Advanced Microstructure and Texture Characterization of Line Pipe Steel Plates Processed With Intensified Accelerated Cooling

Jing Su, Dengqi Bai, and Rick Bodnar

SSAB Americas, R&D
1755 Bill Sharp Boulevard, Muscatine, IA 52761
Phone: 563-889-6457
Email: jing.su@ssab.com

Keywords: Linepipe; Toughness; Texture; Dislocation densities; Grain boundary misorientation angle.

ABSTRACT

The optimization of microstructure through Thermomechanical Controlled Processing (TMCP) is crucial for achieving an optimal strength and toughness balance in linepipe steel plates. This study investigates linepipe steels processed with different rolling and cooling practices using a MULPIC (MULti-Purpose Interrupted Cooling) system. The objective is to explore the effects of finish rolling temperature and cooling conditions on product grain size, grain boundary characteristics, phase constituents, crystallographic texture, and through-thickness microstructure uniformity. The detailed microstructure and texture were characterized using optical microscopy and scanning electron microscopy (SEM) with electron backscattered diffraction (EBSD) techniques. The influences of various metallurgical factors on mechanical properties are discussed.

INTRODUCTION

TMCP is widely used in the production of advanced linepipe steels, requiring a good combination of high strength and toughness [1]. TMCP integrates controlled rolling with in-line accelerated cooling. Numerous studies have explored the effects of rolling parameters [2–4] and cooling conditions including cooling rate, start and finish cooling temperatures [5,6], as well as coiling temperatures [7,8] on microstructure and mechanical properties of linepipe steels [9,10]. Complex microstructures such as acicular ferrite and/or bainite with a mixture of polygonal ferrite, quasi-polygonal ferrite, martensite/retained austenite (M/A) islands and degenerate pearlite have been reported in linepipe steels. The microstructure variations are influenced by TMCP parameters. The effects of microstructure features, such as grain size, phase constituents, dislocation densities, and precipitates, as well as crystallographic texture on the mechanical properties of HSLA steels have been extensively investigated [11–14].

In this study, TMCP with accelerated cooling using a hybrid MULPIC system at the SSAB-Mobile rolling mill was used to produce 0.75” thick API X65 discrete plates. The hybrid cooling system incorporates a section of conventional laminar cooling and a 12 m long MULPIC cooling unit, known as intensified accelerated cooling (ACC) [15,16]. The achievable cooling rate using MULPIC is higher than laminar cooling. Two X65 samples with different finish rolling (finishing) temperatures, coupled with intensified ACC, were chosen. Transfer bar cooling was applied using the MULPIC, which reduces the time for the metallurgical delay. For comparison, an API X52 plate processed through controlled rolling followed by air cooling was also selected for study. The primary objective of this work is to investigate the influence of finishing temperature and cooling path on microstructure and texture through the thickness of the heavy-gauge linepipe plates.

EXPERIMENTAL PROCEDURE

The chemical compositions of the X52 and X65 plates are shown in Table I, with X52 containing slightly higher C and lower Mn contents compared to X65. The carbon equivalent (CE) values calculated according to Equation (1) are 0.34 and 0.37 for X52 and X65, respectively. Both plates were rolled from 6” thick continuously cast slabs down to 0.75” plates according to two different controlled-rolling pass schedules.

\[
CE = C + \frac{\text{Mn}}{6} + \frac{\text{Cr} + \text{Mo} + V}{5} + \frac{\text{Cu} + \text{Ni}}{15}
\]
Table I. Chemical Compositions of X52 and X65, wt. %

<table>
<thead>
<tr>
<th>Grade</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Mo</th>
<th>Nb</th>
<th>V</th>
<th>Ti</th>
<th>N</th>
</tr>
</thead>
<tbody>
<tr>
<td>X52</td>
<td>0.063</td>
<td>1.26</td>
<td>0.20</td>
<td>0.007</td>
<td>0.0004</td>
<td>None</td>
<td>0.039</td>
<td>0.028</td>
<td>0.014</td>
<td>0.0077</td>
</tr>
<tr>
<td>X65</td>
<td>0.055</td>
<td>1.45</td>
<td>0.22</td>
<td>0.008</td>
<td>0.0002</td>
<td>Addition</td>
<td>0.047</td>
<td>0.027</td>
<td>0.015</td>
<td>0.0062</td>
</tr>
</tbody>
</table>

