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Effect of Initial Oriented Columnar Grains on the Texture Evolution and Magnetostriction in Fe–Ga Rolled Sheets

Jiheng Li ^{1,2}, Chao Yuan ³, Qingli Qi ¹, Xiaoqian Bao ¹ and Xuexu Gao ^{1,*}

¹ State Key Laboratory for Advanced Metals and Materials, University of Science and Technology Beijing, Beijing 100083, China; lijh@ustb.edu.cn (J.L.); qiqi@xs.ustb.edu.cn (Q.Q.); bxq118@ustb.edu.cn (X.B.)

² Department of Chemistry-Ångström Laboratory, Uppsala University, Uppsala 75121, Sweden

³ Griem Advanced Materials Co., Ltd., Beijing 100088, China; chaoyuan916@sina.com

* Correspondence: gaoux@skl.ustb.edu.cn; Tel.: +86-10-6233-4343

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Abstract: The effects of initial oriented columnar grains on the texture evolution and magnetostriction in (Fe₈₃Ga₁₇)_{99.9}(NbC)_{0.1} rolled sheets were investigated. The recrystallization texture evolution exhibited the heredity of initial orientations, concerning the formation of cube and Goss textures in the primary recrystallized sheet for the columnar-grained sample. Moreover, the growth advantage of Goss grains was more obvious than that of cube grains during the secondary recrystallization process. Because of the combined effect of this and Nb-rich precipitates as inhibitors, a sharp Goss texture and very large Goss grains were achieved in the secondary recrystallized sheet for the columnar-grained sample. For comparison, the secondary recrystallization in the equiaxed-grained sample was not fully developed although there were Nb-rich precipitates as inhibitors. We think this could be ascribed to the large particle size and premature coarsening of precipitates. Magnetostriction of the secondary recrystallized columnar-grained sheet was up to 232 ppm owing to the ideal Goss texture and quite large grain size. As for the equiaxed-grained sample, the magnetostriction was only 163 ppm in the secondary recrystallized sheet.

Keywords: magnetostriction; Fe–Ga alloy; texture; columnar-grains; rolling

1. Introduction

Magnetostrictive Fe–Ga alloys (Galfenol) have received increasing attention—particularly, a new kind of magnetostrictive smart material for actuator, sensor, and energy harvesting applications [1,2]. These interests stem from the fact that, unlike existing smart material systems, Galfenol is the first kind of material that shows a good combination of magnetostrictive properties and mechanical properties. The addition of Ga increases the magnetostrictive capability of Fe over tenfold up to 400 ppm ($\times 10^{-6}$) along the $\langle 100 \rangle$ direction in single crystal material [3]. Mechanically [4], Fe–Ga alloys are robust, as opposed to materials such as PZT, Ni–Mn–Ga, or Terfenol-D. In addition, Fe–Ga alloys have high permeability [5], a high Curie temperature [6], and highly thermal-stable ferromagnetism [7]. These make Fe–Ga alloys unique.

The high conductivity of Fe–Ga alloy requires its use in the form of thin sheets to avoid eddy current losses when it is used in high frequency. In order to produce thin sheets with reasonable robustness and magnetostrictive properties, efforts have been made to produce the textured sheets by rolling and secondary crystallization processes [8–12]. Up to now, the secondary recrystallization, also named abnormal grain growth (AGG), of Goss-oriented ($\{110\}\langle 001 \rangle$) grains has been considered as the most effective way to achieve the sharp $\langle 001 \rangle$ orientation along the rolling direction in the rolled Fe–Ga

alloy sheets [13–18]. Previous studies have reported the achievement of sharp Goss orientation, by the combined effects of NbC particles as inhibitors and sulfur-induced surface energy [15–17]. Presently, most studies are about the rolling and recrystallization behaviors of the Fe–Ga polycrystalline alloys with equiaxed grains, and mainly focus on the influence of the final annealing process on abnormal grain growth. Recently, we have prepared Goss-oriented Fe–Ga alloy sheets using $\langle 001 \rangle$ oriented columnar-grained alloys for rolling [18]. On one hand, as Fe–Ga polycrystalline alloys exhibit grain boundary embrittlement, the rollability of these alloys with columnar grains can be improved by suppressing the fracture along the transverse boundaries when the long axes of columnar grains are arranged along the rolling direction. On the other hand, the anisotropic feature of columnar grains is different from that of the equiaxed grains with a random texture, and the role of the restriction at the special columnar grain boundaries on orientation formation cannot be ignored during the rolling and recrystallization process. In this work, effects of the initial oriented columnar grains on texture evolution and magnetostriction in the rolled $(\text{Fe}_{83}\text{Ga}_{17})_{99.9}(\text{NbC})_{0.1}$ alloy sheets are investigated. It proves that without a sulfur-induced surface energy effect, the use of initial oriented columnar grains for rolling can improve the secondary recrystallization, and a sharp Goss texture and quite large Goss grains are achieved in the secondary recrystallized alloy sheets, resulting in a high magnetostriction.

