Original Article

The effect of grain size and initial texture on microstructure, texture, and formability of Nb stabilized ferritic stainless steel manufactured by two-step cold rolling

Daniella Gomes Rodrigues a,∗, Cláudio Moreira de Alcântara b, Tarcísio Reis de Oliveira b, Berenice Mendonça Gonzalez a,∗

a Department of Metallurgical and Materials Engineering, Universidade Federal de Minas Gerais, Av. Antônio Carlos, 6627, Pampulha, Belo Horizonte, Minas Gerais, Brazil
b Aperam South America Research Center, Praça 1° de Maio, 09, Centro, Timóteo, Minas Gerais, Brazil

A R T I C L E   I N F O

Article history:
Received 13 October 2017
Accepted 13 July 2019
Available online 5 August 2019

Keywords:
Ferritic stainless steel
Grain size
Texture
Formability
Two-step cold rolling

A B S T R A C T

The present study investigated the influence of grain size and initial texture on the microstructure, texture, and formability of a niobium-stabilized ferritic stainless steel processed by two-step cold rolling. The characterization of the samples was performed by X-ray diffraction, electron backscatter diffraction, and tensile tests in order to evaluate the formability by the average normal anisotropy coefficient. The results showed that the reduction in grain size from 110 ± 21 to 50 ± 7 μm, in the hot rolled band, was favorable to obtain a more homogenous microstructure and smaller grain size after annealing. Coarse grain size and α-fiber (110)∥RD) favored a strong α-fiber after cold rolling. In addition, γ-fiber (111)∥ND) was more intense in the deformation texture of fine grain size samples. The recrystallization texture was constituted by intense γ-fiber (111)∥ND) and weak γ-fiber (110)∥ND) for both conditions. However, the reduction in grain size increased the γ-fiber fraction. Therefore, the r = 1.92 and δ/α = 8.12 were obtained to the sample originated from the initial fine grain size. The fine grain size and weak texture in the starting materials were effective in developing a high r value in the samples manufactured by two-step cold rolling.

© 2019 The Authors. Published by Elsevier B.V. This is an open access article under the CC BY-NC-ND license (http://creativecommons.org/licenses/by-nc-nd/4.0/).

1. Introduction

Ferritic stainless steels (FSSs) are alloys that mainly consist of Cr and Fe. These alloys have been applied in kitchen cookware, sinks, automobile parts, heaters, and equipment for nitric acid plants due to a combination of corrosion resistance, mechanical properties, and surface quality [1–3]. Niobium is an option to reduce sensitization effects and to stabilize the ferrite phase [4].

Many studies have been conducted to improve the formability of FSSs through the application of several different methods. Accordingly, controlling possible crystallographic orientations enhances the formability. This property can be evaluated by the average normal anisotropy coefficient (r value). A high r value is associated with an intense fraction of γ-fiber (111)∥ND) and a weak γ-fiber (110)∥ND) after anneal-
ing [5,6]. Huh and Engler [7] suggested to perform intermediate annealing during cold rolling. Zhang et al. [8] reported an increase of 25% in the $f$ value through the enhancement of in-grain shear bands formed in the hot-rolled band. Gao et al. [9] reported that warm rolling favored the formation of a nearly uniform $\gamma$-fiber recrystallization texture, which significantly improved the steel formability. Yazama et al. [5] pointed out that the refinement of a hot-rolled steel microstructure is one way to control the formation of the $\gamma$-fiber.

The initial characteristics influence the formation of microstructure and texture during the thermomechanical process. Schlippenbach [10] reported an increase in $\alpha$-fiber ($<110>\parallel$RD) and no change in $\gamma$-fiber fraction after 80% deformation, when the grain size was increased from 22 to 150 $\mu$m in a low carbon steel. However, Lee et al. [11] reported that an increase from 150 to 500 $\mu$m in grain size before cold rolling process led to a slight difference in the intensity of the rolling texture after 87% thickness reduction of a non-oriented electrical steel sheet.

Thus, the present study investigated the effect of grain size and initial texture on the microstructure, texture, and formability of a Nb-stabilized ferritic stainless steel manufactured by the two-step cold rolling process.

