Microstructure Control and Toughness of Heavy-Gauge Coiled Plate Produced by the Compact Strip Process

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ABSTRACT

The production of heavy-gauge, high toughness coiled plate by the compact strip process has historically been limited by both cast slab thickness and thermo-mechanical processing capability of stand-alone tandem finishing. By decoupling the roughing and finishing rolling stages of an intermediate thickness slab (4.7" - 5.1") in a manner similar to traditional hot strip rolling, and through optimization of both the microalloying strategy and transfer bar cooling practices, the strength-toughness balance of high strength, low alloy coiled plate is markedly improved for heavy gauges (0.625" - 1.00"). The effect of transfer bar cooling practices on impact toughness is dependent on alloy design. The effects of both reduction and thermal schedules with respective microalloying strategies with niobium and titanium are supported by austenite conditioning modeling. Beneficial effects on strength-toughness balance are observed with titanium additions, presumably due to both a reduction in solute nitrogen prior to hot rolling and limitation of post-recrystallization grain growth prior to finish rolling.

INTRODUCTION

The production of heavy-gauge, hot-rolled carbon-manganese (CMn) and high-strength low-alloy (HSLA) steels is well established in conventional hot strip and plate mills supplied by thick slab continuous casters (6” thickness or greater). In contrast, production of heavy-gauge coiled plate products (greater than 0.375”) with elevated toughness requirements has been limited in compact strip mills due to limitation in total available reduction necessary to achieve optimum processing and associated microstructures. Additionally, compact strip mills must contend with design limitations placed by the necessary avoidance of the peritectic solidification range, elevated nitrogen levels resulting from EAF steelmaking practices, and reduced equalizing furnace temperatures relative to conventional reheating furnaces.

Stated broadly, an optimized alloy and process design maximizes the strengthening effect of grain refinement and limits the detrimental effects on toughness from structural heterogeneities, non-metallic inclusions, and solute nitrogen. Additional strengthening mechanisms such as precipitation and solid solution strengthening must be similarly limited within the constraint of product usage requirements, i.e., strength, ductility, weldability, etc., as these mechanisms reduce toughness as a function of strength [1-7]. Achieving a sufficiently refined structure with limited heterogeneity within the gauge range of interest is further complicated in single-stage tandem finishing operations by an increased difficulty to decouple the necessary austenite conditioning stages of static recrystallization and strain retention prior to austenite decomposition on the runout table [8-9]. Therefore, modern CSPs employing single stage tandem rolling must rely on complex interstand cooling strategies within the tandem finishing mill to ensure that sufficient reduction in both recrystallization regimes is obtained for product requirements. Alternatively, a compact strip mill more closely resembling conventional hot strip mills with respect to slab thickness, decoupled roughing and finishing operations, and controlled cooling of the transfer bar following roughing provides both metallurgical and practical design advantages for the production of heavy-gauge, high toughness coiled plate products. To further this assertion, a brief review of the necessary stages of austenite conditioning prior to runout table cooling is offered.

Following continuous casting and temperature equalization within the tunnel furnace of a CSP, it is necessary for both the refinement and homogeneity of austenite to remove the as-cast structure via one or more cycles of complete static recrystallization. This stage of austenite conditioning, generally referred to as roughing, is intended to prepare the austenite to
a sufficiently refined and equiaxed state prior to application of finish rolling. Failure to achieve a sufficiently fine and uniform austenite microstructure following roughing results in a general degradation of the potential strength-ductility balance of the finished product [6,10]. Recrystallization and grain growth response of the steel following roughing reduction(s) is governed principally by temperature, deformation strain, and alloy content. Accordingly, casting and equalizing furnace practices must be properly specified to ensure correct and consistent microalloy content - in addition to bulk alloy content that remains largely independent of slab temperature - is available to limit post-recrystallization grain growth following each roughing reduction [11-12]. Extensive reporting on post-recrystallization grain grown suppression, a process generally referred to as recrystallization controlled rolling, is available based on previous study, and it is noted briefly that such suppression may be accomplished both by the bulk and microalloy additions in solution as well as the presence of fine carbonitride particles, primarily titanium nitride, evolved at temperatures greater than roughing reduction temperatures [13-17]. An excessive suppression of recrystallization following roughing is unfavorable and results in a non-uniform austenite grain distribution entering finishing. Conversely, a lack of grain size control following roughing may result in excessive post-recrystallization grain growth and general degradation of the strength-ductility balance potential of the steel. To summarize, the goal of the roughing process when maximizing the strength-ductility balance is to achieve the finest recrystallized austenite grain size possible within the constraints of product requirements. Additionally and to enable both greater alloy utilization and process design flexibility with this aim, transfer bar temperature and associated recrystallization and grain growth kinetics may be reduced independently following roughing by means of a transfer bar cooler when alloy effects are not sufficient to achieve the desired refinement prior to finish mill entry.

