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Effect of pre-quenching on martensite–bainitic microstructure and mechanical properties of GCr15 bearing steel

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In order to improve the strength and toughness of steel GCr15 (52100), the effect of different amounts of pre-transformed martensite on the kinetics of isothermal bainitic transformation and the strength and toughness of martensite–bainitic (MB) duplex microstructure has been studied by using pre-quenching after conventional 850 °C heating to different temperatures (220, 200, 180 °C) below M_s and then isothermal treatment at 240 °C. The experimental results show that the accelerating effect of pre-quenched martensite on isothermal bainitic transformation principally depends upon the pre-quenching temperature (the amount of pre-quenched martensite). The MB duplex microstructure with 33% pre-transformed martensite has the optimum combination of strength and toughness.

1. Introduction

As is well known, in order to ensure high surface hardness, wear resistance and a good combination of strength and toughness for rolling bearings, a lot of measures are commonly used such as the production of carburized bearing steel, bearing steel with low hardenability, bainite austempering treatment, etc., instead of the conventional quenching and tempering of high-carbon chromium-bearing steels. Recent studies [1–4] on the martensite–bainite (MB) duplex microstructure and its strengthening and toughening effect on GCr15 steel indicate that an austempering near the M_s point may possibly become a new sort of strengthening and toughening technology to bring out the property potentials of high-carbon chromium-bearing steels. In this paper, the effect of pre-quenched martensite on isothermal bainitic transformation near the M_s point was studied in order to reveal the strength and toughness of the MB duplex microstructure obtained by pre-quenching to different temperatures below M_s, followed by an isothermal treatment.

2. Experimental procedure

Specimens were cut from the same ingot. The chemical composition and metallurgical quality satisfy the requirements for standard GCr15 (identical to 52100) steel.

The kinetic curves of bainitic transformation at different isothermal temperatures near the M_s point were determined by a Formaster-D type dilatometer. The M_s point was determined to be 235 °C for specimens austenitized at 850 °C.

Dimensions of specimens for property measurements are shown in Table I, in which non-standard fracture toughness specimens with a slit produced by spark machining but without pre-created fatigue cracks were used. Low-energy multiple impact specimens with a notch in the middle of the side edge were subjected to unidirectional multiple impact at a rate of 450 min^{-1}.

The morphology of lower bainite in the MB duplex microstructure was revealed by Lepera's etching agent [5] and its volume fraction (an average from five visual fields) was measured by the net-point counting method.

The amount (V_m) of martensite for pre-quenching to different temperatures below the M_s point was estimated by the empirical formula [6]

$$V_m = 1 - 6.956 \times 10^{-15} (455 - \Delta T)^{5.32}$$

in which ΔT is the difference between the M_s and pre-quenching temperatures. The volume fraction of retained austenite was determined by magnetic measurements.

The different heat treatment conditions are shown in Table II. The specimens were pre-quenched to three different temperatures, (C, D, E) below the M_s point for 5 min followed by an isothermal holding at 240 °C. For each pre-quenching temperature, different times for isothermal holding at 240°C were used in order to control nearly the same amount of 30–40% bainite in

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TABLE I Dimensions of specimens for various tests

	Hardness	Bending	Impact toughness	Fracture toughness	Low-energy multiple impact
Dimensions (mm)	ϕ 10 × 15	ϕ 10 × 60	ϕ 10 × 60 (no notch)	8 × 16 × 72	10 × 10 × 55

TABLE II Heat treatment procedures

Procedure	Heat treatment conditions[a]		
A	850 °C, 10 min	→ 240 °C, 23 min → OQ	→ Tempering at 200 °C for 1.5 h
B		→ 220 °C, 5 min → 240 °C, 20 min → OQ	
C		→ 200 °C, 5 min → 240 °C, 18 min → OQ	
D		→ 180 °C, 5 min → 240 °C, 23 min → OQ	
E		→ OQ	

[a] OQ = Oil quench.

all cases. The same austenizing and tempering temperatures and times were selected to eliminate the effect of austenite grain size and the amount of undissolved carbides.

