Microstructure and mechanical properties of 780 MPa high strength steels produced by direct-quenching and tempering process

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Microstructure and mechanical properties of 780 MPa grade steel plate manufactured by conventional reheat-quenching and tempering (RQ-T) and direct-quenching and tempering (DQ-T) processes were investigated. The DQ process was found to enhance the hardenability of steel effectively so that tensile strengths of a range from 780 to 860 MPa have been achieved using DQ-T process, while tensile strength of about 770 MPa has been obtained from the RQ-T sample. In contrast, low temperature toughness of DQ-T samples was generally inferior to that of RQ-T sample, unless hot rolling and cooling processes were optimized in a controlled manner. For example, fracture appearance transition temperature (FATT) of DQ-T samples was varied in a range from −50°C to −120°C, while RQ-T specimens exhibited nearly constant FATT of about −80°C. The finish-rolling temperature (FRT) was one of potential process parameters to determine strength/toughness balance of the steel manufactured by DQ process, while the effect of FRT was closely associated with the cooling rate applied in the process. It has been demonstrated that, for the specimens quenched with a cooling rate higher than 20°C/sec, it may seem to be appropriate to adjust the FRT as low as possible in the non-recrystallization region. In contrast, for the specimens quenched with a low cooling rate of less than 10°C/sec, it may seem to be proper to apply higher FRT to obtain excellent strength/toughness balance of the steel.

1. Introduction

For some time, the steel industry has been seeking combining processing steps as a means of reducing costs and increasing overall efficiency. For example, some steel makers have eliminated a heat treatment step for rolled plates or bars by installing on-line water-quenching units, thereby permitting the quenching of these products immediately after hot deformation (herein referred to as direct quenching or DQ). In conventional off-line quenched and tempered steels (herein referred to as reheat quenching or RQ), the austenite composition and grain size, which play important roles in determining the ultimate structure and properties, are controlled principally by the austenitizing temperature. In contrast, quenching immediately after hot working allows for some control of austenite structure prior to transformation. For example, generating martensite from deformed austenite by controlled rolling below the austenite recrystallization temperature may provide some of the property improvements traditionally associated with ausforming [1, 2].

The microstructure of DQ steels is strongly dependant on processing variables, such as deformation and cooling conditions [3, 4]. This observation has led researchers to investigate the condition of the austenite grain structure after rolling at various temperatures. However, many of these previous investigations have evaluated effects of the process parameters independently; hence, relationships between austenite structure and mechanical properties are difficult to establish because the transformation behavior of DQ steels is influenced by deformation and cooling conditions simultaneously.

In the present study, a 780 MPa grade 1Ni-0.5Cr-0.5Mo-0.05V-B steel (ASTM A514F Grade) which has been well known to possess a good strength-toughness balance as a structural steel, has been investigated to compare the two different quenching processes and optimize the rolling and cooling conditions for the on-line DQ-T process.

2. Experimental procedures

2.1. Material

The 780 MPa grade steel which was produced by continuous casting at Pohang Iron & Steel Co. Ltd., conforms to grade F within the ASTM A 514 specification, and it's compositions are provided in Table I. The steel melt was refined by various methods to reduce the levels of impurity elements. Sulfur and

TABLE I Chemical compositions of ASTM-A514F steel plate (wt%)

C	Si	Mn	Cu	Ni	Cr	Mo	V	s.Al	Ti	B	N
0.117	0.27	0.79	0.19	0.93	0.43	0.41	0.041	0.045	0.016	0.0015	0.0055

TABLE II Nominal rolling schedules for the direct-quenched samples

Pass no.	Thickness (mm)	Finish rolling temperature (°C) 950	900	850	800	750
1	47 (22)	1145	1145	1120	1100	1100
2	35 (26)	1100	1080	1050	980	950
3	27 (23)	1075	1050	980	950	900
4	21 (22)	1030	980	950	900	850
5	16.5 (21)	980	950	900	850	830
6	13 (15)	950	900	850	800	750
Q	13	920	870	825	780	730

Reduction ratio (%).

phosphorous levels were lowered to 50 ppm and 150 ppm respectively.

2.2. Thermomechanical processing

For the RQ-T process, the ingot was rolled to 13 mm thick plate with an approximately 20% reduction in each rolling pass, and was reaustenitized at 930°C for 1 hr and quenched to room temperature with a cooling rate of 40°C/sec. This plate was tempered at 650°C for 1 hr. For the DQ-T process, a slab of 180 mm thickness was firstly rolled to 60 mm thick plate. These samples were instrumented with mid-thickness thermocouples, reheated to 1150°C, and rolled in six passes to a thickness of 13 mm. During hot rolling, the condition of austenite phase prior to quenching was varied by five different rolling schedules shown in Table II. A conventional hot rolling (finish-rolling is carried out at fully recrystallized temperature) and controlled rolling (some of rolling reduction is carried out at non-recrystallized temperature region) conditions were appled. The rolled steel plates were kept at finish-rolling temperature (herein referred to as FRT) for 20 sec and then direct quenched to room temperature with a cooling rate in the range from 5°C/sec to 40°C/sec. Following quenching, the DQ plate samples were tempered with the same condition applied to RQ samples.

