DEVELOPMENT OF FERRITIC STEELS FOR REDUCED ACTIVATION:

THE U.S. PROGRAM[*]

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Abstract

The Cr-Mo ferritic (martensitic) steels are candidates for the structural components of fusion reactors. Irradiation of such steels in a fusion environment produces long-lived radioactive isotopes, which lead to difficult radioactive waste disposal problems once the structure is removed from service. Such problems could be reduced by using steels that contain only elements that produce radioactive isotopes that decay to low levels in a reasonable time (tens of years instead of hundreds or thousands of years). The U.S. Department of Energy has a program to develop steels to meet the criteria for shallow land burial as opposed to deep geologic storage. A review of the alloy development programs indicates that ferritic steels that meet these criteria can be developed.

DISCLAIMER

1. Introduction

The most serious safety and environmental concerns for fusion reactors involve the induced radioactivity in the first-wall and blanket structures [1]. If a "low-activation" material were available for the structural components, such safety concerns would vanish. In addition, such a material would allow for contact maintenance and repair; it would also alleviate the problem of special waste storage of the highly radioactive structures after service (they could be discarded as nonradioactive materials).

As pointed out in a report by a U.S. Department of Energy (DOE) panel set up to study this subject, the technology for commercially producing and fabricating material that would meet the low-activation criteria is not available and is unlikely to be available soon (high-purity silicon carbide was the only material suggested) [1]. This means that other solutions must be sought for the safety and maintenance problems. However, it does appear possible to simplify the disposal of radioactive reactor components by developing materials in which the induced radioactivity decays in a reasonable time to such a level that the waste no longer requires maintenance. Such materials could be disposed of by shallow land burial instead of the much more expensive deep geologic disposal.

The DOE panel called attention to the difficulty of defining low-activation materials by defining four types of fusion reactors and materials. It defined standard-activation, very-low-activation, low-activation, and reference-activation reactors and materials. These were defined according to the guidelines for the classification and disposal of low-level nuclear wastes that have been issued by the U.S. Nuclear Regulatory Commission (10 CFR Part 61) [2].

A standard-activation reactor would be constructed of a standard-activation material such as type 316 stainless steel or 12Cr-1MoVW steel. For such reactors, all maintenance would be performed remotely. These materials are not classified in 10 CFR Part 61, nd deep geologic burial would be required for waste disposal. The very-low-activation and low-activation materials would be classified as Class A and Class B, respectively, in 10 CFR Part 61 and could be disposed of easily. They would also allow certain hands-on maint nance.

Finally, the reference-activation material (Class C of 10 CFR Part 61) could be disposed of by land burial; limited hands-on maintenance would be possible only outside the shield after two days of shutdown. Steels, depending on the alloying elements, fall into the class of reference-activation materials. The Alloy Development for Irradiation Performance task of the U.S. Department of Energy Office of Fusion Energy is funding work to develop such steels. It is these materials that will be discussed in this paper.

The term **low-activation** material is often used to describe alloys that allow simplified waste disposal as well as those that would allow hands-on maintenance. Materials that would allow hands-on maintenance have to have low activation (e.g., pure SiC) or would have to decay to extremely low levels very rapidly. Thus the term **reduced activation** will be used in this paper to describe the materials under discussion. The steels that meet the reference-activation criteria have also been called **fast induced-radioactive decay** (FIRD) steels [3], where **fast** is a relative term that was taken to mean steels with radioactive decay rapid enough to qualify at least for Class C waste disposal criteria.

The ferritic steels now being considered for fusion-reactor applications are the Cr-Mo steels: 2 1/4Cr-1Mo, 9Cr-1MoVNb, and 12Cr-1MoVW steels. Molybdenum and niobium are the primary alloying elements that keep these steels from meeting the criteria for near-surface burial. These elements would need to be eliminated, and in particular, niobium would have to be reduced to extremely low levels [1]. Other alloying elements that must be minimized include nickel, copper, and nitrogen, none of which play a dominant role in determining the properties of most of the Cr-Mo steels. About 0.5% Ni is added to 12Cr-1MoVW steel to avoid delta-ferrite [3] and it would have to be replaced.

Because tungsten is found to behave like molybdenum in simple steels, it is being proposed to replace molybdenum [3]. Vanadium is also being proposed for this purpose [4]. It has been suggested that the strengthening function of niobium could be replaced by a greater use of vanadium and by titanium [3,4]; the use of tantalum is also being investigated. However, one problem with tantalum is that its presence in alloys irradiated in fission reactors results in rapidly decaying radionuclides that make the study of such alloys after irradiation difficult.

In the United States, reduced-activation ferritic steels are being developed at the Oak Ridge National Laboratory (ORNL), Hanford Engineering Development Laboratory (HEDL), and GA Technologies, Inc. (GA). The alloys being investigated will be described, and the initial results will be reviewed.

2. Alloys

The basis for replacing molybdenum and niobium in the Cr-Mo steels of interest in the fusion reactor program has been discussed [3-5]. Tungsten,

vanadium, and tantalum have been chosen to replace these elements in the Cr-Mo steels of interest to provide the strengthening required. The objective of the alloy development program is the development of reduced-activation alloys with unirradiated and irradiated properties as good as or better than those of the Cr-Mo steels.

The 9Cr-1MoVNb and 12Cr-1MoVW steels were chosen for fusion-reactor applications because of studies in the breeder reactor program that showed the steels had excellent swelling resistance and high-temperature creep resistance. However, the proposed breeder reactor applications involve service conditions considerably different from those envisioned for fusion reactors. Because of those differences, it was recommended that 21/4Cr-1Mo steel should be strongly considered for fusion-reactor applications [6]. This follows because of the lower operating temperatures of some proposed fusion reactor designs, possible advantages in welding the low-chromium steel, and conservation of a strategic material. In ion-irradiation studies, the swelling resistance of 21/4Cr-1Mo steel was equivalent to that of high-chromium steels [6]. The reduced-activation development programs consider chromium concentrations over the range 21/4 to 12%.

The three alloy-development programs have taken slightly different approaches to replace the molybdenum and niobium. The ORNL program is considering a range of alloys based on the analogous Cr-Mo steels (Table 1) [3]. The HEDL program has concentrated on replacing molybdenum with vanadium, although several alloys with tungsten are being investigated (Table 2) [4]. The GA program is concentrated on a 9 Cr and a 12 Cr steel with vanadium and tungsten (Table 3) [5]. The three programs will be discussed separately.

