

A COMPARISON OF LOW-CHROMIUM AND HIGH-CHROMIUM REDUCED-ACTIVATION STEELS FOR FUSION APPLICATIONS*

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INTRODUCTION

Ferritic steels have been considered candidate structural materials for first wall and blanket structures for fusion power plants since the late 1970s (1). The first steels considered in the United States were the conventional Cr-Mo steels Sandvik HT9 (nominally 12Cr-1Mo-0.25V-0.5W-0.5Ni-0.2C, here designated 12Cr-1MoVW), modified 9Cr-1Mo steel (9Cr-1Mo-0.2V-0.06Nb-0.1C, designated 9Cr-1MoVNb) and, to a lesser extent, 2 1/4Cr-1Mo steel (2.25Cr-1Mo-0.1C). All compositions are in wt. %. The normalized-and-tempered 9 and 12Cr steels had a tempered martensite microstructure, and the normalized-and-tempered 2 1/4 Cr steel had a tempered bainite microstructure.

In the early 1980s, various national fusion materials alloy development programs began to develop reduced-activation steels. A reduced-activation steel is created by eliminating or replacing those elements that produce long-lived radioactive isotopes during irradiation in a fusion first-wall/blanket structure (2). Typical alloying elements that need to be eliminated include Mo, Ni, Nb, Cu, and N. Reduced-activation steels patterned after conventional Cr-Mo steels were developed by replacing molybdenum with tungsten and niobium with tantalum (1). The initial studies at Oak Ridge National Laboratory (ORNL) were on steels with 2.25 to 12% Cr (3-5). Of the eight ORNL steels studied, a bainitic Fe-2.25Cr-2W-0.25V-0.1C (2 1/4Cr-2WV) steel had the highest strength (4). However, the impact toughness of this steel was inferior to a bainitic Fe-2.25Cr-2W-0.1C (2 1/4Cr-2W) steel and a martensitic Fe-9Cr-2W-0.25V-0.09Ta-0.1C (9Cr-2WVTA) steel, which had tensile and Charpy impact properties comparable to or better than the 9Cr-1MoVNb and 12Cr-1MoVW steels (4).

Neutron irradiation of the type encountered in a fusion power plant causes severe deterioration of impact toughness, as shown by an increase in the ductile-brittle transition temperature (DBTT) and a decrease in the upper-shelf energy (USE) in a Charpy test. The 9Cr-2WVTA steel developed one of the smallest shifts in DBTT (Δ DBTT) ever observed for this type of steel. After irradiation in the Fast Flux Test Facility (FFTF) to \approx 28 dpa, the 9Cr-2WVTA had a Δ DBTT of 32°C, compared to 151°C for 12Cr-1MoVW irradiated to 10 dpa. In addition to the lower Δ DBTT, the unirradiated DBTT of the 9Cr-2WVTA steel was -88°C, compared to -46°C for the 12Cr-1MoVW steel. However, when 2 1/4Cr-2WV steel was irradiated to \approx 28 dpa, it exhibited a Δ DBTT of 145°C.

Although relatively modest shifts in DBTT occur for 9Cr-1MoVNb and 12Cr-1MoVW steels irradiated in FFTF, a fast reactor, much larger shifts (>200 °C) occurred when irradiated in the High Flux Isotope Reactor (HFIR), a mixed spectrum reactor, where displacement damage was accompanied by transmutation-induced helium production (6). The difference was attributed to the higher helium concentration in the steel produced during irradiation in HFIR. These results are important because the high-energy neutrons from the fusion reaction will also produce large amounts of transmutation helium combined with displacement damage in the first wall of a fusion power plant.

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EXPERIMENTAL PROCEDURES

Details on the production of the experimental heats of the new reduced-activation steels, their chemical compositions, and heat treatment procedures have been published (3-5,11,12). All of the steels were tested in the normalized-and-tempered condition. The steels were normalized by austenitizing at 1050°C and then rapidly cooling in air or an inert gas (5,12). Tempering treatments will be discussed as required. Miniature tensile (25.4-mm long with a gage length 7.62 x 1.52 x 0.76 mm) and Charpy (3.3 x 3.3 x 25.4 mm with a 0.51-mm-deep 30° V-notch and a 0.05- 0.08-mm-root radius) specimens were tested; details of the test procedures have been published (11,12).