3. Results and discussion

3.1. Comparison of properties between DQ(T) and RQ(T) steel

Tensile properties and Charpy impact toughness of the two specimens were compared in Fig. 1. The tensile strength of DQ specimen was 1211 MPa, which was about 8% higher than that of the RQ specimen showing 1126 MPa. Following tempering at 650°C, the tensile strength of DQ-T specimen softened to 839 MPa, while it was again higher than that of RQ-T specimen showing 773 MPa, Fig. 1a. As well recognized, DQ process could easily enhance effective hardenability in comparison with the case of RQ when normal hot rolling

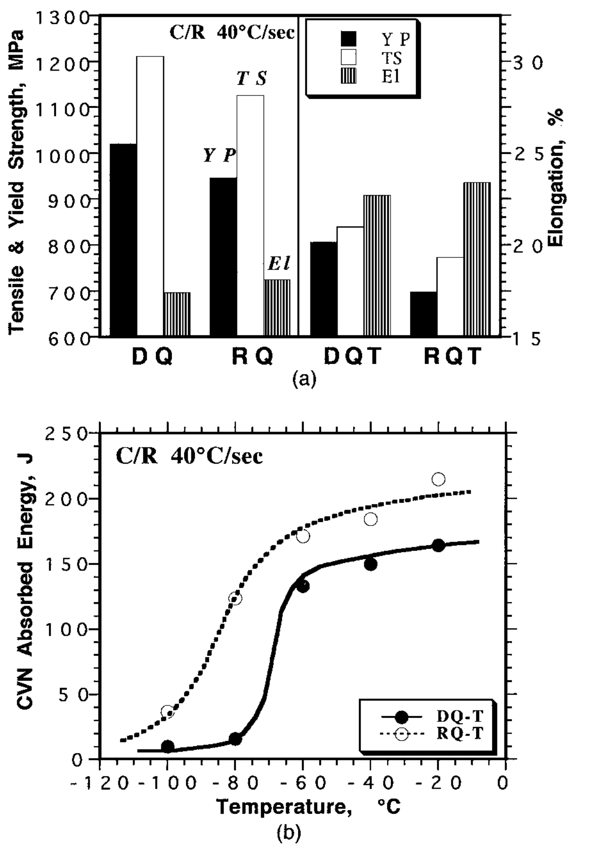

Figure 1 Comparison of (a) tensile and (b) impact properties of DQ(T) and RQ(T) specimens.

conditions are adopted and austenite is recrystallized after hot rolling. This effect is mainly due to higher reheating temperatures in DQ than in RQ, where in DQ both more uniform solution of alloying elements in austenite and coarser austenite grains enhance effective hardenability [5].

In contrast, as illustrated in Fig. 1b, the DQ-T specimen exhibited inferior impact toughness to the RQ-T specimen. Upper shelf energy and fracture appearance transition temperature (FATT) of RQ-T specimen were 215 J and −80°C respectively, while the DQ-T specimen showed 164 J and −70°C. From these observations, it may be deduced that the DQ process may effectively enhance the hardenability of steel, resulting in higher strength than that from conventional RQ process, while, in an attempt to acquire excellent low temperature toughness as well, a variety of process parameters should be carefully optimized during DQ process. In this token, potential effects of process parameters on microstructure and mechanical properties of DQ steels were investigated.

Fig. 2 shows prior austenite grain structures following conventional RQ and DQ process. Both specimens were cooled with 40°C/sec during quenching, whereas the DQ specimen was finish-rolled at 950°C. The RQ steel reveals equiaxed grain morphology due to the complete recrystallization during reheating process at 930°C, while the austenite grains of DQ steel remains as deformed and elongated along the rolling direction.

Fig. 3 illustrates TEM bright field images of RQ-T and DQ-T specimens. In both samples, matrix was composed of parallel martensite laths and the interior of

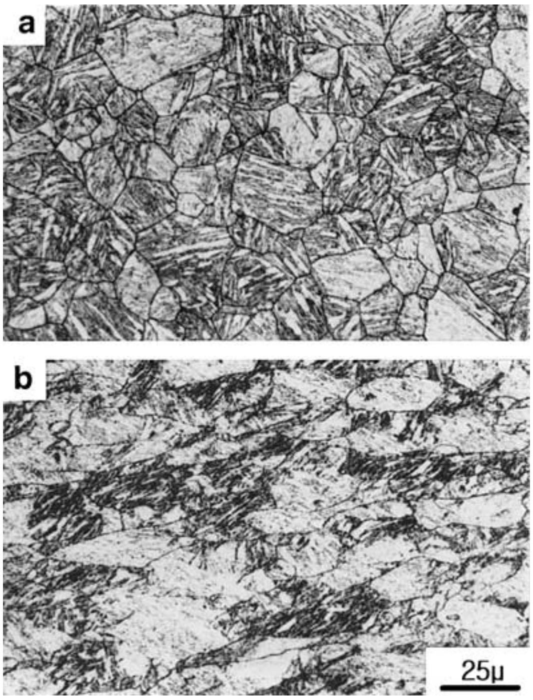

Figure 2 Typical prior austenite grain structures of (a) RQ and (b) DQ specimens.