The slab reheating temperatures for the two steels were in the range of 1230 °C to 1260 °C, exceeding their Nb dissolution temperatures (1120 °C to 1130 °C) calculated by Equation (2) \(^{[17]}\). Hence, Nb is expected to be in solution after reheating.

\[
\log[\text{Nb}] \left[ C + \frac{12}{14} \cdot N \right] = 2.26 - \frac{6770}{T} \tag{2}
\]

Here, \([\text{Nb}], [C], \text{and } [N]\) are alloy concentrations in wt.% and \(T\) is absolute temperature (K). The non-recrystallization temperature \((T_{nr})\) and recrystallization stop temperature \((T_R)\) were estimated using Equations (3) and (4) \(^{[9]}\). The estimated \(T_{nr}\) and \(T_R\) temperatures are 994 °C and 919 °C for X52 and 987°C and 912 °C for X65. Roughing was conducted at temperatures above the \(T_{nr}\) and the resume temperatures for finishing were at or slightly below the \(T_R\) temperatures of the two steels.

\[
T_{nr} = 174 \times \log\left[ \text{Nb} \times \left( C + \frac{12}{14} \cdot N \right) \right] + 1444 \text{ (°C)} \tag{3}
\]

\[
T_R = T_{nr} - 75 \text{ (°C)} \tag{4}
\]

After the last pass of finishing, a temperature rundown along the plate length was observed in the X65 plate with an approximately 50 °C difference between the mid length and the tail of the plate, illustrated in Fig. 1 (a). To examine the effect of finishing temperatures, samples were taken from the mid length and the tail of the X65 plate, which are referred to as HFT-ACC and LFT-ACC, respectively. Accelerated cooling was applied to the X65 plate after rolling using MULPIC, where a cooling rate of 35.6 °F/s (20 °C/s) was achieved. For comparison, an X52 plate produced by controlled rolling followed by air cooling was also selected (Fig. 1 (a)). The \(A_{3}\) temperatures for the two steels were calculated using the equation developed by Ouchi, et al. \(^{[18]}\). The finishing last pass (FLP) temperature of the air-cooled specimen was close to its calculated \(A_{3}\) temperature. For the LFT-ACC specimen, the FLP temperature was approximately 20 °C below the calculated \(A_{3}\) temperature, while the FLP temperature of HFT-ACC was about 30 °C above the \(A_{3}\) temperature. The finish cooling temperature for the X65 plate was 574 °C, which is below its \(B_s\) temperature (682 °C) calculated by the equation developed by Steven and Haynes \(^{[19]}\). It should be noted that these temperatures represent plate surface temperatures measured by a mill pyrometer and the core temperatures of the heavy-gauge plates are normally slightly higher than the surface temperatures. Fig. 1 (b) shows the measured surface temperatures and estimated bulk temperatures at the FLP for the three samples.
Uni-axial tensile tests were performed using standard API 5L tensile specimens along the transverse-to-rolling direction (TRD) of the plates using a 2000 kN Instron machine. Charpy V-notch (CVN) impact tests were conducted using full-size specimens (10 x 10 x 55 mm) according to ASTM E23 standard in the TRD orientation. A total of six temperatures, 24, 0, -20, -40, -60, and -80 °C, were used to construct the ductile-to-brittle transition curves. In addition to the Charpy impact energies, the shear areas and splits on the fracture surfaces were also evaluated.

The metallography samples were ground with diamond discs from 320# to 600# and then polished using 9 µm, 3 µm, and 1 µm diamond suspensions. The final polishing was performed using a non-crystallized colloidal silica solution. Optical microstructures, etched with a 2% Nital solution, were characterized using a Nikon Epiphot 200 optical microscope. For electron backscatter diffraction (EBSD) analysis, ion milling was performed after mechanical polishing using a JEOL Cross-section polisher IB-19530CP. Samples were ion milled in a rotational mode at an ion irradiation angle of 5° using a voltage of 8 kV for 3 hours.

EBSD analysis was conducted using a JEOL JSM-IT700 HR scanning electron microscope (SEM) equipped with Oxford Symmetry 2 EBSD detector and AZtec software for data acquisition. For texture analysis, a low magnification of 150X with a step size of 0.6 µm was used to ensure around 20,000 grains were included for statistical accuracy. Wright et al. [20] compared the texture results measured by EBSD and X-ray diffraction and found that approximately 10,000 grains were required to have a reliable and accurate texture measurement using EBSD. The average values and distributions of grain size and grain morphology were generated based on scans performed at a magnification of 150X. For dislocation density and secondary phase analysis, a high magnification of 500X and a fine step size of 0.1 µm were employed. Post-acquisition data analysis was performed using AZtecCrystal software.

