Efficient Generation of Cube-on-Face Crystallographic Texture in Iron and its Alloys

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A novel method has been discovered for controlling the crystallographic orientation of body-centred cubic crystals of iron. The process allows the cube faces of the crystals to align with the major surface of the polycrystalline sample in sheet form. Such a texture is advantageous in terms of its magnetic properties and has been a long-standing goal for research on electrical steels. The mechanism controlling the favourable orientation obtained is associated with the fact that the cube faces are elastically compliant so that the texture can develop in a manner consistent with minimization of strain energy. For the same reason, the presence of surface oxide which modifies the elastic modulus of metal surface stifles the development of the correct texture. The new process is demonstrated to lead to a dramatic increase in the magnetic flux density and a reduction in iron losses. It also involves a remarkably shorter processing time than conventional texture control in electrical steels.

KEY WORDS: cube-on-face texture; magnetic properties; elastic anisotropy; phase transformation.

1. Introduction

Most polycrystalline metals contain crystals which are not randomly oriented in space, but rather, their axes are approximately aligned with the macroscopic shape of the sample. The non-random distribution arises because of oriented processing, heat-treatment or phase transformation. The sample is then said to be crystallographically textured and exhibits macroscopically anisotropic properties, which reflect the orientation distribution. Such anisotropy can be advantageous. In ferritic iron, the magnetic flux density rises most easily along (100) directions, in contrast to (111) directions which are said to be magnetically hard. Steels which are used for electrical applications involving rapid changes in magnetic field therefore perform better in terms of energy loss, permeability as well as magnetic flux density when the crystals are aligned with (100) directions parallel to the sheet normal. The (100) planes which contain two perpendicular (001) directions and no (111) direction are naturally the planes of easy magnetisation, so a texture in which these planes are aligned to the sheet surface with the cube edges parallel to the sample axes is known as the cube texture, \{100\}(001). A cube-on-face texture on the other hand corresponds to the \{100\}(0\bar{1}v\bar{w}) orientation.

Due to the magnetic anisotropy, the texture control is a fundamental tool to improve magnetic properties of electrical steels and can have a very large impact on energy consumption. Electrical steels are soft magnetic iron–silicon alloys with varying silicon contents, which are used for transformers, motors, generators, alternators, ballasts and a variety of other electromagnetic applications. Since more than half of the electricity generated is used to drive electrical motors, efficiency of motors affects energy consumption significantly and thereby greenhouse gas emissions. Typical efficiency of motors ranges from 83 to 92%, and their operating efficiency is far below, 62%.1,2) The only way to improve motor efficiency is to reduce motor losses. Loss components in an induction motor consist of the core loss in iron cores, the copper loss in rotors and stators, the stray load loss, and the friction and windage loss.1,3) Among them, the copper loss and the core loss, which cover at least 75% of the overall losses, can be reduced significantly by improving magnetic flux density along with reducing iron loss through the texture control of core materials.

Although the ideal texture for rotating machines is known to be the cube-on-face, there is no efficient method for its manufacture. In Fe and Fe-base bcc alloys, the cube-on-face texture does not evolve by the ordinary cold rolling and recrystallization processes. Although a strong \(\alpha\) fiber develops during the cold rolling process, which includes the \{100\}(011) component, it disappears almost completely after the indispensable recrystallization for good magnetic properties. The resulting recrystallization texture is a strong \(\gamma\) fiber, instead.3) Since late 1950’s, various process routes have been proposed to obtain the cube-on-face texture for better soft magnetic properties. The major mechanism to develop the \{100\} texture is to induce preferential growth of \{100\} grains utilizing the secondary recrystallization after cross rolling,4,5) the tertiary recrystallization in sulphur containing atmospheres6,7) the strain induced boundary migra-
tion followed by temper rolling, and the surface segregation of Sb. In addition, phase transformations can lead to the required (100) texture. The most well-known example is the direct casting (the liquid to solid phase transformation) of steel strips which have the same bcc structure up to the melting point. In the cases of the γ to α transformation in Fe and Fe-base alloys, although the texture is known to be rather randomized, several exceptions were reported which show relatively strong (100) texture. Among the proposed methods, only one process is considered to be commercially feasible now. Tomida reported that during the vacuum annealing at γ/α duplex or γ phase temperatures, manganese removal and decarburization induce the γ to α phase transformation at the surface of the sheet. As a result of the phase transformation, a thin bcc layer develops at the surface of the sheets with a strong (100) texture, although it is not clear why this should be in the (100) orientation. Furthermore, the process is very complex and needs a prolonged heat treatment at relatively high temperatures.