3. ORNL Program

The initial consideration of possible replacements for molybdenum and niobium indicated that the Cr-W steels may offer the best possibility as replacements for Cr-Mo steels [3]. Tungsten is in the same group of the periodic table and displays similarities to molybdenum when it is used as an alloying element in steels. This similarity suggested an initial steel composition of 2% W.* This amount of tungsten is required to obtain an atomic concentration similar to that for 1% Mo. It was decided to add vanadium in an amount similar to that present in the Cr-Mo steels of interest [3].

The alloys in Table 1 were selected as a first cut in developing what were termed fast induced-radioactivity decay (FIRD) steels [3]. The steels include Cr-V, Cr-W, Cr-W-V, and Cr-W-V-Ta steels. The chromium compositions were chosen to allow a comparison with the 2¼Cr-1Mo, 9Cr-1MoVNb, and 12Cr-1MoVW steels presently of interest. The tantalum was substituted for the niobium. Tantalum is in the same group of the periodic table as niobium and often displays properties analogous to those of niobium.

The compositional variation was chosen to develop information on the effect of Cr, W, and V on the properties of this class of steel. From such a study, an optimized FIRD steel is to be sought. Such a systematic variation of these elements does not appear to have been conducted for the Cr-Mo steels, and it is hoped that information obtained from these studies will help not only in the development of an optimized FIRD steel, but will also aid in understanding the behavior of the Cr-Mo steels [3]. The existing irradiation studies on the Cr-Mo steels should also accelerate the development of the FIRD steels.

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*All compositions are in weight percent (wt %).

A 15.9-mm-thick plate of each of the eight heats was normalized and then pieces of the plate were tempered 2 h at 600, 650, 700, 750, and 780°C [7]. Microstructures varied according to the chemical composition. The 2¼ Cr alloys demonstrated the effect of the tungsten on hardenability (as determined by amount of bainite in the microstructure) [Fig. 1(a)-(d)]. The 2¼ Cr alloy with 0.25% V (2¼ CrV) and that with 1% W and 0.25% V (2¼Cr-1WV) had 30 to 35% bainite, with the remainder polygonal ferrite. The 2¼Cr-2W steel was almost entirely bainite (less than 1% ferrite) and the 2¼Cr-2WV steel contained 8 to 10% polygonal ferrite. When these steels were heat treated as 0.76-mm-thick sheet, the size of material from which irradiation tensile specimens are generally machined, all of the steels were entirely bainite.

The 5Cr-2WV, 9Cr-2WV, and 9Cr-2WVTa steels [Fig. 1(e)-(g)] were 100% martensite. However, the 12Cr-2WV steel [Fig. 1(h)] contained considerable amounts of delta-ferrite (approximately 26%). This alloy was patterned on the 12Cr-1MoVW steel with tungsten substituted for the molybdenum. The appearance of the delta-ferrite indicates that insufficient amounts of austenite-stabilizing elements were present to form 100% austenite during the austenitization treatment. The 12Cr-1MoVW steel contains approximately 0.5% Ni and 0.2% C, both austenite-stabilizing elements, whereas the 12Cr-2WV steel contains approximately 0.1% C and no nickel. Significant amounts of nickel are not allowed in a FIRD steel, and the carbon content was restricted to 0.1% to promote good weldability.

Figure 2 shows the hardness as a function of tempering temperature after the 2-h tempering treatments. Although it is expected that vanadium carbide will play an important role in the strength of these steels, there

is also apparently a synergistic effect of tungsten and vanadium. The steels with 0, 1, and 2% W and 0.25% V display increasing hardness with increasing tungsten content. The $2\frac{1}{4}$Cr-2W steel is harder than the $2\frac{1}{4}$Cr-2WV steel as normalized; after tempering at higher temperatures, the steel with vanadium is harder. The as-normalized hardness of the $2\frac{1}{4}$Cr-2WV steel is similar to the hardness of the $2\frac{1}{4}$Cr-2W steel despite the fact that it is not entirely bainite. The $2\frac{1}{4}$Cr-2WV steel has the highest hardness of the $2\frac{1}{4}$ Cr steels, regardless of the tempering conditions. Tempering for 2 h at 600°C gives rise to a hardness peak for this steel, and for the highest tempering temperature, the hardness of this steel is similar to that of the high-chromium steels.

The hardness of the 5Cr-2WV steel approached the highest hardness achieved for these steels — similar to that for the as-normalized 9Cr-2WVTa steel. The two 9 Cr steels showed little difference, both having normalized hardnesses near R_c 40, after which the hardnesses decreased continuously with tempering temperature.

The 12Cr-2WV steel contained 26% delta-ferrite (balance martensite), and for that reason it would be expected to have a lower hardness than the steels made up entirely of martensite. This is the case in the normalized condition and after tempering at 650 and 700°C. However, the tempering resistance of the steel is such that, when tempered at the highest temperatures, the hardness approaches that of the 5 Cr and 9 Cr steels that contained no ferrite.

The tempering studies indicated that for long tempering times or high tempering temperatures, the $2\frac{1}{4}$Cr-2WV, 5Cr-2WV, 9Cr-2WV, 9Cr-2WVTa, and 12Cr-2WV steels behaved similarly. This occurred despite the fact that

the 2¼ Cr steel was bainitic and the high-chromium steels were martensitic [7]. Furthermore, the 12 Cr steel also developed hardnesses similar to the other martensitic steels after longer tempering times or higher tempering temperatures. This occurred despite the fact that the steel contained 26% delta-ferrite and was much softer than the other martensitic steels after normalizing and after the low-temperature tempers.

The tempering behavior of the FIRD steels was compared to the 2¼Cr-1Mo, 9Cr-1MoVNb, and 12Cr-1MoVW steels. The 2¼Cr-2W steel behaved like the 2¼Cr-1Mo steel [7]. The 2¼Cr-2WV steel had a greater tempering resistance than 2¼Cr-1Mo steel, probably because of the vanadium. The 9Cr-2WV and 9Cr-2WVTa steels tempered quite similarly to 9Cr-1MoVNb and 12Cr-1MoVW steels. The 12Cr-2WV steel also behaved like these high-chromium steels at long tempering times or higher tempering temperatures (Fig. 3) [7].

