

LOW-CHROMIUM REDUCED-ACTIVATION FERRITIC STEELS*

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ABSTRACT: Steels are being developed for fusion-reactor applications that contain only elements that produce radioactive isotopes that decay to low levels in a reasonable time. These reduced-activation or fast induced-radioactivity decay ferritic steels are being developed to be analogous to the Cr-Mo steels presently in the fusion program, but with molybdenum replaced by tungsten. In this paper, steels with 2% Cr will be discussed. To determine the effect of tungsten and vanadium on these steels, heats were produced with 2% W, with 0.25% V, with 1% W and 0.25% V, and with 2% W and 0.25% V. Tempering and microstructural studies were made and tensile and impact tests were conducted. Preliminary results indicate that it should be possible to develop a low-chromium Cr-W steel without molybdenum or niobium. Such steels should have properties as good as or better than the three Cr-Mo steels presently being considered as candidates for fusion-reactor applications.

KEY WORDS: reduced-activation steels, Cr-Mo steels, Cr-W steels, microstructures, tempering behavior, tensile properties, impact properties.

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Irradiation of alloys that contain molybdenum, nickel, or niobium in a fusion environment produces long-lived radioactive isotopes. This leads to difficult radioactive waste-disposal problems once the structure is removed from service. Such problems can be reduced by developing 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) [1]. We are developing reduced-activation or "fast induced-radioactivity decay" (FIRD) steels similar to the ferritic steels presently in the fusion program [2].

Ferritic steels now being considered in the United States for fusion-reactor applications are the following commercial Cr-Mo steels: 2%Cr-1Mo (2.25%Cr-1%Mo-0.1%C*), 9Cr-1MoVNB (9%Cr-1%Mo-0.2%V-0.06%Nb-0.1%C), and 12Cr-1MoVW (12%Cr-1%Mo-0.25%V-0.5%W-0.5%Ni-0.2%C) steels. Molybdenum and niobium are important alloying elements added to give these steels strength, but they also are the primary alloying elements that keep these steels from meeting the criteria for near-surface burial. It is therefore necessary to eliminate these elements in any alloys used for fusion applications. In fact, residual niobium would have to be reduced to extremely low levels [1]. Other normal alloying elements that must be minimized include nickel, copper, and nitrogen [1], none of which play a significant role in determining the properties of the Cr-Mo steels under discussion, although 0.5% Ni is added to 12Cr-1MoVW steel to avoid delta-ferrite.

Because tungsten often behaves like molybdenum in simple steels [3], it was proposed as a replacement for molybdenum [2]. To replace the

*All compositions are in weight percent.

strengthening function of niobium, the use of other reactive/refractory elements such as vanadium, titanium, and especially tantalum [2], which has many similarities to niobium, has been suggested.

Programs to develop reduced-activation ferritic steels are in progress throughout the world [4-9]. In most of those programs, tungsten is used to replace molybdenum, although vanadium is also being used [4-6]. Most of the programs have limited their studies to only a few alloys with compositions between 8 and 12% Cr [6,7-9].

Most of the work on the conventional Cr-Mo ferritic steels in the fusion program is directed toward the 9Cr-1MoVNb and 12Cr-1MoVW steels. However, there are also possible advantages of a low-chromium steel [10]. Therefore, we proposed to investigate the wider range of compositions given in Table 1.

Compositions for these new steels, referred to here as Cr-W steels, were based on variations of the base compositions of the three commercial ferritic steels currently of interest in the fusion-reactor materials program. A range of chromium compositions from ~2.25 to 12% was proposed [2]. An atom-for-atom replacement of molybdenum by tungsten was chosen; this required ~2% W, since the atomic weight of tungsten is approximately twice that of molybdenum. Vanadium was maintained at 0.25%, similar to the normal amount used in the Cr-Mo steels. A carbon level of 0.1% was maintained for all steels to help ensure weldability.

The behavior of Cr-W steels that contain 9 and 12% chromium have been discussed [11]. This paper will focus on the 2% Cr steels and compare them with 2%Cr-1Mo, 9Cr-1MoVNb, and 12 Cr-1MoVW steels. Alloys

with 2% Cr and 0% W (designated 2%CrV), 1% W and 0.25% V (2%Cr-1W), 2% W and 0.25% V (2%Cr-2WV), and with 2% W and 0% V (2%Cr-2W) were examined to determine the effect of vanadium and tungsten on properties.

EXPERIMENTAL PROCEDURE

The four 2% Cr steels given in Table 1 were electroslag remelted (ESR) by Combustion Engineering, Inc., Chattanooga, Tennessee. Melt compositions are given in Table 2. In addition to the elements of primary interest (Cr, V, W, C, and Ta), other elements such as Mn, P, Si, etc., were adjusted to composition levels typically found in commercial steels. Product forms obtained included 15.9-mm plate and 0.25- and 0.76-mm sheet.

Tensile tests were made on specimens taken from the 0.76-mm-thick sheet with a reduced gage section 20.3-mm long by 1.52-mm wide. Specimens were machined with gage lengths parallel to the rolling direction. Tests were made in vacuum on a 120-kN-capacity Instron universal testing machine at a crosshead speed of 8.5 $\mu\text{m/s}$, which results in a nominal strain rate of $4.2 \times 10^{-4}/\text{s}$.

Full-size Charpy V-notch specimens were made from the 15.9-mm-thick plate in accordance with ASTM specification E 23 with dimensions of 10 by 10 by 55 mm; the specimens contained a 2-mm-deep, 45-degree V-notch with a 0.25-mm-root radius. Specimens were taken along the rolling direction with the notch running transverse to the rolling direction (L-T orientation). Each individual Charpy data set was fitted to a hyperbolic tangent function for obtaining the transition temperature and upper-shelf energy.

Steels were tested in the normalized-and-tempered condition. All but the 2%Cr-2W steel were austenitized at 1050°C; the austenitization temperature for the 2%Cr-2W steel was 900°C. The 15.9-mm plates were austenitized 1 h in air and air cooled, and the 0.76-mm tensile specimens were austenitized for 0.5 h in a helium atmosphere and then cooled in flowing helium. (By definition, a normalized steel is air cooled after austenitization; when pulled into the flowing helium in the cold zone of the tube furnace, rapid cooling occurs, and this heat treatment will here be referred to as normalizing.) Two tempering treatments were tested: 1 h at 700°C and 1 h at 750°C. The only exception to these tempering temperatures was for the 2%Cr-1W steel, which was tempered 1 h at 725°C and 1 h at 750°C.

RESULTS

Microstructures

Detailed results from microstructural studies and precipitate phase identification were previously published [12], and they will be only briefly described here. Phases were identified on extraction replica films using both X-ray diffraction analysis of bulk precipitate extractions and X-ray energy-dispersive spectroscopy (XEDS) studies in an analytical electron microscope.

Specimens from the 15.9-mm-thick plates were examined by optical microscopy (Fig. 1) after normalizing and after normalizing and tempering at 700 and 750°C [12]. Microstructures contained bainite and polygonal (proeutectoid) ferrite. The 2%CrV alloy had 30 to 35% bainite, with the remainder being ferrite. For the 2%Cr-1W steel, the

microstructure contained ~55% bainite and 45% ferrite. The 2%Cr-2W steel was essentially 100% bainite; the 2%Cr-2WV steel contained bainite plus 15 to 20% polygonal ferrite. Optical microstructures of the normalized-and-tempered 0.76-mm-thick sheet from which the tensile specimens were made indicated that all four of the steels were entirely bainitic [12].

Transmission electron microscopy (TEM) of thin-foil specimens showed a range of microstructures. Figure 2 shows examples of fibrous [Fig. 2(a)] and interphase [Fig. 2(b)] precipitates. Such precipitates were observed in the polygonal ferrite regions of the 2%CrV, 2%Cr-1WV, and 2%Cr-2WV steels. Also shown in Fig. 2 is the bainitic structure of the 2%CrV [Fig. 2(a)] and the 2%Cr-1WV [Fig. 2(b)] steels. Bainitic structures of the 2%Cr-2W and 2%Cr-2WV are shown in Fig. 3. These regions contained a high dislocation density, and some large precipitates were visible within these regions. The bainite had the granular appearance typical of that found in 2%Cr-1Mo steel under certain heat-treatment conditions.

