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FRACTURE BEHAVIOR OF BAINITIC CHROMIUM-TUNGSTEN AND CHROMIUM-MOLYBDENUM STEELS

R. L. Klueh, D. J. Alexander, and P. J. Maziasz

Metals and Ceramics Division
Oak Ridge National Laboratory
Oak Ridge, TN 37931-6376

ABSTRACT

Bainitic microstructures formed during continuous cooling of low-carbon alloy steels often appear different from classical upper and lower bainite developed by isothermal transformation. The kind of non-classical bainite produced during transformation determines the fracture behavior in a Charpy impact test. Quenching and normalizing treatments of a 3Cr-1.5Mo-0.25V-0.1C steel gave two different bainitic microstructures: a carbide-free acicular structure formed during quenching and a granular bainite formed during normalizing. The superior impact toughness of the quenched steel over the normalized steel was attributed to the difference in microstructure. A similar observation on microstructure was made for a 2.25Cr-2W-0.1C and a 2.25Cr-2W-0.25V-0.1C steel. These observations were used to develop new Cr-W steels with improved strength and impact toughness.

1. INTRODUCTION

Chromium-molybdenum bainitic steels containing 2-3% Cr and 1-1.5% Mo with small amounts of elements such as V, Ti, Nb, and B have been used extensively in the power-generation and chemical industries. The most widely used of these steels is 2 1/4Cr-1Mo, nominally Fe-2.25Cr-1Mo-0.1C (all compositions are in weight percent). In recent years, several alloy

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development programs have sought to improve the properties of these steels [1-6]. Work on one of those steels, an Fe-3Cr-1.5Mo-0.1V-0.1C (3Cr-1.5MoV) steel, demonstrated the effect of microstructure on properties [6]. Further information on that Cr-Mo steel will be presented here.

Since the late 1970s, Cr-Mo ferritic/martensitic steels have been candidate structural materials for the first wall and blanket structures of magnetic fusion power plants. Ferritic\martensitic steels are more swelling resistant during neutron irradiation than austenitic stainless steels. They also have higher thermal conductivity and lower thermal expansion coefficients than austenitic steels, so that heat flow produces lower thermal stresses. The first ferritic steels considered in the U.S. fusion reactor materials program were Sandvik HT9 (nominally Fe-12Cr-1Mo-0.25V-0.5W-0.5Ni-0.2C, here designated 12Cr-1MoVW) [7] and modified 9Cr-1Mo steel (nominally Fe-9Cr-1Mo-0.2V-0.06Nb-0.1C, designated 9Cr-1MoVNb) [8]. The 2 1/4Cr-1Mo has also been considered [9].

During the mid 1980s, fusion programs around the world began to emphasize the development of "reduced-activation" ferritic steels [10-12]. Irradiation of a fusion reactor structural steel by neutrons generated in the fusion reaction will activate (transmute to radioactive isotopes) elements of the steel. The difference between reduced-activation steels and conventional steels is that induced radioactivity in the reduced-activation steels decays much more rapidly. For that reason, the steels have been referred to as fast induced-radioactivity decay (FIRD) steels [10]. In FIRD or reduced-activation steels, alloying elements that produce long-lived radioactive isotopes during neutron irradiation are eliminated or minimized. Common alloying elements that must be eliminated or minimized include Mo, Ni, Nb, Cu, and N. In FIRD steels, molybdenum is replaced by tungsten in the conventional Cr-Mo steels to produce Cr-W steels; niobium is replaced by tantalum.

Just as the work on conventional steels for fusion emphasized the high-chromium 9Cr-1MoVNb and 12Cr-1MoVW steels [7-9], work on reduced-activation steels has concentrated on 7 to 10% Cr steels [10-12]. Initial studies at Oak Ridge National Laboratory (ORNL) were on steels with 2.25 to 12% Cr [13-15]. Of the original eight ORNL steels, an Fe-2.25Cr-2W-0.25V-0.1C (2 1/4Cr-2WV) steel had the highest strength [13]. However, the impact properties of 2 1/4Cr-2WV in a Charpy test were inferior to those of an Fe-9Cr-2W-0.25V-0.07Ta-0.1C (9Cr-2WVTa) steel [15]. The 9Cr-2WVTa had comparable tensile properties [10] and superior Charpy impact properties [15] to those of 9Cr-1MoVNb and 12Cr-1MoVW.

One problem encountered by ferritic steels for fusion applications is that neutron irradiation causes an increase in the ductile-brittle transition temperature (DBTT) and a decrease in upper-shelf energy (USE). Therefore, steels with low initial DBTTs are sought. Charpy specimens of the 9Cr-1MoVNb and 12Cr-1MoVW steels irradiated to 13 dpa (displacements per atom) at 400°C in the Experimental Breeder Reactor (EBR-II) showed increases in DBTT of 52 and 124°C, respectively [16]. The 9Cr-2WVTa irradiated to 13 dpa showed an increase in DBTT of only 15°C [17].

Despite the excellent behavior of the 9Cr-2WVTa, there are advantages for low-chromium bainitic steels [18]. It might be possible to use a low-chromium steel without a post-weld heat treatment, an important consideration for a complicated structure, such as a fusion power plant. A bainitic steel might also be used as normalized (without tempering), an economic advantage. Finally, a low-chromium steel would conserve chromium, a strategic material.

This paper will discuss work on a 3Cr-1.5MoV steel that indicates how different bainitic microstructures developed by different heat treatments can affect the fracture behavior observed in an impact test. This will be followed by observations on the development of reduced-activation Cr-W steels that show similar microstructure-dependent behavior.

2. EXPERIMENTAL PROCEDURE

Table 1 lists chemical compositions for all steels tested.

The argon-oxygen-decarburized (AOD) heat of 3Cr-1.5MoV steel was processed to 100-mm-thick plate; plates measuring 1.6 x 1.4 x 0.1 m were heat treated. After austenitizing 2 h at 955°C, one plate was water quenched and one was air cooled (normalized); both were stress relieved 2 h at 565°C. Although the as-received plates were technically quenched and stress relieved and normalized and stress relieved, for simplicity they will be referred to as quenched and normalized, respectively, since little precipitation occurred after the stress relief [6]. Plates were tempered at 663 to 704°C and from 1 to 30 h. Tempering conditions were expressed as a tempering parameter, $TP=T(20 + \log t)x10^{-3}$, where T is temperature in Kelvin and t is time in hours.

Impact properties were measured on standard Charpy V-notch specimens (10mm x 10mm x 55mm) taken from the 1/4- and 3/4-thickness depths in the plates. Specimens were taken transverse to the rolling direction, with cracks propagating in the longitudinal direction (T-L orientation). The cooling rates at these plate thicknesses were approximately 0.1 and 1.7°C/s (7 and 100°C/min) for the normalizing and quenching heat treatment, respectively.