In the present work, we demonstrate that the (100) texture can be evolved efficiently through the γ to α transformation by cooling and there is a role to play for the orientation dependence of the elastic energy of Fe at high temperatures.

2. Experimental Procedure

Table 1 shows the chemical compositions of the alloys used in the present study. Ingots were prepared by vacuum induction melting, hot-forged to 20 mm thick plates, followed by hot-rolling to 2 mm thick sheets. These sheets, whose surface scale was removed utilizing a pickling process in 18% HCl at 60°C, were cold-rolled to thicknesses of 0.35 mm, 0.5 mm, and the like.

The measurement of critical temperatures during the phase transformation was conducted with a dilatometer and a DSC. The critical temperatures for Fe, measured at cooling/heating rates of 600°C/h, were \( A_{1/2} = 913.4°C \), \( A_{4/3} = 923.0°C \), \( A_{9/11} = 901.1°C \), and \( A_{11} = 875.5°C \). The critical temperatures for Fe–1%Si, measured at cooling/heating rates of 600°C/h, were \( A_{1/2} = 983°C \), \( A_{3} = 1003°C \), \( A_{11} = 974°C \), and \( A_{11} = 949°C \).

The heat treatments of samples were conducted in a tube furnace under various conditions. The heating and cooling rates were 600°C/h unless otherwise specified.

The textures were measured on the surfaces of specimens. The textures were evaluated by the pole figure, orientation distribution function (ODF) and the texture coefficient measured by X-ray diffraction. The texture coefficient, \( P_{hkl} \), is defined as:

\[
P_{hkl} = \frac{\sum N_{hkl} I_{hkl}}{\sum N_{hkl} I_{hkl}} \times \left( \frac{I_{hkl}}{I_{hkl}} \right)
\]

where \( N_{hkl} \) is the multiplicity factor, \( I_{hkl} \) is the X-ray intensity of the (hkl) plane for a given sample, and \( I_{hkl} \) is the X-ray intensity of the (hkl) plane for a specimen with randomly oriented grains. \( P_{hkl} \) represents an approximate ratio of the surface area covered by the (hkl) plane in the sample of interest to that in a sample with randomly oriented grains. In case of the (100) plane, \( P_{100} \) of 20 represents approximately 100% surface coverage.

3. Results and Discussion

When a Fe sheet was subjected to the γ to α transformation in a reducing gas atmosphere (H₂ gas having the dew point of −54°C), the (100) texture developed predominantly in its surface layer (Fig. 1(a)). The samples were placed in the middle of the furnace when the temperature reached 850°C. After holding at 850°C for 5 min, they were heated to the annealing temperature at the heating rate of 600°C/h. After holding for 1 min at the annealing temperature, the samples were cooled to 850°C at the cooling rate of 600°C/h. At the conclusion of the heat treatment, the samples were pulled out from the furnace and cooled in a chamber at room temperature. A (111) texture dominates when the annealing temperature is below 910°C and ferrite is the stable phase. This is typical of low carbon steel sheets. However, for higher temperatures where austenite is stable, a strong (100) texture develops (more than 60% of the surface area being covered with the (100) grains) and the (111) component almost disappears.

The evolution of strong (100) texture in pure iron containing 7 wppm sulphur is rather exceptional. When the γ to α phase transformation occurs by a diffusive mechanism to polygonal ferrite, the resulting texture is relatively weak and shows little or no variant selection. In our study, instead of the random texture, the strong (100) texture was observed in Fe. Furthermore, to develop the (100) texture in Fe, a heat treatment at 930°C for less than 25 min was sufficient.

Similar behaviour was observed in the Fe–1%Si samples which developed a strong (100) texture when transformed from austenite in a reducing gas atmosphere of H₂ gas having the dew point of −55°C (Fig. 1(b)). The samples were placed in the furnace when its temperature reached 950°C, held there for 5 min and then heated at 600°C/h to the annealing temperature. After holding for 5 min, the samples were cooled to 950°C at 600°C/h, removed and cooled in the chamber at room temperature. Austenite is stable in the Fe–1%Si between 1000 and 1310°C and α+γ mixtures occur between 980 and 1000°C. Annealing below 980°C, where ferrite is stable, leads predominantly to the (111) texture. This is typical of silicon–steel sheets. On the other hand, annealing in the austenite field at temperatures in excess of 1000°C, leads to intense (100) texture (more than 70% of the surface area is covered with the (100) texture) (Fig. 1(b)).

