Design of an Effective Heat Treatment Involving Intercritical Hardening for High Strength/High Elongation of 0.2C–3Al–(6–8.5)Mn–Fe TRIP Steels: Microstructural Evolution and Deformation Behavior

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Abstract: High strength/high elongation continues to be the primary challenge and focus for medium-Mn steels. It is elucidated herein via critical experimental analysis that the cumulative contribution of transformation-induced plasticity (TRIP) and microstructural constituents governs high strength/high elongation in 0.2C–3Al–(6–8.5)Mn–Fe steels. This was enabled by an effective heat treatment involving a combination of intercritical hardening and tempering to obtain high strength/high ductility. An excellent combination of high ultimate tensile strength of 935–1112 MPa and total elongation of 35–40% was obtained when the steels were subjected to intercritical hardening in the temperature range of 700–750 °C and low tempering at 200 °C. The intercritical hardening impacted the coexistence of austenite, ferrite, and martensite, such that the deformation behavior varied with the Mn content. The excellent obtained properties of the steels are attributed to the cumulative contribution of the enhanced TRIP effect of austenite and the microstructural constituents, ferrite and martensite. The discontinuous TRIP effect during deformation involved stress relaxation, which was responsible for the high ductility. Lamellar austenite, unlike the equiaxed microstructure, is envisaged to induce stress relaxation during martensitic transformation, resulting in the discontinuous TRIP effect.

Keywords: microstructure; deformation; mechanical properties; heat treatment; deformation behavior; discontinuous transformation during deformation

1. Introduction

Transformation-induced plasticity (TRIP) steels have good crashworthiness, superior ductility, and high strength. The steels are potential candidates for automotive applications. The enhanced work hardening rate and energy absorption behavior are attributed to the TRIP effect, which is the transformation of retained austenite to martensite [1,2]. The conventional low-alloyed TRIP steels with Mn content of less than ~2.5% and a product of ultimate tensile strength and total elongation (PSE) of ~15 ± 10 GPa% consist of ferrite, retained austenite, a small amount of martensite, and occasionally bainite [3–6]. The high-manganese twinning-induced plasticity (TWIP) steels with PSE of ~60 ± 10 GPa% are generally completely austenite [7,8].
Recently, a number of studies on medium-manganese (4–10%) steels have been conducted [9–16]. The medium-Mn steels with PSE ≥ 30 GPa% are considered promising automobile steels. The composition of austenite greatly influences the austenite stability. It has been suggested [12,13] that austenite stability can be increased by the distribution of manganese between the austenite and ferrite during annealing in the intercritical region with sufficient holding time for the microstructure to reach phase equilibrium. Miller [14] reported that the austenite fraction in 6.0% Mn steels could be as high as 30%. More recently, retained austenite fractions of ~16.8% in 5.2Mn–0.10C steel, ~27.8% in 5.8Mn–0.10C steel, and ~38% in 7.1Mn–0.10C steel were obtained by Merwin [11]. The volume fraction of retained austenite was significantly increased with increasing Mn content, which increases with PSE.

In conventional TRIP steel, a two-stage heat treatment process is adopted to stabilize austenite during austempering [17]. As regards medium-Mn steels, the steel is first heated in the austenitic region and then quenched to obtain a complete martensite microstructure. Subsequently, annealing followed by air cooling stabilizes a high retained austenite fraction. The quenching and partitioning (Q&P) process basically consists of complete austenitization followed by rapid cooling to the region between $M_s$ and $M_f$. Annealing at the quenching temperature or above is subsequently conducted to facilitate the partitioning of carbon from martensite to austenite [18–20]. The process of obtaining austenite from the initial martensite is referred to as “austenite reverted transformation” (ART) [10,14,18–22]. The heat treatments discussed above require cold rolling and a long annealing time (6–12 h). A long annealing time renders the austenite too stable and leads to the TRIP effect being significantly weakened. Herein, for medium-manganese steel, two different traditional heat treatments (Q&P and ART processes) are described. The shortcomings of the two traditional heat treatments are summarized. Our purpose in this is to explain the necessity for putting forward a new heat treatment. Thus, a novel heat treatment is proposed for the experimental medium-Mn steels to obtain high strength/high elongation, constituting the objective of the study.

