1. Introduction

The high strength steels with high formability is the main objective of thermomechanical controlled processing. From the mid-1980s, a new class of TRIP (transformation induced plasticity) steel has been proposed based on Fe–0.2C–1.5Mn–1.5Si (wt%).\(^1\)\(^-\)\(^3\) The microstructure of these new TRIP steels contains a significant amount of retained austenite with a multiphase nature consisting of polygonal ferrite and bainite. An excellent combination of high strength with high formability is achieved by the transformation of metastable retained austenite to martensite during forming of the component at room temperature.

The multiphase microstructures of these new TRIP steels are usually obtained by a standard two stage heat-treatment that consists of an intercritical annealing stage followed by an isothermal stage in the bainite transformation temperature range.\(^1\)\(^-\)\(^3\) Most investigations on these TRIP steels have been concentrated on the effects of heat treatment conditions and alloying elements such as Si and Mn on the microstructure and mechanical properties in cold rolled sheet steels. There have been few studies on hot rolled TRIP steel especially on controlled rolling and cooling. The multiphase microstructures of TRIP steels can be produced in hot-rolled steels by cooling to the bainite transformation temperature after rolling, at which temperature the sheet is coiled. The coiling temperature is 400°C, it is an optimum cooling temperature.\(^4\)

To produce the TRIP microstructures in hot rolled steels, it is necessary to apply an appropriate cooling process on a runout table (ROT) in the hot strip mill. In the present study, the steel was controlled cooling on a ROT stage after hot rolling. The first water cooling stage made high density dislocation remain in the austenite, the second water cooling stage avoided pearlite formation, and it is necessary to hold about 10 s during intermediate air cooling on the ROT, which contributes to enhancement the stability of retained austenite. Excellent mechanical properties were obtained through TMCP due to the transformation of retained austenite into martensite during straining (transformation induced plasticity) for the present Fe–C–Mn–Si multiphase steels, and the specimen 3 shows the highest value of total elongation (37%) and the balance of strength and ductility (30 488 MPa%).

KEY WORDS: thermomechanical controlled processing (TMCP); a three step cooling; TRIP steel; retained austenite; strength; ductility.

2. Experimental Procedure

2.1. Chemical Compositions of Steel and Processing Conditions

The experimental 80-mm thickness steel plate was vacuum melted and forged. The chemical composition of steel is shown in Table 1. In Table 1, the \(T_{\text{cr}}\), which denotes the nonrecrystallization temperature, was measured using a thermomechanical simulator by corresponding double-pass compression testing. Continuous cooling transformation (CCT) diagram was also constructed using this simulator by corresponding double-pass compression testing (Fig. 1).
On the basis of CCT diagram, controlled cooling trials of deformed steel was conducted on a ROT in a laboratory rolling mill.

First the slab was hot-rolled down to 20 mm by a φ450 mm hot rolling mill, and then machined to form four pieces of plates of 80 mm × 20 mm × 300 mm, and finally hot-rolled down to 3.8 mm thickness by a φ300 mm hot rolling mill. A reheating temperature of 1120°C and a finish rolling temperature of 840°C were applied during hot rolling process, and the processing schedule is shown in Table 2.

After finish rolling, controlled cooling was conducted by the three step cooling, i.e. early and late water cooling with intermediate air cooling on a ROT. TMCP schedule in a laboratory was consistent with practice mill, as shown in Fig. 2.

In Fig. 2, first, the hot rolled steel was water cooled to an appropriate intermediate slow cooling region, and then it was air cooled in this region. Start and finish temperature of air cooling stage were controlled about 550–600°C and 500–550°C, respectively. Second, the steel was water cooled again to the coiling temperature. Finally, the steel was placed into an especial isothermal device (approximate 400°C) to simulate the coiling. A cooling path (zigzag with an arrow) on a ROT for a desirable microstructure of TRIP steel is shown in Fig. 1. The surface temperature was measured by a hand pyrometer throughout the experiment. However, it is difficult to control the steel temperature precisely during three step cooling. Water cooling time in first stage for the steel 1 was too long resulted in an over low temperature, and water cooling time in second stage for the steel 2 was too short resulted in an over slow cooling rate. Controlled cooling parameters were nearly the same for the steel 3 and the steel 4. After hot rolling, stay time of all the steels before first water cooled was set as short as possible. Processing parameters of controlled cooling on a ROT in the laboratory rolling mill is shown in Table 3.

