Industrialization of Thermomechanically Rolled C-Mn Steel Plate

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ABSTRACT
Thermomechanical rolling schedules in the low austenite temperature region have been developed with the aid of mathematical modelling and industrial trials to produce as-rolled C-Mn steel plate on an industrial scale to the S355J2 specification with yield strengths approaching 400MPa, high impact toughness energy at sub-zero temperatures and good flatness - without the need for expensive micro-alloying or accelerated cooling. An integrated model that included workpiece heat transfer, austenite evolution, roll force, mechanical property and shape predictions was compiled to optimize the relevant processing parameters to meet the above objectives. The rolling process relies on careful selection of delay times after roughing to condition the austenite during finishing and increase the ferrite nucleation site density prior to transformation during cooling. Following complete recrystallisation after roughing, the finishing schedules applied at low austenitic temperatures were adequate to achieve sufficient grain refinement and the desired mechanical properties by restricting recrystallisation and accumulating sufficient strain. The total rolling strain is an important factor influencing grain size homogeneity, especially in thicker plates, whilst the pass strain sequence and magnitude affects final flatness.

INTRODUCTION
Interest has been shown in increasing the as-rolled strength in economical C-Mn plate steels via some thermomechanical (TM) rolling process to compete at the lower end of the medium-high yield strength market, i.e., above 355MPa, which is generally above that found in conventionally rolled (CR) plates. One TM process is so-called recrystallisation controlled rolling (RCR) which relies on complete austenite recrystallisation at high deformation temperatures and has been successfully applied to Ti-V steels. Grain growth is restricted due to grain boundary pinning by sufficiently small, TiN precipitates with low solubility which form during continuous casting and remain stable throughout the process. RCR, however, is unsuitable for C-Mn steels because of the uncontrollably high austenite grain growth rate due to the absence of stable precipitates that retard grain boundary mobility.

Most medium-high strength steels are produced using costly micro-alloying elements, i.e., Nb, Ti and V by applying TM rolling in the low temperature, no-recrystallisation region. Strong, tough material is obtained from enhanced austenite grain refinement due to extensive strain accumulation and substructure development, the extent of which is defined by the type and volume fraction of carbo-nitride precipitates present in the alloy system. It is postulated that, if low temperature TM rolling is performed on C-Mn steels, significant improvements to the as-rolled microstructure and mechanical properties could be realized. A potential drawback in TM rolling however, is the lengthy holding time required prior to finishing that reduces throughput on conventional mills, unless the plant is specifically designed to accommodate these extended delays. Further, poor shape may result from high forces due to deformation in the low rolling temperature region.

To successfully apply TM rolling, many processing variables need to be controlled. Integrated mathematical models capable of predicting the thermal, microstructural and mechanical property behaviour in steels during hot rolling have been shown to be useful, particularly for micro-alloyed grades and to a lesser extent in C-Mn steels. This work discusses how such a model was used to trial and implement TM rolling of C-Mn plate on a conventional reversing mill in the intermediate thickness range to enhance both strength and toughness. The influence of rolling schedule on final plate shape is also explored.
**Integrated Model**

The main phenomena governing the evolution of austenite during hot rolling of C-Mn steels are static recrystallisation and normal grain growth once recrystallisation has been completed. Both are strongly dependent on the rolling schedule which includes the following variables: reduction (strain), rolling speed (from which inter-pass time and strain rate are determined) and plate temperature. In addition to the final austenite grain size and the extent of strain retention, the cooling rate after rolling plays a major role in determining the final ferrite grain size. The above variables were incorporated into an integrated model describing austenite evolution in C-Mn steels and also included workpiece temperature, roll force, mechanical property and simplified final shape predictions. The model is presented elsewhere but, for convenience, important equations employed are presented below. The model predicts average values and does not consider inhomogeneity along the width or length of the workpiece.

The applied von Mises strain, $\varepsilon$, and mean strain rate, $\dot{\varepsilon}$, in pass $i$ are given by:

$$\varepsilon_i = \frac{2}{\sqrt{3}} \ln \left(\frac{h_{i-1}}{h_i}\right)$$

$$\dot{\varepsilon}_i = 1000 \frac{r_i}{(R_i(h_{i-1} - h_i))^{0.5}} \varepsilon_{\text{eff},i}$$

where $R$ is the flattened work roll radius, $v$ is strip velocity, $h$ is the pass exit thickness, $\varepsilon_{\text{eff}}$ is the effective strain, $\varepsilon_C$ is the critical strain for the onset of dynamic recrystallisation (DRX) and is independent of strain:

$$\varepsilon_C = 1.098 \times 10^{-2} d_i^{0.174} \varepsilon_i^{1.0165} \exp(24327 / (8.314 T_i))$$

