TENSILE BEHAVIOR OF NEUTRON-IRRADIATED MARTENSITIC STEELS: A REVIEW

MATERIALS

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Chromium-molybdenum martensitic (ferritic) steels such as 9 Cr-1 Mo-V-Nb and 12 Cr-1 Mo-V-W are candidates for fast reactor and fusion reactor applications. In a fast reactor, the effect of neutron irradiation is caused by displacement damage, that is, by the interstitials and vacancies that are created by the high-energy neutrons. Increases in strength occur for irradiation up to ~450°C. This hardening is largely attributed to the dislocation loops that form from the agglomeration of the interstitials. Precipitates that form during irradiation can also contribute to the hardening. At higher temperatures, most of the displacement damage anneals out.

Irradiation effects expected in the first wall of a fusion reactor differ from those in a fast reactor. In addition to displacement damage, large amounts of transmutation helium will also be produced. The simultaneous effects of displacement damage and helium can be simulated by irradiating nickel-doped ferritic steels in a mixed-spectrum fission reactor. Helium is produced by transmutation reactions between thermal neutrons and nickel, and displacement damage is formed by the fast neutrons of the spectrum. Results using this technique indicate that hardening occurs as in a fast reactor, but the helium causes a strength increase in addition to that caused by displacement damage alone. This effect of helium could have a significant effect on other properties, especially toughness, and must be considered in the design of fusion reactors.

I. INTRODUCTION

Martensitic steels first became of interest for nuclear applications in the early 1970s when they were considered for duct and cladding applications for the liquid-metal reactor (LMR) program.[1] The first steel investigated extensively was the Sandvik alloy HT-9, a 12 Cr-1 Mo steel containing 0.2% carbon and also containing additions of vanadium, tungsten, and nickel (Table I). For this paper, the steel will be designated generically as 12 Cr-1 Mo-V-W. The 12 Cr-1 Mo-V-W steel was found to have microstructural stability during irradiation and superior resistance to void swelling compared with austenitic stainless steels. As a result, it was also identified as a possible structural alloy for magnetic fusion reactors when the fusion energy program began to consider ferritic steels.[2]

Because the modified 9 Cr-1 Mo steel developed by Combustion Engineering and Oak Ridge National Laboratory showed strength properties comparable to those of 12 Cr-1 Mo-V-W steel, it was also examined for LMR applications and found highly resistant to irradiation. This steel (designated here as 9 Cr-1 Mo-V-Nb) is modified with vanadium, niobium, and nitrogen (Table I), and it is also considered a possible candidate for fusion reactor applications.

Although not a martensitic steel, the ferritic $2\frac{1}{4}$ Cr-1 Mo steel (Table I) has also been considered a possible candidate for fusion applications. Because carbon in this steel is not stabilized by niobium or vanadium, the steel cannot be used in the primary circuit of an LMR. Use of the steel in such a sodium-cooled system containing austenitic steel structural components would result in excessive decarburization.

A new criterion of reduced-activation materials has recently been introduced for alloys for fusion applications.[3] All alloys irradiated in a fusion reactor will become radioactive. Depending on the elements in the steel, the decay of the radioactivity in the activated material can take tens to thousands of years. Reduced-activation alloys are those that would decay rapidly to low levels in such a period (tens of years) that they could be disposed of by shallow land burial instead of the much more expensive deep geological disposal. For ferritic or martensitic steels to meet this criterion,

TABLE I

Chemical Composition of Test Materials

Element	9 Cr-1 Mo-V-Nb (Heat 30176)	12 Cr-1 Mo-V-W (Heat 9607-R2)	$2\frac{1}{4}$ Cr-1 Mo (Heat 56447)
Carbon	0.092	0.20	0.10
Silicon	0.15	0.17	0.23
Manganese	0.48	0.57	0.40
Phosphorus	0.012	0.016	0.009
Sulfur	0.004	0.003	0.006
Chromium	8.32	12.1	2.16
Molybdenum	0.86	1.04	1.03
Nickel	0.09	0.51	0.14
Vanadium	0.20	0.28	0.01
Niobium	0.06	---	---
Tungsten	<0.01	0.45	---
Nitrogen	0.055	0.027	---
Aluminum	0.006	0.006	---
Titanium	0.001	0.001	0.01
Copper	0.03	---	0.03

the amounts of molybdenum, niobium, nitrogen, and nickel in the steel must be minimized. As a result, chromium-tungsten-vanadium steels are being developed.[3] Preliminary mechanical property data indicate that reduced-activation martensitic steels behave similarly to the conventional martensitic steels.

II. RADIATION DAMAGE

When a steel is irradiated in a high-energy neutron field (only neutron irradiation will be considered), neutrons displace atoms from their lattice positions into interstitial positions and leave behind a vacant lattice site. These displaced atoms (termed interstitials) move into and out of vacant lattice sites (vacancies), and this "displacement damage" is described in terms of the average number of times an atom is displaced from its lattice position as displacements per atom (dpa). The dpa can be calculated from the neutron fluence received by the steel.