3. Results

3.1. Kinetics of isothermal transformation of prequenched GCr15 steel

In order to inquire into the effect of pre-quenched martensite on the transformation behaviour of isothermal bainitic at temperatures slightly higher than M_s, specimens were austenitized at 850 °C, pre-quenched to 220, 200, 180, 160 or 120 °C for 5 min and then held isothermally at 240 °C. Fig. 1 shows the kinetic curves of isothermal bainitic transformation of pre-quenched martensite determined by the dilatometric method, where the terrace height of the first part of each curve shows the fraction of pre-quenched martensite obtained at different temperatures. The dotted line connected by the demarcated points on the curves shows the effect of pre-transformed martensite formed at different temperatures on the incubation period of isothermal bainitic transformation at 240 °C. It can be seen that with the decrease of pre-quenching temperature and thus the increase of athermal martensite, the changes of the incubation period of isothermal bainitic transformation at 240 °C give a C-shaped curve. Therefore through the pre-quenching below M_s, because of the existence of a certain amount of pre-transformed martensite and its stress-induced effect, nucleation of the α-phase in bainite is accelerated [7] and the incubation period of bainitic transformation is constantly shortened. When the pre-quenching temperature is about 200 °C and the amount of pre-transformed martensite is about 33% (estimated from the formula and very close to the test result), the incubation period of isothermal bainitic transformation at 240 °C is the shortest. But when the pre-quenching temperature is too high, the retained austenite is seriously partitioned by martensite and bears more complicated stresses which may possibly prolong the incubation period of bainitic transformation. This phenomenon has been studied in more detail elsewhere [8].

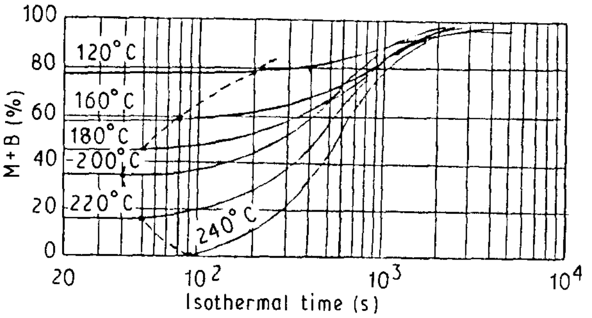

Figure 1 Kinetic curves of isothermal bainitic transformation at 240 °C of steel GCr15 initially pre-quenched to different temperatures.

3.2. Duplex microstructure of pre-quenched and isothermally treated steel GCr15

The study on the MB duplex microstructure and its strength and toughness has been firstly carried out in low- and medium-carbon steels. The amount of bainite in the MB duplex microstructure is usually selected as about 15–20% according to the requirements of mechanical properties. In order to eliminate the tendency to brittle fracture and to meet the needs of suitable hardness values (not less than HRC59) of GCr15 steel, the amounts of bainite duplex microstructure recently investigated differ significantly from each other.

In order to analyse the MB duplex microstructure obtained in this study, first of all, the morphology of conventionally quenched and tempered microstructure related to procedure A in Table I is shortly described as follows. Research [9] showed that there exists a microscopically inhomogeneous distribution of the elements C, Cr due to inhomogeneous dissolution of globular carbide $(Fe, Cr)_3C$ within austenite grains of steel GCr15 with originally spheroidized pearlitic structure after conventional austenitizing in the two-phase region. This leads not only to a quenching structure with a fine martensite, but also to morphological features of black and white regions which arise from inhomogeneous distribution of retained carbides, as shown in Fig. 2. In fact, the cloud-shaped white regions are rich in carbon and chromium, while the black regions possess a higher