RESULTS AND DISCUSSION

Mechanical Properties

In Fig. 2 (a), a comparison of yield strength (YS) and ultimate tensile strength (UTS) along the TRD is presented for air-cooled, HFT-ACC, and LFT-ACC specimens. Among the three conditions, the HFT-ACC specimen demonstrates the highest YS and UTS, followed by LFT-ACC, while the air-cooled specimen exhibits the lowest YS and UTS. Fig. 2 (b) shows the total elongation and Y/T (i.e., YS/UTS) ratio of the three conditions. The air-cooled specimen exhibits the highest total elongation, followed by LFT-ACC, while HFT-ACC shows the lowest elongation. As expected, the total elongation follows an inverse trend to strength. The Y/T ratio of the air-cooled specimen is higher than that of the HFT-ACC and LFT-ACC specimens.

Fig. 3 presents Charpy transition curves, plotting impact energy and shear area percentage versus temperature, for air-cooled, HFT-ACC, and LFT-ACC specimens. The LFT-ACC specimen shows the lowest ductile-to-brittle transition temperature (DBTT), followed by HFT-ACC, while the transition curve of the air-cooled specimen shifts to higher temperatures. The air-cooled specimen presents a lower upper-shelf energy (USE) than the two specimens subjected to ACC. No splits were observed on the fracture surfaces of the air-cooled and LFT-ACC specimens at all testing temperatures, while 1 to 2 splits were observed in the HFT-ACC specimen at -80 °C. Overall, the air-cooled specimen shows the worst Charpy toughness performance, while the LFT-ACC specimen exhibits the best low-temperature Charpy toughness among the three conditions.
Figure 3. Charpy transition curves, plotting (a) impact energy and (b) shear percentage versus temperature, of air-cooled, HFT-ACC, and LFT-ACC specimens.

Optical Microstructure
Fig. 4 shows Nital-etched optical microstructures at the quarter and mid thicknesses of the air-cooled, HFT-ACC, and LFT-ACC specimens. The air-cooled specimen exhibits a polygonal ferrite and pearlite grain structure with pearlite banding at the mid thickness. The HFT-ACC specimen reveals a predominantly acicular ferrite or bainitic ferrite grain structure at the quarter and mid thicknesses. The LFT-ACC specimen exhibits a mixture of bainitic ferrite and polygonal ferrite at both thickness locations.

Figure 4. Nital-etched optical microstructures at the quarter and mid thicknesses of the air-cooled, HFT-ACC, and LFT-ACC specimens.
Secondary Phases
Fig. 5 shows SEM secondary electron (SE) images of the air-cooled, HFT-ACC, and LFT-ACC specimens at the quarter and mid thicknesses. Large pearlite colonies are observed in the air-cooled specimen. In the HFT-ACC and LFT-ACC specimens, M/A islands and degenerate pearlite (DP) are widely dispersed. The high pearlite content and banded structure at the mid thickness of the air-cooled specimen contributed to its inferior toughness, i.e., lower USE and higher transition temperature compared to the two specimens subjected to ACC.

EBSD Results
Fig. 6 shows the average grain sizes (equivalent diameter, µm) at the quarter and mid thicknesses of the air-cooled, HFT-ACC, and LFT-ACC specimens, calculated by arithmetic mean and area-weighted mean methods. The equations for the two calculation methods are provided in Fig. 6 (c). In general, the average grain sizes calculated by the arithmetic mean are smaller than those by area-weighted mean. The area-weighted mean method, in comparison, reflects the contributions of large grains to the average grain size. Figs. 6 (a) and (b) present average grain sizes using the threshold misorientation angles of >5° and >15°, respectively, to define grain boundaries. The grain boundaries with misorientation angles of 5° to 15° are defined as low angle grain boundaries (LAGBs), while grain boundaries with misorientation angles of >15° are defined as high angle grain boundaries (HAGBs). When using the arithmetic mean, the air-cooled specimen shows coarser grain sizes with both threshold misorientation angles compared to the two specimens subjected to ACC, presenting similar grain sizes. When using the area-weighted mean and threshold misorientation angles of >5°, a similar trend is observed for the three specimens. However, HFT-ACC presents the largest area-weighted mean grain sizes with HAGBs, followed by the air-cooled specimen, while LFT-ACC shows the finest grain sizes among all three conditions.