2. Materials and Experimental Procedure

E. De Moor et al. [23] proposed a model which is used to design the composition of steels and predict the heat treatment temperature. The model suggests that the amount of austenite stabilized at room temperature depends on the content of carbon, aluminum, and manganese alloying elements enriched in austenite [15,16,19]. It further suggests that a temperature exists in the intercritical region that yields the maximum fraction of retained austenite at room temperature. In Thermo-Calc software, we input the composition of steel and set the temperature range. The prediction of the phase volume fraction of the experimental steels and composition in austenite could then be calculated.

The cast ingots were refined by Central Iron and Steel Research Institute in Beijing using a vacuum melting furnace. The real composition of the experimental steels was measured by atomic absorption spectrophotometry (SP-3803, Spectrum, Shanghai, China) and is given in Table 1. Then, the ingots were heated at 1250 °C for 3 h and hot forged between 1200 °C and 900 °C into billets of section size 100 mm × 30 mm. Subsequently, the billets were cooled in air to room temperature (RT). Finally, the billets were heated at 1250 °C for 5 h and hot rolled to sheets with 4 mm thickness after 7 passes of hot rolling; the primary rolling temperature was 1150 °C, the final rolling temperature was not lower than 850 °C, and the hot-rolled steel was finally cooled to room temperature at a cooling rate of 15 °C/s.

<table>
<thead>
<tr>
<th>Steel</th>
<th>Mn</th>
<th>Al</th>
<th>C</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>6Mn</td>
<td>6.00</td>
<td>3.20</td>
<td>0.20</td>
<td>90.60</td>
</tr>
<tr>
<td>8.5Mn</td>
<td>8.40</td>
<td>3.11</td>
<td>0.19</td>
<td>88.30</td>
</tr>
</tbody>
</table>

The traditional heat treatment for medium-Mn steels mentioned above requires a long holding time in the intercritical region, and it was proved to be inapplicable to the experimental steels [18–22]. Also, the long annealing time renders the retained austenite too stable. As a result, the TRIP effect is
weakened. Thus, an effective and simple two-stage heat treatment process was adopted (as shown in Figure 1): (1) Intercritical hardening—the as-hot-rolled steels were heated in the intercritical temperature region 600–850 °C for 1 h and then immediately quenched in water to room temperature. (2) Tempering—the quenched steels were tempered at 200 °C for 15 min, followed by air cooling to room temperature to obtain a good balance between ductility and strength. In the processes of annealing and tempering, carbon diffuses from ferrite to austenite, which enhances the stability of austenite, leading to superior ductility [6].

![Figure 1](image-url)

**Figure 1.** A heat treatment schedule for steels using an effective and simple two-stage heat treatment process.

Tensile specimens of a gauge width 12.5 mm and length 50 mm were taken parallel to the rolling direction and were machined from the heat-treated plate. A universal testing machine (SANSMT5000, MTS, Saint Paul, MN, USA) was used at a constant crosshead speed of 3 mm-min⁻¹ for tensile testing at room temperature. The microstructure was studied using an optical microscope (OM, OLYMPUS, OLYMPUS-GSX500, Tokyo, Japan), a field-emission scanning electron microscope (FE-SEM, SSX-550, Supra, Shimadzu, Tokyo, Japan), and a field-emission transmission electron microscope (FE-TEM, TECNAI G2-20, FEI, operated at 200 kV, Hillsboro, MO, USA). The OM and SEM samples were etched in an aqueous solution of 75% H₂O + 25% NaHSO₃. The TEM samples were twin-jet polished (Struers, Tenupol-5, Copenhagen, Denmark) in a solution of 95% alcohol and 5% perchloric acid at ~20 °C.