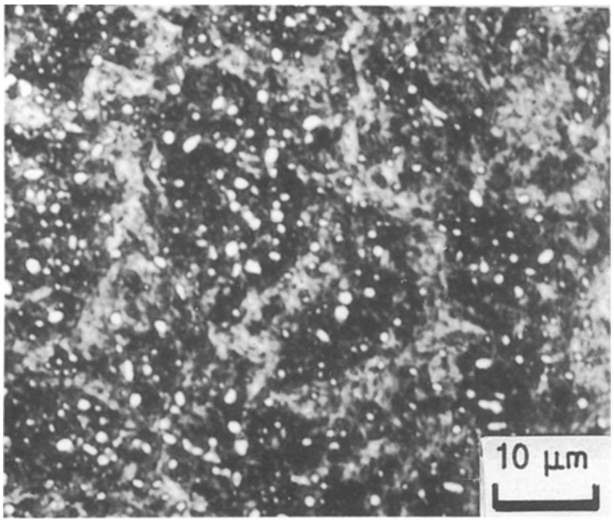

Figure 2 Microstructure of steel GCr15 conventionally quenched and tempered.

density of retained carbides and thus are poor in the elements carbon and chromium.

The duplex microstructures of each procedure given in Table II are shown in Fig. 3. Fig. 3a shows a control austempered structure obtained by pure isothermal quenching at 240 °C for 23 min (with 200 °C, 1.5 h tempering afterwards). In supercooled austenite there firstly forms a certain amount of lower bainite distributed around the retained carbides and then the carbon-rich fine martensite forms upon subsequent cooling. Fig. 3b–d show the microstructures obtained by firstly pre-quenching to different temperatures below M_s and then isothermally treating at 240 °C. Apparently, with a decrease of pre-quenching temperatures, an amount of low-carbon martensite firstly forms in microregions with lower carbon and the chromium content of the supercooled austenite (higher M_s point) and increases correspondingly, and possesses the feature of the black regions with a higher density of retained carbides. Except for a certain amount of bainite which forms in the specimen isothermally treated at 240 °C, the other part of the supercooled austenite will transform into athermal martensite upon subsequent cooling to room temperature.

The two duplex microstructures formed by two different procedures, of which one is martensite first and bainite second and the other is bainite first and martensite second, have been further proved by TEM observations. Fig. 4 is the TEM morphology of a control specimen treated by isothermal quenching at 240 °C for 23 min, in which Fig. 4a shows the lower bainite formed firstly in regions of higher density of retained carbides; Fig. 4b shows the twin-type and