The fraction of static recrystallisation (SRX), $X$, is found from the time for 50% recrystallisation, $t_{50}$. Both parameters are functions of austenite grain size, $d$, and plate temperature, $T$:

$$t_{50,i} = 8.31 \times 10^{-15} d_i^{1.5} \varepsilon_i^{-0.33} \varepsilon_{\text{eff},i}^{-1.5} \exp(263000/(8.314 T_i))$$

$$X_i = 1 - \exp(-0.693(t_i / t_{50,i}))$$

The effective strain is the sum of the applied strain and the amount retained in the previous pass:

$$\varepsilon_{\text{eff}} = \varepsilon_i + (1 - X_{i-1}) \varepsilon_{i-1}$$

The austenite grain size following recrystallisation after each pass is given by:

$$d_{\text{ex},i} = 88.96 d_i^{0.369} \varepsilon_{\text{eff},i}^{-0.368} \exp(-28060/(8.314 T_i))$$

The average combined austenite grain size, $d_{\text{ave}}$, of both recrystallized and un-recrystallised microstructure components - assumed to act independently - can be expressed in simplified form as:

$$d_{\text{ave},i} = X_i^{4/3} d_{\text{ex},i} + (1 - X_i)^2 d_{\text{ave},i-1}$$

When recrystallisation is complete, i.e., taken as $X > 0.98$, grain growth takes place between passes according to:

$$d_{\text{gg},i} = (1.45 \times 10^7 \exp(-100000/(8.314 T_i)) t_i + d_{\text{ex},i}^{0.7})^{1/7}$$

The ferrite grain size, $d_\alpha$, is a function of C and Mn contents, final average austenite grain size, cooling rate after mill exit, $\Psi_N$, and the effective strain after the last pass, $N$:

$$d_\alpha = (6.77 - 10C - Mn) \Psi_N^{-0.175} d_{\text{ave},N}^{0.4 - 0.25 \varepsilon_{\text{eff},N}}$$

Roll force, $F_{F&A}$, which is proportional to plate width, $W$, is estimated from the mean flow stress, MFS, in C-Mn steels:

$$F_{F&A} = 0.25 W \sqrt{R_i(h_{i-1} - h_i)} \left( \pi + \frac{2 \sqrt{2}}{h_i(h_{i-1} - h_i)} \right) \frac{2}{\sqrt{3}} \text{MFS}$$
Yield (YS) and ultimate tensile strength (UTS) were estimated from:

\[
YS = 62.6 + 26.1\text{Mn} + 60.2\text{Si} + 3286\text{N}_{\text{free}} + 759\text{P} + 19.7(d/1000)^{-0.5}
\]

\[
UTS = 164.9 + 634.7\text{C} + 53.6\text{Mn} + 99.7\text{Si} + 3339\text{N}_{\text{free}} + 652\text{P} + 11(d/1000)^{0.5}
\]

Rolling Schedules

In practical terms, primary mill operational objectives include a) low forces: rolling within load capacity is promoted by high rolling temperatures and small strains, b) high throughput: enhanced by high temperatures and fast rolling speeds, and c) plate dimensions: all must be within specified tolerances. Good final shape (flatness) is controlled by the rolling schedule and the effective roll crown. However, to obtain a sufficiently fine final microstructure for enhancing mechanical properties, effective austenite conditioning is necessary via TM rolling that requires both low temperatures and large reductions, contrary to some of the primary mill requirements. The integrated model was employed to find a compromise between the above objectives.

Various simulated TM schedules were tested at increasingly lower plate temperatures by lengthening the delay time after roughing. The aim was to obtain a sufficiently fine \(d_\alpha\) with yield strengths comfortably above 355MPa, but also restrict roll forces to well below the mill capacity of 4000t. From previous work\(^9\), the initial austenite grain size after reheating and soaking 0.2%C-1.5%Mn-0.3%Si full size (240mm) slabs was assumed to be 350 \(\mu\)m. From mill measurements, inter-pass times were taken as 7s during roughing and 4s in finishing due to faster rolling speeds. The cross-over bar thickness, COB, prior to finishing for each plate thickness was selected as a compromise between i) applying adequate through-thickness strain and ii) due to cost implications, limiting the extended delay time after roughing to perform TM rolling during finishing in the low austenitic temperature region. After roughing (including broad sizing) in approximately 11 passes followed by an extended air cooling delay period, finishing was completed within 8 to 13 passes. Applied pass strains were restricted to below 0.25 to limit rolling forces. Table I shows the determined delay times, COB thickness and finishing start temperatures required to exit the mill between 770 and 830°C for plates gauges, \(h_N\), between 12 and 25mm and widths between 2.4 and 3m. This temperature range was above the calculated \(Ar_3\) temperature\(^12\) to avoid rolling in the two-phase region, but sufficiently low to refine the microstructure and improve mechanical properties as a result of austenite substructure development and subsequent ferrite grain refinement. The plate dimensions at any stage of rolling were calculated from volume changes after deformation and width spreading.