Work on reduced-activation steels had two parts: tests on two steels from the original eight heats and on four new heats that were variations on the original steels. The 2 1/4Cr-2W and 2 1/4Cr-2WV steels were from the original eight heats of steel studied [13-15]; details on composition, processing, and microstructure have been published [12]. Small 450-g vacuum arc-melted button heats of 3Cr-2W, 3Cr-3W, 3Cr-2WV, and 3Cr-3WV steels were made using the 2 1/4Cr-2W and 2 1/4Cr-2WV as starting material. Tests were on the normalized-and-tempered steel. The 2 1/4Cr-2W, 3Cr-2W, and 3Cr-3W steels were austenitized 1 h at 900°C, and the 2 1/4Cr-2WV, 3Cr-2WV, and 3Cr-3WV were austenitized 1 h at 1050°C. The higher temperature was used for the latter steels to assure that any vanadium carbide present dissolved during austenitization. Two tempering treatments were used for all of the Cr-W steels: 1 h at 700°C and 1 h at 750°C. One-third-size Charpy specimens of the 2 1/4Cr-2W and 2 1/4Cr-2WV were heat treated directly and compared with specimens from the heat-treated 15.9-mm plate to determine the effect of heat treatment on these steels.

Tensile test specimens from the reduced-activation steels were taken from 0.76-mm-thick sheet; Charpy test specimens were from 15.9-mm-thick plate for the 2 1/4Cr-2W and 2 1/4Cr-2WV and from 6.4-mm plate for the 3Cr-2W, 3Cr-3W, 3Cr-2WV, and 3Cr-3WV. Charpy specimens were one-third the standard size; they 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 the longitudinal orientation with a transverse crack (L-T orientation).

To demonstrate the effect of cooling rate on the microstructure of 2 1/4Cr-2W and 2 1/4Cr-2WV, 10-mm-square bars and 3.3-mm-square bars of each steel were austenitized in a helium atmosphere in a tube furnace and then cooled by pulling into the cold zone. To accentuate the difference in cooling rates, the 3-mm bar was cooled in flowing helium, and the 10-mm bar was cooled in static helium.

3. RESULTS AND DISCUSSION

3.1 3Cr-1.5MoV Steel

The original studies on the 3Cr-1.5MoV steel showed that the Charpy properties of the quenched steel were less sensitive to tempering than were those of the normalized steel [6]. The smallest tempering parameter in those studies was TP=19.6 (8 h at 663°C). Smaller tempering parameters have now been applied, and the trend continued to hold. Figure 1 shows the 54 J DBTT and the USE plotted against tempering time at 663°C for the short-time tempers. A 1 h temper lowered the DBTT of the quenched plate from about 26°C in the untempered condition to -48°C [Fig 1(a)]. After the same tempering treatment for the normalized steel, the DBTT increased slightly. With further tempering, the DBTT of both the quenched and the normalized plates decreased, with the decrease more pronounced for the normalized plate. The relative effect of tempering time on the USE of the two steels was similar to the effect on DBTT [Fig. 1(b)]: there was a rapid increase in USE for the quenched steel, whereas that for the normalized steel increased more slowly with tempering time.

Figure 2 shows all of the DBTT and USE data as a function of tempering parameter. Only after the highest tempering parameter (TP=20.7, 16 h at 701°C) did the DBTT of the normalized steel approach that for the quenched steel [Fig. 2(a)]. The relative effect on the USE was somewhat similar [Fig. 2(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.

Figure 3 shows transmission electron microscopy (TEM) photomicrographs of the quenched and the normalized 3Cr-1.5MoV steel plates. Based on optical microscopy and TEM, both the quenched and air-cooled microstructures were entirely bainite, although neither microstructure was typical of upper or lower bainite. The normalized steel contained mainly regions with a high dislocation density with dark regions or "islands" scattered throughout [Fig. 3(a)]. Most of the subgrains were equiaxed. A few scattered regions contained indistinct laths, indicating a tendency toward elongated subgrains, but these areas also contained the islands, which were also often elongated.

The quenched steel had an acicular or lath structure that contained a high density of dislocations. Figure 3(b) is typical of most of the microstructure, 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.

Bainite forms when steel is transformed from austenite between the temperatures where ferrite and pearlite form (at high temperatures) and martensite forms (at low temperatures). It was originally thought to consist of only two easily distinguishable morphologies, upper and lower bainite, which are defined according to the temperature of formation. Habraken [19] demonstrated that 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 [19,20], where upper and lower bainite formed.

Habraken and Economopoulos [20] 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 (above this temperature) results in upper bainite, and transformation below it results in lower bainite.

For nonclassical bainite, Habraken and Economopoulos [20] found that a CCT diagram could be divided into three vertical regimes (Fig. 4). Different microstructures form for cooling rates that pass through these different zones. They found that a steel cooled rapidly enough to pass through zone I produced a "carbide-free acicular" structure, which consists of side-by-side plates or laths of ferrite containing a high-dislocation density [20]. For an intermediate cooling rate through zone II, a carbide-free "massive" or "granular" structure resulted, which Habraken and Economopoulos designated granular bainite [20]. They described granular bainite as consisting of equiaxed subgrains of low-carbon ferrite with a high dislocation density coexisting with dark "islands" [20]. These islands are enriched in carbon during the bainitic transformation and have been shown to be retained austenite with a high carbon concentration, part of which often transforms to martensite when cooled below M_S. These high-carbon regions are referred to as "martensite-austenite (or M-A) islands" [19-21]. The microstructures developed by still slower cooling through zone III were not observed in this study and will not be discussed.

The microstructure of the normalized 3Cr-1.5MoV steel was characteristic of granular bainite described by Habraken and others [19-21]. Granular bainite has a matrix consisting of equiaxed subgrains of ferrite with a high dislocation density coexisting with the dark high-carbon M-A islands. The high-carbon regions form when carbon is rejected by bainitic ferrite during transformation. More recent work [22] has indicated that the "equiaxed subgrains" of granular bainite described by Habraken and Economopoulos often show lath characteristics. An indication of a lath structure is seen in Fig 3(a) by the appearance of the M-A islands in a parallel array, which could indicate a lath (subgrain) boundary. Despite resistance by some researchers to the term granular bainite [22,23], that term will be used here to describe this microstructure.

Before tempering, there was little indication of precipitation in either the quenched or the normalized steel. After tempering, TEM and extraction replicas indicated the quenched steel developed carbides primarily on the lath boundaries with smaller needle-like precipitates within the matrix [Fig 5(a)]. The normalized steel also developed two types of precipitate morphologies [Fig. 5(b)]: regions containing an accumulation of globular carbides surrounded by a high density of finer needle-like precipitates. The agglomeration of the large carbides was associated with the high-carbon M-A islands [6].

3.2 Chromium-Tungsten Steels

The original studies on the Charpy impact behavior of 2 1/4Cr-2W and 2 1/4Cr-2WV were on specimens taken from normalized 15.9-mm-thick plates [15]. Under these heat-treatment conditions, the 2 1/4Cr-2WV contained about 20% polygonal ferrite and the 2 1/4Cr-2W was ≈100% bainite. When the two Cr-W steels were normalized in the 1/3-size Charpy specimen geometry (a 3.3-mm-square bar), they were 100% bainite.