After conditioning of the transfer bar, finish rolling may be conducted with one of two primary strategies. Firstly, the recrystallization controlled rolling strategy employed in the roughing process may be repeated by rolling to finish gauge above the recrystallization limit temperature (RLT), i.e., the temperature at which recrystallization is at least partially suppressed. Alternatively and most commonly for microalloyed steels, finish rolling is conducted below the recrystallization stop temperature (RST) to ensure sufficient strain retention prior to cooling from austenite on the runout table [18-20]. Recrystallization response of the steel following deformation is again principally dictated, in combination with the degree and rate of deformation and temperature, by alloy content and the effects of both solute and dynamically precipitated microalloy carbonitrides during the rolling process. This process is generally referred to as controlled rolling. As is the case with roughing, the intent of the deformation below RST is to ensure consistent structure and retained strain within the strip and to avoid the emergence of structural heterogeneities and to ensure consistency and refinement of the transformed ferrite structure via well-documented transformation phenomena [8-9].

The current contribution presents an industrial implementation of the previously outlined metallurgical concepts for production of high toughness, heavy-gauge coiled plate from an intermediate slab thickness utilizing a hot rolling process equipped with both decoupled roughing and finish rolling stages and transfer bar cooling capabilities. The hot rolling response, microstructure, and associated mechanical performance of selected alloys are evaluated with specific focus on the effects of microalloy strategy and transfer bar cooling practice.

**EXPERIMENTAL**

**Steel Design and Production**

Three steels were selected to observe the combined effects of transfer bar cooling practice and microalloy strategy on the strength-toughness balance of coiled plate produced by the compact strip process. Table I outlines the selected alloy chemistries and associated industry specification. A low carbon-manganese structural steel (CMn) was chosen for a baseline for comparison of microstructure response to transfer bar cooling to those of niobium (Nb) and niobium-titanium (NbTi) microalloyed steels.

<table>
<thead>
<tr>
<th>Steel</th>
<th>C (max)</th>
<th>Mn (max)</th>
<th>Si (max)</th>
<th>P (max)</th>
<th>S (max)</th>
<th>Nb (max)</th>
<th>Ti (max)</th>
<th>Al (max)</th>
<th>N (max ppm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CMn</td>
<td>0.05-0.07</td>
<td>0.80-0.90</td>
<td>0.04</td>
<td>0.02</td>
<td>50</td>
<td>Res.</td>
<td>Res.</td>
<td>0.02-0.04</td>
<td>110</td>
</tr>
<tr>
<td>Nb</td>
<td>0.05-0.07</td>
<td>0.85-0.95</td>
<td>0.04</td>
<td>0.02</td>
<td>50</td>
<td>0.015-0.025</td>
<td>Res.</td>
<td>0.02-0.04</td>
<td>110</td>
</tr>
<tr>
<td>NbTi</td>
<td>0.05-0.07</td>
<td>1.00-1.10</td>
<td>0.04</td>
<td>0.02</td>
<td>50</td>
<td>0.020-0.030</td>
<td>0.010-0.020</td>
<td>0.02-0.04</td>
<td>110</td>
</tr>
</tbody>
</table>

