The Microstructure and Toughness of High Tensile Strength Steels*

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Synopsis

The impact properties of some high tensile strength steels were investigated in relation to the cooling rate on continuous cooling (after austenitization) and the austenite grain size.

Results showed the existence of the optimum cooling rate for the impact transition temperature. The optimum cooling rate varied with the hardenability of a steel, and as the austenitizing temperature was raised, it shifted to the slower cooling rate.

The electron microscopic observation revealed that the structure with superior impact properties was the duplex martensite-bainite structure. The morphology of carbide in this bainite was similar to that in the lower bainite of a higher carbon steel, but the trace analysis indicated that it had the same lath-like morphology with a $\langle 111 \rangle_{\alpha} \{110\}_{\alpha}$ as that in the upper bainite and the lath martensite. (This bainite in a low-carbon low-alloy steel was designated as Bainite III type in a previous paper.)

The orientation measurements of a facet pit and a fracture surface clarified that the cleavage plane was $\{100\}$. On the basis of the direct observation of fracture surface by the scanning electron microscope, the unit crack path was defined. There was experimentally better correlation between the unit crack path and the transition temperature, that is, the finer the unit crack path, the lower the transition temperature is.

The duplex martensite-bainite III structure had the finest crack path. These results suggest that the role of bainite III in the duplex structure was the partitioning of austenite grain prior to the martensite transformation.

I. Introduction

There have been numerous studies on the toughness of steels in relation to microstructures, and it has been well-known that the low-temperature toughness of a steel increases by refining the austenite grain size and transformation product.¹⁻¹⁵⁾ Consequently, it is believed that the tempered martensite which was produced from the fine grained austenite has the most superior mechanical properties (the combination of strength and toughness). Although the quenched and tempered high tensile strength steels have been widely used,¹⁶⁾ the factors controlling the toughness of quenched and tempered steel have not been clarified. Recently it was reported that the dependencies of toughness on the austenite grain size were different with the steels.⁵⁾ Also, it might be noted that the fully hardened structure (100% martensite) could be scarcely obtained in a low-carbon low-alloy steel,¹⁷⁾ and the microstructure has an important role in the toughness.

In the present experiments, the impact properties of various kinds of microstructures obtained during continuous cooling for some low-carbon low-alloy high tensile strength steels were determined in detail. As the result, a characteristic phenomenon was obtained; the duplex martensite-bainite (Bainite III type) structure formed with a little slower cooling rates than the upper critical cooling rate had the most superior impact properties.

The aims of the present study are to investigate the morphology of bainite in the duplex structure and to reveal the structural factors contributing to the toughness by the observation of the impact fracture surface in relation to microstructure with the scanning electron microscope.

II. Experimental Procedure

1. Materials

The chemical compositions of the steels used are shown in Table 1. The steel *N* is a 60 kg/mm² grade quenched and tempered high strength steel containing small amount of nickel, molybdenum, and vanadium. The steels 7 and 8 are typical of 80 kg/mm² grade quenched and tempered high strength steels containing copper, nickel, chromium, molybdenum, and vanadium, the small amount of boron being added to the steel 8. They were hot rolled to the plate 7 mm in thickness with a constant rolling condition.

2. Heat-treatments

The heating temperatures from 900° to 1 200°C were chosen to vary the austenite grain size. Martensite, bainite, and ferrite-pearlite structures were obtained with various cooling rates on continuous cooling (mean cooling rates between 800° and 500°C were from 200°C/sec to 0.0075°C/sec). The mechanical properties were determined on these microstructures, both as continuously cooled and as tempered. The tensile properties were tested using 4×6×30 mm gauge length specimen. The impact test was made using 5×10×55 mm subsize Charpy specimen with

Table 1. Chemical compositions of steels used

Steel	C	Si	Mn	P	S	Cu	Ni	Cr	Mo	V	B	Ti	sol Al
<i>N</i>	0.15	0.34	1.25	0.027	0.017	0.07	0.19	0.02	0.13	0.03	—	—	0.032
7	0.12	0.30	0.83	0.004	0.005	0.30	1.11	0.53	0.49	0.03	—	—	0.038
8	0.14	0.26	0.89	0.004	0.005	0.27	1.04	0.54	0.50	0.03	0.004	0.03	0.044

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2 mm V-notch.

