Effects of Excess C on New Grain Formation and Static Recrystallization Behavior at Shear Bands in Cold-Rolled Ultra-Low Carbon Steel Sheets

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The effects of excess carbon on static recrystallization of cold-rolled steel sheets were studied using Ti-bearing ultra-low carbon steels with and without carbon in solution. Here, excess carbon means solute carbon added beyond the atomic equivalent ratio with Ti. Static recrystallization was evidently accelerated by the presence of excess carbon. Microstructural observation confirmed that statically recrystallized grains were generated at ferrite grain boundaries and deformation bands. Preexisting fine grains also stimulated static recrystallization. However, nucleation at deformation bands was not particularly significant in Ti-bearing ultra-low carbon steels without excess carbon. This difference derived from excess carbon is discussed in detail in relation to stored energy, preferential nucleation sites of static recrystallization and grain boundary migration.

KEY WORDS: ultra-low carbon steel; cold rolling; carbon; nucleation; recrystallization; shear band; grain boundary; texture.

1. Introduction

Texture in metals and alloys has a large influence on various properties and induces anisotropy. In particular, the texture evolved by static recrystallization (SRX) has been studied extensively, as SRX affects the plastic deformability of low carbon steels. For example, a study of texture evolution during SRX in cold-rolled steel sheets with carbon contents from 0.6 to 46 ppm revealed that the intensity of the \{111\} texture increased with increasing C contents up to 3.5 ppm but decreased when C exceeded 3.5 ppm.1) In Ti-bearing interstitial free (IF) and low carbon (LC) Al-killed steels, the intensity of the \{111\} texture developed by cold rolling followed by annealing increased with increasing annealing temperature, but the major orientation in the Ti-bearing IF steel was \{111\} \textless 112 \textgreater, while the major texture component in the LC Al-killed steel was \{111\} \textless 011 \textgreater.2) However, these studies were conducted using X-ray diffraction (XRD), which is suitable for measurement of the SRX behavior and average texture of a wide range, but does not always provide the localized information necessary for a detailed discussion of the mechanism of texture evolution due to nucleation and growth during SRX.

It is known that nucleation of SRX occurs preferentially at i) sites with a high dislocation density and high strain gradient,3,4) ii) deformation-induced microstructures such as shear bands and deformation bands,5–8) iii) around coarse particles9) and iv) at grain boundaries.7) Due to the large localized deformation in these areas, the crystallographic orientations of the statically recrystallized grains tend to differ largely from those of the surrounding matrix. However, very few studies have investigated the effects of the crystallographic orientations of the new grains on texture evolution during SRX.

Formation of shear bands is promoted by element addition and affects both the SRX behavior and formability of steel.10) Furthermore, it has been reported that solute carbon in IF and LC steels promoted shear band formation by localization of deformation, which promotes the formation of new grains with the \{111\} \textless 112 \textgreater Goss orientation.8) This study suggests the possibility of formation of new grains having specific crystallographic orientations largely different from the mother grains, which affects texture evolution. It is interesting to note that the SRX behavior described

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above may be different from the conventional mechanisms of continuous SRX commonly observed in IF and LC steels.

In this paper, the effects of excess C on SRX behavior during annealing after cold rolling were investigated in detail by observation from the transverse direction (TD), focusing on the formation of statically recrystallized grains at shear bands. Here, excess carbon means solute carbon added in excess of the atomic equivalent ratio with Ti. Furthermore, the characteristic mechanisms of SRX at the shear bands were also investigated.

2. Experimental Procedure

The chemical compositions of the steels used here are listed in Table 1. The ultra-low C steel (hereinafter, ULC) is a Ti-bearing ultra-low carbon steel containing 0.0025 mass% C and 0.020 mass% Ti, which is sufficient to stabilize the total C content, and the excess C added steel (hereinafter, EC-ULC) is an ultra-low carbon steel with a C content larger than the atomic equivalent ratio with Ti. From the chemical compositions in Table 1, the mole fractions of C in ULC and EC-ULC calculated by the following Eq. (1) were 0.019 and 0.017, respectively. The Si and S contents of ULC and EC-ULC were calculated as 0.008 and 0.0005 mass%, respectively.

\[ x_{Ti,eff} = x_Ti - x_N - x_S \] .......................... (1)

where \( x_M \) is the mole fraction of each element. In ULC, the mole fraction of effective Ti (0.019) is larger than that of C (0.012), whereas in EC-ULC, the mole fraction of C (0.019) is larger than that of Ti (0.017), confirming that EC-ULC contains excess C.

