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The effect of finish rolling temperature and tempering on the microstructure, mechanical properties and dislocation density of direct-quenched steel

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ABSTRACT

A unique batch tempering treatment for industrial scale direct-quenched steel coils has been studied using laboratory simulations. The tempering treatment was non-isothermal with slow heating to 570 °C and slow cooling to simulate the tempering of large steel coils. The paper presents the effect of finishing rolling temperature (FRT) relative to the non-recrystallization temperature (T_{NR}) and the effect of long time tempering on the microstructure, dislocation density and mechanical properties of direct-quenched coiled strips. Conditioning austenite below the recrystallization stop temperature resulted in a finer effective grain size distribution, which correlated strongly with the impact toughness of the final product. Furthermore low finish rolling temperature resulted in partially ferritic microstructures while higher finishing rolling temperatures led to mixtures of bainite and martensite. Dislocation densities determined with TEM and XRD showed somewhat different trends regarding the effect of tempering: intra-lath dislocation density, as measured with TEM, showed a statistically significant drop in only one case, while XRD analysis indicated a drop in all cases. Furthermore, no significant correlation between finishing rolling temperature and dislocation density existed in XRD studies. The XRD results indicate that the decrease in dislocation density corresponds to about 100 MPa lower dislocation strengthening. However, precipitation hardening and potential internal micro stress relief compensates this as yield strength remains unchanged or even increases during tempering.

1. Introduction

The growing demand for more energy efficient solutions during the 21st century has led to increased interest in the development of structural steels with higher strength made using novel energy efficient production processes.

Quenching and tempering martensitic structural steel provides an excellent combination of strength and toughness. In the case of direct-quenched low-carbon steels, good combinations of strength and toughness are obtained without tempering as a result of the good toughness associated with low-carbon martensite formed from thermomechanically rolled austenite [1,2]. For a given steel strength, a direct-quenched version has a leaner composition than a quenched and tempered steel; therefore, the direct-quenched material is more prone to softening in weld intercritical and sub-critical heat affected zones. This is a disadvantage in some applications, which means there is also a demand for direct-quenched and tempered steels. Since direct-quenched (DQ) steels are most efficiently produced in the form of large coils on hot strip mills, the tempering of such coils is an interesting potentially cost-efficient method for producing quenched and tempered (DQ-T) high-strength steels that resist softening in the heat affected zone. In this work, batch annealing of full industrial scale steel coils was simulated in the laboratory.

As-quenched martensite is formed by diffusionless transformation and is therefore highly metastable. During tempering supersaturated interstitial carbon redistributes by diffusion and the lattice structure transforms into body-centred cubic. Carbon tends to diffuse to dislocations where it starts forming cementite particles in the case of low-alloy martensite. In addition during tempering, the somewhat tangled dislocation structure of the as-quenched martensite changes into a

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lower energy cell-structure with an overall lower dislocation density. As the strength of martensite is controlled by solid solution, dislocation density, grain size and precipitation strengthening, the microstructural changes occurring during tempering are important factors with respect to the final strength of steel [3–7].

The conventional process route for quenched and tempered steels is hot rolling and cooling followed by re-heating, quenching and furnace tempering. However, direct quenching, where the steel plate is directly quenched to martensite after hot rolling, is an interesting energy efficient alternative to the conventional process route as the reheating process is omitted. Even though conventional quenching and tempering is well understood [8] direct quenching leads to different as-quenched microstructures compared to re-austenitized and quenched steels, which have better strength – toughness combinations [9]. Therefore the changes occurring during the tempering of the industrially relatively new direct-quenched steels requires more research. In direct quenching, the state of the austenite prior to quenching can differ substantially from that in conventional quenching and tempering. For example, it is known that compared to the recrystallized state, pancaked austenite, which can be controlled by thermomechanical treatment prior to direct quenching, improves the toughness of martensite in quenched condition prior to any tempering [2,3]. Furthermore, the tempering of large coils would require much longer heat treatment times than are required for conventional quenched and tempered plate products.

This paper presents the effect of finishing rolling temperature (FRT) relative to the non-recrystallization temperature (T_{NR}) and the effect of long time tempering on the microstructure, dislocation density and mechanical properties of direct-quenched coiled strips. The results provide an important contribution to our understanding of how best to obtain optimum combinations of high strength and toughness.