Extraction replicas were examined by TEM and XEDS, and these showed the diversity of the carbides in the microstructures of these steels. The 2%CrV and 2%Cr-1WV steels contained fibrous precipitates, interphase precipitates, and large blocky precipitates (Fig. 4). The fibrous and interphase precipitates occurred in the polygonal ferrite [12]. There were two types of fibers: the fine fibers that are illustrated in Fig. 4 were found to be vanadium-rich MC, and the coarser fibers or "ribbons," seen in Fig. 5 were chromium-rich M_7C_3 . A morphological characteristic of the interphase precipitation (Fig. 4) is that it generally

consists of rows of fine MC precipitates (Fig. 4). Large, blocky precipitates (Fig. 4), which were mainly rich in iron, suggesting an M_7C_3 phase, appeared in two regions: along prior austenite grain boundaries and within regions of the bainitic matrix in which there were no fibrous or interphase precipitates [12].

As demonstrated in Fig. 4, the large M_7C_3 and M_3C precipitates in the bainitic matrix regions appeared in various shapes. On average, the coarse M_7C_3 and M_3C precipitates were smaller in the 2%Cr-1WV steel than in the 2%CrV steel. In general, the bainitic regions of the 2%CrV and 2%Cr-1WV steels also contained a distribution of finer MC precipitate particles. The 2%Cr-1WV steel contained a higher density of these finer MC particles in the bainitic regions than did the 2%CrV steel.

No fibrous or ribbon precipitates were observed in the 2%Cr-2W steel, which was essentially 100% bainitic. This steel contained two types of precipitate: small needlelike particles that were mainly chromium-rich, but also rich in tungsten, suggesting an M_2X phase, and large blocky particles that appeared to be a mixture of chromium-rich M_7C_3 and iron-rich M_3C phase [Fig. 6(a)]. Only a few fibrous precipitates were observed in the 2%Cr-2WV steel, which contained 15 to 20% polygonal ferrite. Some interphase precipitate regions were detected, but the MC particles were much smaller than those found in the 2%CrV and 2%Cr-1WV steels. The bainite of the 2%Cr-2WV steel, like that of the 2%Cr-2W steel, contained two types of precipitate: large blocky M_7C_3 and M_3C particles, both containing detectable tungsten enrichment,

and a high density of very fine vanadium-rich MC particles that were also quite rich in tungsten [Fig. 6(b)]. The blocky particles in the 2%Cr-2WV were similar in appearance to those in the 2%Cr-2W steel, but the fine MC particles were smaller and more abundant. At high magnification the small particles appeared to be a mixture of needlelike and rectangular (in two dimensions) morphologies, but compositionally they were virtually the same [12].

Tempering Behavior

Tempering curves are given in Fig. 7 for the specimens taken from the 15.9-mm-thick plate and austenitized 1 h at 900°C (the 2%Cr-2W steel) or 1 h at 1050°C (the other three heats) and then air cooled. Data are for pieces of plate tempered 2 h at 600, 650, 700, 750, and 780°C. All heats showed a continuous decrease in hardness when tempered above 600°C.

Tempering resistance was affected by tungsten and vanadium composition. Steels with 1 and 2% W and 0.25% V (2%Cr-1WV and 2%Cr-2WV) displayed increasing hardness with increasing tungsten content. The 2%CrV steel (0.25% V without any tungsten) was clearly the softest. On the other hand, the 2%Cr-2W steel (2% W but no vanadium) had a higher hardness than either the 2%CrV or the 2%Cr-1WV steels after normalizing and after tempering at 600°C. After tempering at 650°C and above, the hardness of the 2%Cr-2W was less than that of the 2%Cr-1WV steel and approached that of the 2%CrV steel.

Despite the fact that the 2¼Cr-2WV steel was not entirely bainite, the as-normalized hardness was similar to the hardness of the 2¼Cr-2W steel. After tempering, the 2¼Cr-2WV steel had the highest hardness of the 2¼ Cr steels, regardless of the tempering conditions.

Tensile Properties

Tensile properties were determined over the temperature range from room temperature to 600°C. Data for the steels tempered at 700 and 750°C are plotted in Figs. 8 and 9, respectively. For all of the steels, the strength was substantially lower after tempering at 750°C than after tempering at 700°C. Relative strengths of the different steels were the same after the two different tempering treatments.

After tempering at 700°C, the effect of vanadium on the 0.2% yield stress (YS) and ultimate tensile strength (UTS) was evident [(Fig. 8(a) and 8(b))]. Except for the UTS at the highest temperature (600°C), the 2¼CrV steel was stronger than the 2¼Cr-2W steel. The vanadium effect was evident when the 2¼Cr-2W steel was compared to the 2¼Cr-2WV steel. An effect of tungsten in the presence of vanadium was also evident for these test conditions: the 2¼Cr-2WV steel was substantially stronger than the 2¼Cr-1WV steel, which, in turn, was stronger than the 2¼CrV steel. A somewhat similar relationship applied to these steels after tempering at 750°C [Fig. 9(a) and 9(b)], although the difference in strengths among the 2¼CrV, 2¼Cr-2W, and 2¼Cr-1WV steels was much smaller. Even after tempering at 750°C, however, the 2¼Cr-2WV steel remained substantially stronger than the other steels.

Total elongation results [Figs. 6(c) and 7(c)] indicated that tungsten had a favorable effect on ductility. The 2%Cr-2W steel, which was often the weakest, had the highest elongation for all test conditions. However, the 2%Cr-2WV steel, which was always the strongest, had a total elongation that was as great or greater than the values for the weaker 2%CrV and 2%Cr-1WV steels.

Impact Properties

A summary of the impact data is given in Table 3, where the ductile-brittle transition temperature (DBTT) and the upper-shelf energy (USE) are given for each steel. A DBTT was determined at the 41 and 68 J level and from lateral-expansion measurements.

Impact curves are shown in Fig. 10 for the 2% Cr steels tempered at 750°C. The 2%Cr-2W steel had the lowest DBTT and highest USE. For the vanadium-containing steels, the 2%Cr-1WV and 2%Cr-2WV steels had similar DBTTs, which were lower than the DBTT of the 2%CrV steel. Data scatter was quite large for the 2%Cr-1WV steel, and the one high point away from the trend of the other data points caused this curve fit to have a lower DBTT than would have been the case if this point had not been included. The USE of the 2%Cr-2WV steel was lower than that for the other two vanadium-containing steels, which had USE values similar to those of the 2%Cr-2W steel. When tempered at 700°C (Table 3), the 2%Cr-2W steel again had the best combination of DBTT and USE. The 2%Cr-2WV steel had the worst properties for these tempering conditions. However, the DBTT values of all four steels were considerably above those obtained after tempering at 750°C.

Discussion

Microstructures

Microstructures of the plate specimens varied according to the chemical composition. The 2% Cr alloys showed the effect of the tungsten on hardenability; the steel without tungsten contained the most polygonal ferrite, followed by the steel with only 1% W. Since no ferrite was detected in the 2%Cr-2W steel, the appearance of ferrite in the 2%Cr-2WV steel was unexpected. One possible explanation is that the vanadium carbide present in the 2%Cr-2WV steel did not completely dissolve during austenitization. This would lower the concentration of carbon in solution, which could then lower the hardenability. This possibility was anticipated when 1050°C was chosen as the austenitization temperature for the 2%Cr-2WV steel and only 900°C for the 2%Cr-2W steel. The 1050°C temperature was considered high enough to dissolve the vanadium carbide, and austenitization tests tended to confirm that conclusion because a maximum hardness was reached after austenitization at 1000°C [13].

