NEUTRON IRRADIATION EFFECTS ON THE DUCTILE-BRITTLE TRANSITION

OF FERRITIC/ MARTENSITIC STEELS*

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Abstract

Ferritic/martensitic steels such as the conventional 9Cr-1MoVNb (Fe-9Cr-1Mo-0.25V-0.06Nb-0.1C) and 12Cr-1MoVW (Fe-12Cr-1Mo-0.25V-0.5W-0.5Ni-0.2C) steels have been considered potential structural materials for future fusion power plants. The major obstacle to their use is embrittlement caused by neutron irradiation. Observations on this irradiation embrittlement will be reviewed. Below 425-450°C, neutron irradiation hardens the steels. Hardening reduces ductility, but the major effect is an increase in the ductile-brittle transition temperature (DBTT) and a decrease in the upper-shelf energy, as measured by a Charpy impact test. After irradiation, DBTT values can increase to well above room temperature, thus increasing the chances of brittle rather than ductile fracture. In addition to irradiation hardening, neutrons from the fusion reaction will produce large amounts of helium in the steels used to construct fusion power plant components. Tests to simulate the fusion environment indicate that helium can also affect the toughness. Steels are being developed for fusion applications that have a low DBTT prior to irradiation and then show only a small shift after irradiation. A martensitic 9Cr-2WVTa (nominally Fe- 9Cr-2W-0.25V-0.07Ta-0.1C) steel had a much lower DBTT than the conventional 9Cr-1MoVNb steel prior to neutron irradiation and showed a much smaller increase in DBTT after irradiation.

DISCLAIMER

Introduction

Ferritic/martensitic steels are used in nuclear fission reactors, such as the light water reactors that generate electricity, fast reactors for power and research, and test reactors. They are also considered as structural materials for fusion power plants planned for the next century. Irradiation by the neutrons generated in these devices affects the viability of the steels for structural applications. This paper will review the effect of neutron irradiation on the steels being considered for fusion applications.

Ferritic/martensitic steels are considered for fusion applications because they are more swelling resistant during neutron irradiation than austenitic stainless steels [1]. 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 the commercial Sandvik HT9 (nominally Fe-12Cr-1Mo-0.25V-0.5W-0.5Ni-0.2C, here designated 12Cr-1MoVW) and modified 9Cr-1Mo steel (nominally Fe-9Cr-1Mo-0.2V-0.06Nb-0.1C, designated 9Cr-1MoVNb) [1]. These steels are also used in the power-generation and petrochemical industries.

During the mid 1980s, fusion programs around the world began to emphasize the development of "reduced-activation" ferritic steels [2-4]. 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 more rapidly, thus simplifying the disposal of the radioactive structure after its service lifetime. In 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 in reduced-activation steels include Mo, Ni, Nb, Cu, and N. Molybdenum is replaced by tungsten in the conventional Cr-Mo steels to produce Cr-W steels; niobium is replaced by tantalum.

The major problem faced by ferritic/martensitic steels for fusion applications is the embrittlement that occurs during irradiation. Irradiation causes an increase in the ductile-brittle transition temperature (DBTT) and a decrease in the upper-shelf energy (USE) as determined in a Charpy impact test. This phenomenon will be the primary subject of this paper.

Background

When a steel is irradiated by high-energy neutrons, interstitials and vacancies are formed when neutrons displace atoms from lattice positions into interstitial positions. It is the disposition of the irradiation-produced interstitials and vacancies that determines the effect of irradiation on properties. This "displacement damage" depends on neutron fluence and is described in terms of the average number of times an atom is displaced from its lattice position as displacements per atom (dpa). For a steel irradiated in a fast reactor such as the Fast Flux Test Facility (FFTF) or the Experimental Breeder Reactor (EBR-II), each atom can be displaced over 30 times in a year. At reactor temperatures (300-600°C), interstitials and vacancies are mobile, and most are eliminated by one-to-one recombination and have no effect on properties.

Defects that do not recombine either form clusters or migrate to "sinks"—surfaces, grain boundaries, dislocations, and existing cavities—where they are absorbed. Physical and mechanical properties are affected by the clusters. Clusters of interstitials can evolve into dislocation loops; vacancy clusters in combination with dissolved or transmutation gases can develop into microvoids

or cavities. The type of cluster that forms depends on irradiation temperature. In irradiated ferritic/martensitic steels, interstitials are mobile even at room temperature, and they can combine to form dislocation loops. These loops cause an increase in strength and a decrease in ductility. Vacancies become increasingly mobile above $\approx 200°C$, which can lead to cavity formation and an increase in volume (swelling). Cavity formation occurs because dislocations have a bias for interstitials, and thus, vacancies are absorbed by existing cavities. If all sinks accepted vacancies and interstitials equally, they would annihilate at a sink, and no swelling would result. One reason martensitic steels are favored for fusion is that they show low swelling. At irradiation temperatures above $\approx 425°C$, dislocation loops are unstable because interstitial cluster formation is inhibited by high equilibrium vacancy concentrations and rapid diffusion that enhance vacancy-interstitial annihilation. At these higher temperatures, displacement damage has little effect on properties.