### 2. Materials and methods

In this study, fine and coarse grain hot-rolled samples of a ferritic stainless steel containing 0.35% of Nb were supplied by Aperam South America. The chemical composition of the steel under investigation is 0.021% C, 0.024% N, 16.0% Cr, 0.35% Nb, 0.30% Si, 0.22% Ni, and 0.20% Mn. These samples were manufactured in an industrial line until being annealed at 970 °C after hot rolling.

As-received samples, with a thickness of 6.0 mm, were cold rolled in a laboratory rolling mill up to 3.0 mm, corresponding to a 50% reduction in thickness. Following this first reduction, the samples were annealed at 880 °C, with a heating rate of 1.8 °C/s and 24 s of soaking time. Subsequently, the samples were cold rolled up to 0.6 mm, corresponding to an 80% reduction in thickness. Finally, the samples were also annealed at 880 °C, with a heating rate of 23.5 °C/s and 24 s of soaking time. The coarse and initial fine grain size samples, with different processing steps, were recorded as hot-rolled samples (CHR and FHR, respectively). The sample codes are listed in Table 1.

The samples were characterized using a Leica DMRM light optical microscopy, X-ray diffraction (XRD) measurement, and electron backscatter diffraction (EBSD). Microstructural characterization and crystallographic texture were carried out in longitudinal sections using EBSD. The samples for microstructural characterization were prepared in a standard manner. For EBSD, the final polishing was performed with a colloidal silica suspension. The samples for X-ray diffraction measurement were mechanically polished and etched with a solution of HNO$_3$ (12%) and HF (3%) at ~55 °C. The grain size was measured using the linear intercept method, in accordance with the standards established by ASTM E112/13 [12].

The crystallographic textures, determined through X-ray diffraction measurement, were done using a Philips X’Pert PRO MPD texture diffractometer. From the incomplete pole figures of (110), (200), and (211), the orientation distribution functions (ODF) were calculated by the series expansion method, according to Bunge ($l_{\text{max}} = 22$) [13]. The EBSD scans were performed in a Philips XL-30 SEM with a LaB$_6$ filament and operated at 20 kV. The range of step size varied from 4 to 8 $\mu$m. The EBSD data were processed using a TSL/EDAX Analysis™ software, considering a confidence index (CI) of >0.095. The clean up procedure applied in the raw data was performed using neighbor CI correlation.

The stored energy after deformation was calculated according to the hardness results, using Eq. (1) [14], where $M = 3.05$ is the Taylor factor, $a = 0.3$ is a constant, $G = 80$ GPa is the average shear modulus for ferritic stainless steel [15], and $b = 2.48 \times 10^{-10}$ m is the Burgers vector [14]. Vickers microhardness was measured using a Leitz–Wetzlar microhardness tester. The average values were defined from ten measurements with a dwell time of 10 s and a load of 100 gf.

$$E_d = \rho^*E = \frac{H_T - H_0}{MaGb} \left( \frac{Gb^2}{2} \right) = \frac{1}{18G} \left( \frac{H_T - H_0}{Ma} \right)$$

The $\bar{r}$ and $\Delta r$ values were determined through tensile tests in order to evaluate the effect of the initial grain size and texture on formability. Specimens were cut from the final annealed sheets at the angles of 0°, 45°, and 90° respecting the rolling direction (RD). The specimens with a gauge length of 50 mm and a width of 20 mm were strained until 15%. The $\bar{r}$ and $\Delta r$ values were calculated according to the ASTM E517 standard using the Eq. (2) and Eq. (3) [16].

$$\bar{r} = \frac{(r_0 + 2r_{45} + r_{90})}{4}$$

$$\Delta r = \frac{(r_0 - 2r_{45} + r_{90})}{2}$$

### 3. Results

#### 3.1. Microstructures throughout the process

The microstructure developed throughout the thermomechanical process of Nb ferritic stainless steel is shown in Fig. 1, where ND is the normal direction and RD is the rolling direction. From Fig. 1a and b, it was possible to observe the dif-