Production heats of 210 short tons were produced through an electric arc/ladle metallurgy furnace routing and were continuously cast into 4.7” (120 mm) thick slabs with liquid core and soft reduction to minimize centerline segregation. As outlined in Figure 1, cast slabs were then hot rolled following equalization through a two-stand tandem roughing mill. Following roughing, transfer bar cooling was applied to a subset of each alloy chemistry. Thereafter, a second equalization
treatment is performed prior to finish rolling in a 6-stand tandem mill, cooling, and coiling. Table II outlines relevant target hot mill processing parameters for selected gauges and alloy/process variants through coiling, and it should be noted that the heaviest gauge steels were subjected to four (4) rather than six (6) reductions in the finishing mill. Production variations in casting speed, mill pacing and hot mill model adaptations are not accounted for in the listed targets, and potential effects of such variation are considered in the context of experimental results. Full width samples of hot rolled coils were collected at a minimum of 20 feet following less than one (1) hour of cooling from the coil tail for mechanical properties and microstructural studies. Prior to manufacture, both microalloy solubility and austenite conditioning response were evaluated to ensure both full austenite recrystallization in roughing and a minimization of partial recrystallization in finishing for the microalloyed steel, Nb and NbTi [12,21]. Simulations were conducted with MicroSim-DSP® software customized to SDI Sinton’s CSP configuration for subsequent comparison to produced microstructure and associated mechanical properties. Fundamentals of this simulation approach are outlined in previous publication [22,23].

<table>
<thead>
<tr>
<th>Steel</th>
<th>TF1 Exit Temp (°C)</th>
<th>Roughing Mill Reduction (%)</th>
<th>TBC Entry Temp (°C)</th>
<th>TF2 Entry Temp (°C)</th>
<th>TF2 Exit Temp (°C)</th>
<th>Finishing Mill Reduction (%)</th>
<th>Finishing Temp (°C)</th>
<th>Coiling Temp (°C)</th>
<th>Target Gauge (in.)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CMn</td>
<td>R1: 1100</td>
<td>30</td>
<td>R2: 35</td>
<td>-</td>
<td>-</td>
<td>1025</td>
<td>37</td>
<td>32</td>
<td>27</td>
</tr>
<tr>
<td>CMn-TBC</td>
<td>R1: 1100</td>
<td>30</td>
<td>R2: 35</td>
<td>1050</td>
<td>925</td>
<td>1000</td>
<td>30</td>
<td>-</td>
<td>27</td>
</tr>
<tr>
<td>Nb</td>
<td>R1: 1125</td>
<td>30</td>
<td>R2: 35</td>
<td>-</td>
<td>-</td>
<td>1025</td>
<td>27</td>
<td>-</td>
<td>22</td>
</tr>
<tr>
<td>Nb-TBC</td>
<td>R1: 1125</td>
<td>30</td>
<td>R2: 35</td>
<td>1025</td>
<td>900</td>
<td>1000</td>
<td>27</td>
<td>-</td>
<td>22</td>
</tr>
<tr>
<td>NbTi-TBC</td>
<td>R1: 1125</td>
<td>30</td>
<td>R2: 35</td>
<td>1025</td>
<td>900</td>
<td>1000</td>
<td>27</td>
<td>-</td>
<td>22</td>
</tr>
</tbody>
</table>

Mechanical Evaluation

Uniaxial tensile specimens were extracted from experimental steels and evaluated according to ASTM E8 in the transverse orientation relative to rolling direction from the center of the width [24]. Sample orientation was determined based on end-use specification requirements. Yield strength, ultimate tensile strength, uniform and total elongations for all tested specimens were recorded for a minimum of 10 specimens per alloy/rolling strategy combination.