2. Materials and Methods

2.1. Experimental Steels

The composition of the vacuum treated continuously cast steel studied in this work is shown in Table 1. Hot rolling experiments were performed on a pilot scale hot rolling mill. After reheating at 1225 °C 210 mm thick slabs were thermomechanically rolled to 6 mm thick plates with three different finish rolling temperatures (775, 865 and 915 °C) and direct-quenched to room temperature (DQ materials). The two lowest finishing rolling temperatures were below the non-recrystallization temperature (T_{NR}), 876 °C (according to Boratto equation, [11]), thereby providing pancaked austenite giving better strength – toughness combinations in the direct-quenched condition [1]. The two lowest finishing rolling temperatures were also below A_3, i.e. austenite-to-ferrite transformation temperature (according to Andrews equation, [12]) leading to the possibility of other non-martensitic phase transformations prior to quenching.

Following direct quenching part of the material was tempered in a laboratory furnace following a thermal cycle designed to simulate the heat treatment of large steel coils, i.e. slow heating at 35 °C/h to a peak temperature at 570 °C, zero soaking time, and then cooling at 40 °C/h to room temperature. The untempered direct-quenched material is referred to as DQ while the tempered DQ material is identified by the code DQ-T.

2.2. Microstructural Evaluation

The direct-quenched microstructures prior to tempering were studied using light optical microscopy (OM) and laser scanning confocal microscopy (LSCM) to identify prior austenite grain morphology parameters. Nital was used to reveal transformation microstructures in OM and field emission scanning microscopy (FESEM).

The prior austenite grain structure was quantified at the quarter-thickness by the mean linear intercept (MLI) grain size $\bar{L}_{RD}$ and $\bar{L}_{ND}$ along the two principal directions—the rolling direction (RD) and the normal direction (ND)—using the sections shown in Fig. 1. Based on these measurements, the mean prior austenite grain size ($\bar{L}$), total reduction below the recrystallization temperature ($R_{tot.}$), the grain boundary surface area per unit volume (S_V) and relative standard error were determined using Eqs. (1)–(3) [13].

$$R_{tot} = \left(1 - \sqrt{\frac{\bar{L}_{ND}}{\bar{L}_{RD}}}\right) \times 100\% \quad (1)$$

$$S_V = 0.429\frac{1}{\bar{L}_{RD}} + 1.571\frac{1}{\bar{L}_{ND}} \quad (2)$$

$$Relative\ standard\ error = \frac{0.65}{\sqrt{number\ of\ measurements}} \quad (3)$$

Electron backscatter diffraction (EBSD) was performed using AztecHKL acquisition and analysis software. The FESEM for the EBSD measurements were made using an accelerating voltage of 15 kV and a step size of 0.2 μm for total area of 80 × 360 μm to obtain large representative area from the whole microstructure. Lath size (d_l) and effective grain size (d) were determined as equivalent circle diameter (ECD) values with low-angle (2.5–15°) and high-angle boundary misorientations (> 15°), respectively.

X-Ray diffraction (XRD) line broadening studies were carried out using Cu K_α radiation on a Rigaku SmartLab 9 kW X-ray diffractometer and PDXL2 analysis software to estimate the lattice parameters, microstrains and crystallite sizes of the experimental steels. Furthermore, dislocation densities were calculated using the Williamson-Hall method (Eq. (4)) [14,15]:

$$\rho = \sqrt{\rho_s \rho_p} \quad (4)$$

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. (5)–(6). Furthermore according to Williamson et al. [14,15]:

$$\rho_s = \frac{k\ \varepsilon^2}{F\ b^2} \quad (5)$$

and

$$\rho_p = \frac{3n}{D^2} \quad (6)$$

where ε is microstrain, b is burgers vector, F is an interaction factor assumed to be 1, factor k is assumed as 14.4 for body-centred cubic metals and D is crystallite size. In the equation n is dislocations per block face, assumed as 1. This assumption is based on Williamson et al. [15], as they state and assume that metal is broken up into blocks and the dislocations are lying in the boundaries between the blocks. Therefore, value 1 can be used as an assumption, which will lead to the minimum dislocation density.

Thin foils for TEM studies were prepared using twin-jet

Table 1
Chemical compositions of experimental steel (in wt%). T_{NR} is 876 °C and A_3 is 883 °C.

C	Si	Mn	V	Cr	Mo	Ti	Al	B	P	N	S
0.1	0.2	1.0	0.08	0.98	0.63	0.012	0.03	0.0016	0.006	0.0053	0.0005