It is the disposition of the interstitials and vacancies created by irradiation that determines the effect of the irradiation on the tensile properties—as well as many other properties. At reactor temperatures, interstitials and vacancies are mobile, and most are eliminated by a one-to-one recombination and therefore have no effect on properties. Those that do not recombine migrate to "sinks," where they are absorbed. Sinks include surfaces, grain boundaries, dislocations, and existing cavities. Tensile and other properties are affected by the defect clusters that can form. Clusters consisting of interstitials can evolve into dislocation loops, and vacancy clusters can develop into microvoids or cavities. Solute clusters can also form under certain conditions.

The type of cluster defect that forms depends on the irradiation temperature. Below $\sim 0.35T_m$, where T_m is the absolute melting point of the irradiated material, interstitials are mobile relative to vacancies, and the interstitials combine to form dislocation loops. This gives rise to an increase in strength and a decrease in ductility.

Vacancies become increasingly mobile above $0.35T_m$, and a dislocation and cavity structure results. This microstructure, which is accompanied by an increase in volume (swelling), occurs because certain sinks have a bias and do not accept vacancies and interstitials equally.[4] If all sinks accepted both defects equally, then the vacancies and interstitials would annihilate at a sink, and no swelling would result. However, within a grain, interstitials are accepted preferentially by dislocations. This leaves an excess of vacancies to be absorbed by cavities, giving rise to the observed swelling.

Finally, at high irradiation temperatures (greater than $\sim 0.6T_m$), defect clusters are unstable. That is, the high equilibrium vacancy concentration and rapid diffusion lead to vacancy-interstitial annihilation, and displacement damage has little effect on properties. However, at elevated temperatures (greater than $\sim 0.5T_m$), any transmutation helium produced during irradiation can lead to problems in tensile ductility.

In addition to the displacement damage, a neutron can be absorbed by an atom of the irradiated alloy, resulting in a transmutation reaction that produces a new metal atom and hydrogen and/or helium gas atoms within the alloy being irradiated. Indications are that small amounts of new metal atoms have little effect on properties. Likewise, hydrogen will probably have little effect on properties because it should diffuse from the alloy at the operating temperatures of most reactors (250 to 550°C). However, helium can affect the properties. It is relatively insoluble in metals and thus will be incorporated into bubbles or voids that can form within the matrix and on grain boundaries and precipitate interfaces.

Although there are subtle differences between the disposition of displacement damage produced in steels in a fast fission reactor and a fusion reactor, the effect of displacement damage on candidate steels for a fusion reactor can be studied by irradiating in a fast fission reactor. However, the much higher energy of the neutrons (up to 14 MeV) produced in a fusion reactor will result in much more transmutation helium than forms in most fission irradiations. Such a simultaneous development of displacement damage and helium can affect both the swelling behavior and the mechanical properties relative to the formation of displacement damage alone.

III. IRRADIATION EFFECTS ON TENSILE PROPERTIES

Irradiation effects on tensile properties will be examined for two processes: the effect of displacement damage of the type generated in a fast reactor and the effect of a combination of displacement damage and transmutation helium of the type that is expected to be important for a fusion reactor. At present, there is no fusion reactor or other 14-MeV neutron source available to study the effect of helium directly, and the effect of the simultaneous formation of displacement damage and helium generation must be simulated using fission reactors.

Tensile studies have been conducted on martensitic steels for the LMR program, but few data from those studies have been published. For that reason, data for both parts of this discussion are taken from studies conducted for the U.S. fusion energy program. Furthermore, the effect of neutron irradiation on martensitic steels will be discussed for the conventional 9 Cr–1 Mo–V–Nb and 12 Cr–1 Mo–V–W steels. It is expected that all martensitic steels, including the reduced-activation steels, will behave similarly when irradiated.

For the steels to be discussed, the 9 Cr–1 Mo–V–Nb and 12 Cr–1 Mo–V–W steels were in the normalized and tempered condition before irradiation. Both steels are normalized by first austenitizing for 0.5 to 1 h at ~1050°C (the time depends on the thickness of material being heat treated) and then either air cooling or cooling rapidly in a flowing gas. Tempering is carried out at 750 to 780°C for 1 to 2.5 h. After the normalization treatment, the microstructure consists of a martensite lath network with the matrix containing a high

of a ferrite matrix with carbide precipitates and a relatively low dislocation density. The carbides are a combination of $M_{23}C_6$ and MC (Ref. 5).

IV. DISPLACEMENT DAMAGE EFFECTS ON TENSILE BEHAVIOR

Irradiation in a fast reactor such as the Experimental Breeder Reactor II (EBR-II) or the Fast Flux Test Facility (FFTF) results in displacement damage with little helium formation. Irradiation temperatures are controlled by the sodium coolant temperature, and tests can be conducted between ~360 to 700°C.

An example of the effect of fast reactor irradiation on tensile properties of 9 Cr–1 Mo–V–Nb steel is shown in Figs. 1 and 2 (Ref. 6). Irradiation was at 390, 450, 500, and 550°C in EBR-II to ~9 dpa (the specimens irradiated at 390°C were tested at 400°C). Also shown are data for the normalized and tempered steel and for specimens thermally aged at the irradiation temperature for 5000 h, the approximate time in reactor.