### Table 1. Chemical compositions of the experimental steel.

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>S</th>
<th>P</th>
<th>Al</th>
<th>T_{re}</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>0.233</td>
<td>1.365</td>
<td>1.540</td>
<td>0.004</td>
<td>0.007</td>
<td>0.080</td>
<td>910°C</td>
</tr>
</tbody>
</table>

### Table 2. Deformation schedules of the experimental steel.

<table>
<thead>
<tr>
<th>Pass no.</th>
<th>F0</th>
<th>F1</th>
<th>F2</th>
<th>F3</th>
<th>F4</th>
</tr>
</thead>
<tbody>
<tr>
<td>Deformation temperature /°C</td>
<td>1120</td>
<td>1060</td>
<td>1000</td>
<td>920</td>
<td>840</td>
</tr>
<tr>
<td>Reduction %</td>
<td>—</td>
<td>45</td>
<td>30</td>
<td>30</td>
<td>30</td>
</tr>
<tr>
<td>Thickness /mm</td>
<td>20</td>
<td>11.0</td>
<td>7.7</td>
<td>5.4</td>
<td>3.8</td>
</tr>
</tbody>
</table>

### Table 3. Controlled cooling parameters of the experimental steels.

<table>
<thead>
<tr>
<th>Specimen</th>
<th>τ₁ /s</th>
<th>τ₂ /s</th>
<th>τ₃ /s</th>
<th>τ₄ /s</th>
<th>T /°C</th>
</tr>
</thead>
<tbody>
<tr>
<td>1°</td>
<td>3</td>
<td>4</td>
<td>1</td>
<td>57600</td>
<td></td>
</tr>
<tr>
<td>2°</td>
<td>1</td>
<td>12</td>
<td>0.5</td>
<td>57600</td>
<td></td>
</tr>
<tr>
<td>3°</td>
<td>1</td>
<td>9</td>
<td>1</td>
<td>57600</td>
<td></td>
</tr>
<tr>
<td>4°</td>
<td>1</td>
<td>6</td>
<td>1</td>
<td>57600</td>
<td></td>
</tr>
</tbody>
</table>

2.2. Testing Procedure

Four kinds of specimens for the tensile test were cut in a longitudinal direction from four pieces of steels, respectively, and then machined to the standard specimens. The microstructure of transverse sections of the specimens before and after tensile testing was observed by light optical microscopy (LOM), scanning electron microscopy (SEM) and transmission electron microscopy (TEM), respectively. The specimens for LOM were colour etched using the LePera method⁶ to reveal the ferrite, bainite, austenite and martensite structures, and the colour etchant was a mixture of 4% (NO₂)₃C₆H₂OH saturated Na₂S₂O₃. With this etchant ferrite appears grey, bainite appears black. Martensite, retained austenite and martensite-austenite (MA) islands appear white since it is difficult for them to be tinted in this etchant. X-ray diffraction analysis was carried out to determine the volume fraction of retained austenite (Vγ) with the modified Miller’s method.⁶

\[ V_\gamma = \frac{1.4I_\gamma}{I_\alpha + 1.4I_\gamma} \]  

Where \( I_\alpha \) and \( I_\gamma \) are the integrated intensities of the (110)_α, (200)_α, (211)_α, (220)_α, (310)_α and (222)_α peaks and the
3. Experimental Results

3.1. Mechanical Properties

Ultimate tensile strength (UTS), yield strength (YS), total elongation (TEL) and the product of ultimate tensile strength and total elongation (UTS×TEL) of the specimens at various controlled cooling parameters are shown in Fig. 3. The specimens 1 and 2 show maximum and minimum UTS and YS, respectively, and the specimens 3 and 4 show nearly same UTS and YS. The specimen 3 shows higher UTS and YS, as well as highest TEL (824 MPa, 536 MPa and 37%, respectively). The mechanical properties of all the specimens are as high as that obtained in the similar chemical composition of cold rolled multiphase steel, even exceed the latter.\(^{11}\) The specimen 2 shows the lowest value of UTS and YS in the four specimens, but they still reach 788 MP and 510 MPa, respectively. TEL of the specimen 1 is minimum value, but it still reaches 22%. In Fig. 3(c), the balance of strength and ductility (UTS×TEL) exhibits a similar trend as TEL shown in Fig. 3(b). The specimen 3 shows the highest value of 30,488 MPa%. 