3. Microstructural Observations

The optical microscopic observation for these specimens was carried out after etching with 2% nital.

The thin foils were used for electron microscopy. For the thin foil preparation, the bulk specimens of 1 mm in thickness was polished down to 0.1 mm chemically. The final electro-polishing was carried out in the chromic acid/phosphoric acid electrolyte. The foils were examined in the HU-200 operated at 200 kV. The habit plane and growth direction of martensite and bainite were determined by the trace analysis.

4. C.C.T. Diagrams

The decomposition of austenite on continuous cooling was examined by means of dilatometry measurements using a Formster (the specimen was 3 mm in diameter and 10 mm in length) and Leitz dilatometer (the specimen was 3.8 mm in diameter and 50 mm in length). They were continuously cooled after austenitized at 900° to 1 200°C. The cooling and temperature/dilatation curves were recorded simultaneously. The C.C.T. curves were determined from the dilatation curves, microscopic observations and hardness measurements.

5. Observations of Fracture Surface

As for the impact fracture surface, the aspect of fracture surface was examined with a scanning electron microscope, and the relation between the surface aspect and the microstructure was shown simultaneously. The nickel coating on the fracture surface was made, then the section which was parallel to the direction of the crack propagation and orthogonal to the fracture surface was polished. It was etched with 2% nital and facet pits on the section were presented using the following reagents;¹⁸⁾ (A) HCl 2 cc + H₂O₂ 10 cc + H₂O 200 cc (B) HCl 2 cc + Sat. FeCl₃ 6H₂O 10 cc + H₂O 200 cc (C) CH₃COOH 100 cc + H₂O 100 cc.

After removing the nickel coating, the fracture surface and the corresponding section were examined with the JSM-2 type scanning electron microscope. The orientation of the cleavage plane of the brittle fracture surface was determined by the orientation measurements of the facet pit and the fracture surface using a gonio-microscope.

III. Experimental Results

1. Effect of Cooling Rate on Mechanical Properties

Figures 1, 2, and 3 show the relations of the cooling rate, which is expressed in cooling time from 800° to 500°C, to the fracture transition temperature in the impact test (vT_s) and tensile strength for steels *N*, 7, and 8, respectively.

In steel *N*, austenitizing at 950° and 1 000°C, the faster the cooling rate, the lower the vT_s is. The vT_s did not vary monotonously with cooling rate after austenitizing at 1 200°C, and there was the optimum cooling rate for vT_s . As the austenitizing temperature

was raised, the hardenability increased. It suggested that even though the cooling rates were same, the microstructures were different in the different austenitizing temperatures.

On the other hand, in steel 7, the optimum point for vT_s was observed even at the low austenitizing temperature and it shifted to the lower cooling rate as the austenitizing temperature was raised.

In steel 8, containing boron, the optimum point for vT_s was observed clearly.

These results that the optimum point varied with the chemical composition and austenitizing temperature suggested strongly that the optimum point depended on the hardenability of a steel.

Fig. 1. Effects of cooling rate and heating temperature on mechanical properties of Ni-Mo-V steel (steel *N*)

Fig. 2. Effects of cooling rate and heating temperature on mechanical properties of Cu-Ni-Cr-Mo-V steel (steel 7)

The strength and toughness after tempering at 650°C for 1 hr are shown also in Figs. 1, 2, and 3. In tempering, the strength decreased, but the vTs was improved markedly, especially in faster cooling. The

Fig. 3. Effects of cooling rate and heating temperature on mechanical properties of Cu-Ni-Cr-Mo-V-B steel (steel 8)

tendencies in the change of vTs with cooling rate after tempering were similar to those as-cooled on continuous cooling. Then it was suggested that the impact properties were largely dependent on the initial microstructure before tempering.

2. Microstructural Observations

Photographs 1, 2, and 3 show the typical microstructural appearance of as continuously cooled specimens at various cooling rates in the steels *N*, 7, and 8, respectively.

In the cooling range experimented, martensite, martensite-bainite, and bainite of carbide free bainitic ferrite,¹⁹⁾ were produced with reducing the cooling rate. Further reducing the cooling rate, ferrite-pearlite in steel *N* and ferrite-bainite in steel 7 were obtained, but the critical cooling rate for beginning of ferrite formation in steel 8 was very small.