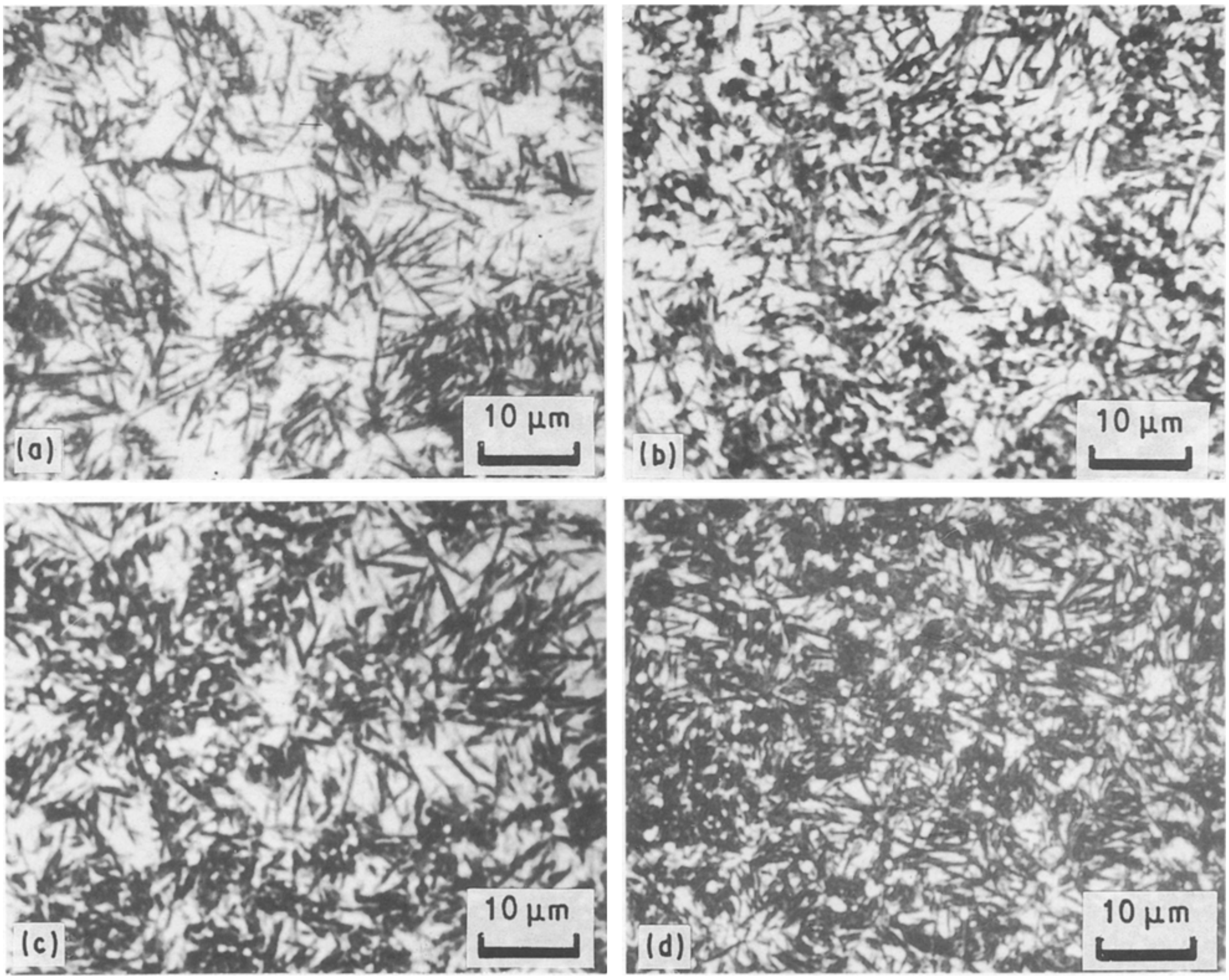

Figure 3 (a–d) MB duplex structure of steel GCr15 variously quenched (for details see text).

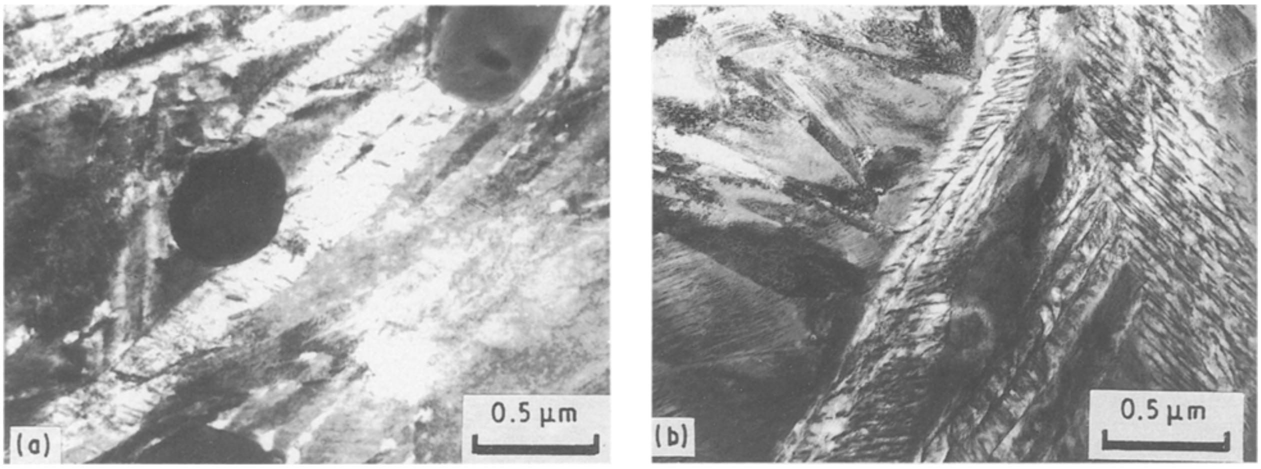

Figure 4 (a, b) MB duplex structure of steel GCr15 control, austempered at 240 °C for 25 min.