Fig. 7 presents grain size distribution maps with HAGBs at the quarter and mid thicknesses of the three specimens. The HFT-ACC specimen exhibits the highest area fractions of coarse grains (>30 µm), contributing to its largest area-weighted mean grain sizes with HAGBs. These coarse grains manifest an elongated shape along the rolling direction. They are mainly granular bainite regions containing LAGBs of 5º to 15º inside the large grains, which are less effective in arresting crack propagation and thus unfavorable for fracture toughness. In contrast, the LFT-ACC specimen contains the highest fraction of fine grains <10 µm among all three conditions, while the air-cooled specimen shows the highest fraction of grains in the range of 10 to 30 µm. Both LFT-ACC and air-cooled specimens have <10% coarse grains (>30 µm) at the quarter and mid thicknesses. It has been reported that grain boundaries with HAGBs are more effective in redirecting crack propagation, thus significantly influencing fracture toughness. The finer effective grain size with HAGBs and narrow grain size distribution of the LFT-ACC specimen contribute to its lowest Charpy transition temperature. Conversely, the presence of high fractions of coarse granular bainite of the HFT-ACC specimen negatively impacts its fracture toughness, i.e., higher transition temperature compared to the LFT-ACC specimen. The upper shelf energies of the two specimens subjected to ACC are similar.
Figure 6. Average grain sizes at the quarter and mid thicknesses of the air-cooled, HFT-ACC, and LFT-ACC specimens using threshold misorientation angles of (a) >5° and (b) >15°; (c) Arithmetic mean and area-weighted mean equations.

Figure 7. Grain size distribution maps with threshold misorientation angle of >15° at the quarter and mid thickness locations of the air-cooled, HFT-ACC, and LFT-ACC specimens.

Fig. 8 shows grain shape aspect ratio plots of the air-cooled, HFT-ACC, and LFT-ACC specimens, using a threshold misorientation angle of >5°. The grain shape aspect ratio is defined as the long axis over the short axis of the grain when fitted...
into an elliptical shape. A smaller grain aspect ratio towards 1 indicates a polygonal grain structure. Both air-cooled and LFT-ACC specimens show higher fractions of polygonal and quasi-polygonal ferrite grains (aspect ratio 1 to 2) than the HFT-ACC specimen at the quarter and mid thicknesses. Specifically, the air-cooled specimen presents higher fractions of polygonal ferrite grains with low aspect ratio <1.5. In contrast, the HFT-ACC specimen exhibits more grains with high aspect ratios, which is due to the presence of large amounts of acicular ferrite and coarse granular bainite grains.

![Grain shape aspect ratio plots using a threshold misorientation angles of >5° at the quarter and mid thicknesses of the air-cooled, HFT-ACC, and LFT-ACC specimens.](image)

Figure 8. Grain shape aspect ratio plots using a threshold misorientation angles of >5° at the quarter and mid thicknesses of the air-cooled, HFT-ACC, and LFT-ACC specimens.

Fig. 9 presents grain boundary misorientation angle distribution plots at the quarter and mid thicknesses of the air-cooled, HFT-ACC, and LFT-ACC specimens. The HFT-ACC specimen shows the highest number of low angle grain boundaries (LAGBs, 5° to 15°) followed by LFT-ACC, while the air-cooled specimen has the lowest number of LAGBs. The HFT-ACC specimen shows the highest fraction of boundaries with misorientation angles between 50° and 60° among all three conditions. It has been reported in the literature that high angle grain boundaries >50° are associated with upper bainite, lower bainite, or martensite depending on the type of phase transformation [21]. On the other hand, the air-cooled specimen, with a dominant polygonal ferrite grain structure, shows the lowest fraction of grain boundaries between 20° and 50°. The grain boundary misorientation angle distributions of the LFT-ACC specimen at both thickness locations lie in between those of the air-cooled and HFT-ACC specimens, reflecting its mixed bainitic ferrite and polygonal ferrite grain structures.

![Grain boundary misorientation angle distribution plots of the air-cooled, HFT-ACC, and LFT-ACC specimens at the (a) quarter and (b) mid thicknesses.](image)

Figure 9. Grain boundary misorientation angle distribution plots of the air-cooled, HFT-ACC, and LFT-ACC specimens at the (a) quarter and (b) mid thicknesses.