<table>
<thead>
<tr>
<th>Table 1 – Processing details and the corresponding sample codes.</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Sample code</strong></td>
</tr>
<tr>
<td>CHR</td>
</tr>
<tr>
<td>FHR</td>
</tr>
<tr>
<td>C-50%CR</td>
</tr>
<tr>
<td>F-50%CR</td>
</tr>
<tr>
<td>C-50%CRA</td>
</tr>
<tr>
<td>F-50%CRA</td>
</tr>
<tr>
<td>C-80%CR</td>
</tr>
<tr>
<td>F-80%CR</td>
</tr>
<tr>
<td>C-80%CRA</td>
</tr>
<tr>
<td>F-80%CRA</td>
</tr>
</tbody>
</table>
ference in the initial grain size, with CHR and FHR presenting average grain sizes of 110 ± 21 μm and 50 ± 7 μm, respectively. After 50% reduction (Fig. 1c and d), the microstructures were formed by the joining of flattened and elongated grains along the rolling direction. In addition, in-grain shear bands were observed in some grains with an inclination of approximately 45° with respect to RD. After the first annealing process (Fig. 1e and f), the samples were fully recrystallized with a more homogenous grain size distribution in the F-50%CRA sample (Fig. 1f). In the C-50%CRA
Fig. 2 – Grain size distribution of the annealed samples. a) After the first annealing. b) After the second annealing.

Table 2 – Stored energy after cold rolling.

<table>
<thead>
<tr>
<th>Sample</th>
<th>Hardness (HV)</th>
<th>F (m^−1)</th>
<th>E_d (MJ/m^3)</th>
</tr>
</thead>
<tbody>
<tr>
<td>C-50%CR</td>
<td>248 ± 5</td>
<td>3.19</td>
<td>0.78</td>
</tr>
<tr>
<td>F-50%CR</td>
<td>269 ± 3</td>
<td>4.41</td>
<td>1.07</td>
</tr>
<tr>
<td>C-80%CR</td>
<td>293 ± 7</td>
<td>6.85</td>
<td>1.66</td>
</tr>
<tr>
<td>F-80%CR</td>
<td>299 ± 4</td>
<td>6.97</td>
<td>1.69</td>
</tr>
</tbody>
</table>

(Fig. 1e), the microstructure was formed by fine (red arrow) and coarse (blue arrow) grain clusters. The C-50%CRA and F-50%CRA average grain sizes were 28 ± 2 µm and 22 ± 2 µm, respectively. Following the second cold rolling (Fig. 1g and h), the microstructure was formed by the joining of more flattened and elongated grains along the rolling direction. After the final annealing (Fig. 1i and j), both samples were fully recrystallized, presenting a similar microstructure. The cluster formation appeared after the first annealing (Fig. 1e), but it disappeared after the second annealing in the C-80%CRA sample (Fig. 1i). After the second annealing, the average grain sizes of C-80%CRA and F-80%CRA samples were 13 ± 1 µm and 10 ± 1 µm, respectively.

The grain size distributions after annealing are shown in Fig. 2. Both distributions showed similar tendencies. The samples originated from coarse grains showed a larger proportional area with the coarse grains. By contrast, fine grain size was more present in the F-50%CRA and F-80%CRA samples.

The density of dislocations and the stored energy after the first and second cold rolling are shown in Table 2. As can be observed, after the thickness reduction, the density of dislocations and the stored energy of the samples coming from FHR proved to be larger than those from CHR.

3.2. Texture throughout the process

3.2.1. Starting materials

Fig. 3 shows the inverse pole figures (IPF) through-thickness and the orientation distribution function (ODF) at φ_2 = 45° section of the CHR and FHR samples. The black boundaries in IPF indicate misorientations of ω > 15° between adjacent grains. The orientation maps, represented in Fig. 3a and b, show that the texture is different from the surface and central layers in both samples. Up to approximately ¼ of the thickness, the texture is characterized by the orientation {110}<uvw>, whereas in the center region, it is characterized by {100}<uvw> and {112}<uvw>, in both samples. It can be observed from φ_2 = 45° ODF (Fig. 3c and d) that the global texture was characterized by α-fiber with a high intensity between the {001}(110) and {112}(110) components and a {110}(001) Goss component in both samples. However, the α-fiber in the CHR sample was more intense than in the FHR sample up to ψ ~ 50°. The maximum of f(g) ~ 4.8 in the CHR sample was observed for the orientation {001}(560) at (5°, 0°, 45°). In the FHR sample, the maximum of f(g) ~ 2.3 was observed for the shifted orientations of α-fiber ideal.