Irradiation at 390°C resulted in an increase in both the 0.2% yield stress and the ultimate tensile strength (UTS) (Fig. 1). At 450, 500, and 550°C, essentially no

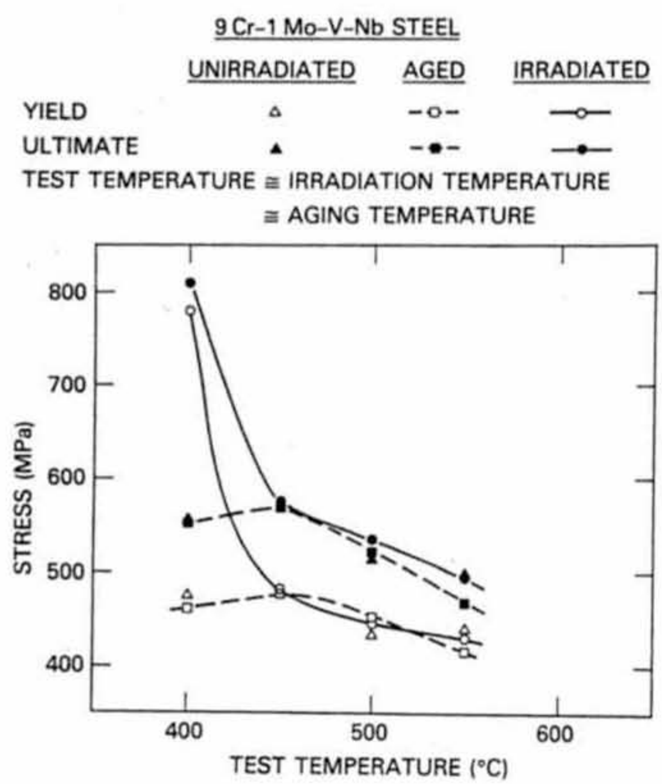

Fig. 1. The 0.2% yield stress and UTS as a function of temperature for 9 Cr–1 Mo–V–Nb steel after irradiation to ≈11 dpa, in the normalized and tempered condition, and after thermal aging for 5000 h at the irra-

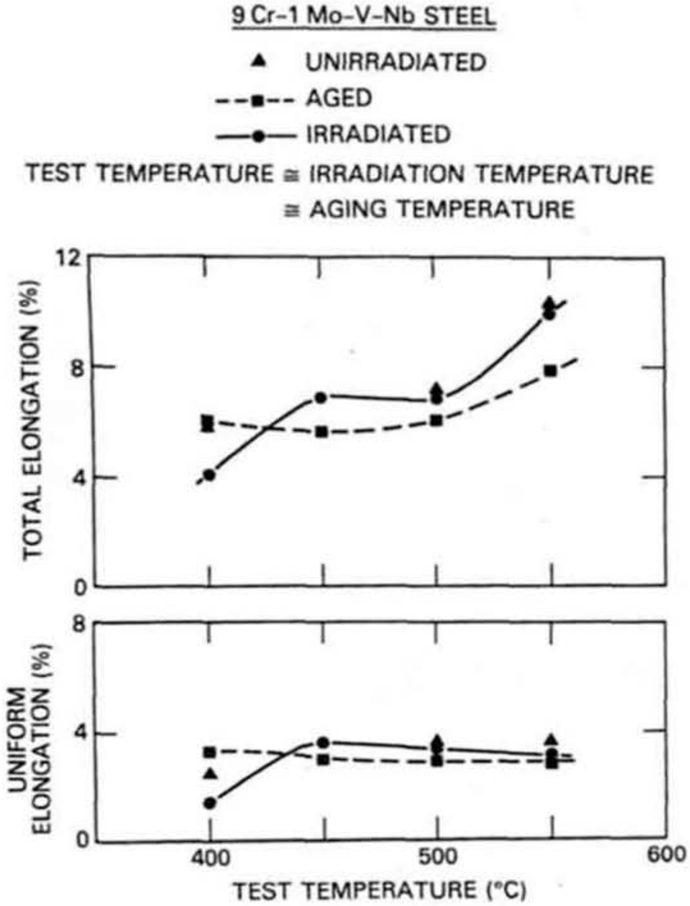

Fig. 2. The uniform and total elongation as a function of temperature for 9 Cr-1 Mo-V-Nb steel after irradiation to ≈11 dpa, in the normalized and tempered condition, and after thermal aging for 5000 h at the irradiation temperature.

change occurred in yield stress or UTS that could be attributed to irradiation or thermal aging. There was essentially no difference in strength among the different specimens at these three temperatures.[6]

The effect of irradiation on ductility reflected the effect on strength (Fig. 2). The uniform and total elongations of the specimens irradiated at 390°C were slightly less than those of the unaged and aged controls. At the three highest irradiation and test temperatures, there appeared to be no effect of irradiation.[6]