Figs. 10 (a-c) show geometrically necessary dislocation (GND) density maps at the quarter thicknesses of the three specimens. The HFT-ACC specimen presents a more uniformly distributed dislocations, whereas the air-cooled and LFT-ACC specimens show inhomogeneous dislocation distributions. The same dislocation distributions were found at the mid thickness of the three specimens (not shown here). Fig. 10 (d) presents the corresponding phase map of the air-cooled specimen, where pearlite was classified based on the band slope and band contrast of the EBSD diffraction patterns. The higher GND densities in the air-cooled specimen are seen primarily located in the vicinities of pearlite phase. Fig. 10 (d) shows average GND densities of the three specimens at both quarter and mid thicknesses. The air-cooled specimen with a polygonal ferrite and pearlite structure exhibits the lowest average GND densities. The HFT-ACC specimen, characterized by a predominantly bainitic ferrite structure, shows higher average dislocation densities than the LFT-ACC specimen.

![Geometrically necessary dislocation (GND) density maps at the quarter thicknesses of the three specimens.](image)

Fig. 10 (a-c) show geometrically necessary dislocation (GND) density maps at the quarter thicknesses of the three specimens. The HFT-ACC specimen presents a more uniformly distributed dislocations, whereas the air-cooled and LFT-ACC specimens show inhomogeneous dislocation distributions. The same dislocation distributions were found at the mid thickness of the three specimens (not shown here). Fig. 10 (d) presents the corresponding phase map of the air-cooled specimen, where pearlite was classified based on the band slope and band contrast of the EBSD diffraction patterns. The higher GND densities in the air-cooled specimen are seen primarily located in the vicinities of pearlite phase. Fig. 10 (d) shows average GND densities of the three specimens at both quarter and mid thicknesses. The air-cooled specimen with a polygonal ferrite and pearlite structure exhibits the lowest average GND densities. The HFT-ACC specimen, characterized by a predominantly bainitic ferrite structure, shows higher average dislocation densities than the LFT-ACC specimen.

![Phase map of the air-cooled specimen showing pearlite.](image)

Fig. 10 (d) presents the corresponding phase map of the air-cooled specimen, where pearlite was classified based on the band slope and band contrast of the EBSD diffraction patterns. The higher GND densities in the air-cooled specimen are seen primarily located in the vicinities of pearlite phase. Fig. 10 (d) shows average GND densities of the three specimens at both quarter and mid thicknesses. The air-cooled specimen with a polygonal ferrite and pearlite structure exhibits the lowest average GND densities. The HFT-ACC specimen, characterized by a predominantly bainitic ferrite structure, shows higher average dislocation densities than the LFT-ACC specimen.
Figure 10. (a-c) GND density maps at the quarter thickness of the air-cooled, HFT-ACC, and LFT-ACC specimens, (d) phase map of the air-cooled specimen, and (e) comparison of average densities of GNDs at both quarter and mid thicknesses of the three conditions.