2. Materials and Methods

The alloys with nominal composition $(\text{Fe}_{83}\text{Ga}_{17})_{99.9}(\text{NbC})_{0.1}$ were prepared from Fe (99.9%, weight percent), Ga (99.99%, weight percent), and master alloys of Nb–Fe and Fe–C. The columnar-grained rod was produced by directional solidification at a growth rate of $720 \text{ mm} \cdot \text{h}^{-1}$. A detailed description of the directional solidification process could be found in Ref. [19]. The as-cast ingot was prepared by induction melting, and then hot forged to reduce casting defects. The slabs with a thickness of $\sim 18 \text{ mm}$ were cut by electrical discharge machining from the directionally solidified and the hot-forged samples, respectively. The long axes of columnar grains were arranged along the rolling direction when the directionally solidified sample was used for rolling, as shown in Figure 1. The slabs were hot-rolled at 1150°C to $\sim 2.1 \text{ mm}$, followed by warm rolling at 500°C to $\sim 1.1 \text{ mm}$. After an intermediate annealing at 850°C for 5 min, further cold rolling was undertaken to make a final thickness of $\sim 0.3 \text{ mm}$. The as-rolled $(\text{Fe}_{83}\text{Ga}_{17})_{99.9}(\text{NbC})_{0.1}$ sheets, $12 \text{ mm} \times 16 \text{ mm}$ cut by electrical discharge machining, were enclosed in quartz ampoules using 0.3 atm Ar as protecting gas. The sheets enclosed in the ampoules were primarily annealed at 850°C for 6 min. After the primary annealing, samples were rapidly heated from 850 to 900°C at a rate of $10^\circ\text{C}/\text{min}$ in the furnace, and they were then slowly heated from 900 to 1080°C at a controlled rate of $0.25^\circ\text{C}/\text{min}$, and they were finally cooled by air to room temperature without dwell at 1080°C .

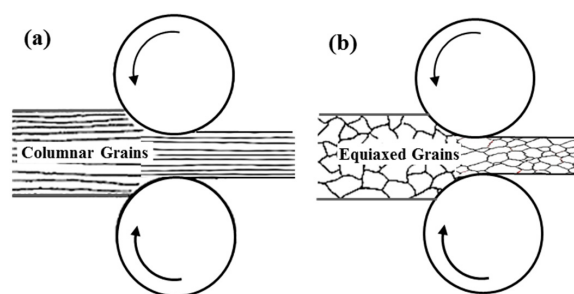


Figure 1. Schematic diagrams of the rolling method: (a) rolling of the columnar-grained sample; (b) rolling of the equiaxed-grained sample.

Microstructures and phases were characterized by optical microscopy (Carl Zeiss AG, Heidenheim, Germany) and X-ray diffraction (XRD) (Rigaku Corporation, Tokyo, Japan) respectively. Precipitates were examined by transmission electron microscopy (TEM) (Technai F30, FEI, Hillsboro, OR, USA), and energy dispersive X-rays spectroscopy (EDS) (Technai F30, FEI, Hillsboro, OR, USA) was used

to identify the composition of precipitates. The texture was analyzed using electron back-scattering diffraction (EBSD). The EBSD was carried out on a SUPRA™ 55 field emission scanning electron microscope (Zeiss Supra 55, Oberkochen, Germany). The EBSD patterns were captured and analyzed to obtain the inverse pole figure (IPF) and the orientation distribution function (ODF). The magnetostriction was measured by strain gauge, and the gauges were positioned along the rolling direction. For the magnetostriction measurement ($\lambda_{//}$ and $\lambda_{\perp}$), a magnetic field parallel and perpendicular to the rolling direction (RD) was applied, respectively.

3. Results and Discussion

Figure 2 shows the microstructures, phases, and orientations of the directionally solidified and as-cast $(\text{Fe}_{83}\text{Ga}_{17})_{99.9}(\text{NbC})_{0.1}$ alloys. In the specimen fabricated by directional solidification, some columnar grains with a width above 1000 μm are distributed homogeneously, as shown in the longitudinal optical photograph (Figure 2a). The grain boundary is parallel to the drawing direction (grain growth direction), because the grain morphology of the directionally solidified specimen is related to the direction of heat dissipation. The XRD pattern of the directionally solidified sample captured from the cross section of the rod is shown in Figure 2b. It demonstrates that the α -Fe phase (A2) is the dominant phase in the directionally solidified sample. The (200) peak dominates the pattern, rather than the (110) peak, indicating the preferred $\langle 100 \rangle$ orientation along the grain growth direction. Moreover, a very weak peak corresponding to NbC at $\sim 40.29^\circ$ is observed. Additionally, an unexpected small peak around 30.7° corresponding to the DO_3 phase (long-range order structure) appears, which could be attributed to the low temperature gradient (about 55 K/cm) and relatively slow cooling rate during the directional solidification process. In order to further detect the orientation information, the EBSD pattern on the cross section of the directionally solidified rod was captured. IPF shows that a strong $\langle 100 \rangle$ orientation was achieved, consistent with the dominant (200) diffraction peak in the XRD pattern, as seen in Figure 2c. By contrast, in Figure 2d, many equiaxed grains can be observed in the as-cast alloy. The main phase of A2 is visible in the XRD pattern of the as-cast alloy, as indicated in Figure 2e. The EBSD pattern of the as-cast alloy was captured on the same surface of plate sample used for XRD measurement. These equiaxed grains are without an obvious preferred orientation although a substantially weak $\langle 110 \rangle$ fiber texture can be seen in Figure 2f. This may be due to the fact that the dominant diffraction peak is usually (110) peak in random oriented Fe–Ga alloys, as shown in Figure 2e. The directionally solidified and the as-cast alloys are called the columnar-grained (CG) and the equiaxed-grained (EG) specimens, respectively.