Table 2 shows data when the steels were normalized as 3-mm-square bars and tempered at 700 and 750°C. Also given are results for the steels when heat treated as 15.9-mm plate (the 2 1/4Cr-2W was tempered at 700°C and the 2 1/4Cr-2WV at 750°C). The DBTT in Table 2 for the 1/3-size Charpy specimens was determined where the impact energy was one-half of the USE. Note that for these miniature specimens different DBTT and USE values apply than for standard Charpy specimens. However, other work showed that a low transition temperature for miniature specimens translates to a low value for a standard specimen [24]. Likewise, a correlation exists for the USE.

The microstructure, as reflected in the size of specimen heat treated, had a large effect on the properties of 2 1/4Cr-2WV, with much less effect on 2 1/4Cr-2W (Table 2). For the 2 1/4Cr-2W, there was little difference when normalized as a 15.9-mm plate (and tempered at 700°C) or as a 3-mm-square bar (and tempered at 700 or 750°C). This contrasts with the 2 1/4Cr-2WV steel, which had a high DBTT for the bainite plus 20% ferrite microstructure of the 15.9-mm plate, the slightly lower DBTT when the 3-mm-square bars were normalized and tempered at 700°C, and then the still lower value for the 3-mm-square bars normalized and tempered at 750°C.

This different behavior of the Charpy properties with tempering temperature between the two steels was similar to the difference observed between the quenched and normalized 3Cr-1.5MoV steel. The normalized-and-tempered microstructures also reflected this difference (Fig. 6). The tempered 2 1/4Cr-2W contained elongated carbides, indicative of a lath structure prior to tempering [Fig. 6(a)]. The 2 1/4Cr-2WV, on the other hand, contained regions with large globular carbides, indicative of a tempered granular bainite structure [Fig. 6(b)].

To show that granular bainite and acicular bainite could be developed in the 2 1/4Cr-2W and 2 1/4Cr-2WV steels, they were examined after cooling at two different rates (different-sized specimens were cooled--see Experimental Procedure section). 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 microstructures were examined by TEM, they were found to be similar to those for 3Cr-1.5MoV. The fast-cooled 2 1/4Cr-2WV steel had a lath structure [Fig. 7(a)], and the slow-cooled steel had an equiaxed structure with dark islands [Fig. 7(b)]. Similar microstructures were observed for the 2 1/4Cr-2W.

Microstructures in the quenched and the normalized 3Cr-1.5MoV steel (Fig 3) and in the different-sized specimens of normalized 2 1/4Cr-2WV (Fig. 7) and 2 1/4Cr-2W are indicative of the differences between acicular and granular bainite. Micrographs of the rapidly cooled specimens [Figs. 7(a)] are characteristic of the carbide-free acicular bainite described by

Habraken and Economopoulos [20]. The dark areas in the slowly cooled specimens [Figs. 7(b)] are M-A islands. When granular bainite is tempered, globular carbides form in the high-carbon M-A islands, whereas elongated carbides form on lath boundaries of acicular bainite, just the morphologies observed when the 3Cr-1.5MoV (Fig. 5) and the 15.9mm plates of 2 1/4Cr-2W and 2 1/4Cr-2WV steel (Fig. 6) were tempered.

3.3 Development of Chromium-Tungsten Steels for Improved Toughness

The Charpy impact studies on the 3Cr-1.5MoV, 2 1/4Cr-2W, and 2 1/4Cr-2WV indicated that the carbide-free acicular bainite formed on continuous cooling had a higher impact toughness (low DBTT and high USE) after tempering at a lower TP (at a lower temperature or for a shorter time) than did granular bainite. Once these toughness properties were reached for the acicular bainite, further tempering had little additional effect. Although the 3Cr-1.5MoV steel had similar tensile properties in the quenched and the normalized conditions and after similar tempering treatments, the observations on toughness mean that a steel consisting of acicular bainite can be optimized for both strength and toughness, because it will not be necessary to temper to a low strength to achieve acceptable toughness. To do this, it was proposed to use this optimization to develop reduced-activation Cr-W steels for fusion reactor applications, where a low DBTT is important because the DBTT will increase during neutron irradiation in a fusion reactor.

As seen in the CCT diagram in Fig. 4, acicular bainite is promoted by rapid cooling through Zone I. Rapid cooling is also necessary to avoid the ferrite transformation. Another way to avoid ferrite formation is to increase hardenability, which moves the ferrite transformation curve to longer times. Although not discussed by Habraken and Economopoulos, such a change should also cause the three cooling zones to move to longer times. If so, this should have the same effect on microstructure as increasing the cooling rate.

Alloying was chosen to improve hardenability. Carbon has a large effect on hardenability and could be used. However, carbon can adversely affect weldability, so the carbon level was maintained at 0.1%. Instead, as a first attempt to vary hardenability, further additions of Cr and W were made to the 2 1/4Cr-2W and 2 1/4Cr-2WV compositions. Alloys with 3% Cr and either 2 or 3% W were prepared (Table 1).

The 3Cr-2W and 3Cr-3W had strengths comparable to 2 1/4Cr-2W (Table 3). The strength of 3Cr-2WV was somewhat less than that of 2 1/4Cr-2WV, but the strength of 3Cr-3WV approached that of 2 1/4Cr-2WV.

Although only small changes were noted in the strength, the addition of 0.75% Cr caused a significant improvement of the impact properties of the new steels over the comparable steels containing 2 1/4% Cr (Table 4). Of special interest is the fact that the 3Cr steels have exceedingly low DBTT values even after tempering at 700°C, indicating that increasing the hardenability has improved the toughness.

Since a steel with 3% W has a higher hardenability than one with only 2% W, the higher DBTT of the 3Cr-3W steel relative to the 3Cr-2W steel was unexpected. The reason for this behavior was discovered when the microstructures of the normalized specimens were examined by TEM. The 3Cr-2W steel contained acicular bainite, as expected for a steel with increased hardenability [Fig. 8(a)]. However, a significant amount of coarse precipitate (M_6C and/or $M_{23}C_6$) formed in the 3Cr-3W steel when normalized and the laths were not too well defined [Fig 8(b)]. It is not known whether the additional tungsten induces the formation of this precipitate, which causes the lower toughness, whether the 900°C austenitization temperature is too low to dissolve all carbides for a steel with this much tungsten. This precipitate evidently does not form in the presence of vanadium when normalized at 1050°C, and TEM indicates that both the 3Cr-2WV and 3Cr-3WV contain carbide-free acicular microstructures (Fig. 9). For these steels, the impact properties of the more hardenable 3Cr-3WV steel are better than those of the 3Cr-2WV.

The most important result in Table 4 is that the steels had low DBTT values even after tempering at 700°C. This indicates that it may be possible to use these steels with a less severe temper than is necessary for most such steels. It may also be possible to use the steels without a temper. A lower tempering temperature or no temper would lead to a higher strength for the steel to go along with the good toughness.

These steels then show promise as FIRD or reduced-activation steels for fusion reactors. If it were possible to use the 3Cr-3WV steel after a 700°C temper, it would have a significant strength advantage over the 9Cr-2WVTa steel presently favored, because the latter martensitic steel would have to be tempered at 750°C or higher to obtain sufficient toughness. For the same reason, such a steel would also have an advantage over 9Cr-1MoVNb and 12Cr-1MoVW steels, the primary conventional steel candidates for fusion reactor applications.