Transmission electron microscopy (TEM) examination of materials irradiated in this experiment were carried out by Gelles and Thomas,[7] who compared normalized and tempered material with irradiated material. In the unirradiated condition, the 9 Cr-1 Mo-V-Nb steel is a tempered martensite structure that contains large blocky $M_{23}C_6$ precipitate particles along with a higher density of small, mainly MC precipitates. After irradiation at 390°C, a high density of dislocation loops and tangles formed.[7] Small, rod-shaped precipitates identified as Cr_2C and a small number of faceted voids were also found. For specimens irradiated at 500 and 550°C (none were examined after irradiation at 450°C), there was very little change in microstructure compared with the unirradiated condition.[7]

The TEM observations fit quite well with the observations on tensile properties. Hardening at 390°C is caused by the irradiation-induced dislocation and precipitate structure. At 450, 500, and 550°C, the lack of any significant change in microstructure coincides with the unchanged tensile properties after irradiation or thermal aging at these temperatures.

Specimens of 12 Cr-1 Mo-V-W steel were irradiated up to 13 dpa in the same experiment as the 9 Cr-1 Mo-V-Nb steel.[8] Similar effects were observed: Hardening occurred at 390°C, and essentially no change resulted from irradiation at 450, 500, and 550°C. The $2\frac{1}{4}$ Cr-1 Mo ferritic steel was also irradiated in this experiment. Hardening was observed for the steel irradiated at 390°C. However, an irradiation-enhanced loss of strength was observed at 550°C, as well as some loss of strength due to thermal aging alone.[9] The difference between the steels is attributed to the presence of the more stable $M_{23}C_6$ and MC in the 9 Cr-1 Mo-V-Nb and 12 Cr-1 Mo-V-W steels, compared with the relatively unstable M_2C and M_7C_3 in $2\frac{1}{4}$ Cr-1 Mo. Irradiation can accelerate thermal aging because the large number of vacancies produced enhances diffusion.

Both the 9 Cr-1 Mo-V-Nb and 12 Cr-1 Mo-V-W steels were further irradiated in EBR-II to 23 to 25 dpa at the same temperatures.[10] At 390°C, there was little change relative to the steels irradiated to 9 to 13 dpa, an indication that the hardening saturates at a fluence of ~10 dpa. After irradiation at 450, 500, and 550°C, there was also little difference in the tensile properties of the irradiated and the normalized and tempered specimens.[10]

V. EFFECT OF HELIUM ON THE TENSILE PROPERTIES OF IRRADIATED STEELS

Because there is no 14-MeV neutron source to produce the helium:dpa ratio expected in the first wall of a fusion reactor, simulation techniques using fission reactors are required to study the effect of helium. A method employed for the martensitic steels is to add small amounts of nickel to the steels and to irradiate them in a mixed-spectrum reactor, such as the High-Flux Isotope Reactor[11] (HFIR). Both fast and thermal neutrons are generated in a mixed-spectrum reactor. Fast neutrons in the spectrum produce displacement damage. Helium is produced when the ^{58}Ni in the steel undergoes the following two-step reaction with thermal neutrons in the spectrum:

$$^{58}Ni + n \rightarrow {}^{59}Ni + \gamma$$

and

$$^{59}Ni + n \rightarrow {}^{56}Ni + {}^{4}He.$$

Helium production in steels in HFIR is illustrated in Fig. 3 for conventional 9 Cr-1 Mo-V-Nb (containing 0.2% nickel) and 9 Cr-1 Mo-V-Nb with 2% nickel (9 Cr-1 Mo-V-Nb-2 Ni). Expected helium production in the first wall of a tokamak-type magnetic fusion reactor is shown, along with the helium produced in steels in fast reactors such as EBR-II and FFTF. As seen, when a steel with ~2% nickel is irradiated in HFIR, helium is produced at a rate similar to that in a tokamak. Also shown is the curve for Type 316 stainless steel. Because of the high nickel content of this steel, a larger helium buildup occurs than would occur in a fusion reactor. Other techniques must be used for such high nickel alloys to obtain the correct helium:dpa ratio for fusion.

To determine whether helium affects tensile properties, 9 Cr-1 Mo-V-Nb, 9 Cr-1 Mo-V-Nb-2 Ni, 12 Cr-1 Mo-V-W (0.5% nickel), 12 Cr-1 Mo-V-W-1 Ni (1% nickel), and 12 Cr-1 Mo-V-W-2 Ni (2% nickel) steels were irradiated in HFIR at ~50°C (the HFIR coolant temperature), and the effect of fluence (0 to ~25 dpa) and helium concentration (up to 327 appm) was determined by room temperature tensile tests.[12] Results for the 12 Cr-1 Mo-V-W, 12 Cr-1 Mo-V-W-1 Ni, and 12 Cr-1 Mo-V-W-2 Ni steels indicated that the yield stress increased with increasing displacement damage (Fig. 4). Above ~5 dpa, the rate of strength increase with fluence decreased as the damage level increased.[12]