To analyze the texture evolution during the rolling process, EBSD patterns were captured on the RD–ND (normal direction) section of the rolled sheets. IPF maps for the hot-rolled sheets and the warm-rolled sheets before and after intermediate annealing are shown in Figure 3. In the IPF maps, the red, blue, and green colors represent the crystal directions of $\langle 001 \rangle$, $\langle 111 \rangle$, and $\langle 101 \rangle$ arranged along the RD, respectively. All samples show the through-thickness structures and texture gradients on the lateral face. The $\langle 001 \rangle$ oriented columnar grains in the CG sample are elongated along the rolling direction during the rolling process, as shown in Figure 3a,b. An increase in the area of near green and blue colors indicates that the orientation begins to deviate from $\langle 001 \rangle$. The deformation microstructure consists of the shear-deformed surface regions due to the friction between roll and sheet and a homogeneously deformed center region in the EG sample, as shown in Figure 3d,e. The grains in the surface region are markedly refined by the shearing stress, while the $\langle 110 \rangle$ fiber textured grains (green color shown in Figure 3d,e) are slightly refined under the compressive stress in the center region. In addition, there are many scanning blind spots shown by the white color in the surface region due to the many deformation defects by shearing, as shown in Figure 3a,b,d,e. Figure 3c,f show that the intermediate annealing, at 850 $^\circ\text{C}$ for 5 min, leads to the partial recrystallization of the warm-rolled sheets. The grain sizes are very inhomogeneous in these intermediate annealed sheets. In Figure 3c, a marked rotation from the $\langle 001 \rangle$ to $\langle 111 \rangle$ can be observed on both sides of the intermediate annealed CG sheet (indicated by black arrow) due to the recrystallization in the strong shear deformation region.

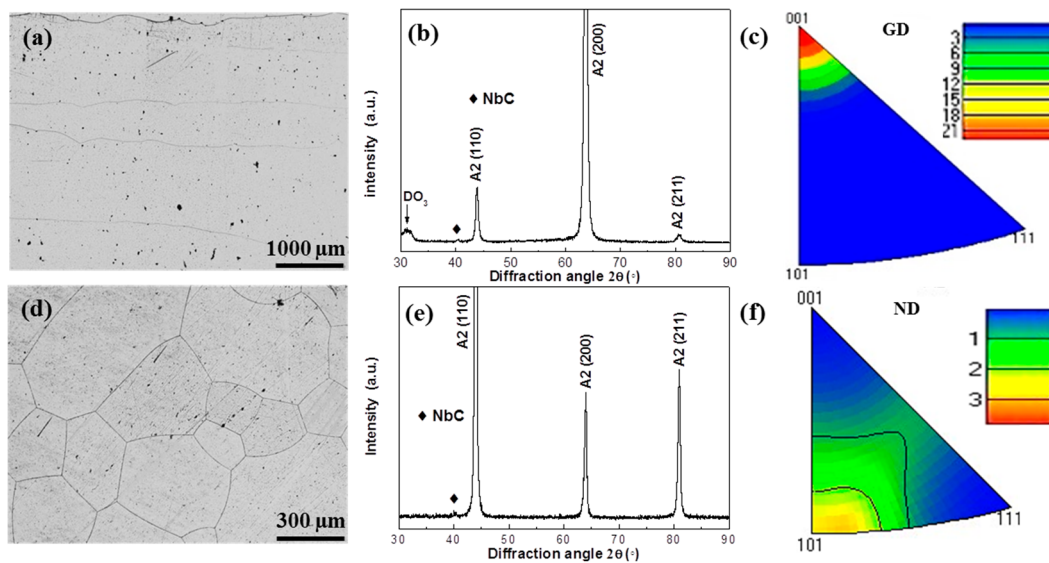


Figure 2. (a) Optical microstructure, (b) XRD pattern, and (c) IPF of the directionally solidified alloy. (d) Optical microstructure, (e) XRD pattern, and (f) IPF of the as-cast alloy. GD: growth direction of the directionally solidified rod; ND: normal direction of the plate sample used for XRD measurement.