Precipitate morphology observed in the 2% Cr steels was characteristic of the precipitation that accompanies the austenite-to-polygonal-ferrite transformation in alloy steels. Because the eutectoid composition in such steels occurs at lower carbon concentrations and higher temperatures than in the Fe-C system, classical pearlite does not appear [14]. Three different types of ferrite-carbide microstructures are observed: fibrous carbides, interphase precipitates, and carbides that form within the supersaturated ferrite during tempering, often

forming on dislocations [14]. The first two types of precipitation occur at a moving austenite-ferrite interface. Fibrous precipitates are morphologically analogous to pearlite and occur at an incoherent phase boundary; they are much closer to equilibrium than the interphase precipitates. Fine fibers, which in these steels were identified as vanadium-rich MC [12], are associated with a high transformation temperature and a slowly moving interface [14]; the coarser ribbons were identified as M_7C_3 , with chromium enrichment and substantial iron content [12]. Interphase precipitates, which were also vanadium-rich MC [12], form at lower temperatures at a low-energy interface by the movement of steps along the boundary [14].

Fibrous and interphase precipitate morphologies were present in larger quantities in the 2%CrV and 2%Cr-1W steels because they contained larger amounts of polygonal ferrite. The large blocky types of precipitate observed in the matrix of these steels occurred in the bainitic regions of the steels. Most of the precipitates in the 2%Cr-2W steel were of the blocky type, because this steel contained essentially no polygonal ferrite.

A synergistic effect of tungsten and vanadium was evident in the microstructures and tensile properties of the 2% Cr steels. The overall carbide distribution of the 2%Cr-2W steel was coarser than the carbide distributions in the steels containing vanadium, including the steel without tungsten. The tensile properties reflected this coarser precipitate distribution, because the 2%Cr-2W steel was generally the weakest or had properties approaching those of the 2%CrV steel (Figs. 8 and 9).

Tempering Studies

Although vanadium carbide is expected to play an important role in the strengthening of these steels, there was apparently an additional synergistic effect between tungsten and vanadium. The hardness of the 2%Cr-2WV steel exceeded that for the 2%Cr-1WV steel, and both of these steels had higher hardnesses than the 2%Cr-2W after tempering at 650°C and above. The greater hardness of the 2%Cr-2W steel after normalizing and after tempering at 600°C was probably due to this steel being entirely bainitic, whereas the 2%Cr-1WV steel contained significant amounts of polygonal ferrite. Early precipitation of abundant M_2C or M_2X may also have played a role in the higher hardness of 2%Cr-2W steel at the lower tempering temperatures. At the higher tempering temperatures, the presence of MC in the vanadium-containing steels with tungsten presumably caused these steels to have the higher hardness. The fact that the 2%CrV and 2%Cr-2W steels have similar hardnesses at 650°C and above is an indication that it is neither the vanadium nor tungsten by itself that provides the tempering resistance. Likewise, this shows that the 100% bainitic structure is not inherently the strongest (the 2%Cr-2W steel was the only steel with a 100% bainitic microstructure).

Although no Cr-Mo steels were tempered for comparison, some tempering data were available for 2%Cr-1Mo steel obtained under similar circumstances [13]. These data showed that the hardness of the 2%Cr-1Mo steel is similar to the 2%Cr-2W steel -- especially at the higher tempering parameters [13]. The 2%Cr-1WV and 2%Cr-2WV steels were more

resistant to tempering than 2%Cr-1Mo steel, probably because they contained vanadium. Since the 2%Cr-2W steel was taken as analogous to 2%Cr-1Mo steel, the equivalent tempering response suggests that on an atom-for-atom basis, molybdenum and tungsten play similar roles in these steels, which was the objective in replacing the molybdenum by tungsten.

Tensile Properties

Tensile properties were also affected by the combination of tungsten and vanadium. The steel with 0.25% V and no tungsten had strength properties similar to the steel with 2% W and no vanadium. The steel with 2% W had excellent ductility, and with a combination of tungsten and vanadium, a synergistic effect occurred to produce a steel with excellent strength and ductility.

The high strength of the 2%Cr-2WV steel is probably due to the fine MC that formed in this alloy. Note that the tensile properties of the 2%Cr-2WV steel were not affected by polygonal ferrite, since the tensile specimens were taken from 0.76-mm-sheet that was entirely bainite.

The objective in the development of the Cr-W steels was properties comparable to the Cr-Mo steels presently in the fusion-reactor program. This objective would be met if the 2%Cr-2W steel has properties comparable to the analogous 2%Cr-1Mo steel. This was basically the case (Fig. 11), indicating that the replacement of molybdenum by tungsten can result in steels with comparable tensile properties, although the UTS values for the 2%Cr-1Mo steel are somewhat stronger than the 2%Cr-2W. An addition of vanadium then results in steels with improved properties over those for 2%Cr-1Mo steel.

To compare the strongest 2% Cr steel with the strongest high-chromium Cr-W steel previously discussed [11] and the strongest Cr-Mo steels, Fig. 12 shows the tensile behavior of the 2%Cr-2WV steel with the 9Cr-2WVTa steel [11] and the 9Cr-1MoVNb and 12Cr-1MoVW steels. All steels were tempered at 750°C. The YS of the 2%Cr-2WV steel was higher than that of the 9Cr-2WVTa steel, which was the strongest high-chromium reduced-activation steel [11] at all test temperatures, and the UTS was higher at the highest test temperatures. Further, the two reduced-activation steels compared favorably with 9Cr-1MoVNb and 12Cr-1MoVW steels. Although the low-temperature strengths (YS and UTS) for the Cr-Mo steels were greater than those for the 2%Cr-2WV steel, at the elevated temperatures the strengths of the 2%Cr-2WV and the 9Cr-2WVTa steels were similar to or higher than the strengths of the Cr-Mo steels.

In general, the ductility values of the Cr-W steels as measured by total elongation exceeded those of the two Cr-Mo steels [Fig. 12(c)]. The 9Cr-1MoVNb steel had the lowest total elongation at all test temperatures. By contrast the 12Cr-1MoVW steel had the highest total elongation at room temperature, but then it decreased to values as low or lower than that of the Cr-W steels at higher temperatures. The 2%Cr-2WV steel had the highest total elongation at all temperatures above room temperature.

The observation that the 2%Cr-2WV steel had the highest strength of the low- and high-chromium Cr-W steels investigated was not totally unexpected, if we accept the assumption that 2% W will play the same role as 1% Mo does in the Cr-Mo steels. It is known that the major effect of chromium above 2% in the Cr-Mo steels is to increase the

hardenability and corrosion resistance [15]. Observations of equivalent strength of Cr-Mo steels of varying chromium concentration (no V, Nb, etc.) are made when appropriate microstructures are compared [15] (i.e., tempered bainite in the 2% Cr steel is compared to tempered martensite in the high-chromium steel); similar results would not be expected if the 2% Cr steel contained large amounts of polygonal ferrite.

The strengths of 9Cr-1MoVNb and 12Cr-1MoVW steels have often been compared to the strength of 2%Cr-1Mo steel and found to be superior [16]. However, such comparisons ignore the presence of the strong carbide-forming elements like vanadium and niobium in the 9Cr-1MoVNb and 12Cr-1MoVW steels, and ignore the important role these elements play in determining the strength of the higher chromium steels. It has been pointed out that the addition of a strong carbide former to 2%Cr-1Mo steel should also result in improved properties relative to the higher chromium steels [10]. The use of a lower chromium ferritic steel for first walls of future fusion reactors may have several advantages [10]. As shown in the present study, not only is the development of such steels possible, but they can have strengths similar to higher-chromium steels. Presumably this would also apply to Cr-Mo steels as well as Cr-W steels, and could be important for non-nuclear, elevated-temperature applications.

It is difficult to adequately compare the properties of the steels examined in the present work with the reduced-activation steels being developed elsewhere [4-9], because different heat treatments are used by the different investigators, thus not allowing for a common comparison.

