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Reduced-activation steels: Future development for improved creep strength

R.L. Klueh *

Oak Ridge National Laboratory, P.O. 2008, MS 6138, Oak Ridge, TN 37831-6138, United States

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ABSTRACT

Reduced-activation steels for fusion applications were developed in the 1980s to replace the elevated-temperature commercial steels first considered. The new steels were patterned after the commercial steels, with the objective that the new steels have yield stress and ultimate tensile strength and impact toughness in a Charpy test comparable to or better than the steels they replaced. That objective was achieved in reduced-activation steels developed in Japan, Europe, and the United States. Although tensile and impact toughness of the reduced-activation steels exceed those of the commercial steels they were patterned after, their creep-rupture properties are inferior to some commercial steels they replaced. They are even more inferior to commercial steels developed since the 1980s. In this paper, compositional differences between reduced-activation steels and new commercial steels are examined, and compositions are proposed for development of new-and-improved reduced-activation steels.

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1. Introduction

Development of reduced-activation ferritic/martensitic steels for first wall and blanket structures of future fusion reactors began in the mid-1980s. The objective was that when a reduced-activation steel was irradiated in a fusion neutron environment, nuclear transmutation of alloying elements in the steel would produce relatively short-lived radioactive isotopes. This rapid radioactivity decay would allow reactor components constructed from reduced-activation steels to be disposed of by shallow land burial when removed from service, instead of by the much more expensive deep geologic storage. According to the original investigations, reduced-activation status could be achieved if the typical alloying elements Ni, Mo, Nb, Cu, and N were eliminated or restricted [1]. These conclusions were based on US Nuclear Regulatory Commission Guidelines 10CFR Part 61 that were developed for fission reactor waste. With respect to nitrogen, subsequent calculations based on updated activation cross sections and decay data libraries demonstrated that limits on nitrogen are not as restrictive as the original calculations indicated [2–4].

The elevated-temperature ferritic/martensitic steels are used in the normalized-and-tempered condition. Normalization involves austenitization at 1000–1075 °C followed by an air cool to produce martensite or bainite, depending on the composition; normalization of 5–9% Cr steels will produce martensite, while 2–5% Cr steels will produce bainite – and in some cases ferrite, depending on the chemical composition and size of the cross section heat treated. Instead of air cooling, water or oil quenching is sometimes used, leading to a quenched-and-tempered steel. Tempering is at 700–790 °C to produce adequate ductility and toughness.

Before the development of reduced-activation steels, commercial elevated-temperature Cr–Mo steels were considered for fusion applications. The first commercial steels considered in the international fusion materials programs beginning in the late 1970s were tempered martensitic steels that had been investigated the previous decade in the international fission fast breeder reactor programs [5]. These steels, which were developed by the steel industry for the conventional power-generation and petrochemical industries, had a range of compositions that included 2.25–12% Cr, 1–2% Mo, 0–0.3% V, 0–0.5% W, 0–0.5% Ni, 0–0.2% Nb, 0.1–0.15% C, 0–0.07% N (all compositions are in wt%). Steels chosen for the international fusion programs were EM-12 in France, FV448 in the United Kingdom, DIN 1.4914 in Germany, JFMS in Japan, and Sandvik HT9, modified 9Cr–Mo, and 2¼Cr–1Mo in the United States. Compositions are given in Table 1.

Reduced-activation steels were developed in Europe, Japan, and the United States [5–14]. The development strategy involved patterning the new steels after the commercial steels they were to replace. Generally, this involved replacing molybdenum in the commercial steels by tungsten, niobium by tantalum, and nickel by manganese on an atom-for-atom basis. Eventually, the international steel development programs settled on 7–9% Cr steels, and F82H [7–9], EUROFER 97 [10–12], and ORNL 9Cr–2WVTa [6,13,14] steels were developed in Japan, Europe, and the United States, respectively. Nominal compositions of these reduced-activation steels are given in Table 2. In both Japan and Europe, multi-ton heats of their steels were produced, different product forms fabricated, and detailed test programs conducted.

Irradiation effects experienced by conventional and reduced-activation ferritic/martensitic steels when irradiated with neutrons

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* Tel.: +1 865 574 5111; fax: +1 865 241 3650.
E-mail address: kluehrl@ornl.gov

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Table 1
Compositions of selected commercial steels (wt%)

Steel	C	Cr	Mo	V	Nb	W	Ta	Mn	Si	Ni	N	B
Mod 9Cr–1Mo	0.086	8.44	0.89	0.24	0.08			0.37	0.16	0.11	0.054	–
Sandvik HT9	0.21	12.11	1.03	0.33		0.53		0.50	0.21	0.58	0.004	–
EM12	0.086	9.58	1.91	0.28	0.41			0.92	0.37	0.12		–
FV448	0.10	10.7	0.60	0.14	0.26			0.86	0.46	0.65	0.050	–
DIN 1.4914	0.15	11.0	0.50	0.30	0.25			0.40	0.50	0.75	0.030	–
E911	0.11	9.0	1.0	0.20	0.08	1.0		0.45	0.20	0.20	0.06	–
NF616	0.07	9.0	0.50	0.20	0.05	1.80		0.45	0.06		0.06	0.005

Table 2
Compositions of reduced-activation steels (wt%)

Steel	C	Cr	V	W	Ta	Mn	Si	Ni	N	B
F82H	0.093	7.50	0.14	2.01	0.02	0.17	0.16	0.02	0.006	0.003
EUROFER 97	0.10	8.82	0.19	1.10	0.07	0.37	0.01	0.02	0.021	0.005
9Cr–2WVTa	0.11	8.90	0.23	2.01	0.06	0.44	0.21	<0.01	0.022	–

are similar. The major effect for both types of steel is hardening at irradiation temperatures below 425–450 °C, depending on composition [5]. Hardening is measured as an increase in yield stress (YS) and ultimate tensile strength (UTS), and it causes embrittlement, which is measured in a Charpy impact test as an increase in the ductile–brittle transition temperature (DBTT) and a decrease in upper-shelf energy (USE). Therefore, irradiation resistance to embrittlement was emphasized in the development of the reduced-activation steels, with the objective of developing steels with irradiation resistance as good as or better than those of the commercial steels that were being replaced.