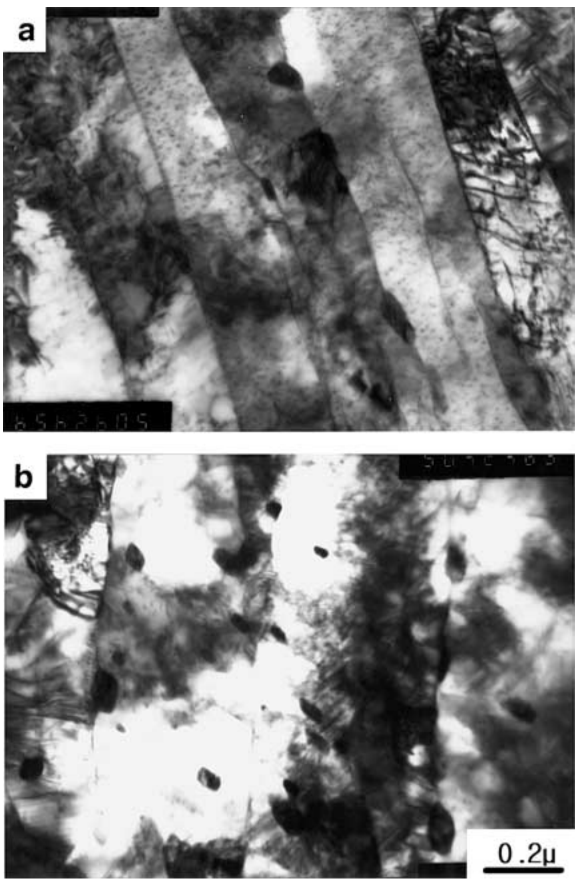

Figure 3 TEM bright field images of (a) DQ-T and (b) RQ-T specimens both tempered 1 hour at 650°C.

lath was heavily dislocated. However, it could be confirmed that the DQ-T specimen had shorter lath spacing and higher dislocation density than the RQ-T specimen, indicating that austenite deformation and direct quenching could induce fine and highly deformed martensite lath structure.

The microstructures of 780 MPa grade QT steels can be characterized by dislocation substructures and precipitates in tempered martensite or bainitic ferrite and by the fineness of microstructures in terms of lath, bundle, packet size or prior austenite grain size. These microstructural features are closely related with mechanical properties of QT steels such that strength is determined by dislocation substructures and precipitates in ferrite, while low temperature toughness in terms of FATT is dependent on the effective grain size. By adopting DQ process, the former microstructural features can be easily achieved, however, the latter microstructural feature are more sensitively influenced by rolling and cooling conditions. In an attempt to get excellent strength/toughness balance of DQ steels, the effect of process parameters on mechanical properties should be clearly understood.

3.2. Effect of cooling rate on microstructure and properties of DQ steel

To investigate the relationship between microstructure and strength/toughness of DQ and DQ-T specimens, the specimens were finish rolled at 950°C and quenched with various cooling rates within the range between 5°C/sec and 40°C/sec. The rolling conditions are typical hot rolling (HR) conditions of steel plates. Fig. 4a shows the variation of tensile properties