Figure 2 shows the (100) pole figures and \( \varphi_2 = 45°-sec-
tions of the orientation distribution function (ODF) for both the pure Fe sample annealed at 930°C for 1 min and the Fe–1%Si sample annealed at 1030°C for 5 min, which underwent the γ to α transformation during the heat treatment. The (100) pole figure of Fe shows that almost all surface grains have the (100) orientation. Also, the ϕ2=45°-section of the ODF indicates that the texture of the Fe sample has the (100) texture (Fig. 2(a)). The density along the ϕ=0 axis is very high and evenly distributed (θ-fiber). This result indicates that the (100) plane of almost all surface grains is parallel to the sheet surface, while the (100) directions are distributed randomly in the sheet plane (the (100) fiber texture). Similar result is observed in the Fe–1%Si sample (Fig. 2(b)). The results indicate that the γ to α transformation in a reducing atmosphere leads to a strong (100) fiber texture.

Figure 3 shows large columnar ferrite grains formed from austenite during cooling in the hydrogen atmospheres, with the grain size exceeding the sample thickness. The columnar structure implies that the (100) texture is homogenous through the thickness, and the scale is attributed to a high mobility of phase boundaries in relatively pure samples. Because of reduced constraint, the γ to α transformation begins at the sample surfaces which also attain the lowest temperature. The surface nucleation of (100) ferrite grains can also be supported by the effect of the annealing atmospheres on the evolution of the (100) texture (Figs. 4(a) and 4(b)), which will be discussed later. In the high-purity iron, the large columnar grains are results of the high velocity of mobile α–γ interface. As Abiko et al. showed, it can exceed 0.5 mm/s. In the iron–silicon alloy, the γ to α transformation seems to be accomplished by massive transformation which is characterised by the lack of bulk solute partitioning and short-range diffusion, so that the growth rate is large.

The dependence of the desirable texture on oxygen content was studied by placing samples of Fe–Si in a furnace when the furnace temperature reached 1150°C, holding for 15 min and cooling in the chamber at room temperature. Figure 4(a) shows that the (100) texture evolves as the vacuum pressure is reduced below 1×10−3 Pa, with the intensity increasing with decreasing pressure. These experiments show that the (100) texture evolves even with the high heating and cooling rates.

To ascertain the effect of oxygen, heat treatments were conducted in H2 gas with various dew points (Fig. 4(b)). The dew point of a hydrogen gas can be converted into the oxygen partial pressure at a given temperature in 100% H2 gas (Fig. 5). Samples were placed in a furnace at 950°C for 5 min and heated to the annealing temperature of 1030°C at 600°C/h. After holding for 10 min at 1030°C, samples were cooled to 950°C at a rate of 600°C/h; they were finally cooled in the chamber at room temperature. Figure 4(b)
shows that in this case the dew point of H₂ gas should be controlled at less than −50°C, suggesting the existence of a critical oxygen partial pressure at the annealing temperature.

This critical oxygen partial pressure corresponds to the onset of oxidation (Fig. 5), which proves that a clean bare metal surface is essential for the generation of the (100) texture. Figure 5 shows the results of a thermodynamic calculation on effect of dew point in 1 atm H₂ on oxygen partial pressure and oxygen partial pressure for oxidation of SiO₂ in fcc and bcc Fe–1%Si. The detailed thermodynamic calculation will be discussed elsewhere. At temperatures where the dew point curves are placed below the oxidation curve, the oxidation of alloy will not take place. For instance, the oxidation of Si will not take place at 1 030°C in 1 atm hydrogen of −50°C in dew point because the partial pressure of oxygen in the hydrogen atmosphere is lower than the oxygen partial pressure for the oxidation of Si at 1 030°C. Figures 4 and 5 indicate that the cleanness of the sample surfaces is essential for the evolution of the (100) texture.