Extensive irradiation-effects studies were conducted on the conventional and reduced-activation steels. These investigations have been reviewed [5], and they will not be discussed here. In this paper, mechanical properties of unirradiated and irradiated reduced-activation steels will be compared to the commercial steels they replaced. This will be followed by examination of the mechanical properties of new commercial steels developed in recent years for non-nuclear power plant applications. These new steels will also be compared to the reduced-activation steels, especially the creep properties. Finally, based on literature information derived from experience gained in developing the latest commercial steels, computational thermodynamics calculations will be presented that lead to a proposal for compositions for new and improved reduced-activation steels.

2. Comparison of reduced-activation and commercial steels

For comparison of the commercial and reduced-activation steels, the steels in the US fusion materials program will be used. Microstructures, mechanical properties, and radiation resistance of the US reduced-activation and commercial steels are similar to steels in the fusion materials programs in Japan and Europe. Although some work in the United States on commercial steels prior to development of reduced-activation steels was carried out on bainitic 2¼Cr–1Mo steel, most work was on martensitic Sandvik HT9 and modified 9Cr–1Mo steels (Table 1), because of their greater strength due to the presence of the strong carbide-forming element V in both and Nb and N in modified 9Cr–1Mo. The latter steel was developed in the 1970s in the US fast breeder reactor program for steam generator applications, and in the 1980s-1990s, it became a preferred steel for replacement of 2¼Cr–1Mo in conventional power plants throughout the world [15–17].

When ferritic/martensitic steels were introduced into the US fusion materials program, Sandvik HT9 was the primary candidate, because of the large amount of irradiation-effects data collected during the fast breeder reactor program. In early irradiation studies in the US fusion program, modified 9Cr–1Mo was shown to be superior to HT9 in most respects [5]. Although room-temperature strength (YS and UTS) of normalized-and-tempered HT9 is greater than that of modified 9Cr–1Mo [18], the creep-rupture properties of the modified 9Cr–1Mo are superior (Fig. 1). The improved elevated-temperature properties of the modified 9Cr–1Mo are attributed to the presence of the combination of 0.2–0.25%V, 0.05% Nb, and 0.05% N. Although HT9 contains 0.25% V, the combination of the three elements in modified 9Cr–1Mo, especially the nitrogen, produce higher precipitation hardening at elevated temperatures [19].

Since irradiation embrittlement causes an increase in DBTT (ΔDBTT), normalized-and-tempered properties are important (i.e., a lower DBTT in the unirradiated condition is conducive to a lower post-irradiation DBTT). Again, modified 9Cr–1Mo has the better properties. Typical USE and DBTT values for HT9 and modified 9Cr–1Mo for 1/3-size Charpy specimens are 6.0 J and −32 °C and 10.5 J and −64 °C, respectively [20], with the difference attributed to the higher carbon concentration in HT9 [5]. Higher carbon also makes HT9 more difficult to weld, an important consideration, since welding is required in construction of a fission or fusion plant.

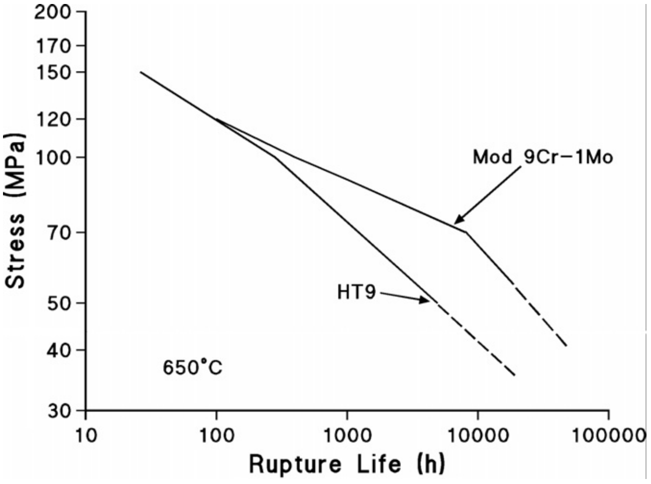

Fig. 1. Creep-rupture curves at 650 °C for Sandvik HT9 and modified 9Cr–1Mo steels.

Although the two steels exhibited a similar amount of hardening when irradiated in the Experimental Breeder Reactor EBR-II, the modified 9Cr–1Mo steel also had better resistance to radiation embrittlement. After irradiation to 13 and 26 dpa at 390 °C, the ΔDBTT was, respectively, 52 and 54 °C for modified 9Cr–1Mo and 124 and 144 °C for HT9 [21]. Similar differences were observed in other fast-reactor irradiations at 300–400 °C [5,20].