For specimen preparation, vacuum melted ingots were forged to a thickness of 30 mm, then cut to a length of 100 mm. The slabs were reheated to 1 523 K for 3 600 s and hot rolled in 6 passes to a thickness of about 4.0 mm at a finish temperature of about 1 203 K. During hot rolling, part of the Ti should form precipitates of TiN and then TiS. After rolling, the hot-rolled sheets were cooled to 873 K, which corresponds to the coiling temperature, by immersion in an alumina fluidized-bed furnace at room temperature. The sheets were kept at 873 K for 3 600 s in an electric furnace, followed by furnace cooling. After pickling, the sheet thickness was reduced to 2.8 mm by grinding both surfaces to remove scale and surface defects, and the ground sheets were cold rolled to a thickness of 0.7 mm by 75% cold-rolling reduction.

Small samples cut from the two cold-rolled sheets were sealed in a vacuumed quartz tube and annealed in a muffle furnace at temperatures from 1 73 K to 1 203 K. During hot rolling, part of the Ti should form precipitates of TiN and then TiS. After rolling, the hot-rolled sheets were cooled to 873 K, which corresponds to the coiling temperature, by immersion in an alumina fluidized-bed furnace at room temperature. The sheets were kept at 873 K for 3 600 s in an electric furnace, followed by furnace cooling. After pickling, the sheet thickness was reduced to 2.8 mm by grinding both surfaces to remove scale and surface defects, and the ground sheets were cold rolled to a thickness of 0.7 mm by 75% cold-rolling reduction.

Small samples cut from the two cold-rolled sheets were sealed in a vacuumed quartz tube and annealed in a muffle furnace at temperatures from 1 023 K and annealing times from 0 to 900 s. After annealing, the samples were removed from the furnace and then air-cooled in a quartz tube.

The microstructures of each steel sheet were observed with an optical microscope (OM) and SEM-EBSD (Electron Backscatter Diffraction; SEM/ SU-5000; Hitachi High-Tech Corporation). Backside mapping was conducted by mechanical polishing, followed by electrical polishing with a solution of 2-n-butoyxethanol and perchloric acid (volume ratio of 10:1) at 293 K and a voltage of 12 V for 300 s using a stainless steel plate as a cathode. For the SEM-EBSD observation and analysis from TD, the color mapping was decoded along ND for easier evaluation of the evolved texture on the RD (Rolling Direction) plane. Observation was conducted along TD because the shear bands were easily visible.

Changes in hardness during annealing were measured using a micro-Vickers hardness tester (HMV-G 20 S; Shimadzu Corporation). Ten points were measured under a load of 2.942 N for 15 s, and the mean hardness value of 8 measurements, excluding the maximum and minimum values, was derived.

The aging index (AI) of the hot-rolled samples was evaluated by the increment of the lower yield stress after aging at 373 K for 3 600 s at 8% prestrain. JIS No. 5 tensile test pieces (gauge length: 50 mm, width of parallel part: 25 mm) were prepared from the hot-rolled sheets, and tensile tests were conducted at a crosshead speed of 10 mm/min.

3. Experimental Results

3.1. Microstructural Observation of Cold-Rolled Steels

Figures 1 and 2 are OM images of the cold-rolled specimens on the ND and TD planes, respectively, where (a) is the ULC steel and (b) is the EC-ULC steel. Figure 1 shows a pancake-like microstructure elongated in the RD, which is often observed in cold-rolled samples. Figure 2 shows the development of lamellar microstructures and shear bands. The grain size was widely dispersed in the cold-rolled microstructure of ULC, while small grains were uniformly distributed in EC-ULC. The initial grain size was measured from Fig. 1 by the linear intercept method11) assuming that the grains did not grow in the TD direction during cold rolling. The initial grain sizes of ULC and EC-ULC were estimated to be 57 μm and 34 μm, respectively, or difference of about 1.7 times. The AI values of the hot-rolled ULC and EC-ULC were 0.2 and 23.4 MPa, respectively. These results confirmed that ULC did not contain excess C, but excess C existed in EC-ULC.