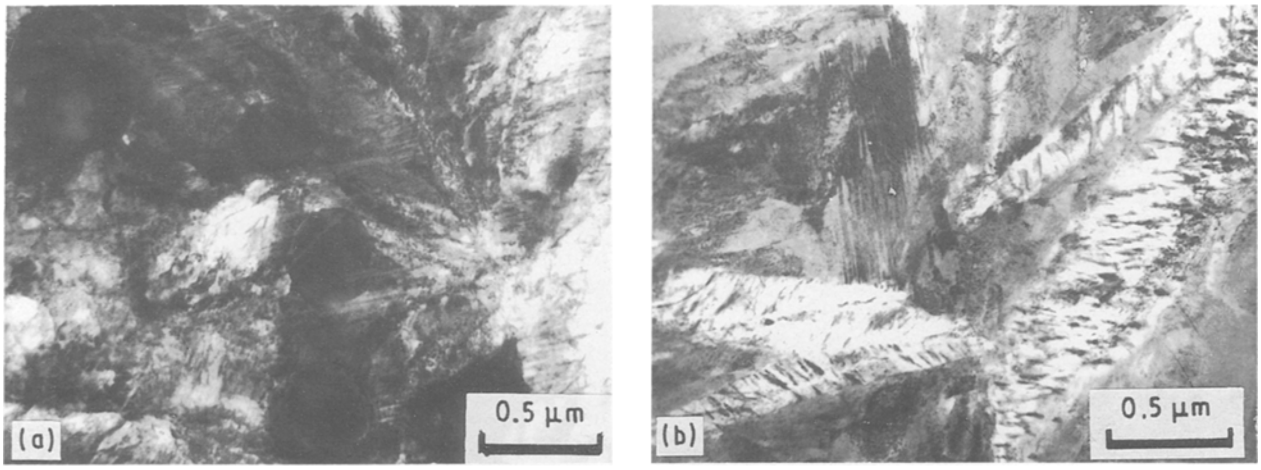

Figure 5 (a, b) TEM microstructure of steel GCr15 pre-quenched to 200 °C for 5 min and isothermally treated at 240 °C for 18 min.

dislocation-type martensite in areas far from the regions of high density of retained carbides. Fig. 5 shows the TEM morphology of material treated by pre-quenching at 200 °C for 5 min and then isothermal treatment at 240 °C for 18 min, in which Fig. 5a shows the pre-quenched dislocation-type martensite around the regions of high density of retained carbides, and Fig. 5b shows isothermally formed lower bainite and a small amount of twin-type martensite in the duplex microstructure. Fig. 6 shows the retained austenite distributed between martensite and bainite in duplex microstructures.

The estimated amounts of microstructural constituents (pre-quenched martensite, isothermal bainite, retained austenite, retained carbide and athermal martensite) obtained by different treatments are shown in Table III. In this table, the data for the amount of retained carbide are quoted from our previous work [9] as determined by electrolytic dissolution.

3.3. Strength and toughness of pre-quenched and isothermally treated steel GCr15

3.3.1. Strength and toughness

The values of mechanical properties (hardness, bending strength, impact toughness and fracture toughness) are shown in Table IV for specimens treated by the procedures shown in Table II. It can be seen that the hardness values of specimens with different microstructures are all higher than HRC 59, i.e. they all satisfy the needs of technical conditions. The effect of pre-quenching temperature on the strength and toughness is illustrated by Figs 7 and 8, drawn according to the data given in Table IV. The properties of the