It should be emphasized that the superior toughness was obtained in the duplex martensite-bainite structure. The bainite in such a duplex structure could be scarcely distinguished from the martensite under an optical microscope. Careful observation, however, indicated that the region with the similar orientation

Photo. 1. Optical microstructures of Ni-Mo-V steel (steel *N*) cooled from 1200°C with

- (a) 200°C/sec (T.S.=132.9 kg/mm², vTs =-6°C)
- (b) 75°C/sec (T.S.=129.9 kg/mm², vTs =-66°C)
- (c) 14°C/sec (T.S.=76.5 kg/mm², vTs =16°C)
- (d) 0.0075°C/sec (T.S.=50.7 kg/mm², vTs =32°C)
- (a)-(d) (×500) (×3/4)

Photo. 2. Optical microstructures of Cu-Ni-Cr-Mo-V steel (steel 7) cooled from 1200°C with

- (a) 200°C/sec (T.S.=121.0 kg/mm², vTs =12°C)
- (b) 23°C/sec (T.S.=120.3 kg/mm², vTs =-54°C)
- (c) 3.4°C/sec (T.S.=76.7 kg/mm², vTs =32°C)
- (d) 0.0075°C/sec (T.S.=67.5 kg/mm², vTs =92°C)
- (a)-(d) (×500) (×3/4)

of martensite laths in the former austenite grain was large for the fully martensitic structure, but the austenite was finely divided in the duplex structure. The toughness became much inferior in the bainite with bainitic ferrite transformed on further reducing the cooling rate.

Photograph 4 shows a typical low carbon lath martensite. Photograph 5 shows the bainite appearing in the du-

plex martensite-bainite structures which had the superior impact properties. Here (a) and (b) are the bright field image and the dark field image using a cementite reflection, respectively. The morphology of cementite precipitation is very similar to that in the case of lower bainite in higher carbon steels. Photograph 6 shows the bainitic ferrite without carbide precipitation.

Photo. 3. Optical microstructures of Cu-Ni-Cr-Mo-V-B steel (steel 8) cooled from 1 200°C with
 (a) 200°C/sec (T.S.=132.2 kg/mm², $vTs=34^{\circ}\text{C}$)
 (b) 23°C/sec (T.S.=124.6 kg/mm², $vTs=-6^{\circ}\text{C}$)
 (c) 3.4°C/sec (T.S.=83.8 kg/mm², $vTs=72^{\circ}\text{C}$)
 (d) 0.0075°C/sec (T.S.=75.1 kg/mm², $vTs=92^{\circ}\text{C}$)
 (a)-(d) ($\times 500$) ($\times 3/4$)

Photo. 4. Martensite formed in Ni-Mo-V steel (steel N) cooled from 1 200°C with 200°C/sec

Photo. 5. Bainite formed in Ni-Mo-V steel (steel N) during continuous cooling from 1 200°C with 75°C/sec
 (a) Bright field image
 (b) Dark field image using a cementite reflection
 (a) and (b) ($\times 20\,000$) ($\times 2/3$)

Photo. 6. Bainitic ferrite formed in Ni-Mo-V steel (steel N) cooled from 1 200°C with 14°C/sec

The trace analysis for such martensite, bainite, and bainitic ferrite indicated the $\langle 111 \rangle_\alpha$ growth direction and $\{110\}_\alpha$ habit plane.¹⁹⁾ Consequently, the bainites described above could be considered as the upper bainite concerning the morphology of ferrite.²⁰⁾ In the previous paper,¹⁹⁾ the bainites in low-carbon low-alloy steels were classified to three types; Bainite *I* type (a carbide free bainitic ferrite), Bainite *II* type (cementite precipitates mainly along the ferrite lath interface) and Bainite *III* type (cementite precipitates within a ferrite grain and the morphology of cementite was similar to that in lower bainite). Photographs 5 and 6 show the Bainite *III* and Bainite *I*, respectively.

3. Transformation

Figures 4, 5, and 6 show the C.C.T. diagrams after austenitizing at 950°C for steels *N*, 7, and 8, respectively. As the austenitizing temperature was

Fig. 4. C.C.T. diagram for Ni-Mo-V steel (steel *N*)

Fig. 5. C.C.T. diagram for Cu-Ni-Cr-Mo-V steel (steel 7)

raised, the critical cooling rate for the beginning of bainite formation (C_B) generally shifted to the slower cooling rate.