RESULTS AND DISCUSSION

Charpy properties of the bainitic 2 1/4Cr-2W and 2 1/4Cr-2WV steels and the martensitic 9Cr-2WVTa steel from the original eight heats of reduced-activation steel (3-5) were determined on specimens taken from normalized-and-tempered 15.9-mm plate. The 2 1/4Cr-2WV contained \approx 20% polygonal ferrite and 80% bainite, the 2 1/4Cr-2W contained 100% bainite, while the 9Cr-2WVTa contained 100% martensite (3). It was concluded that the inferior impact toughness of the 2 1/4Cr-2WV was caused by the ferrite in the otherwise bainitic microstructure (5). Charpy properties of 2 1/4Cr-2WV improved after a heat treatment to produce 100% bainite (13), but properties were still inferior to those of the 2 1/4Cr-2W and the 9Cr-2WVTa steels. Transmission electron microscopy examination indicated a difference in the tempered bainitic microstructures of the 2 1/4Cr-2W and 2 1/4Cr-2WV (13). Elongated precipitates and substructure were observed in the 2 1/4Cr-2W [Fig. 2(a)], suggesting that a lath microstructure was present prior to tempering. Large globular precipitates appeared in patches in the 2 1/4Cr-2WV [Fig. 2 (b)]. The reason for these microstructural differences was attributed to the different kinds of bainite that formed in the two steels when they were normalized (13). Based on work by Habraken and Economopoulos (14), the difference was attributed to how rapidly the steels were cooled from the austenitizing temperature.

To demonstrate the effect of cooling rate, pieces of 10 x 10 mm and 3.3 x 3.3 mm bars of 2 1/4Cr-2W and 2 1/4Cr-2WV were austenitized in a helium atmosphere in a tube furnace and cooled by

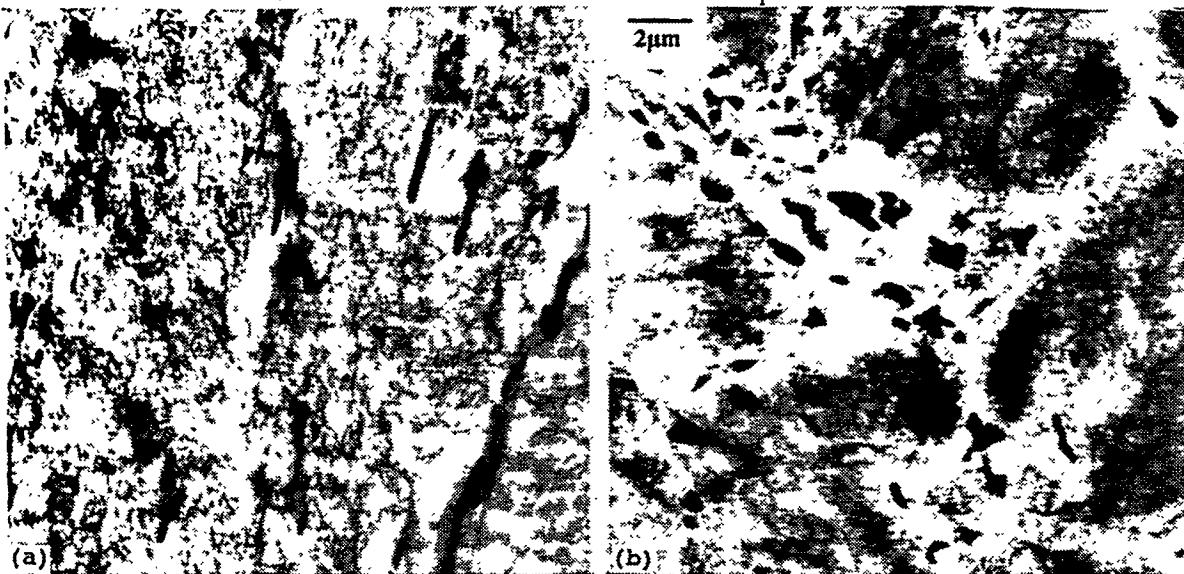


Fig. 2 Microstructure of normalized-and-tempered (a) 2 1/4Cr-2W and (b) 2 1/4Cr-2WV steels.