Besides causing displacement damage, an atom of the steel can absorb a neutron and undergo a transmutation reaction to produce new metal atoms and gaseous hydrogen and helium inside the steel. New radioactive metal atoms that take a long time for the radioactivity to decay were the impetus for developing reduced-activation steels. New metal atoms have little effect on properties, because relatively small quantities form, and the characteristics of the new atoms are generally similar to the atoms they replace. Hydrogen has generally been assumed to have little effect, because it will diffuse from the alloy at operating temperatures. More work is required to verify this. Helium, however, is insoluble in metals and alloys and will be incorporated into bubbles or voids that form within the metal, and it can affect the mechanical properties.

<div align="center">Experimental Procedure</div>

Neutron Irradiation

One difficulty in developing materials for fusion is that no fusion reactors are available to test materials. Therefore, neutron irradiation effects expected to occur in a fusion reactor must be simulated in fission reactors. Displacement damage formed in fusion and fast fission reactors is similar, and fast reactors, such as the FFTF and the EBR-II, have been used to study the effect of displacement damage. However, neutrons with much higher energy (up to 14 MeV) are produced by the fusion reaction than in a fast fission reactor, and they will produce more transmutation helium within material irradiated in a fusion reactor than a fast fission reactor. The simultaneous development of displacement damage and transmutation helium in a fusion power plant could affect both the swelling behavior and the mechanical properties relative to the formation of displacement damage alone. Therefore simulation techniques are required to study helium effects.

One method to simulate helium effects in martensitic steels is to irradiate them in a mixed-spectrum reactor, such as the High Flux Isotope Reactor (HFIR) [5]. Both fast and thermal neutrons are generated in a mixed-spectrum reactor. Displacement damage is produced by the fast neutrons, and transmutation helium is produced when the ^{58}Ni in the steel (present in the composition or is added) undergoes a two-step reaction with thermal neutrons to produce an α-particle—a helium atom. For a steel containing 2% Ni irradiated in the HFIR, the same amount of helium (in appm) per atom displaced (He/dpa ratio) occurs as is expected in a Tokamak fusion plant [5].

Test Procedure

Typical compositions of the 9Cr and 12Cr steels to be discussed are given in Table 1. Details on the production of the experimental heats of the new reduced-activation steels, their chemical compositions, and heat treatment procedures have been published [6]. 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. Tempering was 1 h at 760°C for the 9Cr-1MoVNb steel, 2.5 h at 780°C for the 12Cr-1MoVW steel, and 1 h at 750°C for the 9Cr-2WV and 9Cr-2WVTa steels. Miniature tensile (44.5 mm long with a gage length 7.62 x 1.52 x 0.76 mm) and miniature Charpy (⅓ size: 3.3 x 3.3 x 25.4 mm or ½-size: 5 x 5 x 25.4mm) specimens were tested. Details of the test procedures have been published [6,7].

Table I. Chemical Composition of Steels (wt. %)

Element	9Cr-1MoVNb	12Cr-1MoVW	9Cr-2WV	9Cr-2WVTa
C	0.092	0.20	0.12	0.1
Si	0.15	0.17	0.25	0.23
Mn	0.48	0.57	0.51	0.43
P	0.012	0.016	0.14	0.15
S	0.004	0.003	0.005	0.005
Cr	8.32	12.1	8.73	8.72
Mo	0.86	1.04		
Ni	0.09	0.51		
V	0.20	0.28	0.25	0.23
Nb	0.06			
W	0.01	0.45	2.09	2.09
Ta				0.075
N	0.055	0.027		

