Effect of Coiling Procedure on Microstructure and Mechanical Properties of C-Si-Mn Hot-rolled TRIP steel

Bin Zhang, Linxiu Du* and Ying Dong

The State Key Laboratory of Rolling and Automation, Northeastern University, Shenyang 110819, China
E-mail: zhangbinneu@163.com

Abstract. The effect of coiling procedure on microstructure evolution and mechanical properties were experimentally investigated. The results showed that the strength of experimental steels was increased and the elongation was decreased with temperature below Ms point. Among all the experimental steels, the mechanical properties of steel A and steel C were imperfect. It can be attributed to the high coiling temperature (500 °C) and low coiling temperature (285 °C). The mechanical properties of the experimental steel B were the most balanced and excellent. The yield strength, tensile strength and impact energy of the 1/4-Charpy size specimen were 685 MPa, 1010 MPa and, 16.3 J respectively, which could meet the demand of high strength structural steel in civilian applications the total elongation at -20 °C was 15.3% and TRIP effect caused by RA was obvious.

1. Introduction
Currently, advanced high strength steels (AHSSs) play a significant role in weight loss and improving product life for vehicles [1-2]. Transformation induced plasticity (TRIP) steel is one of AHSSs with better combination of strength and ductility. TRIP steels have a microstructure composed of bainite and retained austenite (RA) embedded in a ferrite matrix. When the deformation occurs, the metastable RA subjected to strain is progressively transformed to martensite, thus resulting a high hardening rate at high strain levels and contributing to improved ductility-known as the “TRIP” effect [3]. The RA with appropriate stability in TRIP steel can transform to martensite under a failure or crash, which improves the safety when steel is used for automotive parts. Therefore, the volume fraction and stability of retained austenite must be carefully controlled in order to gain the required combination of strength and ductility of TRIP steels.

In general, TRIP steel is mainly produced by cold rolling and hot rolling. For production of cold-rolled TRIP steel, as-rolled plates are usually processed with a two-step heat treatment i.e. intercritical annealing and isothermal bainite treatment [4]. The production cycle of cold rolling and annealing is too long, which is not conducive to industrial production. For conventional hot-rolled TRIP steels, the control of RA and other phases is normally realised by changing the cooling rate and time through multistage cooling, after hot rolling of austenite recrystallization zone [5]. However, it is not easy to adjust the cooling parameters and complicated to adopt multistage cooling. There are few articles on the production of 1000 MPa grade TRIP steel by reasonable design of chemical composition combined with thermo-mechanical control process (TMCP) and subsequent coiling process.

The present work was aimed at adjusting the proportion of RA and other phases in microstructure of C-Si-Mn TRIP steels; evaluated the effect of the coiling procedures. The volume fraction and
stability of RA were studied in detail. The relationship of microstructure and mechanical properties was investigated to reveal the effect of coiling temperature.

2. Experimental
The nominal chemical composition of the studied C-Si-Mn TRIP steel in weight % is 0.08-0.12C, 1.0-1.6Si, 1.7-2.3Mn, 1.3(Cr+Mo), 0.02Ti, and balance Fe. The martensite start transformation (M$_s$) was calculated to be 421 °C using the empirical Eq. (1) [6]:

\[
M_s = 539 - 432C - 7.5Si - 30.4Mn - 17.7Ni - 12.1Cr - 7.5Mo + 30Al \text{ (wt%)}
\]  

(1)

where C, Si, Mn, Cr, Ni, and Mo are the contents of the elements in wt%

The experimental steel was melted in a vacuum induction furnace and cast as 45 kg ingots. The ingot was forged into block of thickness of ~ 50 mm. The 50 mm thick slab was heated to 1200 °C for 2 h for austenization, and then air-cooled to 990 °C and subjected to two stages of controlled-rolling. The slabs were rolled and thinned to plates of 4.0 – 5.5 mm via eight passes on φ 450 mm trial rolling mill. The finish rolling temperature was controlled about 840 °C. Finally, the plates were water-cooled at the rate of 50 °C/s, 60 °C/s and 65 °C/s to 500 °C, 460 °C (above the M$_s$) and 280 °C (below the M$_s$) and cooled slowly in a furnace to room temperature. The specific schematic diagram of two different TMCP procedures is illustrated in Fig. 1.