Full-size Charpy impact specimens were machined from experimental steels and tested according to ASTM E23 in either the longitudinal or transverse orientation relative to rolling direction from the width quarter-point [25]. Sample temperature and orientation for testing, 30 °F and transverse for CMn and -20 °F and longitudinal for Nb and NbTi, was determined based on end-use requirements. Charpy impact energies for all tested specimens were recorded for a minimum of 10 specimens per
alloy/rolling strategy combination. For clarity, a sample orientation schematic is shown in Figure 2 for both uniaxial tension and Charpy impact tests.

Figure 2: Sample Orientation Schematic. Sample nomenclature based on long axis of sample.

Microstructure Evaluation
Metallography specimens were obtained by sectioning experimental steels in either the longitudinal or transverse direction based on the associated Charpy specimen orientation. Samples were then mounted and polished in accordance with standard metallographic techniques. A 2 vol pct nital solution was used to etch the polished specimens, and microstructural constituents, notably primary ferrite and lower temperature transformation products containing fractions of pearlite, bainitic or Widmanstätten ferrite, were evaluated using comparative methods for area fraction estimations. Mean ferrite grain size was determined according to ASTM E112 [26]. Mean aspect ratio of and size of non-polygonal ferrite microconstituents were also evaluated. A minimum of ten (10) fields of view were used for each quantitative microstructure evaluation with the aid of ImageJ image analysis software.

RESULTS

Mechanical Properties
Uniaxial tensile properties of the experimental hot-rolled steels are shown in Table III, where \( N \) denotes the number of tests performed for each experimental condition. Transfer bar cooling practices display no statistically significant effect on the average yield strengths (\( \langle \sigma_Y \rangle \)), tensile strengths (\( \langle \sigma_{UTS} \rangle \)), and elongations (\( \langle El. \rangle \)) of the respective steel grades relative to baseline practices without transfer bar cooling. Standard deviations of the respective distributions are listed as well and denoted by ‘\( \Delta \)’. Addition of Ti to the Nb-microalloyed steel results in a reduction in both yield and tensile strength of the NbTi steel relative to the Nb steel with transfer bar cooling. All produced coils conformed to the respective industry specification.

Table III. Uniaxial Tensile Properties of Experimental Steels

<table>
<thead>
<tr>
<th>Steel</th>
<th>Orientation</th>
<th>Gauge (in.)</th>
<th>( N )</th>
<th>( \langle \sigma_Y \rangle ) (ksi)</th>
<th>( \Delta \sigma_Y ) (ksi)</th>
<th>( \langle \sigma_{UTS} \rangle ) (ksi)</th>
<th>( \Delta \sigma_{UTS} ) (ksi)</th>
<th>( \langle El. \rangle ) (%)</th>
<th>( \Delta El. ) (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CMn</td>
<td>T</td>
<td>0.410</td>
<td>84</td>
<td>53</td>
<td>1.9</td>
<td>65</td>
<td>1.5</td>
<td>36</td>
<td>3.1</td>
</tr>
<tr>
<td>CMn – TBC</td>
<td>T</td>
<td>0.625-0.875</td>
<td>18</td>
<td>53</td>
<td>2.0</td>
<td>64</td>
<td>1.3</td>
<td>43</td>
<td>2.2</td>
</tr>
<tr>
<td>Nb</td>
<td>T</td>
<td>1.00</td>
<td>85</td>
<td>63</td>
<td>2.9</td>
<td>74</td>
<td>1.9</td>
<td>38</td>
<td>2.6</td>
</tr>
<tr>
<td>Nb - TBC</td>
<td>T</td>
<td>1.00</td>
<td>34</td>
<td>65</td>
<td>2.4</td>
<td>73</td>
<td>1.7</td>
<td>40</td>
<td>1.9</td>
</tr>
<tr>
<td>NbTi - TBC</td>
<td>T</td>
<td>1.00</td>
<td>6</td>
<td>56</td>
<td>1.6</td>
<td>66</td>
<td>1.2</td>
<td>45</td>
<td>1.8</td>
</tr>
</tbody>
</table>

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Figures 3(a)-(c) show Charpy impact data obtained for experimental steels presented for determination of the effect of transfer bar cooling and alloy strategy. Table IV tabulates the average (\(\langle CVN\rangle\)) and standard deviation (\(\Delta CVN\)) of Charpy impact energy distributions shown in Figure 3. Again, \(N\) denotes the number of tests performed for each experimental condition.