The objective for development of reduced-activation steels was that they have strength (in a tensile test) and impact toughness (in a Charpy test) as-good-or-better than the HT9 and modified 9Cr–1Mo steels they were to replace. This proved to be the case for the final reduced-activation steels developed in the international programs–EUROFER 97, F82H, and ORNL 9Cr–2WVTa. Irradiation resistance to embrittlement was also better, as seen in Fig. 2, where Charpy curves for 1/3-size specimens of HT9 and ORNL 9Cr–2WVTa are compared in the normalized-and-tempered condition and after irradiation at 365 °C in the fast flux test facility (FFTF) [5]. The 9Cr–2WVTa is also better than modified 9Cr–1Mo, primarily because of the much lower DBTT in the unirradiated condition. For 1/3-size Charpy specimens, DBTT values are ≈−64 °C [20] and −88 °C [22] for the modified 9Cr–1Mo and 9Cr–2WVTa, respectively. After irradiation in FFTF to ≈20 dpa at 365 °C, the DBTT values were −19 and −67 °C (ΔDBTTs of 45 and 21 °C), respectively, indicating an advantage for the reduced-activation steel. Based on the strength and toughness criteria set forth, reduced-activation steels developed in the international programs successfully met the objectives for the development efforts.

In recent years, emphasis of fusion reactor designers has been directed toward improved reactor efficiency, which would require higher operating temperatures. The reduced-activation steels are projected to have an upper-use temperature of ≈550 °C [23]. This temperature is determined by creep strength, and in this respect, steels such as EUROFER 97 and F82H, for which extensive creep data are available, are not as strong as some of the commercial steels they were patterned after. This is demonstrated in Fig. 3, where creep-rupture properties for the two reduced-activation steels are compared to HT9 and modified 9Cr–1Mo.

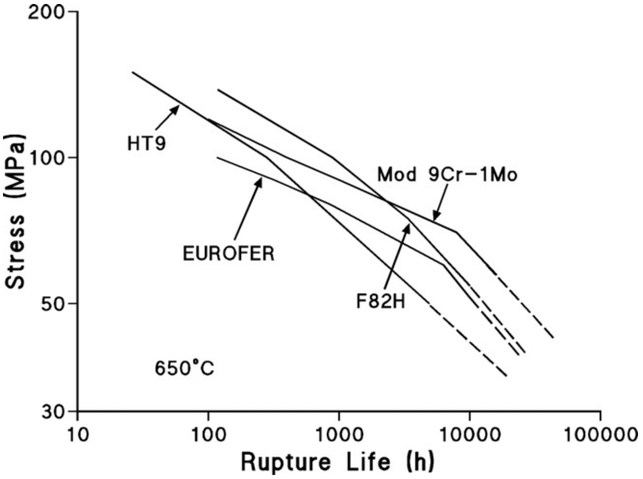

Fig. 3. A comparison of the creep-rupture curves at 650 °C for the reduced activation steels F82H and EUROFER 97 with the commercial steels Sandvik HT9 and modified 9Cr–1Mo.

The difference in creep-rupture behavior of the reduced-activation and commercial steels can be understood based on compositions. As noted above, the difference in the creep-rupture properties of HT9 (12Cr–1MoVW) and modified 9Cr–1Mo (9Cr–1MoVNb) is due to the presence of nitrogen and niobium in the latter. The 7–9% Cr reduced-activation steels were modeled after modified 9Cr–1Mo with molybdenum replaced by tungsten and niobium replaced by tantalum, the replacement elements being chosen because they are in the same columns of the periodic table of the respective elements and have similar effects on the mechanical properties in steels to those of the elements they replaced [6]. Therefore, the major difference between reduced-activation steels and modified 9Cr–1Mo is the nitrogen. The ORNL 9Cr–2WVTa, for example, contains ≈0.025% N compared to 0.05% N in modified 9Cr–1Mo. EUROFER 97 and F82H contain even less nitrogen than

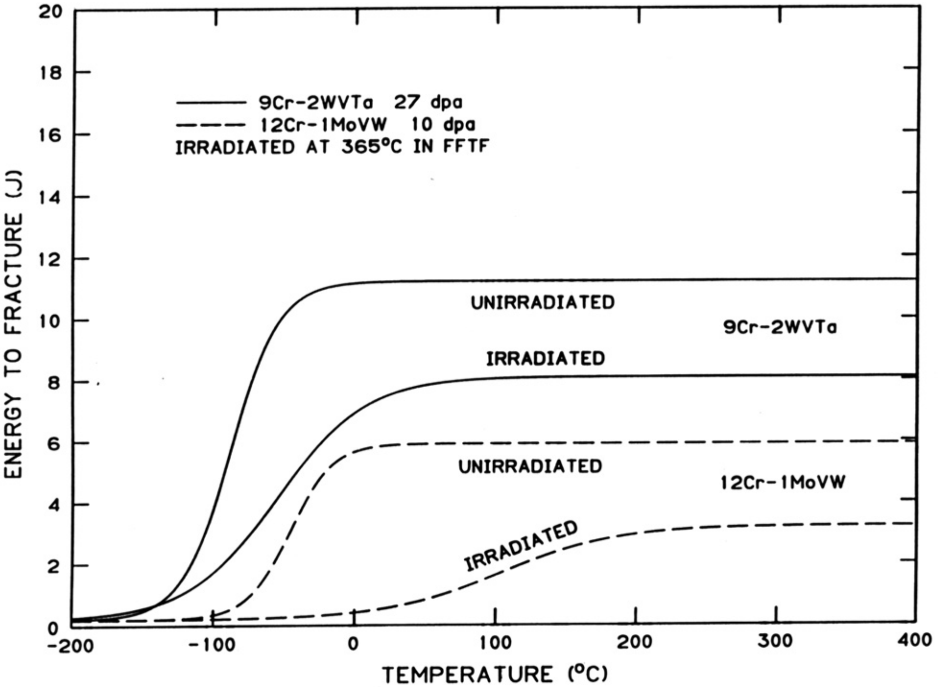

Fig. 2. Charpy impact curves for Sandvik HT9 steel and reduced-activation 9Cr–2WVTa steel in the normalized-and-tempered condition and after irradiation at 365 °C in the fast flux test facility (FFTF).

