REDUCED-ACTIVATION BAINITIC AND MARTENSITIC STEELS FOR NUCLEAR FUSION APPLICATIONS—R. L. Klueh (Oak Ridge National Laboratory)*

OBJECTIVE

Since the mid 1980s, research programs have been in progress in Europe, Japan, and the United States to develop reduced-activation steels for fusion applications. This report reviews the recent work in the development of those steels.

SUMMARY

Reduced-activation steels were developed to enhance safety and reduce adverse environmental effects of future fusion power plants. Martensitic and bainitic steels were developed during the 1985–90 timeframe, and the feasibility of their use for fusion was investigated in an international collaboration from 1994 to present. Work continues to improve the steels and understand the effect of neutron irradiation on them.

Introduction

Development programs for materials for future fusion power reactors are being carried out in Japan, the European Union, and the United States. Possible structural materials for the reactor chamber first wall and blanket structure are limited, because the material will operate at elevated temperatures while being exposed to a flux of high-energy neutrons. Ferritic steels were first considered for fusion in the late 1970s, and the first steels considered were commercial 9–12%Cr martensitic steels and 2.25% Cr bainitic steels (all compositions are in wt %) [1].

Beginning in the mid-1980s, fusion reactor materials research programs throughout the world began developing "low-activation" structural materials [2–21]. Low-activation materials were defined as materials that during irradiation would either not activate or any induced radioactivity caused by transmutation of elements in the material by interaction with high-energy (14 MeV) neutrons from the deuterium-tritium fusion reaction would decay rapidly to allow safe operation and hands-on reactor maintenance [2,3]. True low-activation structural materials were not feasible, and "reduced-activation" steels were proposed that would not contain elements that would result in long-lived, transmutation-produced, radioactive elements. This would allow for safer and more economical disposal of radioactive reactor components after service exposure. Radioactivity from reduced-activation steels should decay to low levels in ≈100 y, compared to thousands of years for some non-reduced-activation steels.

The conditions (temperatures, stresses, and environment) reduced-activation steels would be exposed to in an operating fusion reactor are not well established, although simplified design studies have been carried out to investigate various concepts [22]. Coolants for systems in which the ferritic and martensitic steels would be used include helium and liquid lead-lithium (Pb-Li) eutectic alloy. As with all power-generation systems, designers push for high temperatures for increased efficiency. Operating stresses for systems where the steels are generally considered candidates are around 50 MPa, perhaps slightly higher for helium-cooled systems and slightly lower for liquid-metal systems. Under these conditions, the upper operating temperature for the ferritic/martensitic steels would be similar to those of conventional power plants when creep is limiting, about 550–600ºC. Oxidation in the helium system should not be a limiting factor, since oxygen impurities in the helium should be controllable to very low levels. Liquid-metal corrosion by Pb-Li coolant could reduce the upper temperature limit to about 450ºC if the liquid is in direct contact with the steel [23]. Insulating coatings on the exposed steel are proposed to suppress magnetohydrodynamic effects due to the flowing liquid [23]. One proposed design uses silicon-carbide inserts to isolate the steel from the liquid Pb-Li, thus increasing the upper operating temperature back to where creep limits the design stress (550–600ºC) [22].

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The other important environmental condition in a fusion reactor is irradiation. Because of its importance for evaluating steels for fusion, this subject will be discussed in some detail prior to the detailed discussion of the development of martensitic and bainitic reduced-activation steels for fusion applications.

Irradiation Effects on Ferritic/martensitic Steels

High-energy neutron irradiation in a fusion reactor displaces atoms from their normal matrix positions to form vacancies and interstitials. It is the disposition of the “displacement damage,” measured as displacements per atom (dpa), that affects mechanical properties. The progressive change in microstructure of ferritic steels with irradiation dose and temperature involves the agglomeration of vacancies and interstitials into voids and dislocation loops that cause swelling. Loop size increases and loop number density decreases with increasing temperature, eventually becoming unstable around 400–450ºC [1, 24–28]. In ferritic/martensitic steels, agglomeration of vacancies can lead to void swelling up to about 500ºC.

Ferritic steels became of interest because they are low swelling compared to conventional austenitic stainless steels when irradiated in a fast reactor. Besides swelling, the production of vacancies during irradiation can accelerate recovery and precipitation processes (e.g., coarsening) and induce non-equilibrium phases, which can affect properties [1,24–26,28,29].

Transmutation reactions of neutrons with metal atoms produce a new atom (usually another metal atom) with a smaller atomic number and a gas atom—helium or hydrogen. The small numbers of new non-gaseous atoms that form are generally thought not to affect properties, although these atoms are the source of radioactivity that provided impetus for the development of “low-activation” materials. The effect of hydrogen was generally thought to be minimal, because most of it was expected to migrate out of the lattice at reactor operating temperatures. However, recent ion [30] and proton [31] irradiations have produced evidence for retention of considerable amounts of the hydrogen in austenitic stainless steel and martensitic steel, respectively. Recent neutronic calculations, dosimetry analysis, and gas measurements for proton-irradiated martensitic steels suggested that hydrogen effects should not be an issue above 250ºC [32]. Indications are that helium can affect swelling [25], and this will be discussed below. This is important for a fusion reactor, where high helium concentrations will be produced, although ferritic steels will still remain low swelling relative to austenitic stainless steels.