After optimizing the rolling schedules for each plate thickness through simulation, small-scale industrial rolling trails were initially performed followed by larger-scale production. Standard tensile (transverse direction) and impact toughness tests were performed at 0°C and -20°C across the plate front-end. Microstructures were revealed using a 2% nital etchant solution from which the ferrite grain size was determined using the mean linear intercept method.

**Table I. TM rolling parameters.**

<table>
<thead>
<tr>
<th>(h_N) mm</th>
<th>COB mm</th>
<th>COB/(h_N)</th>
<th>Delay time, min.</th>
<th>TM start temperature,°C</th>
</tr>
</thead>
<tbody>
<tr>
<td>12</td>
<td>40</td>
<td>3.3</td>
<td>2.5</td>
<td>890</td>
</tr>
<tr>
<td>16</td>
<td>49</td>
<td>3.1</td>
<td>3.5</td>
<td>865</td>
</tr>
<tr>
<td>20</td>
<td>58</td>
<td>2.9</td>
<td>5.0</td>
<td>835</td>
</tr>
<tr>
<td>25</td>
<td>68</td>
<td>2.7</td>
<td>8.5</td>
<td>810</td>
</tr>
</tbody>
</table>

RESULTS AND DISCUSSION
Figs. 1-4 are simulated and actual mill outputs for selected plate rolling schedules in the 12-25mm thickness range and demonstrate how some potential quality problems were avoided. Most of the parameters calculated from Eqsns. 1-14 are included in the figures, as well as the predicted plate surface temperatures as a function of pass number. The critical strain to initiate dynamic recrystallisation, $\varepsilon_c$, was generally well above $\varepsilon_{app}$, but close to, or above $\varepsilon_{eff}$ in some cases during low temperature finishing. Muojekwu showed that DRX only occurs when i) $\varepsilon_{eff} > \varepsilon_c$ and ii) the applied strain rate is lower than the estimated boundary strain rate, i.e., the critical strain rate at the transition between peak stress (DRX) and no-peak stress (SRX) response during deformation. In this work the applied strain rate during finishing was 5-7s^{-1}, which is significantly higher than the boundary strain rate - below 1s^{-1} in the 780-820°C temperature region - in a plain C steel similar to that used here. Thus, all calculations were simplified to only consider static recrystallisation when describing the evolution of austenite during rolling.

Micrographs taken at the quarter and mid-thickness of rolled plates are also shown, together with the corresponding yield and ultimate tensile strengths. The ferrite grain size distributions were generally non-uniform after all TM schedules due to the occurrence of partial recrystallisation at low rolling temperatures, where the resulting deformation bands have different ferrite nucleation potentials which results in mixed grain sizes.

**Conventional vs Thermomechanical Schedules**

Fig.1a compares the thermomechanical behaviour in 12mm C-Mn plates subjected to both CR (schedule 1A) and TM rolling following a 2.5min delay after roughing (schedule 1B). In the early roughing passes, recrystallisation in both schedules was limited by the coarse initial austenite grain size formed after reheating and soaking. High temperatures and strain accumulation steadily increased $X$ to 1 by the fourth pass where $d_\gamma$ had decreased to below 100µm. The remainder of roughing consisted of alternate passes of partial or full recrystallisation followed by grain growth. No fine, stable particles were present to effectively restrict austenite grain boundary mobility at these high temperatures and recrystallisation was rapidly completed - a desirable condition for avoiding a fraction of large, unrecrystallised austenite grains that may persist downstream and potentially have a negative effect on toughness.

In schedule 1A, the extent of recrystallisation decreased steadily during finishing, but $X$ remained relatively large due to high rolling temperatures in the absence of a significant holding time after roughing. In schedule 1B, $X$ decreased significantly at the start of finishing due to the lower temperature and long $t_{so}$ as well as the smaller applied strain. In subsequent passes, $t_{so}$ was longer and $\varepsilon_{eff}$ higher than in schedule 1A, whilst $X$ and $d_\gamma$ were substantially smaller which increased the effective interfacial area for ferrite nucleation sites and, upon transformation, produced a finer $d_\alpha$. Despite the increase in MFS during finishing in the TM schedule, all roll forces remained well within the mill capacity of 4000t.