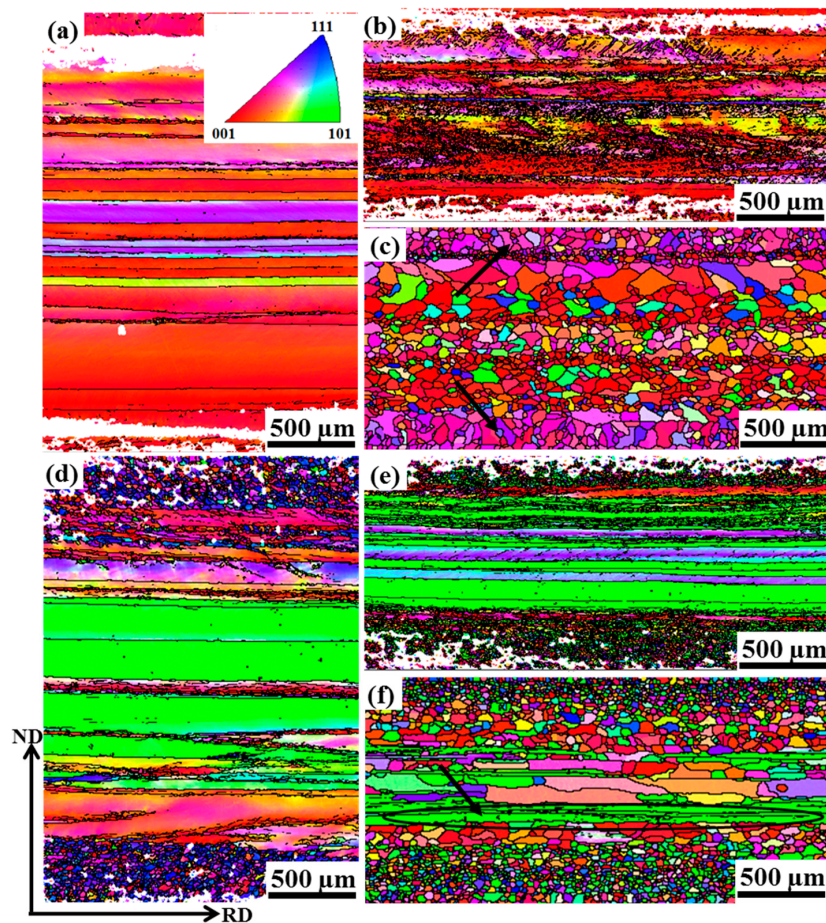


Figure 3. IPF maps of (a) the hot-rolled sheet and warm-rolled sheet (b) before and (c) after intermediate annealing for the CG sample; IPF maps of (d) the hot-rolled sheet and warm-rolled sheet (e) before and (f) after intermediate annealing for the EG sample. Grain boundaries are high angle boundaries ($\omega \geq 15^\circ$).

Figure 4 presents the ODF ($\phi_2 = 45^\circ$) plots corresponding to Figure 3. It shows that the major texture components in the hot and warm deformed CG samples are a strong Goss texture and a weaker near-cube texture $\{100\}\langle 001 \rangle$ (or $\{100\}\langle 021 \rangle$), as seen in Figure 4a,b. In general, the initial texture has less influence on the formation of shear texture components, which is accordant with the study results of Shimizu et al. [20]: that any initial orientation can induce a sharp Goss texture after the hot rolling of a single crystal silicon–iron alloy. The near-cube texture likely comes from the slight rotations of the initial cube grains. The rotation from the $\langle 001 \rangle$ to $\langle 111 \rangle$ on both sides of the intermediate annealed CG sheet, which is reflected in the decrease in intensity of the Goss texture and the appearance of the $\{110\}\langle 113 \rangle$ texture, as shown in Figure 4c. In contrast, the rolling textures are dominated by $\{111\}\langle uvw \rangle$ and $\{100\}\langle 011 \rangle$ (also named by the 45° rotated cube texture) in the hot- and warm-rolled EG sample, as shown in Figure 4d,e. In this case, the change of grain orientation takes place as a consequence of shear on specific favorably oriented crystal planes and directions, and slip preferentially occurs on $\{hkl\}\langle 111 \rangle$ slip systems where $\{hkl\}$ could be $\{110\}$ or $\{112\}$. As is well known, this is common in body-centered cubic (BCC) metals. In addition, the texture of $\{100\}\langle 011 \rangle$, which is a typical component in BCC metals at high reduction, is retained in the warm-rolled EG sample, which is attributed to the fact that the $\{100\}\langle 011 \rangle$ oriented grains do not lead to shear strain during deformation. After an intermediate annealing, orientations of the deformed $\{111\}\langle 112 \rangle$ textured grains change due to the recrystallization in the strong shear deformation region on both sides of the intermediate annealed EG sheet, while the retained strong $\{100\}\langle 011 \rangle$ texture could be attributed to the poor recrystallization behavior of the deformed grains (indicated by the black arrow in Figure 3f), as shown in Figure 4f.

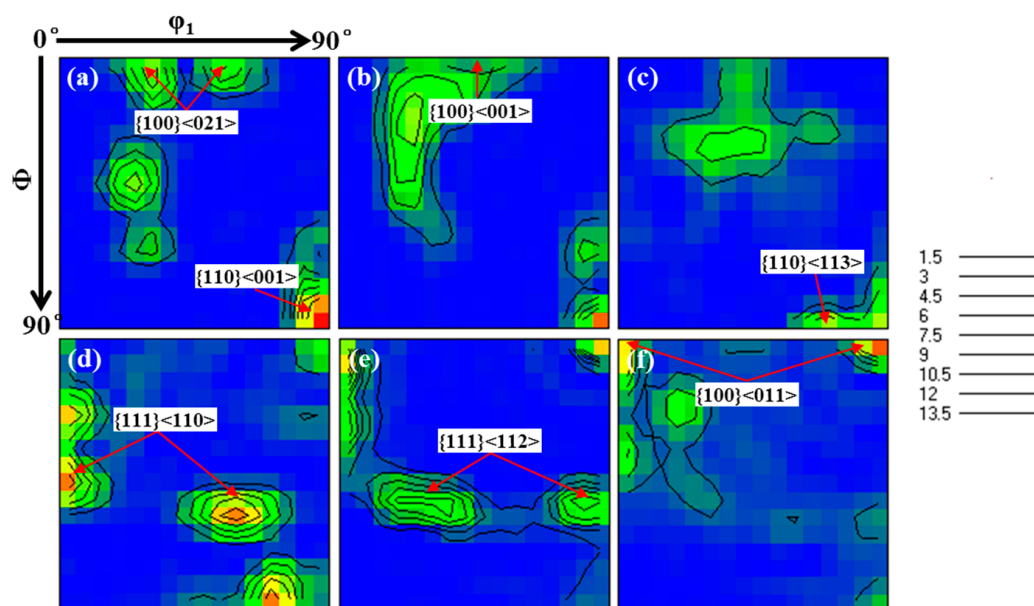


Figure 4. ODF plots at $\phi_2 = 45^\circ$ section of (a) the hot-rolled sheet and warm-rolled sheet (b) before and (c) after intermediate annealing for the CG sample; ODF plots at $\phi_2 = 45^\circ$ section of (d) the hot-rolled sheet and warm-rolled sheet (e) before and (f) after intermediate annealing for the EG sample.

