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RELATIONSHIP OF BAINITIC MICROSTRUCTURE TO IMPACT TOUGHNESS IN Cr-Mo AND Cr-W STEELS

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ABSTRACT

Non-classical bainite microstructures can develop during continuous cooling of low-carbon alloy steels. These differ from classical upper and lower bainite developed by isothermal transformation. Two non-classical bainite microstructures were produced in a 3Cr-1.5Mo-0.25V-0.1C steel using different cooling rates after austenitizing--water quenching and air cooling. The carbide-free acicular bainite formed in the quenched steel had a lower ductile-brittle transition temperature (DBTT) than the granular bainite formed in the air-cooled steel. With increasing tempering parameter (defined by tempering time and temperature), the DBTT of both decreased and approached a common value, although the final value occurred at a much lower tempering parameter for the quenched steel than for the air-cooled steel. The upper-shelf energy was similarly affected by microstructure. These observations along with similar observations in two Cr-W steels indicate that control of the bainite microstructure can be used to optimize strength and toughness.

1. INTRODUCTION

A previous paper [1] demonstrated that the type of bainite microstructure developed in a 3Cr-1.5MoV steel during continuous cooling from the austenitization temperature affected the impact toughness. In this paper, an extension of that work is discussed along with similar observations on Cr-W steels being developed for fusion reactors [2].

2. EXPERIMENTAL

Details on the chemical composition of the argon-oxygen-decarburized (AOD) heat of 3Cr-1.5MoV steel (3Cr-1.5Mo-0.1V-0.1C, composition in wt %) used for this study have been published [1]. The ingot was processed to 100-mm-thick plate, and plates measuring 0.1 x 1.6 x 1.4 m were heat treated. One plate was air cooled (normalized) and one was water quenched after annealing 2 h at 955°C; both were stress relieved 2 h at 565°C. Despite the stress relief, these materials will be referred to as normalized and quenched, respectively (there was little indication of precipitation after the stress relief [1]). Impact properties were measured with standard Charpy V-notch specimens (10 mm × 10 mm × 55 mm) taken from the 1/4- and 3/4-thickness depths in the plates. Cooling rates at these thicknesses were approximately 0.12 and 1.67°C/s (7 and 100°C/min) for the normalizing and quenching heat treatment, respectively. The plates were tempered at 663 to 704°C and from 1 to 30 h. Tempering conditions were expressed as a tempering parameter, $TP = T(20 + \log t) \times 10^{-3}$, where T is temperature in Kelvin and t is time in hours.

The Cr-W steels were experimental heats with the following nominal compositions (in wt %): 2¼Cr-2W-0.1C (designated 2¼Cr-2W) and 2¼Cr-2W-0.25V-0.1C (designated 2¼Cr-2WV). Details on composition, microstructure, and processing have been published [2]. To demonstrate the effect of cooling rate on the steels, 10-mm-square and 3.3-mm-square bars were normalized by first austenitizing in a helium atmosphere in a tube furnace and then cooled by pulling into the cold zone. To austenitize, the 2¼Cr-2W steel was annealed 1 h at 900°C, and the 2¼Cr-2WV was annealed 1 h at 1050°C. The higher temperature was used for the latter steel to assure that any vanadium carbide present dissolved during austenitization.

Miniature Charpy specimens of the Cr-W steels were machined from 15.9-mm-thick plate. The specimens were essentially one-third the standard size and measured 3.3 x 3.3 x 25.4 mm and contained a 0.51-mm-deep 30° V-notch with a 0.05 to 0.08-mm-root radius. Specimens were machined from 15.9-mm plate in the longitudinal orientation with a transverse crack (LT). To determine the effect of heat treatment, specimens taken directly from 15.9-mm heat-treated plate were tested. Tests were also carried out on specimens that were normalized directly. The normalizing treatments were the same as those used for the bars. Tempering conditions were 1 h at 700°C and 1 h at 750°C.

3. RESULTS

3.1 Microstructures

Transmission electron microscopy (TEM) of the quenched and the normalized 3Cr-1.5MoV steel plates is shown in figure 1. Based on optical microscopy, TEM, and the continuous-cooling transformation diagram [3], both the quenched and the air-cooled microstructures were entirely bainite, although neither microstructure was typical of classical upper or lower bainite.