In both CR and TM schedules, $d_\gamma$ decreased asymptotically to a size which was increasingly difficult to refine despite further decreases in thickness and/or temperature. The diminishing returns in the final austenite grain size was also observed in other schedules implying that nucleation site saturation had occurred for those particular set of conditions.

The final microstructures in Fig.1b contain a combination of polygonal ferrite and banded pearlite colonies, the size of which depended on the rolling schedule. The CR and TM schedules had predicted final $d_\alpha$ sizes of 25µm and 10µm respectively, which corresponded to $d_\alpha$ values of 11µm and 7µm (Eqn.10), similar in size to measurements. Grain refinement increased the yield strength from 325MPa after CR to 407MPa in the TM schedule. The corresponding predicted values from Eqsns.11 and 12 gave conservative estimates of YS and UTS. The flatness after visual inspection was regarded as “good” in both schedules.

**Plate Flatness**

Initial attempts at additional grain refinement by increasing the strain magnitude in the final stages of rolling were unsuccessful, despite predictions of a finer $d_\alpha$ and higher YS in schedule 2A, fig.2a. The calculated roll force in these passes were, however, significantly higher than the corresponding reductions in schedule 1B, but were still below the mill limit. A plate subjected to schedule 2A enhanced the yield strength, produced excessive forces and poor final flatness when compared to schedule 1B. This result prompted the inclusion of a shape-correction routine into the model to improve flatness for any reduction sequence. Centre buckles or edge waves in a rolled plate are caused by a differential elongation of the plate across its width and is directly related to the plate crown change, $\delta$, during the pass reduction, Eqn.15.

$$\delta = \frac{C_1}{h_1} - \frac{C_2}{h_2}$$

where $C_1$, $C_2$ are incoming and outgoing crowns and $h_1$ and $h_2$ are the entry and exit plate gauges. Due to internal stresses, the deterioration of plate flatness does not occur as long as the values for the change in relative plate crown are within a certain range that is known as the “flatness dead band” as described by Shohet and Townsend. The plate flatness calculation for the above two schedules is shown in fig.2b. Good shape is expected after the reduction sequence in schedule 1B compared to severe wavy edges predicted in the plate rolled to schedule 2A, where $\delta$ lies outside the bottom dead band limit in the final three
passes. Thus, reduction sequences that ensure $\delta$ lies within the dead band for a given plate crown were subsequently employed in all TM rolling schedules to obtain a compromise between a sufficiently fine $d_a$ and good flatness.

Figure 1a. C-Mn rolling schedules: model outputs for 12mm thick plate.
Schedule 1A: CR, 21 Passes $d_a = 11\mu m$, YS=308MPa, UTS =472MPa.
Schedule 1B: TM, 23 Passes $d_a = 7\mu m$, YS=370MPa, UTS =514MPa.

Figure 1b. Final microstructures, grain size (mid thickness) and measured strength after schedules in Fig.1a. Width = 3000mm.
Figure 2a. TM schedules for 12mm C-Mn plate. Influence of strain distribution on final flatness.
Schedule 2B: Predicted $d_\alpha = 6 \mu m$, YS = 447MPa, UTS = 587MPa. Properties not measured.

Figure 2b. Shohet and Townsen flatness calculation for 12mm x 3000m plates. Crown = 0.5mm (concave)
Microstructural Homogeneity

Fig.3 shows a) model predictions and b) final microstructures and properties after TM rolling of 16mm and 20mm plates. Compared to the 12mm schedule 1B in Fig.1, trends in both austenite evolution and flow/force behaviour were similar, but the ferrite grains were more mixed in these thicker gauges. This was attributed to smaller COB/hN ratios (Table I), where less total strain was available for achieving a uniform through-thickness grain size distribution. Good agreement between predicted and measured values of both $d_\alpha$ and strength was obtained.

Figs.4a,b illustrated how the number of rolling passes in TM rolling influenced the austenite evolution and final microstructure and properties of thicker (25mm) plate in which the total available reduction is limited. Plates rolled according to schedule 4A (21 passes) contained local regions, particularly at the quarter thickness, where some grains were much coarser compared to the average matrix size. The significant difference in schedule 4B was the application of fewer passes - and slightly larger applied strains - that provided more extensive $\varepsilon_{\text{eff}}$ in finishing and was probably responsible for a more homogeneous final grain size distribution. Fig.4a shows that, unlike the schedules for thinner gauges in Figs.1-3, austenite recrystallisation was never completed during roughing, which may also have contributed to the mixed final grain size in both 25mm schedules. The model does not predict localized variations in $d_\alpha$ and further work is needed to address this shortcoming.