Figure 7 indicates that C_0 is slower than C_B but faster than the critical cooling rate for the beginning of bainitic ferrite formation (C_{BF}). The structures obtained at the optimum cooling rate for vT_s included martensite and bainite. The bainite obtained at the cooling rate C_0 was Bainite *III* type as shown in Photo. 5. C_{BF} nearly corresponded to the cooling rates at which hardness of 300 H_V could be obtained, and the vT_s already began to rise at this cooling rates. To obtain the duplex martensite-bainite structure (predominant

Fig. 6. C.C.T. diagram for Cu-Ni-Cr-Mo-V-B steel (steel 8)

Fig. 7. Relation of optimum cooling rate for impact transition temperature (C_0), critical cooling rate for beginning of bainitic ferrite (C_{BF}), and critical cooling rate for beginning of bainite (C_B). (C_0 , C_{BF} , and C_B are expressed in cooling time from 800° to 500°C.)

amount of martensite) which had the superior impact properties, the bainite was Bainite *III* type on the continuous cooling (Figs. 4, 5, and 6). In the C.C.T. diagrams (Figs. 4, 5, and 6), the regions of Bainite *I*, *II*, and *III* and the approximate range of optimum cooling rate are shown.

4. Effect of Austenite Grain Size

It has been well known that the finer the austenite grain size (d_T), the more superior the impact properties are.¹⁻⁸⁾ Figures 8 and 9 show the relations between $\log d_T^{-1/2}$ and vTs . In these figures, the cooling rate had been employed as a parameter, but even though the

Fig. 8. Relation between austenite grain size and transition temperature for Ni-Mo-V steel (steel *N*)

Fig. 9. Relation between austenite grain size and transition temperature for Cu-Ni-Cr-Mo-V steel (steel 7)

cooling rates were same, the microstructures with the different austenite grain sizes were not always the same. The microstructures which had the linear correlation to the austenite grain size were the martensite, the bainite formed at high temperature, and the ferrite-pearlite. The dependency of vTs on the austenite grain size for these structures was about 15°C for every ASTM grain size number, as has been reported previously.⁴⁾ Recently the detailed investigation on the relation of microstructure and fracture characteristics of low carbon tempered martensite was carried out by the independent observations of the microstructure and the fractured surface, and the effective grain size was proposed.²¹⁾ From the direct observation of fracture surface and microstructure in ferrite-pearlite, it was noted that the fracture unit was composed of several ferrite grains.²²⁾ The minimum fracture unit of pearlite in high carbon steel was observed to be a pearlite colony.^{23,24)} It can be considered that since there may be good correlation between the fracture unit and austenite grain size, vTs is closely dependent on austenite grain size.

In the duplex martensite-bainite structure the dependency of vTs on the austenite grain size was small. In steel *N* with a small hardenability, at the same cooling rate, the microstructures were varied markedly by the austenitizing temperature. In steel 7, the dependency of vTs on the austenite grain size in duplex structure was also smaller than that of martensite. This suggests that there is a case (cooling rate) where an increased austenite grain size results in improving vTs .

5. Scanning Electron Microscopy of Fracture Surface

Photographs 7, 8, and 9 show the direct correspondence between the Charpy impact fracture surfaces and microstructures of as continuously cooled specimens after austenitized at 1 200°C in steel *N* (corresponding to Photos. 4, 5, and 6, respectively). The microscopies were performed with (a) fracture surface, (b) 45° inclination to fracture surface, (c) 90° inclination to fracture surface (cross section to the fracture surface). All were typical of quasi-cleavage surface with finely rugged patterns like river-mark and discontinuous aspects. Photograph 7 shows the fractography of martensite. The shapes of facet pits in the fracture unit appeared same, indicating that the unit was consisted of laths with almost the same crystallographic orientation. Photographs 8 and 9 show the fractographies of duplex martensite-B *III* bainite and B *I* type bainite respectively. The fracture unit in Photo. 8 was smaller than that in Photo. 7, which in turn was smaller than that in Photo. 9. The fracture planes {112} and {123} besides {100} were reported in a 0.3% carbon martensite,²⁵⁾ but the cleavage plane determined in this study by the orientation measurements of facet pits and corresponding fracture surface with a goni-microscope represented {100}. The results are illustrated in Fig. 10, which are the same with those obtained in the single crystal of pure iron.²⁶⁾