Since nickel lowers the A_{c1} temperature, the steels with 2% nickel were tempered at a lower temperature (700°C) than the conventional steels or the 12 Cr-1 Mo-V-W-1 Ni steel (780°C). This resulted in different strengths in the unirradiated condition and caused some difficulty in interpreting the results. The differences for the different nickel compositions after irradiation were most easily seen by comparing the behavior of the 12 Cr-1 Mo-V-W and 12 Cr-1 Mo-V-W-1 Ni steels (Fig. 4), which had similar strengths in the unirradiated condition. When irradiated to ~20 dpa to produce ~80 and 160 appm helium in 12 Cr-1 Mo-V-W and 12 Cr-1 Mo-V-W-1 Ni, respectively, the

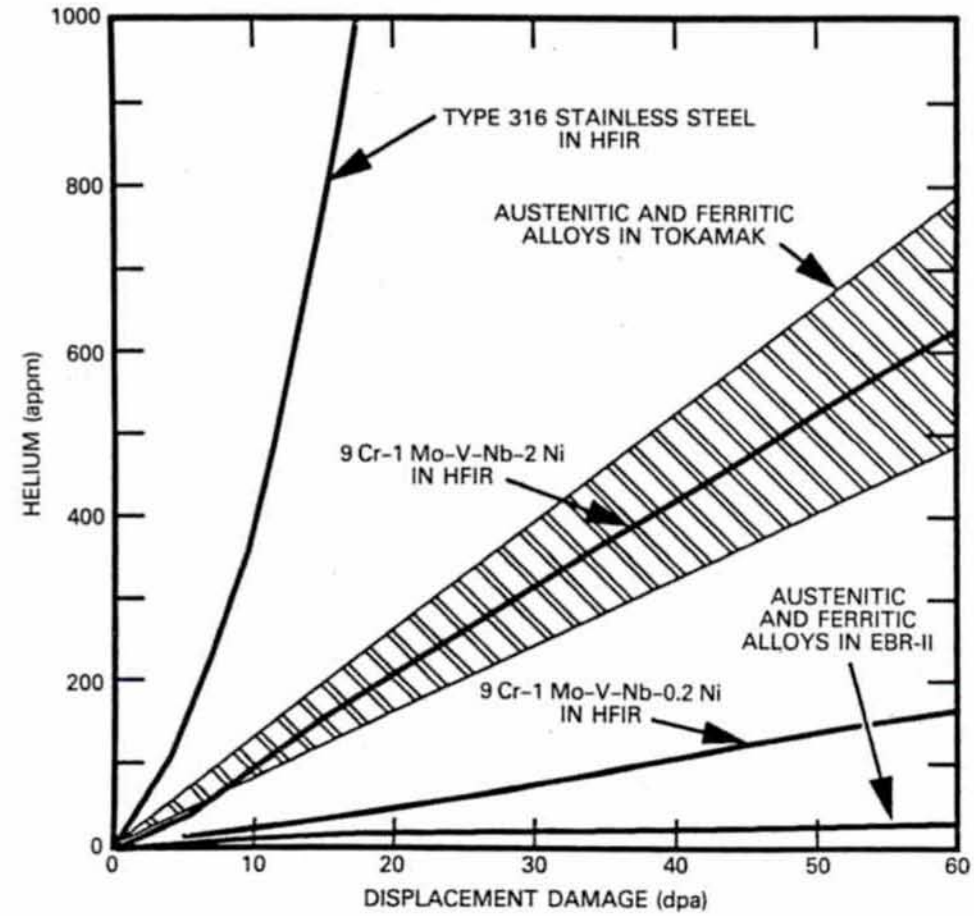

Fig. 3. Relationship between helium concentration and displacement damage for austenitic and ferritic steels irradiated in HFIR, EBR-II, and the first wall of a tokamak fusion reactor.