Using a direct comparison method [24], the austenite volume fraction was determined by X-ray diffraction (XRD, D/Max2250/PC, Rigaku, Tokyo, Japan) with Cu Kα radiation. The volume fraction of austenite was measured using the integrated intensities of the (311)α, (220)α peaks of ferrite and the (211)γ, (220)γ peaks of austenite. The austenite volume fraction, \( V_A \), was calculated using the equation [25]

\[
V_A = \frac{1}{N} \sum_{i=1}^{N} \frac{l_{i\gamma}}{R_{i\gamma}} + \frac{1}{M} \sum_{i=1}^{M} \frac{l_{i\alpha}}{R_{i\alpha}}
\]

where \( l_{i\gamma} \) and \( l_{i\alpha} \) are the integrated intensity values of the austenite and α-phase, respectively; \( N \) and \( M \) are the numbers of peaks of austenite and peaks of ferrite, respectively; and \( R_{i\gamma} \) and \( R_{i\alpha} \) are the standardization constant of austenite and ferrite, respectively, such that \( R_{(311)\gamma} = 2.282 \text{ m}^{-1}, R_{(220)\gamma} = 1.796 \text{ m}^{-1} \). 

\( R_{(211)\alpha} = 2.932 \text{ m}^{-1} \) and \( R_{(200)\alpha} = 1.269 \text{ m}^{-1} \).

**3. Results and Discussion**

**3.1. Design of the Chemical Composition of Steels**

The fraction of retained austenite as a function of the annealing temperature was predicted using the model proposed by De Moor et al. The modeling steps of the calculation process are as follows [23]:

```latex
\[ V_A = \frac{1}{N} \sum_{i=1}^{N} \frac{l_{i\gamma}}{R_{i\gamma}} + \frac{1}{M} \sum_{i=1}^{M} \frac{l_{i\alpha}}{R_{i\alpha}} \]
```
According to the Koistinen–Marburger (K–M) equation, the amount of transformation from austenite to martensite during the cooling process calculated using Equations (2) and (3), the fraction of retained austenite can be calculated using Equation (2) [26]:

\[
M_s = 550 - 350C - 40Mn + 30Al. \quad (2)
\]

According to the Koistinen–Marburger (K–M) equation, the amount of transformation from austenite to martensite during the cooling process calculated using Equation (3):

\[
f_m = 1 - \exp[-0.11 (M_s - T)] \quad (3)
\]

where \(T\) is the room temperature, which is taken as 20 °C.

![Figure 2. Schematic illustrations of the predictive model for austenite stabilization as a function of temperature for 6Mn steels and 8.5Mn steels: (a,d) phase fractions; (b,e) C, Al, and Mn content in austenite; and (c,f) calculated volume fraction of retained austenite.](image)

According to the initial austenite fraction (as shown in Figure 2a,d) and the newly formed martensite during the cooling process calculated by Equations (2) and (3), the fraction of retained austenite at room temperature can be obtained, as shown in Figure 2c,f. For 8.5Mn steel, a pronounced peak (~65.2 vol.%) was observed at ~755 °C, and for 6Mn steel, the maximum fraction of retained austenite reached ~37.7 vol.% at about 690 °C.