9Cr–2WVTa. The upper operating temperature for modified 9Cr–1Mo in conventional power plants is ≈593 °C, compared to ≈550 °C for HT9 [15–17]. From these observed differences, it appears elevated-temperature creep strength of reduced-activation steels can be improved. It must be improved if operating temperatures of fusion power plants are to be increased and ferritic/martensitic steels are to be used as structural materials.

3. Improved commercial steels

Steels such as HT9, EM12, and JFMS (Table 1) that were the first steels considered in the international fusion programs in the late 1970s were developed in the 1960s or earlier by the steel industry for conventional power plant applications. The improved properties of modified 9Cr–1Mo relative to these steels demonstrated the success of development by selected changes in composition. Because of the improved properties of normalized-and-tempered modified 9Cr–1Mo developed for fission applications, it subsequently became the choice for conventional power plants in the 1990s that were designed for higher temperatures (up to 590 °C) than the preceding plants [15–17].

Steel development programs in the steel industry from 1985 to 1995 sought improved properties compared to modified 9Cr–1Mo, and new steels were developed that were modifications of modified 9Cr–1Mo. A steel developed in Europe, known as E911, involved the addition of 1% W to the modified 9Cr–1Mo composition. In Japan, NF616 steel was developed by replacing half of the molybdenum with 1.8% W and adding ≈0.004% B (see Table 1). The creep-rupture strength of NF616 is a significant improvement over modified 9Cr–1Mo and, in turn, over the reduced-activation steels and HT9 (Fig. 4). The 10^5-h rupture strength at 600 °C for NF616 is ≈140 MPa compared to ≈90 MPa for modified 9Cr–1Mo. The upper-use temperature for NF616 in conventional power plants is 620 °C compared to 590 °C for modified 9Cr–1Mo [15].

4. Creep-strengthening mechanisms

Creep strength of normalized-and-tempered 7–9% Cr steels is determined by the $M_{23}C_6$ and MX precipitates formed during tempering and by solid-solution strengthening attributed mainly to molybdenum and/or tungsten [24–26]. The $M_{23}C_6$ forms on prior-austenite grain boundaries and on martensite lath boundaries, and MX forms mainly in the matrix. The $M_{23}C_6$ on lath (subgrain) boundaries stabilizes the subgrain structure. The small MX particles, which are a combination of vanadium-rich nitrides and niobium-rich carbonitrides, provide precipitation strengthening during creep [24–26].

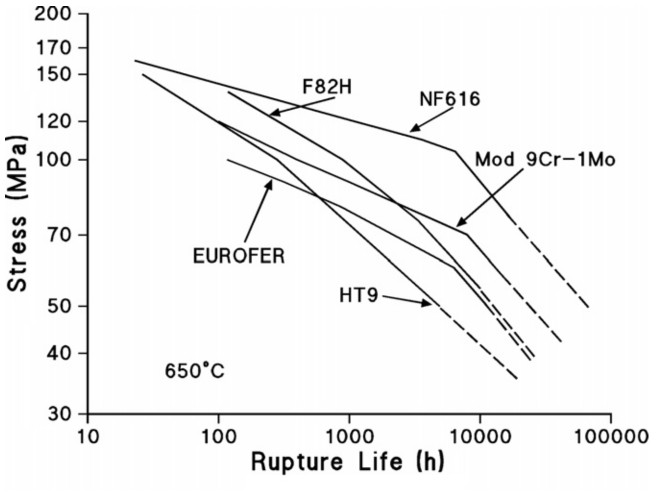

Fig. 4. A comparison of the creep-rupture curves at 650 °C for the reduced activation steels F82H and EUROFER 97 with the commercial steels Sandvik HT9, modified 9Cr–1Mo, and NF616.

With time at elevated temperatures (in a creep test or during service), Ostwald ripening causes precipitate coarsening. Coarsening of the $M_{23}C_6$ particles on subgrain boundaries reduces their pinning effect, thus destabilizing the subgrain structure, and subgrain growth reduces creep strength. Similarly, coarsening of MX reduces the primary precipitate-strengthening effect, although MX coarsening does not occur as rapidly as for $M_{23}C_6$. In addition, Laves phase [Fe_2(Mo,W)] forms, which removes molybdenum and/or tungsten from solution, reducing their solid-solution hardening effect. Once the effects of precipitate strengthening and strengthening due to the subgrain structure are exhausted by precipitate coarsening, creep strength for long-time (low-stress) elevated-temperature exposure is determined by the solid-solution strength.