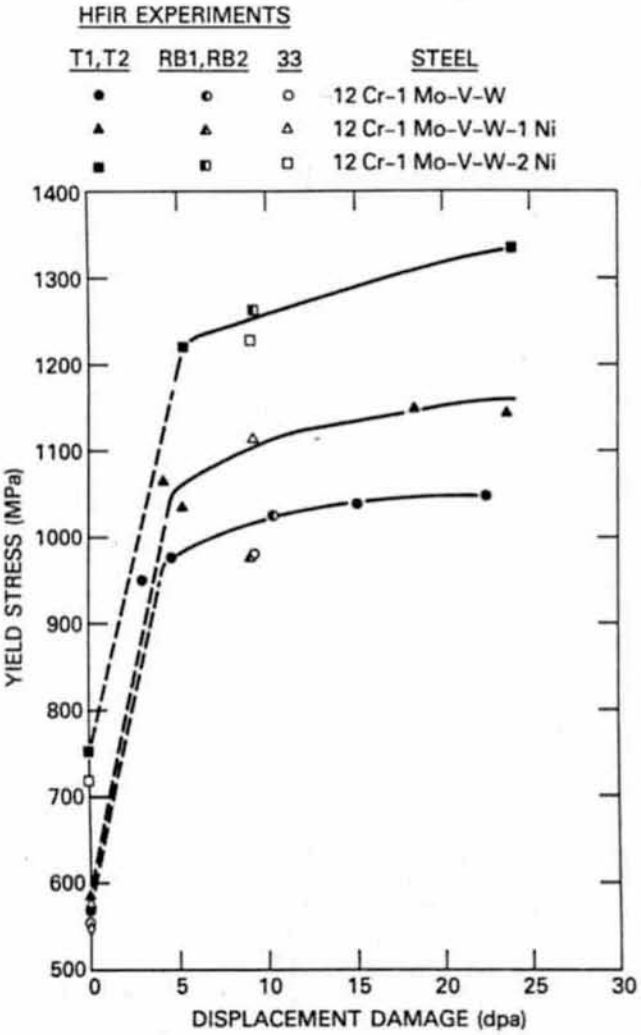

Fig. 4. The yield stress of 12 Cr-1 Mo-V-W, 12 Cr-1 Mo-V-W-1 Ni, and 12 Cr-1 Mo-V-W-2 Ni steels as a function of fluence after irradiation in HFIR at ≈50°C.

yield stress of the 12 Cr-1 Mo-V-W-1 Ni steel increased ~100 MPa more than that for 12 Cr-1 Mo-V-W steel (Fig. 4). Likewise, although the yield stress of the 12 Cr-1 Mo-V-W-2 Ni steel was ~185 MPa above that for the 12 Cr-1 Mo-V-W steel in the unirradiated condition, the difference was ~256 MPa at 20 dpa, where the 12 Cr-1 Mo-V-W-2 Ni steel contained ~320 appm helium.[12] The difference between the 12 Cr-1 Mo-V-W-1 Ni and 12 Cr-1 Mo-V-W-2 Ni steel after ~20 dpa was only ~20 MPa more than before irradiation, indicating perhaps that a saturation with helium concentration was occurring. Differences in the UTS behavior for the different steels showed a similar trend to that of the yield stress.[12]

Results for the yield stress behavior of 9 Cr-1 Mo-V-Nb and 9 Cr-1 Mo-V-Nb-2 Ni steels are shown in Fig. 5. Again, although there is a difference of 220 MPa between the yield stress of the two steels in the normalized and tempered condition, the difference is ~350 MPa after irradiation to ~20 dpa, where the 9 Cr-1 Mo-V-Nb and 9 Cr-1 Mo-V-Nb-2 Ni steels contained ~20 and 220 appm helium, respectively. A similar observation was made for the UTS behavior.[12]

At low fluences, a ductility decrease accompanied the strength increase for the three 12 Cr-1 Mo-V-W steels.[12] Because the steel with no nickel added had the highest ductility before irradiation, it showed the largest decrease, followed by the steel with 1% nickel. The total elongation showed indications of a minimum after displacement damage levels of 5 to 10 dpa. After 15 dpa or more, all three steels had elongations above those at 10 dpa.[12] Similar results were observed for the

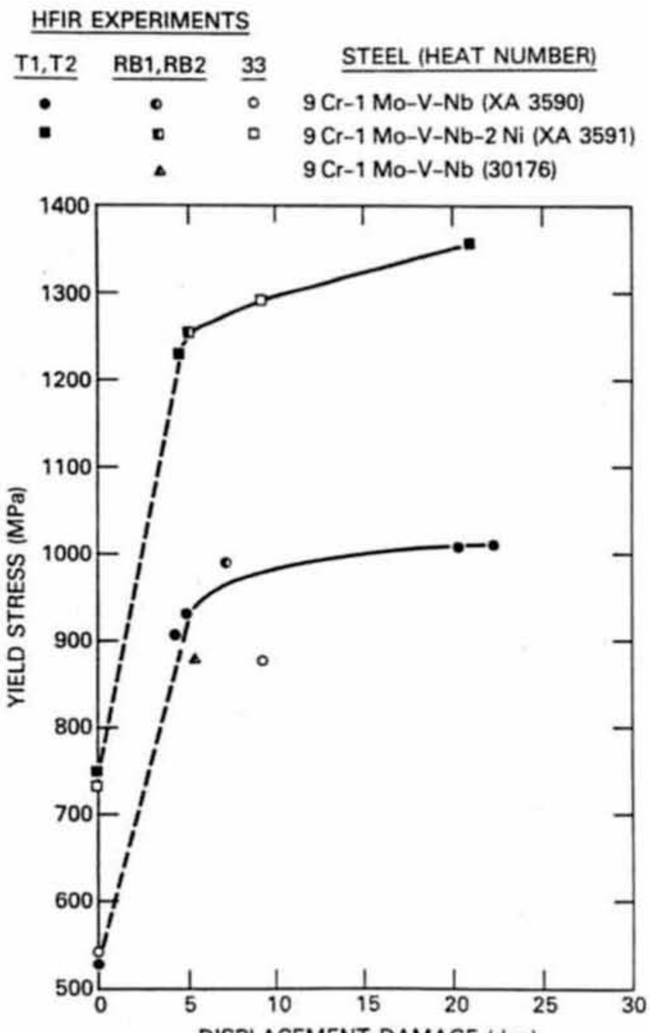

Fig. 5. The yield stress of 9 Cr-1 Mo-V-Nb and 9 Cr-1 Mo-V-Nb-2 Ni steels as a function of fluence after irradiation in HFIR at ≈50°C.