Figure 4 shows the stress–strain curves (up to the maximum stress) from tensile test of the specimens after TMCP. The stress–strain curves are characterized by low yield ratio. Continuous yielding appears on the curves for almost all the specimens. All the curves present characteristics of intensive serrations, this is because the interrupted plastic straining occurred in the localization of the specimens during straining. The curves are prolonged afterwards reach the peak values, especially in the specimen 3. It can be seen from Fig. 4 that the strain hardenability is maintained constant in a large strain range prior to necking for the stress–strain curve of the specimen 3.

3.2. Microstructures

Figure 5 shows LOM micrographs of the specimens after various controlled cooling. Specimen 1 contained granular bainite (black), polygonal ferrite (grey) and a larger amount of white martensite (Fig. 5(a)). Specimen 2 contained granular bainite, polygonal ferrite (grey) and white martensite–austenite (MA) islands, moreover, fine pearlite (black) is appeared in the microstructure (Fig. 5(b)). The microstructure of specimen 3 is similar to that of specimen 4, it is composed of polygonal ferrite, granular bainite and retained austenite (Figs. 5(c), 5(d)), and ferrite (grey) and retained austenite (white) and/or MA islands (white) are obvious in Fig. 5(c) and they are not in Fig. 5(d).

All constituents (polygonal ferrite, granular bainite and retained austenite) can clearly be seen by SEM. In SEM micrographs (Fig. 6) with Nital etching of specimen 3 after TMCP, the structure of the multiphase microstructures consist of polygonal ferrite (black), retained austenite (grey), and granular bainite (black). A larger amount of austenite (grey) can be retained for the specimen 3 before tensile testing, and an amount of them decreased obviously after tensile testing. On the one hand, the polygonal ferrite and austenite grains were elongated in the longitudinal direction after tensile testing, thus, these grains became fine in the transverse section. On the other hand, martensite and/or MA islands (white) increased after tensile strain, and MA islands are usually quite fine. As a result, the microstructure became finer by tensile testing.

Figure 7 shows retained austenite distribution for the same specimens as above. Comparing with the ferrite around, the contrast of retained austenite is dark because carbon concentration in it is relatively high in TEM micrographs. The presence of the retained austenite is confirmed by its diffraction pattern. Martensite is discovered accompany martensite–austenite islands increasing obviously on the TEM micrograph for the same specimens after tensile testing (Fig. 8). Martensite is dark similar to retained austenite because carbon concentration of it is higher than that of the ferrite. Similarly, the phases can be distinguished between martensite and retained austenite by their diffraction pattern.

These micrographs confirm that the strain-induced transformation to martensite from the retained austenite occurs during tensile deforming.

In Fig. 9, TEM micrograph is presented for the same
specimen as above, and this micrograph confirms the conclusion from the optical micrograph that bainite is granular, which is to say, ferrite takes the shape of platelets and the bar-like islands are generally parallel distributed.

The TEM analysis reveals that TMCP results in the formation of intragranular defects for the specimens. A high density of dislocation tangles is found within the austenite.
grain, and retained austenite is labeled (Fig. 10). The foil for TEM is thin because viewing field is close to the edge of selected specimen, besides, the contrast was adjusted to attain the brightest by rotating the foil. At the same time, the amplification times of Fig. 10 is enhanced. Therefore, austenite in Fig. 10 and Fig. 7 look very different.

The distribution of the retained austenite is further confirmed by the X-ray diffraction studies. The volume fractions of retained austenite before and after tensile testing for the specimens are summarised in Table 4.