Figure 5 displays the changes of magnetostriction in the CG and EG samples during the rolling process. The observed magnetostriction values are an average of three similarly treated samples, and the error bars show the standard deviation. As for the bulk samples, the magnetostriction of $\lambda_{//}$ is only measured for the as-cut CG and EG slabs, while the magnetostrictions of $\lambda_{//}$ and $\lambda_{\perp}$ are measured for the rolled and annealed sheets. The magnetostrictive value decreases sharply from 215 ppm in the as-cut CG slab to 24 ppm in the cold-rolled sheet, which is mainly ascribed to the deviation of orientation from the $\langle 001 \rangle$ direction and many severely deformed grains. Magnetostriction of the as-cut EG slab, which has no preferred orientation, is relatively small, and that of the rolled sheets

decreases during the rolling process. The change of magnetostriction in the CG samples is markedly larger than in the EG samples, which could be ascribed to the non-uniform deformation of large columnar grains during the rolling process. This is likely another effect of the CG samples' having a different favorable initial texture, which changes upon rolling. Overall, after a heavy rolling reduction, magnetostriction of the cold-rolled CG and EG sheets are both very low.

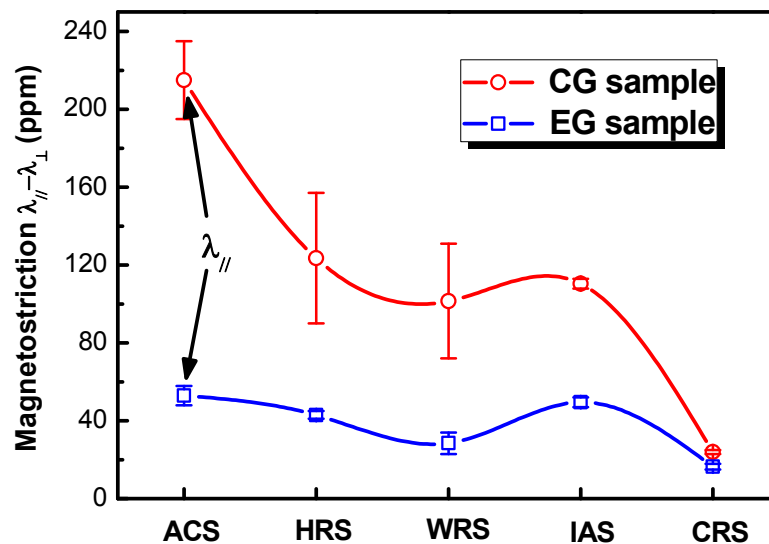


Figure 5. Magnetostriction of the CG and EG samples during rolling process. ACS, HRS, WRS, IAS, and CRS denote the sample of as-cut slabs, hot-rolled sheets, warm-rolled sheets, intermediate annealed sheets, and cold-rolled sheets, respectively.

The warm-rolled sheets with intermediate annealing are further cold-rolled to a final thickness of 0.3 mm, and the primary annealing is carried out at 850 °C for 6 min. The IPF maps and ODF plots of the cold-rolled sheets and the primary recrystallized sheets are shown in Figure 6. The figure shows obvious differences in the cold rolling texture and the primary recrystallization texture between CG and EG samples owing to their different initial orientation and grain morphology. The dominant texture is a sharp $\{223\}\langle 362 \rangle$ (near $\{113\}\langle 361 \rangle$) with minor components of $\{100\}\langle 023 \rangle$, $\{111\}\langle 112 \rangle$, and $\{110\}\langle 001 \rangle$ in the cold-rolled CG sample, as shown in Figure 6a,b. With a heavy rolling reduction, most initial cube grains gradually rotate to $\{223\}\langle 362 \rangle$. The rotation to $\{113\}\langle 361 \rangle$ is a typical path of cube grains during rolling, whereas a part of the initial Goss grains rotates to $\{111\}\langle 112 \rangle$ [21]. In addition, the shear deformation texture of $\{110\}\langle 001 \rangle$ (red color indicated by the black arrow in Figure 6a) remains in the surface and subsurface regions of the cold-rolled CG sample. After the primary recrystallization, in addition to a weak Goss texture, the CG sample shows predominantly a γ -fiber texture ($\{111\}\langle 110 \rangle$ and $\{111\}\langle 112 \rangle$) and a near-cube texture, as seen in Figure 6c,d. Hu [22] has pointed out that recrystallization proceeds easily in the deformed γ -fiber- and cube-textured grains, whereas $\{100\}\langle 011 \rangle$ and $\{100\}\langle 023 \rangle$ grains can hardly recrystallize. As such, the primary recrystallized texture in the CG sample originates from the recrystallization of the deformed γ -fiber, cube, and Goss grains. The deformed $\{100\}\langle 023 \rangle$ grains are so stable in resisting recrystallization that they could only be consumed slowly by other oriented grains.