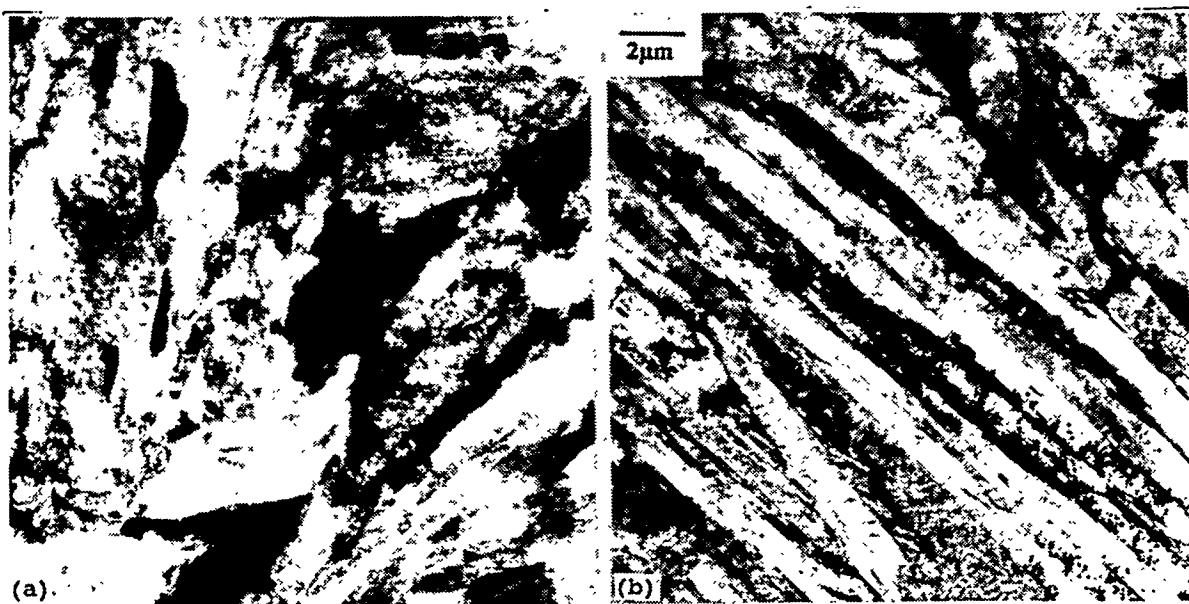


Fig. 3 Microstructure of normalized 2 1/4Cr-2WV after (a) slow cool and (b) fast cool.

pulling into the cold zone (13). To increase the cooling rate of the small specimens, flowing helium was used for cooling, whereas the large specimens were cooled in static helium to further reduce the cooling rate. The microstructure after the slow cool [Fig. 3(a)] had an equiaxed or granular structure, compared to an acicular structure after the fast cool [Fig. 3(b)]. Figure 3 shows the results for the 2 1/4Cr-2WV steel, but similar observations were made for the 2 1/4Cr-2W steel.

Bainite is generally defined as a ferrite matrix containing carbides that transformed from austenite in the temperature range ≈ 250 - 550°C . There were originally thought to be only two morphological variations—upper and lower bainite—which were defined according to the temperature of formation and differentiated by the appearance of the carbide particles relative to the axis of the bainite plate or needle. Upper bainite forms as a collection of ferrite plates or laths, with the carbide particles arranged parallel to the plates. Lower bainite consists of ferrite plates or needles with carbides that form within the ferrite at about a 60° angle to the axis of the plate or needle.