The effect of irradiation on tensile behavior of the 7–12% Cr ferritic/martensitic steels depends on temperature [33–36]. Hardening caused by irradiation-induced dislocation loops and precipitate changes occurs at irradiation temperatures up to 425–450ºC. It is measured as an increase in yield stress and ultimate tensile strength (Fig. 1) and a decrease in ductility [34]. Hardening saturates with increasing fluence, and saturation occurs by 10 dpa [35]. For irradiation above 425–450ºC, properties are generally unchanged (Fig. 1), but there may be radiation-enhanced softening, depending on fluence [34,35].

Irradiation hardening affects other properties, such as fatigue and toughness, with the latter of greatest concern. The effect on toughness is observed in Charpy impact tests as an increase in ductile-brittle transition temperature (DBTT) and a decrease in upper-shelf energy (USE) [37–42] (Fig. 2). The shift in DBTT varies inversely with irradiation temperature and saturates with neutron fluence for irradiation in a fast reactor, where little helium forms (indicated in Fig. 2, which shows the same shift in DBTT after 10 and 17 dpa). Saturation may not occur under conditions where high-helium concentrations form, as discussed below.

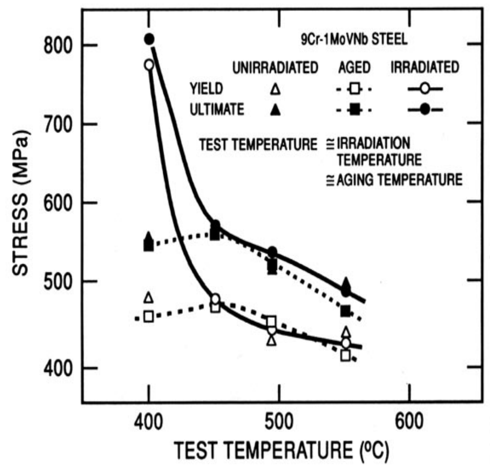

Fig. 1. Yield stress and ultimate tensile strength of normalized-and-tempered, thermally aged, and irradiated modified 9Cr-1Mo (9Cr-1MoVNb) steel. Irradiation was in Experimental Breeder Reactor (EBR-II) to 9 dpa [34].

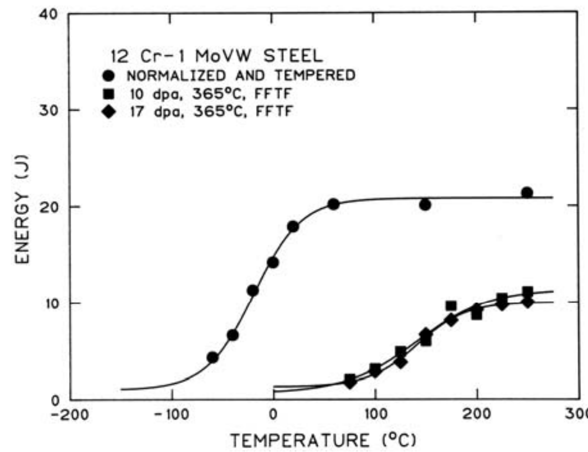

Fig. 2. Charpy curves for half-size specimens of Sandvik HT9 (12Cr-1MoVW) steel before and after irradiation to 10 and 17 dpa at 365ºC in the Fast Flux Test Facility FFTF [35].

Reduced-activation Steels

Martensitic Steels

For steels to meet reduced-activation criteria, calculations indicated that the typical steel-alloying elements Mo, Nb, Ni, Cu, and N must be eliminated or minimized [1–3]. Proposals for reduced-activation ferritic steels involved the replacement of molybdenum in conventional Cr-Mo steels by tungsten [4,6–10] and/or vanadium [7,11]. Niobium was replaced by tantalum.

Research programs in the 1980s in Japan, European Union, and United States developed 7–12% Cr-V and Cr-W-V reduced-activation steels [4–15] to which tantalum are sometimes added as a replacement for niobium [5,7–9,15]. Steels with 7–9% Cr were emphasized over 12% Cr, because of the difficulty of eliminating δ-ferrite in 12% Cr steels without increasing carbon or manganese for austenite stabilization. Nickel is usually used for austenite stabilization (instead of manganese), but it was prohibited by the reduced-activation criteria. Delta-ferrite can lower toughness, and manganese promotes chi-phase precipitation during irradiation, which causes embrittlement [11].

In Japan, nominally Fe-7.5Cr-2.0W-0.2V-0.04Ta-0.10C (F82H) [8,16,17] and Fe-9Cr-2W-0.2V-0.07Ta-0.05N-0.10C (JLF-1) [14,18,19] steels were chosen, and in Europe, an Fe-9.3Cr-1.0W-0.25V-0.04Cu-0.10C (OPTIFER Ia) and Fe-9.4Cr-1.1Ge-0.30V-0.13C (OPTIFER II) were originally chosen and investigated [13,20]. The steel with the best properties in the United States was ORNL 9Cr-2WVTa steel [10,15,21].

Reduced-activation 7–12% Cr steels were patterned after commercial 9–12% Cr steels with the objective that they have properties as good or better than the commercial steels they were to replace. They had microstructures similar to the commercial steels, as seen in Fig. 3, where the ORNL 9Cr-2WVTa steel is compared with modified 9Cr-1Mo steel after which it was patterned [1]. The microstructures consist of tempered martensite, and the predominant precipitates are fairly large $M_{23}C_6$ carbides on prior-austenite grain boundaries and martensite lath (subgrain) boundaries with a few small MX precipitates in the matrix.