![Figure 1. Schematic diagram of TMCP experimental steels of two different coiling procedures.](image)

The samples for metallographic studies were cut from the plate along the rolling direction (RD) and polished using standard metallography procedure and etched with 3% nital solution. Etched samples were examined for microstructure by OLYMPUS BX53M optical microscope (OM) and Zeiss Ultra 55 scanning electron microscope (SEM). A φ 3 mm foil with 45 μm was twin-jet polished using 12.5 vol% perchloric acid solution and characterized by FEI G2 F20 transmission electron microscope (TEM). In addition, the surface of samples was prepared by fine grinding and final electropolishing, in order to eliminate the residual stress. Volume fraction of RA was quantified from integrated intensity of (200)$_\alpha$, (211)$_\alpha$, (200)$_\gamma$, (220)$_\gamma$, and (311)$_\gamma$, diffraction peaks using a Cu target by X-ray diffraction (XRD) technique. The detailed calculation was in the light of Rietveld whole profile fitting method [7].

Dilatometry experiment was carried out to verify the M$_s$ temperature using cylindrical specimen of φ 3 mm×10 mm cut from the plate with length direction parallel to the RD. The specimen was heated to 1100 °C, held for 5min and cooled at a rate of 10 °C/s to ambient temperature.

The full-thickness tensile specimens with dimensions of 25 mm length were machined from the plates parallel to the RD. Tensile tests were carried out on a CMT5105 universal testing machine at room temperature with a crosshead speed of 3 mm/min according to ISO 6892-1:2009 [8]. The 1/4-Charpy impact specimens were cut from plates perpendicular to the RD with dimensions of 2.5
Impact tests were performed at 20 °C, 0 °C and -20 °C, respectively, on a MTS ZBC2452-B pendulum impact tester. The all data of mechanical properties was based on an average of three measurements for accuracy.

3. Results and Discussion

3.1. Alloy Design and Ms Temperature
In order to obtain good welding performance, the carbon content was controlled at a low range of 0.08 - 0.12. The high silicon of 1.0 - 1.6 was designed to suppress the precipitation of cementite. Because Si has low solubility in cementite and Si can improve the activity of carbon in ferrite with consequential influence on stabilization of austenite through carbon accumulation of austenite. Mn was added to decrease the Ms temperature so that low temperature bainitic transformation can be obtained. In addition, Cr and Mo were added to improve the hardenability. The addition of 0.02% Ti was used to enhance grain refinement and refine the microstructure of coarse grain heat affected zone.

The Ms temperature was calculated as 430 °C by tangent method, as shown in Fig. 2. The result of dilatometry experiment is consistent with the 421 °C estimated by empirical equation. In actual controlled cooling process, the finish cooling temperature must be strictly controlled above 430 °C in order to obtain ideal bainite microstructure with certain amount of RA.

![Figure 2. Dilation vs. temperature curves of experimental steel during continuous cooling to room temperature at cooling rate of 110 °C/s.](image)