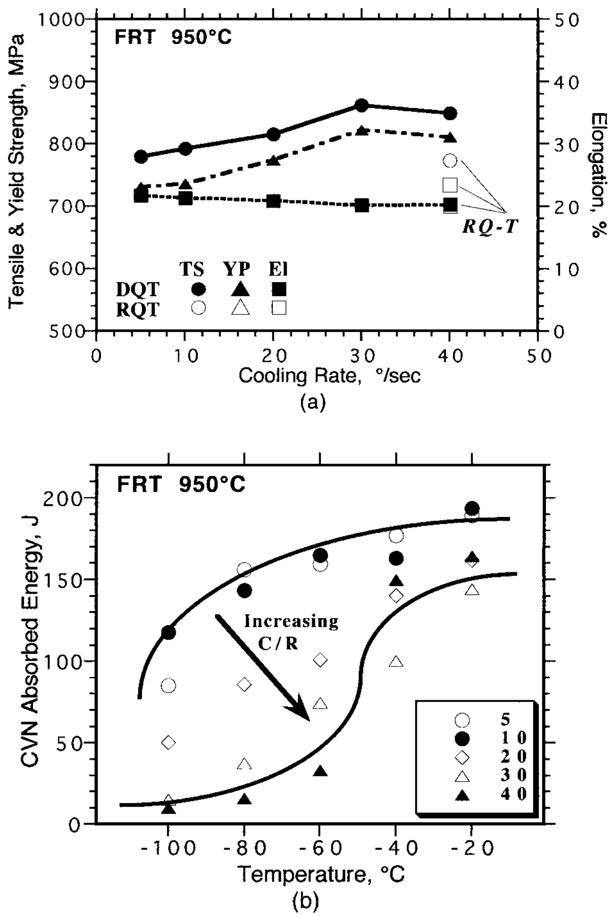

Figure 4 Variations in (a) tensile and (b) impact properties of DQ-T specimens with cooling rate (FRT 950°C).

(TS, YP, El) of quenched and tempered specimens with increasing cooling rate. Over the cooling rate range examined, tensile and yield strengths increased monotonically with a increase in cooling rate. However, the increasing rate appeared to be saturated with the cooling rate being over 30°C/sec. The HR + DQ specimen quenched with the same rate (40°C/sec) as applied to RQ specimen revealed much higher tensile and yield strengths than those of RQ specimen. With the slowest cooling rate (5°C/sec), the HR + DQ specimen could achieve a strength level equivalent to the RQ specimen.

Fig. 4b illustrates the variation of impact toughness of quenched and tempered specimens with increasing cooling rate. For the range of cooling conditions employed, the variation in impact toughness with cooling rate was significant. The upper shelf energy and FATT of the specimen quenched with 5°C/sec were 189 J and −100°C respectively, while those of the specimen quenched with 40°C/sec were 164 J and −50°C respectively, i.e. the tempered specimens quenched with less than 10°C/sec revealed higher upper shelf energy and lower FATT compared to specimens quenched with more than 20°C/sec, indicating that the impact toughness of DQ-T specimens was decreased with increasing cooling rate. It thus seems likely that impact toughness of DQ-T specimens may be strongly dependent upon cooling conditions during quenching after hot rolling.

To investigate the change in mechanical properties of DQ specimens associated with cooling conditions, microstructural variations in HR + DQ and CR + DQ specimens with cooling rate were observed using optical microscopy. Fig. 5 includes optical micrographs showing the variation of microstructures with cooling rate. All the specimens finish rolled at 950°C revealed austenite grain structures heavily deformed and elongated along the rolling direction. While the specimen quenched with 40°C/sec showed typical martensitic lath structure, whereas the specimens quenched with lower cooling rates of 5–10°C/sec revealed the bainite-dominant microstructures. These microstructural features were confirmed by transmission electron microscopy.

Fig. 6 illustrate (a) tensile properties and (b) impact toughness of DQ-T specimens as a function of cooling rate for the specimens finish rolled at 750°C. For the range of cooling conditions employed, tensile and yield strengths of DQ and DQ-T specimens were again linearly increased with increasing cooling rate, Fig. 6a. However, most notably, the impact toughness of DQ-T specimens appeared to be improved with increasing cooling rate during direct quenching, implying the contrary result to the previous HR + DQ specimens finish rolled at 950°C.

Based on the combination of above two contradictory results, it is apparent that the effect of cooling rate during quenching is closely related with the austenite deformation conditions. To investigate the correlation between cooling and deformation conditions in further detail, the specimens deformed in different ways and quenched with the same cooling rate.