Habraken and Economopoulos (14) found morphological variations in the bainite transformation products that differed from upper and lower bainite, although they still formed in the bainite transformation temperature regime. Such "nonclassical" bainites formed more readily during continuous cooling than during an isothermal transformation, which favors the classical bainite. Classical upper and lower bainite form when the steel is transformed in different temperature regimes of the bainite transformation temperature region, as defined on an isothermal-transformation (IT) diagram. The bainite transformation region of an IT diagram can be divided into two temperature regimes by a horizontal line, above which upper bainite forms and below which lower bainite forms. For the non-classical bainites, Habraken and Economopoulos (14) showed that a continuous cooling transformation (CCT) diagram could be divided into three vertical regimes (Fig. 4). A steel cooled through zone I produces a "carbide-free acicular" structure, consisting of side-by-side plates or laths (14). When cooled through zone II, a carbide-free "massive or granular" structure results, generally referred to as granular bainite. It has been determined that granular bainite consists of a bainitic ferrite matrix with a high dislocation density that contains martensite-austenite (M-A) "islands" (14). The M-A islands are enriched in carbon during the formation of the bainite. The microstructures formed in zone III were not observed in this study and will not be discussed.

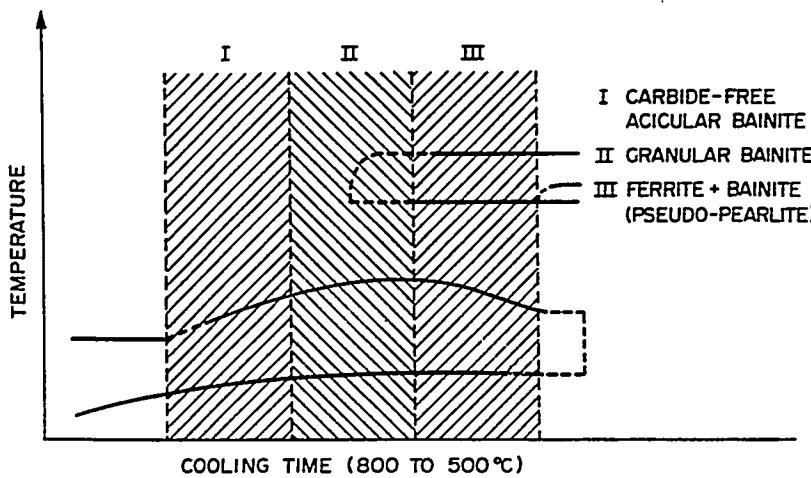


Fig. 4 CCT diagram with cooling zones of Habraken and Economopoulos.

Work on a 3Cr-1.5Mo-0.25V-0.1C steel indicated that tempered carbide-free acicular bainite had better impact properties (low DBTT and high USE) than granular bainite after a low-temperature or short-time (constant temperature) tempering treatment (15). Furthermore, optimum strength and toughness were obtained for the carbide-free acicular bainite under modest tempering conditions, and further tempering had little additional effect on toughness. Considerably more tempering (longer time or higher temperature) was required for similar properties optimization in the granular bainite, which meant optimum impact toughness was only achieved at the expense of lower strength (15).

No CCT diagrams are available for the reduced-activation steels under discussion. Figure 4 is schematic, but even when CCT diagrams are available, details on the three cooling zones must be experimentally determined. However, if the understanding of the effects of cooling rate on microstructure are correct, it would appear that the toughness of the 2 1/4Cr-2WV could be improved by cooling still more rapidly, perhaps by quenching instead of normalizing (air cooling). Another possibility to promote carbide-free acicular bainite is by improving the hardenability (11). Hardenability is defined as the relative ability of a steel to avoid forming the soft ferrite phase when it is cooled from the austenitizing temperature. Increasing hardenability has the same relative effect as increasing the cooling rate: it moves the transformation of ferrite to longer times so the steel can be cooled more slowly and still obtain bainite. This should also move the zone for the formation of acicular bainite to longer times and allow it to form at slower cooling rates.

Hardenability can be altered by changing the chemical composition. As a first attempt to vary hardenability, more chromium and small amounts of boron were added to the 2 1/4Cr-2WV composition (11). Boron was chosen because small amounts increase hardenability (16). The chromium concentration was increased to 2.6% and $\approx 0.005\%$ B was added, which resulted in a marked improvement in the Charpy properties (11). However, there were still indications of granular bainite in the microstructure of an Fe-2.6Cr-0.25V-0.004B-0.1C steel. Therefore, it was decided to add still more chromium and to also add additional tungsten.