The most widely used steel up to ≈560°C for non-nuclear power-generation industry and in the petrochemical industry is 2 1/4Cr-1Mo. However, that steel does not have strength for use at higher temperatures, where high-chromium steels (e.g., 9Cr-1MoVNb and 12Cr-1MoVW) are used. Previous work showed that 2 1/4Cr-2WV steel had strength properties from room temperature to 600°C that are comparable to those of 9Cr-1MoVNb and 12Cr-1MoVW [13], when all steels were tempered at 700 and 750°C. The strength of 3Cr-3WV approaches that of 2 1/4Cr-2WV, and it therefore has a strength comparable to the strength of the 9Cr-1MoVNb and 12Cr-1MoVW steels. Furthermore, the 3Cr steel may have a strength advantage if it does not have to be tempered at temperatures as high as those necessary for the conventional 9Cr-1MoVNb and 12Cr-1MoVW steels, where tempering temperatures of 760°C or greater are required. The lower chromium steel would also provide an economic advantage.

4. SUMMARY

Charpy impact properties of a Cr-Mo steel and several Cr-W steels were investigated and shown to depend on the bainitic microstructure developed during continuous cooling. The bainitic microstructures of these low-carbon (0.1% C) steels can differ from the classical upper and lower bainite that forms during isothermal transformation. The microstructure that forms determines the fracture behavior in a Charpy test.

Continuous cooling (normalizing or quenching) resulted in two microstructural variations of bainite, depending on cooling rate: an acicular (lath) structure and an equiaxed structure with high-carbon martensite-austenite islands, a structure termed granular bainite by previous investigators. When tempered, both contained similar matrix precipitates, but carbides precipitated on lath boundaries in the acicular structure, and large globular carbides precipitated in the high-carbon martensite-austenite islands of the granular bainite. The Charpy impact behavior of the acicular structure could be maximized with considerably less tempering than the granular bainite, thus allowing a better combination of strength and toughness.

The formation of acicular bainite or granular bainite depends on the cooling rate from the austenitization temperature, with the acicular bainite forming during the faster cooling rate. It was postulated that increasing the hardenability should have a similar effect. Several 3Cr-W steels were prepared and the properties of these steels were an improvement over the steels with only 2.25% Cr.

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Table 1 Chemical Composition of Steels (wt. percent)

Steel	C	Si	Mn	P	S	Cr	V	W	Cu	N
3Cr-1.5MoV	0.12	0.27	0.84	0.011	0.002	2.86	0.09		0.06	0.014
2 1/4Cr-2W	0.11	0.15	0.39	0.016	0.006	2.48	0.009	1.99	0.03	0.017
2 1/4Cr-2WV	0.11	0.20	0.42	0.016	0.006	2.41	0.24	1.98	0.03	0.008
3Cr-2W	0.099	0.12	0.38	0.016	0.008	3.08	0.008	2.04	0.03	0.008
3Cr-3W	0.097	0.12	0.37	0.016	0.009	3.08	0.009	3.03	0.03	0.013
3Cr-2WV	0.094	0.13	0.40	0.016	0.007	3.08	0.24	2.12	0.03	0.015
3Cr-3WV	0.095	0.12	0.38	0.016	0.009	3.09	0.25	3.02	0.04	0.012

Table 2. Charpy Impact Properties of Reduced-Activation Steels[a]

	Heat Treatment Geometry					
	3-mm-square bar				15.9-mm plate[b]	
	1 h 700°C		1 h 750°C			
Alloy	DBTT(°C)	USE(J)	DBTT(°C)	USE(J)	DBTT(°C)	USE(J)
2 1/4Cr-2W	-56	11.5	-77	10.1	-48	9.6
2 1/4Cr-2WV	-9	7.0	-52	11.0	0	9.7

[a] Tests were on 1/3-size Charpy specimens
[b] 2 1/4Cr-2W was tempered 1 h 700°C; 2 1/4Cr-2WV was tempered 1 h 750°C.

Table 3. Room Temperature Tensile Properties for FIRD Steels[a]

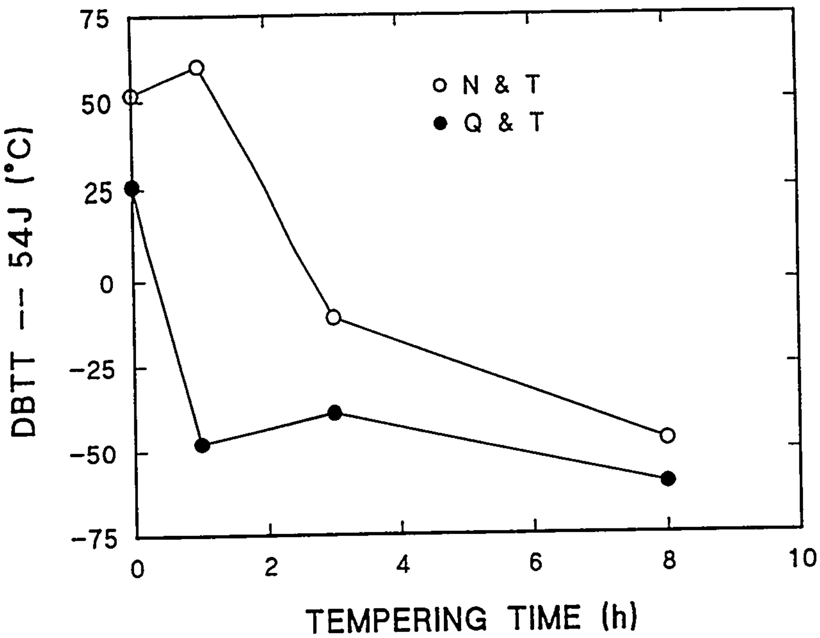
Alloy	Tempered at 700°C			Tempered at 750°C		
Designation	YS (MPa)	UTS (MPa)	El (%)	YS (MPa)	UTS (MPa)	El (%)
2.25Cr-2W	594	677	9.5	554	626	13.2
2.25Cr-2WV	889	978	7.5	684	758	9.8
3Cr-2W	592	709	10.2	520	642	11.4
3Cr-3W	606	730	9.9	505	656	11.8
3Cr-2WV	781	865	7.8	590	689	9.4
3Cr-3WV	868	953	7.8	604	710	8.8

[a] Steels were normalized and tempered.