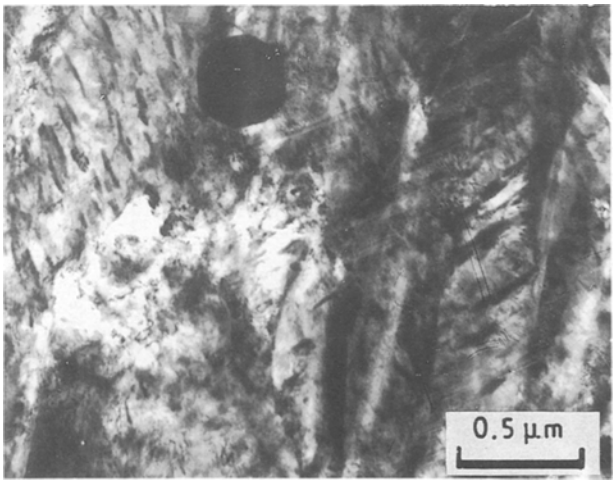

Figure 6 Retained austenite distributed in the MB duplex structure.

TABLE III Amount of various microstructural constituents for different procedures

Procedure	Content (%) Pre-quenched martensite	Isothermal bainite	Retained austenite	Retained carbide	Athermal martensite
A			7–8	6–7	85–87
B		40	8–9	6–7	44–46
C	18	30	7–10	6–7	35–39
D	33	35	9–10	6–7	13–15
E	45	33	10–12	6–7	3–6

TABLE IV Various properties after different treatments

Procedure	Hardness (HRC)	Bending strength (MPa)	Deflection (mm)	Impact work (J)	Fracture toughness (Mpa $m^{1/2}$)
A	61.5	4353	1.60	48.20	42.4
B	60.0	4794	2.40	67.60	50.2
C	59.5	5059	2.75	69.09	51.0
D	59.5	5129	2.87	71.39	52.0
E	59.0	4822	2.35	66.84	49.4

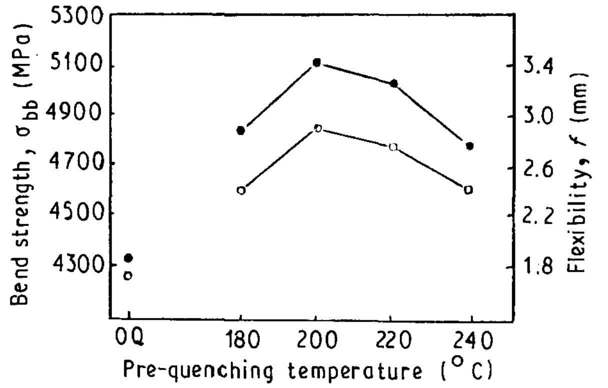

Figure 7 (●) Flexural strength and (○) deflection of variously quenched specimens.

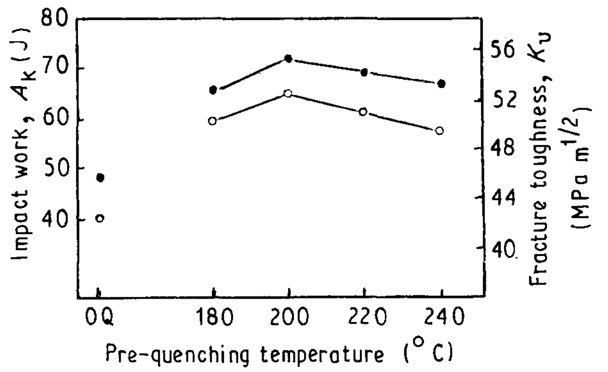

Figure 8 (●) Impact toughness and (○) fracture toughness of variously quenched specimens.

directly quenched or austempered control specimens are also given for comparison.

The experimental results show that the properties of the specimens treated firstly by pre-quenching to different temperatures below M_s and then isothermally treated have better combinations of strength and toughness than those of the directly oil-quenched or austempered ones. Apparently, this is connected with the fact that the pre-quenched low-carbon martensite (naturally tempered during isothermal treatment at 240 °C) and a lesser amount of athermal martensite were formed upon subsequent cooling, besides the existence of a suitable amount of lower bainite (distributed in microregions with a higher carbon content).