3.2.2. Sample after 50% cold rolling and annealing

The texture obtained from XRD measurement on the surface and central region after 50% cold rolling is shown in Fig. 4. From Fig. 4a and b, it can be observed that, in the C-50%CR sample, the texture in the central region was stronger than that on the surface. In the central region, the texture was dominated by an intense α-fiber, with a peak at {001}(110) (f(g) ~ 17.9), and a weak γ-fiber. On the surface, the α-fiber peak occurred at {112}(110) (f(g) ~ 7.7), while the γ-fiber was slightly more intense than in the central region. Similarly, in the F-50%CR sample, the α-fiber in the central region was add intensity than on the surface, with a peak at {001}(110) (f(g) ~ 9.6). On the other hand, the γ-fiber was more developed in the F-50%CR sample than in the C-50%CR sample, especially in the (111) (112) components.

Fig. 5 shows the texture determined by EBSD after 50% reduction and annealing of C-50%CRA and F-50%CRA. In the reverse pole figure map (Fig. 5a and b), the texture is heterogeneous across the entire thickness, and the {011} <uvw> grains are distributed throughout it. From the C-50%CRA sample (Fig. 5a), various fine grains adjacent to coarse grains with random orientation can be observed, as indicated by the black arrows. By contrast, in the F-50%CRA sample (Fig. 5b), grain size and texture distributions were more uniform throughout the thickness. The global texture showed a remarkable difference in recrystallization texture, as illustrated in Fig. 5c and d. In the C-50%CRA sample (Fig. 5c), it can be noted that the α-fiber and Goss component were developed with a maximum of f(g) ~ 1.7 in the {011} (001) Goss component. On the other hand, in the F-50%CRA sample (Fig. 5d), the γ-fiber and Goss component proved to be more developed. A maximum of f(g) ~ 2.0 was observed for the orientation {554}(225) at (90°, 60°, 45°).
3.2.3. Sample after 80% cold rolling and annealing

Fig. 6 shows the deformation textures determined by XRD measurement on the surface and central regions after 80% reduction. The C-80%CR sample (Fig. 6a and b) presented a stronger α-fiber in the central region with a high intensity of {001}<110> at (0°, 0°, 45°) and (223)<110> at (0°, 40°, 45°). The γ-fiber was slightly more intense in the surface region with a peak of (111)<011> at (60°, 54.7°, 45°). In the F-80%CR sample (Fig. 6c and d), the α-fiber and γ-fiber were well developed in both regions. The peak of α-fiber occurred at (223)<011> at (0°, 40°, 45°), and for the γ-fiber, at (111)<011> at (60°, 54.7°, 45°).

The recrystallized texture, determined by EBSD, is shown in Fig. 7 after 80% reduction and annealing. Fig. 7a and b shows the inverse pole figure maps of the C-80%CRA and F-80%CRA samples, respectively. The texture is heterogeneous throughout the thickness of both samples, but it is more heterogeneous in the C-80%CRA sample (Fig. 7a) than in the F-80%CRA one (Fig. 7b). The texture gradient can be expressed by a uniformity factor (UF) obtained from the EBSD data. The value of UF = 0 describes a perfect homogeneous distribution, while UF = 1 describes a heterogeneous distribution. The C-80%CRA sample (Fig. 7a) displayed a UF = 0.032, whereas the F-80%CRA sample (Fig. 7b) showed a UF = 0.024, both of which obtained through the thickness. At φ2 = 45° section (Fig. 7c and d), the γ-fiber is not completely uniform in both samples. However, its intensity in the F-80%CRA sample proved to be stronger, with a peak at (111) (121) at (30°, 54.7°, 45°). The maximum of f(ψg) was observed at (554) (225) at (90°, 60°, 45°) (Fig. 7d). Moreover, in the C-80%CRA sample (Fig. 7c), it could be observed that some α-fiber components remained after the annealing process.