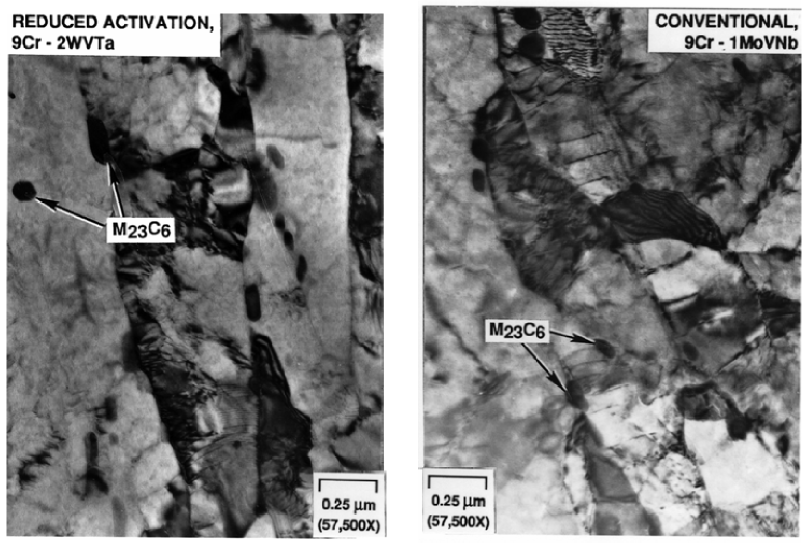

Fig. 3. Electron micrographs of normalized-and-tempered reduced-activation 9Cr-2WVTa and conventional 9Cr-1MoVNb steels [1].

Although all 7–12% Cr conventional and reduced-activation steels irradiated to high displacement damage (>100 dpa) demonstrate the effect of irradiation on toughness (Fig. 2), there are differences among steels, as shown in Fig. 4, where shifts in DBTT for Sandvik HT9 (Fe-12Cr-1Mo-0.5W-0.5Ni-0.25V-0.1C) and ORNL 9Cr-2WVTa are compared after irradiation in a fast reactor at 365ºC [1]. The reduced-activation steel developed much less shift (10ºC vs. 125ºC). Part of this difference was attributed to the larger carbon concentration in HT9 (0.2%) than in 9Cr-2WVTa (0.1%). Modified 9Cr-1Mo has a DBTT shift about half as large as HT9 for similar test conditions, which is still more than twice that for 9Cr-2WVTa. In this case, carbon cannot explain the difference, since these two steels have similar amounts of carbon. The tantalum in 9Cr-2WVTa was shown to have a favorable effect that could be used to explain the difference in these two steels [43].

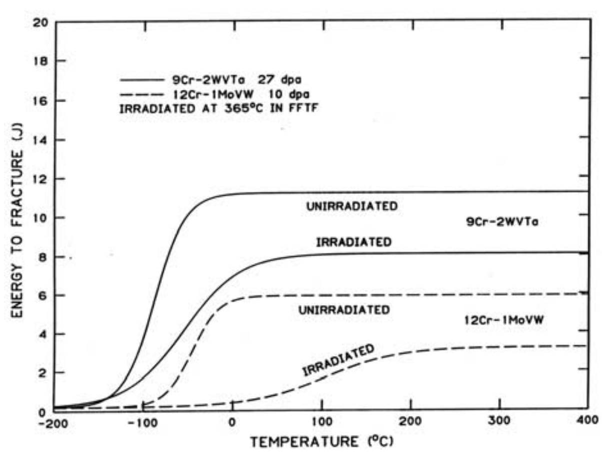

Fig. 4. Comparison of Charpy impact curves for one-third-size specimens of unirradiated and irradiated of HT9 (12Cr-1MoVW) and ORNL 9Cr-2WVTa steels irradiated in FFTF at 365ºC [1].

At an International Energy Agency (IEA) "Workshop for Ferritic/Martensitic Steels for Fusion" in 1992, a proposal was made for an international collaboration to determine the feasibility of using ferritic steels for fusion. The Japanese delegation at the meeting proposed making available large heats of reduced-activation steel for collaboration between Japan, Europe, and the United States under IEA auspices. A modified F82H composition was determined, and two 5-ton heats of this F82H-IEA (Fe-7.5Cr-2W-0.2V-0.04Ta-0.1C) steel were produced by the Japan Atomic Energy Research Institute (JAERI) and NKK Corporation.

Testing of the large heats of F82H-IEA to determine the unirradiated properties was completed during the last two years [44,45]. Physical properties (density, specific heat, thermal expansion, thermal conductivity, Young's modulus, modulus of rigidity, Poisson's ratio, and magnetic hysteresis) and mechanical properties (tensile, creep, Charpy impact, fracture toughness, isothermal fatigue, thermal fatigue, and low-cycle fatigue) were obtained. A computerized database on base metal and welds was developed that is available to the international community on the internet [46].

Besides determining baseline data, F82H-IEA specimens were included in over twenty neutron-irradiation experiments conducted in the High Flux Isotope Reactor (HFIR) in the U.S., in the Japan Research Reactor (JRR-4) and the Japan Materials Test Reactor (JMTR), and the High Flux Reactor (HFR) in The Netherlands. Results were in general agreement with results for other experimental 7–9Cr-WV-type reduced-activation steels that show improved irradiation resistance over conventional Cr-Mo steels [46].