Figure 3(a) shows the concurrent increase and reduction in variation in Charpy impact energy for the CMn steel. It should be additionally noted that the gauges compared are not equal. Total reduction during the hot rolling process is greater for the samples tested with no transfer bar cooling, a variance that would generally indicate the potential for greater overall grain refinement (and associated greater impact toughness) for the lighter gauge steel. Here, however, significantly greater average impact energies are observed for the heavier gauge steels which underwent transfer bar cooling.

Figure 3(b) shows a moderate increase in CVN impact energy for the Nb steel (3/4” thickness) with application of transfer bar cooling. Significant deviation in the sample set is observed, however, in Nb steels both with and without transfer bar cooling, a behavior which will be addressed subsequently with regard to predicted recrystallization behavior in the finishing mill.

Lastly, Figure 3(c) again shows a moderate increase in CVN impact energy for the Nb (1” thickness) with the application of transfer bar cooling. Deviation remains a concern in the tested steels, and a general decrease in the CVN impact energy is observed in the 1” specimens relative to the 3/4” specimens, as expected due to smaller hot reduction. Addition of Ti in the 1” NbTi steel is shown to markedly improve both the average impact energy as well as the deviation. The presumed effect of Ti is discussed in subsequent sections.

Microstructure

Microstructures of the experimental hot rolled steels were observed by light optical microscopy, and representative micrographs are shown in Figures 4 and 5 for samples processed without and with transfer bar cooling, respectively. The CMn alloy presents a primarily polygonal ferrite microstructure, and the HSLA steels present mixed ferrite grain structures with both equiaxed polygonal grains as well as ferrite grains elongated in the rolling direction. Additionally, significant fractions of low temperature transformation products (LTTPs), presumably mixtures of quasi-polygonal, Widmanstätten ferrite, and bainitic ferrite, are present in the HSLA steels. The prevalence of LTTPs is greater in the Nb steel in comparison to the NbTi steel. Results of quantitative microstructural analyses are summarized below in Table V.

Calculations of microalloy solubility indicate that all Nb remains in solution prior to roughing at the target equalization temperatures. Ti additions in the NbTi steel, however, are consumed to stoichiometry with N, resulting in negligible solute Ti prior to hot rolling. MicroSim-DSP® simulations of austenite conditioning in Nb and NbTi steels utilizing the target parameters shown in Table II indicate full recrystallization of the deformed as-cast structure during roughing and small fractions of partial recrystallization in the early finishing passes. Using the actual mill parameters obtained following production, however, MicroSim-DSP® simulations indicate the potential for both retained as-cast fraction and partial recrystallization in the experimental setups selected for additional microscopy analysis. Representative results for recrystallized austenite fraction in Nb and NbTi steels are shown in Figure 6. Here, temperature and reduction in R1 and R2 (roughing passes) are critical variables to ensure full recrystallization and elimination of the as-cast austenite structure, and variation to lower values of temperature and reduction from those specified in setup are shown to result in the potential for retention of as-cast fractions (non-recrystallized austenite) into runout table cooling.
Figure 3. Charpy impact energy for experimental steels with and without application of TBC. Impact energies for 0.410-0.875” CMn steels tested at 30 °F are shown in (a) for the transverse samples. Impact energies for 0.750” Nb steels tested at -20 °F are shown in (b) for the longitudinal samples. Impact energies for 1” Nb and NbTi steels tested at -20 °F are shown in (c) for the longitudinal samples.