Figure 11 shows the X-ray diffraction profile of ferrite and austenite in the specimen 3. After tensile testing, the amount of retained austenite at necking of the specimen 3 decreased from 19.6 to 8.8%.

The fracture morphologies of the specimens after tensile test are accorded with the behavior of tensile properties shown in Figs. 3 and 4. In Fig. 12, a lot of dimples are found when the specimen 3 was broken after tensile test, and the dimple is very deep. Figure 12 confirms that, a three step cooling pattern on the ROT after hot rolling for the specimen 3 results in higher ductility.

### 4. Discussion

#### 4.1. Microstructural Evolution during TMCP

TMCP affected all constituents of the multiphased TRIP steel microstructure. Austenite pancaking is achieved by deforming in the final pass at temperatures below the non-recrystallization temperature. The relevant microstructural features of pancaked austenite are an increase in dislocation density, formation of deformation bands, the destruction of the coherency of annealing twins with the matrix and the formation of ledge-like austenite grain boundaries. The increase in the deformation band will divide the austenite grain into several grains with smaller size. Ferrite nucleation occurs not only on austenite grain boundaries but also within grains in the deformed austenite, because the deformation bands and twin boundaries formed during deformation of austenite act as effective ferrite nucleation sites. The deformation bands produced by the final reduction in non-recrystallization temperature region of austenite refine the ferrite grains. Therefore, finer ferrite grains size were obtained in almost all the specimens (Fig. 5), and that excellent mechanical properties of the specimens 3 and 4 are likely related to finer ferrite grains.

All the specimens were fast cooled from the finish mill exit temperature (Fig. 2), which avoids relaxation after the deformation, because relaxation significantly reduces the effect of strain on transformation and the deformed microstructure is easily recovered and recrystallized during...
the relaxation time. Dynamic recovery is far from complete and static recovery is made impossible due to the fast cooling for all the specimens, therefore, more dislocations are retained after a three step cooling at room temperature (Fig. 10).

Substructure affects austenite stabilization in the present TRIP steels. Tsuzaki and Raghavan\(^{11}\) found that a large dislocation density hinders the growth of the martensitic plates. The stability of retained austenite against the martensitic transformation is improved by increasing the dislocation density.

Cooling processing on the ROT after hot rolling is of great importance for the present steels since the final microstructure can be controlled during this stage. As mentioned above, in order to prevent relaxation after the deformation, the first water cooling stage for the specimen 1 was overage, which resulted in the appearance of martensite as white area. Its morphology is quite different from the others. Such difference in the microstructure would result in the higher ultimate tensile strength (UTS), yield strength (YS) and the lower total elongation (TEL) for the specimen 1 and for the other specimens since the formation of martensite consumes an amount of metastable retained austenite. The second water cooling stage for the specimen 2 was shortage resulted in a slower cooling rate, thereby, a small amount of pearlite formed during controlled cooling in the specimen 2. The TRIP effect of the specimen 2 would be diminished since pearlite consumes a large amount of carbon and weakens carbon enrichment in untransformed austenite, thus reduces the stability of the remaining austenite. Therefore, the specimen 2 shows lowest UTS and YS and low TEL. The desirable microstructure containing polygonal ferrite, granular bainite and retained austenite for specimens 3 and 4 was obtained, this is likely to be due to the cooling path on the ROT of specimens 3 and 4 accords with zigzag in Fig. 1. The microstructure affected the corresponding tensile properties, thus higher value of TEL and UTS×TEL for specimens 3 and 4 was obtained. TEL and UTS×TEL of the specimen 3 presented the highest values. One possible reason for superior tensile properties in specimen 3 may be due to the fact that there is higher retained austenite content of the specimen 3 (Table 4 and Fig. 11), and the stability of retained austenite is the highest, and therefore, the strain-induced transformation of austenite developed gradually during plastic straining.

### 4.2. The Formation of the Polygonal Ferrite

An amount of polygonal ferrite can form during intermediate air cooling stage, although fast cooling after deformation can suppress ferrite transformation. The ferrite formation can be explained by solid-state transformation theory.

