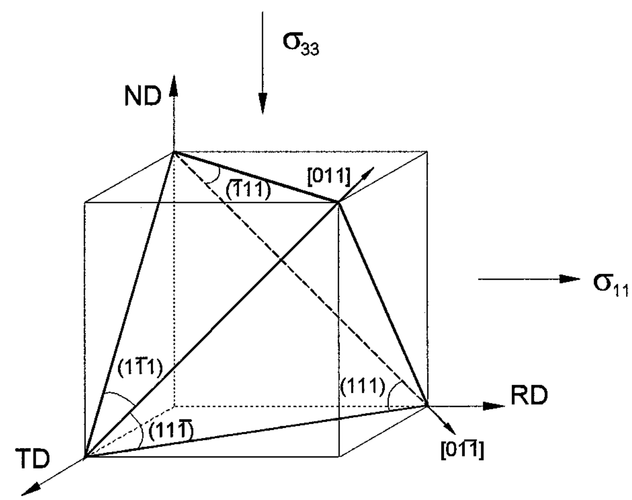

Fig. 3 The possibly activated slip systems in a cube grain during rolling

4.3 Growth of Cube Nuclei

As discussed previously, few cube substructures can keep their stability around the cube orientation during rolling, while most of the neighboring cube grains leave it (Ref 7,17). Figure 4(a) shows the orientation factor of the activated slip systems while {001} ⟨100⟩ is rotated to {011} ⟨100⟩. It is seen that a softer orientation can be obtained if the cube grain is turned toward {011} ⟨100⟩. The rate of orientation change relative to the

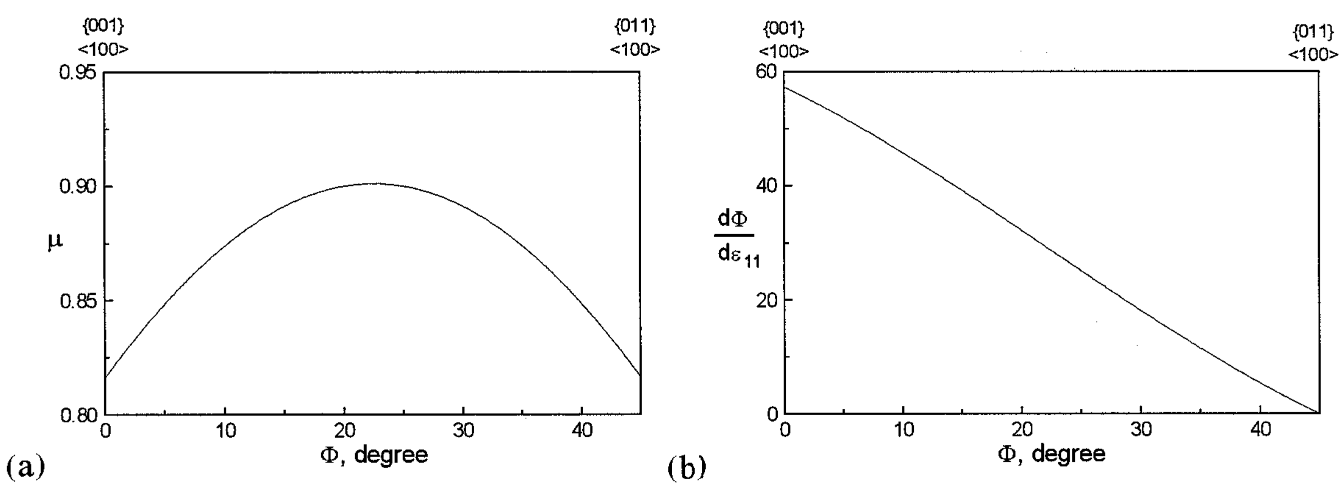

Fig. 4 The orientation factor and corresponding rate of orientation change during the rotation of {001} ⟨100⟩ to {011} ⟨100⟩. (a) Orientation factor. (b) Rate of orientation change: $d\Phi/d\varepsilon_{11}$

strain is a function of the grain orientation under the rolling condition. Concerning Fig. 4(a), Fig. 4(b) gives the corresponding rate of orientation change $d\Phi/d\varepsilon_{11}$. According to Fig. 4, cube grains neighboring a stable cube substructure will have softer orientation, be deformed relatively easily by activation of two slip systems, and have a high rate to leave cube orientation so that the orientation difference between the stable cube substructure and the fast rotated initial cube grains becomes very high and induces a high orientation gradient. Therefore high angle boundaries will be formed very easily around the retained cube substructures if they are transformed into recrystallization nuclei during annealing, which will obviously benefit the rapid growth of the cube nuclei later. A similar case should also appear if the cube grain is rotated toward {123} ⟨634⟩ instead of {011} ⟨100⟩ during rolling.

The common rolling textures in fcc metal sheets are {123} ⟨634⟩, {112} ⟨111⟩, and {011} ⟨211⟩ , with which the deformed grains always have high angle boundaries to the cube nuclei. Therefore there will be no obvious obstruction for cube nuclei to grow into the deformed matrix. Conversely, the orientation differences between other recrystallization and deformation orientations, for example, between {124} ⟨211⟩ and {123} ⟨634⟩, {001} ⟨110⟩ and {112} ⟨111⟩, and {011} ⟨100⟩ and {011} ⟨211⟩, are relatively low. So the cube nuclei in high purity fcc metal sheets will have obvious growth advantages over other nuclei during recrystallization.

