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Influence of intercritical tempering temperature on impact toughness of a quenched and tempered medium-Mn steel: Intercritical tempering versus traditional tempering

C. Sun[a,*], S.L. Liu[b,c], R.D.K. Misra[c], Q. Li[a], D.H. Li[a]

[a] Research Institute, Nanjing Iron and Steel Co., Ltd., Nanjing 210035, China
[b] Collaborative Innovation Center of Steel Technology, University of Science and Technology Beijing, 30 Xueyuan Road, Beijing 100083, China
[c] Laboratory for Excellence in Advanced Steel Research, Materials Science and Engineering Program, Department of Metallurgical, Materials and Biomedical Engineering, University of Texas at El Paso, 500W. University Avenue, El Paso, TX 79968, USA

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ABSTRACT

The influence of intercritical tempering temperature on impact toughness of quenched and tempered 0.05C-5.42Mn medium-Mn steel was studied and compared with traditional tempering. The experimental steel had high hardenability because of high Mn-content. Lath-like α'-martensite without retained austenite was obtained over a wide range of quenching rate of 0.5–30 °C/s, and the quenched steel showed high strength but low impact toughness. On intercritical tempering at 625 °C and 665 °C, the impact toughness was enhanced, as compared to traditional tempering at 570 °C. The reversed austenite enriched with Mn and C formed between the martensite laths was the underlying reason for the increased absorbed crack propagation energy, and the ductile-brittle transition temperature (DBTT) was reduced because of increased stability of reversed austenite. Compared to the steel tempered at 625 °C, the steel tempered at 665 °C contained more reversed austenite, but the reversed austenite was less stable because of reduced enrichment of Mn and C. The enrichment or depletion of Mn and C in austenite and martensite was thermodynamically studied by DICTRA. In striking contrast to the steels tempered between 625 and 665 °C, twinned martensite was formed in the steel tempered at high temperature of 700 °C, and the steel exhibited impact toughness lower than the quenched steel.

1. Introduction

Strength, ductility and toughness are the most important mechanical properties for steels. In recent years, advanced high strength steels with remarkable cryogenic toughness have been extensively studied for cryogenic applications. These steels generally contain nickel [1–5]. However, the high price of Ni has shifted the focus towards Mn. In this regard, medium-Mn steels with 3–10 wt% Mn have received significant attention [6,7]. In majority of the cases, these steels are considered and designed as third-generation advanced high strength steels (3rd-G AHSS) with high product of tensile strength and elongation, for application in the automotive industry. The Fe-Mn alloys exhibit lath-like martensitic microstructure and room temperature mechanical properties similar to Fe-Ni steels [8,9]. However, Fe-Mn martensitic steels exhibit lower cryogenic toughness and higher ductile-to-brittle transition temperature (DBTT) compared to Fe-Ni martensitic steels, because of severe intergranular fracture [10,11].

A novel medium-Mn offshore platform steel was recently studied [12–14]. The carbon content of the steel was ∼ 0.1% and the manganese-content was ∼ 5%. Using TMCP followed by tempering or annealing, an excellent impact toughness at − 60 °C and high yield strength greater than 690 MPa was obtained in steels of thickness of 80 mm and greater. TRIP effect associated with reversed austenite was proven to be beneficial for the enhancement of ductility [14]. More importantly, the combined effect of reversed austenite and ferrite/martensite with low dislocation density in the tempered steel led to excellent toughness at − 60 °C [12]. Furthermore, it was demonstrated that the film-like austenite greatly improved cryogenic impact toughness but its contribution to strain hardening capacity was limited, while the block-type austenite greatly enhanced strain hardening capacity but its contribution to low-temperature impact toughness was poor [15].

Considering the aforementioned studies, our objective here is to obtain high toughness at cryogenic temperatures in medium-Mn steels via controlling reversed austenite in the martensite matrix for cryogenic applications. In the present study, the influence of tempering temperature on impact toughness of 0.05C-5.42Mn (wt%) steel is studied,

* Corresponding author.
E-mail address: chaos_83@163.com (C. Sun).

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and the intercritical tempering is compared with the traditional tempering. A theoretical analysis of the observations is presented.

2. Experimental procedure

The chemical composition of the experimental steel in wt% was 0.05 C, 5.42Mn, 0.09Si, 0.02Al, 0.004 P, 0.001 S and balance Fe. The steel was melted in a vacuum induction furnace and cast into a 50 kg ingot. The ingot was hot forged into a square billet to 140 mm thickness. The billet was homogenized in a box-type electrical resistance furnace at 1100 °C for 2.5 h, then hot rolled via 7 passes to a 12 mm thickness plate, and finally air-cooled to room temperature.

The rolled plate was reheated to the quenching temperature of 820 °C, and held isothermally for 30 min in a box-type electrical resistance heat treatment furnace, and then water-quenched to room temperature. Next, the quenched steel was reheated to different tempering temperatures of 570 °C, 625 °C, 665 °C and 700 °C, and held isothermally for 60 min, and then air-cooled to room temperature. The steels tempered at 570 °C, 625 °C, 665 °C and 700 °C are henceforth referred as T570, T625, T665 and T700, respectively. Cryogenic treatments were also carried out, in which the tempered steels were cooled to − 40 °C and − 110 °C, and held isothermally for 30 min, and then cooled to room temperature.