Full-Scale Production

Fig 5 shows the mechanical properties after full-scale TM rolling of 12 to 25mm thick plates. The yield strength was significantly higher after CR of a specific thickness. The YS decreased with thickness but always ranged between 380 and 420MPa - in agreement with predicted values from Eqn.13. Similar results were found for the UTS using Eqn.14, although the variance was smaller compared to that found in the YS. Total elongation was comfortably above the minimum specification. Impact toughness values were generally excellent, despite the non-uniformity in grain size found in thicker plates. In 16-25mm plates, the impact energy at -20°C was consistently above 150J.

In summary, enhanced grain refinement together with good flatness can be achieved by i) ensuring complete recrystallisation after roughing using sufficient reduction, ii) restricting recrystallisation as much as possible and promoting strain accumulation, albeit it limited, in low temperature finishing. iii) avoiding excessive or sudden strain variations in the final passes and iv) applying adequate through-thickness strain to improve microstructural homogeneity. Although the grain sizes and distributions achieved here were sufficient to significantly improve strength and toughness, a higher COB/hN ratio above 3 may be necessary to achieve further through-thickness grain refinement in thicker plates for adequate toughness since CVN energy is inversely proportional to $d_\alpha$. TM rolling is feasible in C-Mn grades provided extended delay times do not restrict throughput or increase scale losses to unacceptable levels. Finally, it is envisaged that similar schedules can be applied to achieve higher strengths in C-Mn steels by simply alloying with controlled quantities of vanadium and nitrogen, an example of which is shown in Fig.6.[15]

V(C,N) precipitation, the main contributor to additional strengthening in these steels, occurs predominantly after rolling and has limited influence on both austenite evolution, MFS and roll force - even at medium to low rolling temperatures, i.e., the material behaves similarly to a C-Mn grade during both CR and TM rolling.

CONCLUSION

An off-line temperature-microstructure-property model was compiled and applied to successfully implement full-scale production of thermomechanically rolled C-Mn plates in the 12-25mm thickness range to achieve yield strengths approaching 400MPa, good sub-zero impact toughness and sufficiently flat final products. Optimized rolling schedules included a combination of fully recrystallised austenite after all roughing passes, an extended air-cooling delay after broad sizing followed by limited static recrystallisation and subsequent strain accumulation during low temperature finishing for sufficient grain refinement. After transformation during air cooling, the resultant ferrite grain size was sufficiently fine to produce mechanical properties that conform to the S355J2 specification. Strength was increased significantly whilst total elongation was relatively insensitive to thickness. Impact values were consistently above 150J at -20°C. The number of rolling passes should be restricted, especially in thicker plate, to provide sufficiently large pass strains in both roughing and finishing for avoiding excessively mixed final grain sizes. Erratic strain distribution during the last rolling passes can lead to poor plate flatness.

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Schedule 3A, predicted: $d_e=6\mu m$, $YS=389$ Mpa, $UTS=532$ MPa. Schedule 3B, predicted: $d_e=6\mu m$, $YS=379$MPa, $UTS=531$MPa

Figure 3a. Model outputs from 16mm (Schedule 3A, 20 passes) and 20mm (Schedule 3B, 22 passes) TM schedules.

Mid-thickness 1/4 Thickness

a) Schedule 3A - 16mm x 2.4m

Measured: $d_e=7\mu m$, $YS=392$MPa, $UTS=541$MPa

Mid-thickness 1/4 Thickness

b) Schedule 3B - 20mm x 3m

Measured: $d_e=7\mu m$, $YS=392$MPa, $UTS=551$MPa

Figure 3b. Microstructures and strength in 16 and 20mm TM rolled plates.
Figure 4a. Model outputs for 25mm C-Mn TM rolling schedules.

Figure 4b. 25mm C-Mn plate: Microstructures after industrial TM schedules in Figure 4a. Numbers in parentheses are predicted values.
Figure 5. Mechanical properties of TM rolled C-Mn plate as a function of thickness – average values.

Exit = 825°C.  \( d_\text{e} = 6.5\mu\text{m} \)  
YS = 452MPa, UTS = 571MPa

Figure 6. Microstructure and tensile strength after TM rolling of 0.055%V-0.1%N steel to 20mm\(^1\).  
Similar schedule to 3B in Fig.3a.
REFERENCES