On the basis of the direct observation of fracture

Photo. 7. Fractographs of martensite in Ni-Mo-V steel (steel *N*) broken at -60°C ($\times 800$) ($\times 4/5$)

Photo. 8. Fractographs of duplex martensite-bainite structure in Ni-Mo-V steel (steel *N*) broken at -100°C ($\times 800$) ($\times 4/5$)

Photo. 9. Fractographs of bainite in Ni-Mo-V steel (steel *N*) broken at -60°C ($\times 800$) ($\times 4/5$)

Fig. 10. Stereographic analysis of facet pit and cleavage plane

characteristics in detail, 'the region consisting of $\{100\}$ fracture planes of which orientation differences are within a small angle' was taken as a fracture unit.²²⁾ The fracture initiated within the region at the neighbourhood of boundary of this region and propagated radially to the boundary of adjacent fracture unit. The aspects of boundary which formed the fracture unit were discussed in another paper,²²⁾ and it was certified

that the unit did not change with the test temperature. Under these definition, the unit crack path (d_c) of brittle fracture surface was measured, and correlated with vT_s .

In Fig. 11, vT_s was plotted against $\log d_c^{-1/2}$ for the continuously cooled steel 7. There was experimentally more superior correlation in the unit crack path to the transition temperature than that of the austenite grain size to the transition temperature. It was seen that there were two groups characterized by tensile strength level. It is interesting to note that conventionally 50% martensite has been used as a criterion of hardening for the evaluation of a hardenability of steel²⁷⁾ and the strength with 50% martensite was about 100 kg/mm² for this steel. In the case that the strength level was less than 100 kg/mm², the microstructures were consisted of B I type bainite or ferrite-bainite, and the river patterns were clear and the fracture units appeared comparatively block-like, as was shown in Photo. 10. On the other hand, in the case that the strength level was more than 100 kg/mm², the microstructures were consisted of martensite or martensite-B III type bainite. The fractographs revealed that the river patterns were not linear-like and the fine microstructure might have the resistance against propagation of microfracture.

The size of fracture unit did not vary with the test temperature, but as the temperature was raised, the fine ductile zones were observed to be networks around the brittle fracture region.²⁸⁾ As the temperature was raised still more, the ductile fracture region with the size of fracture unit was observed.²⁸⁾

IV. Discussions

In the relation of the impact transition temperature

Fig. 11. Relation between unit crack path and vT_s for continuously cooled Cu-Ni-Cr-Mo-V steel (steel 7)

with the cooling rate on continuous cooling, similar characteristics were observed between the as-cooled specimen and the tempered one. Consequently, it is considered that the transition temperature essentially depended on the microstructures formed by transformation during continuous cooling from the austenitizing temperature. It was clarified that the duplex martensite-B III type bainite had superior impact properties.

It was difficult to compare the toughness of martensite with that of fully B III type bainite, because of difficulties to obtain the fully B III type bainite by continuous cooling in such low-carbon low-alloy steels. Although the carbide morphology of bainite was quite different from that of martensite,¹⁹⁾ the aspects of fracture surfaces scarcely revealed different appear-

Photo. 10. Unit crack path of typical microstructures formed in Cu-Ni-Cr-Mo-V steel (steel 7) during continuous cooling from 1200°C ($\times 600$) ($\times 3/4$)

- (a) Duplex martensite-bainite structure broken at -100°C
- (b) Ferrite-bainite structure broken at -20°C

ances. The habit plane and growth direction of both martensite and bainite were $\{110\}_a\langle 111\rangle_a$ and the thickness of lath in bainite was not less than that in martensite,¹⁹⁾ so there could be no substantial differences contributed to the toughness. The dislocation densities should have some effects on the toughness essentially, but it should be pointed out that the dependencies of transition temperature on the cooling rate were similar before and after tempering.

It has been shown previously that the bainite had superior strength and toughness.^{13,29,30)} In evaluation of toughness of different microstructures, it has to be made in the same strength level, but to make the strength level of martensite as same as that of untempered bainite, martensite has to be tempered at the low temperature which is not preferable to the toughness. So, generally it is difficult to evaluate the superiority of toughness among different microstructures. In the present experiment, however, the strength was almost the same in the range near the optimum cooling rate for vT_s , and there was little strength difference among various structures after tempering, so it was able to evaluate the toughness at the same strength level.