One difference between NF616 and modified 9Cr–1Mo, in addition to tungsten, is the presence of boron (0.005%) in NF616. Boron enters the $M_{23}C_6$ lattice to form $M_{23}(C,B)_6$ [25–28], and it has been credited with stabilizing the $M_{23}C_6$, thus decreasing the rate of coarsening at elevated temperatures [25–28]. Further, by stabilizing $M_{23}C_6$, boron also stabilizes the subgrain structure.

The other difference between modified 9Cr–1Mo and NF616 and E911 involves the replacement of some molybdenum by tungsten in NF616 and the addition of 1% W to the modified 9Cr–1Mo composition in E911. Tungsten was used in these steels because it was assumed to provide more solid-solution strengthening than molybdenum. Although there does not appear to be any direct evidence for this, it does not matter for reduced-activation steels, since they cannot contain molybdenum.

5. Analysis of improved reduced-activation steels

The above discussion demonstrates that impact toughness of ORNL 9Cr–2WVTa steel is superior to that of commercial HT9 and modified 9Cr–1Mo steels it is meant to replace. Likewise, tensile properties of 9Cr–2WVTa were as good as or better than those of HT9 and modified 9Cr–1Mo [14]. Based on neutron irradiation effects on Charpy impact properties, all indications are that properties of irradiated 9Cr–2WVTa are better than those of the conventional steels it replaced. Therefore, the objective in the development of the 9Cr–2WVTa steel was met. A similar conclusion applies to the Japanese (F82H) and European (EUROFER 97) reduced-activation steels.

With respect to creep behavior, however, reduced-activation steels are not as good as modified 9Cr–1Mo (Fig. 3), and this latter steel can, therefore, be used to higher temperatures. Furthermore, the creep-rupture properties of the newer NF616 steel are far superior to those of the reduced-activation steels (Fig. 4). Thus, given the need for higher operating temperatures for fusion systems and the desire to use reduced-activation steels, a second iteration in development of reduced-activation steels is required if properties similar to those of NF616 are to be realized.

As stated above, the main difference between ORNL 9Cr–2WVTa and modified 9Cr–1Mo is that no nitrogen was added to the former. The MX precipitates that strengthen modified 9Cr–1Mo are a combination of vanadium-rich nitrides and niobium-rich carbonitrides [24–26]. By increasing nitrogen from 0.01 to 0.02% in the present reduced-activation steels to the 0.05–0.06% present in modified 9Cr–1Mo and NF616, the number of MX precipitates should be increased, and there should be a commensurate increase in creep strength to approach that of modified 9Cr–1Mo. Based on recent calculations for the limits on nitrogen allowed in a reduced-activation steel, this level of nitrogen would not compromise the reduced-activation character for waste disposal or recycling [2–4].

An addition of boron would also not compromise the reduced-activation character. However, to achieve the desired effect, isotopic ^{11}B should be added instead of natural boron. The ^{10}B constitutes 20% of natural boron, but it transmutes rapidly during neutron irradiation by a (n,α) reaction to produce helium and lithium. Adding ^{11}B should not be a problem, since the isotope is readily available and relatively cheap.

If the disposition of the boron in the microstructure is examined by computational thermodynamics calculations with the JMatPro computer program [30], some interesting observations are made concerning the amount of boron incorporated in $M_{23}C_6$. In particular, for the NF616 composition with 0.06% N and 0.004% B, JMatPro predicts that BN is stable over the range 1378–644 °C (Fig. 5). Between ≈900 and 650 °C, essentially all the boron is in BN.

Experimental evidence exists for the presence of BN at elevated temperatures. Coarse (up to 4 μm) BN-type inclusions were observed in normalized-and-tempered NF616 and in an experimental 9Cr–3W–3Co–0.2V–0.5Mn–0.005N–0.007B–0.004C steel [31]. Heat treatment studies were conducted on the latter low-nitrogen steel to determine if BN precipitates could be eliminated. Rod specimens were austenitized 0.5–16 h in the range 1000–1250 °C and water quenched. The specimens were fractured, and the inclusions were identified by SEM at the bottom of dimples on the ductile fracture surface. The BN inclusions were only eliminated after 0.5 h at 1250 °C, and they were not eliminated at 1000–1200 °C, even for long hold times. The steels are usually austenitized around 1050 °C. These observations do not necessarily contradict the calculations that the precipitate becomes unstable at 1378 °C, given the experimental method for detecting the precipitates and the fact that even in the range where essentially all the boron is in BN, the calculated amount of precipitate formed is less than 0.01 wt%.

The $M_{23}C_6$ precipitates form during tempering at ≈750 °C. Thermodynamics calculations indicate that at 750 °C, 98.9% of the boron in NF616 is in BN and about 1.1% is in the $M_{23}C_6$. Furthermore, for NF616 at 600 °C, the calculations predict that ≈99.8% of boron is incorporated in molybdenum-rich M_3B_2, and only 0.2% is in $M_{23}C_6$ (Fig. 5). As discussed above, boron was found in $M_{23}C_6$ in NF616 [25–29]. No observations of M_3B_2 have been reported in these steels, which is not too surprising, since only 0.05 wt% is calculated at 600 °C for NF616.

Results are somewhat different if the calculations are made for 9Cr–2WVTa steel with 0.005% B and 0.05% N (9Cr–2WVTaNB) (Fig. 6). Calculations again indicate that BN is stable between 1383 and 682 °C, but as the temperature is decreased from 854 °C, the amount of the boron in BN decreases from ≈100% to 0 at 682 °C, below which BN is unstable.