These preliminary tests indicated that the 24Cr-2WV and 9Cr-2WVTa steels have tensile properties comparable to the strongest Cr-Mo steels in the fusion program -- 9Cr-1MoVNb and 12Cr-1MoVW steels. With more development work, it should be possible to further improve on the properties. As noted above, the 24Cr-2W steel has properties similar to those of 24Cr-1Mo steel. In commercial practice, the 24Cr-1Mo steel is used for elevated-temperature non-nuclear applications up to 550 to 600°C.

Impact Properties

For fusion-reactor applications, impact properties are expected to be crucial. When irradiated by neutrons, the DBTT of 12Cr-1MoVW steel can increase by over 200°C [17,18]. Therefore, it may be desirable that any steel used for such applications have as low a DBTT as possible in the unirradiated condition.

Although these steels are at an early stage of development, the impact properties of the 24Cr-2WV steel were somewhat disappointing, in view of the excellent tensile properties of this steel. In particular, the DBTT of the 24Cr-2WV steel was substantially above that of 9Cr-2WVTa steel [11].

Table 3 lists the impact properties for 9Cr-1MoVNb and 12Cr-1MoVW steels with heat treatments similar to the Cr-W steels [11]. After tempering at 750°C, the impact properties of the 24Cr-2WV steel were equivalent to those of the 9Cr-1MoVNb and 12Cr-1MoVW steels. Granted, the 750°C tempering temperature does not give optimum impact properties for these two steels [19,20], but these heat-treating conditions were

used as a common basis of comparison. However, just as the impact properties of the 9Cr-1MoVNb and 12Cr-1MoVW steels can be improved by appropriate heat treatment, so can those of the 2%Cr-2WV steel.

A mixed structure of tempered bainite and polygonal ferrite (the microstructure of the 2%Cr-2WV) is known to cause an adverse effect on impact behavior compared to a microstructure made up of a single constituent [21]. This may explain why the 2%Cr-2W steel had the best impact behavior of the 2% Cr steels (Table 3 and Fig. 10). The observation that the impact behavior of a 100% bainitic steel is superior to one containing polygonal ferrite means that it should be possible to improve the impact properties of the 2%Cr-2WV steel by heat treatment and by increasing the hardenability. By proper alloying, the hardenability of the steel can be increased and the ferrite eliminated. Finally, the type of bainite that forms can also affect the impact properties [22]; the type of bainite, in turn, depends on the hardenability.

Summary and Conclusion

Four reduced-activation (FIRD) ferritic steels with 2% Cr and varying amounts of tungsten and vanadium were obtained. Optical microstructures of the normalized-and-tempered steels showed an effect of tungsten on hardenability. The 2%CrV and 2%Cr-1WV contained 70% and 45% polygonal ferrite, respectively, the balance bainite, while the 2%Cr-2WV steel contained 15 to 20% ferrite, the balance bainite. The 2%Cr-2W steel was 100% bainite.

Hardness measurements as a function of different tempering treatments indicated that the tempering resistance of the 2%Cr-2WV exceeded that of the other steels. Tensile studies indicated that the 2%Cr-2W steel with its atom-for-atom replacement of molybdenum by tungsten resulted in properties similar to those of the analogous 2%Cr-1Mo steel, when both were heat treated similarly. The tensile properties of the 2%Cr-2WV steel compared favorably to those of 9Cr-1MoVNb and 12Cr-1MoVW steels -- the strongest steels presently in the fusion-reactor program.

Despite excellent tensile properties, impact properties of the 2%Cr-2WV steel were inferior to those of the high-chromium Cr-W steels. This difference was attributed to microstructure (15-20% polygonal ferrite), and the development of a low-chromium Cr-W steel with good strength and impact behavior should be possible with further alloying and proper heat treatment.

Results from these preliminary tests indicate that it should be possible to develop low-chromium FIRD ferritic steels with properties similar to or better than those of the ferritic steels presently being considered in the fusion program.

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TABLE 1—Proposed nominal compositions for fast induced-radioactivity decay steel development program

Alloy	Nominal Chemical Composition ^a (wt %)				
	Cr	W	V	Ta	C
2½CrV	2.25	...	0.25	...	0.1
2½Cr-1W	2.25	1	0.25	...	0.1
2½Cr-2W	2.25	2	0.1
2½Cr-2WV	2.25	2	0.25	...	0.1
5Cr-2WV	5	2	0.25	...	0.1
9Cr-2WV	9	2	0.25	...	0.1
9Cr-2WVTa	9	2	0.25	0.12	0.1
12Cr-2WV	12	2	0.25	...	0.1

^aBalance iron.

TABLE 2—Compositions for fast induced-radioactivity
decay (FIRD) ferritic steels

Element	Chemical Composition ^a (wt %)			
	2¼CrV Heat 3785	2¼Cr-1WV Heat 3786	2¼Cr-2W Heat 3787	2¼Cr-2WV Heat 3788
C	0.11	0.10	0.11	0.11
Mn	0.40	0.34	0.39	0.42
P	0.015	0.015	0.016	0.016
S	0.006	0.006	0.005	0.006
Si	0.17	0.13	0.15	0.20
Ni	0.01	0.01	<0.001	<0.001
Cr	2.36	2.30	2.48	2.42
Mo	0.01	<0.01	<0.01	...
V	0.25	0.25	0.009	0.24
Nb	<0.01	<0.01	<0.001	...
Ta	<0.01	<0.01	<0.001	...
Ti	<0.01	<0.01	<0.001	...
Co	0.005	0.006	0.008	...
Cu	0.02	0.025	0.03	...
Al	0.02	0.02	0.02	0.021
B	<0.001	<0.001	0.001	...
W	...	0.93	1.99	1.98

^aBalance iron.

TABLE 3—Impact properties of 16 -chromium FIRD steels

Steel	Tempering Temperature ^b (°C)	Impact Properties ^a			
		TT _{41J} (°C)	TT _{68J} (°C)	TT _{Le} (°C)	USE (J)
2½CrV	700	85	86	85	240
	750	66	69	70	318
2½Cr-1WV	725	52	53	52	220
	750	8	23	38	340
2½Cr-2W	700	24	24	12	260
	750	-41	-30	-31	324
2½Cr-2WV	700	85	110	112	131
	750	31	31	31	265
9Cr-1MoVNb	700	56	68	68	161
	750	27	41	41	199
12Cr-1MoVW	700	33	68	64	99
	750	4	29	26	115

^aTT_{41J} is 41-J (30 ft-lb) transition temperature; TT_{68J} is 68-J (50 ft-lb) transition temperature; TT_{Le} is lateral expansion transition temperature as determined by 0.889-mm expansion; USE is upper-shelf energy.

^bAll steels were tempered 1 h; before tempering all but the 2½Cr-2W were normalized at 1050°C; the 2½Cr-2W was normalized at 900°C.

List of Figures

Fig. 1. Optical microstructures of normalized-and-tempered 15.9-mm-thick plate of (a) 2%CrV, (b) 2%Cr-1WV, (c) 2%Cr-2W, and (d) 2%Cr-2WV steels.

Fig. 2. Transmission electron microscopy photomicrographs of (a) 2%CrV and (b) 2%Cr-1WV steels showing examples of fibrous [left side of (a)] and interphase [left side of (b)] precipitation in the polygonal ferrite along with the bainitic structure (right side of each photomicrograph) of each steel.

Fig. 3. Transmission electron microscopy photomicrographs of the bainitic microstructures of (a) 2%Cr-2W and (b) 2%Cr-2WV steels.

Fig. 4. Extraction replica that shows an example of the fibrous, interphase, and blocky precipitates that occur in the polygonal ferrite of the normalized-and-tempered 2%CrV.

Fig. 5. Extraction replica that shows an example of coarser fiber or "ribbon" precipitates that occur in the polygonal ferrite of the normalized-and-tempered 2%CrV.