ductility of the 9 Cr-1 Mo-V-Nb and 9 Cr-1 Mo-V-Nb-2 Ni steels.[12]

There are indications that the hardening in chromium-molybdenum steels at 390°C irradiated in EBR-II saturates with increasing fluence,[10] as has been observed for austenitic stainless steels.[13] These 50°C results for the 9 Cr-1 Mo-V-Nb and 12 Cr-1 Mo-V-W steels indicated that the rate of increase in strength decreased significantly with increasing fluence between about 5 and 10 dpa (Figs. 4 and 5) (Ref. 12). Although data are limited, the results were interpreted to indicate that after the rather rapid decrease in hardening rate beyond ~5 dpa, there might still be a slight upward slope to the strength-displacement damage curves. However, as pointed out earlier, the comparison of the results for the 12 Cr-1 Mo-V-W steels with different nickel contents indicates that a saturation may be occurring as the helium concentration increases, just as saturation in hardening appears to occur with increasing displacement damage in the absence of helium for steels irradiated at 390°C (Ref. 10).

The nickel-doped steels were also irradiated in HFIR at 300, 400, and 500°C up to ~11 dpa (Ref. 14) and in EBR-II at 390, 450, 500, and 550°C up to ~16 dpa (Ref. 15). When irradiated at 300°C in HFIR, all of the steels showed an increase in strength relative to thermally aged steels. The increase was slightly greater for the higher nickel-containing (helium-containing) steels, indicating that part of the strength increase may have been caused by the helium produced during irradiation. If this is true, the strength increase due to helium is in addition to that caused by the dislocation loops and precipitates that form by displacement damage alone. Irradiation at 400°C also caused a strength increase for all of the steels, with an effect of helium again evident.[14] When the results for irradiation in HFIR were compared with specimens irradiated in EBR-II at 390°C, where little helium was produced,[15] it also appeared that helium played a role in that larger strength increases were observed for HFIR irradiation than for EBR-II irradiation. There was no indication that the presence of nickel in the steels had a significant effect on the irradiated properties. After irradiation in HFIR at 500°C, there was little or no change in strength relative to the thermally aged strength. This agreed with the results of the EBR-II-irradiated steels and indicated no strengthening role of helium at this temperature. The change in ductility reflected the strength changes.

As discussed in the Sec. IV, the irradiation hardening of the steels by displacement damage is the result of the dislocation and precipitate structure that forms during irradiation. The HFIR experiments discussed in this section indicate that helium superimposes a further increment of hardening. Mechanisms by which helium might affect strength have been speculated upon.[12,14] Helium in either an interstitial or substitutional position could affect strength, as could the small bubbles and voids that are promoted in the martensitic steels in the presence of helium.[16]

When helium is in an interstitial position, it can readily diffuse through an alloy.[17] However, interstitial helium can become immobilized when trapped by a vacancy into a substitutional position. To the extent that helium preempts vacancies from recombining with mobile self-interstitials, either more self-interstitials will be present at a given time or more self-interstitials will migrate and cluster to form and grow sinks (i.e., loops). Helium vacancy clusters can form at temperatures where both the helium and vacancies are mobile.[18] There are also indications that helium refines the scale of the damage microstructure produced (i.e., loops) and prolongs loop stability to high fluences.[19] Such processes would cause additional hardening as helium builds up. Excess self-interstitials could cause more loop nucleation and/or growth relative to lower helium concentrations. Helium-enhanced loop formation and growth and a buildup of helium vacancy clusters in the matrix could cause additional strengthening. Eventually, these additional sources of hardening would be expected to saturate.

At irradiation temperatures greater than $\sim 0.5 T_m$, displacement damage is no longer stable, and flow properties are basically unaffected by irradiation. In many irradiated alloys that contain helium, the ductility after irradiation decreases when tensile tested at elevated temperatures.[20] Such helium embrittlement leads to intergranular fracture, and the loss of ductility is thought to be caused by helium on grain boundaries.[20] For austenitic stainless steels, the effect occurs with the presence of only a few appm helium—even the small amounts formed during fast reactor irradiation.[20] All indications are that this effect is not important in the martensitic steels,[21] and it will not be discussed further.

VI. IMPLICATIONS OF HARDENING FOR OTHER PROPERTIES

Observations presented earlier indicate that martensitic steels are hardened by irradiation at temperatures below ~450°C. Despite the hardening, the loss of ductility would not appear to eliminate the steels for applications in irradiation environments. However, hardening that causes an increase in yield stress and UTS often causes a loss of toughness, as indicated by Charpy impact tests. An increase in the ductile-brittle transition temperature (DBTT) and a decrease in the upper-shelf energy are observed in tests of steels irradiated at temperatures up to 450°C (Refs. 22, 23, and 24). It is the loss of toughness that is of greatest concern for martensitic steels to be used in nuclear applications. Recent experiments indicate that helium may also play a role in this loss of toughness.[25,26]