There are other differences between the calculated equilibrium phases of 9Cr–2WVTaNB and NF616. In particular, M_3B_2 does not form in the 9Cr–2WVTaNB. Furthermore, at a 750 °C tempering temperature, ≈29% of the boron is in $M_{23}C_6$, and at $T < 682$ °C, all boron is calculated to be in $M_{23}C_6$. If boron stabilizes $M_{23}C_6$, which in turn stabilizes the subgrain structure, this effect should be much more effective in a steel that does not contain molybdenum. This information could also be used to devise a heat treatment that allowed all of the boron to be incorporated in $M_{23}(CB)_6$.

When reduced-activation steels were developed, tungsten was substituted for molybdenum on an atom-for-atom basis. Since the atomic weight of tungsten is approximately twice that of molybdenum, 2 wt% W was substituted for 1 wt% Mo. Both of these elements promote Laves phase [$Fe_2(Mo,W)$] formation. Thermal aging experiments on F82H to 30000 h at 400–650 °C demonstrated a correlation between an increase in DBTT (Fig. 7) and the amount of Laves phase formed during aging (Fig. 8) [32,33]. A ΔDBTT of 105 °C was observed after 30000 h at 650 °C, even though there was a simultaneous reduction in yield stress of 180 MPa.

Because of the effect of temperature on precipitation kinetics, the amounts of Laves phase formed depended on aging temperature and time. This is demonstrated in Fig. 8, where the calculated amount of Laves is compared to that extracted after aging 30000 h at 400–650 °C. Because of increased kinetics with increasing temperature, equilibrium was reached by 30000 h at 650 °C. Further, there is good agreement between calculated and extracted amounts at 650 °C. It also appears that equilibrium is being approached at 600 °C. The amounts of extracted precipitate decrease with decreasing temperature from 550 to 400 °C, and the extracted amounts at these temperatures are much less than the calculated amounts.

Laves phase formation in F82H would be expected to be affected by neutron irradiation. Increases in Charpy transition

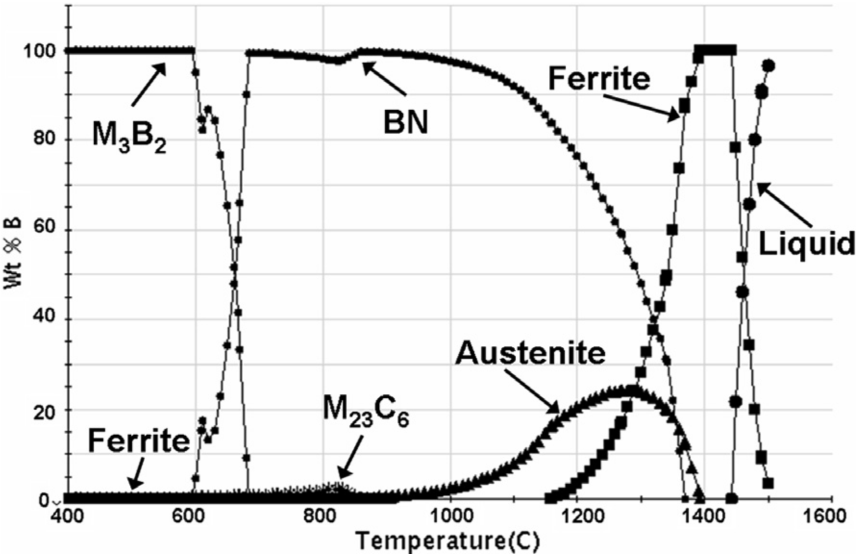

Fig. 5. A plot of the distribution of boron among the phases in NF616 steel as a function of temperature as calculated by JMatPro.

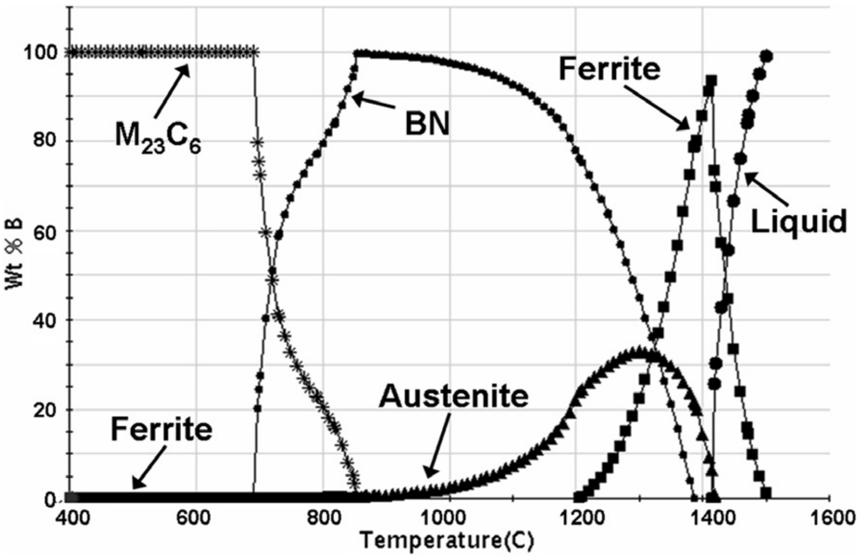

Fig. 6. A plot of the distribution of boron among the phases in ORNL 9Cr–2WVTa steel as a function of temperature as calculated by JMatPro.