Table 1. Chemical compositions of steels used: ultra-low carbon steel (ULC) and excess carbon added ultra-low carbon steel (EC-ULC).

| Steels     | Unit     | C   | Si  | Mn  | S   | Al  | N   | Ti | Effective Ti |
|------------|----------|-----|-----|-----|-----|-----|-----|----|--------------|
| ULC        | Mass fraction (mass%) | 0.0025 | <0.008 | 0.15 | <0.0005 | 0.014 | 0.0009 | 0.020 | –            |
|            | Mole fraction  | 0.012 | <0.016 | 0.15 | <0.0009 | 0.029 | 0.0036 | 0.023 | 0.019        |
| EC-ULC     | Mass fraction (mass%) | 0.0041 | <0.008 | 0.15 | <0.0005 | 0.021 | 0.0011 | 0.019 | –            |
|            | Mole fraction  | 0.019 | <0.016 | 0.15 | <0.0009 | 0.043 | 0.0044 | 0.022 | 0.017        |

* Mole fraction of effective Ti \( x_{Ti,eff} = x_{Ti} - x_{N} - x_{S} \)
3.2. Static Recrystallization Behavior and Microstructural Observation from ND Plane

Figure 3 shows the change in Vickers hardness ($H_V$) with increasing annealing time at 1023 K. In both ULC and EC-ULC, $H_V$ decreased rapidly at about 60 s after the start of annealing, and the dispersion of hardness increased around that time due to the mixture of deformed and statically recrystallized microstructures. However, the hardness converged to almost a constant value after full SRX. The times to complete softening for ULC and EC-ULC were about 300 s and 80 s, respectively. Thus, complete softening of EC-ULC was achieved in one-quarter the time of ULC.

In order to investigate the difference in softening behavior, microstructural observation was conducted from the ND plane by SEM-EBSD. Figures 4(a) to 4(c) show typical IPF (Inverse Pole Figure) maps during annealing of ULC at 1023 K. The annealing times were 0 s (as-cold rolled), 65 s and 180 s. Figures 4(d) to 4(f) show the IPF maps of EC-ULC under the same conditions. Figures 4(a) to 4(c) show that the fraction of statically recrystallized grains increased with increasing annealing time, while the fraction of the deformed microstructure decreased. In Fig. 4(b), statically recrystallized grains appeared to form preferentially from grain boundaries in deformation bands, and some of statically recrystallized grains also appeared to form from grain interiors. Comparing Figs. 4(b) and 4(e), the fraction of SRX was larger in EC-ULC than in ULC, and SRX was also completed earlier in EC-ULC.

The fraction of SRX was measured by using the IPF maps of the samples annealed at 1023 K for various annealing times in addition to the IPF maps shown in Fig. 4. Figure 5 shows the results summarized as the change in the fraction of SRX with annealing time at 1023 K. The fraction of SRX in EC-ULC increased rapidly at about 60 s, and this change was more rapid in EC-ULC than in ULC. SRX was completed earlier in the EC-ULC (60 s) than in ULC (180 s). This result agreed with the tendency of the softening curve shown in Fig. 3.

The pole figures obtained from the same IPF image as in Fig. 4 and changes in the $\{111\}$ texture intensity with increasing annealing time are summarized in Fig. 6. In ULC, the $\{111\}$ texture intensity decreased once at about 60 s and then increased, whereas in EC-ULC, the $\{111\}$ texture intensity was almost the same as that of the cold-rolled sample until about 60 s and then increased gradually. Thus, in both steels, the $\{111\}$ texture intensity increased with increasing annealing time, and this behavior was deeply related to the progress of SRX and grain growth. In particular, the increased $\{111\}$ texture intensity during the grain growth
Fig. 4. IPF maps during annealing at 1 023 K of ULC and EC-ULC; annealing times of (a), (d) 0 s (as-cold rolled), (b), (e) 65 s and (c), (f) 180 s, respectively.

Fig. 5. Change in the recrystallized fraction with annealing time at 1 023 K.

Fig. 6. Change in the [111] texture intensity evolved on the ND plane with increasing annealing time at 1 023 K.

Fig. 7. IPF maps of ULC annealed at 1 023 K for (a) 0 s (as-cold rolled), (b) 65 s and (c) 180 s, and those of EC-ULC annealed at 1 023 K for (d) 0 s (as-cold rolled), (e) 65 s and (f) 180 s. Color map decoding was changed to along the ND while observation was conducted along the TD.
The grain size is fine. Energy is introduced with a higher density when the initial inhomogeneous deformation zone with high stored strain that SRX of the finer grains occurred faster\(^\text{12}\) because an inhomogeneous deformation zone with high stored strain energy is introduced with a higher density when the initial grain size is fine.\(^\text{13}\) Past studies have reported that excess C promotes the formation of shear bands during cold rolling,\(^\text{8,14,15}\) but did not investigate the effects of SRX nucleation from shear bands or the shear bands themselves on SRX. In our research, the TD plane was observed in detail in order to focus on the inhomogeneous deformation zone formed in cold rolling, and especially the formation of statically recrystallized grains from shear bands. The results will be presented in Section 3.3 below.