3.2. Microstructure Characterization
The microstructures of different TMCP were characterized by secondary electron images, as shown in Fig. 3. With increasing cooling rate, the low temperature transformation microstructure changed from bainite to martensite. The steel A with high coiling temperature consisted of granular bainite, coarse M-A constituents and bainitic ferrite matrix (Fig. 3a). The complex phase structure of many islands is distributed in massive ferrite. The coarse structures formed at the top of the bainite transition temperature region. The steel B with medium cooling rate and secondary coiling temperature was composed of granular bainite, a few lath bainite and fine M-A constituents (Fig. 3b). No carbides were precipitated from granular bainite microstructure, which had a positive effect on the mechanical properties of experimental steel. When the finish cooling temperature was below the Ms temperature, the microstructure of steel C was comprised of lath martensite and ferrite (Fig. 3c). The prior austenite grain was filled with two or more packets corresponding to certain habit plane. [10].
Fig. 4 exhibits the TEM morphology of experimental steel subjected to different finish rolling temperature. The morphology of bainitic ferrite in steel A was shown in Fig. 4(a). The dislocations pile up on the grain boundary and the dislocation wall is formed. The precipitation marked by the yellow dotted circle was Ti(C,N) precipitating in rolling process, of which composition was presented in Fig. 4d. The Ti(C,N) has an obvious effect on grain refinement in high rolling temperature [11]. The TEM morphology of granular bainite in steel B was lath-like microstructure and there were a lot of dislocations on the laths (Fig. 4b). It can be seen from Fig. 4c that the microstructure of steel C was characterized by coarse martensite lath with high density of dislocations. The dislocation strengthening is one of the strengthening mechanisms to enhance strength [12].

$$V = \frac{1.4I_{\gamma}}{I_{\alpha} + 1.4I_{\gamma}}$$

3.3. The Relationship of Microstructure and Mechanical Properties

The XRD spectra of experimental steels subjected to different TMCP are presented in Fig. 5. The volume fraction of RA of sample surface was calculated using Eq. (2) [13]:

$$V = \frac{1.4I_{\gamma}}{I_{\alpha} + 1.4I_{\gamma}}$$
where $V_\gamma$ is the volume fraction of RA, $I_\gamma$ is the integrated intensity of $\gamma$ peaks, and $I_\alpha$ is the integrated intensity of $\alpha$ peaks. The volume fraction of RA was estimated to be 13.3%, 9.3% and 2.1% for steel A, steel B and steel C, respectively. The RA preferentially retains at prior austenite boundaries and then at lath boundaries and RA in laths has a high stability [14]. Moreover, the RA with certain fraction and stability can disperse strain and delay necking so as to obtain a certain elongation.

The volume fraction of RA was estimated to be 13.3%, 9.3% and 2.1% for steel A, steel B and steel C, respectively. The RA preferentially retains at prior austenite boundaries and then at lath boundaries and RA in laths has a high stability [14]. Moreover, the RA with certain fraction and stability can disperse strain and delay necking so as to obtain a certain elongation.

The engineering stress-strain curves and strain hardening rate curves are shown in Fig. 6. The yield strength of steel A, steel B and steel C were 808 MPa, 685 MPa and 1011 MPa. Discontinuous yielding phenomena have been observed in Fig. 6a, it is generally associated with low mobile dislocation density [15]. The tensile strength of steel A (1199 MPa) was higher than that of steel B (1010 MPa) and steel C (1160 MPa). When the coiling temperature was 500 °C, the steel A was held in the bainite region for too long, however, more of the intercritical austenite transformed into bainite with high strength. The elongation of steel A, steel B and steel C were 12.1%, 15.3% and 9.1%, respectively. The mechanical stability of RA is important in determining the tensile properties of TRIP steels. It should be noted that the lath microstructure exhibits a significantly larger strength but a poor elongation. The fraction of RA of steel A and steel B were more than steel C, thereby possibly generating sufficient TRIP effect. The granular bainite structure exhibits a higher strain hardening rate at low strain, as shown in Fig. 6b. The strain hardening rate is obtained by differentiating the true stress to the true strain \( \frac{d\sigma}{d\varepsilon} \). The strain value corresponding to the intersection of the true stress-strain curve and the strain hardening rate curve is equal to the uniform elongation. In all experimental steels, the uniform elongation of steel A (6.6%) was larger than that of steel B (6.0%) and steel C (2.3%) due to high volume fraction of RA.