An attempt was made to eliminate the 26% delta-ferrite in the 12Cr-2WV steel by adding more manganese and carbon. With the 12Cr-2WV (0.1% C, 0.5% Mn) steel as the base composition, alloys with 2.8% Mn, 5.6% Mn, and 0.2% C were produced [8]. Manganese is used in place of the nickel (not allowed in a FIRD steel) used for this purpose in the 12Cr-1MoVW steel; higher manganese concentrations are required because manganese is not as strong an austenite former as nickel. After a normalizing heat treatment, the 2.8% Mn and 5.6% Mn steels contained 20% and <1% delta-ferrite, respectively; the steel with 0.2% C contained about 5% delta-ferrite.

When tempered, the steel with 0.2% C had the highest hardness at the low tempering temperatures [8]. Although the steels with 0.2% C and

5.6% Mn had higher hardnesses than the 2.8% Mn and the base composition for the highest tempering temperature, the differences were small [8]. Therefore, all of the compositions might be expected to develop similar strengths after long service times.

As pointed out above, the steel with 26% delta-ferrite had hardness values after long tempering times or high tempering temperatures similar to the values for the all-martensitic steels with 5 and 9% Cr. Thus, it appears that it might be possible to consider a duplex steel for a fusion reactor first wall. Because of their better welding characteristics, the 9Cr-2Mo duplex steels are being considered by the Japanese fusion program as first-wall candidates. One problem with such steels is that the presence of delta-ferrite increases the ductile-brittle transition temperature (DBTT).

The tensile behavior was determined for the eight heats of steel in the normalized-and-tempered condition. The preliminary data verified the strengthening effect of vanadium and tungsten that was observed in the tempering studies [8]. They also revealed that the 2¼Cr-2WV steel had tensile properties similar to or slightly better than those of the 5 Cr, 9 Cr, and 12 Cr steels when the steels were given similar heat treatments.

In addition to further tensile property determinations, impact properties of the unirradiated steels will also be determined. Tensile, impact, and TEM specimens are being irradiated in FFTF to obtain the first information on the effect of irradiation on these steels.

4. HEDL Program

The alloys being investigated by HEDL (Table 2) are based on chromium compositions used in the Cr-Mo steels [4,9]. The choice of alloying elements

was based on a pseudo-carbide phase diagram (Fig. 4) derived from the work in the literature on Fe-Cr-V-C alloys [10,17]. The emphasis in this program was on replacing the molybdenum with vanadium rather than with tungsten, although some tungsten is also being used (1 wt % W was used). There has been some discussion about whether 1 or 2% W should be used [4]. Less tungsten would minimize the possibility of forming intermetallic phases during irradiation, which could lead to embrittlement. Manganese was used in the high-chromium alloys to avoid delta-ferrite formation.

The 2¼Cr-0.5 V, 2¼Cr-1V, and 2¼Cr-1.5V-0.3Mn were used to determine an optimum level of vanadium for such an alloy. The fact that the 9Cr-0.5V steel includes virtually no manganese means that the addition of vanadium must be kept below 0.5% (based on Fig. 4) in order to eliminate delta-ferrite. The 9Cr-0.5V-2.5Mn and 1.3V-1Mn steels examine the effect of manganese and vanadium. In the latter steel, the vanadium level was increased with a compensating increase in the carbon level. The 9Cr-0.3V-1W-2.5Mn steel was intended to investigate the effect of tungsten. The 12Cr-1V-6.5Mn and 12Cr-0.3V-1W-6.5Mn steels were similar to 12Cr-1MoVW steel, except that molybdenum was excluded and carbon levels reduced. In order to compensate for the lower carbon levels, the manganese content was increased significantly. These steels also were used to examine the effects of tungsten and vanadium.

A series of tensile specimens and transmission electron microscopy (TEM) disks were either quenched and tempered or normalized and tempered. The tensile specimens were taken from 0.76-mm-thick sheet and the TEM disks from 0.02-mm-thick sheet. The 2¼ Cr steels were water quenched and

then tempered 2 h at 650°C. The high-chromium steels were normalized and then tempered 2 h at 700°C.

The microstructures of the as-heat-treated steels were found to be typical of bainitic (low-chromium steels) and martensitic (high-chro.ium steels) steels (Fig. 5). However, differences between the alloys could be found, and in the 12Cr-1V-6.5Mn steel, the amount of delta-ferrite present was excessive.

Alloys in the 2¼ Cr range showed structures typical of tempered bainite, whereas structures typical of tempered martensite were found in 9 and 12 Cr alloys. Careful examination of the 12Cr-1V-6.5Mn steel across the thickness of the billet revealed that there was more delta-fe⁻ite at the edge of the billet than at the center, indicating a composition gradient; the exact nature of the gradient was not determined. From Fig. 5, the delta-ferrite content in the 12Cr-1V-6.5Mn steel was estimated to be 5%. Therefore, it is apparent that the compositional specification for this alloy requires more austenite stabilizer if a martensitic steel is to be produced. To achieve this, it was proposed that the carbon content could be increased by 0.025%, the manganese content increased by 0.75%, or the chromium content decreased by 0.35% [9].

The tensile properties of the as-heat-treated reduced-activation steels were in most cases comparable to their commercial-alloy counterparts. Exceptions were found for the bainitic alloys and the 12 Cr steel. The low-vanadium bainitic alloy (2¼Cr-0.5V) was 65% stronger than 2¼Cr-1Mo, but the other bainitic alloys were weaker, having much lower yield strengths, slightly lower ultimate tensile strengths, and much higher elongations. The 12 Cr martensitic alloys had significantly higher

strength properties with comparable elongation response in comparison with 12Cr-1MoVW steel. This may be due in part to differences in heat treatment. The properties for the commercial alloys were taken from the literature [12].

