



## Microstructure and mechanical properties of hot-rolled and heat-treated TRIP steel with direct quenching process

He-song Wang<sup>a</sup>, Jian Kang<sup>a</sup>, Wei-xue Dou<sup>b</sup>, Yuan-xiang Zhang<sup>a</sup>, Guo Yuan<sup>a,\*</sup>, Guang-ming Cao<sup>a</sup>, R.D.K. Misra<sup>c</sup>, Guo-dong Wang<sup>a</sup>

<sup>a</sup> State Key Laboratory of Rolling and Automation (RAL), Northeastern University, Shenyang 110819, China

<sup>b</sup> School of Metallurgy, Northeastern University, Shenyang 110819, China

<sup>c</sup> Laboratory for Excellence in Advanced Steel Research, Department of Metallurgical, Materials and Biomedical Engineering, University of Texas at El Paso, El Paso, TX 79968, USA

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### ABSTRACT

In the present study, a hot-rolled and heat-treated TRIP steel (Fe-0.25C-1.23Si-2.09Mn-2.92Al wt%) was successfully produced by a simple process that involved intercritical heat treatment (IHT) followed by direct quenching without bainite reaction step. The microstructure of the as-cast ingot, as-hot-rolled sheet and heat-treated sheet was studied and related to mechanical properties of the heat-treated sheet. It was observed that  $\delta$ -ferrite was retained from the solidification stage and remained as a stable phase during different stages of processing because of the alloy design that included  $\sim 3$  wt% Al. Banded structure was obtained in both as-hot-rolled sheet and heat-treated sheet. The microstructure of 770–820 °C heat-treated sheet consisted of  $\delta$ -ferrite, retained austenite (RA),  $\alpha$ -ferrite and martensite. Based on the alloy design and direct quenching process,  $\sim 20$ –30% volume fraction of RA was obtained in the heat-treated sheet. The 780 °C heat-treated sheet exhibited an excellent combination of tensile strength and total elongation of 880 MPa and 28%, respectively.

### 1. Introduction

In recent years, energy saving and environment concern have led to significant demand for reducing the weight of the automotive parts [1]. Advanced high strength steels (AHSS) are being actively considered to reduce weight and improve safety of automobiles [2]. Transformation-induced plasticity (TRIP) steels are typical AHSS because of good combination of strength and ductility [3,4]. TRIP steels have been widely studied and developed. The microstructure of traditional TRIP steels mainly consists of ferrite, retained austenite (RA), bainite and some amount of martensite [5–8]. RA is a major constituent of TRIP steels. During tensile deformation, RA transforms to martensite producing a transformation-induced plasticity effect (TRIP effect) [9,10]. The TRIP effect delays necking and strengthens the steel [11,12]. Thus, an adequate amount of RA with appropriate stability is the key factor that governs the mechanical properties of TRIP steels.

The traditional process for TRIP steels includes a two-stage heat treatment following by hot rolling or cold rolling [13]. In the first stage, TRIP steels are cooled through or held in the austenite-ferrite two phase temperature region. The ferrite fraction can be controlled via adjustment of holding time, holding temperature and cooling rate [8,14]. In

the second stage, in order to avoid pearlite formation, TRIP steels are fast cooled to the bainite reaction temperature region and are held at a selected temperature for an appropriate time to ensure bainite formation and enrichment of RA with carbon. Generally, isothermal bainite reaction holding temperature and time are used to adjust RA fraction and its stability [15,16]. RA is obtained after cooling to ambient temperature from the bainite reaction temperature. Besides traditional TRIP steels, a low-alloy duplex, directly quenched TRIP steel is also considered for automobile applications [17]. The processing of low-alloy duplex, directly quenched TRIP steels is different from the traditional process. They are processed by a simple process, which involves intercritical annealing followed by direct quenching. Thus, it is easier to be produced because it does not require the bainite reaction step during processing. The duplex structure of ferrite and RA obtained by direct quenching after intercritical annealing is governed by the alloy design that includes  $\sim 3$  wt% Al and 0.4 wt% C. Al is beneficial in obtaining desired ferrite in TRIP steels such that RA is stabilized by partitioning of C that diffuses from ferrite into intercritical austenite on holding the steel in the austenite-ferrite two phase temperature region [18,19]. This is a potential and viable approach to process TRIP steel by direct quenching without the bainite reaction step. However, there are a few

\* Corresponding author.