Results and Discussion

Displacement Damage Effects on Tensile Behavior

The 9Cr-1MoVNb steel tensile specimens were irradiated in EBR-II to ≈11 dpa at 390, 450, 500, or 550°C and tested at the irradiation temperature [8]. Both the 0.2% yield stress (YS) and the ultimate tensile strength (UTS) increased for specimens irradiated at 390°C, but little change occurred at the other three temperatures (Fig. 1). Thermal aging for times equivalent to those in the reactor had little effect over the entire temperature range [8]. Ductility changes reflected the effect on strength: the uniform and total elongations at 390°C were slightly less than those of the unaged and aged controls with little effect at the highest temperatures [8]. Similar effects were observed for 12Cr-1MoVW steel irradiated to 13 dpa in EBR-II [9]. Hardening occurred at 390°C, with essentially no change after irradiation at 450, 500, and 550°C. Both steels were also irradiated in EBR-II to 23 to 25 dpa [10]. At 390°C there was little change relative to the steels irradiated to 11 to 13 dpa, an indication that the hardening saturated by a fluence of ≈10 dpa. After irradiation at 450, 500, or 550°C, there was also little difference in the tensile properties of the irradiated and the normalized-and-tempered specimens [10].
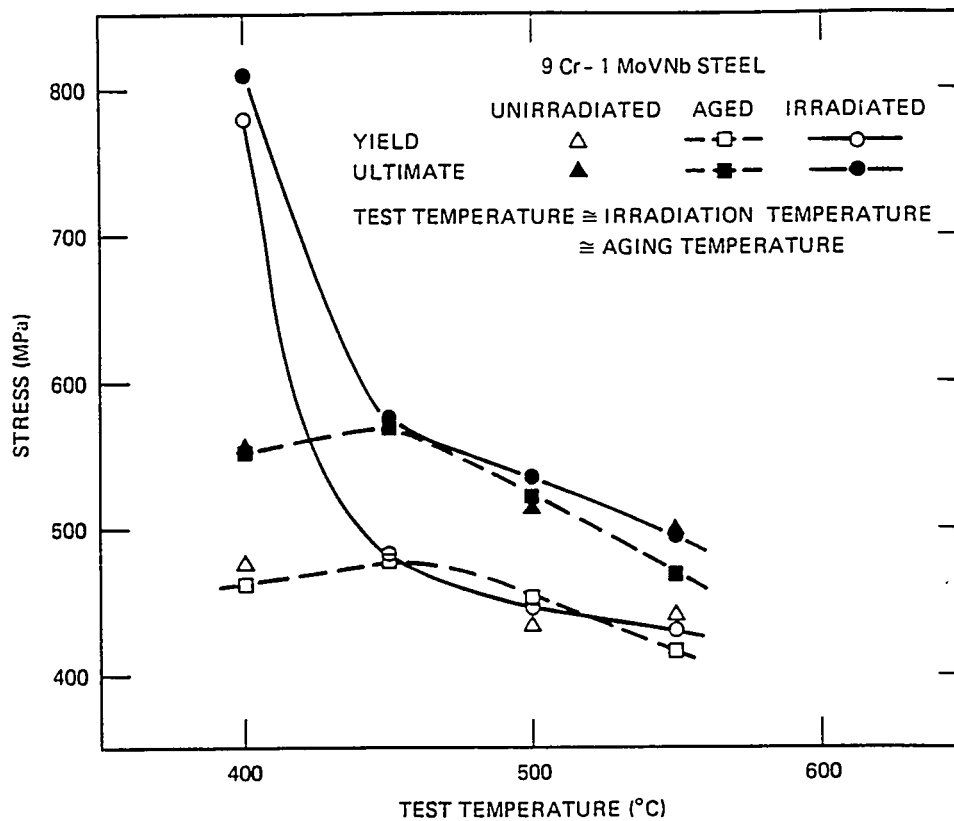
Figure 1 - The 0.2%yield stress and ultimate tensile strength of 9Cr-1MoVNb steel before and after thermal aging and after irradiation to ≈11 dpa

Transmission electron microscopy (TEM) examination of materials irradiated in this experiment were carried out by Gelles and Thomas [11], who compared unirradiated with irradiated material. Unirradiated 9Cr-1MoVNb and 12Cr-1MoVW steels have a tempered martensite structure with large blocky $M_{23}C_6$ precipitates along with small, mainly MC precipitates. After irradiation at 390°C, a high density of dislocation loops and tangles formed, which caused the hardening. Small, rod-shaped precipitates identified as Cr_2C and a small number density of faceted voids were found. For specimens irradiated at 500 or 550°C (none were examined after irradiation at 450°C), there was very little change in microstructure compared to the unirradiated condition [11], in agreement with the unchanged tensile properties at these temperatures.

Temperatures in a fast reactor are controlled by the sodium coolant temperature and irradiation can be carried out between about 360 to 700°C. However, in HFIR, lower irradiation temperatures are possible, and it is found that below 360°C hardening increases somewhat with decreasing temperature [12].