The alloy response following irradiation at 420°C to between 6 and 10 dpa could be divided into two types. The bainitic alloys all showed increases in strength and decreases in elongation, but the 2¼Cr-0.5V and 2¼Cr-1V steels were considerably stronger than 2¼Cr-1Mo after irradiation at 420°C. In comparison, the 9 Cr and 12 Cr martensitic steels showed only a moderate increase in strength. This is quite different from the behavior of 9Cr-1MoVNb and 12Cr-1MoVW steels, where significant hardening occurred following irradiation at 400°C [13]. However, in all cases, elongation values for the reduced-activation alloys do not appear to be as good as those of their commercial counterparts. It should be pointed out that differences in heat treatment may account for some of the differences observed.

Figure 6(a) and (b) provides a comparison of yield strength response as a function of irradiation temperature. For the bainitic alloys [Fig. 6(a)], the yield strength increased 200 MPa when irradiated at 420°C, but following irradiation at 585°C, the strength decreased. Hardening at 420°C was much less in the martensitic alloys [Fig. 6(b)] and reductions in strength for the 9 Cr alloys after irradiation at 585°C were not as large as those for the bainitic alloys. However, the strength remained high for the 12 Cr alloys after irradiation at 585°C. Figure 6(c) demonstrates that the elongation was reduced from the unirradiated values

more in the bainitic alloys than the martensitic alloys following irra-
diation at 420°C, but the lower elongation levels were still satisfactory.

Electron microscopy studies of the as-heat-treated steels showed that
the microstructures of the reduced-activation steels were characteristic
of bainitic and martensitic steels. The bainitic alloys were found to be
much more heterogeneous than the martensitic alloys with regard to carbide
distributions. Also, recovery was more complete in the 9 Cr martensitic
alloys than the 12 Cr alloys. Coarse vanadium carbides were observed in
the bainitic steels. Dislocation structures in the 9 Cr steels were
relaxed and formed well-defined cell walls; carbide particles had diameters
on the order of 130 nm. The 12 Cr steels had similar structures, except
the dislocation structures that formed during the martensitic transfor-
mation recovered less during tempering.

After irradiation, TEM observations of specimens irradiated at 420
and 520°C indicated that the effect of irradiation was similar to that
observed on the Cr-Mo steels irradiated to these levels. At 420°C, most
of the alloys developed perfect dislocation loops. A few scattered voids
were found in several of the steels, but swelling was negligible. All of
the alloys developed considerable precipitation. Although no detailed
precipitation identification was carried out, evidence for α' precipita-
tion was detected in the 12 Cr steels.

The alloys irradiated at 520°C to 14 dpa were found to be relatively
unaffected by irradiation. No void swelling was observed, and only a few
examples of irradiation-induced dislocation loops were found. It was
found that the dislocation and subgrain structures had relaxed and that
precipitate coarsening had occurred. Similar changes were expected to be

responsible for the reductions in strength found after irradiation at 585°C. The better strength properties found in the 12Cr-0.3V-1W-6.5Mn steel can be traced to more limited subgrain growth and the formation of new small precipitate particles. In general, the precipitate structure after irradiation at 520°C was similar to the preirradiation structure. However, compositional analysis of extracted particles revealed that phases other than those expected (based on the carbide diagram) were observed. A precipitate tentatively identified as M_{23} was found, which does not agree with the carbide phase diagrams. An intermetallic phase rich in iron was also observed in the 12Cr-0.3V-1W-6.5Mn steel.

5. GA Program

The GA Technologies program has focused on a 9Cr-2.5W-0.3V-0.15C (GA3X) and a 11Cr-2.5W-0.3V-0.15C (GA4X) steel with basically similar nominal concentrations of carbon (0.15%), tungsten (2.5%), and vanadium (0.3%) [4,14,15]. The first attempt to produce the alloys resulted in compositions that did not reach the proposed nominal compositions (Table 3). Because of the low carbon content of GA4X (0.018% vs. nominal 0.15%), the alloy contained large amounts of delta-ferrite; no further work was carried out on that heat. Heat-treatment studies were carried out on heat GA3X, and tensile and impact properties were determined.

Heat-treatment studies on the steel indicated that an optimum heat-treatment condition was a 1000°C austenitization temperature and a 700°C tempering temperature. After this heat treatment, the microstructure was essentially 100% tempered martensite. The mechanical properties tests

were carried out on steel that was austenitized 1 h at 1000°C and then water quenched. It was tempered 1 h at 700°C.

The precipitate phases were extracted from the quenched-and-tempered steel and found to be exclusively $M_{23}C_6$. Similarly, after aging for 100 and 1000 h at 600°C, the precipitates were essentially all $M_{23}C_6$, although the weight percent of the precipitates increased from 2.0 to approximately 3% after the 100 h aging treatment. Little further change occurred between 100 and 1000 h [14].

Replicas and foils were examined by TEM. Analysis of the carbides on the replicas indicated that the precipitates were predominantly $M_{23}C_6$; as opposed to the x-ray analysis of extracted precipitates, a few MC particles were tentatively identified. The scarcity of the latter carbides was attributed to the small amount of vanadium present (0.017% vs. the nominal 0.3% aimed for during melting). The TEM studies indicated that the structure was a lath martensite, similar to that obtained for a 9Cr-1MoVNb steel. When aged, large precipitates formed, primarily along the former austenite boundaries and along lath boundaries.

Tensile tests on the steel over the range room temperature to 650°C showed that the steel had strengths slightly above the average properties for modified 9Cr-1Mo and similar to those for one heat of 9Cr-1MoVNb steel that has been tested in the fusion program (Fig. 7). However, it should be noted that GA3X was quenched and tempered at 700°C, whereas the 9Cr-1Mo steels were normalized and then tempered at 760°C. Nevertheless, the fact that the steel achieved these strengths with the lower vanadium concentrations indicated properties similar to 9Cr-1MoVNb should be achieved with an alloy with a higher vanadium concentration after similar heat

treatments. The total elongation of the reduced-activation steel was about 50% less than the 9Cr-1MoVNb steels (Fig. 8). The uniform elongation was also less at temperatures up to 450°C, after which similar values were observed.

Charpy impact tests were conducted on the steel. A DBTT value of −25°C was observed, which compares favorably with 9Cr-1MoVNb steel.

The results were taken to provide some clues concerning an optimum alloy composition [15]. It was felt that significant strengthening in GA3X was provided by the 0.17% C. A lower carbon level may be desirable to ensure weldability and ductility [15]. A carbon content of 0.15% or less was proposed. Higher chromium, vanadium, and titanium was also suggested. The possibility of degradation of the aged-and-irradiated steel by the formation of Fe_2W was suggested [15]. However, no such precipitate was observed, and thus the advisability of using less than 2.5% W is yet to be determined.