Fig. 6. Extraction replicas showing the general precipitate distribution of normalized-and-tempered (a) 2%Cr-2W and (b) 2%Cr-2WV steels.

Fig. 7. Rockwell hardness plotted against tempering temperature for 15.9-mm-thick plates of the four 2% Cr steels after normalizing (N) and tempering 2 h at 600, 650, 700, 750, and 780°C.

Fig. 8. Tensile properties as a function of temperature for the 2% Cr steels tempered at 700°C: (a) 0.2% yield stress, (b) ultimate tensile strength, and (c) total elongation.

Fig. 9. Tensile properties as a function of temperature for 2% Cr steels tempered at 750°C: (a) 0.2% yield stress, (b) ultimate tensile strength, and (c) total elongation.

Fig. 10. The Charpy V-notch impact curves for 2%CrV, 2%Cr-1WV, 2%Cr-2W, and 2%Cr-2WV steels; all steels were tempered 1 h at 750°C.

Fig. 11. A comparison of tensile properties for 2%CrV, 2%Cr-1WV, 2%Cr-2W, and 2%Cr-2WV steels with 2%Cr-1Mo steel: (a) 0.2% yield stress, (b) ultimate tensile strength, and (c) total elongation. Steels were normalized and then tempered 1 h at 700°C.

Fig. 12. A comparison of the tensile properties for 2%Cr-2WV and 9Cr-2WVTa steels -- the strongest Cr-W steels -- with the properties for 9Cr-1MoVNb and 12Cr-1MoVW steels: (a) 0.2% yield stress, (b) ultimate tensile strength, and (c) total elongation. Steels were normalized and then tempered 1 h at 750°C.

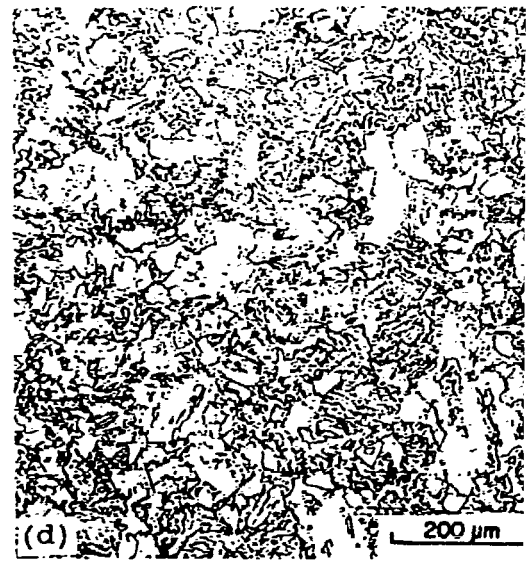
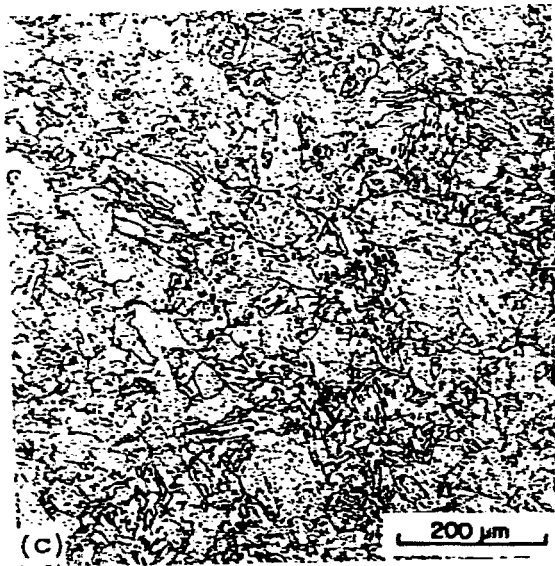
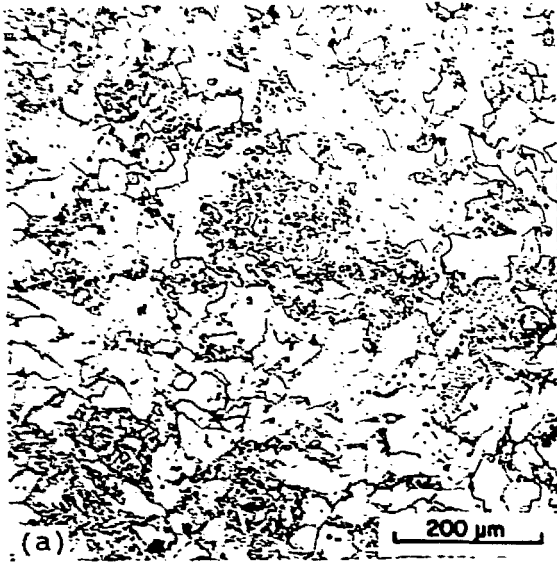


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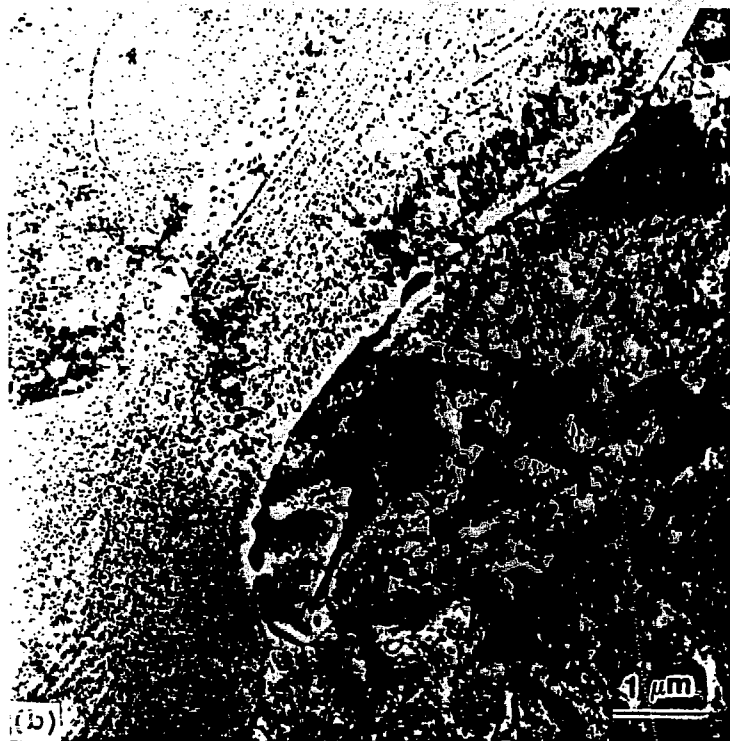
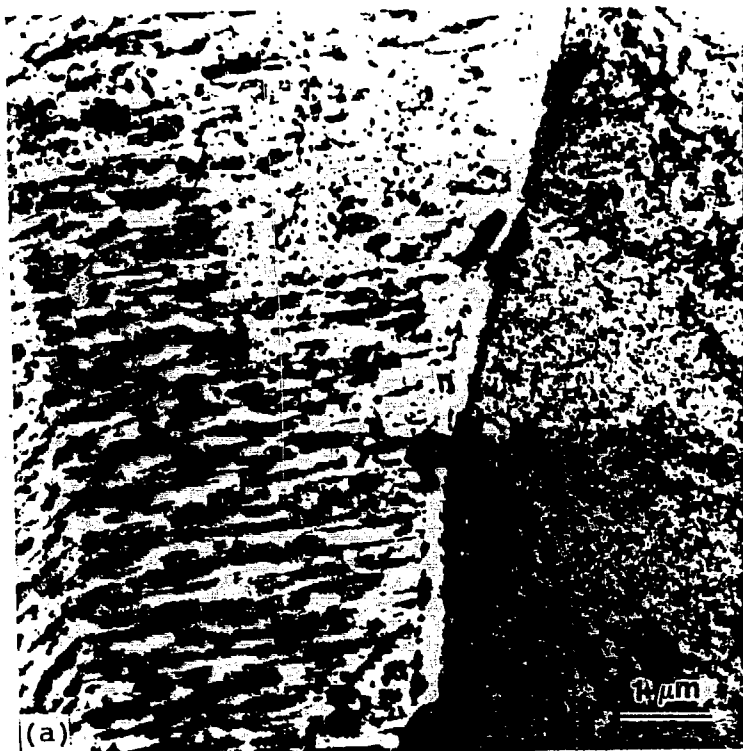