**Figure 5.** XRD spectra of experimental steels

**Figure 6.** (a) The Engineering stress-strain curves of experimental steels (b) the true stress-strain curves and corresponding strain hardening rates of experimental steels
The total impact absorbed energy of steel A tested at -20 °C was 9.1 J and for steel B and steel C were 16.3 J and 11.9 J. Fig. 7 shows the fractographic features of the Charpy impact specimens. The fractographs of steel A and B are similar and there are observable cleavage facets (Fig. 7a and Fig. 7c). It is proposed that the cleavage facet can cause brittle fracture and reduce impact toughness. The fractograph of steel B is ductile fracture in which there are a lot of small and flat dimples (Fig. 7b). According to the classical Griffith theory, in plastic deformation, the fine M-A constituents can avoid the microcrack initiation by enhancing the initiation energy. Therefore, the impact toughness of steel B is superior to that of steel A and steel C.

![Figure 7. SEM observations on fracture surface of experimental steels](image)

(a) steel A (b) steel B (c) steel C

From the above observations and analyses, it can be summarized that high strength, excellent toughness, and good ductility can be obtained at the cooling rate of 60 °C/s and coiling temperature of 460 °C/s in steel B.

4. Conclusions
In this article, the microstructure and mechanical properties of C-Si-Mn hot-rolled TRIP steel on three different cooling procedures were studied under laboratory conditions. All of the conclusions are shown as follows:

1. The granular bainite was observed in steel A and steel B subjected to cooling temperature above Ms temperature, mainly ascribed to slow cooling at high temperature after rolling. The microstructure of steel C consisted of lath martensite and ferrite.

2. Of all experimental steels, the mechanical properties of experimental steel B are the most balanced and excellent. The yield strength, tensile strength, and total elongation were 685 MPa, 1010 MPa, and 15.3% respectively. The impact energy of 1/4-Charpy size specimen at -20 °C was 16.3 J.

3. The volume fraction and stability of steel B were the most suitable and TRIP effect caused by RA was obvious.

4. The fractographs demonstrated that the microstructure of dimples exhibited ductile fracture conducive to improve toughness.

5. Acknowledgements
This study is based on the work supported by the Natural Science Foundation of China under Grant no. 51604072. The financial and equipment support of the State Key Laboratory of Rolling and Automation is also gratefully acknowledged.

6. Reference
[1] Zhao J, Jiang Z. Thermomechanical processing of advanced high strength steels[J]. Progress in Materials Science, 94 (2018) 174-242.
[2] G. Jha, S. Das, S. Sinha, A. Lodh, A. Haldar, Mater. Sci. Eng. A 561 (2013) 394–402.
[3] Yi HL. Review on d-transformation-induced plasticity (TRIP) steels with low density: the concept and current progress. JOM 2014;66:1759–69.
[4] Fu B, Yang WY, Li LF and Sun ZQ 2014 Mater. Sci. Eng. A 603 134
[5] Hashimoto S, Ikeda S, Sugimoto KI and Miyake S 2004 ISIJ Int. 44 1590
[6] Lee SJ, Lee S and Cooman BC 2012 Int. J. Mater. Res. 103 1
[7] McCusker LB, Von Dreele RB, Cox DE, Louer D and Scardi P 1999 Appl. Crystallogr. 32 36
[8] ISO 6892-1. Metallic materials-tensile testing-Part 1: method of test at room temperature; 2009.
[9] ISO 148. Steel-Charpy impact test (V-notch); 1983
[10] Suikkanen PP, Cayron C, DeArdo AJ and Karjalainen LP 2011 J. Mater. Sci. Technol. 27 920
[11] Han Y, Shi J, Xu L, Cao WQ and Dong H 2012 Mater. Design 34 427
[12] Yakubtsov IA and Boyd JD 2008 Mater. Sci. Technol. 24 221
[13] Sugimoto K, Usui N, Kobayashi M and Hashimoto S 1992 ISIJ Int. 32 1311
[14] Sakuma Y, Matlock D, Krauss G 1992 Metall. Trans. A 23 1221
[15] Wang C, Shi J and Wang CY 2011 ISIJ Int. 51 (2011) 651