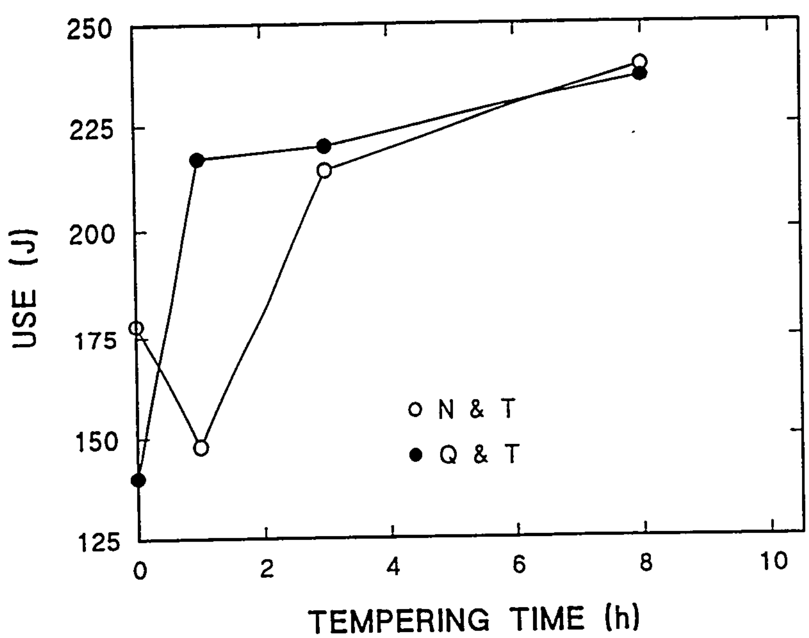
Table 4. Charpy Impact Data for FIRD Steels[a]

Alloy Designation	Tempered at 700°C DBTT (°C)	Tempered at 700°C USE (J)	Tempered at 750°C DBTT (°C)	Tempered at 750°C USE (J)
2.25Cr-2W	-56	11.5	-77	10.1
2.25Cr-2WV	10	8.4	-78	12.7
3Cr-2W	-126	11.9	-115	12.2
3Cr-3W	-65	11.0	-85	13.1
3Cr-2WV	-36	8.9	-85	14.7
3Cr-3WV	-70	10.7	-130	13.5

[a] Steels were normalized and tempered.

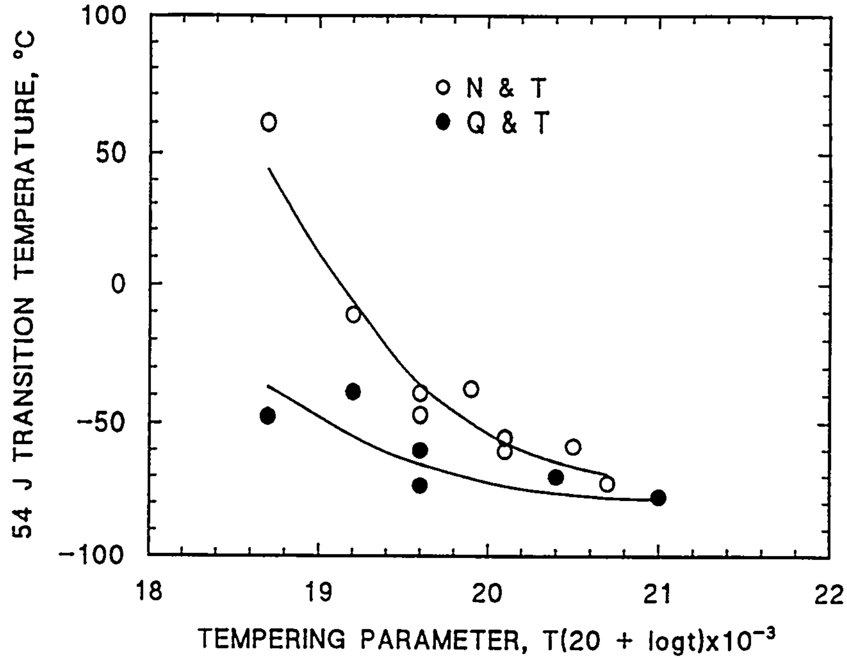


(a)

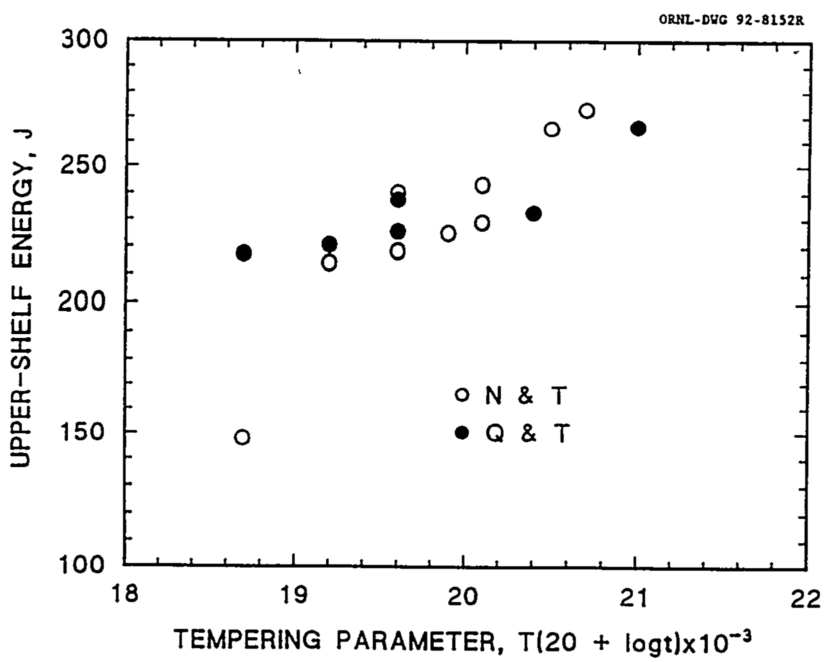


(b)

Figure 1. (a) 54-J Charpy ductile-brittle transition temperature and (b) upper-shelf energy plotted against tempering time at 663°C for normalized-and-tempered and quenched-and-tempered 3Cr-1.5MoV steel.

