Ridging-free Ferritic Stainless Steel Produced through Recrystallization of Lath Martensite

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Ridging phenomenon was successfully suppressed in a ferritic stainless steel by controlling microstructure through recrystallization of lath martensitic structure. Fe–12Cr–1Ni alloy was quenched after the solution treatment in an austenite single phase region to obtain lath martensitic structure. Cold rolling was performed to the quenched materials up to 80% reduction in thickness before the annealing for recrystallization. With increasing the reduction by cold rolling, the recrystallization was promoted and ferrite grain size was decreased to 20 μm after recrystallization in the 80% pre-cold-rolled material. A weak (111)/ND recrystallization texture was formed by the cold rolling, but no grain colonies existed in the microstructure. As a result, the materials produced through the recrystallization of lath martensite did not cause ridging during tensile deformation, although an orange peel appeared when the grain size was not refined enough.

KEY WORDS: ridging; ferritic stainless steel; austenitization; lath martensite; colony; texture; recrystallization; grain size; isotropic structure; misorientation.

1. Introduction

Ferritic stainless steel sheets often form corrugations parallel to the rolling direction on the sheet surface during deep drawing or press forming, so-called ‘ridging’ or ‘roping’.1–11) The products having ridging require additional polishing to keep the exterior appearance fine, and sometimes cause cracking along the ridges. Therefore, the prevention of ridging is one of the important target for improving the productivity and economics of ferritic stainless steel sheets. In general, the ridging phenomenon is understood to be caused by anisotropic plastic flow due to coarse banded structure composed of ‘colonies’.1–8,11) The colony is defined as a cluster of grains with similar crystallographic orientation, whose size sometimes reaches the order of several millimeters.2) Since the dislocations move to the same direction on the same slip planes within each colony, the material deforms anisotropically owing to the differences of r value,3,4) crystal rotation,5) contraction to width direction6) among colonies.

In order to suppress the formation of ridging, the coarse colonies must be subdivided and the texture should be controlled so as to reduce the crystallographic anisotropy. However, it is basically difficult to control the strong texture derived from solidification structure, because the ferritic stainless steel has no austenite-ferrite phase transformation and the texture has to be controlled only by using recrystallization. Some attempts have been done to suppress the ridging for conventional ferritic stainless steels in previous studies.6–11) For example, it has been reported that the recrystallization after cold rolling is promoted by dispersing hard martensite phase,8) the columnar structure formed through solidification is refined by inoculation9) or electromagnetic stirring,10) the colonies are scattered by spread rolling,11) and so on. Nevertheless, the formation of ridging has never been completely suppressed.

A new method for alloy design and microstructure control using the recrystallization of lath martensite is proposed in this paper for developing a complete ridging-free ferritic stainless steel. Since the martensitic transformation proceeds so as to accommodate anisotropic lattice displacement by forming several fine grains with different crystallographic orientations (variants),12) the banded structure can be subdivided effectively and the anisotropy is reduced through the multi-variant transformation on cooling, even if the coarse austenite colonies have been formed during hot rolling after solidification. As a result, the recrystallized ferritic structure formed from such a martensitic structure is also expected to have crystallographic isotropy. In this work, the condition to obtain martensitic single structure was examined in a 12%Cr+1 steel containing a small amount of nickel. The recrystallization behavior and the
crystallographic character of the material were then investigated by means of microstructure observation and EBSP analysis. The ridging property of the developed material was evaluated in comparison with the result of a conventional ferritic stainless steel.