### 3.2. Microstructure Evolution and Mechanical Properties

SEM micrographs of the 8.5Mn and 6Mn intercritical hardened and tempered steels are shown in Figure 3. The microstructures of the 8.5Mn and 6Mn steels comprised intercritical ferrite (\(\alpha\)-ferrite), \(\delta\)-ferrite, retained austenite, and martensite, and the morphology and content of each phase varied significantly with change in the intercritical hardening temperature. The amount of stripelike ferrite decreased with increasing temperature, while the thin lath martensite gradually thickened with increasing temperature. Similar morphological changes were observed for martensite. The thin lath martensite was dispersed in packets (marked rectangle in Figure 3) in the parent austenite matrix.
which was broadly divided into granular and thin layers in the temperature range of 750–800 °C for 6Mn steel and at 800 °C for 8.5Mn steel, respectively. The content of retained austenite in 8.5Mn steel was higher than that in 6Mn steel, while the ferrite content in 8.5Mn steel was less than that in 6Mn steel. According to the Fe–C phase diagram, with increasing annealing temperature, the amount of ferrite decreased, the amount of austenite increased, and the austenite grains grew. With increased austenite grain size, the stability of austenite decreases. During quenching, more austenite was transformed into martensite. Thus, with increasing annealing temperature, the amount of martensite increased after quenching. Austenite in the two steels comprised a high fraction in the temperature ranges of 650–700 °C for 6Mn steel and 650–750 °C for 8.5Mn steel, but when they were treated at 750 °C and 800 °C, respectively, the amount of austenite decreased markedly because of extensive martensitic transformation. Moreover, we performed nanoindentation (Keysight G200XT nanoindenter system) of the 0.20C–8.65Mn–4.12Al–Fe medium-manganese steels after quenching from 750 °C and tempering from 200 °C (as shown in Figure 4), which could also explain the hardness differences between the austenite, δ-ferrite, α-ferrite, and martensite.

Figure 3. SEM micrographs of hot-rolled 6Mn and 8.5Mn steels after quenching from different temperatures. (a) 6Mn, 650 °C; (b) 6Mn, 700 °C; (c) 6Mn, 750 °C; (d) 6Mn, 800 °C; (e) 8.5Mn, 650 °C; (f) 8.5Mn, 700 °C; (g) 8.5Mn, 750 °C; and (h) 8.5Mn, 800 °C.
while 8.5Mn-750 steel (when 8.5Mn steel was heat treated at 750 °C) were examined in the experiment. The standard deviation is also presented in the results in Figure 6. When 8.5Mn steel and 6Mn steel were intercritically annealed at 600–750 °C, Mn promoted an increase in the retained austenite volume fraction. Comparing Figure 2c,f with Figure 5c, we note that the theoretical predictions are in good agreement with the experimental results. With increasing annealing temperature, the austenite volume fraction increased and the volume fraction of ferrite decreased. Thus, the austenite volume fraction increased with increasing temperature when 8.5Mn steel and 6Mn steel were intercritically annealed at 600–750 °C and 650–700 °C, respectively. Compared with the 6Mn steel, the 8.5Mn steel had a larger austenite content. With increasing temperature, the grain size of the austenite increased gradually, which resulted in a decrease in the austenite stability [27,28]. The higher the annealing temperature, the more austenite transformed to martensite during quenching when 6Mn steel was intercritically annealed in the temperature range of 750–850 °C. Because of martensitic transformation, the volume fraction of austenite decreased for 8.5Mn steel intercritically annealed at 800 °C.

The mechanical properties are presented in Figure 6. At least five samples for each temperature were examined in the experiment. Figure 5 shows the XRD pattern and measured fraction of austenite in heat-treated 8.5Mn and 6Mn steels at room temperature. The standard deviation is also presented in the results in Figure 6. Figures 6a and 5b show that ultimate tensile strength (UTS) increased continuously with increasing temperature, whereas the total elongation (TEL) almost decreased with increasing temperature after attaining peak values of 35.6% for 6Mn steel at 700 °C (6Mn-700 steel) and 39.5% for 8.5Mn steel at 750 °C (8.5Mn-750 steel). From the plots of TEL, UTS, and product of UTS and TEL in Figure 6, it can be seen that the mechanical properties of the 8.5Mn steel were obviously superior to those of the 6Mn steel, which mainly resulted from a greater austenite amount in the 8.5Mn steel, as apparent from Figure 5c. A comparison between Figure 6a,b and Figure 5c show that the total elongation had a trend similar to the austenite volume fraction. As shown in Figure 6, the 6Mn-700 steel (when 6Mn steel was heat treated at 700 °C) exhibited a TEL of 35.2%, UTS of 935 MPa, and PSE of 32.9 GPa%, while 8.5Mn-750 steel (when 8.5Mn steel was heat treated at 750 °C) was characterized by a TEL of 39.4%, UTS of 1113 MPa, and PSE of 43.9 GPa%. In contrast, as shown in Table 2 [9,11,13,29], other medium-Mn TRIP steels had lower or similar ductility with similar strength but required additional cold rolling work and a prolonged annealing time. Therefore, 8.5Mn and 6Mn steels heat treated at 750 °C and 700 °C, respectively, and tempered at 200 °C are significantly important. In particular, the two samples had a superior product of ultimate tensile strength and total elongation.

