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The effect of tempering temperature on microstructure, mechanical properties and bendability of direct-quenched low-alloy strip steel

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ABSTRACT

The tempering of re-austenized, quenched and tempered (RAQT) martensitic steels is an extensively studied and well understood field of metallurgy. However, a similar understanding of the effect of tempering on direct-quenched (DQ) high-strength steels has been lacking. Now, for the first time, the effect of tempering in the range of 250–650 °C on the strength, toughness, bendability, microstructure, crystallography and dislocation density of a DQ steel is reported. In the case of tempering at 570 °C, the effects of having a RAQ or DQ starting condition are compared. For the composition and thermal cycles studied, it was found that a peak tempering temperature in the range of 570–600 °C resulted in a DQT steel with an optimal balance of strength, bendability and toughness, i.e. a yield strength greater than 960 MPa, a minimum usable bending radius of 2 times the sheet thickness and T28J of − 50 to − 75 °C depending on the test direction. Crystallographic texture, dislocation density and the distribution of carbides are important factors affecting the bendability of DQT strip. Tempering had no effect on texture, but strongly influenced the size and distribution of carbides thereby resulting in differences in bendability and impact toughness transition temperature.

1. Introduction

The conventional process route for quenched and tempered steels is hot rolling and cooling followed by re-heating, quenching and furnace tempering (RAQT). However, direct quenching (DQ), where the steel plate is directly quenched to martensite after hot rolling, is an interesting energy efficient alternative to the conventional process route as one reheating process is omitted. While conventional RAQT is well understood, the industrially relatively new DQT process requires more research as the state of the austenite prior to quenching can differ substantially from that in conventional quenching and tempering. Furthermore, better understanding for fundamental microstructural differences between RAQT and DQT is valuable in developing new materials.

Already in 1971 Caron and Krauss [1] published an article on the effect of tempering between 400 and 700 °C on the microstructure of 0.2%C RAQT martensite. Krauss continued researching tempered martensite publishing review papers in 1999 [2] and 2017 [3]. In general, RAQT martensite follows a tempering-temperature dependent microstructural evolution including carbon segregation and ε-carbide precipitation (< 200 °C), rod shaped carbide precipitation (200–350 °C) and recovery with spheroidal carbide precipitation (> 330 °C). Regarding mechanical properties, it has been reported that low-temperature tempering leads to an increase in yield strength, tensile strength, hardness and elastic limit until the tempering temperature exceeds 300 °C after which softening processes lead to a decrease in strength. The strength and hardness of martensite increase strongly with increasing carbon content.

The effect of different tempering temperatures on the toughness of RAQT steels has also been studied. Comparisons between low-temperature tempering (LTT) and high-temperature tempering (HTT) were reported by Kennett and Findley [4]. In 1978 Horn and Ritchie [5] described the mechanisms of tempered martensite embrittlement in medium-carbon steels. According to their study, toughness is improved when the tempering temperature is less than 300 °C, whereas toughness is decreased between 300 and 500 °C due to tempered martensite embrittlement, i.e. precipitation of rod-like carbides. On the other hand, in the Gleeble study by Kennett and Findley [4] on the effect of austenite grain size on the properties of quenched and tempered 0.2%C steel, no difference existed in toughness between the quenched and 200 °C tempered (LTT) conditions.

Regardless of the wide and respectable work done by many research

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Table 1
Chemical compositions of experimental steel (in wt%).

C	Si	Mn	V	Cr	Mo	Ti	Al	B	P	N	S
0.1	0.2	1.0	0.08	1.0	0.63	0.012	0.03	0.0016	0.006	0.0053	0.0005

groups across world, the effect of tempering temperature on bendability has not yet been reported. Furthermore, the effect of tempering over a wide range of temperatures has not been fully studied in the case of direct quenched steels. An earlier paper [6] showed that high-temperature tempering increased the bendability of DQ steel. Now this paper reports for the first time the effect of different tempering temperatures on the strength, toughness and bendability of DQ steel.

2. Experimental

2.1. Experimental steels

The experimental composition given in Table 1 was cast on a pilot plant. After reheating, slabs were rolled to 6 mm thick strips using a pilot scale hot rolling mill followed by direct quenching. Finishing rolling temperature (FRT) prior to direct quenching was 915 °C, which is above the A_3 austenite-to-ferrite transformation temperature, i.e. 883 °C according to the Andrews' equation [7]. Further details of the rolling procedure are given in the paper by Saastamoinen et al. [8].

To obtain fully re-austenitized and quenched (RAQ) material for comparative studies, DQ specimens were re-austenitized at 910 °C for 30 min and subsequently quenched into a tank of water.