Hu and Gelles irradiated 12 Cr-1 Mo-V-W steel at 390°C in EBR-II to 13 and 26 dpa and observed a shift in DBTT (ΔDBTT) of 124 and 144°C, respectively.[23] For 9 Cr-1 Mo-V-Nb steel, shifts of 52 and 54°C were observed after 13 and 26 dpa, respectively. Both observations were taken to mean that the shift had saturated with fluence.[25]

When 12 Cr-1 Mo-V-W steel was irradiated in HFIR at ~400°C, where ~25 appm helium was generated, a ΔDBTT of 195°C was observed.[25] In a second experiment, both 9 Cr-1 Mo-V-Nb and 12 Cr-1 Mo-V-W steel were irradiated in HFIR at 400°C to ~40 dpa, and the results were substantially different from the results obtained in EBR-II (Fig. 6) (Ref. 26). The ΔDBTT for the 9 Cr-1 Mo-V-Nb, which contained ~35 appm helium, was 204°C, while that for 12 Cr-1 Mo-V-W with ~110 appm helium was 242°C. Thus, it appears that the saturation with fluence observed for irradiation in EBR-II does not apply for HFIR irradiation. This difference between HFIR and EBR-II has been tentatively attributed to helium.[25,26] If this is a helium effect, the results indicate that helium has a greater effect on 9 Cr-1 Mo-V-Nb than 12 Cr-1 Mo-V-W; that is, the 12 Cr-1 Mo-V-W contained about three times as much helium as the 9 Cr-1 Mo-V-Nb, but the ΔDBTT values for the two steels were only slightly different. This greater effect of helium in 9 Cr-1 Mo-V-Nb is qualitatively consistent with the yield stress behavior[12] in Figs. 4 and 5 and with swelling data obtained for these steels.[16]

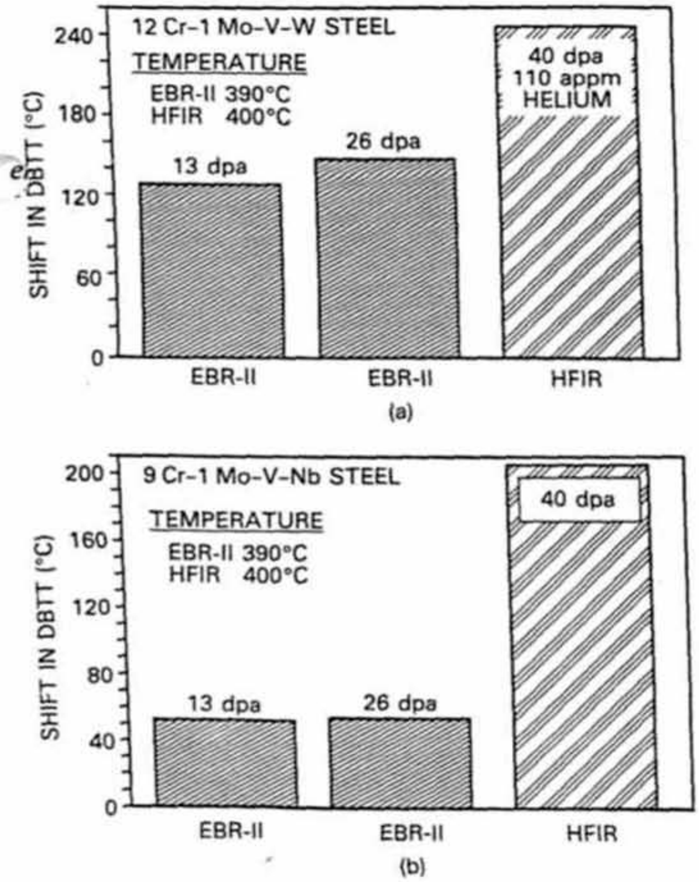

Fig. 6. A comparison of the shift in DBTT for (a) 12 Cr-1 Mo-V-W and (b) 9 Cr-1 Mo-V-Nb steels irradiated in EBR-II and HFIR.

Although this observation should not affect the use of these steels for fast reactor applications, it could be important in a fusion reactor first wall, where large amounts of helium will form. Further tests are required to determine if a saturation in DBTT with fluence occurs in the presence of helium. If a saturation does not occur without a substantial further increase in the DBTT over that observed in the preliminary tests,[25,26] then the use of martensitic steels as structural materials for fusion reactors will need to be reevaluated.

VII. SUMMARY AND CONCLUSIONS

Tensile properties of the martensitic steels are affected by neutron irradiation. Hardening occurs for irradiation temperatures up to ~450°C; the hardening appears to saturate with fluence. At higher temperatures, irradiation has little effect, although irradiation-enhanced diffusion can accelerate the thermal aging process. The presence of helium appears to enhance the irradiation hardening. Although minor losses of ductility accompany the hardening, the effect of irradiation on the tensile properties should not seriously affect the use of these steels for nuclear applications. However, hardening also causes a loss of toughness, as measured by Charpy impact testing. Initial observations indicate that helium enhances this loss of toughness. Furthermore, the saturation with fluence that is observed for the shift in DBTT in the absence of helium does not apply when helium is present. Further studies are required to verify this helium effect and to determine if the martensitic steels will be viable structural materials for fusion reactor first walls, where large amounts of helium will be generated during operation.

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