### 3.3. Observation of TD Plane of Static Recrystallization Microstructure

Figures 7(a) to 7(c) show typical IPF maps of the ULC steels annealed at 1 023 K, where (a) to (c) are annealing times of 0 s (as-cold rolled), 65 s and 180 s, respectively. Figures 7(d) to 7(f) also show the IPF maps of the EC-ULC steels under the same annealing conditions. In the IPF maps of the early stage of SRX in Figs. 7(b) and 7(e), statically recrystallized grains formed preferentially in shear bands or at lamellar boundaries. This is attributed to preferential nucleation of SRX from regions with high strain energy and a high dislocation density gradient, i.e., shear bands and inhomogeneous deformation zones along grain boundaries.\(^\text{7,16,17}\) Further analysis confirmed that nucleation of SRX from lamellar boundaries was dominant in ULC (Fig. 7(b)), whereas SRX nucleation from shear bands was dominant in EC-ULC (Fig. 7(e)). From the difference in the color of the IPF maps, it is readily apparent that the new grains that formed at grain boundaries and shear bands have a large misorientation from their surrounding matrices. Thus, microscopic observation on the TD plane suggests that the mechanism of SRX is the “discontinuous type.” A comparison of Figs. 7(c) and 7(f), which are the microstructures after full SRX, shows that ULC contains many statically recrystallized grains elongated in the RD direction, while EC-ULC is characterized by uniform equiaxed grains. This is thought to be affected by the difference in the grain sizes and distribution densities of the deformation-induced microstructures in the two steels, as described in Section 3.1. When the dispersion of grain size was large, as in ULC, the statically recrystallized grains which formed preferentially at the grain boundaries of fine grains initiated and grew rapidly along lamellar microstructures while preferentially encroaching on the surrounding deformation zone.\(^\text{13}\) However, when the grain size was uniformly small and the density of the shear bands was high (Section 4.2), as in EC-ULC, nucleation occurred earlier and was relatively uniform, the growth rate was large, and differences in the growth rate were small, resulting in the equiaxed statically recrystallized grains observed in EC-ULC. The mechanism of SRX is discussed in detail in the following.

To investigate the change of grain boundary misorientations with SRX, the average misorientations were calculated from IPF maps obtained by SEM-EBSD, as shown in Fig. 8. In both steels, the average misorientation angle tended to increase with increasing annealing time at 1 023 K. However, the average misorientation angle increased continuously and gradually in ULC but increased rapidly at around 60 s in EC-ULC. That is, in ULC, statically recrystallized grains grow with small grain boundary misorientations with respect to the matrix. This behavior can be regarded as a typical “continuous SRX type,” in which grain boundary misorientations increase continuously in the grain growth process. In contrast, the SRX behavior of EC-ULC can be regarded as the “discontinuous SRX type,” as statically recrystallized grains grow with a large grain boundary misorientation from the surrounding matrix.

### 4. Discussion

As described above, the SRX behavior, microstructural changes in the ND and TD planes and differences in the texture evolution of ULC and EC-ULC were investigated using cold-rolled sheets annealed under various annealing temperature and time conditions. As a result, it became clear that the microstructural change observed in the TD plane differed greatly from that in the ND plane. Namely, SRX of both ULC and EC-ULC began preferentially at lamellar boundaries and shear bands, but statically recrystallized grains mainly formed at lamellar boundaries in ULC and at shear bands in EC-ULC. Moreover, the change of grain boundary misorientations in the process from the formation to growth of statically recrystallized grains suggested continuous SRX behavior in ULC and discontinuous SRX behavior in EC-ULC. The reasons for this extremely large difference in the SRX behaviors of the two steels are discussed below.

#### 4.1. Effect of Excess C on New Grain Formation and Unique Recrystallization Behavior at Shear Bands

Microstructural observation of the cold-rolled sheets from
the TD plane revealed that statically recrystallized grains formed preferentially at grain boundaries and shear bands, and the misorientations of the new grains were larger than those of the surrounding matrix (Figs. 7 and 8). At a glance, this SRX behavior may be judged to be discontinuous SRX. The causes of this behavior are examined in detail below.