4.4 Formation of Dominant Cube Texture

Because of the advantages of cube recrystallization grains on nucleation and growth mentioned previously, not only a strong cube texture can be obtained after the primary recrystallization, but the size of cube grains can also be clearly larger than those of other grains, which has been indicated in the cold-rolled, high purity aluminum sheet (Ref 7). In this case it is easily understood that the cube texture is strengthened by grain growth during further annealing and becomes a dominant final texture (Fig. 1) with the help of surface effect (Ref 2).

5. Conclusions

The following conclusions can be drawn:

The formation mechanism of the cube recrystallization texture in high purity fcc metal sheets is discussed in a statistical macroview, in which the phenomena that have been observed in the microstructure during the process of cube texture formation are explained. It was demonstrated that some substructure with cube orientation has certain stability during rolling, can remain in the deformed matrix after heavy rolling, and has some advantages in becoming recrystallization nuclei and growing in comparison to other substructures. All these characteristics of cube substructure or cube grains are closely related to the condition of crystallographic geometry in rolling deformation and lead to a final dominant cube texture.

Acknowledgment

The author would like to thank Dr. H. Hammelrath for supplying the experimental data on pure copper.

References

1. W. Köster, Beobachtungen an Kupfer zum Gesetzmäβigen Gefügeaufbau nach der Rekristallisation, *Z. Metallkde.,* Vol 18, 1926, p 112-116 (in German)
2. G. Ibe, Capacitance and Texture Formation in Aluminum Capacitor Foils, in *Directional Properties of Material,* H.J. Bunge, Ed., DGM-Informationsgesellschaft, Oberursel, 1988, p 145-156
3. A. Goyal, D.P. Norton, J.D. Budai, M. Paranthaman, E.D. Specht, D.M. Kroeger, et al., High Critical Current Density Superconducting Tapes by Epitaxial Deposition of $YBa_2Cu_3O_x$ Thick Films on Biaxially Textured Metals, *Appl. Phys. Lett.,* Vol 69, 1996, p 1795-1797
4. T. Tomida and T. Tanaka, Development of {100} Texture in Silicon Steel Sheets by Removal of Manganese and Decarburization, *ISIJ Int.,* Vol 35, 1995, p 548-556
5. H. Hu, Recovery, Recrystallization and Grain Growth, *Metallurgical Treatises,* J.K. Tien and J.F. Elliott, Ed., Metallurgical Society of AIME, 1981, p 385-407
6. G. Ibe, W. Dietz, A.C. Fraker, and K. Lücke, Vorzugsorientierungen bei der Rekristallisation Gedehnter Einkristalle aus Reinst-Aluminium, *Z. Metallkd.,* Vol 61, 1970, p 498-507 (in German)
7. W. Mao, J. Hirsch, and K. Lücke, Influence of the Cube Starting Texture on Rolling and Recrystallization Texture Development, *Proceedings of 8th International Conference of Textures of Materials,* J.S. Kallend and G. Gottstein, Ed., The Metallurgical Society Inc., Pennsylvania, 1988, p 613-618
8. W. Mao, Model for Rapidly Moving Boundaries, *Science in China,* Vol 35, 1992, p 336-343
9. I.L. Dillamore and H. Katoh, The Mechanisms of Recrystallization in Cubic Metals with Particular Reference to Their Orientation Dependence, *Met. Sci.,* Vol 8, 1974, p 73-83
10. A.A. Ridha and W.B. Hutchinson, Recrystallization Mechanisms and the Origin of Cube Texture, *Acta Metall.,* Vol 30, 1982, p 1929-1939
11. W. Mao, Texture in Inhomogeneously Rolled Aluminum Sheet, *Trans. Nfsoc.,* Vol 2, 1992, p 98-103
12. K. Ito, R. Musick, and K. Lücke, The Influence of Iron Content and Annealing Temperature on the Recrystallization Textures of High-Purity Aluminum-Iron Alloys, *Acta Metall.,* Vol 31, 1983, p 2137-2149
13. C.B. Thomson and V. Randle, A Study of Twinning in Nickel, *Scr. Metall. Mater.,* Vol 35, 1996, p 385-390
14. G.D. Köhlhoff, B. Krentscher, and K. Lücke, Texture Development in a Cube Oriented Copper Single Crystal, *Proceedings of the 7th International Conference on Textures of Materials,* Noordwijkerhoud, 1984, p 95-100
15. W. Mao, Rolling Texture Development in Aluminum, *Chin. J. Met. Sci. Technol.* Vol 7, 1991, p 101-112
16. H.J. Bunge, General Outline and Series Expansion Method, *Quantitative Texture Analysis,* H.J. Bunge and C. Esling, Ed., DGM-Informationsgesellshaft, Oberursel, 1986, p 1-72
17. W. Mao, Influence of Widening on the Rolling and Recrystallization Texture in High Purity Al with Initial Cube Texture, *Chin. J. Met. Sci. Technol.,* Vol 6, 1990, p 325-332
18. W.B. Hutchinson and E. Nes, Evolution of Cube Texture in Copper and Aluminum, *Proceedings of the 7th Ris∅ International Symposium on Metallurgy and Materials Science* (Denmark), 1986, p 107-122
19. C.N. Reid, *Deformation Geometry for Materials Scientists,* Pergamon Press, 1973, p 145-178
20. J. Hirsch and K. Lücke, Mechanism of Deformation and Development of Rolling Textures in Polycrystalline fcc Metals—II. Simulation and Interpretation of Experiments on the Basis of Taylor-Type Theories, *Acta Metall.,* Vol 36, 1988, p 2883-2904