Fig. 8 shows the results of grain boundary distribution (mesotexture) after the final annealing. The mesotexture differentiates the misorientations existent between grains, and the measurements are sensitive to the step size, especially in the subgrains [17]. Therefore, subgrain structure does not
Fig. 4 – Texture from XRD measurement at different regions after a 50% reduction. a) C-50%CR surface; b) C-50%CR center; c) F-50%CR surface; d) F-50%CR center.

Fig. 5 – Inverse pole figure map after 50% cold rolling and annealing. a) C-50%CRA; b) F-50%CRA; c) $\phi_2 = 45^\circ$ C-50%CRA; d) $\phi_2 = 45^\circ$ ODF F-50%CRA.

appear when larger step sizes are used. It can be noted from Fig. 8 that the C-80%CRA sample showed 45% of low angle grain boundaries (LAGBs), in which 30% corresponded to an angle comprehended between $2^\circ < \omega < 5^\circ$. By contrast, the F-80%CRA sample showed a high fraction of high angle grain boundaries (HAGBs).

An enlarged area is shown in Fig. 9 to closely examine the LAGBs of the C-80%CRA sample. The LAGBs with $2^\circ < \omega < 5^\circ$ are
Fig. 6 – Texture from XRD measurement at different regions after second cold rolling. a) C-80%CR surface; b) C-80%CR center; c) F-80%CR surface; d) F-80%CR center.

Fig. 7 – Inverse pole figure map after the second cold rolling and annealing. a) C-80%CRA; b) F-80%CRA; c) $\phi_2 = 45^\circ$ ODF C-80%CRA; d) $\phi_2 = 45^\circ$ ODF F-80%CRA.

indicated by white lines. This figure shows the IPF with the projections of normal (Fig. 9a) and rolling (Fig. 9b) directions. Both projections were performed to better visualize the misorientations inside the grains. It can be noted that the LAGBs appear mainly in $\gamma$-fiber grains. Furthermore, the LAGBs are present between two regions with the same plane, but different direction. The other observations include LAGBs inside grains with the same plane and direction.

The misorientation inside the grains was evaluated by grain orientation spread (GOS). Fig. 10 exhibits the GOS distribution for the C-80%CRA and F-80%CRA samples. The GOS distribution was very different in both samples. In the F-80%CRA sample, most of the grains showed a GOS $<1^\circ$. By contrast, the C-80%CRA sample displayed a GOS up to $4^\circ$ and few areas with GOS $<1^\circ$. The average GOS values then obtained were $1.40^\circ$ and $0.32^\circ$ for C-80%CRA and F-80%CRA, respectively.

3.3. Texture, average normal and planar anisotropy coefficient

Fig. 11 shows the relationship between $\bar{t}$ values, $\Delta t$ values, $\gamma$-fiber, and $\theta$-fiber for the final annealed samples. The F-80%CRA exhibited a $\bar{t}$ value notably higher than the one found in the
Fig. 8 – Grain boundary distribution after the second annealing.

C-80%CRA sample, while the $\Delta r$ value was approximately the same in both samples. Furthermore, the F-80%CRA sample showed a predominance of $\gamma$-fiber. It is important to note that both samples displayed a similar $\delta$-fiber fraction. Hence, the highest $\gamma/\delta$ ratio was obtained for the F-80%CRA, as indicated in Fig. 11. The grain size (GS), $\gamma$ and $\delta$-fiber volume fractions, important orientations of $\gamma$-fiber, and normal ($\bar{r}$) value and planar ($\Delta r$) anisotropy coefficients are presented in Table 3.

4. Discussion

The microstructure formed after the cold rolling process can be significantly affected by the initial features before deformation. From the deformed microstructure (Fig. 1c, d, g, and h), it could be observed that the spacing between the strained grains was lower in the samples from FHR than in those from CHR after cold rolling. In both microstructures, the boundaries tend to arrange themselves almost parallel to the cold rolling axis. As a result, pancake-shaped grains are formed, which are typical after cold rolling, leading to an increase in the boundary area [18]. The in-grain shear bands, presented in some grains, were frequently formed in $\gamma$-fiber grains [19,20]. This deformation heterogeneity produces microstructural fragmentation and strain located within the grain [21]. Campos et al. [22] reported evidence of strong strain inhomogeneities in larger grain size ($500 \mu m$). According to the authors, some grains showed a larger number of in-grain shear bands, while others had none.