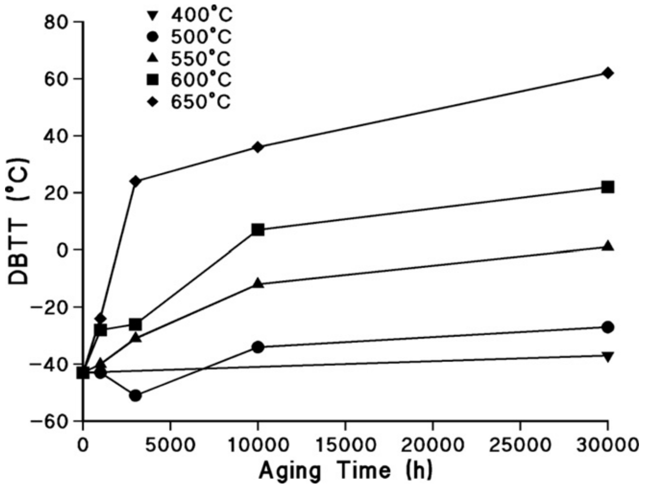

Fig. 7. Ductile–brittle transition temperature of F82H steel thermally aged at 400, 500, 550, 600, and 650 °C for up to 30000 h.

temperature of 33 and 38 °C were observed for F82H irradiated at 500 °C in the High Flux Isotope Reactor (HFIR) to 5 and 20 dpa, respectively, with no change in yield stress (i.e., no irradiation hardening) [34,35]. The effect on DBTT was evidently observed at 500 °C because of irradiation-accelerated Laves-phase precipitation [35].

In addition to promoting embrittlement, Laves phase formation removes tungsten and molybdenum from solution, thus reducing the amount of these elements available for solid-solution strengthening, the primary strengthening mechanism for low-stress (long-time), high-temperature service conditions. The amount of Laves phase formed can be reduced by reducing tungsten concentration. Thermodynamics calculations with JMatPro predict a significant reduction in Laves phase when the tungsten is reduced from 2% to 1% (Fig. 9). At 600 °C, 9Cr–2WVTa is predicted to contain 1.85% Laves phase, whereas if tungsten is reduced by 50% to form 9Cr–1WVTa, only 0.23 wt% Laves is calculated to form.

Intuitively, it appears reducing tungsten content from 2 to 1% would significantly reduce the amount of tungsten in solution. However, with less Laves phase present, a greater percentage of

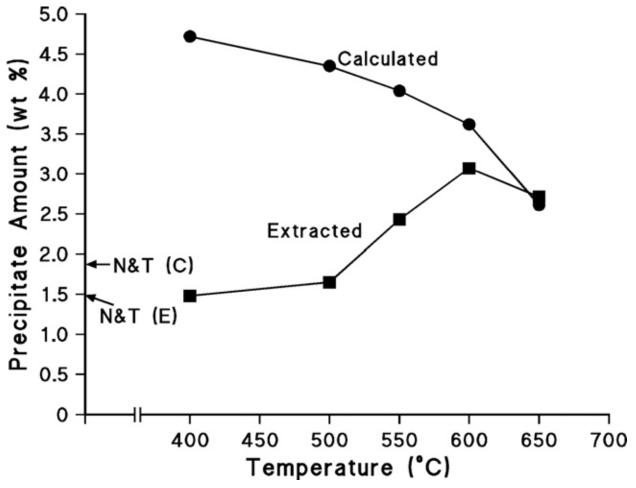

Fig. 8. Amount of precipitate extracted from F82H Steel thermally aged 30000 h at 400, 500, 550, 600, and 650 °C compared to amount calculated using the computational thermodynamics program JMatPro.

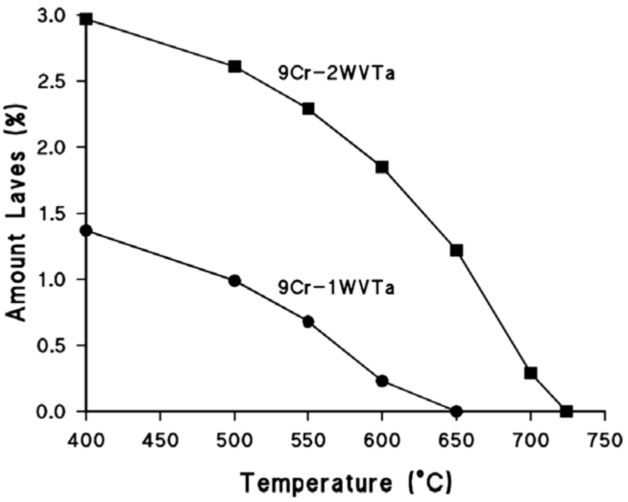

Fig. 9. Amount of Laves phase that forms above 400 °C for a reduced-activation steel containing 1% (9Cr–1WVTa) and 2% (9Cr–2WVTa) tungsten calculated using the computational thermodynamics program JMatPro.

the total tungsten will be in solution – 69% with 1% W versus 34% with 2% W at 600 °C, which means that in both cases ≈0.69 wt% W is available for solid-solution strengthening in 600 °C. Likewise, there is no difference at the lower temperatures (400–600 °C). Thus, reducing tungsten from 2% to 1% reduces the amount of Laves-phase precipitate by over 50%, while leaving the amount of tungsten in solution essentially unchanged.

On the basis of total tungsten in the composition, EUROFER 97 with 1% W has an advantage relative to the amount of Laves phase that will form and, presumably, the amount of embrittlement. [32–34]. However, it will not have an advantage for long-time, high-temperature creep strength, since both contain the same amount of tungsten in solution, and this was demonstrated in the similarity in the extrapolated long-time creep-rupture strength observed for EUROFER 97 with 1% W and F82H with 2% W (Fig. 3).