2. Alloy Design and Experimental Procedure

The material applied to the examination of microstructure control has to undergo martensitic transformation on cooling; that is, austenite single phase has to be obtained at elevated temperature, and the Ms and Mf temperatures should be higher than ambient temperature. In addition, the reversion temperature (As) is also desired to be sufficiently high because the annealing for recrystallization has to be performed without re-austenitization. However, it is generally impossible to obtain martensitic single structure in the conventional low-carbon ferritic stainless steel because it does not have austenite phase region at any temperature. In this study, therefore, a moderate amount of nickel was added to 12%Cr steel without adding carbon so as not to reduce formability and corrosion resistance. Figure 1 shows an equilibrium phase diagram of Fe–12%Cr–Ni ternary alloy system calculated with Thermo-Calc. It is found that nickel markedly enlarges the austenite phase region and the addition of 1% Ni enables the use of austenite phase in the temperature range between 1120 K and 1370 K. According to the approximate assessments of alloying elements on transformation temperatures by Pickering,13) Ms and As temperatures of the 12Cr–1Ni steel are estimated at about 590 K and 1040 K, respectively, which are high enough to complete the martensitic transformation on cooling and perform the annealing for recrystallization without re-austenitization.

The chemical compositions of the 12Cr–1Ni steel and JIS SUH409 steel used as a reference are listed in Table 1. The ingots of 1.5 kg were prepared by melting elemental metals in an induction furnace under an argon gas atmosphere, and then hot-rolled to 10 mm thick at 1373 K. The steel plates obtained were firstly subjected to the austenitization at 1273 K for 1.8 ks followed by water quenching to obtain fully martensitic structure, and then cold-rolled by 5 to 80% as the pre-treatment for recrystallization. The annealing for recrystallization was performed at 973 K or 1023 K, which is below the As temperature, for the as-quenched specimen (undeformed material) and the cold-rolled specimens (deformed materials). Microstructure was observed with an optical microscope and transmission electron microscope (TEM). Distribution of crystallographic orientation in the microstructure was detected by Electron Back-Scattering Pattern (EBSP) method with a scanning electron microscope. The data obtained by the EBSP method was analyzed with the TSL Orientation Imaging Microscopy (OIM) system. Hardness was measured by Vickers hardness testing (load: 98 N). Ridging property was evaluated by observing the surface appearance of plate specimens in which 20% of strain was given by tensile testing. In addition, the depth and distribution of the ridges was measured with a roughness tester.

3. Results and Discussion

3.1. Martensitic Structure and Its Recrystallization Behavior in 12Cr–1Ni Steel

Figure 2 shows microstructure of the 12Cr–1Ni steel which was water-quenched after the solution treatment of 1273 K–1.8 ks. Typical lath martensitic structure is observed, which is characterized by the stratified structure composed of prior austenite grains, packets, blocks in the OM image (a), and laths containing high-density of disloca-

![Image](https://via.placeholder.com/150)

**Fig. 1.** Equilibrium phase diagram of Fe–12mass%Cr–Ni ternary alloy system.

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**Table 1.** Chemical compositions of steels used in this study (mass%).

|       | C    | N    | Cr   | Ni   | Mn   | Si   | P    | S    | Al  | Ti  | Fe   |
|-------|------|------|------|------|------|------|------|------|-----|-----|------|
| 12Cr-1Ni | 0.005| 0.014| 11.99| 0.99 | 0.21 | 0.02 | 0.013| 0.011|     |     |      |
| SUH409  | 0.008| 0.002| 11.22| 0.25 | 0.20 | 0.63 | 0.028| 0.001| 0.02| 0.21| Bal  |

![Image](https://via.placeholder.com/150)

**Fig. 2.** Optical micrograph (OM image) (a) and transmission electron micrograph (TEM image) (b) of 12Cr–1Ni steel quenched after solution treatment at 1273 K for 1.8 ks.
tion in the TEM image (b). The dislocation density was estimated at $2.0 \times 10^{15}/\text{m}^2$ by X-ray diffractometry. This dislocation density is large enough in terms of the driving force for recrystallization. Figure 3 represents the orientation imaging micrograph and the inverse pole figure of the as-quenched material. The crystallographic orientation shown by color in the orientation imaging micrographs is of the normal direction to specimen surface (ND). Since the martensitic transformation occurs in multi-variant mode to cancel out the crystallographic anisotropy, the distribution of crystallographic orientation is almost random and no texture appeared although the variants in each packet are limited to a few kinds. Figure 4 shows the hardness of quenched and cold-rolled specimens as a function of thickness reduction by cold rolling. The martensite in this alloy is not so hard originally (HV 2.7 GPa) because of its ultralow concentration of carbon, and also exhibits little work hardening during the cold rolling. This deformation character is important in terms of the deformability for cold rolling and press forming.