The evolution of the (100) texture with columnar grains via the γ to α transformation is rarely seen in steels and the origin of the (100) texture was not explained clearly. Aspden et al. observed this in Fe by annealing for 12 h at 1 050°C or 1 150°C and cooling at 4°C/h through the γ to α transformation. However, they claimed that a critical concentration of sulphur reduces the surface energy of the {100} planes, thus promoting the nucleation of suitably favoured orientations. Hashomoto et al. reported that the (100) texture evolved in the surface layer of manganese-containing low carbon steels (0.001 wt% C and 0.34 wt% Mn) by the α→γ→α phase transformation in Ar atmosphere with slow heating and slow cooling. They explained this in terms of a variant selection mechanism resulting from the release of elastic stresses towards the normal direction. This model will be discussed later in this paper. The (100) texture has been reported to occur in the surface layer of manganese-containing low carbon steels that underwent the α→γ→α phase transformation in vacuum and H₂–95%N₂. Small equi-axed grains arise with rather week (100) texture limited to the proximity of the surface, indicating a role of surface anisotropy. Kestens et al. claimed that the orientation selection occurs by mini-

The evolution of the (100) texture should be the result of
certain anisotropic nature of ferrite nuclei during the \( \gamma \) to \( \alpha \) transformation on surfaces. We compared our results with two plausible anisotropic natures, elastic strain energy and surface energy. The mechanism to develop the \( \langle 100 \rangle \) texture is associated with the fact that the cube faces are elastically compliant so that the texture can develop in a manner consistent with minimization of strain energy.

There exists severe elastic anisotropy in Fe at around the transformation temperature, which has not been recognized before. In our calculation, the compliance components of Fe by Dever, \( s_{11}=25.9 \text{ TPa}^{-1} \), \( s_{44}=10.1 \text{ TPa}^{-1} \), and \( s_{12}= -11.7 \text{ TPa}^{-1} \), were used. The values of anisotropy parameters, \( 2C_{44}/(C_{11}-C_{12}) \), which is 1 for isotropic materials, are 2.4 at 25°C, 3.1 at 500°C, and 7.4 at 900°C for Fe. Furthermore, addition of Si to iron is known to increase the transformation temperature, which has not been recognized before. In our calculation, the compliance components of Fe by Dever, \( s_{11}=25.9 \text{ TPa}^{-1} \), \( s_{44}=10.1 \text{ TPa}^{-1} \), and \( s_{12}= -11.7 \text{ TPa}^{-1} \), were used. The values of anisotropy parameters, \( 2C_{44}/(C_{11}-C_{12}) \), which is 1 for isotropic materials, are 2.4 at 25°C, 3.1 at 500°C, and 7.4 at 900°C for Fe. Furthermore, addition of Si to iron is known to increase the elastic anisotropy. Therefore, the strain field of a ferrite nucleus would vary significantly depending on the plane orientation on the surfaces. The severe elastic anisotropy of Fe can be more clearly demonstrated by the three dimensional shape of Young’s modulus at 900°C, where the surface represents the magnitude of Young’s modulus in the direction from the origin to the corresponding point on the surface (Fig. 6(a)). Young’s modulus was calculated using the following equation,

\[
\frac{1}{E} = s_{11} - [s_{44} - 2(s_{11} - s_{12})](\alpha^2 \beta^2 + \beta^2 \gamma^2 + \gamma^2 \alpha^2)
\]

Here \( s_{11}, s_{44}, \text{ and } s_{12} \) are components of compliance tensor, and \( \alpha, \beta, \text{ and } \gamma \) are direction cosines. The Young’s modulus is minimum along \( \langle 100 \rangle \) (38.6 GPa) and maximum along \( \langle 111 \rangle \) (237.3 GPa).

Since the elastic modulus is related to the strain energy density, it is necessary to calculate the relevant elastic moduli due to the ferrite nucleation. When the \( \gamma \) to \( \alpha \) transformation occurs in a steel sample, the nucleation will take place in its surface layer because the surface temperature is lower than the interior temperature and the free surface has reduced constraint. Since the density of \( \alpha \) phase differs from that of \( \gamma \) phase, stresses develop in \( \alpha \) nucleus and \( \gamma \) matrix, resulting in the strain energy in the nucleus and the matrix. The stress state of the nuclei can be approximated by stresses along the very thin ferrite film on the surface of the \( \gamma \) matrix. In this case, the stress acting on the surface is zero and so constrains along the surface are much higher than along the thickness direction. Therefore, the strain energy of the nuclei can be proportional to relevant elastic moduli along the surface. We considered three types of biaxial elastic modulus, the average Young’s modulus (\( M_y \)), the biaxial elastic modulus under equibiaxial stress (\( M_{xy} \)), and the biaxial elastic modulus under equibiaxial strain (\( M_e \)). \( M_y \) and \( M_e \) has been discussed in detail by Lee. The average Young’s modulus (\( M_y \)) of the \( (hkl) \) plane is given by

\[
M_{y,hkl} = \frac{1}{2\pi} \int_0^{2\pi} E_{hkl} d\theta
\]