According to Verhoeven\(^{21}\) ferrite growth kinetics, the transformation of the ferrite formation is controlled by diffusion. When the experimental steel of composition \(C_0\) is held at temperature \(T_1\), a portion of the Fe–Fe\(_3\)C diagram and a chunky ferrite (\(\alpha\)) precipitate particle nucleates and grows into the austenite (\(\gamma\)) phase is shown in Fig. 13. In Fig. 13, the carbon concentrations in \(\alpha\) and \(\gamma\) at the \(\alpha/\gamma\) interface is denoted by \(C_{\alpha}\) and \(C_{\gamma}\), respectively; \(Z\) is a distance of the \(\alpha\) precipitate grew, and Area \(A_{1}\) and \(A_{2}\) are proportional to the mass of solute rejected from the \(\alpha\) precipitate and the pile of the solute in front of the interface, respectively. \(A_{1}\) must equal \(A_{2}\).

The solute flux with respect to the growing \(\alpha/\gamma\) interface for a local equilibrium at this interface may be given by Eqs. (2) and (3):

\[
|\text{Flux into interface}| = V C_{\gamma} \frac{dC}{dZ} \quad \text{................(2)}
\]

\[
|\text{Flux away from interface}| = V C_{\alpha} - D \frac{dC}{dZ} \quad \text{................(3)}
\]

Where \(V\) is the growth velocity of the \(\alpha\) precipitate; \(D\) is the diffusion coefficient, and \(\frac{dC}{dZ}\) is the interface concentration gradient, which may be determined as a function of the parameter \(L\) from the construction shown in Fig. 13. Two fluxes will remain balanced for all times at the interface plane, hence the growth velocity is:

\[
V = \frac{D \frac{dC}{dZ}}{(C_{\gamma} - C_{\alpha})} \quad \text{................(4)}
\]

If we approximate area \(A_{1}\) as \(A_{1} = L \cdot (C_{\gamma} - C_{\alpha}) / 2\) and recognize that \(V = dZ/dt\), and \(A_{1} = Z \cdot (C_{\gamma} - C_{\alpha})\), for this case Eq. (5) becomes:

\[
\frac{dZ}{dt} = \frac{D \frac{dC}{dZ}}{2Z(C_{\gamma} - C_{\alpha})^2} \quad \text{................(5)}
\]

Integrating this equation we obtain:

\[
Z = A \sqrt{\frac{t}{D}} \quad \text{...................(7)}
\]

Where

\[
A = \frac{C_{\gamma} - C_{\alpha}}{[(C_{\gamma} - C_{\alpha})(C_{\gamma} - C_{\alpha})]^{1/2}} \quad \text{................(8)}
\]

As discussed above, growth of these precipitates requires long-range diffusion and the growth rate is time dependent. According to the growth kinetics of the \(\alpha\) precipitate,\(^{12}\) in order to obtain an amount of polygonal ferrite, it is necessary to be provided with the duration of intermediate air cooling during the three step cooling pattern on the ROT.

The optimum amount of retained austenite is the critical aspect for TRIP steels. The amount of austenite that can undergo the bainite transformation at simulated the cooling is reduced when polygonal ferrite forms during intermediate air cooling. As a result, a corresponding decrease may be expected in the volume of austenite retained following the
bainite transformation. However, growth of ferrite takes place by rejection of carbon into the untransformed austenite, such that the carbon content of the austenite increases with increasing the polygonal ferrite. The experimental finding of Tamura et al.\textsuperscript{13} shows that the greater the amount of ferrite, the greater the level of C enrichment in the remaining austenite. The stability of the remaining austenite against bainite formation would be enhanced for the present TRIP steels. Taking into account of the formation of ferrite, comparison of the specimen 3 and the others shows that the optimum hold time for air cooling stage might be about 10 seconds (Fig. 3 and Table 3).