Observations on hardening similar to those on the 9Cr-1MoVNb and 12Cr-1MoVW steels have been made on the 9Cr-2WV and 9Cr-2WVTa reduced-activation steels after all four steels were irradiated in FFTF to ≈7 dpa at 365°C (Fig. 2) [13]. The increase in YS for 9Cr-2WVTa after 7 dpa was slightly less than for the other two 9Cr steels. However, after irradiation to 14 dpa, there was little difference between 9Cr-2WV and 9Cr-2WVTa [14] (data for 7 dpa are shown because no 14 dpa data are available for 9Cr-1MoVNb and 12Cr-1MoVW steels). It appears, therefore, that the YS in the unirradiated condition and the change in YS after irradiation for the 9Cr-2WV, 9Cr-2WVTa, and 9Cr-1MoVNb steels are similar. The 12Cr-1MoVW steel showed the largest increase, being over twice that of the 9Cr steels. Similar comparative observations were made for the UTS, uniform elongation, and total elongation before and after irradiation [13].
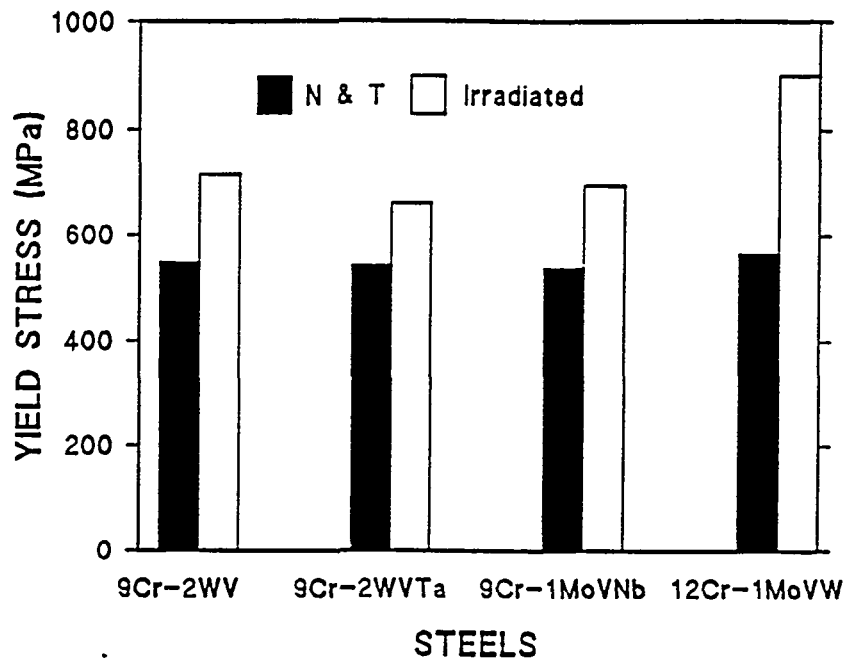
Figure 2 - Yield stress of steels at 365°C before and after irradiation to ≈7 dpa.

Displacement Damage Effects on Impact Behavior

A major concern for ferritic/martensitic steels in light-water reactors, fast reactors, and fusion power plants is the effect of irradiation on impact toughness as measured in a Charpy V-notch test as an increase in DBTT and a decrease in USE. Even if the DBTT is below room temperature before irradiation, it can be well above room temperature after irradiation. Irradiation embrittlement is related to hardening caused by the radiation-produced dislocation loops that form below ≈0.35 T_m (Fig. 3); irradiation-induced precipitates can also have an effect. Irradiation increases the
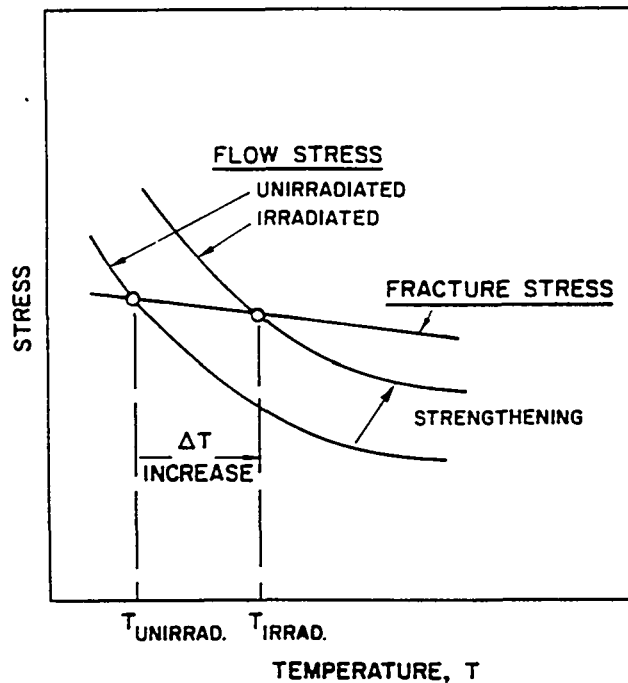

Figure 3 - Schematic diagram that illustrates how strength increase due to irradiation causes a shift in ductile-brittle-transition temperature.

flow stress, and under the assumptions that the fracture stress is unaffected by irradiation and that the intersection of the fracture stress curve and the flow stress curve is the ductile-to-brittle transition temperature for the unflawed condition, the increase in flow stress causes a shift in the DBTT (shown schematically in Fig. 3) [15].