In order to understand the difference in the microstructures of the shear bands formed during cold rolling of ULC and EC-ULC, the average spacing of the shear bands and the change in the misorientation within the shear bands were analyzed using IPF maps of three fields. Figure 9 shows the method used to measure the misorientation change in the shear bands. In Fig. 9, (a) is an IPF map of ULC annealed at 1023 K for 65 s, (b) shows the magnified IPF map of the area surrounded by the red square in (a), and (c) shows the profile indicating the misorientation change along the red arrow in (b) on the (001) pole figure. First, the spacings of the shear bands in ULC and EC-ULC were 1.82 μm and 1.59 μm, respectively. The shear band spacing was smaller and the shear bands formed more densely in EC-ULC. It was also found that the shear bands analyzed in Fig. 9 were close to the {111} <112> Goss orientation, the misorientation between the start (origin) and end points of the arrow in Fig. 9(b) was 22.8°, and the average misorientation between the lamellar microstructures in adjacent shear bands was 7.0°. Therefore, if new grains are generated in these shear bands by the “continuous SRX mechanism,” it is easy to estimate that new grains near the Goss orientation will develop with a misorientation of about 7°, and will develop with a high misorientation as much as a maximum of 23°.

The misorientation in the shear bands just before the end of primary SRX of ULC and EC-ULC was measured by measuring the misorientation change in the shear bands shown in Fig. 9 and is summarized in Fig. 10. ULC was annealed for 150 s and EC-ULC steel was annealed for 125 s, both at 1023 K. In Fig. 10, the misorientation change in the shear bands of EC-ULC was larger than that of ULC, indicating that the shear bands in EC-ULC have higher angle grain boundaries than those in ULC. Because a general high angle grain boundary is judged to be a high energy grain boundary,18) the grain boundary migration rate during SRX is considered to be high.19) That is, the onset and progress of SRX can be expected to be faster in EC-ULC, as shear bands containing higher angle grain boundaries develop more densely. This agrees with the experimental result shown in Fig. 5. These results are also consistent with a report10) that excess C promotes the formation of shear bands and SRX due to localization of deformation.

Figures 11 and 12 show the hardness distributions of the microstructure just before the end of primary SRX for ULC and EC-ULC, respectively. The samples evaluated here were ULC annealed at 1023 K for 150 s and EC-ULC annealed...
at 1 023 K for 125 s. In Figs. 11(a) and 12(a), larger circles indicate higher $H_v$ values, and the colors of the circles show the results of measurements of recrystallized grains in blue, unrecrystallized grains in white, shear bands in red and mixed structures in black. The hardness test results are classified into recrystallized grains, unrecrystallized grains and shear bands, and are summarized in Figs. 11(b) and 12(b). The hardnesses of the recrystallized grains, unrecrystallized grains and shear bands in ULC were 1.4, 1.7 and 1.9 GPa, respectively, whereas those in EC-ULC were 1.2, 1.6 and 1.9 GPa, respectively. Therefore, the difference in the hardnesses of these microstructures was greater in EC-ULC than in ULC. Since a hard microstructure with a high $H_v$ value has a high dislocation density and strain energy, nuclei of SRX are easily generated. Because a difference in the $H_v$ value between microstructures means that the driving force for grain boundary migration during SRX is large,\textsuperscript{20} the difference in the hardness distribution inside the microstructure also supports the conclusion that SRX occurs and is completed earlier in EC-ULC than in ULC.

These results indicate that addition of excess C to Ti-bearing ultra-low carbon steel promotes the formation of denser shear bands during cold rolling and the formation of a larger number of statically recrystallized grains in the shear bands during annealing, as well as faster grain growth due to the larger change in misorientation within shear bands and the larger hardness difference with the surrounding microstructure. Moreover, the initial grain size of EC-ULC was significantly smaller than that of ULC, suggesting that the difference in the initial grain size may affect the SRX behavior of the two steels. The larger misorientation change in the shear bands in EC-ULC caused behavior that could be judged to be discontinuous SRX in the early stage of SRX, since the statically recrystallized grains generated

Fig. 11. Hardness distribution of the microstructures evolved in ULC annealed at 1 023 K for 150 s: (a) IPF map and (b) hardness of each microstructure described in (a). Mixed indicates a point of some mixed structures of recrystallized and unrecrystallized areas, shear bands and grain boundaries. These results are excluded from (b).