Fig. 9a–c show, respectively, SEM fractographs taken from fracture toughness specimens treated by direct oil quenching, pure austempering, and pre-quenching to 200 °C followed by 240 °C isothermal treatment. It is revealed that although they all have quasi-cleavage fractures, more obvious tearing edges consisting of microdimples are present in the MB duplex microstructure because of the existence of a suitable amount of lower bainite, indicating that the specimen had already experienced a certain amount of plastic deformation when fracture occurred. This reflects indirectly the good combination of strength and toughness of the MB duplex microstructure.

3.3.2. Low-energy multiple impact properties

The service life of bearings under conditions of impacting loads, such as railway bearings, rolling mill bearings and so on, is usually associated with the accumulation of damage resulting from multiple impact. It is, therefore, of practical value to examine the low-energy multiple impact properties of steel GCr15. Fig. 10 shows the multiple impact lives of the steel after three kinds of heat treatment procedure under different impact energies, drawn according to the data in Table V. It can be seen that all the multiple impact lives increase with decreasing impact energy. However, the impact lives of MB duplex microstructures are much longer than those of the conventionally quenched and tempered specimen under the same

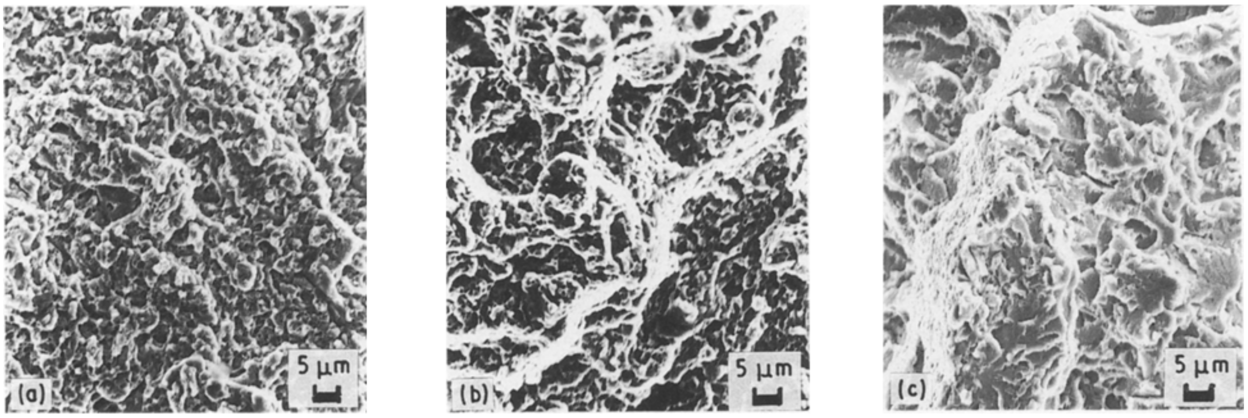

Figure 9 (a–c) SEM fractographs of three comparative heat treatment procedures (for details see text).

TABLE V Low-energy multiple impact life for different heat treatment procedures

Procedure	Heat treatment condition	Multiple impact lives at different impact energies		
		1.37 J	1.08 J	0.74 J
A	850 °C, 10 min → OQ → tempering at 200 °C for 1.5 h	2430	8325	14 587
B	850 °C, 10 min → 240 °C, 23 min → OQ → tempering at 200 °C for 1.5 h	6925	16 650	29 250
C	850 °C, 10 min → 200 °C, 5 min → 240 °C, 18 min → OQ → tempering at 200 °C for 1.5 h	8910	21 615	36 681