6. Proposed compositions for improved reduced-activation steel

The results of this analysis provide a starting point for development of an improved reduced-activation steel. From the comparison of the present reduced-activation steels with the commercial steels with better properties developed during the last 20 years for conventional power plants and using the knowledge available on the origin of the improved properties, it is concluded that an improved reduced-activation steel would contain 1% W, 0.005–0.01% B, and 0.05–0.07% N. Accordingly, the new 9Cr–1WVTaNB reduced-activation steel would have a nominal composition of Fe–9.0Cr–1.0W–0.25V–0.10Ta–(0.005–0.010)B–(0.05–0.07)N–0.10C.

In Table 3, a comparison is shown for the expected (calculated) amount of Laves phase and the amount of tungsten in solution at 400, 500, and 600 °C in going from 9Cr–2WVTa to 9Cr–1WVTaNB steel. Calculations were made for going from 9Cr–2WVTa with 0.02% N to that composition with 0.05% N (9Cr–2WVTaN) to the latter steel containing the higher nitrogen but with tungsten decreased to 1% (9Cr–1WVTaN), and finally to the latter composition with 0.005% B (9Cr–1WVTaNB). Also shown is the calculated amount of MX, for which only one value is shown, since it does not vary significantly between 400 and 600 °C.

The increase in nitrogen from 0.02 to 0.05% will increase the amount of MX from 0.18 to 0.32% with the change independent of W and B concentrations. At 600 °C, Laves phase will be reduced from 1.85% to 0.23% in going from 9Cr–2WVTa to 9Cr–1WVTaN. When boron is added, the calculated amount of Laves increases to 0.36%. This latter change occurs because boron replaces tungsten in $M_{23}C_6$, and the tungsten eliminated from $M_{23}C_6$ is incorporated in additional Laves phase. Similar relative changes are observed at 400 and 500 °C. Tungsten in solution is similar for all compositions, with the amount increasing with temperature (Table 3).

Obviously, a program for a second iteration of reduced-activation steel development would need to determine optimum ranges of B, N, and W. Appropriate additions of Mn and Si would also be made. Several compositions with varying W, B, and N should be explored to determine the effect of these elements on elevated-temperature strength and toughness. Once a satisfactory composition is established, creep-rupture tests on selected compositions will be required to adequately assess elevated-temperature properties. Finally, long-time, low-stress creep-rupture tests will be required to adequately assess the suitability of the chosen composition.

Table 3
Calculated amounts of precipitates and tungsten in solution

Steel	MX (%)	Laves (%)			W in Ferrite (%)		
		400 °C	500 °C	600 °C	400 °C	500 °C	600 °C
9Cr–2WVTa	0.18	2.97	2.61	1.85	0.096	0.28	0.68
9Cr–2WVTaN	0.32	2.97	2.61	1.85	0.096	0.28	0.68
9Cr–1WVTaN	0.32	0.22	0.99	0.23	0.094	0.28	0.69
9Cr–1WVTaNB	0.32	1.44	1.10	0.36	0.094	0.28	0.69

The most difficult aspect of such a development program is the question of irradiation resistance. In the irradiation-hardening temperature regime, little difference between 9Cr–1WVTaNB and 9Cr–2WVTa is expected due to reduction of tungsten and addition of ^{11}B. There is considerable similarity between the proposed 9Cr–1WVTaNB steel and modified 9Cr–1Mo (9Cr–1MoVNb). Therefore, the effect of 0.05–0.07% N addition should not cause a large difference in irradiation-hardening resistance, based on the excellent irradiation resistance of modified 9Cr–1Mo, for which considerable radiation-effects data have been obtained [5]. Likewise, there should be little difference at temperatures above the hardening range, since the calculations indicate that 9Cr–1WVTaNB should have similar amounts of Laves phase to that observed in 9Cr–1MoVNb. No indication of embrittlement was observed when 9Cr–1MoVNb was irradiated in EBR-II to 26 dpa at 450, 500, and 550 °C [21], which contrasts with the embrittlement observed on F82H after irradiation in HFIR to 5 and 20 dpa at 500 °C [34,35].

7. Summary and conclusions

When reduced-activation steels were developed for fusion, the objective was to obtain steels with room-temperature strength and toughness similar to the conventional steels they replaced. Steels developed in Japan, Europe, and the United States met those objectives. However, for increased efficiency, reactor designers have sought to increase fusion reactor operating temperatures beyond the upper-use temperature of reduced-activation steels, which is ≈550 °C. This upper-temperature limit is not as high as for some commercial steels the reduced-activation steels replaced. Properties are even more inferior to commercial steels developed in recent years. Therefore, for increased fusion-reactor operating temperatures, a second iteration of reduced-activation steel development is required.

Compositions of present reduced-activation steels were compared with those of the commercial steels that have better elevated-temperature properties. Differences were attributed to the presence of boron and nitrogen in the strongest commercial steels. Based on the comparison of properties and compositions, it was concluded that a second iteration development program for reduced-activation steels should be focused on a nominal composition of Fe–9.0Cr–1.0W–0.25V–0.10Ta–(0.005–0.010)B–(0.05–0.07)N–0.10C. Several variations around this composition should be investigated. Based on tests to determine elevated-temperature strength and toughness, appropriate compositions can be selected for long-time, low-stress creep tests, which are required to establish the upper-temperature limits for the new steels.

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