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Effect of direct quenching on microstructure and mechanical properties of medium-carbon Nb-bearing steel

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Abstract: The influence of direct quenching (DQ) on microstructure and mechanical properties of 0.19C-1.7Si-1.0 Mn-0.05Nb steel was studied. The microstructure and mechanical properties of reheat quenched and tempered (RQ&T) steel plate were compared with those of direct quenched and tempered (DQ&T) steel plates which were hot rolled at different finish rolling temperatures (1173 K and 1123 K), i.e., recrystallization-controlled-rolled direct-quenched (RCR&DQ) and controlled-rolled direct-quenched (CR&DQ), respectively. The strengths generally increased in the following order: RQ&T<RCR&DQ&T<CR&DQ&T. Strength differences between the CR&DQ&T and RQ&T conditions as high as 14% were observed at the tempered temperature of 573 K. The optical microscopy of the CR&DQ&T steel showed deformed grains elongated along the rolling direction, while complete equiaxed grains were visible in RQ&T and RCR&DQ&T steels. Transmission electron microscopy (TEM) and electron backscattering diffraction (EBSD) of the DQ steels showed smaller block width and higher density of dislocations. Inheritance of austenite deformation substructure by the martensite and differences in martensite block width were ruled out as major causes for the strength differences between DQ and RQ steels.

Key words: Direct quenching (DQ), Block, Mechanical properties, Electron backscattering diffraction (EBSD), Nb steel
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1 Introduction

Recently, some steel producers have been seeking to eliminate a heat treatment step as a means of reducing costs and increasing efficiency by installing in-line water cooling units (Taylor and Hansen, 1991). The objective is to produce a martensite or bainitic on immediate quenching after hot deformation (referred to as “direct quenching (DQ)”). In conventionally off-line quenched-and-tempered steels (referred to as “reheat-quenched (RQ)”), the austenite composition and grain size, which play important roles in determining the ultimate structure and properties, are controlled largely by the austenitizing temperature. In contrast, quenching immediately after hot deformation allows for some control of the “condition” of the austenite prior to transformation. For example, generating martensite from deformed austenite by rolling below the austenite recrystallization temperature may provide some of the property improvements traditionally associated with ausforming (Taylor and Hansen, 1990; Weiss and Thompson, 1992; Chang, 2002).

To date, in the investigations of DQ steels, most studies have concentrated on steel tempered at relatively high temperatures (Morikawa and Hasegawa, 1986; Taylor and Hansen, 1988; Mekkawy *et al.*, 1990). Hence, little information is available on these steels after low temperature tempering treatments, and the strengthening mechanism is not yet clear. This study emphasizes the strengthening mechanism of DQ tempered at low temperature. The correlations between microstructure and mechanical properties of direct quenched and tempered (DQ&T) and reheat quenched and tempered (RQ&T) conditions were investigated.

2 Experimental

The chemical composition of the steel plate is shown in Table 1. The steel was melted in a vacuum induction furnace and forged to a 50-mm thick plate. Fig.1 is a schematic illustration of different heat-treatment processes. For the RQ&T process, the ingot was rolled to a 12-mm thick plate, the rolling ratio was 20% in each rolling pass, the plate was air cooled from the finish rolling temperature to room temperature after rolling, and was reaustenitized at 1193 K for 1 h and quenched to room temperature. This plate was tempered at 573 K for 3 h. For the DQ&T, the ingot was rolled to a 12-mm thick plate, the rolling ratio was 20% in each rolling pass, and the finish rolling temperature was 1123 K (CR&DQ&T) and 1173 K (RCR&DQ&T). The plate was directly quenched in water with an approximate cooling rate of 303 K/s after finish rolling. This plate was tempered at 573 K for 3 h.