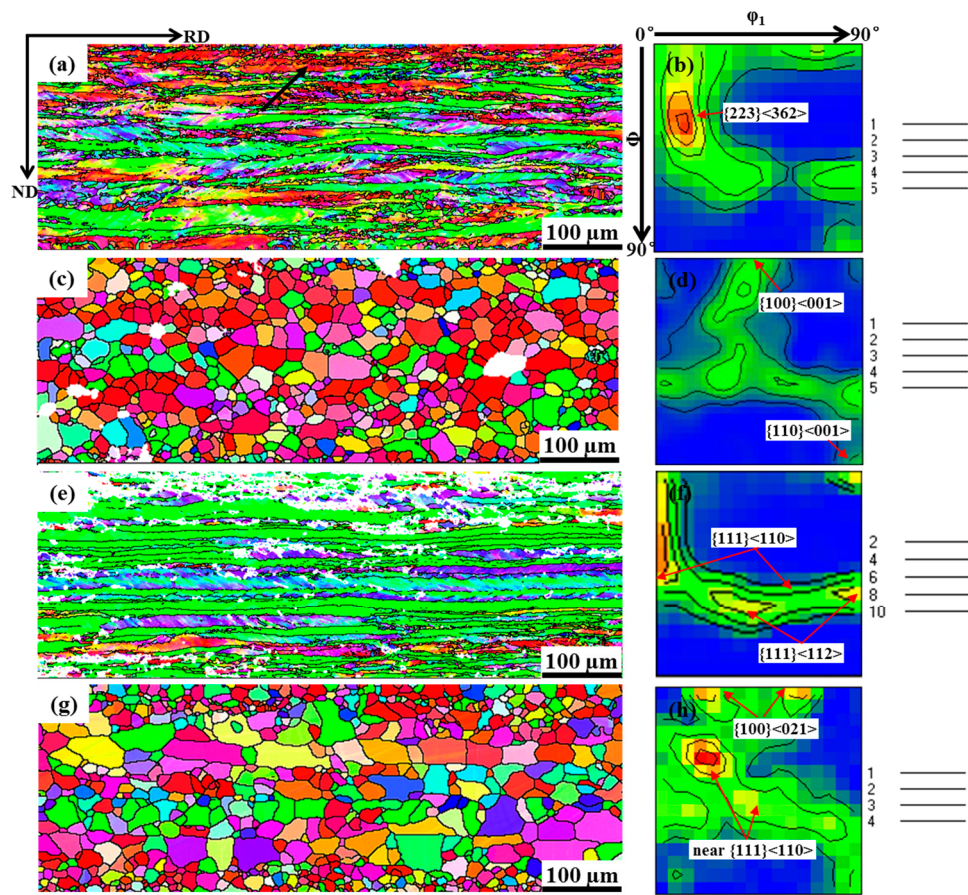


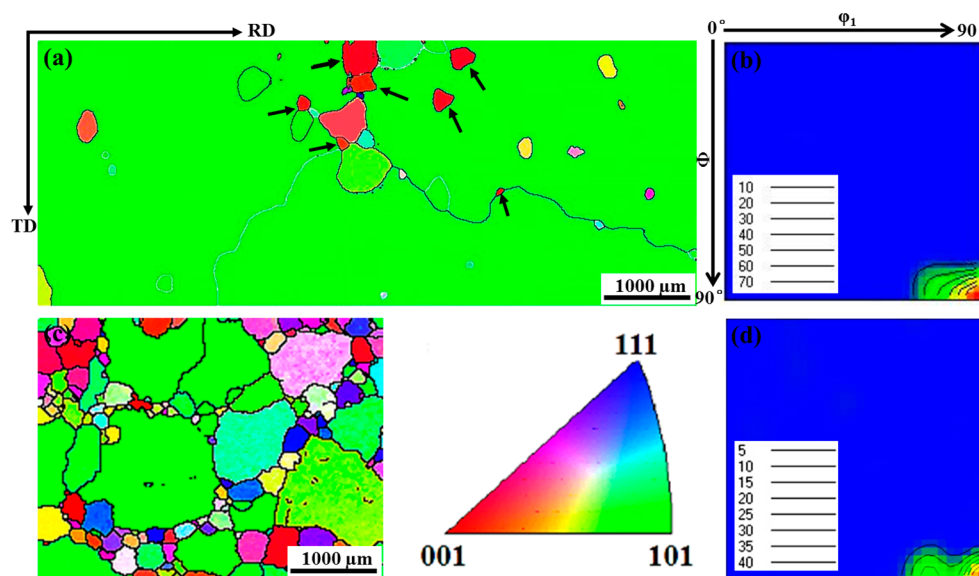
Figure 6. (a) IPF map and (b) ODF plot ($\phi_2 = 45^\circ$) of the cold-rolled sheet for the CG sample; (c) IPF map and (d) ODF plot ($\phi_2 = 45^\circ$) of the primary recrystallized sheet for the CG sample; (e) IPF map and (f) ODF plot ($\phi_2 = 45^\circ$) of the cold-rolled sheet for the EG sample; (g) IPF map; (h) ODF plot ($\phi_2 = 45^\circ$) of the primary recrystallized sheet for the EG sample. Grain boundaries are high angle boundaries ($\omega \geq 15^\circ$).