Figure 4. (a) A SEM image showing the indentation impressions on austenite grains, δ-ferrite grains, α-ferrite grains, and martensite grains in hot-rolled 0.20C–8.65Mn–4.12Al–Fe steels after quenching from 750 °C and tempering from 200 °C. (b) The nanohardness values of different phase grains. Each error bar represents a standard deviation.
Figure 4. (a) A SEM image showing the indentation impressions on austenite grains, δ-ferrite grains, α-ferrite grains, and martensite grains in hot-rolled 0.20C–8.65Mn–4.12Al–Fe steels after quenching from 750 °C and tempering from 200 °C. (b) The nanohardness values of different phase grains. Each error bar represents a standard deviation.

Figure 5 shows the XRD pattern and measured fraction of austenite in heat-treated 8.5Mn and 6Mn steels at room temperature. For the 8.5Mn steel, the retained austenite volume fraction increased to 58.5 vol.% at 750 °C, followed by a significant decrease to 24% at 800 °C. A similar variation in austenite content as a function of temperature occurred for the 6Mn steel. It attained a peak of 33 vol.% at 700 °C. Thus, Mn promoted an increase in the retained austenite volume fraction. Comparing Figure 2c,f with Figure 5c, we note that the theoretical predictions are in good agreement with the experimental results.

Table 2. Comparisons of medium-Mn transformation-induced plasticity (TRIP) steels.

<table>
<thead>
<tr>
<th>Composition</th>
<th>Initial Condition</th>
<th>UTS (MPa)</th>
<th>TEL (%)</th>
<th>PSE (GPa%)</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.13C–5Mn–0.6Si–1.1Al</td>
<td>CR/annealing</td>
<td>988/1461</td>
<td>16.7/8.6</td>
<td>16.5/12.6</td>
<td>9</td>
</tr>
<tr>
<td>0.13C–6Mn–0.5Si–3.1Al</td>
<td>CR/annealing</td>
<td>854/1161</td>
<td>21.7/12.4</td>
<td>18.5/14.4</td>
<td>11</td>
</tr>
<tr>
<td>0.2C–5Mn</td>
<td>HR/ART (12 h)</td>
<td>890</td>
<td>33</td>
<td>29.4</td>
<td>13</td>
</tr>
<tr>
<td>0.15C–4Mn</td>
<td>CR/annealing</td>
<td>1084/1187</td>
<td>19.1/15.9</td>
<td>20.7/18.9</td>
<td>29</td>
</tr>
<tr>
<td>0.2C–6Mn–3Al</td>
<td>HR/QT (1 h)</td>
<td>935</td>
<td>35.2</td>
<td>32.9</td>
<td>present work</td>
</tr>
<tr>
<td>0.2C–8.5Mn–3Al</td>
<td>HR/QT (1 h)</td>
<td>1113</td>
<td>39.4</td>
<td>43.9</td>
<td>present work</td>
</tr>
</tbody>
</table>

UTS, ultimate tensile strength; TEL, total elongation; PSE, product of strength and elongation; HR, hot rolled; CR, cold rolled; QT, quenching and tempering; ART, austenite reverted transformation.