The quenched steel had a lath structure that contained a high density of dislocations. Figure 1(a) is typical of most of the specimen, although a few isolated areas contained laths that were not as developed and were similar to the more equiaxed microstructure of the normalized steel. However, there were no indications of the dark islands that appeared in the normalized steel [figure 1(b)].

The normalized steel microstructure consisted mainly of regions containing a high dislocation density in which dark regions or "islands" were scattered throughout [figure 1(b)]. The subgrain structure was equiaxed, rather than the lath-type morphology of the quenched steel. A few scattered regions contained indistinct laths, indicating a tendency toward elongated subgrains. The lath structure was not as developed as it was during quenching, and it contained islands, which were often elongated.

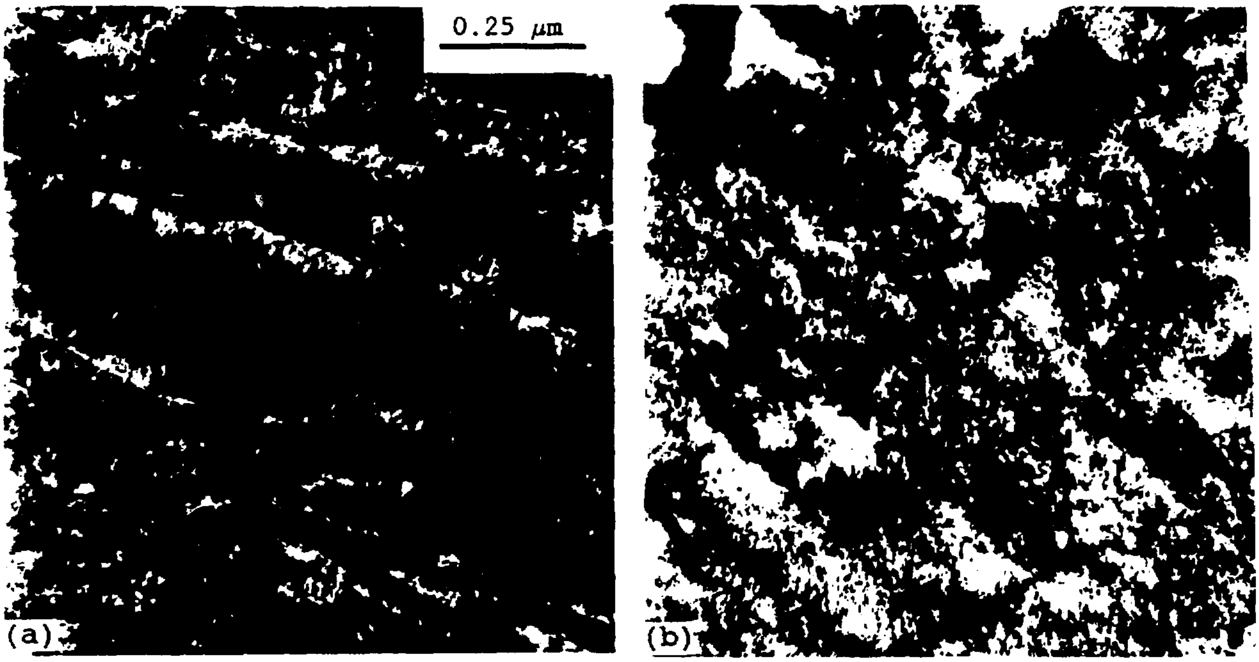

Figure 1. Transmission electron micrographs of (a) quenched and (b) normalized 3Cr-1.5MoV steel.

Before tempering, there was little indication of any precipitate present in either the quenched or the normalized steel [1]. After tempering, TEM and carbide extraction replicas indicated that the quenched steel developed carbides on the lath boundaries with smaller needlelike precipitates within the matrix. The normalized steel also developed two types of precipitate morphologies: regions containing an agglomeration of globular carbides surrounded by a high density of finer needlelike precipitates. The agglomeration of globular carbides was associated with the islands in figure 1(b) [1].

The 2¼Cr-2W and 2¼Cr-2WV steels were examined after cooling at two different rates (different sized specimens were cooled). Optical metallography indicated both steels were 100% bainite after both the fast and slow cool, although there were differences in appearance. The specimen given the fast cool appeared more acicular. When the microstructures were examined by TEM, the results were similar to those for the 3Cr-1.5MoV steel. The fast-cooled steel had a lath structure [figure 2(a)], and the slow-cooled steel had an equiaxed structure with the dark islands [figure 2(b)].