By contrast, strong γ -fiber and $\{100\}\langle 011 \rangle$ textures are obtained in the cold-rolled EG sample, as shown in Figure 6e,f. Figure 6h indicates that near $\{111\}\langle 110 \rangle$ and $\{100\}\langle 021 \rangle$ textures and weak $\{110\}\langle 001 \rangle$ textures are formed in the primary recrystallized EG sample. For the equiaxed crystals in BCC metals, following the start of three main slip system families of $\{110\}\langle 111 \rangle$, $\{112\}\langle 111 \rangle$, and $\{123\}\langle 111 \rangle$, the crystal orientation tends to $\{100\}\langle 112 \rangle$, $\{111\}\langle 110 \rangle$ or $\{111\}\langle 112 \rangle$, and the α -fiber and γ -fiber textures form easily after recrystallization. Table 1 shows the area fractions of differently oriented grains in the primary recrystallized sheets. It can be seen that, with a deviation from the ideal texture within 20° , the area fractions of $\{100\}\langle 001 \rangle$, $\{100\}\langle 011 \rangle$, $\{111\}\langle 110 \rangle$, and $\{111\}\langle 112 \rangle$ textures in the CG sample are higher than that in the EG sample. Especially, the area fraction of $\{100\}\langle 001 \rangle$ texture with a deviation within 20° is up to 16.8% in the CG sample, which is nearly twice that of in the EG sample. Moreover, the slightly higher area fraction of Goss texture is also visible in the CG sample. A part of the primary Goss-oriented grains may come from the initial cube grains. Because, in addition to the rotation to $\{223\}\langle 362 \rangle$, the cube grains also rotate to Goss orientation by shearing during the rolling process, and the recrystallization grains have the same orientation as the sub-band, which serves as the point of origin of the nucleus in deformed Goss grains. This observation agrees with the results obtained in the rolled single crystal silicon–iron alloy [23]. Overall, the primary recrystallization texture in the CG sample presents the heredity of initial orientations, which is reflected in the formation of cube and Goss textures in the primary recrystallized sheet. This could be explained by the theory of oriented nucleation.

Table 1. Area fractions of different oriented grains in the primary recrystallized sheets ($0^\circ\sim 20^\circ$ is deviation degree from the ideal texture).

Texture Component	CG Sample		EG Sample	
	$0^\circ\sim 15^\circ$	$15^\circ\sim 20^\circ$	$0^\circ\sim 15^\circ$	$15^\circ\sim 20^\circ$
$\{110\}\langle 001\rangle$	3.56	5.19	3.43	3.77
$\{100\}\langle 001\rangle$	6.89	9.91	4.29	5.11
$\{100\}\langle 011\rangle$	2.44	1.91	0.44	1.15
$\{111\}\langle 112\rangle$	8.10	8.80	7.10	8.70
$\{111\}\langle 110\rangle$	7.03	4.27	4.21	4.47

As is well-known, the primary recrystallized microstructure and texture are important for the final formation of a sharp Goss texture by abnormal grain growth. The primary recrystallized sheets go through a continuous heating process from 900 to 1080 °C to induce the abnormal grain growth. IPF maps and ODF plots for the secondary recrystallized sheets are shown in Figure 7. In IPF maps, the red, blue, and green colors represent the crystal directions of $\langle 001\rangle$, $\langle 111\rangle$, and $\langle 101\rangle$ arranged along the ND, respectively. Figure 7a demonstrates that most of the grains in the CG sample are very large, up to several centimeters, although some small grains still exist. We think this is attributed to the abnormal growth of Goss grains. As a result, a sharp Goss texture is achieved in the secondary recrystallized CG sample, as shown in Figure 7b. By contrast, it can be seen from Figure 7c that the size of the $\{110\}$ textured grains is not very large in the secondary recrystallized EG sample, and many grains with other orientations remain. This indicates that, although the abnormal grain growth also takes place in the EG sample during the continuous heating process, the secondary recrystallization is not complete. Correspondingly, Figure 7d displays a texture of $\{110\}\langle 113\rangle$ in addition to a strong Goss texture.

**Figure 7.** (a) IPF map and (b) ODF plot ($\phi_2 = 45^\circ$) of the secondary recrystallized sheet for the CG sample; (c) IPF map and (d) ODF plot ($\phi_2 = 45^\circ$) of the secondary recrystallized sheet for the EG sample. The low angle boundaries ($\omega < 15^\circ$) are shown in white; the high angle boundaries ($\omega \geq 15^\circ$) are shown in black.

In Goss-oriented silicon steels, precipitates as inhibitors are used to inhibit the grain growth of primary recrystallization, and an inhibitor is one of the basic conditions for the occurrence of secondary recrystallization. The abnormal grain growth accompanied by precipitated particles of sizes smaller