Table 1 Chemical composition of the steel (%,w/w)

C	Si	Mn	Nb	S	P	N
0.19	1.70	1.00	0.05	0.005	0.006	0.0037

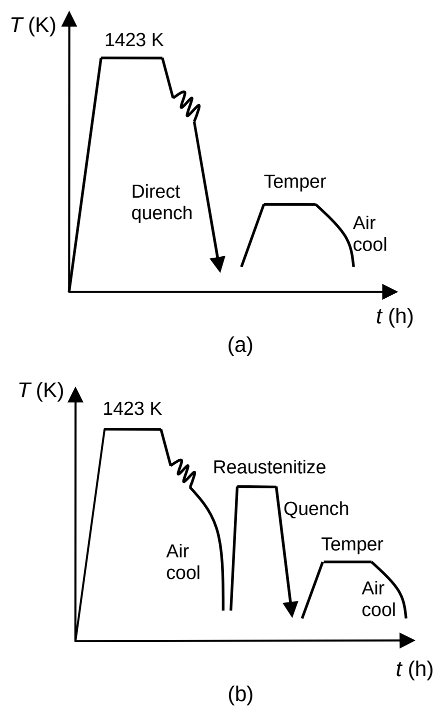

Fig. 1 Schematic diagrams of different heat-treatment processes. (a) DQ&T; (b) RQ&T

The tensile tests of all the specimens were undertaken at room temperature. Charpy V-notch impact testing of the specimens was conducted at 293 K. The average of three consistent test results was recorded as the value for the corresponding specimens. To show the microstructure, the tempered specimens were etched in 4% (w/w) Nital. Subsequently, the microstructure was observed by a Leica MEF-4M microscope (Leica Corp., Germany). The tempered martensite structure including precipitates and carbides was examined in an H-800 transmission electron microscope (TEM) (Hitachi Ltd., Japan). Electron backscattering diffraction (EBSD) (EDAX Corp., USA) installed in a high resolution field emission scanning electron microscopy (SEM) (JEOL Ltd., Japan) was used to analyze crystallographically the lath martensite structure.

3 Results and discussion

3.1 Optical microscopy

Fig. 2 shows optical photomicrographs of the RQ and the DQ steels. There is tempered martensite in both DQ&T and RQ&T. The RCR&DQ&T processing produced a predominantly equiaxed prior austenite grain structure, as shown in Fig. 3a, due to the recrystallization during hot deformation. Fig. 3b shows an elongated or “pancaked” austenite grain structure which resulted from controlled rolling, where the finish rolling temperature is below the austenite recrystallization temperature. The conventional process (RQ&T) resulted in a comparatively fine equiaxed austenite grain structure, as shown in Fig. 3c.

3.2 Transmission electron microscopy

Fig. 4 illustrates TEM images of DQ&T and RQ&T specimens. In all samples, the matrix was composed of parallel martensite laths and the interior of laths was heavily dislocated. However, it is difficult to confirm which specimen has a higher density of dislocation. Fig. 5 shows precipitates distributed throughout the matrix. The precipitation behavior of NbC particles varies with the fabrication process of DQ&T and RQ&T steels. Fig. 5 shows that precipitates of DQ&T steel are a little smaller than those of

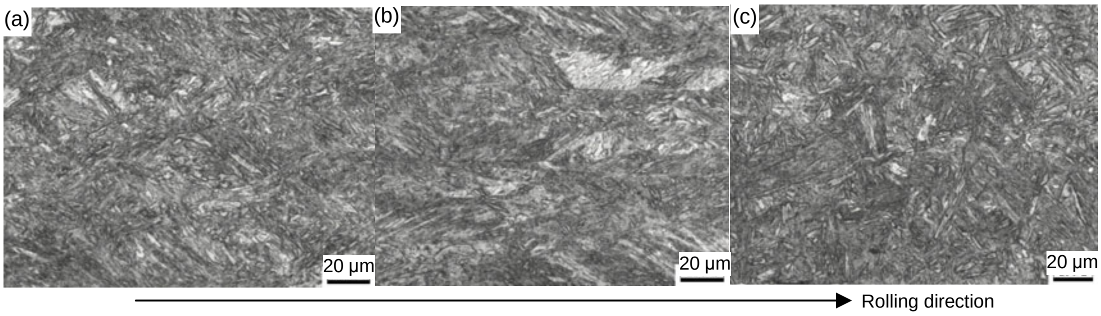

Fig. 2 Typical tempered martensite structure in the tested steel. (a) RCR&DQ&T; (b) CR&DQ&T; (c) RQ&T