3.2 Impact Toughness

Charpy impact tests were conducted on the quenched and the normalized plates of 3Cr-1.5MoV steel after various tempering treatments (different tempering parameters) (figure 3). Tempering had a larger immediate effect on the ductile-brittle transition temperature (DBTT) of the quenched plate than the normalized plate [figure 3(a)]. A tempering parameter of 18.7 x 10^3 (1 h at 663°C) lowered the DBTT of the quenched plate from 67 to -48°C, whereas the same tempering treatment for the normalized steel lowered it from 98 to 60°C. With continued tempering, the DBTT of both the quenched and the normalized plates decreased, with the decrease more pronounced for the normalized plate. Only after the highest tempering parameter used (20.7 x 10^3, 16 h at 701°C), however, did the DBTT of the normalized steel approach that for the quenched steel. The relative effect on the

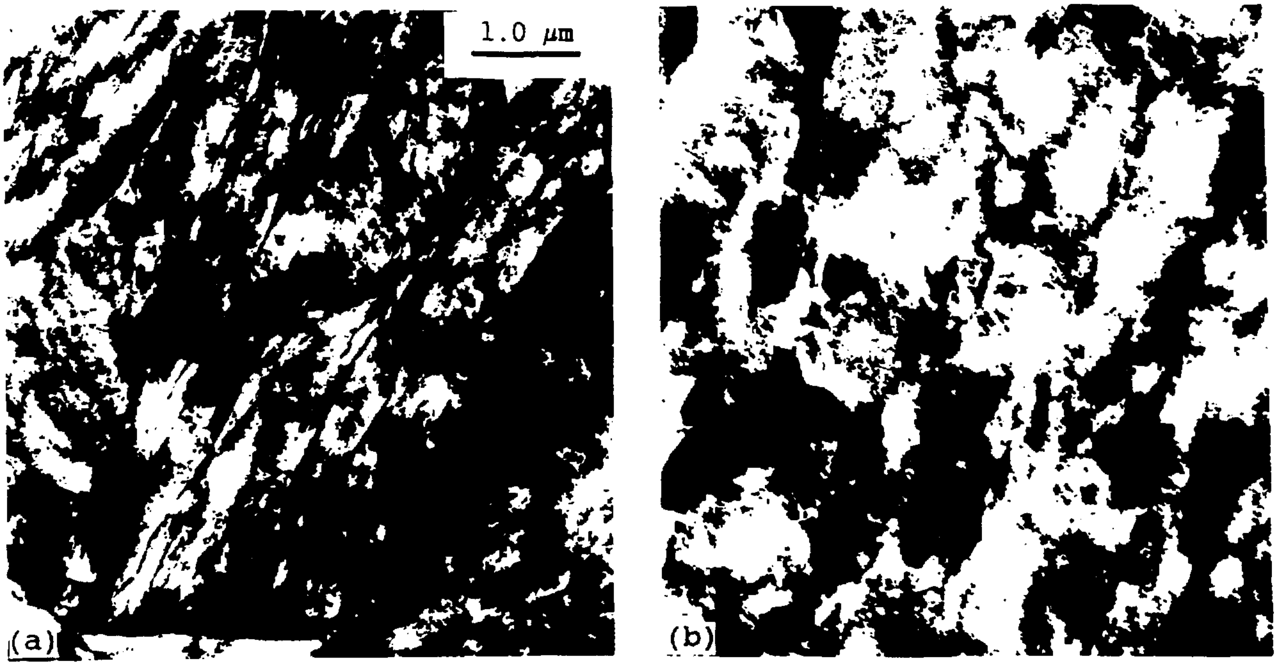

Figure 2. Transmission electron micrographs of (a) fast-cooled and (b) slow-cooled $2\frac{1}{4}$Cr-2W steel.