6. Summary and Conclusions

Alloy development programs are in progress at three U.S. DOE installations to develop reduced-activation ferritic steels to minimize the impact of the radioactive waste generated when the first wall and blanket structures of fusion reactors are discarded after service. The programs seek to develop the alloys by modifying the Cr-Mo steels that are presently of interest in the fusion program. The modifications require the removal of molybdenum and niobium from the Cr-Mo compositions of interest.

The replacement of molybdenum is most crucial. Tungsten and vanadium are being used as replacements. Alloys with chromium concentrations between 2¼ and 12% are being evaluated. Tungsten concentrations of 1 to 2% are being used, and vanadium is being varied between 0.25 to 1.5%. The selected use of tantalum as a replacement for niobium is also being investigated.

The preliminary test results on unirradiated and irradiated experimental alloys indicate that a reduced-activation ferritic steel can be developed that will have unirradiated properties similar to 2¼Cr-1Mo, 9Cr-1MoVNb, and 12Cr-1MoVW steels. Tempering studies identified a 2¼Cr-2WV steel with long-term tempering properties similar to those of a 9Cr-2WV steel. The tempering characteristics of both were similar to those of 9Cr-1MoVNb and 12Cr-1MoVW steels. The 2¼Cr-2W steel behaved similarly to 2¼Cr-1Mo steel. The results indicated the effectiveness of tungsten as a replacement for molybdenum and the effect of vanadium in steels that also contained tungsten.

The development of a Cr-W martensitic steel with 12% Cr as a replacement for 12Cr-1MoVW steel will require the use of manganese or carbon concentrations above the values generally used in these martensitic steels. Without the addition of carbon or manganese, the microstructure will contain considerable amounts of delta-ferrite. A 12Cr-1W-0.3V-6.5Mn steel had excellent unirradiated and irradiated tensile properties. However, the use of this much manganese may degrade the corrosion properties of such a steel in a lithium coolant.

The preliminary studies on the experimental alloys that have been produced to date indicate than an optimized reduced-activation ferritic

steel is possible. It should be possible to develop steels with properties at least as good as those of the Cr-Mo steels. A systematic study of these steels in conjunction with the continued study of the Cr-Mo steels should lead to such an optimized steel; an improved understanding of the Cr-Mo steels may also be a product of this work.

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[15] C. Y. Hsu and T. A. Lechtenberg, in: ADIP Semiannu. Prog. Rep. Sept. 30, 1985, DOE/ER-0045/15 (U.S. DOE, Office of Fusion Energy, Washington, DC) p. 137.

Table 1. Chemical composition of experimental reduced-
activation steels in ORNL program

Alloy	Chemical composition[a],[b] (wt %)						
	Cr	W	V	Ta	C	Mn	Si
2 1/4 CrV	2.33		0.25		0.11	0.40	0.17
2 1/4 Cr-1 WV	2.30	0.93	0.25		0.10	0.34	0.13
2 1/4 Cr-2 W	2.48	1.99	0.009		0.11	0.39	0.15
2 1/4 Cr-2 WV	2.42	1.98	0.24		0.11	0.42	0.20
5 Cr-2 WV	5.00	2.07	0.25		0.13	0.47	0.25
9 Cr-2 WV	8.73	2.09	0.24		0.12	0.51	0.25
9Cr-2 WVTa	8.72	2.09	0.23	0.075	0.10	0.43	0.23
12 Cr-2 WV	11.49	2.12	0.23		0.10	0.46	0.24

[a]P = 0.014 to 0.016, S = 0.005 to 0.006, Ni < 0.01,
Mo < 0.01, Nb < 0.01, Ti < 0.01, Co = 0.005 to 0.008,
Cu = 0.02 to 0.03, Al = 0.02 to 0.03, B < 0.001.

[b]Balance iron.

Table 2. Chemical composition of experimental reduced-activation steels in HEDL program

Heat Number	Composition (wt %)								
	Cr	C	V	W	Mn	Si	N	P	S
2 1/4 Cr-0.5 V	2.32	0.086	0.50	<0.01	<0.01	0.08	0.004	<0.005	0.003
2 1/4 Cr-1 V	2.38	0.093	1.01	<0.01	<0.01	0.08	0.003	<0.005	0.002
2 1/4 Cr-1.5 V-0.3 Mn	2.46	0.11	1.50	---	0.30	0.30	0.015	0.007	0.015
9 Cr-0.5 V	9.13	0.096	0.52	0.01	0.02	0.09	0.003	<0.005	0.003
9 Cr-1.3 V-1 Mn	9.02	0.097	0.51	0.01	2.68	0.09	0.003	<0.005	0.003
9 Cr-0.5 V-2Mn	9.14	0.197	1.23	0.02	1.08	0.09	0.003	<0.005	0.003
9 Cr-0.5 V-1 W-2.5 Mn	9.14	0.197	1.23	0.02	1.08	0.09	0.003	<0.005	0.003
12 Cr-1 V-6.5 Mn	8.82	0.101	0.27	0.89	2.44	0.10	0.002	<0.005	0.004
12 Cr-0.3 V-1W-6.5 Mn	12.19	0.089	1.05	<0.01	6.47	0.10	0.003	<0.005	0.005

Table 3. The nominal and actual chemical compositions of
reduced-activation steels in GA program

Heat Number		C	Cr	W	V	Ti	Mn	S
GA3X	Melt composition	0.165	7.5	1.95	0.015	--	0.002	0.005
	Nominal[a] composition	0.15	9	2.5	0.3	--	--	--
GA4X	Melt composition	0.01	10.2	1.1	0.2	--	0.011	<0.003
	Nominal composition[a]	0.15	11	2.5	0.3	--	--	--