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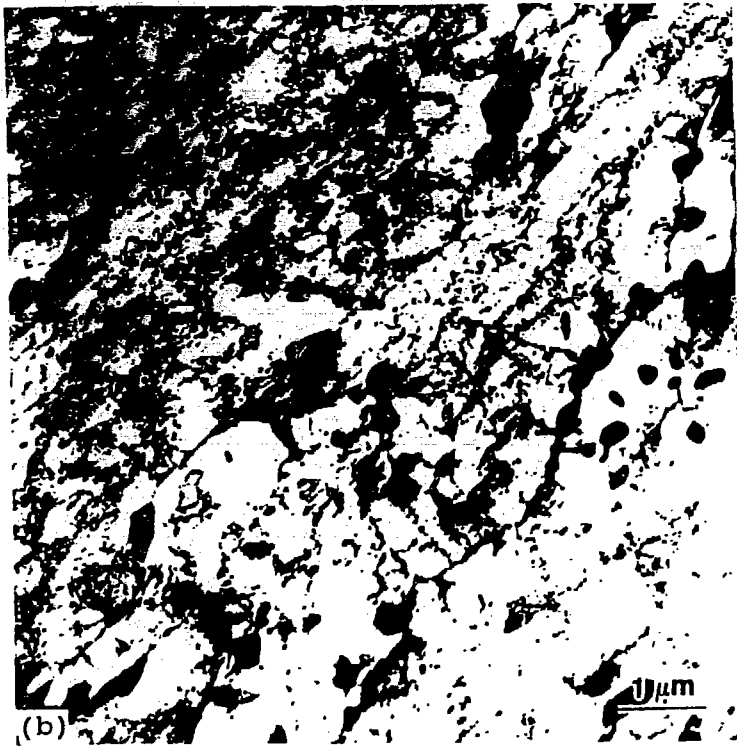
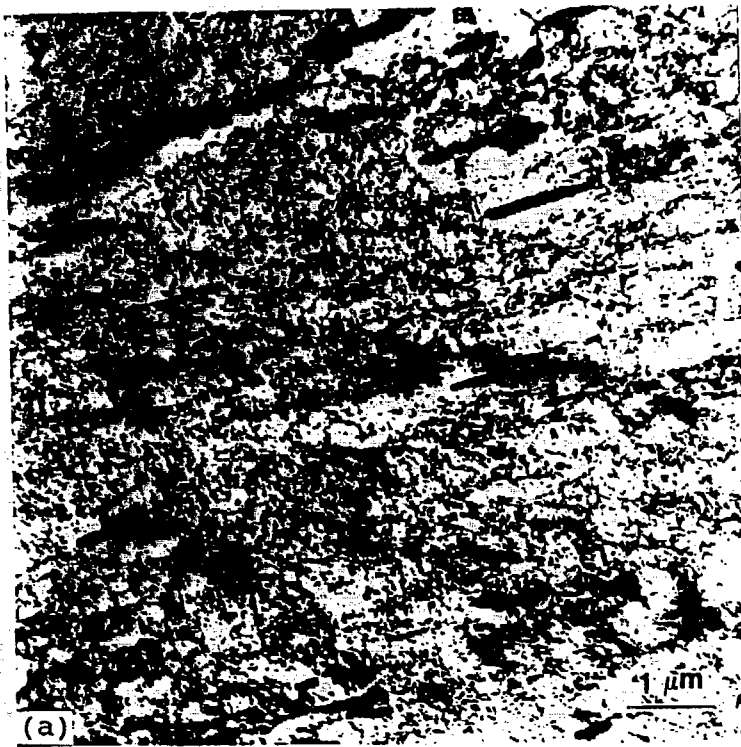


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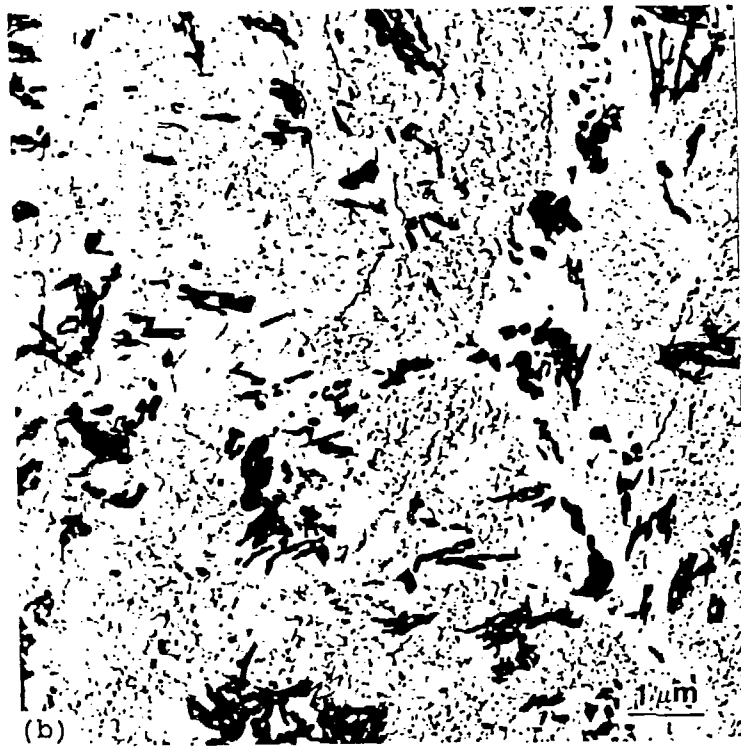


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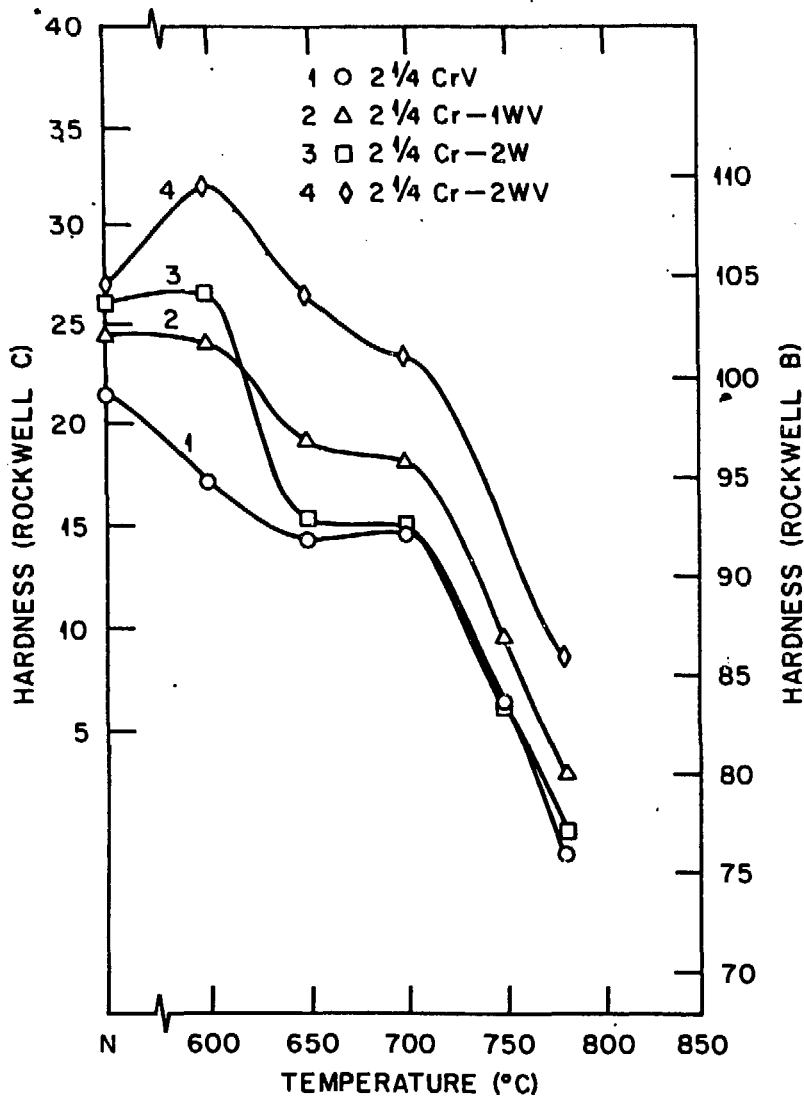