3.3. Deformation Behavior

A comparison of the XRD patterns of undeformed 8.5Mn-750 and 6Mn-700 steels with fractured samples is shown in Figure 7. The results showed that the amount of austenite decreased to a small amount after tensile failure. The retained austenite volume fractions measured by XRD analysis of the 8.5Mn-750 and 6Mn-700 fractured samples were ~9.8 and ~7.7 vol.%, respectively, and the original austenite volume fractions of the 8.5Mn-750 and 6Mn-700 samples were 58.5 and 33 vol.%, respectively. Thus, it is indicated that a significant TRIP effect occurred during tensile deformation.
During quenching, the martensitic transformation occurs preferentially in low-carbon-containing austenite, and higher-carbon-content austenite is inactive. As shown in Figure 9, the strain in Stage III from each peak point of the fluctuation can be divided into two stages (rise and drop of the WH curve). In the rising stage, when a certain critical stress is attained, martensitic transformation is activated, and retained austenite with similar stability is retained. The reason for this is that carbon can effectively improve the stability of austenite [9]. Thus, the strain-induced martensite is the residual white lathy martensite. Comparing the micrographs of the undeformed steels with the fractured surface, it can be inferred that the black lath martensite is the martensite that was formed during heat treatment, while the white lath martensite with higher carbon content corresponds to strain-induced deformation. Black and white lath martensite subdivided the blocky austenite into thin film and granules. 

Based on Figures 7 and 8, there was a marked decrease in the variation in the martensite fraction was the opposite. Black and white lath were fluctuations. Based on previous studies [33–37], this behavior is related to the discontinuous TRIP effect. As shown in Figure 9, the strain in Stage III from each peak point of the fluctuation can be related to straining of martensite and ferrite because martensite phase transition in this stage is active. The WH behavior for the two steels is presented in Figure 9. A similar three-stage WH was observed in 6Mn steel, but in Stage III, there was a small difference: the WH decreased and there was a marked decrease in the WH. 

Figure 8 shows the microstructures of the 6Mn-700 and 8.5Mn-750 steels in the vicinity of the fracture position after tensile deformation. Based on Figures 7 and 8, there was a marked decrease in austenite, while the variation in the martensite fraction was the opposite. Black and white lath martensite subdivided the blocky austenite into thin film and granules. Comparing the micrographs of the undeformed steels with the fractured surface, it can be inferred that the black lath martensite is the martensite that was formed during heat treatment, while the white lath martensite with higher carbon content corresponds to strain-induced deformation. During quenching, the martensitic transformation occurs preferentially in low-carbon-containing austenite, and higher-carbon-content austenite is...
retained. The reason for this is that carbon can effectively improve the stability of austenite [9]. Thus, the strain-induced martensite is the residual white lath martensite.

Figure 7. XRD patterns of the fractured and undeformed steels: (a) 6Mn-700 steel and (b) 8.5Mn-750 steel.

According to previous studies [10,30], the work hardening rate evolution presents three domains in steels containing 5–7% Mn content: (1) a rapid decrease of the work hardening rate (WH), (2) an increase of the WH, and (3) a final smooth decrease of the WH. Shi et al. [10] suggested that the three-stage WH exists only when the volume fraction of retained austenite is greater than ~15%. A majority of studies [10,30–32] have concluded that the first stage is mainly associated with ferrite deformation, the second intermediate stage is related to the occurrence of the TRIP effect, and the third stage may be related to straining of martensite and ferrite because martensite phase transition in this stage is inactive.