The microstructure of the slowly cooled 2 1/4Cr-2W and 2 1/4Cr-2WV [Fig. 3(a)] was characteristic of granular bainite; the dark areas are the M-A islands. The microstructure of the specimen cooled rapidly [Fig. 3(b)] was characteristic of carbide-free acicular bainite (14). When granular bainite is tempered, large globular carbides form in the high-carbon M-A islands, whereas elongated carbides form on lath boundaries of acicular bainite, just the types of morphology formed when the 2 1/4Cr-2WV [Fig. 2(a)] and 2 1/4Cr-2W [Fig. 2(b)] steels were normalized and tempered.

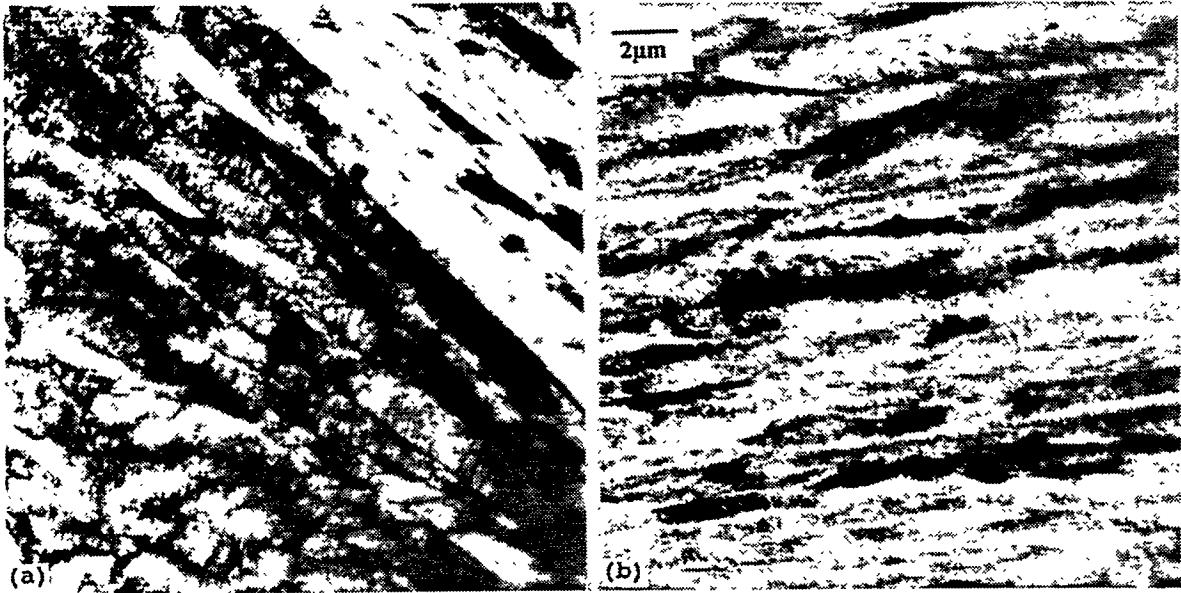


Fig. 5 Microstructure of normalized (a) 3Cr-2WV and (b) 3Cr-3WV steels.

Steels with 3% Cr were produced with the following nominal compositions: Fe-3Cr-2W-0.1C (3Cr-2W), Fe-3Cr-3W-0.1C (3Cr-3W), Fe-3Cr-2W-0.25V-0.1C (3Cr-2WV), and Fe-3Cr-3W-0.25V-0.1C (3Cr-3WV) (12). Since the 3Cr-2WV and 3Cr-3WV had the best properties, only those two steels will be discussed here. The normalized 3Cr-2WV [Fig. 5(a)] and 3Cr-3WV [Fig. 5(b)] steels contained carbide-free acicular microstructures, with the lath size decreasing with increasing tungsten. Thus, the combination of additional chromium and tungsten produced the preferred acicular bainite structure. The 3Cr-2WV and 3Cr-3WV steels contrasted with the 2 1/4Cr-2WV steel, which had a granular bainite structure under similar heat treatment conditions (12).

