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The determining role of reversed austenite in enhancing toughness of a novel ultra-low carbon medium manganese high strength steel

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The ultra-low carbon medium Mn steel was subjected to quenching and intercritical tempering process. The 650 °C tempering route led to 15% volume fraction of austenite with combination of high yield strength of 650 MPa, low yield ratio of 0.86, and excellent elongation of ~31%. The impact energy at −20 °C was enhanced more than twenty times from 10 J to 213 J. The formation of reversed transformation austenite during the intercritical tempering process contributed to the observed superior mechanical properties.

Keywords: Medium manganese steel; Reversed transformation austenite; Toughness; Phase stability

Advanced heavy steel plates with combination of high strength and superior low-temperature toughness are preferred as constructural materials for ship hull, bridges, buildings, pressure vessels, and offshore structures. In TMCP (thermo-mechanically controlled processing) and Q&T (quenching-tempered) of advanced thick steel plates, the deformation and cooling behavior is non-uniform over the entire thickness, because of which inhomogeneous microstructure and mechanical properties are obtained in the plate. Although large amount of Ni, Cr, Mo, and Cu are added to increase the hardenability of austenite, unfavorable microstructures such as coarse granular bainite and Widmanstatten ferrite are generated at small reductions and low cooling rate in the center of the heavy steel plate. Moreover, the yield ratio is 0.92 to 0.94 in the Q&T thick plate steel [1–2].

The austenite phase of Hadfield steel was adequately stabilized by the addition of high amount of Mn and C (10–15 wt.% Mn and 0.9–1.4 wt.% C). These steels are used for wear-resistant applications because of their high strain hardenability. The twin-induced plasticity steel, referred as the second generation advanced high strength steel (AHSS) and designed for automotive steel sheets, exhibits excellent combinations of strength and ductility, but the addition of significant amount of Mn (15–30 wt.%) makes the steel less cost-effective [3–4]. Thus, medium Mn automotive steels containing 5–8 wt.% Mn are being developed as third generation AHSS via intercritical annealing of original martensitic microstructure in the ferrite–austenite region, and metastable austenite was stabilized at room temperature because the austenite becomes highly enriched in Mn during the annealing process [5–10]. The improved combination of strength and ductility was obtained in the cold-rolled and annealing thin steel plate, while the study of medium-Mn alloying to heavy plate steel is not reported in the literature to the best of our knowledge.

We explore here the effect of alloying with 5 wt.% Mn in enhancing hardenability and improving the homogeneity of microstructure and mechanical properties throughout the thickness of the heavy steel plate. Ductility and toughness were studied in relation to the formation of reversed transformation austenite by tempering at different temperatures.

The experimental steel was melted in vacuum induction furnace and cast as 150 kg ingot. The chemical composition of steel in wt.% was 0.04 C, 0.2 Si, 5.0 Mn, 0.002 S, 0.003 P, 0.01 Al, 0.2 Cu, 0.3 Ni, 0.2 Mo, and balance Fe. Ultralow carbon design was preferred from the view point of weldability and reduced cementite content. Cu and Ni were added to enhance corrosion resistance, and a small amount of Mo to suppress martensitic temper embrittlement. The 140 mm thick slab was heated to 1200 °C for 3 h. After air-cooling to 960 °C, followed by rolling via 5 passes on a trial rolling mill with roll diameter of 450 mm, the slab was rolled to a plate of ~80 mm thickness. The finish rolling was controlled at 900 °C. Subsequently, the plate was directly water-quenched (DQ) to room temperature using an ultra-fast cooling system. The DQ plates were again reheated to the tempering temperature of 600 °C and

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650 °C, respectively, for 10 min, and it took ~50 min to reach the isothermal temperature. Then, the plate was air-cooled to room temperature at a cooling rate of ~0.125 °C/s. Another 50 mm thick DQ plates was subjected to tempering at 650 °C for 1 h.

The tensile specimens of dimensions 6 mm diameter and 25 mm length were machined from the plates parallel to the rolling direction. The tensile tests were conducted at room temperature using a crosshead speed of 3 mm/min using a Shimadzu AG-X universal testing machine. Charpy v-notch impact tests were performed at 20 °C, −20 °C, −40 °C, −60 °C, respectively, with standard specimens (dimensions: 10 × 10 × 55 mm) with a v-notch parallel to the rolling direction using Instron Dynatup 9200 series instrumented drop weight impact tester, consistent with ASTM E23 specification. The samples were cooled to −5 °C below the test temperature to take into consideration the rise in temperature during transfer of cooled sample to the Charpy v-notch impact tester. The 140 mm thick slab was rolled to a plate of ~80 mm thickness, and the total reduction was ~43%. Therefore, there was no difference in Charpy impact toughness in the longitudinal and transverse directions. The Vickers microhardness was measured utilizing an FM700 hardness-testing machine employing a load of 500 g. The toughness data are an average of five measurements for each test condition, and strength and hardness data are an average of three measurements.