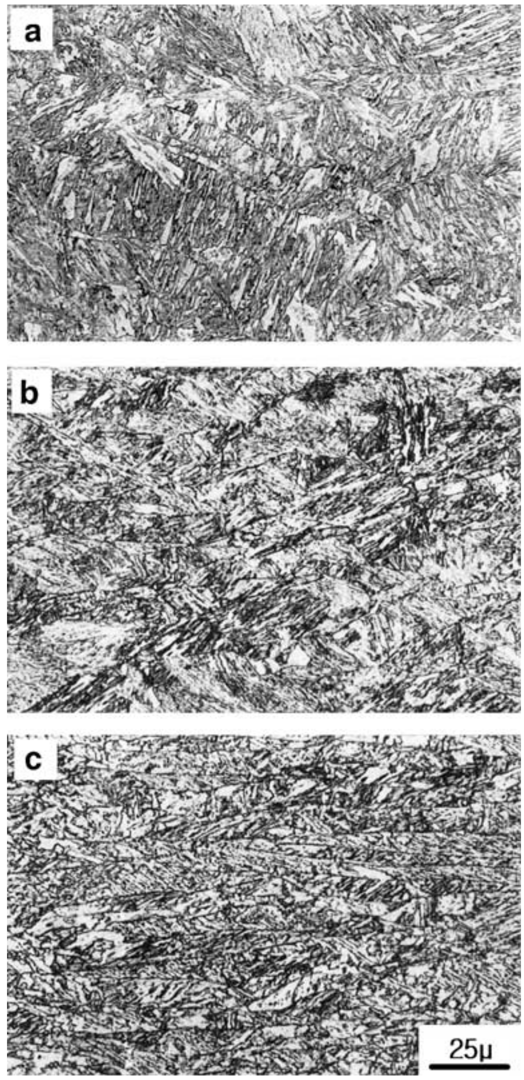

Figure 5 Variations in direct-quenched microstructures with cooling rate of (a) 40°C/sec, (b) 10°C/sec and (c) 5°C/sec.

3.3. Effect of finish rolling temperature on strength and toughness

Fig. 7a shows tensile properties of DQ and DQ-T specimens as a function of finish rolling temperature. The specimens were quenched with 30°C/sec and tempered 1hr at 650°C. The variation of FRT appeared to be effective on the tensile properties of DQ-T specimens. The tensile strength increased from 820 MPa for the specimen finish rolled at 750°C to 847 MPa by increasing FRT to 800°C, while the tensile strength was linearly decreased with further increasing FRT to 950°C. Even though the specimen direct quenched with relatively high cooling rate of 30°C/sec, the specimen finish rolled at 950°C showed the tensile strength of 758 MPa which was below the requirement of ASTM A514F specification (≥760 MPa).

Fig. 7b illustrates the variation in impact toughness with FRT for the DQ-T specimens. The impact toughness was deteriorated with increasing FRT. Following tempering 1 hr at 650°C, all the specimens showed similar upper shelf energy in the range of 145–168 J, while the FATT was remarkably changed with the FRT such that the specimens finish-rolled at 900–950°C transited to brittle fracture mode at approximately −50°C, while

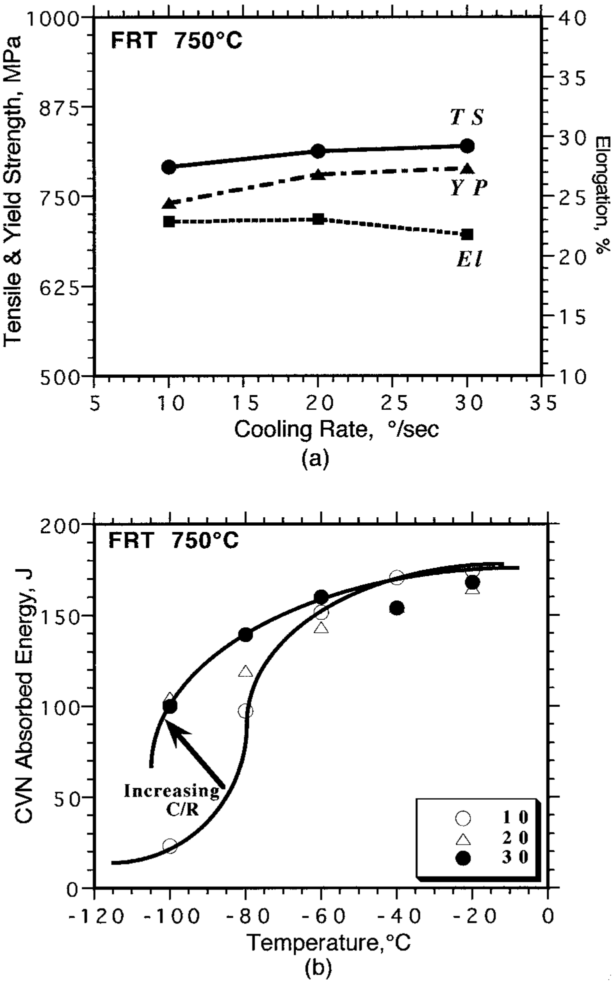

Figure 6 Variations in (a) tensile and (b) impact properties of DQ-T specimens with cooling rate (FRT 750°C).