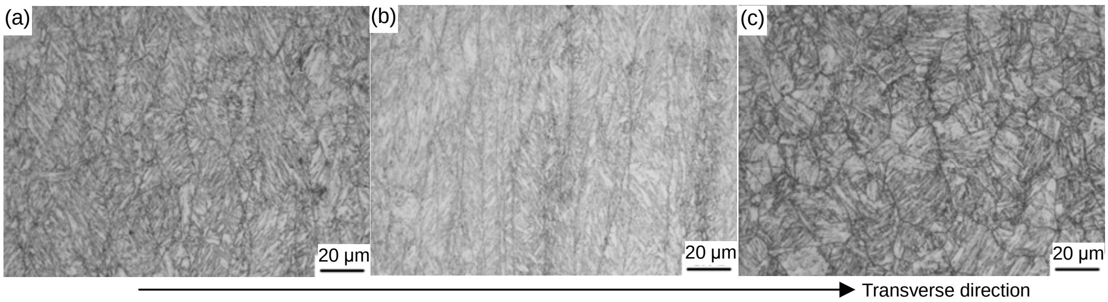

Fig. 3 Typical prior austenite grain structures in the tested steel. (a) RCR&DQ&T; (b) CR&DQ&T; (c) RQ&T

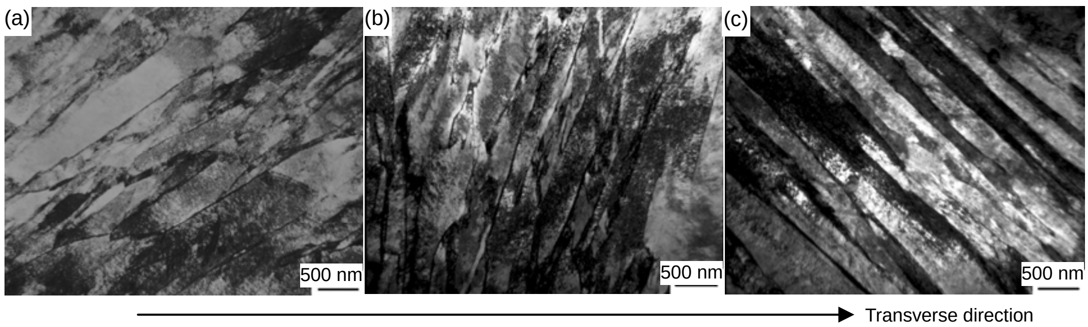

Fig. 4 TEM images of tempered martensite structure. (a) RCR&DQ&T; (b) CR&DQ&T; (c) RQ&T

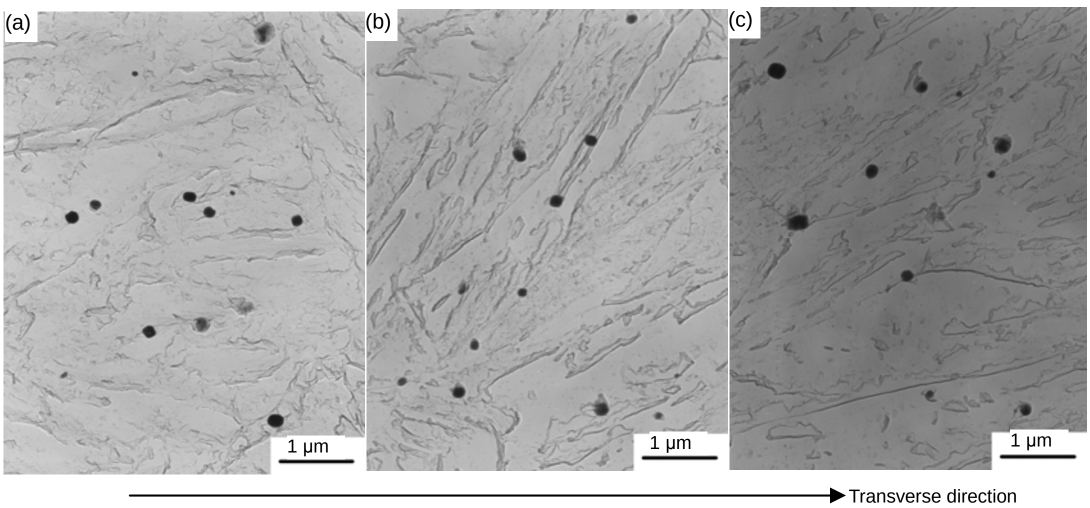

Fig. 5 TEM micrographs of extraction replicas prepared from as-quenched and tempered at 573 K. (a) RCR&DQ&T; (b) CR&DQ&T; (c) RQ&T

RQ&T steel. The mean precipitates sizes were found to be above 60 nm both in DQ&T and RQ&T steels, also the volume fraction of NbC is smaller, the maxima is 0.047%. Thus, precipitation strengthening in austenite is comparatively small, and there is little difference between DQ&T and RQ&T steels.