Studies on irradiated F82H-IEA are continuing to determine irradiation effects on a range of properties for base metal and weldments. As discussed above, irradiation embrittlement causes an increase in DBTT in a Charpy test. Charpy data cannot be used for design. Instead, fracture toughness data are required, and such data for irradiated F82H were obtained [47]. In this work, six disk-compact tension [DC(T)] specimens were irradiated in HFIR to ≈3.8 dpa at two temperatures. Three of the specimens in the low-temperature capsule were at an average of 261°C and another three at 240°C. In the high-temperature capsule, all six specimens were irradiated at an average temperature of 377°C. Fracture-toughness-transition temperatures were evaluated, and the master curve concept was used to evaluate the shift of the fracture toughness transition temperature. Specimens irradiated at the higher temperature exhibited a relatively modest shift of about 40°C. However, the shift of fracture-toughness-transition temperature of specimens irradiated below 300°C was much larger—about 180°C. This shift is in general agreement with the DBTT shift observed for F82H Charpy specimens irradiated in HFIR at 300ºC [48].

Reduced-activation steel development began with small experimental heats to determine compositions with mechanical properties as good or better than the Cr-Mo steels they were to replace. Once that was achieved, the large heats of F82H-IEA were used to establish the feasibility of using the steels for fusion. The next step was to use what was learned to produce advanced steels, and the European Union fusion program has taken that step and produced a 3.5-ton heat of a new steel designated EUROFER 97 [49].

Nominal target chemistry for EUROFER (Fe-9.0Cr-1.1W-0.2V-0.07Ta-0.03N-0.11C) differed from that of F82H-IEA (Fe-8.0Cr-2.0W-0.2V-0.04Ta-0.10C). Actual compositions differed from the target composition, and F82H contained less Cr (7.5%), Ta (0.023%), and V (0.14%) than the target composition; the EUROFER contained more tantalum (0.14%) than the target composition [50]. Therefore, major differences in the steels are the Cr, W, and Ta compositions. Lower tungsten was used in the EUROFER specification (1% vs. 2% for F82H) because the tritium breeding rate is higher for the lower tungsten. A higher breeding rate was considered necessary for the reactor design being considered in the European program at the time the steel was ordered. Lower tungsten will also reduce the amount of Laves phase formed relative to steels with higher tungsten. Despite differences in composition, the two steels had similar microstructures and tensile, impact, and creep properties [50–52]. Data on impact properties indicated that the EUROFER has a lower DBTT and a similar to slightly lower USE [50–52].

In addition to studies to determine displacement damage effects, effect of the simultaneous production of displacement damage and transmutation helium in a fusion reactor first wall on swelling and mechanical properties has been a source of uncertainty, and the subject continues to receive attention. In the absence of a 14 MeV (energy of neutrons from the deuterium-tritium reaction) neutron source, simulation experiments using ion implantation [53], boron-doping [54–58], and nickel-doping [42,59] have been used.

Recent results with boron doping indicated increased swelling due to helium [55,56] when F82H specimens with ≈30 and ≈60 ppm natural boron and ≈65 ppm ^{10}B were irradiated in HFIR at 400ºC up to 52 dpa to produce ≈30, ≈60, and ≈330 appm He, respectively (Fig. 5). Helium is produced by an (n,α) reaction where α is an α-particle (^{4}He). The reaction is:

$$^{10}B + n \rightarrow {}^{4}He + {}^{7}Li \qquad (1)$$

where n is the neutron (fast or thermal[†]). Natural boron contains ≈20% ^{10}B.