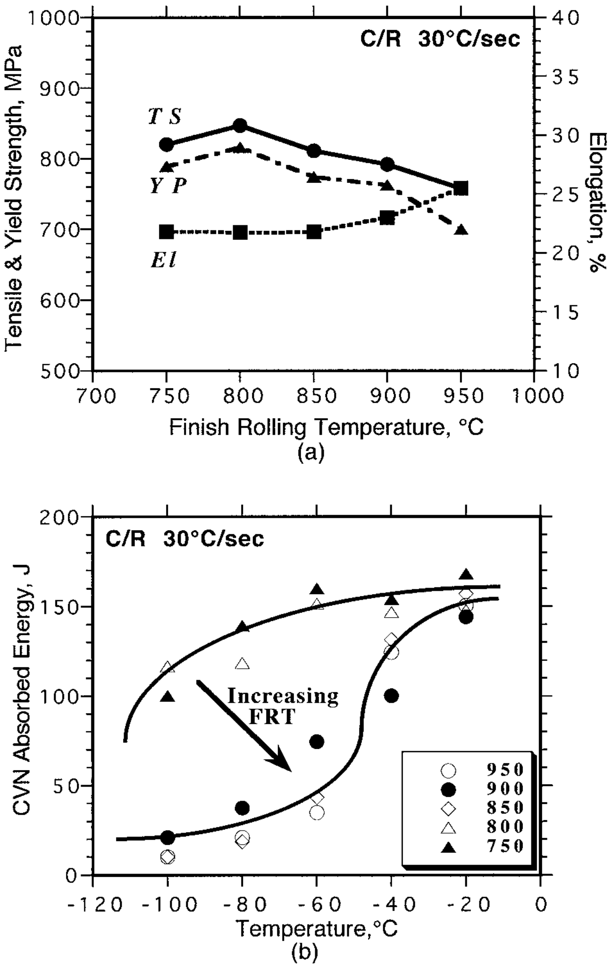

Figure 7 Variations in (a) tensile and (b) impact properties of DQ-T specimens with FRT (C/R 30°C/sec).

the specimens finish-rolled at 750–800°C preserved ductile fracture mode up to the lowest test temperature of −100°C. From the specimens quenched with 20°C/sec, very similar correlations between FRT and mechanical properties have been observed.

Based on the combination of tensile and impact test results, it is apparent that the FRT is a potential process parameter to determine strength and toughness balance of the specimens manufactured by DQ process. It has been demonstrated that, using the present alloy composition and direct quenching with a cooling rate higher than 20°C/sec, it is possible to generate and effective strength and toughness balance by adopting intensive rolling in the non-recrystallization region. Therefore, for the specimens quenched with a cooling rate higher than 20°C/sec, it may seem to be more effective to apply CR + DQ process instead of HR + DQ process for pursuing excellent strength and toughness balance.

To confirm the effect of FRT on the mechanical properties of DQ specimens, austenite grain structures and microstructures were investigated. For the specimens subjected to direct quenching with 30°C/sec under a range of different finish-rolling temperatures, the resulting austenite grain structures have a similar appearance showing elongated or pancaked morphology, Fig. 8. However, as the FRT decreased from 950°C to 750°C, the average thickness of pancaked austenite grains was significantly reduced, and the as-quenched microstructure was changed from fully martensitic structure to martensite/lower bainite duplex structure with lowering FRT.

Fig. 9 shows TEM bright field images for the specimens finish rolled at (a) 950°C and (b) 750°C respectively. A coarse martensite packet composed of a number of parallel arrays of laths can be seen from the specimen finish-rolled at high FRT (≥900°C). In this case, the length of lath was approximately consistent with the austenite grain size. In the specimen finish-rolled at low FRT (≤800°C), on the other hand, many small martensite or bainite packets exist, which consist of small numbers of laths refined by the heavy deformation in the non-recrystallization region. Furthermore, it was found that the number of laths in a packet remarkably decreased by heavy deformation of austenite, and this caused the decrease in the packet size.

A similar attempt changing FRT from 950°C to 750°C was applied to the specimens quenched with relatively slow cooling rate of 10°C/sec. Fig. 10a shows tensile properties as a function of FRT for the DQ-T specimens. For the range of FRT employed, the variation in tensile properties of DQ-T specimens with FRT was negligible. The tensile strength of all the specimens was in the range 775–809 MPa. Fig. 10b illustrates the variation in impact toughness with FRT for DQ-T specimens quenched with 10°C/sec. A particularly interesting observation to emerge from these specimens