3.3 Electron backscattering diffraction

EBSD orientation maps of the lath martensite structure in different manufacturing processes are shown in Fig. 6. The misorientation of the packet boundaries and block boundaries should be larger than 15°. The block is an aggregation of the laths with the same crystallographic orientation (Krauss, 1999; Kitahara *et al*., 2006). Black lines in Fig. 6 show the boundaries with misorientation angles larger than 15°. The block width was measured and evaluated using the average linear intercept method. The block width for all processing conditions is given in Table 2.

Table 2 Block width for different processes

	RQ&T	RCR&DQ&T	CR&DQ&T
Block width (μm)	2.9	2.5	1.7

Fig. 7 shows the distribution of boundary misorientation angles in different manufacturing processes. It is obvious that the DQ&T's number fraction of low angle boundary (5.5%) is more than RQ&T process (3.5%), which means that the density of dislocation associated with DQ&T is higher than that of RQ&T process, because the low angle boundary consists of a series of dislocations.

3.4 Mechanical properties

Table 3 lists the tensile test results and impact test results. The strengths associated with RQ&T are lower than those of the same steels after DQ, the strength generally increases in the following order: RQ&T<RCR&DQ&T<CR&DQ&T. The variation in impact toughness was not significant.

Table 3 Mechanical properties of the tested steel in different processes

Process	R_m (MPa)	$R_{p0.2}$ (MPa)	A_{kv} (J)
RCR&DQ&T	1570	1315	30
CR&DQ&T	1645	1345	23
RQ&T	1435	1175	30

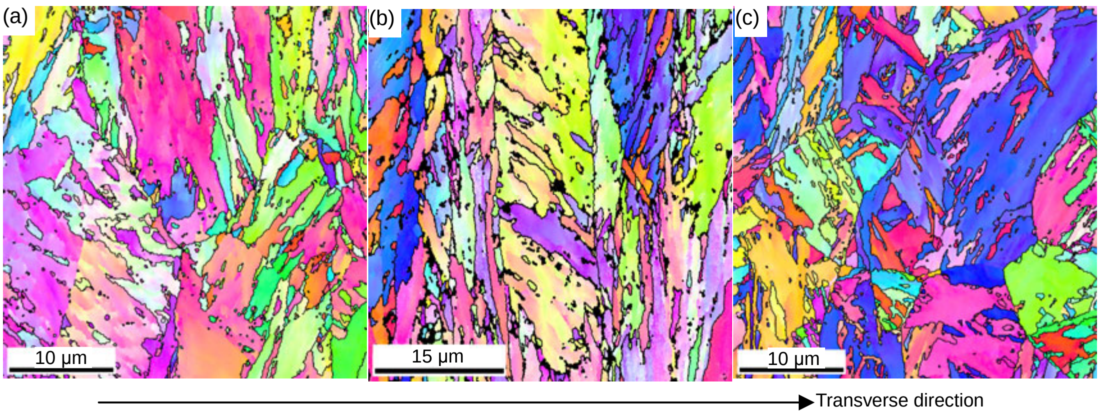

Fig. 6 EBSD orientation maps in different manufacture processes. (a) RCR&DQ&T; (b) CR&DQ&T; (c) RQ&T

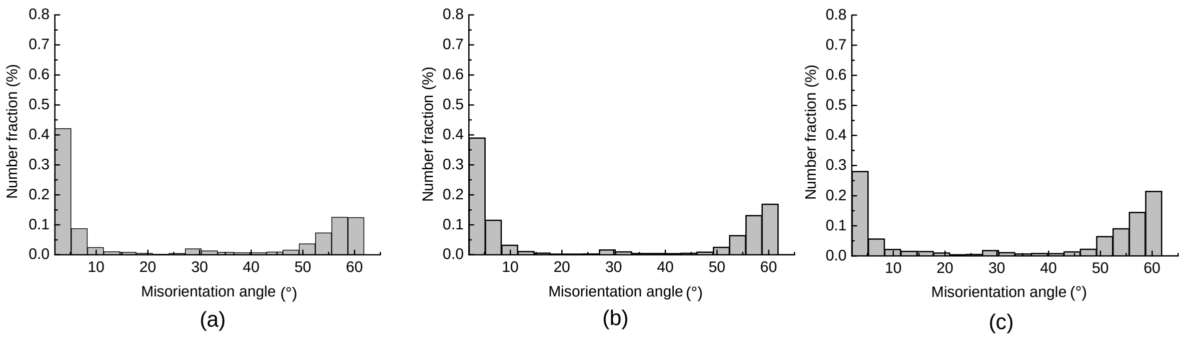

Fig. 7 Distribution of boundary misorientation angles in different manufacture processes. (a) RCR&DQ&T; (b) CR&DQ&T; (c) RQ&T