The room-temperature yield stress for the two bainitic 3Cr-WV steels were compared with the martensitic 9Cr-2WVTa, 9Cr-1MoVNb, and 12Cr-1MoVW steels, the latter two being the conventional steels considered for fusion [Fig. 6 (a)]. The 3Cr-2WV, 3Cr-3WV, and 9Cr-2WVTa were tested after normalizing and then tempering 1 h at 700°C and 1 h 750°C. The 9Cr-1MoVNb and 12Cr-1MoVW steels were tempered at 750°C but not at 700°C, because to produce sufficient toughness for most applications, the standard temper for 9Cr-1MoVNb is 1 h at 760°C and for 12Cr-1MoVW is 2.5 h at 780°C. After tempering all the steels at 750°C, 9Cr-2WVTa, 9Cr-1MoVNb and 12Cr-1MoVW steels were slightly stronger than the 3Cr-2WV and 3Cr-3WV steels. The 3Cr-WV steels were comparable to or slightly stronger than the 9Cr-2WVTa when all three were tempered at 700°C. When the Cr-Mo steels in their standard heat-treatments were compared with the 3Cr-WV steels [Fig. 6 (a)], the strength of the 3Cr-WV steels after tempering at 750°C were greater than those of the Cr-Mo steels. As discussed below, it should be possible to use the 3Cr-WV steels with either a 700°C temper or with no temper at all, which would give the 3Cr-WV steels a distinct strength advantage. Tensile properties were determined from room temperature to 600°C (12), and the yield stress of the 3Cr-WV steels approached those of 9Cr-2WVTa at 500-600°C. Previous work indicated that over the range of room temperature to 600°C, the 9Cr-1MoVNb and 12Cr-1MoVW tempered at 750°C were similar to those for 9Cr-2WVTa tempered at 750°C (4). Therefore, the strength of the 3Cr-WV steels compared favorably to the conventional steels as well as the 9Cr-2WVTa. The ductility of the 3Cr-WV steels was better than that of the high-chromium steels (12).

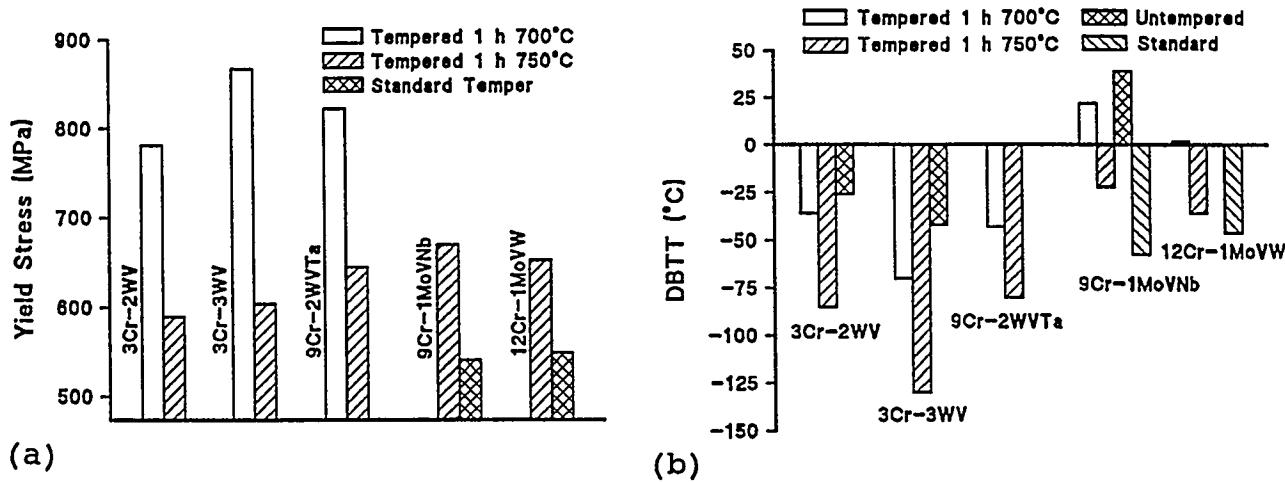


Fig. 6 Comparison of (a) room temperature yield stress and (b) DBTT values of steels.