Table IV. Charpy Impact Properties of Experimental Steels

<table>
<thead>
<tr>
<th>Steel</th>
<th>Orientation</th>
<th>Gauge (in.)</th>
<th>N</th>
<th>(CVN) (ft-lb)</th>
<th>ΔCVN (ft-lb)</th>
</tr>
</thead>
<tbody>
<tr>
<td>CMn</td>
<td>T</td>
<td>0.410</td>
<td>42</td>
<td>185</td>
<td>68</td>
</tr>
<tr>
<td>CMn – TBC</td>
<td>T</td>
<td>0.625-0.875</td>
<td>48</td>
<td>270</td>
<td>10</td>
</tr>
<tr>
<td>Nb</td>
<td>L</td>
<td>0.750/1.00</td>
<td>54/12</td>
<td>170/19</td>
<td>106/22</td>
</tr>
<tr>
<td>Nb - TBC</td>
<td>L</td>
<td>0.750/1.00</td>
<td>66/18</td>
<td>205/60</td>
<td>99/33</td>
</tr>
<tr>
<td>NbTi - TBC</td>
<td>L</td>
<td>1.00</td>
<td>14</td>
<td>278</td>
<td>8</td>
</tr>
</tbody>
</table>
### Table V. Microstructural Parameters Determined by Analysis of LOM Images

<table>
<thead>
<tr>
<th>Steel</th>
<th>Gauge (in.)</th>
<th>ASTM Grain Size</th>
<th>Area Fraction of LTTP (%)</th>
<th>Average Feature Size (Long Axis) of LTTP (μm)</th>
<th>Average Aspect Ratio of LTTP</th>
</tr>
</thead>
<tbody>
<tr>
<td>CMn</td>
<td>0.410</td>
<td>9.3</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>CMn – TBC</td>
<td>0.875</td>
<td>9.7</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Nb</td>
<td>1.00</td>
<td>10.4</td>
<td>11</td>
<td>273</td>
<td>5.1</td>
</tr>
<tr>
<td>Nb - TBC</td>
<td>1.00</td>
<td>10.7</td>
<td>7</td>
<td>444</td>
<td>6.4</td>
</tr>
<tr>
<td>NbTi – TBC</td>
<td>1.00</td>
<td>10.3</td>
<td>&lt;5</td>
<td>458</td>
<td>6.3</td>
</tr>
</tbody>
</table>

Figure 4. Representative light optical micrographs of the (a) base CMn, and (b) Nb steels following hot rolling *without* transfer bar cooling. Indications of low temperature transformation products are indicated by red ellipses.
DISCUSSION

Effect of Transfer Bar Cooling

Principally, transfer bar cooling is employed to slow both recrystallization and the post-recrystallization grain growth of austenite prior to the onset of finish rolling. Additionally, it may be used to control the finish mill entry temperature for prevention of partial recrystallization during finish rolling if mill practice demands. As such, the expected outcome when employed in a fully recrystallized transfer bar would be for the achieved refinement and microstructural homogeneity to translate to the finished ferritic product if additional processing variables are held equal.

Acquired data on the subject experimental steels, CMn and Nb, indicate no statistically significant effect of transfer bar cooling practice on uniaxial tensile properties. Both microstructural refinement and associated Charpy impact energies, however, show improvement in direct comparison for both the CMn and Nb steels. The improvement in Charpy impact energy - and especially the variation in this measurement - is particularly significant with transfer bar cooling for the CMn steel. The improvement in impact energy for the Nb steel at ¾” and 1” is less distinct by comparison. This discrepancy in effect of transfer bar cooling is attributed in part to the relative effect of temperature on the recrystallization and grain growth kinetics of the CMn steel relative to that of the Nb microalloyed steel. Utilizing post-production data, austenite recrystallization following deformation in R2 is
simulated as complete in the CMn steel as it is in the Nb steel as shown in Figure 6. Niobium, however, is independently effective regardless of thermal practice at reducing the relative coarsening kinetics of austenite following recrystallization, and for this reason the Nb steel is likely less sensitive to thermal pathway during cooling at temperatures near the recrystallization limit temperature (RLT) which simulations indicate is near the TBC entry temperature. The CMn steel, in contrast, has a calculated RLT approximately 70 °C lower than the Nb steel and therefore would require considerable cooling to effectively arrest post-recrystallization grain growth.