The relationship between mechanical properties and microstructure is complicated for the tested steel. The main strengthening mechanisms of relevance in this case are grain size strengthening, dislocation hardening, to a lesser extent, solid solution hardening and precipitation strengthening. The yield strength of martensite can be calculated using a number of intrinsic components (Sugden and Bhadeshia, 1988):

$$\sigma_{\mathrm{y}} = \sigma_{\mathrm{i}} + \sigma_{\mathrm{ss}} + \sigma_{\mathrm{p}} + \sigma_{\mathrm{d}} + k_{\mathrm{y}} d^{-1/2}, \quad (1)$$

where σ_{i} is the strength of pure annealed iron, σ_{ss} is the solid solution strength, σ_{p} is the precipitation strength, σ_{d} is the dislocation strength, $k_{\mathrm{y}}d^{-1/2}$ is the grain size strength, and d is the effective grain size, which could be prior austenite grain size, packet size or block width.

Both DQ&T and RQ&T have almost identical solid solution strength and precipitation strength. Thus, dislocation strengthening and grain size strengthening are the major strengthening mechanisms. More recently, Wang *et al.* (2007) reconfirmed that the block width can be regarded as a microstructural factor controlling the strength of the steel. In an attempt to determine the reasons for the strengthening increment associated with DQ&T compared with the conventional RQ&T, yield strength is plotted as a function of the inverse of the square root of the block width in Fig. 8, i.e., a Hall-Petch plot. This plot includes data from previous work (Morito *et al.*, 2006) on RCR&DQ&T, CR&DQ&T, RQ&T in this study. Morito *et al.* (2006) conducted on steels which were reaustenitized at various temperatures and quenched. Two important conclusions are obtained from Fig. 8. First, as well as the block width becoming smaller, dislocation strengthening is the major mechanism to account for the strength increment of DQ steels compared with RQ steels, because the dislocation density in DQ martensite was higher due to the inheritance of deformed substructures of austenite through transformation (Weiss and Thompson, 1992). Second, for DQ steels, the relationship of yield strength and block width basically follows a linear relationship, yield strength increasing with decreasing the finish rolling temperature, because of a finer block width.

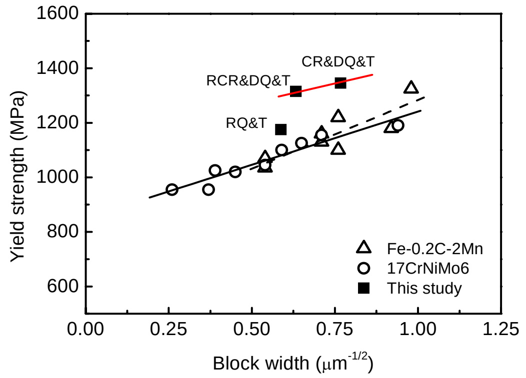

Fig. 8 Dependence of the yield strength on the block width

4 Conclusions

1. DQ martensitic steels, produced with a variety of finish rolling temperatures, exhibit higher strengths than RQ steels. For DQ steels, strength increases with the decrease of the finish rolling temperature.

2. TEM microstructure shows that precipitates of DQ&T steels are a little smaller than those of RQ&T steels, and precipitates size was found to be above 60 nm both in DQ&T and RQ&T steels. Distribution maps of boundary misorientation angles show that DQ&T's number fraction of low angle boundary (misorientation angles less than 15°) is more than that for RQ&T process.

3. Dislocation strengthening is the major mechanism to account for the strength increment of DQ steels compared with RQ steels, because the dislocation density in DQ martensite was higher due to the inheritance of deformed substructures of austenite through transformation. The strength increases with decreasing finish rolling temperature can be explained by differences in block width.

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