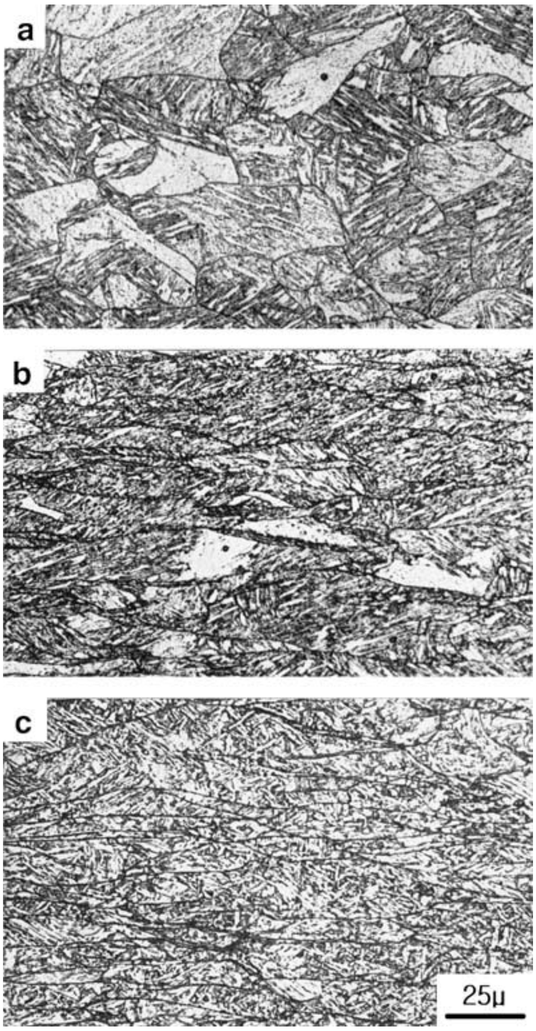

Figure 8 Optical micrographs showing variations in direct-quenched microstructures with FRT of (a) 950°C, (b) 850°C and (c) 750°C (cooling rate: 30°C/sec).

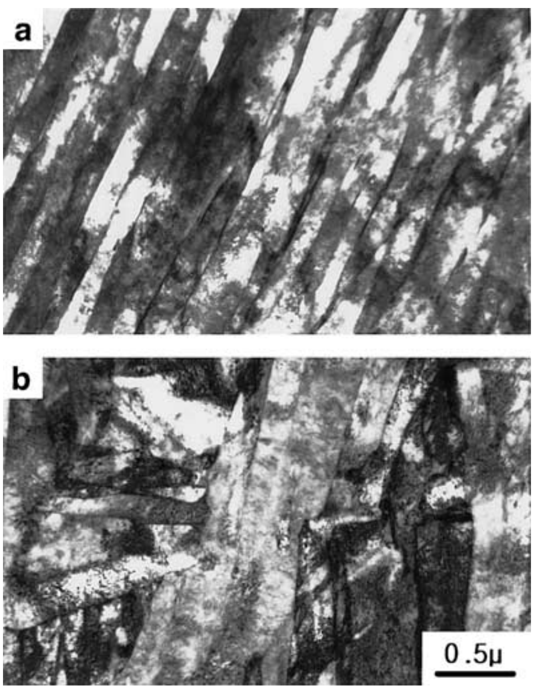

Figure 9 TEM bright field images of direct quenched specimens finish-rolled at (a) 900°C and (b) 800°C and quenched with cooling rate of 30°C/sec.

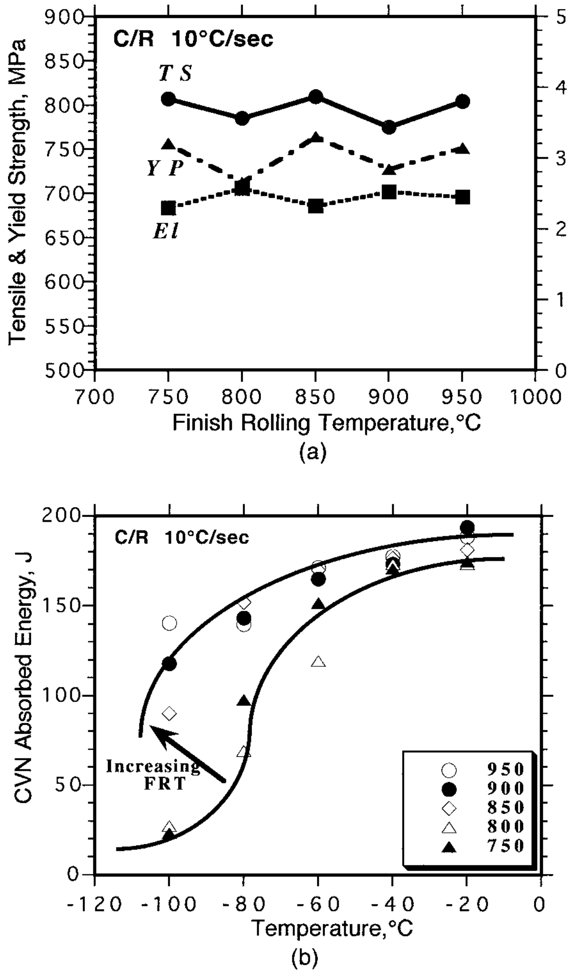

Figure 10 Variations in (a) tensile and (b) impact properties of DQ-T specimens with FRT (C/R 10°C/sec).

quenched with 10°C/sec was an opposite effect of FRT on impact toughness. In both conditions, the impact toughness was improved with increasing FRT, which is contrary to the results obtained from the specimens quenched with 30°C/sec. Following tempering 1hr at 650°C, all the specimens showed similar upper shelf energy in the range of 173–194 J, while the FATT was changed with the FRT such that the specimens with FRT 750–800°C transited to brittle fracture mode at approximately −70°C, while the specimens with FRT 850–950°C preserved ductile fracture mode up to the lowest test temperature of −100°C.