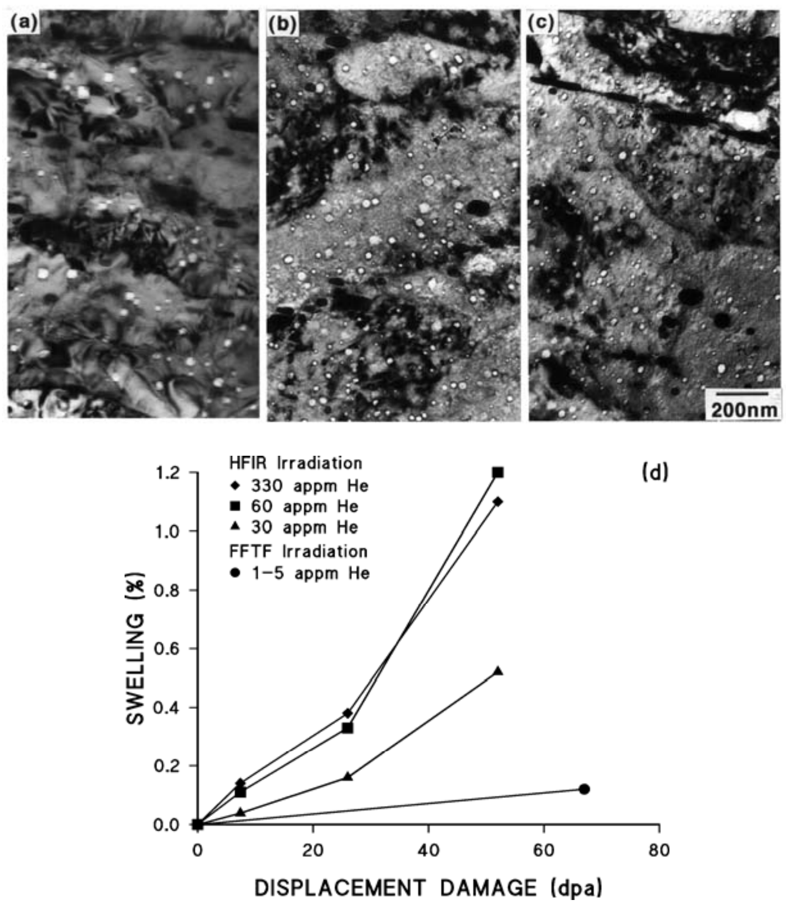

Fig. 5. Cavities in normalized-and-tempered F82H with different boron levels irradiated to 52 dpa in HFIR at 400ºC to produce (a) 30, (b) 60, and (c) 330 appm He. (d) Swelling as a function of displacement damage for the different helium levels for irradiation in HFIR compared to irradiation in the Fast Flux Test Facility (FFTF), where little helium was formed [55].

[†]A thermal neutron is a free neutron with kinetic energy <0.025 eV (≈4.0x10^{-21} J)—the average kinetic energy of a room-temperature gas. A fast neutron is a free neutron with kinetic energy >1MeV. Fast neutrons are produced by nuclear processes such as nuclear fission and fusion.

Cavity size, number density, and swelling after 52 dpa for the three steels were respectively 12.7 nm, 6.1×10^{20} m^{-3}, and 0.52%; 10.6 nm, 2.4×10^{21} m^{-3}, and 1.2%; and 7.6 nm, 6.1×10^{21} m^{-3}, and 1.1%. Swelling of 1.2 and 1.1% after only 52 dpa for the steels with 60 and 330 appm He, respectively, is much larger than generally expected for these steels. The higher number density of smaller cavities for the steel with 330 appm He than the steel with only 60 appm He evidently suppressed swelling by acting as neutral sinks for vacancies and interstitials compared to the lower number density of larger cavities in the steel with 60 appm He. Most previous high-dose data that showed low swelling came from fast reactor irradiation (little helium generated), and as seen in Fig. 5(d), the swelling of F82H irradiated in the Fast Flux Test Facility (FFTF) to 67 dpa (<5 appm He) was about an order of magnitude less (0.12%) [60].

All simulation techniques have problems. Boron, for example, is a reactive element that can be associated with precipitates and prior austenite grain boundaries; also, lithium from the transmutation of ^{10}B could cause problems. Furthermore, during fission reactor irradiation, all the boron is quickly transformed into helium, which differs from what happens in a fusion reactor, where helium forms simultaneously and more gradually with the displacement damage.

When nickel-doped steels are irradiated in a mixed-spectrum reactor such as HFIR, displacement damage and helium form simultaneously. Helium forms by a two-step (n,α) reaction of ^{58}Ni with thermal neutrons in the mixed spectrum: displacement damage forms from fast neutrons in the spectrum [1]. The reactions are:

$$^{58}Ni + n \rightarrow {}^{59}Ni + \gamma \tag{2}$$

$$^{59}Ni + n \rightarrow {}^{56}Ni + {}^{4}He \tag{3}$$

where γ is a gamma ray. This technique also has problems [61]. In recent work, a 9Cr-2W reduced-activation steel with and without 1% Ni was irradiated to 2.2 and 3.8 dpa at 270 and 348ºC, respectively, in the Advanced Test Reactor (ATR), a fast reactor where little helium forms.‡ The nickel-containing steel hardened about 20% more than the steel without nickel at 270ºC, but strengths were similar after irradiation at 348ºC [61]. Likewise, there was a larger shift in DBTT for the nickel-containing steel than the one without nickel when irradiated at 270ºC, but not after irradiation at 348ºC. TEM analysis indicated nickel refined the size of defect clusters, which were more numerous in the nickel-containing steel [61].

These results indicated that nickel doping should be used with caution below ≈300ºC. Results with nickel doping that most strongly indicated helium caused an increase in DBTT above that caused by displacement damage alone and without a saturation with fluence were high-dose tests at 400ºC [40]. Tensile tests of specimens irradiated in a fast reactor gave no indication of hardening due to helium (or nickel). TEM studies of nickel-doped steels irradiated in HFIR and FFTF showed a high density of M_6C formed in the nickel-doped steel but not in the undoped steel [25]. Since the DBTT shift in HFIR, where helium forms, was larger than in FFTF, where little helium forms, the results indicated that helium caused the shift [42,59].

‡A fast reactor neutron spectrum does not contain a significant number of thermal neutrons relative to a mixed-spectrum reactor, which contains both fast and thermal neutrons, nor does it contain the high-energy 14.1 MeV neutrons of a fusion spectrum. The average creation energy of neutrons in a fast reactor is about 2 MeV, and the energy spectrum contains neutrons mainly in the range above about 10 keV. Little transmutation helium is produced in ferritic steels by neutrons in that energy range.

Another method used recently to study helium effects is irradiation with 600–750 MeV protons in PIREX (Proton Irradiation Experiment) and SINQ (Swiss Spallation Neutron Source) at the Paul Scherrer Institute in Switzerland [62,63]. Microstructural observations indicated that during the simultaneous production of displacement damage and helium, helium is distributed in the pre-existing cavities produced by displacement damage [62]. Therefore, helium does not produce hardening above that due to displacement damage alone. Finally, irradiation in SINQ of Charpy specimens of F82H, modified 9Cr-1Mo, and OPTIFER produced an increase in DBTT that had a linear dependence on helium concentration up to 600 appm He, demonstrating the importance of helium on embrittlement [63]. For the irradiation conditions examined, there was no indication of saturation with dose.