E-mail addresses: [wanghesongsy@163.com](mailto:wanghesongsy@163.com) (H.-s. Wang), [yuanguoral@163.com](mailto:yuanguoral@163.com) (G. Yuan).

**Table 1**  
Chemical composition of experimental steel (wt%).

C	Si	Mn	Al	Fe
0.25	1.23	2.09	2.92	Bal.

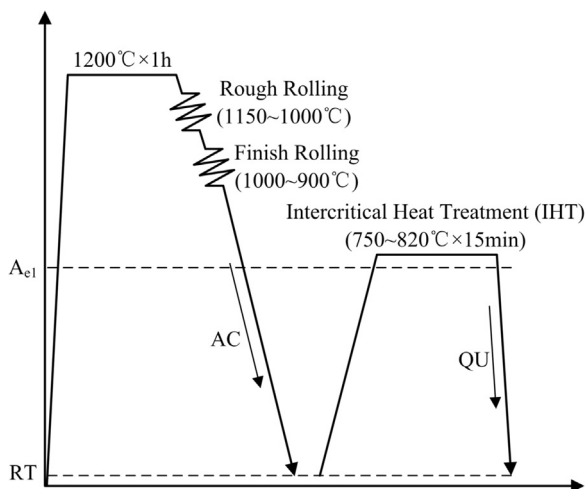
studies that dealt with directly quenched TRIP steel. Hot-rolled and heat-treated TRIP steels processed via direct quenching process has not been reported.

In the present study, a hot-rolled and heat-treated TRIP steel with direct quenching process was produced in the laboratory. The microstructure of the as-cast ingot, as-hot-rolled sheet and heat-treated sheet was characterized using optical microscope, scanning electron microscope, electron probe microanalyzer, electron backscattered diffraction and X-ray diffraction. Additionally, the mechanical properties of heat-treated sheet obtained by different intercritical heat treatment (IHT) temperature were studied. The objective of the study described here is to explore the microstructure and mechanical properties of hot-rolled and heat-treated TRIP steel with direct quenching process.

## 2. Experimental material and procedure

The chemical composition of the experimental steel is listed in Table 1. The steel was melted in a vacuum induction furnace. A 50-kg ingot of experimental steel was cast and forged into a billet with a section dimension of 80 mm × 70 mm. A schematic diagram of the processing route is presented in Fig. 1. The billet was homogenized at 1200 °C for 1 h, and then hot rolled to 2.8-mm thick sheet via 8 passes with rolling temperature in the range of 1150 °C to 900 °C. After finish rolling, the as-hot-rolled sheet was cooled to room temperature in air and cut into some small samples. Subsequently, the as-hot-rolled sheet was subjected to the intercritical heat treatment (IHT). The as-hot-rolled sheet samples were heated between 750 and 820 °C for 15 min, followed by direct quenching using oil. For convenience, the as-hot-rolled sheet samples heat treated at 750, 770, 780, 800 and 820 °C are referred as H75, H77, H78, H80 and H82, respectively.

Specimens were cut from the as-cast ingot, as-hot-rolled sheet and heat-treated sheet. They were first mechanically polished using standard metallography procedure and then etched with 4% nital (time 30 s) for microstructural characterization using OLYMPUS-BX53M optical microscope (OM), FEI QUANTA 600 scanning electron microscope (SEM), JXA-8530F electron probe microanalyzer (EPMA) and Zeiss Ultra-55 field emission scanning electron microscope (SEM) equipped with an electron backscattered diffraction (EBSD) system. Specimens



**Fig. 1.** Schematic illustration of hot rolling and heat treatment condition for the experimental steel. RT is room temperature, AC is air cooling and QU is quenching.

for EBSD studies were first mechanically polished and then electro-polished (current ~ 1.0 A, time ~ 20 s) using an electrolyte consisting of water, perchloric acid and alcohol in the proportion of 1:13:2 at room temperature. The distribution of elements in the selected samples was studied by EPMA equipped with energy dispersive X-ray spectrum (EDS) system (voltage 20 kV, current  $2 \times 10^{-8}$  A, step size 50 nm). The fraction of  $\delta$ -ferrite in the as-cast ingot was obtained using the image processing software, Image Pro Plus.