4.3. Effect of Coiling Condition

The bainitic transformation of deformed austenite is of great importance in the processing of TRIP steels. Austenite transforms to bainite during the simulated coiling at about 400°C (Figs. 5, 6 and 9). It is well known that bainitic ferrite rejects C which is accumulated in the remaining austenite during the isothermal holding at the bainite transformation temperature. The carbon concentration in the remaining austenite increases. This increases the stability of the remaining austenite, thereby leading to a higher volume fraction of the retained austenite at the end of the process.

The above discussion clearly show that the isothermal bainitic transformation stage (the simulated cooling \( t_4 \) stage) for all the specimens is quite beneficial to the enhanced of the stability of the remaining austenite despite prolonged the isothermal holding times at the bainite transformation temperature might lead to carbide precipitation, which reduces the stability of retained austenite.\textsuperscript{14} A larger amount of retained austenite has been observed that is related to the simulated coiling (Figs. 5, 6, 7, 11 and Table 4).

4.4. Factors Affecting the TRIP Effect

As mentioned previously, the ferrite grain size is refined through TMCP. After deformation, the prior austenite grains are fragmented by deformation bands. As a result, the small particles of austenite tend to be stable. It has been reported that smaller austenite size helps the retention of austenite instead of martensite formation.\textsuperscript{15} However, the contribution of the grain refinement to the mechanical properties of TRIP steel is less than the effect of the stability of retained austenite.\textsuperscript{16} The amount and stability of the retained austenite controls the mechanical properties of TRIP steels.\textsuperscript{17}

Chemical composition and thermal processing history strongly influence austenite retention.\textsuperscript{18–20} Silicon refines the austenite microstructure, is a ferrite stabiliser, and accelerates the polygonal ferrite transformation. It also increases the volume fraction of retained austenite by retarding the precipitation of carbides during the bainite transformation.\textsuperscript{21} Manganese makes \( M_t \) temperature lower and causes the retained austenite content increase. As a result, a larger amount of the stable austenite with carbon enrichment was retained (Figs. 5, 6, 7, 11 and Table 4).

Increased dislocation density, grain refinement and carbon enrichment are the main factors that govern the retained austenite stability.\textsuperscript{22} The first water cooling stage resulted in a high dislocation density inside the grains (Fig. 10). An amount of polygonal ferrite formed during air cooling stage (Figs. 5 and 6), which enhanced the carbon concentration in the remaining austenite. The second water cooling stage avoided pearlite formation which consumes the carbon and therefore decreases the stability of austenite. The formation of the bainitic ferrite resulted in the carbon concentration enrichment in austenite further during the simulated coiling. Therefore, excellent mechanical properties were obtained at the optimum TMCP parameter due to the TRIP effect of the stable retained austenite for the specimen 3 (Figs. 3, 4 and 12).

5. Conclusions

Effects of thermomechanical controlled processing on the state of retained austenite and the corresponding mechanical properties in Fe–C–Mn–Si multiphase steels were studied. The main conclusions of the present work are as follows.

1) TMCP lead to ferrite grain refinement, and this might contribute to the improvement of the mechanical properties of the present Fe–C–Mn–Si multiphase steels.

2) The microstructure containing polygonal ferrite, granular bainite and a significant amount of the stable retained austenite can be obtained through three step cooling on the ROT after hot rolling, which contributes to achievement a satisfactory TRIP effect.

3) The first water cooling stage made high density dislocation remain in the austenite, and subsequent air cooling stage resulted in the formation of the polygonal ferrite, which enhanced the carbon concentration in the remaining austenite. The second water cooling stage avoided pearlite formation. The final cooling stage lead to bainite transformation, and the carbon concentration in the remaining austenite increases further. As a result, the stability of the remaining austenite was increased at the end of the process.

4) The ferrite precipitates requires long-range diffusion and its growth rate is time dependent in three step cooling on the ROT. It is necessary to hold about 10 s during intermediate air cooling in order to form the polygonal ferrite, which contributes to enhancement the stability of retained austenite.

5) Excellent mechanical properties were obtained through TMCP for the present Fe–C–Mn–Si multiphase steels, and the specimen 3 shows the highest value of total elongation (37%) and the balance of strength and ductility (30 488 MPa%).

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REFERENCES

14) Z. Li and D. Wu: *ISIJ Int.*, **46** (2006), 121.