A shift in DBTT (ΔDBTT) of $\approx 160°C$ was observed on 12Cr-1MoVW irradiated to 10 dpa at 365°C in FFTF (Fig. 4) [16]. Irradiation to 17 dpa gave a similar shift, indicating that a saturation with fluence occurred (Fig. 4), just as saturation occurred for hardening in a tensile test [8]. Similar observations of saturation were made by Hu and Gelles [17] for 9Cr-1MoVNb and 12Cr-1MoVW steels irradiated in EBR-II at 390°C to 13 and 26 dpa. Saturation values of 124-144 and 54°C were observed for 12Cr-1MoVW and 9Cr-1MoVNb, respectively. Irradiation at 450, 500, or 550°C showed little effect on the DBTT, in agreement with the observation that most hardening vanishes above 425-450°C [17].
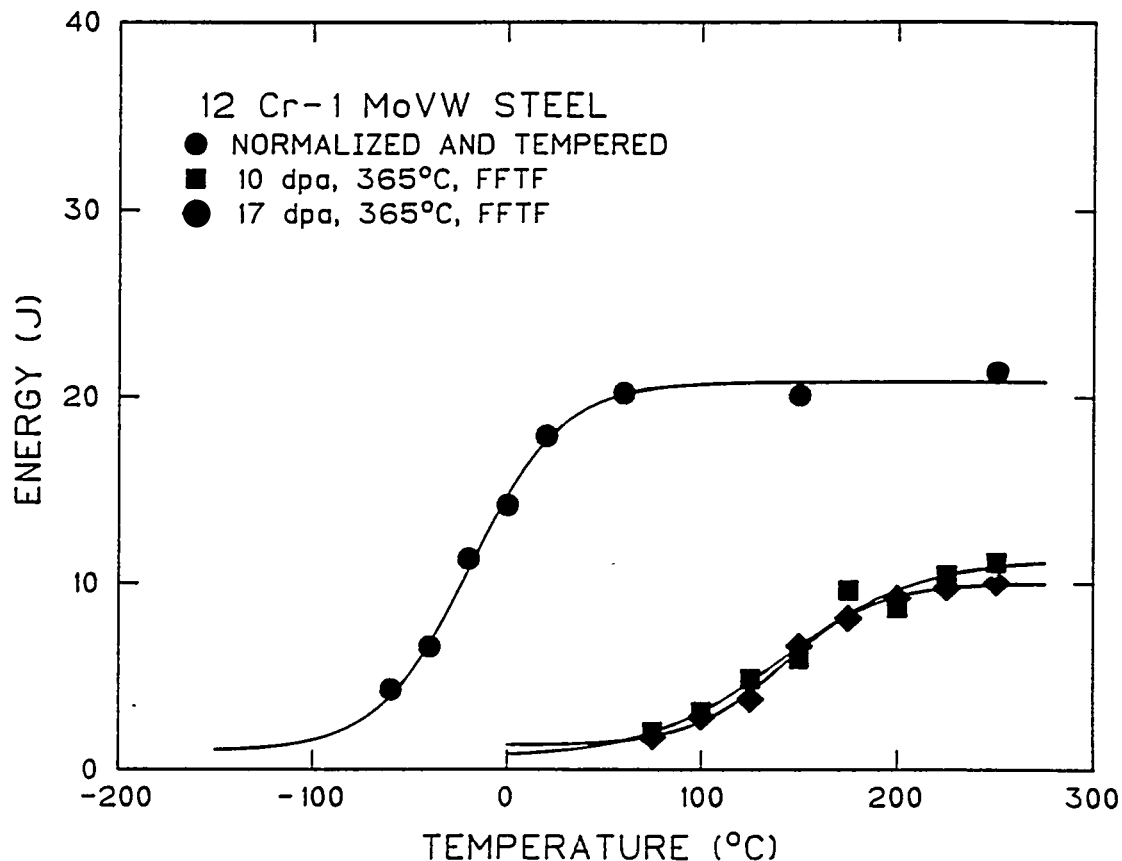

Figure 4 - Charpy curves for half-size specimens of 12Cr-1MoVW steel before and after irradiation to 10 and 17 dpa at 365°C in FFTF.

Although the Charpy curves for the steels are shifted by irradiation, the fracture mode is unaltered between the irradiated and unirradiated steels, with cleavage-type failure on the lower shelf and ductile void coalescence on the upper shelf.