Based on the combination of tensile and impact test results, it is apparent that the FRT is a potential process parameter to determine strength and toughness balance of the specimens manufactured by DQ process. It has been demonstrated that, using the present alloy composition and direct quenching with a cooling rate higher than 20°C/sec, it is possible to generate an effective strength and toughness balance by adopting intensive rolling in the non-recrystallization region. Therefore, for the specimens quenched with a cooling rate higher than about 20°C/sec, it may seem to be appropriate to control the FRT as low as possible in the non-recrystallization region. In contrast, for the specimens quenched with relatively low cooling rate lower than 10°C/sec, it is again apparent that the FRT is a potential process parameter in DQ process. however, it

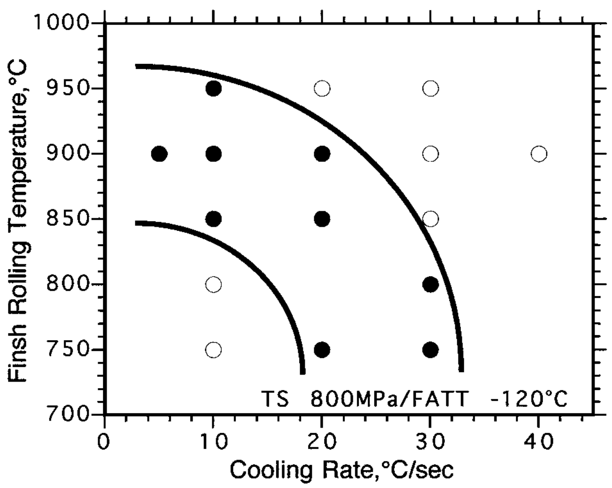

Figure 11 Correlation between cooling rate and FRT showing a region with good strength/toughness balance for the DQ-T steel.

may seem to be appropriate to control the FRT as high as possible.

The interrelationships between rolling and cooling conditions in DQ process was summarized by defining a region with excellent strength/toughness balance (closed circles) as shown in Fig. 11. These plots include data from specimens of all processing conditions which have been tempered at 650°C. This plot clearly shows grouping of data based on processing history, where the strength/toughness balance is best for the higher cooling rate + lower FRT or lower cooling rate + higher FRT conditions.

For the DQ specimens with higher cooling rate, CR + DQ process induces ausforming effect. The ausforming effect was reported as early as 1960 [6, 7]. The effect which can enhance both strength and toughness was explained by reason that the transformed martensite from deformed austenite was refined and the dislocation density in martensite was higher due to the inheritance of deformed substructures of austenite through transformation. After tempering for ausformed steel, carbide precipitates were distributed homogeneously in the structures due to increased precipitation sites while they precipitated preferentially on the various boundaries in RQ steel resulting in deteriorated toughness. More recently, Maki [8] reconfirmed the effect that cells of deformed austenite were inherited by martensite and that, although the packet size of martensite became larger, block width, which controlled toughness, became smaller.

Apart from the strengthening effect of direct quenching after controlled rolling (CR + DQ), a large research effort has been concentrated in optimization of the CR + DQ processing parameters in order to produce materials of high strength without deterioration of toughness. A number of studies have found that the increase of strength by direct quenching need not necessarily be accompanied by a detrimental effect on toughness. It is particularly noteworthy that the promising improvements in the combination of strength and toughness obtained in these studies are always related to microstructural refinement, which is well consistent with the present results showing enhanced toughness by adopting duplex microstructure of martensite and bainite.

Low temperature toughness is mainly dependent upon an effective grain size of the microstructures, so that austenite grain size prior to quenching is directly related with the transformed microstructural features, and hence to low temperature toughness. In the present CR + DQ specimens with higher cooling rates, austenite grains subdivided by partial bainitic transformation prior to martensite formation, which led to effectively finer overall microstructures beneficial to low temperature toughness. In contrast, for the specimens quenched with a lower cooling rate, deformation of austenite in lower FRT conditions tends to promote the ferrite transformation to result in decreased effective hardenability.

4. Conclusions

The DQ process was found to enhance the hardenability of steel effectively, resulting in higher strengths than RQ samples of the same chemistry, while the low temperature impact toughness of DQ-T samples were generally inferior to that of RQ-T sample, suggesting that potential process parameters should be optimized to obtain a good strength/toughness balance.

The finish rolling temperature (FRT) was found to be a potential process parameter in DQ process, while the effect of FRT was closely associated with the cooling rate applied in the process. For the specimens quenched with a cooling rate higher than 20°C/sec, it may seem to be appropriate to adjust the FRT as low as possible in the non-recrystallization region. In contrast, for the specimens quenched with a relatively low cooling rate of less than 10°C/sec, it may seem to be proper to adopt conventional hot rolling with higher FRT for the good strength and toughness balance of the steel.

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