where \( E_{hkl} \) represents Young’s modulus of the \( (hkl) \) plane. Figure 6(b) shows the dependence of \( M_{y} \) on temperature. At room temperature, the elastic anisotropy is not significant, the ratio of the maximum to the minimum \( M_{y} \) being about 1.3. However, at 900°C, the ratio increases to 1.9, showing a significant increase in the planar anisotropy. Also, as the temperature increases, the plane of the maximum \( M_{y} \) changes from \( \{110\} \) to \( \{111\} \). The order of \( M_{y} \) values at 900°C is \( \{100\} < \{310\} < \{321\} \leq \{111\} \leq \{211\} < \{110\} \). In the cases of \( M_{xy} \) and \( M_e \), the orders are \( \{100\} < \{310\} < \{211\} < \{321\} < \{110\} < \{310\} < \{211\} < \{321\} < \{111\} \), respectively. In the case of \( M_{xy} \), the \( \{110\} \) plane is excluded because the directional dependence of \( M_{xy,110} \) yields negative values along certain directions above the Curie temperature.

The elastic strain energy of the nucleus is minimized with the \( \{100\} \) texture because the elastically compliant plane is parallel to the sheet surface, which favours the selection of the \( \{100\} \) oriented nuclei. According to our results (Figs. 1(a), 1(b), 4(a), and 4(b)), the order of texture developed by the \( \gamma \) to \( \alpha \) transformation is \( \{100\} > \{310\} > \{211\} > \{111\} \). Since the \( \{100\} \) orientation develops predominantly, peak intensities of \( \{321\} \), \( \{111\} \), and \( \{211\} \) are so small that the hierarchy among them is rather obscure. In addition, the least observable plane is almost always \( \{110\} \). For Fe at 900°C, the above three biaxial
moduli yield the lowest values for the ⟨100⟩ orientation. Therefore, regardless of the biaxial moduli considered, the ⟨100⟩ texture is expected to develop in the surface layer. Among the three biaxial moduli, the hierarchy of \( M_f \) correlates best with the texture evolution; the order of \( M_f \) is similar to the inverse order of texture developed by the \( γ \) to \( α \) transformation. According to \( M_f \) and \( M_γ \), the ⟨111⟩ texture is expected to be the least observable texture. However, our experimental results show that, although the ⟨111⟩ texture is weak, it is almost always more pronounced than the ⟨110⟩ texture, which is the weakest texture. The inaccuracy in \( M_f \) and \( M_γ \) appears to stem from different stress–strain states of anisotropic ferrite nucleus on anisotropic austenite surface. Neither the stress components nor the strain components of the nucleus can be the same along the surface during the isotropic volume change by the \( γ \) to \( α \) transformation.

Hashimoto et al. also tried to explain their observation of the ⟨100⟩ texture evolution in the manganese containing steel during the \( α \)→\( γ \)→\( α \) transformation utilizing the minimization of elastic strain energy.\(^{[7]}\) Although they considered the elastic strain energy as a controlling factor for developing the ⟨100⟩ texture, there are differences between their and our models. The main differences lie in the consideration of the Kurdjumov–Sachs (K–S) orientation relationship and the concept of the elastic strain energy. Hashimoto et al. assumed that the K–S orientation relationship (OR) operates between ferrite and austenite.

We do not consider any OR because of the high transformation temperatures, the slow cooling rates (≤600°C/h) and the resultant large columnar grains in pure Fe and Fe–1%Si (>500 μm). Considering the experimental conditions, the austenite can undergo complete reconstruction into the massive ferrite, by considerably uncoordinated, diffusive mixing-up of atoms at the transformation interfaces.\(^{[29]}\) In fact, no OR is known to exist in the massive transformations.\(^{[30]}\) The K–S relationship is associated with the transformed structures at low temperature during relatively rapid cooling. Widmanstätten ferrite, acicular ferrite, and bainitic ferrite hold the K–S relationship with austenite matrix. Therefore, the ⟨100⟩ texture evolution at slow heating and cooling rates might not be attributed to the K–S relationship.

In case of the elastic strain energy, we are considering the in-plane stiffness to minimize the elastic strain energy of the surface nucleus, whereas Hashimoto’s model concerns about large elastic work in normal direction of sheet surface. The in-plane stiffness means the easiness of developing two-dimensional ferrite nuclei on the austenite surfaces, whereas the large elastic work in normal direction of sheet surface means the minimization of strain energy by the three-dimensional ferrite particles in austenite matrix. In other words, we are emphasizing the incipient stage of ferrite nucleation on surfaces (stability of ferrite embryos).