The 9Cr-2WV and 9Cr-2WVTa steels were irradiated at 365°C in FFTF. The 9Cr-2WVTa showed exceptionally small shifts in DBTT: 4, 14, and 21°C after 6.4, 15.4, and 22.5 dpa, respectively [14,18,19]. The ΔDBTT for the 9Cr-2WV saturated at $\approx 52°C$ after 23.9 dpa [19]. This compares with the 9Cr-1MoVNb and 12Cr-1MoVW irradiated at 365°C in FFTF that saturated at ≈ 45 and 140°C, respectively [14]. Thus, the 9Cr-1MoVNb and 9Cr-2WV show similar shifts; they also have similar unirradiated DBTTs (-64°C for the 9Cr-1MoVNb and -60°C for the 9Cr-2WV) and similar tensile properties (Fig. 2). Not only does the 9Cr-2WVTa show a

very small ΔDBTT (21°C), but because it has a very low DBTT in the unirradiated condition (-88°C), the DBTT after irradiation is considerably below that for the other steels after irradiation. An examination of the ΔDBTT data for the three irradiations for the 9Cr-2WVTa steel indicated that there was a gradual increase in the DBTT with increasing dpa [19]. However, even at the highest dpa, the DBTT for 9Cr-2WVTa is lower (-67°C) than for any of the other steels before irradiation.

The difference in Charpy properties of the 9Cr-2WV and 9Cr-2WVTa before and after irradiation occurred despite little difference in strength before and after irradiation (Fig. 2). Further, transmission electron microscopy of the normalized-and-tempered 9Cr-2WV and 9Cr-2WVTa indicated only minor differences prior to irradiation [20]. Tantalum refines the prior austenite grain size [19], which can affect the DBTT, but the lath size of the two steels was similar [20], and the 9Cr-2WVTa contained a slightly larger amount of MC. There was also no marked difference in microstructure after irradiation [20]. Therefore, the only major difference in the two steels to account for the difference in Charpy properties is the tantalum in solid solution. An atom probe analysis of the unirradiated 9Cr-2WVTa steel indicated that ≈90% of the tantalum remained in solution in the normalized-and-tempered condition [21].

A smaller grain size in the 9Cr-2WVTa, which could be caused by the tantalum, was originally used to explain the difference between the 9Cr-2WV and 9Cr-2WVTa steels [14]. This explanation was subsequently questioned because as normalized-and-tempered the two steels had a similar YS, and they also had a similar YS to 9Cr-1MoVNb (Fig. 2), which had the smallest grain size of the three steels [19,20]. This leads to the conclusion that microstructure (grain size, precipitate type, etc.) is not the sole explanation of the mechanical property changes, and that in addition to affecting grain size, tantalum in solution must cause an increase in fracture stress for 9Cr-2WVTa over 9Cr-2WV, and the combination of tungsten and tantalum in the 9Cr-2WVTa leads to a higher fracture stress than produced by molybdenum and niobium in 9Cr-1MoVNb [19]. The observation that the ΔDBTT of the 9Cr-2WVTa increased slightly with fluence is in agreement with that explanation. This increase in DBTT would follow if tantalum is being removed from solution during irradiation and incorporated in the precipitates, thus decreasing the matrix cleavage fracture stress.

Helium Effects

The effects of the simultaneous formation of displacement damage and transmutation helium was studied by irradiating standard 9Cr-1MoVNb and 12Cr-1MoVW steels in HFIR at 400°C to ≈40 dpa [22]. This produced ≈35 appm He in 9Cr-1MoVNb and ≈110 appm He in the12Cr-1MoVW. The helium came from the ≈0.11 and 0.43% Ni in the respective heats of 9Cr-1MoVNb and 12Cr-1MoVW steel that were irradiated in these experiments [22]. The ΔDBTT for 9Cr-1MoVNb and 12Cr-1MoVW after HFIR irradiation were 202 and 242°C [24], respectively, compared to the saturated values of ≈54 and 124-144°C obtained by Hu and Gelles [17] after irradiation at 390°C in EBR-II. The same heat of 12Cr-1MoVW steel that was irradiated in HFIR was also irradiated to ≈12 dpa at 390°C in EBR-II and had a ΔDBTT of 122°C [23], in agreement with the saturated values obtained by Hu and Gelles. These results indicate that the saturation values found for the steels irradiated in EBR-II do not apply when they are irradiated in HFIR.

Similar differences in HFIR and EBR-II irradiation were observed for a different heat of 12Cr-1MoVW steel irradiated at 400°C to 4-9 dpa (≈25 appm He) in HFIR [24]. A ΔDBTT of 195°C was observed, which is larger than the 124-144°C saturation value observed in EBR-II and lower than the 242°C after ≈40 dpa in HFIR [22]. Not only do larger DBTTs occur for irradiation in HFIR, but if a saturation fluence exists for HFIR, it is higher than for irradiation in EBR-II.