These recent results on helium effects on swelling [57,58,62] and toughness [63] of reduced-activation steels are disturbing, given the large amount of helium that will form in a fusion reactor first wall. Furthermore, the observations agrees with those in nickel- [42,59] and boron-doping [56] experiments that indicated increased swelling with increased helium and an increase in DBTT with increased helium above that observed by displacement damage alone with no increase in strength attributable to helium. More work on helium effects is required to validate the ferritic/martensitic steels for fusion applications.

Bainitic Steels

Although reduced-activation 7–9% Cr martensitic steels were eventually chosen for further development, low-chromium bainitic steels were also studied [4, 5,11,14,15] during the original development phase. Steel development programs in the United States and Japan examined chromium concentrations over the range 2.25 to 12% [11,14,15]. An Fe- 2.25Cr-2.0W-0.25V-0.07Ta-0.1C (2¼Cr-2WVTa) steel had a higher strength than the ORNL 9Cr-2WVTa steel for similar heat treatment conditions [21], but the 9Cr-2WVTa steel had better toughness [37]. Similar results were obtained in Japan [14]. In this case, irradiation in the JMTR resulted in essentially no shift in DBTT for the 2.25Cr steel compared to a shift of 22ºC for the 9Cr steels. However, the 9Cr steels still had the lowest DBTT after irradiation because of the lower value before irradiation [14].

The Oak Ridge National Laboratory (ORNL) reduced-activation steel development program was the only one that pursued bainitic steels beyond the first iteration [64,65]. Prior to work on reduced-activation steels, work at ORNL demonstrated that quenched-and-tempered plates of an Fe-3Cr-1.5Mo-0.25V-0.1C steel unexpectedly had better Charpy properties than normalized (air-cooled)-and-tempered plates [66]. Different tempered bainitic microstructures were observed (Fig. 6), and they were related to microstructures observed by Habraken and Economopoulos (H&E) [67], who found morphological variations in the bainite transformation products that differed from classical upper and lower bainite, although they formed in the bainite transformation temperature regime.

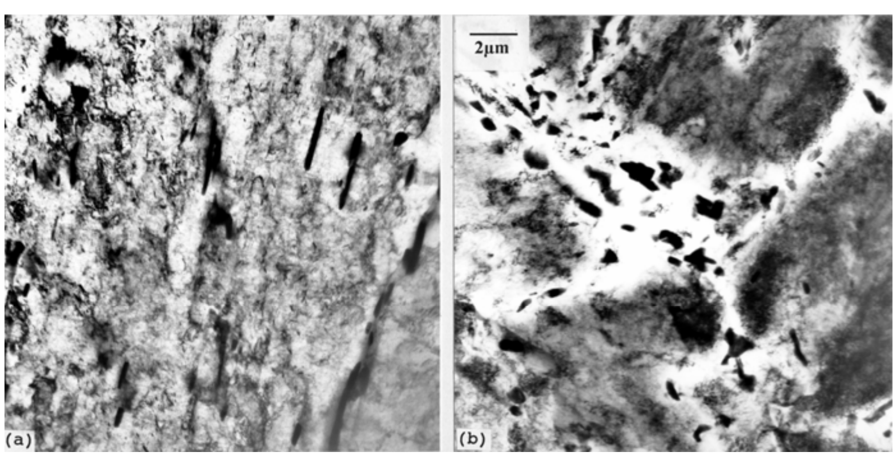

Fig. 6. Microstructures of (a) quenched-and-tempered and (b) normalized-and-tempered 3Cr-1.5MoV steel [64].

The bainite transformation region of an isothermal-transformation diagram (≈250–550ºC) 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 that formed during continuous cooling, H&E showed that a continuous cooling transformation diagram could be divided into three vertical regimes (Fig. 7). A steel cooled through zone I produced a "carbide-free acicular" structure, consisting of side-by-side plates or laths. When cooled through zone II, a carbide-free "massive or granular" structure resulted, referred to as granular bainite. It was determined that granular bainite consists of a bainitic ferrite matrix with a high dislocation density containing martensite-austenite (M-A) "islands" enriched in carbon during formation of the bainite [67]. Microstructures formed in zone III are not relevant to this discussion.

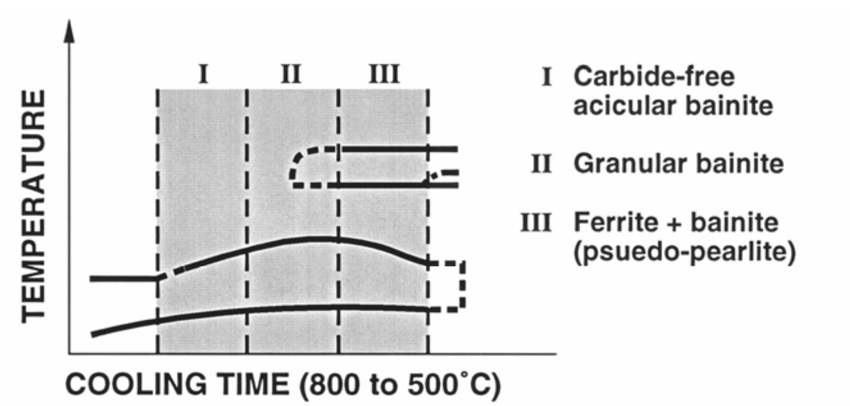

Fig. 7. Schematic representation from Habraken and Economopoulos of a continuous-cooling transformation diagram with cooling zones for the formation of three morphological variations of bainite during continuous cooling.