The WH behavior for the two steels is presented in Figure 9. A similar three-stage WH was observed in 6Mn-700 steel, but in Stage III, there was a small difference: the WH decreased and there were fluctuations. Based on previous studies [33–37], this behavior is related to the discontinuous TRIP effect. As shown in Figure 9, the strain in Stage III from each peak point of the fluctuation can be divided into two stages (rise and drop of the WH curve). In the rising stage, when a certain critical stress is attained, martensitic transformation is activated, and retained austenite with similar stability continuously transforms to martensite. The WH increases rapidly in this stage because of the occurrence of the TRIP effect. As shown in Figure 10, the granular and lamellar grains located between adjacent martensite laths are austenite grains, as apparent from the diffraction pattern (inset in Figure 10b). It is known that the stability of granular and lamellar retained austenite is greater than that of equiaxed austenite [38]. Chiang et al. [39] showed that lamellar austenite exhibits high and sustained work hardening ability at high strain, while equiaxed austenite exhibits a relatively rapid transformation rate. It can be deduced that the fluctuations in Stage III are because of stress relaxation in the martensitic transformation process, which is due to the presence of austenite types of different morphology.
Therefore, a majority of lamellar austenite was transformed into martensite in Stage III. Thus, the lamellar austenite was retained. A majority of the austenite retained after tensile fracture was granular. It can be seen from the results in Figure 9 that the discontinuous TRIP effect in the 8.5Mn-750 sample contributed to its superior mechanical properties. With change in the intercritical temperature, the maximum volume fraction of retained austenite in good agreement with the experimental data. The predicted results provided a basis for the design of a heat treatment scheme. The higher Mn content in 8.5Mn steel was responsible for its higher austenite content.

In the case of the 8.5Mn-750 sample, as shown in Figure 9b, the WH behavior showed only two stages, which were similar to Stages II and III of the 6Mn-700 steel. The absence of the rapidly decreasing Stage I in the 8.5Mn-750 sample may be related to two main factors: first, a small amount of soft ferrite; second, a larger amount of austenite (Figures 3g and 5c) which is largely equiaxed and rarely transformed into martensite during the commencement of plastic deformation. It is reasonable to conclude that the equiaxed austenite first transformed into martensite intensely in Stage II, and lamellar austenite was retained. A majority of the austenite retained after tensile fracture was granular. Therefore, a majority of lamellar austenite was transformed into martensite in Stage III. Thus, the

**Figure 9.** Work hardening rate and stress–strain curves of 6Mn-700 steel (a) and 8.5Mn-750 steel (b).

**Figure 10.** TEM micrographs of undeformed 6Mn-700 steel and 8.5Mn-750 steel: (a,c) bright-field (BF) images and (b,d) dark-field (DF) images.
fluctuation in this stage results from the discontinuous TRIP effect associated with lamellar austenite, which relieves the local stress concentration produced by martensite laths. In contrast, Stage III of the 6Mn-700 sample fluctuated to a small degree because of the large amount of ferrite effectively relieving the internal stress induced by volume expansion associated with the martensite transformation process. Thus, the cracks marked with a rectangle in Figure 8a are in δ-ferrite grains. It can be seen from the results in Figure 9 that the discontinuous TRIP effect in the 8.5Mn-750 sample was stronger than that in the 6Mn-700 sample. Moreover, the products of the ultimate tensile strength and total elongation of the 8.5Mn-750 sample and 6Mn-700 sample were 43.9 GPa% and 32.9 GPa%, respectively. Thus, the strong discontinuous TRIP effect in the 8.5Mn-750 sample contributed to its superior mechanical properties.

4. Conclusions

(1) With change in the intercritical temperature, the maximum volume fraction of retained austenite stabilized at room temperature, calculated based on the model proposed by De Moor [23], was in good agreement with the experimental data. The predicted results provided a basis for the design of a heat treatment scheme. The higher Mn content in 8.5Mn steel was responsible for its higher austenite content.

(2) Optimal mechanical properties were obtained in 8.5Mn and 6Mn steels when they were subjected to intercritical hardening at 750 °C and 700 °C, respectively, and identical tempering at 200 °C. The 8.5Mn-750 steel was characterized by an excellent combination of a TEL of 39.4%, UTS of 1113 MPa, and UTS × TEL of 43.9 GPa%. The mechanical properties of the two experimental steels were superior to those of other medium-Mn TRIP steels, with the advantage of reduced cold rolling work and a shorter annealing time.

(3) Lamellar austenite, unlike the equiaxed microstructure, is more likely to cause stress relaxation during martensitic transformation, resulting in a discontinuous TRIP effect. The superior product of ultimate tensile strength and total elongation of the 8.5Mn steel is significantly related to a strong discontinuous TRIP effect.

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