Photograph 11 shows the direct correspondence between fracture surface and the duplex structures etched by 2% nital, exhibiting somewhat different etching sensitivity. The direction of crack propagation (shown by arrows) changed at the boundary of the region etched darkly (B in Photo. 11). This structure appeared apparently prior to martensite transformation (M in Photo. 11). Although it appeared to

Photo. 11. Correspondence of duplex martensite-bainite structure to fractured surface ($\times 10\,000$) ($\times 4/5$)

be uncertain that this is bainite or autotempered martensite, the carbide morphology was quite similar to that of B III type bainite. The fracture path changed its direction at the boundary of bainite but within the lath (B) there was no evidence of the resistance against fracture, and consequently, the lath itself could not seem particularly effective against brittle fracture.

Recently it was reported³¹⁾ that martensite was tougher than lower bainite but duplex martensite-bainite structure was tougher than those structures. In this case, however, although the bainite was lower bainite different from upper bainite, it was suggested that the boundary between different microstructures might play important role to enhance the toughness.

The observation of fracture surface revealed that there was the correlation between dc and vT_s , and the martensite-bainite structure with superior toughness showed the finest fracture unit. The superior toughness of the duplex structure is probably related to the mode of bainite formation. In other words, the bainite partitioned the austenite grain prior to the martensite transformation as well as providing the increased transformation sites.

To obtain such a duplex structures on continuous cooling, the bainite should be B III type bainite. Because the impact properties depended on the fine substructure, the partitioning by bainitic ferrite formed at the high temperature seems not effective for the toughness. It is because that the lath of bainite itself and the fracture unit are large and the austenite with enriched carbon produces the high carbon transformation products. The partitioning by martensite seems effective, but the controlling of martensite transformation in low-carbon low-alloy steel is difficult.

V. Summary

The impact properties of some high tensile strength steels were investigated in relation to the cooling rate on continuous cooling and the austenite grain size. The direct observations of impact fracture surface in relation to microstructures were performed with the scanning electron microscope. The results obtained are as follows:

- (1) There was the optimum cooling rate for the impact transition temperature.
- (2) The transition temperature characteristics against the cooling rate after tempering were similar to that of as continuously cooled ones.
- (3) The optimum cooling rate varied with the hardenability of steel and as the austenitizing temperature was raised, it shifted to the lower cooling rate.
- (4) The optimum cooling rate was smaller than the critical cooling rate for the beginning of bainite formation, and the electron microscopy revealed that the microstructure obtained on the optimum cooling rate was consisted of martensite and bainite.
- (5) The morphology of carbide in the bainite was similar to that in lower bainite of a higher carbon steel, but the trace analysis for such bainite revealed that it had the same lath-like morphology with a $\langle 111\rangle_a\{110\}_a$ habit as that in the upper bainite and the

lath martensite. (This bainite in a low-carbon low-alloy steel was designated as Bainite III type in the previous paper.¹⁹⁾

(6) The transition temperatures of microstructures consisted of ferrite-pearlite, martensite and bainite formed at higher temperature had the correlation to the austenite grain size, but in the duplex martensite-bainite structure, the austenite grain size dependency of transition temperature was small. There was a case that, even though the austenite grain coarsening occurred, the transition temperature was improved.

(7) The direct observations of fracture surface in relation to the microstructure with a scanning electron microscope were performed. The orientation measurements of facet pit and fracture surface revealed that the crystallographic orientation of fracture surface was {100}.

(8) On the basis of the direct observation of fracture characteristics in detail, "the region consisting of {100} fracture planes of which orientation differences are within a small angle" was taken as a fracture unit and in this unit a crack nucleated, propagated, and was arrested. It was experimentally clarified that there was a better correlation of the transition temperature to the unit crack path than to the austenite grain size. The fine aspect of fracture surface, however, in a fracture unit was dependent on the microstructure, such as martensite and ferrite-pearlite.

(9) The fracture unit became finest in the duplex martensite-bainite structure, and it was extremely large in the bainite mainly consisted of bainitic ferrite.

(10) These results suggested that the role of bainite in the duplex structure was the partitioning of austenite grain prior to the martensite transformation.

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