For the LFT-ACC specimen, the non-uniform distribution of GNDs is related to its mixed microstructure, which comprises of acicular ferrite, bainite, polygonal and quasi-polygonal ferrite. Figs. 11 (a) and (b) show the grain shape aspect ratio map and corresponding GND map of the LFT-ACC specimen at the quarter thickness. Polygonal and quasi-polygonal ferrite grains usually show low aspect ratios of 1 to 2, while acicular ferrite and bainite grains normally show high aspect ratios >2. A grain shape aspect ratio of 2 was used here to separate these two different grain structures. Figs. 11 (c) and (d) present the partitioned GND maps separated by the aspect ratio of 2. Figs. 11 (e) and (f) show the GND density distribution plots and grain size distribution plots of the two partitioned GND maps. The polygonal and quasi-polygonal ferrite grains exhibit an average GND density of $3.18 \times 10^{14}$ /m$^2$, which is 22% lower than the acicular ferrite and bainite grains ($4.06 \times 10^{14}$ /m$^2$). The polygonal and quasi-polygonal ferrite grains also show higher fractions of fine grains <10 µm, while acicular ferrite and bainite grains exhibit a wider grain size distribution containing coarse grains 20 to 40 µm. The mid thickness has a similar grain structure where the average GND density of the polygonal and quasi-polygonal ferrite grains is $3.67 \times 10^{14}$ /m$^2$ and that of the acicular ferrite and bainite grains is $4.01 \times 10^{14}$ /m$^2$ (not shown here). The formation of polygonal ferrite grains in the LFT-ACC specimen can be related to its lower finishing last pass temperature. As aforementioned, the measured surface temperature at the FLP of LFT-ACC was below the calculated $A_3$ temperature, indicating a possibility of two-phase rolling where polygonal ferrite grains formed and underwent plastic deformation. Given that the temperature at the core of the plate is higher than that at the surface, the bulk temperature can be close to or slightly higher than the $A_3$ temperature. Thus, dynamic transformation at the roll bite during FLP and/or early transformation after FLP and before entering ACC may also occur, both contributing to the formation of polygonal ferrite grains. The polygonal/quasi-polygonal ferrite grains subjected to two-phase rolling are expected to contain higher dislocation densities compared to those formed during and after last pass finishing. In Fig. 11 (b), some of the polygonal and quasi-polygonal ferrite grains show higher GND densities than the rest of the grains, which can be related to two-phase rolling. The formation of these polygonal/quasi-polygonal ferrite with fine grain sizes and relatively lower dislocation densities in the acicular and bainitic ferrite structure are beneficial for fracture toughness but decrease yield strength.
Figure 11. (a) Grain shape aspect ratio map and (b) GND density map of LFT-ACC at the quarter thickness; Partitioned GND maps of grains with aspect ratios of (c) 1-2 and (d) >2; (e) GND density distribution plots and (f) grain size distribution plots of the two partitioned GND density maps shown in (c) and (d).

Texture

Fig. 12 shows the orientation distribution function (ODF) maps at $\phi_2=0^\circ$ and $45^\circ$ of the air-cooled, HFT-ACC, and LFT-ACC specimens at the quarter and mid thicknesses. The LFT-ACC specimen shows the lowest maximum texture intensity, followed by the air-cooled specimen, while the HFT-ACC specimen exhibits the strongest texture. In general, the maximum texture intensities at the mid thickness are higher than those at the quarter thickness. The air-cooled specimen presents dominant transformed copper I and II textures at both the quarter and mid thicknesses. The HFT-ACC specimen presents a predominant $\{332\}<113>$ texture at the quarter thickness, while the dominant texture at the mid thickness is a rotated cube texture $\{001\}<110>$. For the LFT-ACC specimen, a dominant Goss $\{110\}<001>$ texture is seen at the quarter thickness. The texture at the mid thickness of LFT-ACC is similar to that of HFT-ACC, showing dominant rotated cube $\{001\}<110>$.

It has been reported that the $\{001\}<110>$ rotated cube texture exhibits the most pronounced mechanical anisotropy, which is detrimental to fracture toughness [23,24]. Additionally, the $\{001\}$ plane family has been linked to the occurrence of delamination/splits on fracture surfaces of linepipe steels, as it is the weakest cleavage plane for the BCC structure [14]. In contrast, the $\{332\}<113>$ component demonstrates low anisotropy and proves beneficial for fracture toughness [23,24]. Fig. 13 presents the plots of the $\alpha$, $\gamma$, and $\epsilon$-fiber textures of the three conditions. In the $\alpha$-fiber plots, HFT-ACC shows the highest intensities of rotated cube texture among all three samples at both quarter and mid thicknesses. The three specimens exhibit similar $\gamma$-fiber textures. In the $\epsilon$-fiber plots, HFT-ACC shows the highest intensities of $\{332\}<113>$ texture; it surpasses the rotated cube texture at the quarter thickness but is lower than the rotated cube at the mid thickness. The strong rotated cube texture of the HFT-ACC specimen contributes to the observed splits on the fracture surfaces of the CVN specimens tested at -80 °C.
Figure 12. ODF maps at $\varphi_2=0^\circ$ and $45^\circ$ of the air-cooled, HFT-ACC, and LFT-ACC specimens at the quarter and mid thicknesses.

Figure 13. Plots of $\alpha$-, $\gamma$-, and $\varepsilon$-fiber textures of the air-cooled, HFT-ACC, and LFT-ACC specimens at the quarter and mid thicknesses.