Fig. 3. Orientation imaging micrograph and inverse pole figure of the as-quenched 12Cr–1Ni steel.

Fig. 10. Exterior appearance of tensile test pieces strained by 20% in 12Cr–1Ni steels recrystallized after cold rolling at different reductions and JIS SUH409 steel.

Fig. 7. Orientation imaging micrographs and inverse pole figures in 12Cr–1Ni steel annealed at 1 023 K for 1.2 ks without cold rolling (a), 12Cr–1Ni steel annealed at 1 023 K for 60 s after 80% cold rolling (b) and JIS SUH409 steel annealed at 1 223 K for 60 s after 80% cold rolling (c).
Figure 5 shows the softening behavior during annealing in the as-quenched specimen (0%) and the cold-rolled specimens (5%, 20%, 80%). The hardness is plotted as a function of annealing parameter, $T \left( \log \frac{t+20}{H_{11001}^2} \right)^2$, as in the previous studies$^{15,16}$; where $T$ and $t$ are temperature in K and time in hour, respectively. In all of specimens, it is confirmed that the hardness varies with the variation of annealing parameter and drops abruptly owing to the discontinuous recrystallization after the gradual softening by recovery.$^{15}$ The time to the beginning and completion of recrystallization is significantly shortened with increasing the reduction by cold rolling. Since the dislocation density, which corresponds to the driving force for recrystallization, is hardly increased by the cold rolling up to 40% in ultralow carbon martensite,$^{17}$ the promotion of recrystallization by cold rolling is thought to be due to the increase in the density of formation site for the recrystallized grains.$^{18}$ On the other hand, it should be noted that the hardness of the fully recrystallized materials is about 1.2 GPa which is softer than that in conventional ferritic stainless steels with annealing. Figure 6 represents optical micrographs obtained before and after recrystallization for the materials with different pre-cold rolling. Although almost equiaxed ferritic structure is formed in all specimens after the recrystallization, the grain size of ferrite is reduced with increasing the amount of pre-cold rolling. The grain refinement to 20 μm is achieved by 80% of pre-cold rolling.

3.2. Crystallographic Characters of Ferritic Structure Formed by Recrystallization of Lath Martensite

The specimens selected for investigating the crystallographic characters are: (1) 12Cr–1Ni steel annealed at