Because the objective was to simulate the effects on steels in a fusion reactor, 9Cr-1MoVNb and 12Cr-1MoVW with 2% Ni (9Cr-1MoVNb-2Ni and 12Cr-1MoVW-2Ni) were irradiated in HFIR[22], since steels with 2% Ni irradiated in HFIR develop a He/dpa ratio similar to that in a Tokamak fusion reactor [5]. The 9Cr-1MoVNb-2Ni and 12Cr-1MoVW-2Ni steels had ΔDBTTs of 348 and 328°C, respectively, well above the saturated values in EBR-II. When the 12Cr-1MoVW-2Ni was irradiated to ≈12 dpa at 390°C in EBR-II, where little helium was produced, a ΔDBTT of only 90°C was observed [23], similar to the 12Cr-1MoVW steel in EBR-II [17, 23].

Unfortunately, the radioactive fractured specimens from HFIR became unavailable for scanning electron microscopy (SEM). To obtain information on the fracture mode, TEM specimens were fractured in the hot cell. The specimens were held in a fixture, cooled in liquid nitrogen, fractured by striking with a hammer, and examined by SEM [22]. Such a fracture will occur on the lower shelf and should give an indication of the fracture mode. A 12Cr-1MoVW-2Ni steel specimen irradiated in HFIR to ≈74 dpa and 760 appm He had a fracture that was ≈75% intergranular [Fig. 5(a)], compared to a normalized and tempered specimen, where the fracture was essentially all cleavage [Fig. 5(b)]. Specimens irradiated in EBR-II also showed cleavage fracture [22].
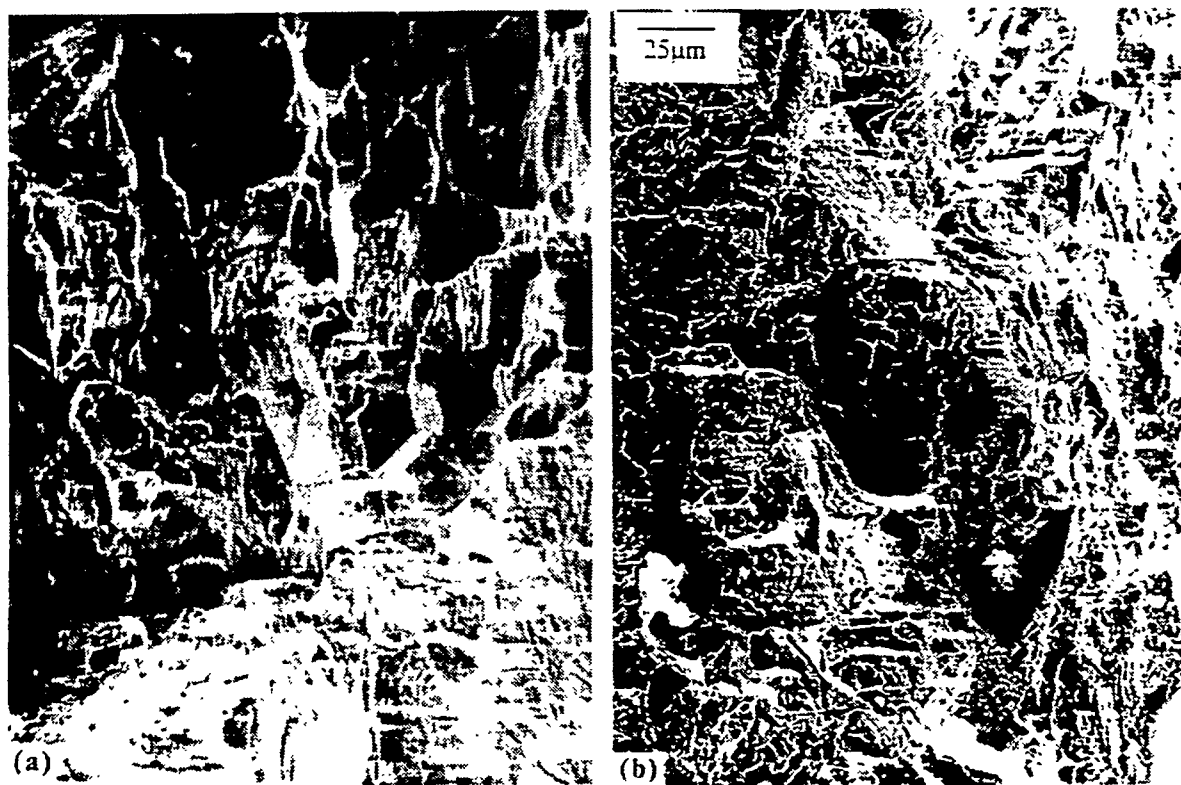

Figure 5 - Scanning electron micrographs of fractured TEM disks of 12Cr-1MoVW-2Ni steel (a) irradiated to 74 dpa (760 appm He) in HFIR and (b) as normalized and tempered.