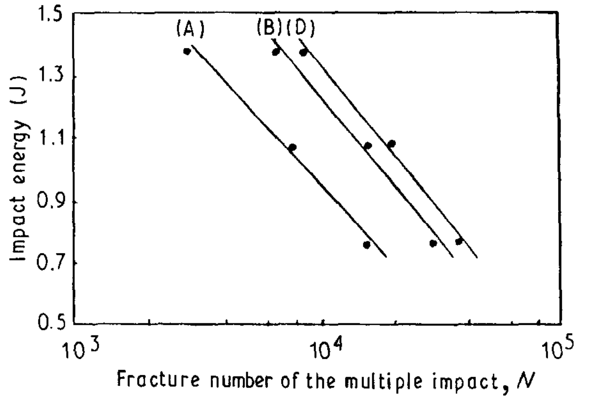

Figure 10 Multiple impact life of variously quenched specimens (for details see text).

impact energy. Evidently, this results from the presence of a suitable amount of lower bainite and its contribution to strength and toughness.

As is shown above, the duplex microstructures produced by either bainite first and martensite second or martensite first and bainite second treatments possess higher strength and toughness than the conventionally quenched and tempered ones, due to the presence of suitable amounts of lower bainite. The distribution of a certain amount of lower bainite in the quenched matrix can hinder the propagation of cracks, resulting in a zig-zag path and hence an increase in the total crack area. As a result, more work must be done to compensate for the increased strain-field energy and crack surface energy. On propagating through MB interfaces of different orientations, the increase in resistance will slow down the propagating rate of the crack. Consequently, the appearance of large amounts of tearing edges and local microdimples is attributed to the plastic deformation of lower bainite under the action of three-dimensioned stresses at the tip of the propagating cracks.

4. Discussion

In previous studies on the strengthening and toughening mechanisms of martensite–bainite duplex microstructures in low- and medium-carbon steels it was suggested that the original austenite grains are partitioned by the initial formation of isothermal bainite and the refinement of martensite transformed upon subsequent cooling. However, it is worth doubting whether the partitioning and refining model can be used to explain the combination of high strength and high toughness of the MB duplex microstructures. As is shown above, the strength and toughness values of duplex microstructures obtained either by controlled austempering at 240 °C (first bainite, second martensite), or by pre-quenching at 180–220 °C followed by isothermal treatment at 240 °C (first martensite, second bainite) after austenitization at 850 °C, show no marked difference. This indicates that the strengthening and toughening model, by which the initially formed bainite partitions the austenite grains and thus refines the martensite, has no marked effect on improving the strength and toughness under the condition of conventional quenching from 850 °C. But it has been shown [4] by the authors previously that in the case of higher-temperature quenching at 890 °C (M_s = 175 °C), under conditions of the same amount of bainite, duplex microstructures obtained either by direct isothermal quenching at 240 °C or by pre-quenching to 150 °C followed by isothermal treatment at 240 °C have obvious differences in strength and toughness, with the latter procedure giving lower values. Therefore, in the case of higher-temperature quenching, the increase of carbon content and more homogeneous distribution of C, Cr elements in austenite grains and, particularly, the occurrence of coarse-grained twinned martensite will undoubtedly decrease the strength and ductility. In this case, the initially formed lower bainite partitions and effectively refines the austenite grains, resulting in relatively higher strength and toughness values.

5. Conclusions

1. After 850 °C austenitization, pre-quenching to temperatures lower than M_s can remarkably influence the incubation period of the subsequent isothermal bainitic transformation at 240 °C. The shortest incubation period can be obtained after pre-quenching to 200 °C. When the pre-quenching temperature is too low, and the amount of pre-transformed martensite is too high and pre-quenching is not beneficial to accelerate the bainitic transformation.

2. Excellent combinations of strength and toughness have been shown in a suitable martensite–bainitic duplex microstructure of steel GCr15 obtained by pre-quenching to 200 °C below the M_s temperature, followed by isothermal quenching at 240 °C for 18 min (followed by 200 °C tempering for 1.5 h).

3. Microstructures obtained by either isothermal quenching at 240 °C or by pre-quenching 180–220 °C, below M_s, followed by isothermal treatment at 240 °C after austenitization at 850 °C, show no remarkable difference, i.e. the partioning of austenite grains by initially formed bainite does not contribute markedly to the improvement of strength and toughness of bearing steel GCr15.

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