The microstructure of slowly cooled reduced-activation 2¼Cr-2WVTa [Fig. 8(a)] steel was characteristic of granular bainite; the dark areas are the M-A islands. The microstructure of the specimen cooled rapidly [Fig. 8(b)] was characteristic of carbide-free acicular bainite, as defined by H&E [67]. When granular bainite is tempered, large globular carbides form in the high-carbon M-A islands [Fig. 6(b)], whereas elongated carbides form on lath boundaries of acicular bainite [Fig. 6(a)] [65]. It is these different carbide microstructures that give rise to the different Charpy properties (the globular carbides of the normalized-and-tempered steel lower the toughness relative to the quenched-and-tempered steel) [66].

If the understanding of the effects of cooling rate on microstructure is correct, then toughness of the bainitic steels could be improved by cooling rapidly—quenching instead of normalizing (air cooling). It was reasoned, therefore, that another possibility to promote carbide-free acicular bainite is by improving hardenability [65]. Hardenability is the relative ability of a steel to avoid forming the soft ferrite phase when 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 chemical composition, and based on the 2¼Cr-2WVTa having excellent tensile properties but reduced toughness after normalizing and tempering, the composition was varied to demonstrate how the properties could be improved based on the above reasoning [65]. This eventually led to two nominal compositions that have been investigated in some detail: Fe-3.0Cr-3.0W-0.50Mn-0.25Si-0.25%V (3Cr-3WV) and this composition with 0.1% Ta (3Cr-3WVTa).

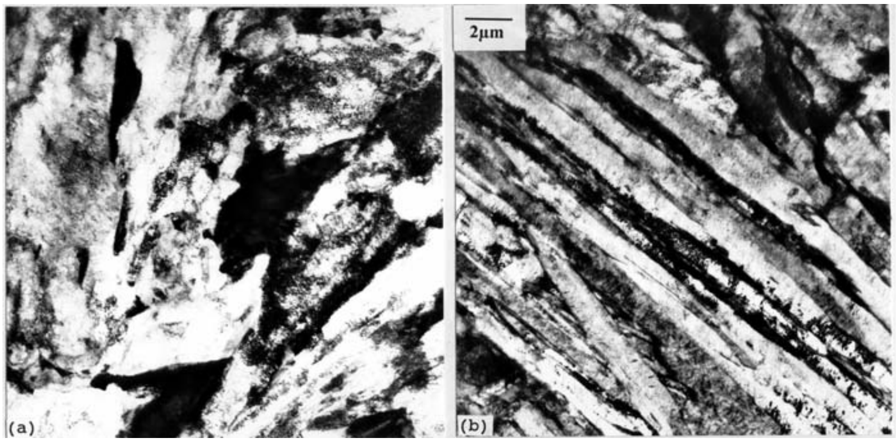

Fig. 8. Microstructure of a 2¼Cr-2WVTa steel after (a) a slow cool to develop granular bainite and after (b) a fast cool to produce carbide-free acicular bainite.

The 3Cr-3WV and 3Cr-3WVTa steels have strengths from room temperature to 600ºC that exceed those of the advanced commercial 2.25Cr steels T23 (Fe-2.25Cr-1.6W-0.25V-0.05Nb-0.07C) and T24 (Fe-2.25Cr-1.0Mo-0.25V-0.07Ti-0.005B-0.07C). The 3Cr-3WVTa also has strength approaching and exceeding that of some of the high-chromium steels such as F82H and modified 9Cr-1Mo. A similar advantage is exhibited during creep at 600ºC [Fig. 9(a)]. Creep tests at 650ºC [Fig. 9(b)] showed that the 3Cr-3WVTa steel has properties comparable to modified 9Cr-1Mo (T91) (no data were available for T23 at this temperature).