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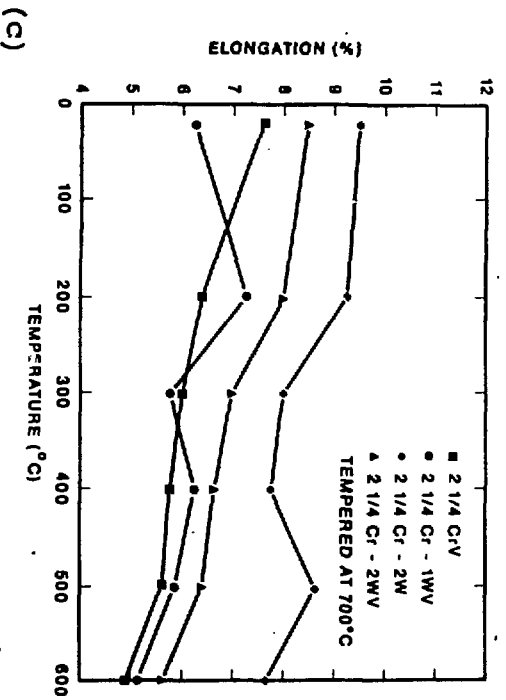
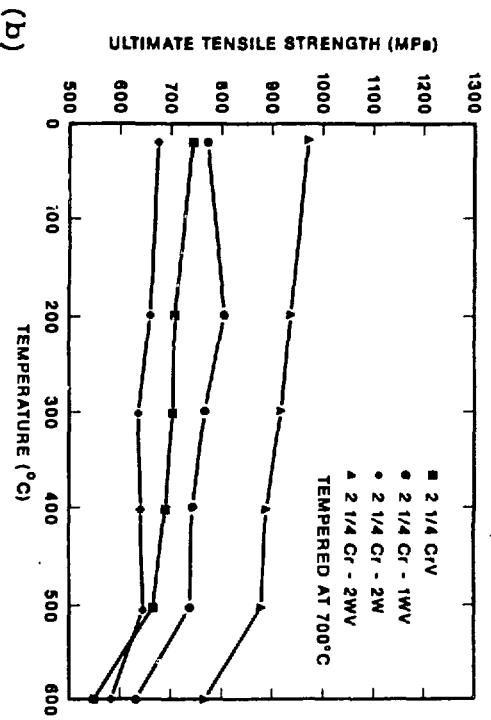
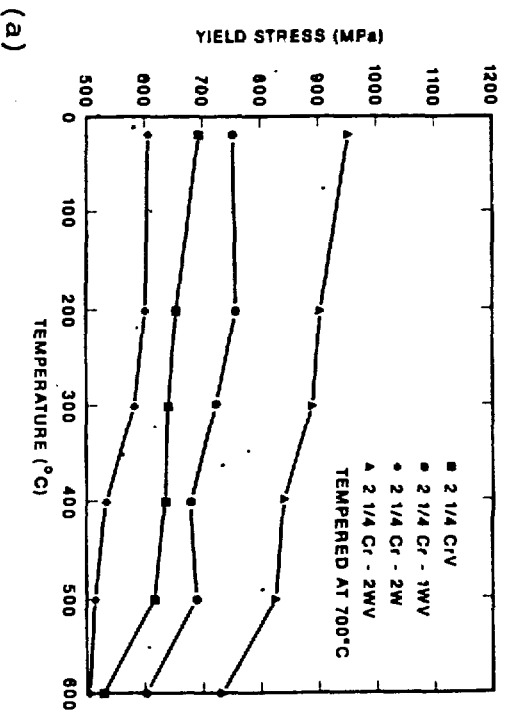
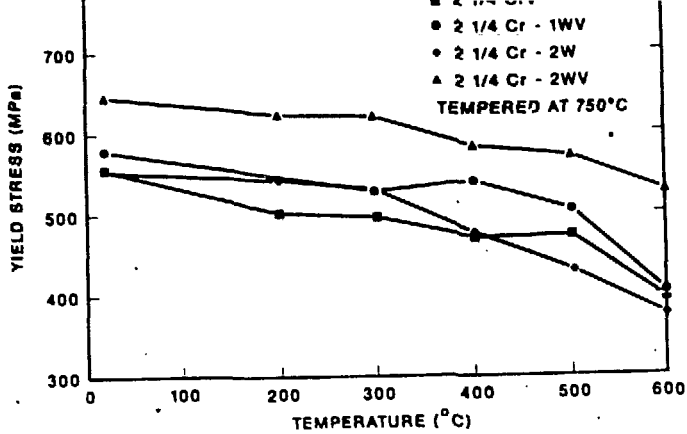
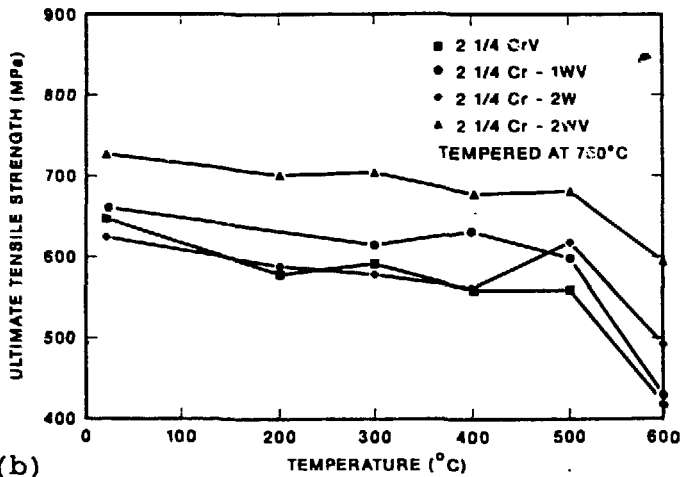


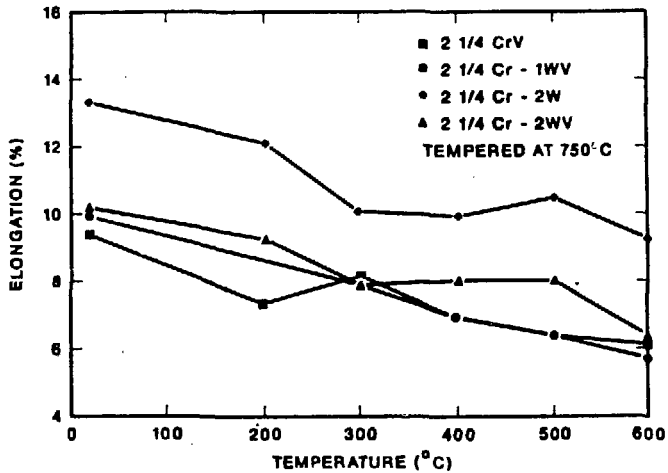
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(a)



(b)



(c)