[a]The nominal composition is the value specified; the melt compositions did not meet these specifications for some elements.
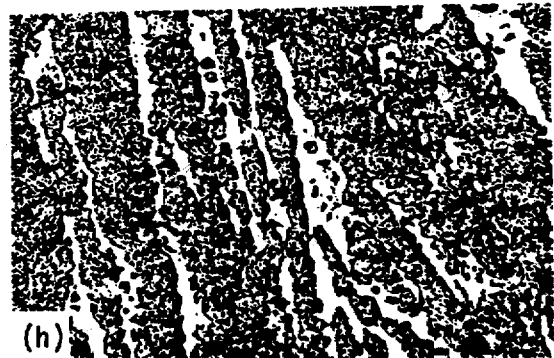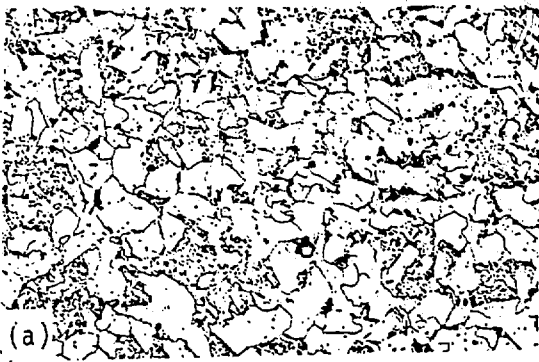
Fig. 1. Microstructures of normalized-and-tempered ORNL reduced-activation steels: (a) 2¼ CrV, (b) 2¼Cr-2WV, (c) 2¼Cr-2W, (d) 2¼Cr-2WV, (e) 5Cr-2WV, (f) 9Cr-2WV, (g) 9Cr-2WVTa, and (h) 12Cr-2WV steels.
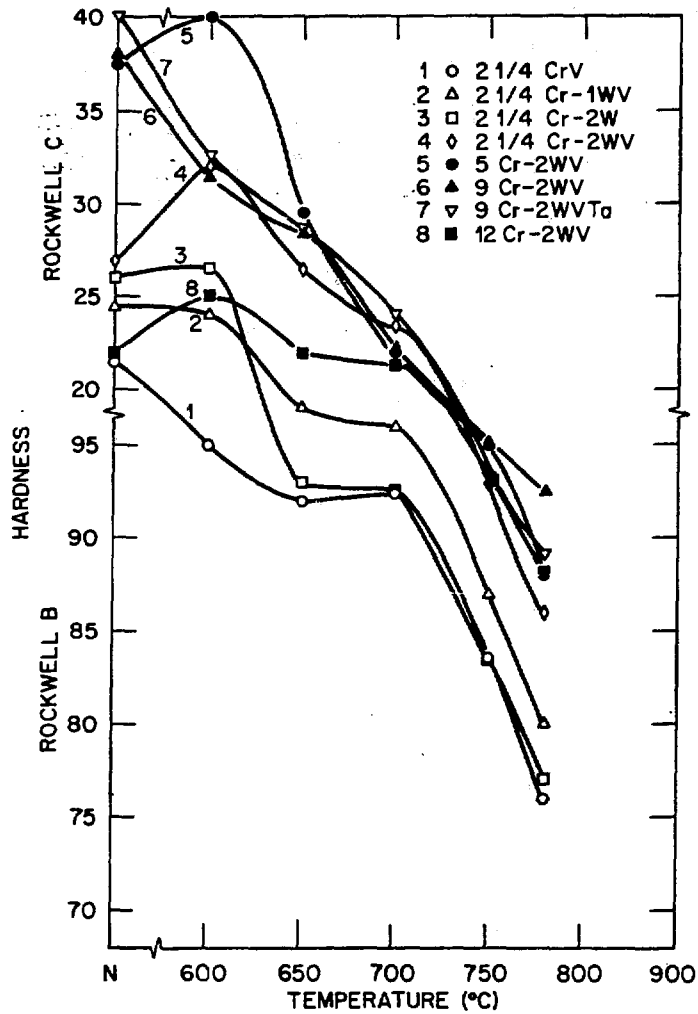
Fig. 2. Rockwell hardness plotted against tempering temperature for ORNL reduced-activation steels. Steels were normalized (N) and pieces were tempered 2 h at 600, 650, 700, 750, and 780°C.