The Ac_1 and Ac_3 temperatures were determined by a DIL805A dilatometer, and the specimens were slowly heated at a linear rate of 0.05 °C/s. The CCT diagram was determined by Gleeble-3800 thermal simulator. The specimens were heated to 820 °C and held isothermally for 30 min, and then cooled at different linear rates of 0.5 °C/s, 1 °C/s, 3 °C/s, 5 °C/s, 10 °C/s and 30 °C/s. Vickers hardness tests and optical metallographic observations were carried out. For optical metallographic observations, the specimens were mechanically polished and etched using 3.5 vol% nital solution.

Standard Charpy v-notch (CVN) impact tests were performed at 25 °C, − 40 °C, − 80 °C, − 110 °C, − 150 °C and − 192 °C, using a NI500A impact tester on specimens of dimensions 10 × 10 × 55 mm^3. Impact energies were determined and force-displacement curves were recorded.

Transmission electron microscopy (TEM) specimens were mechanically thinned to ~ 50 µm thickness, followed by twin-jet electropolishing at a voltage of 30 V and at − 20 °C, using an electrolyte consisting of 9% perchloric acid and 91% absolute ethyl alcohol. The specimens were examined using a field-emission transmission electron microscope (FEI Tecnai G^2 F20) operated at an accelerating voltage of 200 kV. The content of Mn in austenite and martensite was determined by energy-dispersive X-ray spectroscopy (EDS).

In addition, the volume fraction of austenite was determined by a XRD-7000 diffractometer using CuKα radiation at a scanning speed of 1°/min and step size of 0.02°. The specimens were first mechanically polished and then electropolished using an electrolyte consisting of 12% perchloric acid and 88% absolute ethyl alcohol at room temperature. The integrated intensities of (200)γ, (220)γ, (311)γ, (200)α, and (211)α diffraction peaks were used to determine the austenite volume fraction [16].

3. Results and discussion

3.1. CCT diagram and quenched steel

The CCT diagram of the experimental steel is presented in Fig. 1. The two-phase region was determined to be between 576 °C and 790 °C. At a cooling rate in the range of 0.5–30 °C/s, only α'-martensite transformation was obtained and no retained austenite was detected by XRD. The martensite start temperature (M_s) was ~ 381 °C, and the martensite finish temperature (M_f) was ~ 220 °C. Both M_s and M_f changed to a small extent with cooling rate. The Vickers hardness of transformation product increased from 332 to 347 as the cooling rate was increased

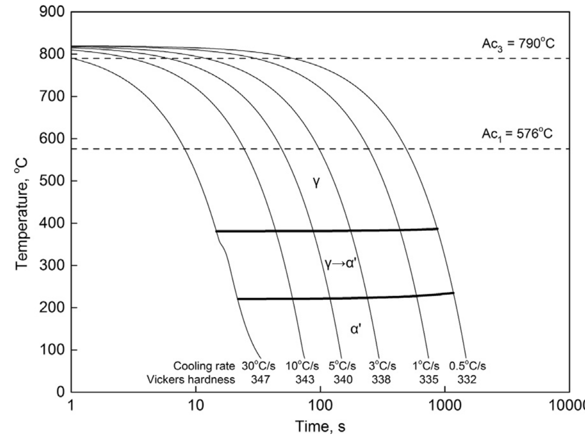

Fig. 1. CCT diagram of experimental steel. The austenitizing temperature was 820 °C.

from 0.5 °C/s to 30 °C/s, but the increase was not very significant. Optical micrographs of phase transformation products are presented in Fig. 2. Lath-like α'-martensite was obtained, and there was no apparent difference between the samples subjected to different cooling rates.

The microstructure of as-quenched steel is presented in Fig. 3. The microstructure consisted of long and straight lath-like martensite with dimensions of ~250 nm width (Fig. 3a), and high density of dislocations were present within the lath (Fig. 3b). No retained austenite was detected by XRD in the as-quenched steel. The tensile and impact properties of as-quenched steel are presented in Table 1. A high yield strength (951 MPa) and a high tensile strength (1062) were obtained after quenching, but the elongation (9.2%) and impact energy (32 J) was low.

3.2. Impact toughness of tempered steels

The CVN impact energy of tempered steels in the temperature range of − 196 °C and 25 °C (RT) is presented in Fig. 4. The impact toughness was a strong function of tempering temperature. The steel T665 showed higher impact energy at 25 °C than steel T625. With decreased test temperature, both steels showed a decrease in impact energy. This was particularly true for steel T665. Accordingly, steel T665 had lower impact energy below − 80 °C, and also exhibited higher ductile-to-brittle transition temperature (DBTT) of ~ − 60 °C compared to steel T625 (~ − 120 °C). Here the DBTT was determined as the temperature corresponding to the half of the sum of the impact energy measured at 25 °C and the impact energy measured at − 192 °C. At − 110 °C, the steel exhibited quite low impact energy of 16 J, while the steel T625 still possessed high impact energy of 126 J. The steels T570 and T700 showed low impact energies even at 25 °C, and the values were 22 J and 34 J respectively.

The force-displacement curves of impact tests at 25 °C (RT) and − 110 °C are presented in Fig. 5. At 25 °C (Fig. 5a), the unstable crack propagation of steel T625 and steel T665 was not a significant observation. The crack propagation energy of steel T665 corresponding to the area under the force-displacement curve was higher than steel T625. When the impact test temperature was lowered to − 110 °C (Fig. 5b), the unstable crack propagation of steel T625 was enlarged, and the stable crack propagation continued to be obvious. For steel T665, however, no plastic deformation occurred prior to the maximum force and only unstable crack propagation occurred, which resulted in low crack propagation energy.

3.3. Microstructure of tempered steels

The TEM micrographs of steel T570 are presented in Fig. 6. In steel T570, the morphology of long and straight martensite laths changed