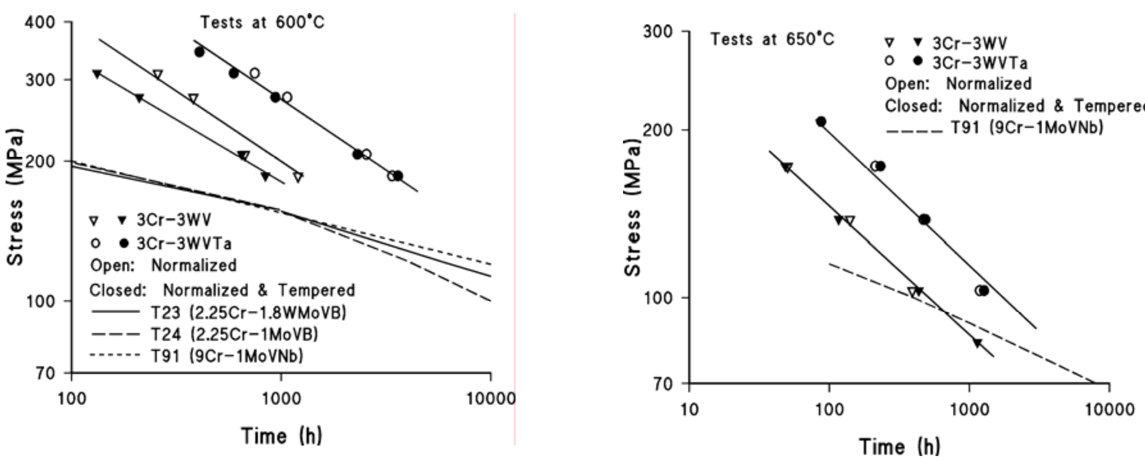

Fig. 9. Creep-rupture curves for 3Cr-3WV and 3Cr-3WVTa steels in the normalized and normalized-and-tempered conditions (a) at 600ºC compared to T23 (2.25Cr-1.6WVNb), T24 (2.25Cr-1MoVTi), and modified 9Cr-1Mo (9Cr-1MoVNb) steels and (b) at 650ºC compared to modified 9Cr-1Mo steel.

Elevated-temperature strength in these steels is obtained from a bainitic microstructure with a high number density of fine MX precipitates in the matrix (Fig. 10). Both the 3Cr-3WV and 3Cr-3WVTa steels contain needle-like precipitates, but the precipitates are considerably finer in the latter steel, indicating an effect of the tantalum. Indications are that during creep, coarsening of these fine matrix precipitates is more rapid in the steel without tantalum.

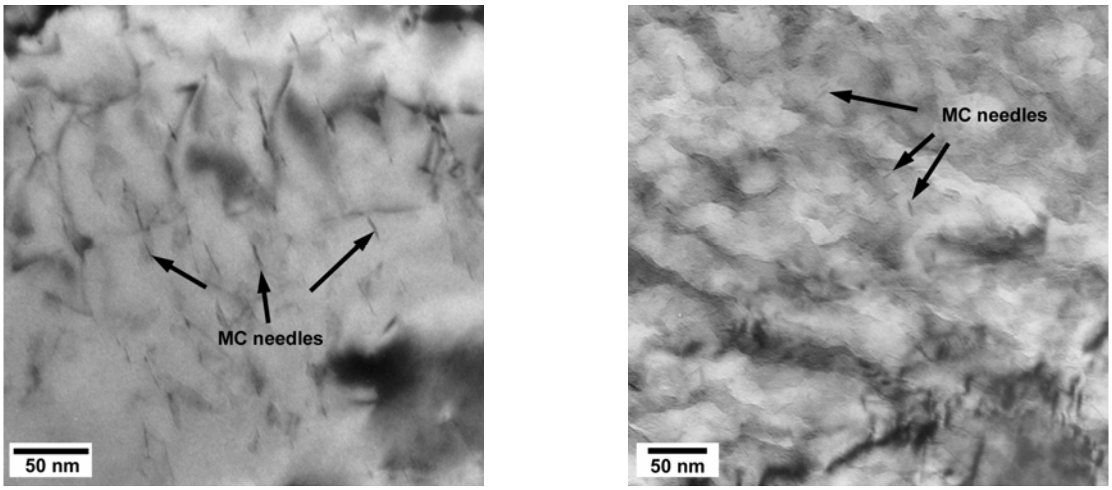

Fig. 10. Photomicrographs of (a) 3Cr-3WV and (b) 3Cr-3WVTa steels showing the fine needle precipitates that provide creep strength.

Conclusions

Reduced-activation steels for fusion applications were developed as replacements for conventional steels for reduced radioactivity for waste disposal after service. New 7–9% Cr martensitic steels with molybdenum replaced by tungsten and niobium replaced by tantalum were developed with microstructures and properties equivalent to or better than the conventional steels they replaced. Emphasis was on 7–9% Cr martensitic steels because 9–12% Cr conventional Cr-Mo steels had been considered primary candidate materials for fusion prior to the development of reduced-activation steels.

Although 2–3% Cr bainitic steels were studied in the early stages of the development program, with one exception, they were abandoned for high-chromium steels. A nominally Fe-3.0Cr-3.0W-0.25V-0.10Ta-0.10C bainitic steel was developed with excellent high-temperature strength. Depending on the design temperatures for fusion reactors, a low-chromium steel could offer advantages, especially for fabrication of the complicated structures envisioned. Low-chromium steels are more easily welded, since the welds and heat-affected zones will not contain hard and brittle martensite.

Reactor designers may push to higher operating temperatures for increased efficiency. The original conventional steels (e.g., modified 9Cr-1Mo) considered were developed for fossil-fired power plants with a maximum operating temperature of 550-600ºC. Reduced-activation steels were patterned after these steels and generally have creep and tensile properties similar to modified 9Cr-1Mo. In recent years, improved conventional steels for fossil-fired plants have been developed with maximum temperatures of 600-630ºC [68]. Therefore, methods used to improve the conventional steels need to be examined to determine if similar improvements can be made in reduced-activation steels.

In addition to the elevated-temperature requirements, steels for fusion also require resistance to high-energy neutron irradiation. Martensitic and bainitic steels appear sufficiently resistant to radiation damage in the absence of large amounts of helium. Because high-energy neutrons produced by the fusion reaction will also produce large amounts of transmutation helium, it is important to understand the effect of simultaneous helium and displacement damage production in a steel reactor first wall. There are indications that helium can affect toughness of martensitic and bainitic steels, and more work is required to understand that potential problem.

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