The influence of grain size before cold rolling on the stored energy of deformation can be seen in the results projected in

Fig. 9 – Enlarged area of the C-80%CRA sample. a) Projection of the normal direction. b) Projection of the rolling direction.

Fig. 10 – Grain orientation spread distribution after the second annealing.
The stored energy showed a difference mainly after the first cold rolling. During deformation, the grain boundaries make the slip processes that occur in the grains difficult. The reduction of the previous grain size increases the grain boundary constraints and the uniformity of deformation [23]. According to Humphreys [18], the stored energy tends to increase with a decrease in grain size resulted from low and medium deformations. Furthermore, the stored energy associated with this increase in grain boundary area represents an important part of the stored energy, which will be higher for fine grains and large strains [18].

The recrystallized grain size and uniformity of the microstructure can be linked to the stored energy after cold rolling. The grain size depends on the nucleation and growth rates [23]. The literature demonstrates that high stored energy increases the nucleation rate more than the growth rate [23]. According to the experimental results shown in Table 2, it could be observed that the stored energy after the first cold rolling was larger in the F-50% CR sample than in the C-50% CR one. Thus, the nucleation rate’s probability was higher and more uniform in the matrix for the F-50% CRA sample. This condition is consistent with the microstructural results shown in Figs. 1e and f.

Landgraf et al. [24] reported the formation of grain clusters in silicon steels, which showed the same orientation after recrystallization. In their study, the sample with an initial grain size of 500 μm was cold rolled until reaching a deformation of ε = 2.1. According to the authors, clusters are formed by the nucleation of various nuclei in the same deformed grain. The recrystallization in the large deformed grains occurs independently in each grain.

The difference among the stored energies was not accentuated after the second cold rolling. The recrystallized microstructure showed a difference in the distribution and average grain size (Fig. 11 and j). Furthermore, a difference was also observed in the distribution of grain boundary fractions and in the misorientation within the grains. The C-80% CRA sample revealed a fraction of 45% of LAGBs after the second annealing, while the F-80% CRA exhibited only 17.5% (see Fig. 8). Fig. 9 demonstrated that this high LAGBs fraction occurred due to the nucleation of adjacent grains with very close orientation. This suggests that the distribution of nuclei during the onset of recrystallization was different in the samples. In other words, in the F-80% CRA sample, the distribution of grains during recrystallization was possibly more random than the one that occurred in the C-80% CRA sample. The high fraction of LAGBs led to the enhancement of misorientation in the C-80% CRA sample’s grains.

From the results of deformation texture after cold rolling, it was clear that the development of deformation texture was influenced by the initial features. As shown in Figs. 4 and 6, after both cold rolling processes, the α-fiber was significantly more intense in the CHR sample, reaching a peak at (001) (110) component. By contrast, the development of γ-fiber was slightly more intense in the FHR sample. An important detail was the reduction of α-fiber and the increase of γ-fiber after the second cold rolling process in the CHR sample. It is worth noting that the difference in average grain size before deformation was approximately 60 μm and 6 μm for the first and second cold rolling processes, respectively. Different initial grain size results in different constraints of grain boundaries to the plastic deformation; hence, it can influence to the grains rotation. The rotation during deformation occurs more easily in the coarse-grained sample. According to the literature, the coarse grains have a tendency to deform heterogeneously through deformation banding, while for the fine grains, the deformation is more homogenous, and the constraints of grain boundary are much stronger [18,25]. Therefore, the development of deformation texture shows a close relation to the mode deformation of each grain.