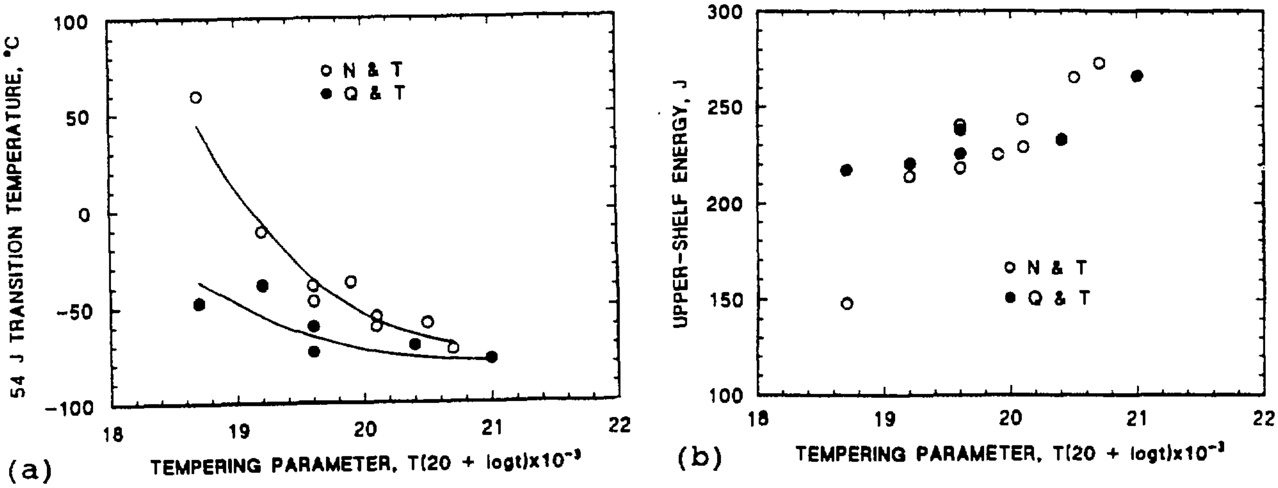

Figure 3. (a) 54 J Charpy ductile-brittle transition temperature and (b) upper-shelf energy plotted against the tempering parameter for normalized-and-tempered and quenched-and-tempered 3Cr-1.5MoV steel.

upper-shelf energy (USE) was somewhat similar [figure 3(b)]. The lowest tempering parameter produced a high USE for the quenched plate relative to that for the normalized plate. However, for higher tempering parameters, the USE of the quenched and the normalized plates had similar values.

When the two Cr-W steels were heat treated in the 1/3-size Charpy specimen geometry (3.3-mm square cross section), they were 100% bainite. Table 1 shows the Charpy impact results when heat treated as 1/3-size Charpy specimens and tempered at 700 and 750°C.

Table 1. Charpy Impact Properties of Reduced-Activation Steels

Alloy	Heat Treatment Geometry					
	1/3-Size Specimen				15.9-mm Plate[a]	
	1 h at 700°C		1 h at 750°C			
	DBTT (°C)	USE (J)	DBTT (°C)	USE (J)	DBTT (°C)	USE (J)
2¼Cr-2W	-56	11.5	-77	10.1	-48	9.6
2¼Cr-2WV	-9	7.0	-52	11.0	0	9.7

[a]2¼Cr-2W was tempered 1 h 700°C; 2¼Cr-2WV was tempered 1 h 750°C.

respectively. Also given are results when heat treated as 15.9-mm plate (the 2¼ Cr-2W was tempered at 700°C, and the 2¼Cr-2WV at 750°C). For the latter heat treatment, the 2¼Cr-2W was 100% bainite, and the 2¼Cr-2WV contained 80% bainite and 20% polygonal ferrite. The DBTT in Table 1 was determined where the impact energy was one-half of the USE.

The microstructure, as reflected in the size of the specimen heat treated, had a large effect on the properties of 2¼Cr-2WV, with much less effect on the 2¼Cr-2W (Table 1). Whereas tempering the 1/3-size specimen of 2¼Cr-2WV at 750°C significantly lowered the DBTT and raised the USE relative to values obtained when tempered at 700°C, the effect of this tempering treatment had much less effect for the 2¼Cr-2W. In fact, for the 2¼Cr-2W, there was little difference when heat treated as 15.9 mm plate (and tempered at 700°C) or as a 1/3-size Charpy specimen (and tempered at 700 or 750°C). This contrasts with the 2¼Cr-2WV steel with a high DBTT for the bainite plus polygonal ferrite microstructure of the 15.9-mm plate (tempered at 750°C), the slightly lower DBTT when the 1/3-size Charpy specimen was normalized and tempered at 700°C, and then the still lower value for the 1/3-size Charpy specimen normalized and tempered at 750°C.