Goss texture is observed at the quarter thickness of both air-cooled and LFT-ACC specimens but not at the mid thickness. For the HFT-ACC specimen, Goss texture is not evident at either the quarter or mid thickness. It has been reported in the literature that the formation of Goss texture is associated with the nucleation of grains within shear bands [25] and in the coarse initial austenite grain structures during hot rolling [26]. Additionally, Goss texture is typically found at the surface of steel plates due to high shear deformation induced by the friction between the rolls and the steel [27]. The formation of the Goss texture at the quarter thickness of the LFT-ACC and air-cooled specimens can be related to the lower finishing temperatures and higher shear
strain penetration. Specifically, the dominant Goss texture formed at the quarter thickness of the LFT-ACC specimen can also be attributed to Transfer-Bar Cooling where the surface temperature rapidly dropped, resulting in a harder surface and higher shear strain penetration to the quarter thickness. In the α-fiber plots at the quarter thickness, the peaks of the HFT-ACC and air-cooled specimens are found around the transformed copper components, while the LFT-ACC specimen presents a small peak at {223}<110>. Literature showed that when rolling in the two-phase region, the peak in the α-fiber shifted from rotated cube/transformed Cu to {223}<110> [38]. This suggests that two-phase rolling has occurred at the quarter thickness of the LFT-ACC specimen.

Effect of Microstructure and Texture on Mechanical Properties
The yield strength of low carbon micro-alloyed steels can be assessed by summarizing the contributions from various strengthening mechanisms, as expressed in Equation (5) [29].

$$\sigma_{YS} = \sigma_0 + \Delta\sigma_{SS} + \Delta\sigma_{ppt} + \Delta\sigma_{dis} + \Delta\sigma_d + \Delta\sigma_{pearlrite} + \Delta\sigma_{text}$$

(5)

Here, $\sigma_0$ is the ferrite lattice friction stress, $\Delta\sigma_{SS}$ and $\Delta\sigma_{ppt}$ are the contributions by solid solution and precipitation strengthening, $\Delta\sigma_{dis}$ and $\sigma_d$ are the contributions of dislocations and grain boundaries, $\Delta\sigma_{pearlrite}$ is the contribution of pearlite, and $\Delta\sigma_{text}$ is the contribution of texture to the strength. $\Delta\sigma_{d}$ can be calculated according to the Hall-Petch relationship, where $d$ is the average grain size determined by the mean linear intercept method. It has been reported that grain boundaries with low misorientation angle of $>2^\circ$ are effective in contributing to yield strength by influencing dislocation mean free path [39], whereas high angle grain boundaries ($>15^\circ$) are more effective in impacting fracture toughness.

The HFT-ACC specimen exhibited the highest YS along the TRD among all three conditions, which was 15 MPa higher than the LFT-ACC specimen. For the two ACC specimens, the friction stress and solid solution strengthening are assumed to be the same given that they were both produced from the same slab. The HFT-ACC and LFT-ACC specimens exhibited similar arithmetic mean grain sizes with a misorientation angle of $>2^\circ$. Hence, the grain boundary strengthening effects of the two specimens are expected to be comparable. A review of the literature suggested that in Nb-Ti-V micro-alloyed steels, high cooling rates ($\geq 20 ^\circ C/s$) could suppress nano-precipitate formation in ferrite during or after transformation and keep the microalloying elements in solution [30]. This can be the case for the HLT-ACC specimen in this study. For the LFT-ACC specimen, two-phase rolling at the finishing last pass could lead to nano-precipitate formation in the ferrite, potentially enhancing YS. However, the extent of this enhancement depends on the amount of Nb and V remaining in solution after hot rolling in austenite. Nevertheless, the GND densities of the HFT-ACC specimen are 24% higher than those of LFT-ACC at both quarter and mid thicknesses, which primarily contributed to the higher YS of the former. The stronger texture developed in the HFT-ACC specimen resulted in higher strength anisotropy compared to the LFT-ACC specimen, which could also contribute to the observed YS difference.