Lee et al. [11] studied the effect of grain size in a non-oriented electrical steel sheet, using two samples with initial grain sizes of 150 μm and 500 μm, with an almost random initial texture. The samples were cold rolled with 87% reduction and annealed at 700 °C, 800 °C, and 900 °C for 5 min. According to the authors, the cold-rolled textures consisted of α-fiber and γ-fiber with similar intensities; therefore, the initial grain size did not influence the deformation texture. In contrast, Schlippenbach et al. [10] showed an increase of α-fiber and a slight γ-fiber fraction when the initial grain size was enhanced from 22 to 150 μm in a low carbon steel.

According to Inagaki [25], the rolling texture would not be largely influenced by the grain size when it is smaller than 50 μm, due to a few changes in the deformation mode. The influence of grain size consists of the fact that the deformation mode of each grain changes when the grain size is approximately 65 μm. Hence, the effect of grain size is not simple. Moreover, the effect of the initial texture should also

<table>
<thead>
<tr>
<th>Table 3 - Grain size (GS), texture, and anisotropy coefficient.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Code</td>
</tr>
<tr>
<td>-----------------------</td>
</tr>
<tr>
<td>C-80% CRA</td>
</tr>
<tr>
<td>F-80% CRA</td>
</tr>
</tbody>
</table>
be considered in the development of the deformation texture.

From the initial texture presented in Fig. 3, it was observed that, after the first deformation, the (011)-fiber disappeared, and the α-fiber and γ-fiber increased (Fig. 4). During the cold rolling process, the (011)-(uvw) fiber is unstable under plain strain deformation, and it tends to rotate toward the α-fiber and γ-fiber, favoring the sharpening of both fibers in the cold-rolled sample [7,26–28], as shown in Fig. 4. After the second cold rolling, the intensity of α-fiber in the C-80%CR sample (Fig. 5a and b) was notably stronger than that in the F-80%CR sample (Fig. 5c and d). The presence of α-fiber in the previous sample (Fig. 5c), as well as after the cold rolling process, intensified its intensity in the center region (Fig. 6b). Another contribution is that the grains with random orientation had a tendency to rotate to the α-fiber and γ-fiber, mainly to the (111)-<011> component at (60°, 54.7°, 45°) with 80% reduction [7]. On the other hand, the presence of the (111)-<112> component in Fig. 5d contributed to the high intensity of the γ-fiber after cold rolling (Fig. 6c and d). The grains (111)-<112> in poly-crys-tals are unstable and tend to rotate toward (111)-<110> [25].

After the first annealing, the recrystallization texture in the samples was very different. In the C-50% CRA sample (Fig. 5a), the α-fiber and the (011)-<001> Goss component were developed. The α-fiber present after annealing (Fig. 5c) could be retained from the cold-rolled sample (Fig. 4b). During cold rolling, the α-fiber grains stored low energy [7,18,29,30]. Hence, during annealing, the recrystallization rate developed in these grains is low, so this orientation may be retained or consumed by other orientations in the recrystallization process. By contrast, in the F-50% CRA sample (Fig. 5b), the γ-fiber and (554)-<225> component were dominant. The γ-fiber fraction tends to increase with the decreasing in the grain size of the hot-rolled band sample, due to the lower probability of a shear band development during cold rolling [5,29]. Furthermore, the formation of γ-fiber can be attributed to the grain boundary constraints [25]. Additionally, the (111)-(112) recrys-tallized grains, developed in Fig. 5b, nucleate in the (111)-<011> deformed grains or in boundary grains in ferritic stainless steel [9].

The γ-fiber was strengthened mainly in the F-80% CRA after the second annealing, as shown in Fig. 7. According to the literature, the recrystallization texture depends on the cold-rolled texture and microstructure [31]. The development of the strong γ-fiber, with a peak at (111)-(112) (Fig. 7d), can be attributed to the high intensity of the γ-fiber developed after the second cold rolling process (Fig. 6c and d), mainly in the (111)-<110> component. As mentioned above, the (111)-(112) recrystallized grains are formed by preferential nucleation at (111)-<110> deformed grains [9]. Furthermore, the nucleation of γ-fiber grain occurs adjacent to the previous high angle grain boundaries. The combination of fine grain, γ-fiber, and weak α-fiber (Fig. 5b), before cold rolling, was favorable to the development of the stronger γ-fiber and proved to be more homogeneous throughout the thickness (Fig. 7b).