Fig. 12. Hardness distribution of the microstructures evolved in EC-ULC annealed at 1 023 K for 125 s: (a) IPF map and (b) hardness of each microstructure described in (a). Mixed indicates a point of some mixed structures of recrystallized and unrecrystallized areas, shear bands and grain boundaries. These results are excluded from (b).
in shear bands resulted in a larger change in misorientation with respect to the surrounding matrix and, furthermore, the formation of new grains in these shear bands was dominant.

4.2. Microstructure and Recrystallization Behavior on TD Plane

SEM-EBSD observation on the TD plane revealed preferential SRX at grain boundaries and shear bands and the unique crystal orientation of these new grains (Fig. 7). Since these new grains have large grain boundary misorientations from the surrounding matrix, the mechanism of SRX may be misinterpreted as the discontinuous SRX type in both steels. However, this large misorientation can be explained by the fact that subgrains are formed by recovery in regions of where a high strain energy and dislocation density gradient exist in the deformed microstructure, then grow by encroachment on the surrounding deformed microstructure, causing a large change in misorientation in the surrounding deformed microstructure (Fig. 10), and thus already have a large misorientation with the adjacent microstructure during the growth period. This behavior is very similar to the formation of statically recrystallized grains with significantly different orientations from the matrix in the deformation zone in aluminum alloys.9 This is misinterpreted as discontinuous SRX in macroscopic microstructure observation after coarsening because part of the microstructure in the deformation zone grows with an orientation greatly different from that of the matrix. Annealing of ultra-fine-grained 304 stainless steel processed by large strain deformation also shows discontinuous behavior with a rapid change in the high misorientation at the moving grain boundary front due to coarsening of ultra-fine grains and encroachment on adjacent grains, but this is reported to be the continuous SRX type.21 Discontinuous SRX is essentially caused by twin formation during grain growth in low stacking fault energy metals and alloys.5 Therefore, although the mechanism of SRX in both steels in this study is basically the continuous SRX type, the new grain growth which progresses by encroachment on the surrounding deformed microstructure with large local misorientation appears to be discontinuous SRX.

The results of these investigations facilitate the understanding, for example, that some shear bands that develop in the ULC steel are close to the {111} <112> Goss orientation (Fig. 9), which is a site where statically recrystallized grains are likely to form, and the new grain formation and growth which occur there lead to the formation of Goss-oriented new grains. However, gradual encroachment by the {111} texture during grain growth (Fig. 6) prevents new grains with these unique orientations from becoming the major orientation. The results of the microstructural observation on the TD plane in this study might appear to differ greatly from the general knowledge obtained by XRD, but are essentially the same because this is one result of observation by extracting part of the early stage of SRX. The detection of microstructures with these unique orientations is important for understanding the SRX texture and can be clarified for the first time by microstructural observation on the TD plane. In the future, systematic multiscale research from the micro- to the macroscale orders by combining TD plane observation with XRD will be necessary.

5. Conclusions

The effects of excess C on the static recrystallization (SRX) behavior at shear bands were examined by comparing the new grain formation behavior in a Ti-bearing ultra-low carbon steel (ULC) and an excess C added ultra-low carbon steel (EC-ULC). The conclusions obtained in this study are summarized as follows.

(1) The onset of SRX occurred preferentially at shear bands in both steels. The onset and progress of SRX were faster in EC-ULC than in ULC, and, SRX was completed earlier in EC-ULC. Nevertheless, the evolved textures, with a rather sharp intensity at $\{111\}$, displayed the same appearance in both steels after full SRX, indicating that excess C did not significantly affect the finally evolved texture.

(2) The newly formed grains in EC-ULC appeared at glance to evolve by the mechanism of discontinuous SRX because of their large misorientations from the matrix, while those in ULC possessed much smaller misorientations with their parent grains, and thus were understood to be formed by classical continuous SRX. The large misorientations of the statically recrystallized grains in EC-ULC were attributed to deformed microstructures with large misorientations in the shear bands, as the growth of new grains in the shear bands resulted in an increase in misorientation during invasion of adjacent areas with large misorientation. Therefore, it was concluded that the SRX behavior in EC-ULC was also basically the continuous type.

(3) Addition of excess C to ULC steel caused more uniform and denser formation of shear bands by cold rolling and enhanced the formation of statically recrystallized grains at the shear bands, causing faster new grain formation and growth. These changes were caused by the larger misorientation within shear bands and the larger hardness difference due to the larger dislocation density change compared with other deformed microstructures.

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