Transmission electron microscopy (TEM) studies were conducted using 3 mm diameter thin foils, electropolished using a solution of 8% perchloric acid and alcohol, and examined by FEI Tecnai G^2 F20 TEM at an accelerating voltage of 200 kV. The Mn-content in austenite after tempering was determined by energy-dispersive X-ray spectroscopy (EDX). The fracture surface of impact specimens was studied by a FEI Quanta 600 scanning electron microscope (SEM). The volume fraction of austenite were determined by X-ray diffraction (XRD) using a Cu-Kα radiation source. The theoretical calculations concerning evolution of various phases with temperature, such as fcc, bcc, and cementite, were studied using Thermocalc combined with TCFE6 database for thermodynamic calculation in equilibrium.

The TEM micrographs of experimental steel are presented in Figure 1. The DQ microstructure consisted of lath martensite and high density of dislocations (Fig. 1a). Austenite nucleated at a few martensitic laths during tempering at 600 °C for 10 min, and the density of dislocations was reduced (Fig. 1b). When the tempering temperature was increased to 650 °C, a high content of film-shaped and globular-shaped reversed austenite formed along the martensitic laths. The reversed transformation austenite had a width of ~100–150 nm, while the width of martensitic lath was 200–300 nm (Fig. 1c). The Mn content in austenite on tempering for 10 min was ~10.3–13.7 wt.%, consistent with 9.36 wt.% calculated by Thermocalc. The nucleation and growth of austenite involve migration of interface boundary into the ferrite phase and diffusion of interstitial C and substitutional Mn between ferrite and austenite phases [5], which was confirmed by atom probe tomography and simulation[11–12]. It was proposed in 0.6C–1.8Ni wt.% steel that austenite formed by martensitic reversion follows an athermal, i.e. displacive mechanism [13]. The reversion mechanism is changeable between diffusive and displacive in low-alloyed steels, which depends on the heating rate and that martensitic reversion occurs if the heating rate is adequate to avoid diffusive reversion [8,11,14]. The reverse transformation proceeds diffusively when the heating rate is less than 15 °C/s in 0.05C–(5–9)Mn in wt.% steel [8]. The diffusive reversion occurs at a heating rate of 300 °C/s in 0.15C–5Mn wt.% steel, while the mechanism changes to displacive at 450 °C/s [14]. Given the very slow heating rate of the experimental heavy steel plate, the diffusive mechanisms dominate the reverse transformation. The diffusionless reverted austenite generally exhibited lath-shaped grains, while the globular-shaped reverted γ-grains were formed by diffusive mechanism [8]. The film-shaped and globular-shaped austenite in the experimental steel had similar Mn content because Mn further diffusion during 650 °C isothermal process after heating process. The measured ~15.8–17.5 wt.% Mn concentration in steel tempered at 600 °C that formed fine austenite was significantly higher than that predicted by partitioning between austenite and ferrite of 12.7 wt.%, and the reason is discussed below.

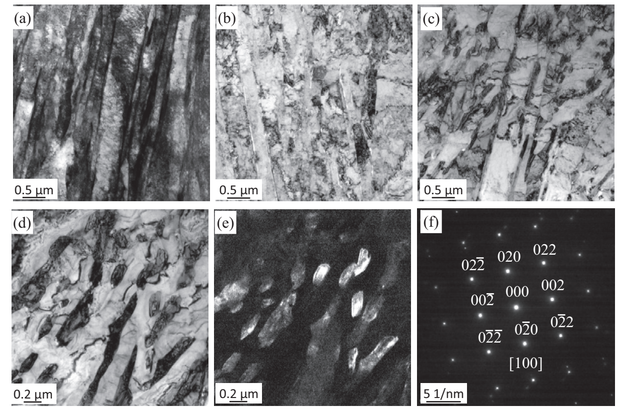

Figure 1. TEM micrographs: (a) Direct-quenching; (b) tempering at 600 °C; (c) tempering at 650 °C; (d) bright field image of 650 °C reversed austenite at high magnification; (e) dark field image of 650 °C reversed austenite at high magnification; and (f) selected area electronic diffraction pattern of austenite.

The XRD spectra of experimental steels are presented in Figure 2. The XRD peak of DQ steel indicated α-phase (black plot), implying that the microstructure consisted of martensite. When the steel was tempered at 600 °C, γ peak was slightly enhanced (red plot), indicating the formation of a very small amount of austenite, whose content was difficult to be determined. As tempering temperature was

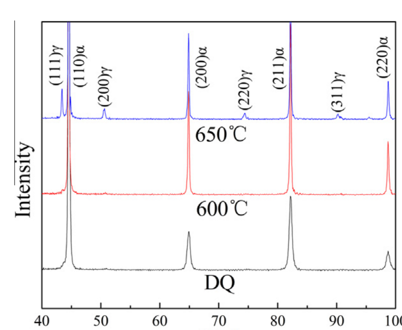

Figure 2. XRD spectra of experimental steels. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)