4. DISCUSSION

Bainite develops in steels when transformation from austenite occurs between the temperatures where ferrite and pearlite form, at high temperatures, and martensite forms, at low temperatures. It was originally thought to consist of two easily distinguishable morphological variations, upper and lower bainite, defined according to the temperature of formation. Habraken [3] demonstrated there were microstructural variations on these classical bainite microstructures that formed in the bainite transformation temperature regime. Such "nonclassical" bainite formed more easily during continuous cooling than during isothermal transformation [3,4], where upper and lower bainite formed.

Habraken and Economopoulos [4] contrasted classical and nonclassical bainite using the isothermal transformation (IT) and continuous-cooling transformation (CCT) diagrams. The bainite transformation region of an IT diagram can be divided into two temperature regimes by a horizontal line. Transformation above this line results in upper bainite and below it in lower bainite. For nonclassical bainite, Habraken and Economopoulos [4] showed that a CCT diagram could be divided into three vertical regions. Three different microstructures form when cooling rates are such as to pass through these different zones. A steel cooled rapidly enough to pass through zone I produces a "carbide-free acicular" structure, which consists of side-by-side plates or laths of ferrite containing a high-dislocation density [4]. For an intermediate cooling rate through zone II, a carbide-free "massive" or "granular" structure results, which is generally referred to as granular bainite [4]. Granular bainite has a ferrite matrix consisting of equiaxed subgrains of ferrite with a high dislocation density coexisting with dark "islands" [4]. These islands are enriched in carbon during the bainitic transformation and have been shown to be high-carbon retained austenite, part of which can transform to martensite when cooled below M_s. These regions are referred to as martensite-austenite (or M-A) islands [4]. Since the microstructures developed by slow cooling through zone III were not observed in this study, they will not be discussed.

Microstructures observed by TEM on the quenched and the normalized (figure 1) 3Cr-1.5MoV steel and on the different-sized specimens of normalized 2¼Cr-2W and 2¼Cr-2WV (figure 2) are indicative of the differences between acicular and granular bainite. Micrographs of the rapidly cooled specimens [figures 1(a) and 2(a)] are characteristic of carbide-free acicular bainite [3,4]. The dark areas in the slowly cooled specimens of figures 1(b) and 2(b) are M-A islands. When granular bainite is tempered, large globular carbides form in the high-carbon M-A islands, whereas elongated carbides form on the lath boundaries of acicular bainite. These are just the morphologies observed when the 3Cr-1.5MoV steel [1] and the 15.9-mm plates of 2¼Cr-2W and 2¼Cr-2WV steel [2] were tempered.

The impact studies indicated that carbide-free acicular bainite had a high impact toughness (low DBTT and high USE) after tempering at a lower temperature or for a shorter time (constant temperature) than for granular bainite (figure 3 and Table 1) [2]. Once these toughness properties were reached at a low TP for the acicular bainite, further tempering had little additional effect. Although the quenched and the normalized steels had similar tensile properties in the as-heat-treated conditions and after similar tempering treatments [1], the observations on toughness mean that a steel consisting of acicular bainite can be optimized for strength and toughness, because it will not be necessary to temper to a low strength to achieve acceptable toughness. This optimization is being applied to develop the Cr-W steels for fusion reactor applications. A low DBTT is important for this application, since the DBTT will increase during neutron irradiation in a fusion reactor.

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REFERENCES

1. Klueh, R. L. and Nasreldin, A. M.: Met. Trans. A, 1987, 18A, 1279.
2. Klueh, R. L. and Maziasz, P. J.: Met. Trans. A, 1989, 20A, 373.
3. Habraken, L. J.: Proc. 4th Intern. Conf. Electron Microscopy, Vol 1, (Springer-Verleg, Berlin), 1966, p. 621.
4. Habraken, L. J. and Economopoulos, M.: Transformation and Hardenability in Steels, (Climax Molybdenum Co., Ann Arbor, MI), 1967, p. 69.