Wang et al. [32] reported that the strong γ-fiber in the deformation texture is favorable to obtain an ideal γ-fiber after annealing, and with the combination of a strong α-fiber and a weak γ-fiber, a heterogeneous γ-fiber is thereby obtained. On the other hand, in the C-80% CRA sample (Fig. 6a), the γ-fiber showed a slight displacement of the ideal γ-fiber, and some α-fiber components remained after the second annealing. The α-fiber components between (001)-(011) and (112)-(011) have low stored energy, hence, low recrystallization rate during annealing [7,33]. Therefore, these components tend to recover and prevail in the steel or can be consumed during recrystallization. From Fig. 7c, it can be observed that some α-fiber orientations were retained from the starting materials (Fig. 3b) throughout the thermomechanical process.

Lee et al. [34] reported that the occurrence of plastic anisotropy in steel is correlated to the crystallographic texture. Regarding the i value exhibited in Fig. 11, it was observed that the i value = 1.92 was obtained for the F-80% CRA sample, with 59% of γ-fiber and 7% of θ-fiber. The planar anisotropy was low and positive for both samples. According to Du et al. [35], high i value is obtained for materials which possess a strong γ-fiber and a weak θ-fiber. Hence, the γ/θ ratio is a good parameter to evaluate the deep drawability of the steel. The γ-fiber grains have a high i value and a small planar anisotropy, according to the results calculated using the relaxed constraint method of crystal plasticity [29]. For a good deep drawability, a high i value, a minimum |Δr|, and the (111)-(110) and (111)-(112) components are favorable [29,34].

Gao et al. [9] cited that the presence of non-uniform γ-fiber in the recrystallization texture can deteriorate the formability of ferritic stainless steels. They reported that the warm rolling process improves the uniform γ-fiber recrystallization texture, thus enhancing its formability in the final sheet. Therefore, the i value result obtained in the F-80% CRA sample combines a strong γ-fiber, mainly (111)-(112), and a weak θ-fiber.

5. Conclusions

The grain size and initial texture exert a direct effect on the development of microstructure, texture, and formability of a ferritic stainless steel manufactured by two-step cold rolling. The main results can be summarized as follows:

i. The present study demonstrated that the difference in the initial grain size affected both the final microstructure and the recrystallized grain size after annealing. The influence was more pronounced after the first cold rolling process and annealing. The initial coarse grain size produced a heterogeneous microstructure formed by fine grain cluster adjacent to coarse grain cluster. On the other hand, the initial fine grain size formed a homogenous microstructure. These results point to differences in nucleation rate due to the effect of the previous grain size before the cold rolling process.

ii. The deformation texture was mainly constituted by a strong α-fiber and a weak γ-fiber after both cold rolling process. However, there was a remarkable difference in the intensity of the α-fiber between the samples. The CHR samples showed an α-fiber more intense than FHR ones. This difference was associated with the initial texture present in the hot-rolled band sample.

iii. The γ-fiber was predominant in the recrystallization texture of both samples after the second annealing. However,
the coarse grain and the α-fiber present in the starting materials led to the development of slightly shifted components of γ-fiber and α-fiber retained after the second annealing, while the fine grain and weaker α-fiber produced a stronger γ-fiber after annealing. These differences occurred due to the influence of grain size on the development of γ-fiber and the low recrystallization rate of α-fiber.

iv The fine grain and weak texture in starting materials were advantageous in order to develop a high ϱ value. The ϱ value of 1.92 was achieved for 59% of γ-fiber and 7% of θ-fiber. This finding was mainly attributed to the presence of a strong γ-fiber and a weak θ-fiber.

Conflicts of interest

The authors declare no conflicts of interest.

Acknowledgments

The authors would like to thank Fundação de Amparo à Pesquisa de Minas Gerais (Fapemig), Conselho Nacional de Desenvolvimento Científico e Tecnológico (CNPq, 142380/2013-0), Comissão de Aperfeiçoamento de Pessoal do Nível Superior (Capes-PROEX, 0835/2018), and Aperam South America.

REFERENCES