The LFT-ACC specimen exhibited a 12 MPa higher YS than the air-cooled specimen. The LFT-ACC steel has a 0.20% higher Mn content, giving a minor strength increase of 0.9 ksi (6 MPa) over the air-cooled steel calculated by Pickering’s equation [31]. The air-cooled specimen exhibited much larger grain sizes, measuring 35% and 21% coarser at the quarter and mid thicknesses, respectively, in comparison to the LFT-ACC specimen. Additionally, the former showed 46% and 30% lower densities of GNDs at the quarter and mid thicknesses, respectively, compared to the latter. Both the coarser grain sizes and lower GND densities contribute to the relatively lower YS observed in the air-cooled specimen compared to the LFT-ACC specimen. A review of the literature suggested that in ferrite-pearlite steels, <20% pearlite does not significantly contribute to strength [17]. Thus, the presence of <5% pearlite in the air-cooled specimen is expected to have a limited impact on YS. Based on the JMatPro calculations, the LFT-ACC and air-cooled specimens have similar equilibrium fractions of carbonitride precipitates (Figs. 14 (a) and (b)). Figs. 14 (c) and (e) show the elemental distribution of M(C,N) and MN precipitates as a function of temperature in the two steels. Here, M denotes metal elements, C represents carbon, and N represents nitrogen. The MN precipitates are primarily titanium nitride (TiN) for both alloys. The M(C,N) precipitates are predominantly Nb-rich carbonitrides (Nb(C,N)) at temperatures above ~800 °C. Strain-induced Nb(C,N) precipitates are expected to form in both alloys during hot rolling. At temperatures below 800 °C, the M(C,N) precipitates are mainly carbon-rich Nb,V(C,N). It has been reported in the literature that the cooling rate affects nano-precipitate formation in ferrite during transformation [30]. Nb, V)C nano-precipitates are kinetically more favorable to form in ferrite during air cooling compared to accelerated water cooling [32]. As aforementioned, the LFT-ACC specimen may contain strain-induced nano-precipitates formed in ferrite due to two-phase rolling. Therefore, a scanning transmission electron microscopy (STEM) analysis is necessary to properly quantify the strengthening contributions due to Nb,V(C,N). Additionally, the air-cooled specimen showed a stronger texture and thus a higher anisotropy in YS than the LFT-ACC specimen, which could also affect its YS.
The LFT-ACC specimen with a well-balanced mixture of polygonal/quasi-polygonal ferrite, acicular ferrite, and fine and uniformly dispersed M/A islands exhibited a superior toughness, evident in the lowest Charpy transition temperature among all three conditions and a high level of USE. This can be related to its finest effective grain size with HAGBs, a low fraction of coarse grains resulting in a narrow grain size distribution, a lower dislocation density, and a weak texture. In contrast, the HFT-ACC specimen, characterized by a predominant bainitic ferrite structure, shows a higher transition temperature compared to the LFT-ACC specimen. This is primarily associated with the high fraction of coarse granular bainite grains, a wide grain size distribution, and a strong texture, especially the high intensity of the rotated cube component. The air-cooled specimen showed inferior toughness performance compared to the two ACC specimens with a higher transition temperature and a lower USE. This is attributed to the presence of a high pearlite content, especially the banded pearlite structure at the mid thickness, and a coarse effective grain size associated with the polygonal ferrite grains.

CONCLUSIONS

1. The microstructure and texture results of the LFT-ACC specimen indicated a two-phase rolling and/or early transformation during finishing last pass, resulting in a mixture of acicular ferrite and quasi-polygonal/polygonal ferrite grain structure through the thickness. The superior Charpy toughness achieved in the LFT-ACC specimen, evident by its lowest transition temperature among all three conditions and a high level of USE, is attributed to the smallest effective grain size with HAGBs, fine and uniformly dispersed M/A islands, a relatively low dislocation density, and a weak texture.

2. The HFT-ACC specimen with a predominant acicular ferrite and granular bainite structure exhibited the highest yield strength, mainly due to its high dislocation density. However, the HFT-ACC specimen showed a higher transition temperature and a similar USE compared to the LFT-ACC specimen, which is mainly due to a higher fraction of coarse granular bainite grains, a higher dislocation density, and a stronger texture, particularly due to the rotated cube component.

3. The air-cooled specimen showed a lower YS and inferior Charpy toughness, i.e., a higher transition temperature and lower USE, compared to the two specimens subjected to ACC. The lowest YS of the air-cooled specimen can be related to its larger grain size and lower dislocation density. The presence of a higher volume fraction of pearlite, especially the formation of pearlite bands at the mid thickness, along with a coarse effective grain size with HAGBs and a relatively strong texture, contributed to the poor toughness of the air-cooled specimen.
ACKNOWLEDGEMENTS

The authors would like to thank rolling mill personnel and test lab staff, as well as R&D analysts at SSAB Americas for their great support in conducting mill trials and providing test data. The authors are also grateful to Dr. Sunday Abraham for his critical review of the manuscript, and SSAB Americas senior management for their permission to publish this work.

REFERENCES