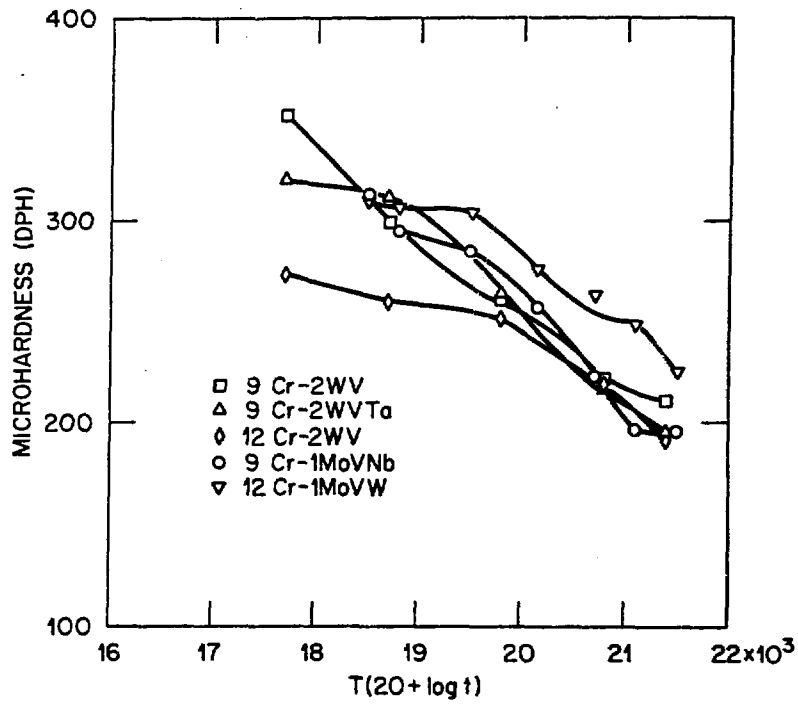
Fig. 3. Microhardness plotted against the Holloman-Jaffee tempering parameter for 9Cr-2WV, 9Cr-2WVTa, 12Cr-2WV, 9Cr-1MoVNb, and 12Cr-1MoVW steels.
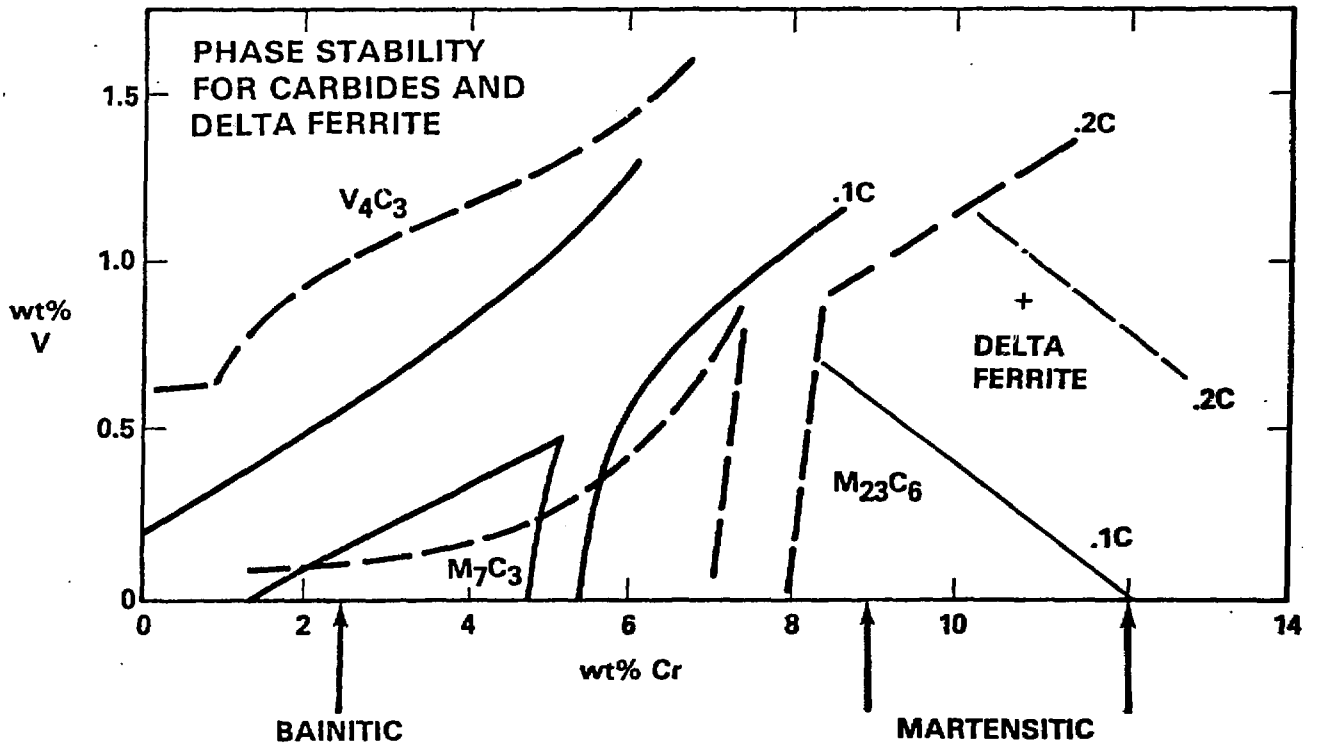
Fig. 4. Carbide phase diagram for Fe-Cr-V system after 1000 h at 700°C. The solid lines are for 0.12% C and the dashed lines for 0.2% C.
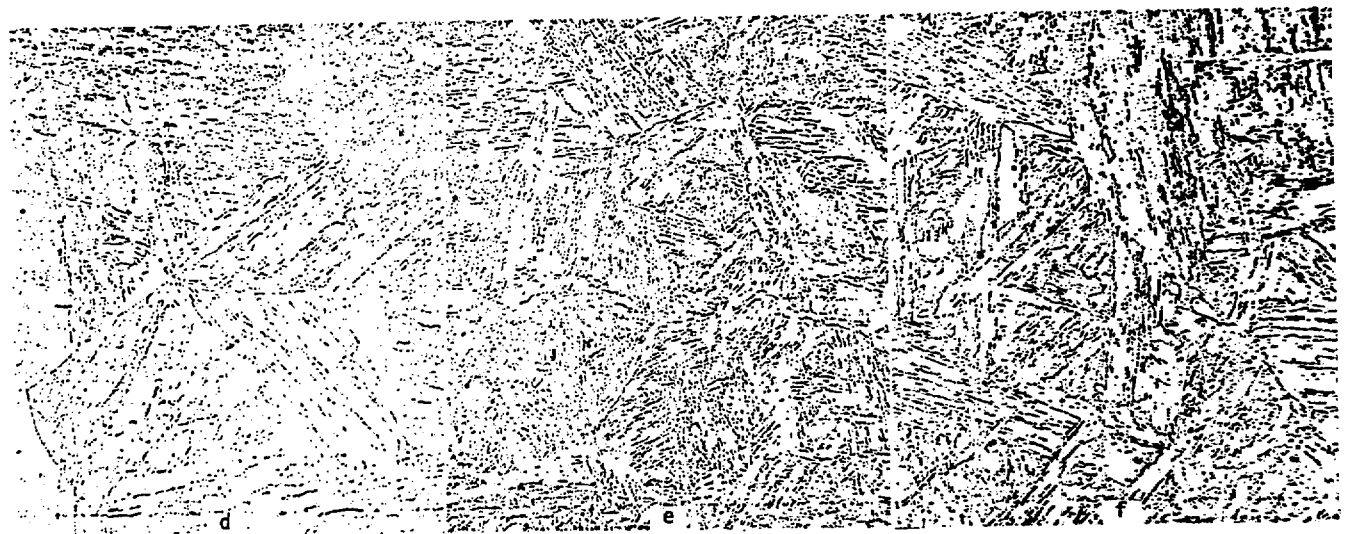
Fig. 5. Microstructures of Normalized-and-tempered HEDL reduced-activation steels: (a) 2 1/4Cr-0.5V, (b) 2 1/4Cr-1V, (c) 2 1/4Cr-1.5V-0.3Mn, (d) 9Cr-0.5V, (e) 9Cr-1.3V-1Mn, (f) 