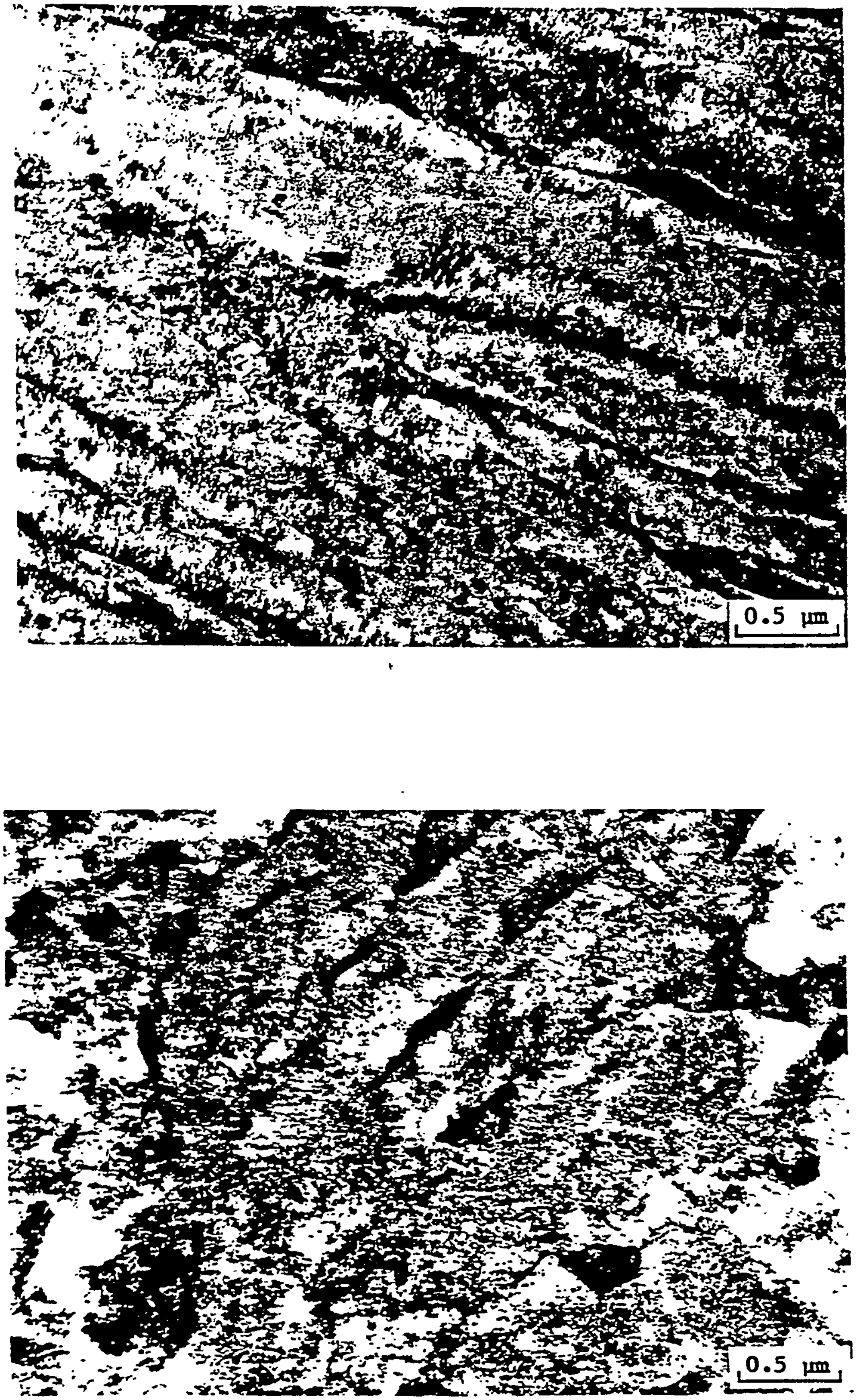


(a)



(b)

Figure 2. (a) 54-J Charpy ductile-brittle transition temperature and (b) upper-shelf energy plotted against tempering parameter for normalized-and-tempered and quenched-and-tempered 3Cr-1.5MoV steel.



(a)

(b)

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

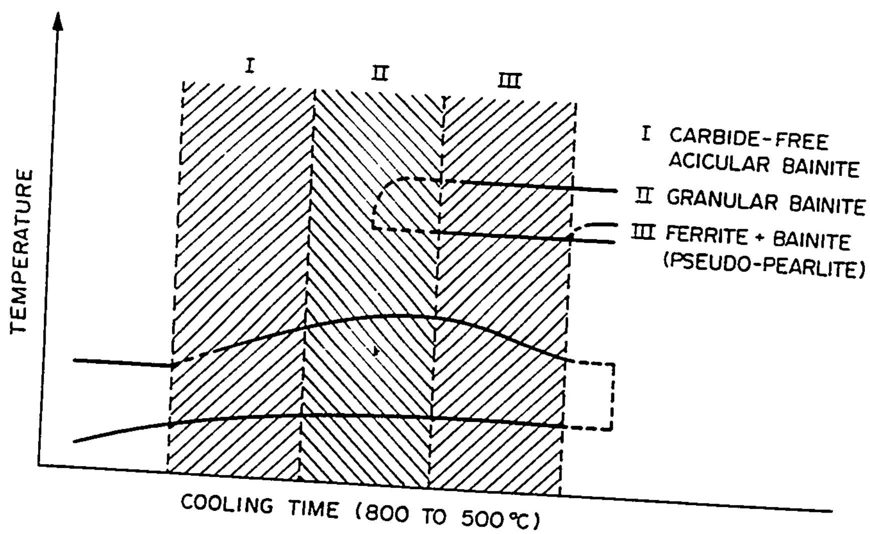
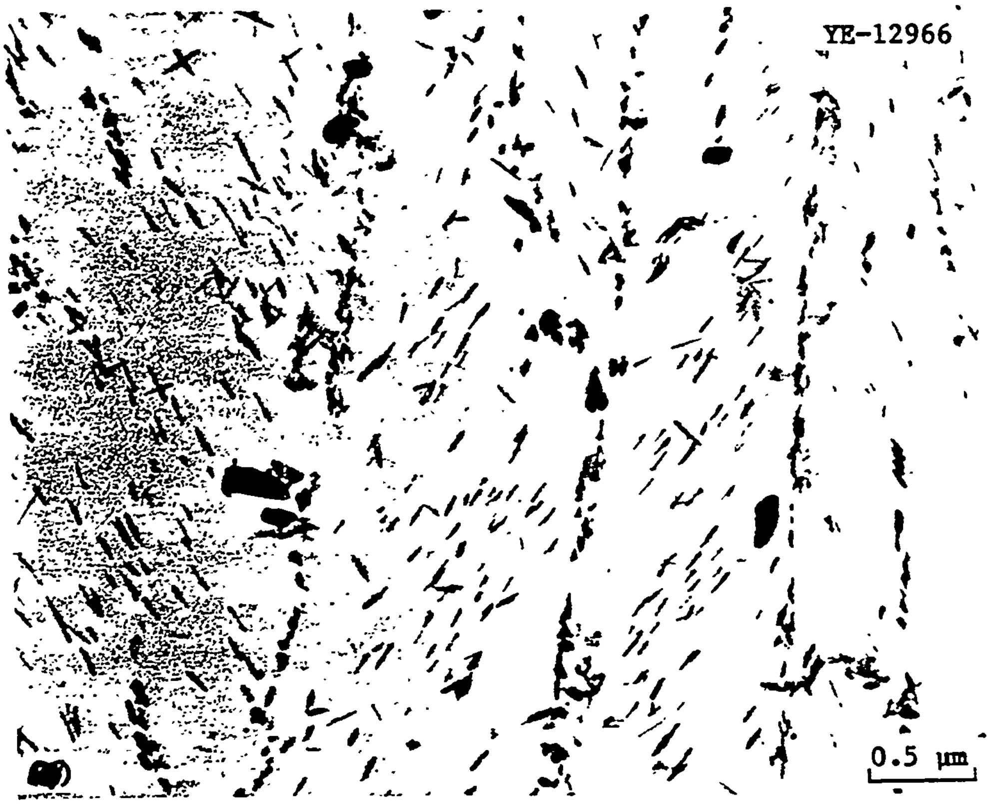


Figure 4. Schematic representation of a continuous-cooling transformation diagram for the formation of three morphological variations of bainite. Taken from reference 20.