Transmutation helium can affect the behavior of an irradiated alloy in three ways [25]. First, helium stabilizes vacancy clusters, which, in turn, cause an increase in the number of interstitial clusters (i.e., helium ties up vacancies and reduces interstitial-vacancy recombination). Interstitial clusters grow into dislocation loops and increase the strength. Second, helium stabilizes the clusters to a higher temperature. Third, helium migrates to grain boundaries during irradiation, which can then affect mechanical properties [25].

The first two effects lead to hardening. Tensile studies on the nickel-doped and undoped steels indicate that helium might provide an increment of hardening above that due to radiation-produced

defects and precipitates [12]. At 400°C the scatter in those data makes a definite conclusion concerning the extent of hardening uncertain, beyond saying there is probably a slight increment of hardening due to the helium. Therefore, the results indicated that the large shift in DBTT observed in HFIR was not caused by the increase in flow stress alone, as illustrated in Fig. 3, but rather, was attributed to a lower fracture stress caused by a change in fracture mode from cleavage to intergranular fracture [26].

An increase in DBTT can be caused by: (1) more or larger flaws,(2) less resistance to the initiation of a flaw, and (3) less resistance to the propagation of a flaw. Inclusions or carbides are likely sources of microcracks that initiate fracture in steels [27]. The larger ΔDBTT for 12Cr-1MoVW than 9Cr-1MoVNb in FFTF (little helium) was attributed to the larger amounts of large precipitate particles in 12Cr-1MoVW steel [26]. The 12Cr-1MoVW contains twice as much precipitate as 9Cr-1MoVNb because it contains twice as much carbon.

To explain the helium effect, it was proposed that when the steels contain sufficient helium the microcrack source could be a helium-containing bubble or bubbles on a prior-austenite grain boundary (or on a martensite lath boundary). Helium is envisioned to collect into small cavities that under stress become nuclei for fracture and/or enhance crack propagation, explaining why fracture surfaces of HFIR-irradiated, helium-containing steels contain intergranular facets.

Considerable speculation is inherent in this discussion. However, if helium plays the role postulated, then impact toughness could be affected in a fusion reactor, where large amounts of helium would be generated in the first wall. With enough helium generated during irradiation, the ΔDBTT below 400°C could become as large or larger than at 400°C. It is unclear what might happen above 400°C, since no HFIR experiments were conducted at these temperatures. The diffusion rate increases with temperature, thus increasing the rate at which helium can migrate to boundaries. On the other hand, irradiation hardening decreases rapidly above 400°C and disappears above ≈450°C. Therefore, even if helium is present on boundaries, the reduced yield stress may preclude a larger ΔDBTT than observed for fast reactor irradiation.

At this juncture, a critical need exists for fracture toughness data, and an understanding of how the Charpy data are related to fracture toughness needs to be developed. Only with such an understanding can the implications of the helium effects observed in these studies be evaluated in the context of fusion reactor design.

Summary and Conclusions

High-chromium martensitic steels have properties that make them candidate structural materials for future fusion power plants. The major problem faced by the martensitic steels for this application is the reduction in toughness that is caused by neutron irradiation. When a steel is irradiated by neutrons, atoms are displaced into interstitial positions (displacement damage), which harden the steel. Hardening gives rise to a decrease in impact toughness, as indicated by an increase in the DBTT and a decrease in the USE in a Charpy impact test. Transmutation helium also forms from the high-energy neutrons generated by the fusion reaction. This helium can also affect the toughness. In the absence of a fusion reactor or other high-energy neutron source, the effects can only be studied by simulating possible fusion effects by irradiation in fission reactors.

Neutron irradiation of the conventional 9Cr-1MoVNb and 12Cr-1MoVW steels in a fast reactor, where little helium is generated over the temperature range 365-550°C, has shown that these steels harden below ≈425°C; at higher temperatures there is little change in strength. The steels show an increase in the DBTT and decrease in the USE at the temperatures where the hardening occurs.

Increases in DBTT of 54°C for the 9Cr-1MoVNb steel and 124-144°C for the 12Cr-1MoVW steel were observed when the steels were irradiated in a fast reactor. A reduced-activation 9Cr-2WVTa steel was developed that had a much lower DBTT than the conventional steels and showed a shift in DBTT that was less than half that of the conventional steels after fast reactor irradiation.

When transmutation helium was introduced into the 9Cr-1MoVNb and 12Cr-1MoVW steels by irradiation in HFIR, where helium is generated by the interaction of thermal neutrons in the HFIR spectrum with nickel in the steel, an approximate four-fold increase in the DBTT of the 9Cr-1MoVNb steel and an approximate doubling of the DBTT of the 12Cr-1MoVW steel were observed. These increases were tentatively attributed to a change in the fracture mode of the steel caused by the helium.

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