Although the addition of 0.75% Cr to the 2 1/4Cr-2WV composition caused only minor changes in strength, there was a significant improvement of the impact properties of the new steels over the original reduced-activation steels containing 2 1/4% Cr (12). The 3Cr-2WV and 3Cr-3WV were improvements over 9Cr-2WVTa when tempered at the same temperature [Fig. 6 (b)]. Of special interest is that the 3Cr-WV steels have exceedingly low DBTTs even after tempering at 700°C, indicating that increasing the hardenability significantly improved the impact toughness. The 3Cr-3WV steel had a DBTT of -70°C after the 700°C temper, compared with -88°C for the 9Cr-2WVTa steel after a 750°C temper. This indicates that it may be possible to use these steels with less tempering than is necessary for a high-chromium steel, and it presents the possibility of using them as normalized (untempered). Although the room-temperature strengths of the 9Cr-1MoVNb and 12Cr-1MoVW steels after tempering at 750°C slightly exceeded those of the 3Cr-2WV and 3Cr-3WV steels tempered at 750°C [Fig. 6 (a)], the Charpy impact behavior of the 3Cr-WV steels [Fig. 6(b)] was superior to that of the Cr-Mo steels after tempering at 750°C. Furthermore, the DBTT and USE of the 3Cr-WV steels after tempering at 700°C were as good or better than those of the Cr-Mo steels tempered at 750°C (12). For this latter comparison, the strength of the 3Cr-WV steels was also superior to that of the Cr-Mo steels. The USE and DBTT of the 3Cr-3WV tempered at 700°C were comparable to 9Cr-2WVTa tempered at 750°C (12).

When the 3Cr-WV steels were tested in the normalized condition, DBTT values well below room temperature were obtained [Fig. 6(b)]. The 3Cr-3WV steel again had the best properties. The use of a lower tempering temperature or eliminating the need to temper would lead to a higher strength to go along with the good toughness. Even in the untempered condition, the DBTT [Fig. 6 (b)] and USE (12) of the 3Cr-WV steels compared favorably with the Cr-Mo steels and with the 9Cr-2WVTa steel that have been toughened by tempering at 750°C [Fig. 6(b)], further indicating that the new steels may be practical for use in the untempered condition for some applications. Note in Fig. 6(b) the high DBTT of the untempered martensitic 9Cr-1MoVNb steel. This is typical of untempered martensitic Cr-Mo or Cr-W steels. Even when the 9Cr-1MoVNb and 12Cr-1MoVW steels were given their standard tempering treatments to further improve toughness, the DBTT of the 3Cr-3WV in the untempered condition is comparable (12). Likewise, the USE values of the 3Cr-2WV and 3Cr-3WV steels in the untempered condition compare favorably with the 9Cr-2WVTa after a 750°C temper and with the 9Cr-1MoVNb and 12Cr-1MoVW after a standard temper (12).

The mechanical property results on the 3Cr-WV steels demonstrated that composition modifications to increase hardenability promoted an acicular bainite microstructure with improved strength

and toughness. In this work, alloying additions were limited to chromium and tungsten. However, it should be possible to further improve the mechanical properties of the 3Cr-WV steels. Tantalum improved the impact toughness of the 9Cr steels (5), and tantalum and boron improved the toughness of 2.25Cr steels (11). Work is in progress to explore the effect of tantalum, boron, and other elements added to the 3Cr-2WV and 3Cr-3WV compositions to further improve the properties.

A major disadvantage of ferritic steels for fusion is the need for a PWHT that must be carried out in the field during construction. Based on the poor impact properties of untempered martensite, a PWHT for the high-chromium steels cannot be eliminated. Preliminary work on these new low-chromium steels indicates that they have excellent impact toughness without tempering, but further work is required to determine whether the PWHT can be eliminated.

SUMMARY

At present, the high-chromium, martensitic reduced-activation ferritic steels are favored as possible structural materials for fusion applications. In particular, a 9Cr-2WVTA steel has excellent strength and toughness, and the properties show little change during neutron irradiation. Despite this, low-chromium, bainitic steels may offer advantages. Preliminary work has shown that steels with 3% Cr, 2-3% W, 0.25% V, and 0.1% C have tensile and Charpy properties as good or better than those of the 9Cr-2WVTA steel prior to irradiation, and these properties are developed with less tempering than is required for the high-chromium steels. Performance of the new low-chromium steels during neutron irradiation still needs to be evaluated.

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