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*$^2$ The parameter $T \left( \log \frac{t+20}{H_{11001}^2} \right)^2$ is conventionally referred to as ‘tempering parameter’. However, the heat treatment performed here aims to soften the steel sheets for improving deformability; thus, the term ‘annealing parameter’ is used in this article to make the aim clear.
1023 K for 1.2 ks without pre-cold rolling (undeformed material); (2) 12Cr–1Ni steel annealed at 1023 K for 60 s after 80% cold rolling (deformed material); and (3) commercial JIS SUH409 steel annealed at 1223 K for 60 s after 80% cold rolling (reference). Orientation imaging micrographs and inverse pole figures for these materials are shown in Fig. 7. Two kinds of specimens obtained from martensite ((a) and (b)) have no colony. In particular, the undeformed material exhibits completely isotropic ferritic structure with random distribution in crystallographic orientation. The reason for the formation of isotropic structure could be explained by the facts that (1) lath martensitic structure originally has little crystallographic anisotropy, and besides, (2) it recrystallizes by the grain boundary bulge mechanism\cite{15} resulting in the preservation of isotropy even after recrystallization. As for the deformed material, a weak $\{111\}/ND$ texture is formed because of the heavy pre-cold rolling, but the $\{111\}/ND$ grains are not clustered but scattered in the microstructure. On the other hand, a strong texture of $(111)/ND$ is developed in the SUH409 (c) and most of the grains is colored blue. In addition, it is found that some $(001)/ND$ grains colored red form a cluster as pointed by the arrow, which also corresponds to a small colony. It should be mentioned here that Tsuzaki et al.\cite{19} reported that the $(111)/ND$ recrystallization texture is easily developed in the deformed martensite rather than deformed polygonal ferrite in a low alloy ultra-low carbon steel. Although the microstructure of SUH409 was formed from polygonal ferrite through recrystallization, it is thought that the extremely anisotropic initial structure in the hot-rolled plate resulted in the strong texture formation after recrystallization. Figures 8 and 9 show the distribution of high angle boundaries having misorientation more than 15 degrees, and the relation between the misorientation angle and its number fraction for detected grain boundaries (>5 degree), respectively. The data was obtained from the specimens shown in the Fig. 7. In the two 12Cr–1Ni steels ((a) and (b)), almost all grain boundaries have large misorientation (high angle boundary) and the distributions of misorientation angle are just similar to the random distribution shown by the open circles, which was theoretically calculated by Mackenzie.\cite{20} On the other hand, the SUH409 (c) contains a high frequency of low angle boundaries. Considering that the grain boundaries with small misorientation (subgrain boundaries) do not work as effective barriers to dislocation motion,\cite{21} the grain size in the SUH409 should not be measured with an optical microscope but be evaluated with the high angle boundary map of Fig. 8 obtained by EBSP analysis.

3.3. Ridging Property of 12Cr–1Ni Steel Microstructure—Controlled by Recrystallization of Lath Martensite

To demonstrate the effect of isotropic structure on ridging property, the appearance of tensile test pieces strained by 20% is shown for the developed 12Cr–1Ni steels and the SUH409 steel in Fig. 10. The test pieces of 0%, 80% and SUH409 correspond to the materials shown in the Fig. 7,
and that of 20% is an additional specimen of 12Cr–1Ni steel which was annealed at 1 023 K for 60 s after 20% cold rolling. The SUH409 clearly exhibits ridging pattern parallel to the tensile direction on the specimen surface, while the other three specimens produced by recrystallization of lath martensitic structure never cause ridging. Although some roughness is observed in the undeformed and the 20% deformed materials, this is other kind of surface defect called as ‘orange peel’, which appears when the grain size is large. The 80% deformed material having fined ferritic structure with the grain size of 20 μm exhibits excellent surface condition without ridging nor orange peel. It has been thought that grain refinement prevents ridging in ferritic stainless steel. However, the result in this study indicates that the grain refinement itself dose not directly suppress ridging, but it indirectly improves the ridging property through the subdivision of colonies and the dispersion of ⟨111⟩//ND and ⟨001⟩//ND grains. Figure 11 represents the roughness profiles of the specimen surface for the 12Cr–1Ni steels recrystallized after cold rolling at different reductions and JIS SUH409 steel. It is found that the SUH409 has caused considerable ridging with a depth of 100 μm or more, while the developed steels produced by recrystallization of lath martensitic structure have flat surfaces, especially in the 80% deformed material. These results suggest that an excellent exterior appearance can be also expected in deep drawing or press forming for the developed isotropic-structured ferritic stainless steel with fine grain size.

4. Conclusion

A new method for alloy design and microstructure control has been proposed for ferritic stainless steel to suppress ridging phenomenon. The process and the characteristics of the obtained materials are summarized as follows:

1) 12Cr–1Ni steel is fully austenitized at elevated temperature and undergoes martensitic transformation on following quenching.

2) Recrystallization of the lath martensitic structure forms crystallographically isotropic or near-isotropic ferritic structure with no colony, because the lath martensitic structure is originally isotropic owing to the multi-valiant transformation mechanism.

3) The ferritic stainless steel sheets produced through the recrystallization of lath martensite never cause ridging during tensile deformation. In particular, the exterior appearance is successfully kept fine when the grain size is refined by sufficient prior cold rolling before recrystallization.

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