than 0.2 μm is commonly observed in grain-oriented silicon steels. In addition, various mechanisms for the development of secondary recrystallization have already been proposed [24–26]. The coincidence site lattice (CSL) model and the high energy grain boundary (HEGB) model are frequently used to quantify the grain boundary characteristics in Goss-textured silicon steels during the abnormal grain growth process. Theoretical interpretations for the development of secondary recrystallization in the oriented silicon steels are also applicable for Fe–Ga alloys, because they have the same BCC structure. Figure 8 shows the TEM image of Nb-rich precipitates in the primary recrystallized EG sample. It can be found that the sizes of most Nb-rich precipitates are larger than 0.2 μm in the primary recrystallized EG sample. Furthermore, Nb-rich precipitates would be further coarsened during the continuous heating process, resulting in a decrease in inhibition force. Therefore, the large particle size and premature coarsening of precipitates would be the reason for the incomplete secondary recrystallization in the EG sample. In addition, there are more Goss-oriented grains in the primary recrystallized CG sample, which could provide more nuclei for the secondary recrystallization. On the other hand, although the orientation of most initial Goss grains rotated to $\{111\}\langle 112 \rangle$ during the rolling process, Goss grains easily increase in size during the recrystallization annealing process due to the high mobility boundaries between them and the deformed matrix. The $\{111\}\langle 112 \rangle$ oriented grains are rotated by about 35° with respect to the $\{110\}\langle 001 \rangle$ orientation, which produces boundaries close to the $\Sigma 9$ coincident orientation. Thus, higher diffusivity can be obtained on the boundaries between $\{111\}\langle 112 \rangle$ and Goss grains [27]. Moreover, Figure 7a suggests that, in the late stage of secondary recrystallization, Goss grains grow competitively with cube grains (shown by black arrows), and the growth advantage of Goss grains is more obvious than that of cube grains. This is consistent with the result obtained in silicon steel by Zhang et al. [28]. Goss grains annex other textured grains, and several adjacent Goss grains then meet and merge into a large Goss grain, resulting in a sharp Goss texture finally. This growth pattern is perhaps the reason that the secondary recrystallization could be quickly completed from the beginning to the end, and very large Goss grains can be obtained. In addition, the heredity of initial oriented columnar grains, which leads to strong cubic and Goss orientations in the primary recrystallized sheet, possibly compensate to some extent for the adverse effect of the insufficient inhibition of inhibitors on the development of secondary recrystallization.

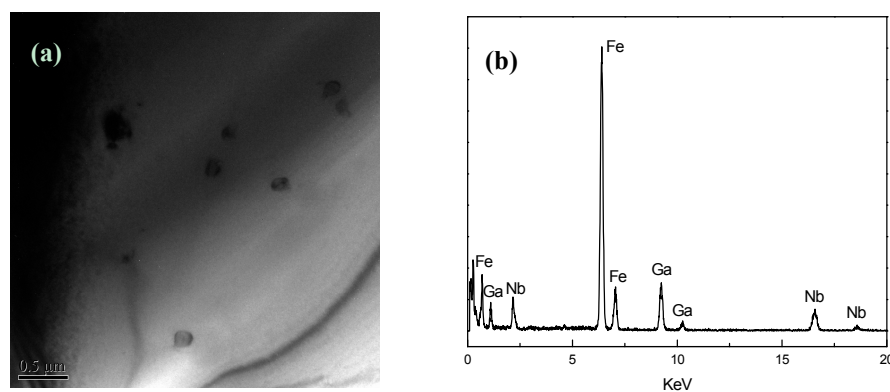


Figure 8. (a) TEM image and (b) EDS profile of Nb-rich precipitates in the primary recrystallized EG sample.

Figure 9 shows magnetostrictive curves of the secondary recrystallized sheets. After the secondary recrystallization annealing, the magnetostriction ($\lambda_{//}-\lambda_{\perp}$) of the CG sample has an increase up to 232 ppm due to the ideal Goss texture and the improvement in the crystal perfection after recrystallization. For comparison, the magnetostriction is only 163 ppm in the secondary recrystallized EG sample. Moreover, there is little hysteresis in the magnetostriction vs. magnetic field, resulting in an almost reversible magnetostriction.

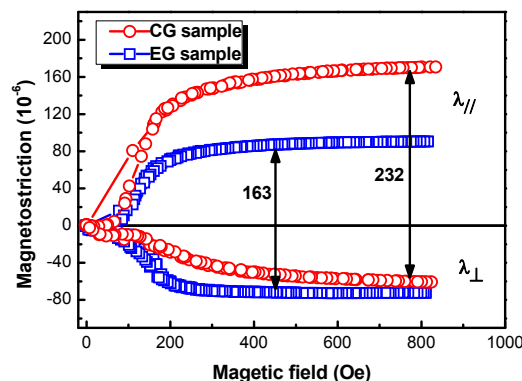


Figure 9. Magnetostrictive curves of the secondary recrystallized sheets.

4. Conclusions

(1) The dominant texture in the cold-rolled CG sample is a sharp $\{223\}\langle 362 \rangle$ with minor components of weak $\{100\}\langle 023 \rangle$, $\{111\}\langle 112 \rangle$, and $\{110\}\langle 001 \rangle$, whereas a strong γ -fiber and $\{100\}\langle 011 \rangle$ textures can be obtained in the cold-rolled EG sample. During the rolling process of the CG sample, most initial cube grains rotate gradually to $\{223\}\langle 362 \rangle$ or to Goss orientation by shearing; meanwhile, the orientation of initial Goss grains rotates to $\{111\}\langle 112 \rangle$. By contrast, there is a conventional rotation path from α -fiber to γ -fiber during the rolling process of the EG sample.

(2) The recrystallization texture evolution in the CG sample presents the heredity of initial orientations, concerning the formation of cube and Goss textures in the primary recrystallized sheet. Moreover, the growth advantage of Goss grains is higher than that of cube grains during the secondary recrystallization process. By the combined effect of this and Nb-rich precipitates as inhibitors, a sharp Goss texture and some very large Goss grains are achieved in the secondary recrystallized CG sample. In contrast, the development of secondary recrystallization in the EG sample was not fully completed, which could be attributed to the large particle size and premature coarsening of precipitates.

(3) Magnetostriction of the secondary recrystallized CG sample is up to 232 ppm due to the ideal Goss texture and very large grain size. As for the EG sample, the magnetostriction is only 163 ppm in the secondary recrystallized sheet.

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Conflicts of Interest: The authors declare no conflict of interest.

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