9Cr-0.5V-2Mn, (g) 9Cr-0.5V-1W-2.5Mn, (h) 12Cr-1V-6.5Mn, and (i) 12Cr-0.3V-1W-6.5Mn steels.
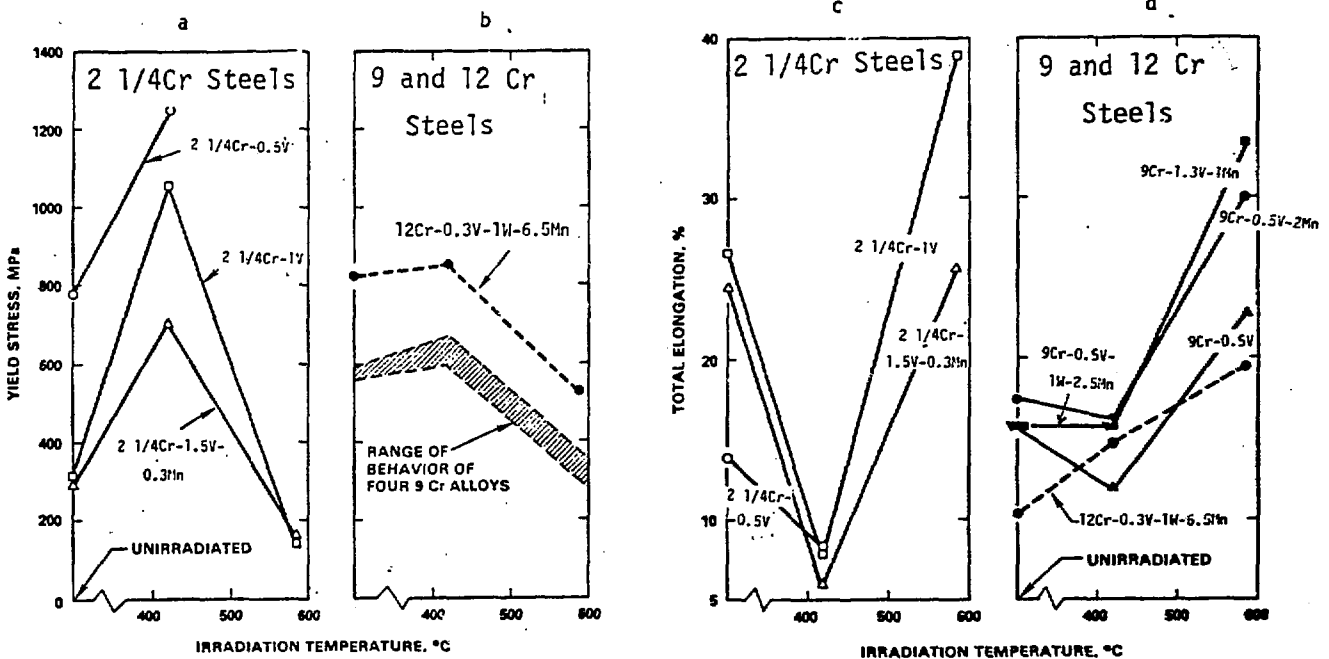
Fig. 6. Uniaxial tensile test results for HEDL reduced–activation alloys for room-temperature tests.
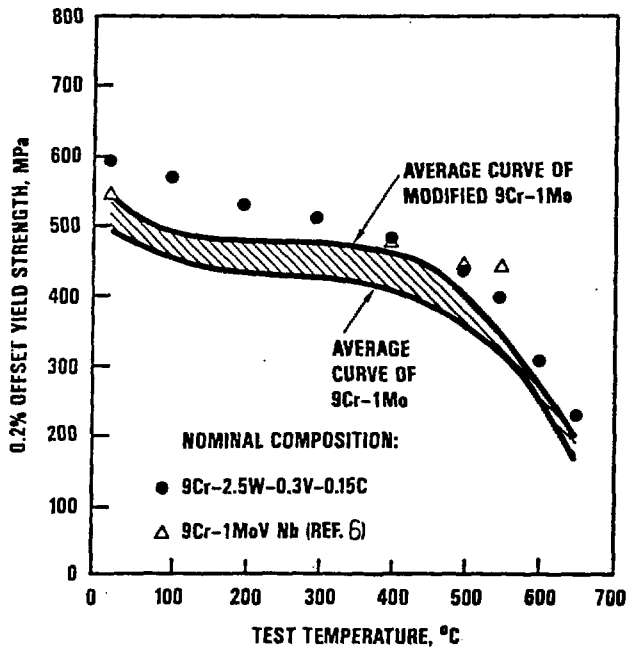
Fig. 7. The yield strength vs. temperature for the GA reduced-activation steel compared with the average curves for modified 9Cr-1Mo steel and an experimental heat of 9Cr-1MoVNb steel.
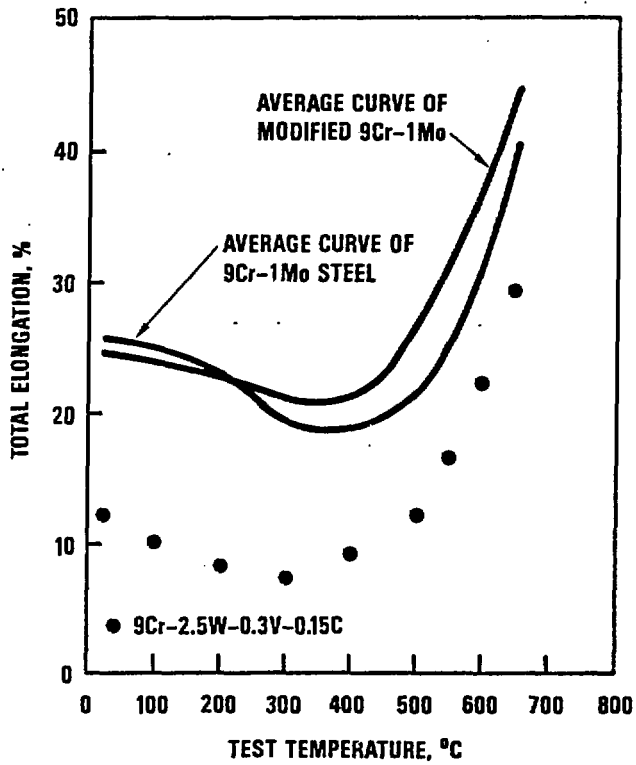
Fig. 8. Total elongation vs. temperature for the GA reduced-activation steel compared with 9Cr-1Mo steels.