According to the Hashimoto’s model, heating rate is essential to develop the ⟨100⟩ texture. However, as shown in our results (Fig. 4(a)), the sample with a fast heating rate develops strong ⟨100⟩ texture. This means that slow heating rate is not a prerequisite for the ⟨100⟩ texture evolution. According to our results, the crucial conditions for the ⟨100⟩ texture evolution by the \( γ \)→\( α \) transformation are clean metal surfaces and relatively slow cooling.

Although many researchers have regarded the surface energy minimization as a major driving force for the orientation selection during the \( γ \) to \( α \) transformation, it does not seem to play an important role in developing the ⟨100⟩ texture. According to this mechanism, surface grains with low surface energies may nucleate and/or grow preferentially to form low surface energy planes predominantly. The surface energies of Fe are ordered as \( γ_{210} > γ_{111} > γ_{310} > γ_{211} > γ_{100} > γ_{110} \) i.e., they increase with surface roughness.\(^{[31]}\) The hierarchy of the surface energies does not match with the order of planes developed by the \( γ \) to \( α \) transformation, ⟨100⟩→⟨310⟩→⟨321⟩→⟨111⟩→⟨211⟩→⟨110⟩. If this mechanism is active, the ⟨110⟩ and ⟨100⟩ planes are expected to develop dominantly after the \( γ \) to \( α \) transformation. However, in the present results, the lowest surface energy plane, ⟨110⟩, disappears almost completely, and is the least observable plane. Since we used very clean annealing atmospheres as well as high purity samples, contamination with the surface active elements such as sulphur, oxygen, tin, or antimony, which are known to alter the hierarchy of the surface energies, is not likely. Furthermore, the degree of surface energy anisotropy is insignificant. The ratio of
the maximum to the minimum surface energy is about 1.1 (2.66 vs. 2.38 J/m²). For these reasons, we believe that the surface energy does not seem to play an important role in developing the (100) texture.

The disappearance of the (100) texture by surface oxidation can be explained in terms of the modification of the inherent elastic anisotropy of metal surface. The present results suggest that the surface oxidation prevents the Fe-base alloys from developing the (100) texture. Since the elastic modulus represents the strength of atomic bonding, if impurity atoms such as oxygen make a strong bonding with surface atoms, this will change the elastic properties of the metal surfaces, thereby leading to a loss of the inherent planar elastic anisotropy. For example, Young’s modulus of β cristobalite (SiO₂) at 912°C (32) and wüstite (FeO) at 25°C (33) are obviously quite different from that of iron. The elastic anisotropy of both materials is not large. The anisotropy parameters are 2.05 for β cristobalite and 0.96 for wüstite, and the ratios of the maximum to the minimum Mₐ are 1.15 for β cristobalite and 1.01 for wüstite. These results suggest that surface chemistry can affect the elastic anisotropy of metal surfaces significantly, and, furthermore, the anisotropy in near surface layers can be utilized as a tool to induce an orientation selection during phase transformation.

Figure 7 shows the magnetic properties of 0.5 mm-thick Fe-1%Si with the cube-on-face texture. The magnetic properties are W₁₅₅₀ (core loss at 1.5 T, 50 Hz)=3.7 W/kg and Bₐ (magnetic flux density at 5 000 A/m)=1.83 T. Compared with randomly oriented materials with similar chemical compositions, the (100) textured materials not only reduce the core losses by more than 30% but also increase the magnetic flux density by about 0.09 T. Special attention should be paid to the superior magnetic flux density. According to Senda et al., (34) the efficiency of an induction motor is improved by low iron losses as well as a high magnetic flux density. The improvement of magnetic flux density together with reducing iron loss by texture control will make a significant contribution to improving the electric motor efficiency.

4. Conclusions

A novel method has been discovered for controlling the crystallographic orientation of body-centred cubic crystals of iron and its alloys using the γ to α phase transformation.

(1) When iron and its alloy sheets with clean metal surfaces undergo the γ to α phase transformation, they develop a strong (100) texture with grain size being larger than the sheet thickness.

(2) The mechanism controlling the (100) orientation obtained is associated with the fact that the cube faces are elastically compliant so that the texture can develop in a manner consistent with minimization of strain energy.

(3) The presence of surface oxide which modifies elastic modulus of metal surface stifles the development of the (100) texture.

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