Phases present in the specimens were identified by X-ray diffraction (XRD) (Cu K $\alpha$  radiation, operating voltage 56 kV, current 182 mA, room temperature). In order to minimize the possible error originating from the mechanically-induced transformation of RA during sample preparation, XRD samples were electropolished using the process similar to EBSD samples. All XRD samples were scanned over the  $2\theta$  range from 40° to 101° (step rate 0.04°/s). Integrated intensities of (200) $\alpha$  and (211) $\alpha$  peaks as well as (200) $\gamma$ , (220) $\gamma$  and (311) $\gamma$  peaks were considered for calculating the volume fraction of RA via Origin 8.6. The volume fraction of RA ( $V_\gamma$ ) was obtained using the following equation [20]:

$$V_\gamma = 1.4I_\gamma / (I_\alpha + 1.4I_\gamma) \quad (1)$$

Where  $I_\gamma$  and  $I_\alpha$  are the average integral intensities of austenite and ferrite peaks, respectively.

The average carbon concentration of RA in the specimens was calculated using the following equation [21]:

$$a_\gamma = 3.556 + 0.0453x_c + 0.0056x_{Al} + 0.00095x_{Mn} \quad (2)$$

where  $a_\gamma$  is the lattice parameter of austenite in Angstroms,  $x_c$ ,  $x_{Al}$  and  $x_{Mn}$  are concentration of carbon, aluminum and manganese in austenite in wt%, respectively. The value of  $x_{Al}$  and  $x_{Mn}$  used in Eq. (2) was thermodynamically calculated by Thermo-Calc software combined with TCFE6 database (see next section). Silicon was not considered in Eq. (2), because it has insignificant effect on the lattice parameter of austenite [22].  $a_\gamma$  was calculated using Eq. (3) based on (200) $\gamma$ , (220) $\gamma$  and (311) $\gamma$  peaks and was determined from Nelson-Riley extrapolation method [23].

$$a_\gamma = \frac{\lambda \sqrt{h^2 + k^2 + l^2}}{2 \sin \theta} \quad (3)$$

where  $\lambda$ ,  $\theta$ , and  $(hkl)$  are the wavelength of the radiation, Bragg angle and three Miller indices of a plane, respectively.

Tensile specimens of dimensions 25 mm (gauge length) × 6.25 mm (width) × 2.8 mm (thickness) were machined with longitudinal axis parallel to the rolling direction. The tensile test was performed using a CMT5105-SANS machine at room temperature at a crosshead speed of 2 mm/min. The strain hardening exponent ( $n$ ) was calculated using the following equation:

$$n = d(\ln \sigma) / d(\ln \epsilon) \quad (4)$$

where  $\sigma$  is true stress and  $\epsilon$  is true strain.

## 3. Results

### 3.1. Alloy design and microstructure of the as-cast ingot

Fig. 2(a) shows the fraction of equilibrium phases in the temperature range of 600–1600 °C calculated by Thermo-Calc software combined with TCFE6 database. The solidification of the alloy started with the formation of  $\delta$ -ferrite. The maximum fraction of  $\delta$ -ferrite was ~ 80%. With decrease in temperature, some  $\delta$ -ferrite transformed to austenite. There was no fully austenitic region and the maximum predicted amount of austenite was ~ 90%. This is related to the alloy composition that includes ~ 3 wt% Al. Previous studies suggested that high Al content added to steel is beneficial in obtaining  $\delta$ -ferrite at room temperature [24,25]. Moreover, from Fig. 2(a) it can be seen that the critical temperature, start temperature and finish temperature of

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