Direct-quenched and tempered (DQT) specimens were produced by tempering the DQ material in a laboratory furnace with peak temperatures at 250, 400, 500, 570, 600 and 650 °C. A slow heating rate of 35 °C/h and no holding time at the peak temperature was employed to simulate the tempering of industrial-scale steel coils. After reaching the peak temperature, specimens were cooled at 40 °C/h to room temperature. To compare the effects caused by the initial microstructure, i.e. DQ or RAQ, RAQT specimens were produced by tempering RAQ material at 570 °C with the same heating and cooling rates. The test matrix is presented in Table 2.

2.2. Microstructural evaluation

To study the evolution of microstructure as a function of tempering temperature, a Zeiss UltraPlus field emission scanning microscope (FESEM) was used. The effect of as-quenched condition, i.e. DQ or RAQ, on texture and grain sizes was revealed with the aid of electron backscatter diffraction (EBSD) measurements and analyses using AZtec*HKL* acquisition and analysis software at both the quarter-thickness position and just below the top surface of the steels. In the EBSD measurements, the FESEM was operated at 15 kV and the step size employed was 0.2 µm. The scanned field size was 80 × 480 µm for subsurface measurements and 80 × 360 µm for quarter thickness measurements.

For EBSD polished samples, X-Ray diffraction (XRD) line broadening studies were carried out at subsurface (150 µm) using Cu K_α radiation on a Rigaku SmartLab 9 kW X-ray diffractometer and PDXL2 analysis software was used to estimate the lattice parameters, microstrains and crystallite sizes of the experimental steels. To determine these parameters, Rietveld refinement was used. As in earlier studies, [6,8], dislocation densities were calculated using the Williamson-Hall method (Eq. (1)) [9,10]:

$$\rho = \sqrt{\rho_s \rho_p} \tag{1}$$

where ρ_s is dislocation density calculated from strain broadening and ρ_p is dislocation density calculated from particle i.e. crystallite size, see Eqs. (2) and (3). According to Williamson et al. [9,10]:

$$\rho_s = \frac{k\varepsilon^2}{Fb^2} \tag{2}$$

and

$$\rho_p = \frac{3n}{D^2} \tag{3}$$

where ε is microstrain, b is burgers vector, F is an interaction factor assumed to be 1, factor k is taken as 14.4 for body-centered cubic metals and D is crystallite size. In Eq. (3), n is the number of dislocations per block face, which is taken as 1. This assumption is based on Williamson et al. [10], as they assume that the metal is broken up into blocks and the dislocations are lying in the boundaries between the blocks. Therefore, n = 1 can be used as an assumption, which will lead to the minimum dislocation density.

2.3. Mechanical testing

Three parallel tensile tests were carried out at room temperature for the DQ, DQT and RAQT conditions in accordance with the European standard EN 10002 using flat specimens (6 × 20 × 120 mm^3), cut with their axes both at 0° and 90° to the rolling direction (RD). For specimens both parallel to and transverse to the RD, three Charpy-V experiments at each of six test temperatures from − 140–0 °C were performed with 6 mm thick specimens in the DQ, DQT and RAQT conditions according to the European standard EN 10045.

Tanh-fitting as described by Oldfield [11] using the procedure later introduced by EricksonKirk et al. [12] was used for defining Charpy V transition curves and determining both 28 J transition temperatures and 95% confidence intervals. To convert sub-sized T28J values to full size equivalents, the method presented by Wallin [13,14] was used.

Three-point bending tests were carried out in an Ursviken Optima 100 bending machine to a bending angle of 90 degrees. Plate specimens, 6 × 100 × 300 mm^3, were bent with the bend axes parallel to both the transverse and rolling directions. The die opening width (W) employed was 75 mm and the punch radii (r) ranged from 6 mm to 30 mm. After bending, the quality of the bent surface was examined visually, as described in Ref. [15]. In the evaluation, the formation of a slight nut-shape did not lead to rejection, but if any surface waviness or more severe surface defects appeared, the bend test was considered as failed. The minimum bending radius was that resulting in a defect-free bend.

Table 2
Experimental matrix.

	Starting condition	Tempering conditions
DQ	Direct-quenched	As-quenched + tempered at 250, 400, 500, 570, 600, 650 °C
RAQ	Re-austenized and quenched	As-quenched + tempered at 570 °C

3. Results and discussion

3.1. Microstructure

3.1.1. The effect of tempering temperature on microstructure of DQ material

Previous studies on the bendability of modern high-strength steels have shown that the microstructure and texture near the surface of the steels are the main factors correlating with bendability [16]. On the