(a)



(b)

Figure 5. Extraction replicas of 3Cr-1.5MoV steel (a) after quenching and tempering 8 h at 663°C and (b) after normalizing and tempering 8 h at 663°C.

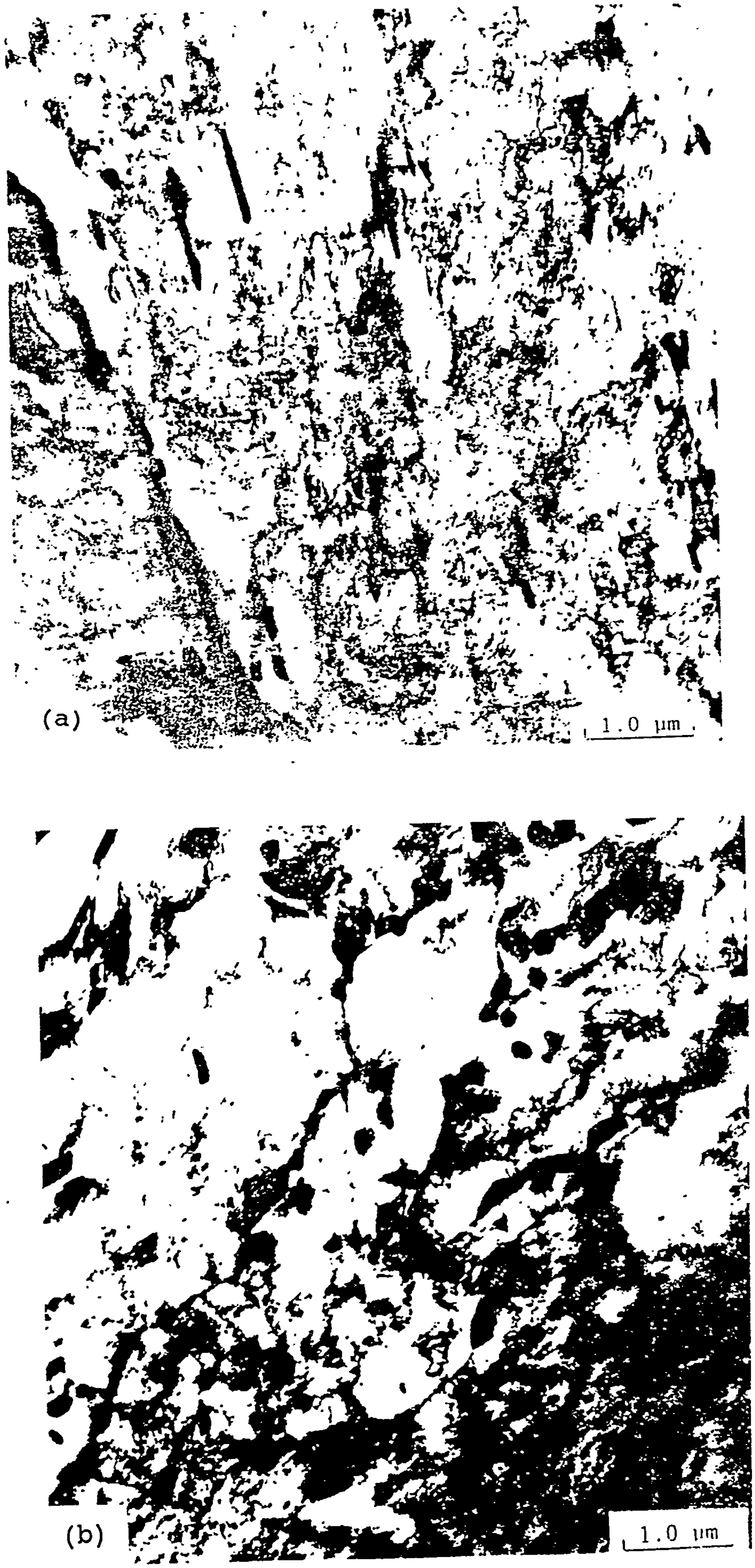
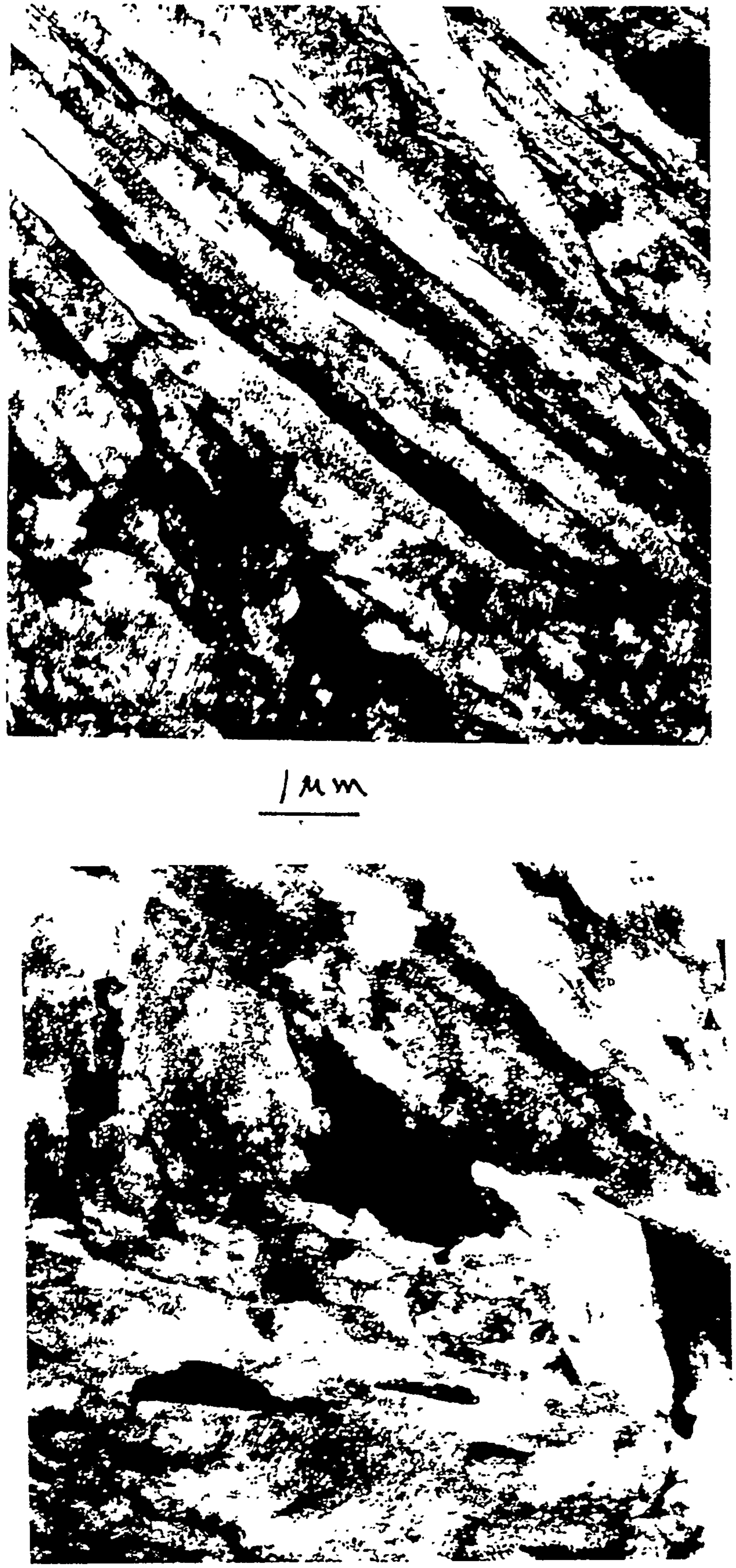


Figure 6. Normalized-and-tempered microstructures of (a) 2 1/4Cr-2W and (b) 2 1/4Cr-2WV steels.