**Effect of Steel Composition & Rolling Practice**

The primary variation in impact behavior on the base CMn is attributed to transfer bar cooling practice, but the use of transfer bar cooling is shown to have subsequent effects in the recrystallization response of the steel in finishing. Simulations using post-production rolling data indicate a significantly greater amount of partial recrystallization in the CMn steel without TBC relative to that with TBC. This is a result of higher finishing mill entry temperature. Based on this calculation, it is presumed that the greater degree of partial recrystallization in the CMn steel without TBC is, at least in part, the cause for the greater scatter in toughness data shown in Figure 3(a) as outlined mechanistically previously [6,10].

The marginal effect of TBC practice in the Nb steel has been discussed, but the absolute variability in the Charpy impact data requires evaluation. As shown in Figure 6, considerable amounts of partial recrystallization are observed in the Nb steel. Non-uniformity of the austenite generated from the finishing mill practice is, therefore, the presumed cause of the fraction of LTTPs observed the Nb steel as shown in Figures 4(b) and 5(b). Although simulations predict complete recrystallization of austenite following deformation in R2, a distinct possibility exists, given the proximity of R2 entry temperature to the RLT of the Nb steel, that unrecrystallized fractions of the as-cast structure may remain in the Nb steel, similarly to those observed in the NbTi steel. Regardless of the source of the austenite heterogeneity, however, it is evident in the observed structures that contributes to the variability of the Charpy impact toughness in the Nb steel, both with and without transfer bar cooling.

The greatest contrast in impact toughness due to steel composition is observed in comparison of the Nb and NbTi steels. Unlike the Nb steel, the NbTi steel exhibits an increased and consistent Charpy impact toughness. Similar effects have been observed in previous study, and the mechanism for the positive effect of Ti is generally attributed to both the removal of solute nitrogen and refinement of the recrystallized austenite following roughing due to the presence of fine TiN, especially in CSP products [16,17,27]. Additionally, it is shown in Figures 4 and 5 and Table V that the observed area fraction of LTTPs is reduced in the NbTi steel relative to the Nb steel, and the general microstructure in the NbTi steel appears more polygonal. Referring to Figure 6, a reduced amount of partial recrystallization is predicted in finish rolling of the NbTi steel relative to the Nb steel, and this is a potential cause of the observed toughness improvements. It must be acknowledged that retention of the as-cast austenite structure is predicted to be greater in the NbTi steel relative to the Nb steel. Despite this, however, mechanical results do not indicate a deterioration or variability in the impact toughness of the NbTi steel. Lastly and in addition to the microstructural studies described in this contribution for correlation to mechanical performance, additional work is ongoing to characterize the potential effects of recrystallization texture on impact toughness of the Nb and NbTi steels.

In summary, complete recrystallization of the deformed as-cast structure during roughing requires greater combinations of temperatures, roughing reduction, and inter-pass time than those trialed in the subject steels. Alternatively, transfer bar cooling may be employed at reduced alloy levels and similar deformation conditions to arrest post-recrystallization grain growth effectively following a complete recrystallization cycle. Transfer bar cooling may then be considered as an additional and decoupled control variable for austenite conditioning within the compact strip process.

**CONCLUSIONS**

The production of heavy-gauge coiled plate by the compact strip process is demonstrated by decoupling the roughing and finishing rolling stages of an intermediate thickness slab with the following conclusions:

- Both transfer bar cooling practices and alloying strategy are shown to influence the strength-toughness balance of subject steels, and the beneficial effect of transfer bar cooling on impact toughness is most pronounced in the base carbon-manganese steel.
- Additional beneficial effects on toughness are observed with a dual niobium-titanium microalloying strategy, presumably due to both a reduction in solute nitrogen prior to hot rolling and a reduction of the austenite heterogeneity during hot rolling.
- Austenite heterogeneity is presumed in part as the cause of low temperature transformation products in the finished steels.
- Austenite conditioning simulation utilizing post-production process data supports the observed alloy and processing effects on microstructure and associated mechanical performance.
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