(a)

(b)

Figure 7. Microstructures of (a) fast-cooled and (b) slow-cooled 2 1/4Cr-2WV steel.

(a)

(b)

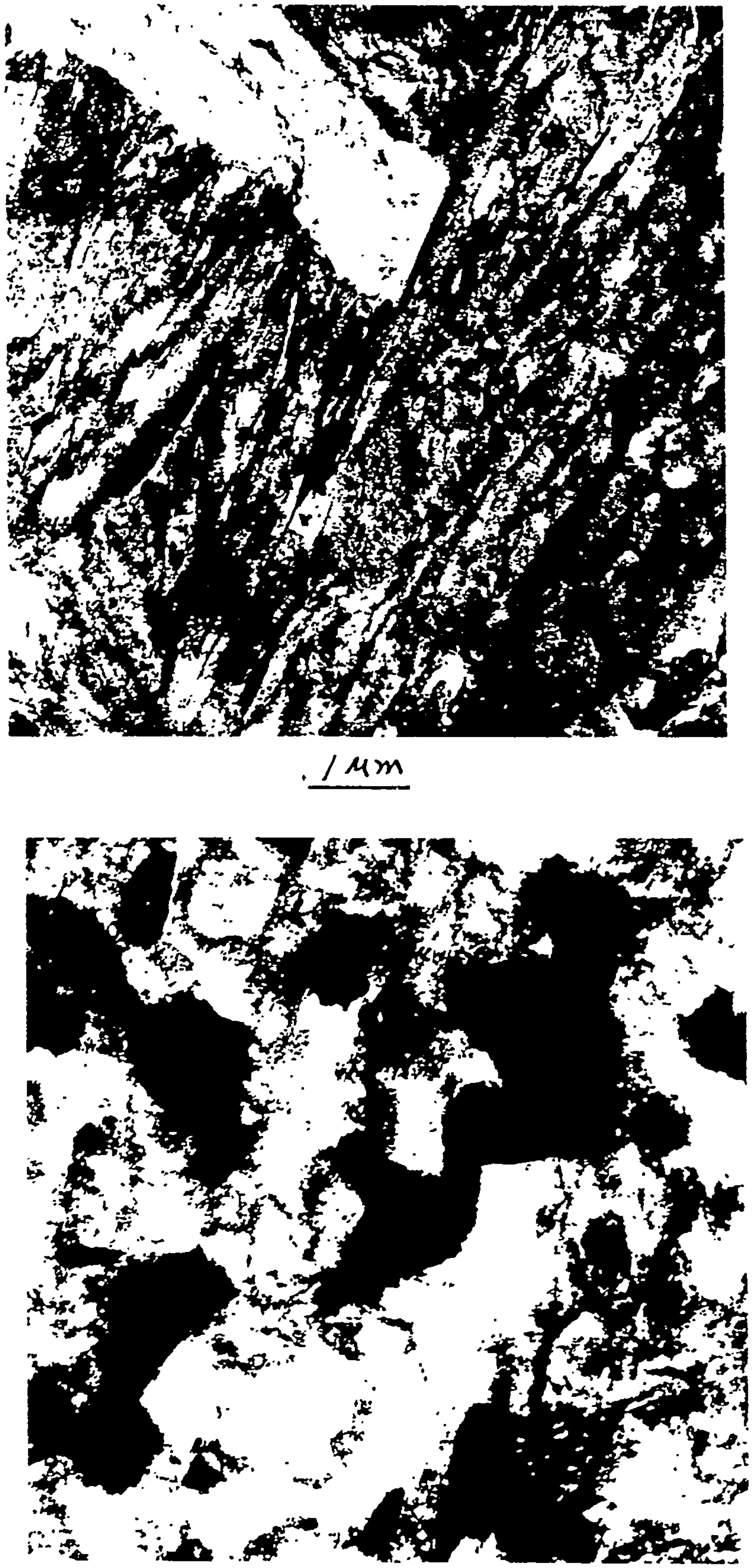
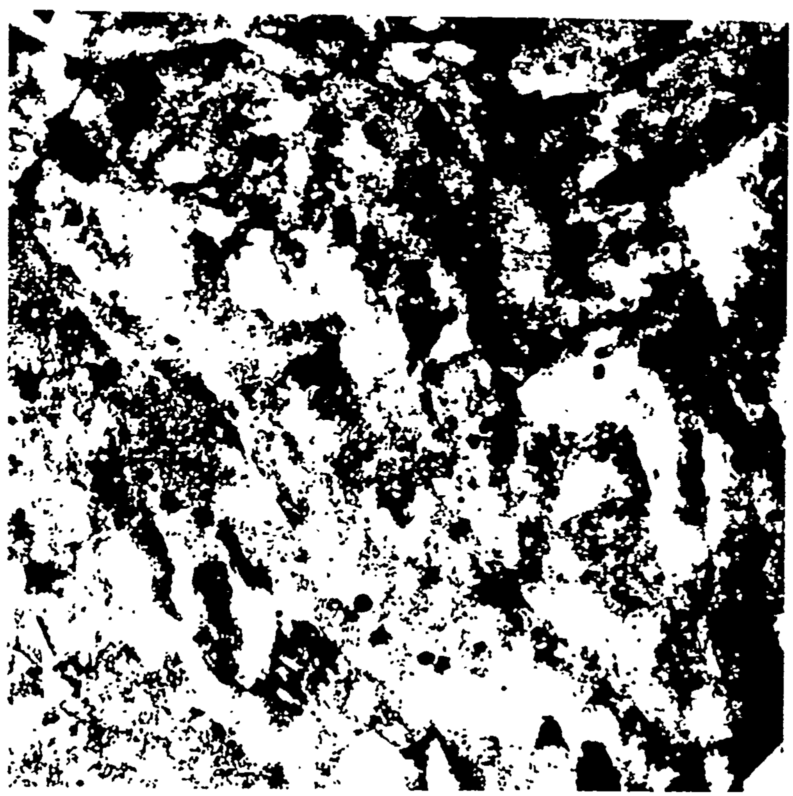


Figure 8. Transmission electron micrographs of normalized (a) 3Cr-2W and (b) 3Cr-3W steels.

(a)

1 μm



(b)

Figure 9. Transmission electron micrographs of normalized (a) 3Cr-2WV and (b) 3Cr-3WV steels.