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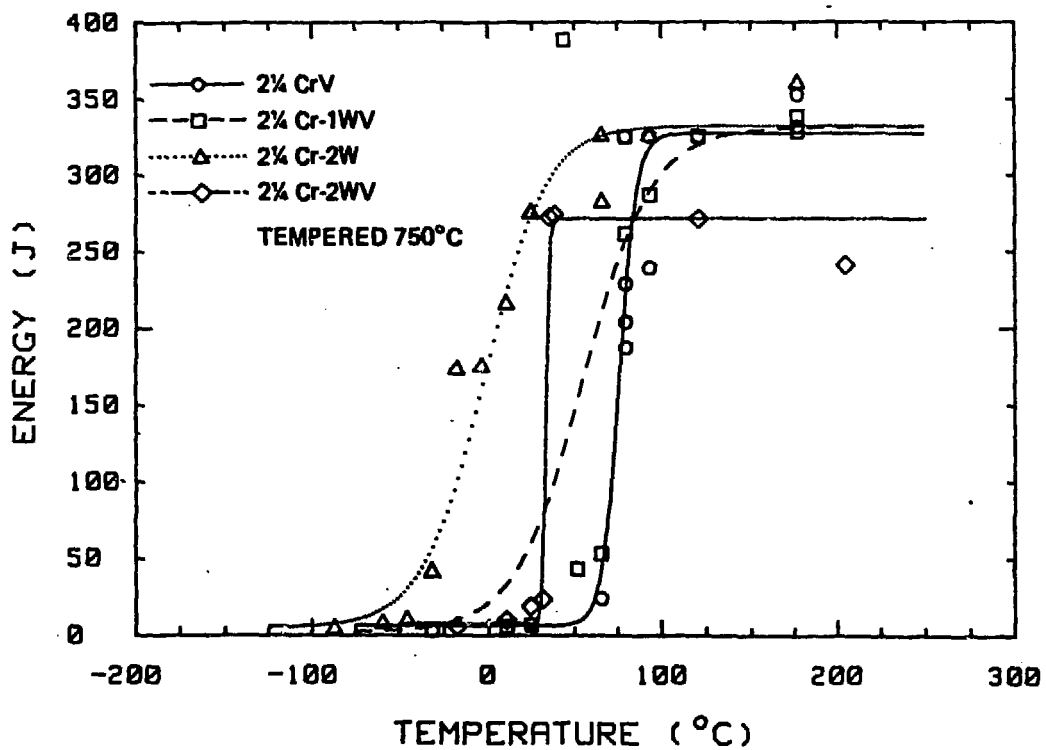
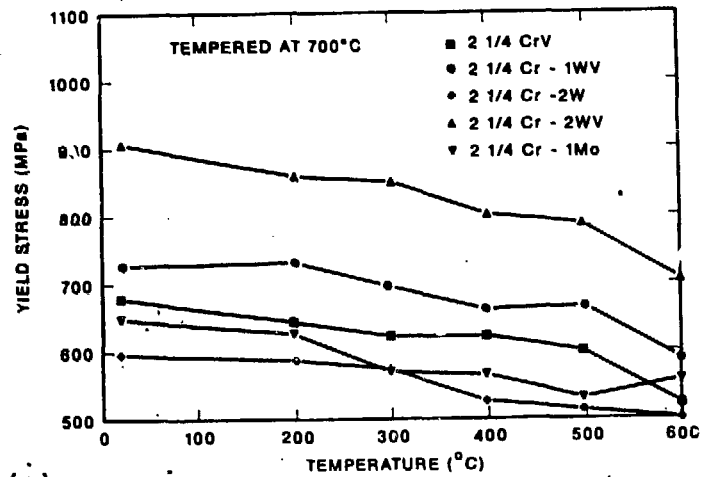
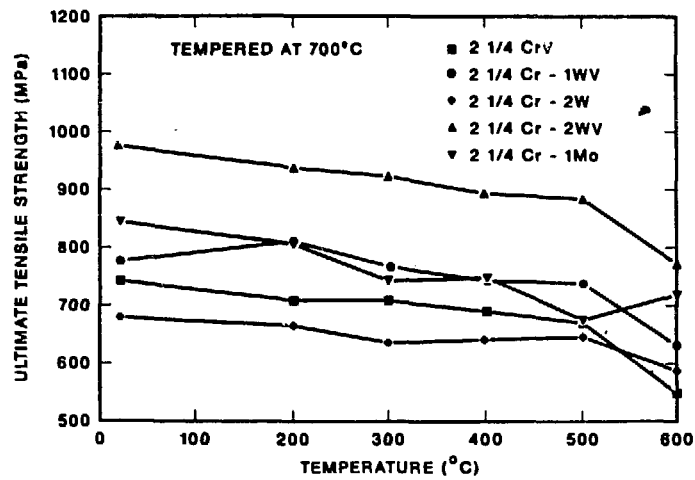


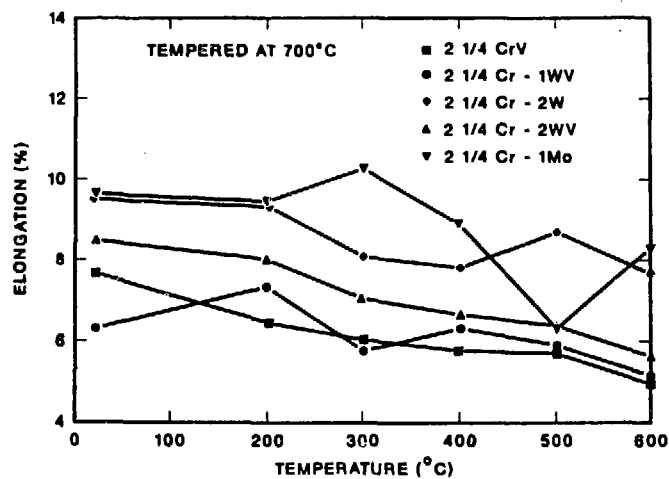
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(a)

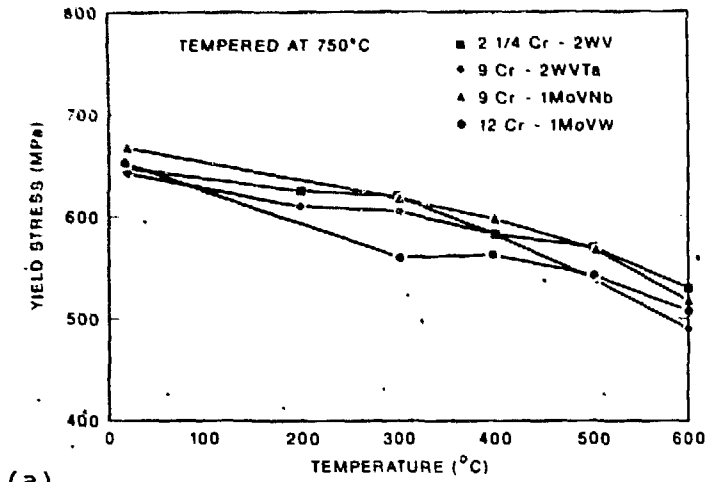


(b)

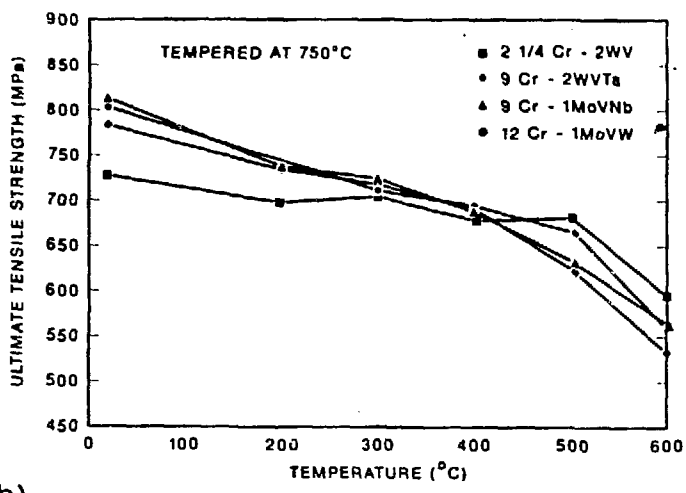


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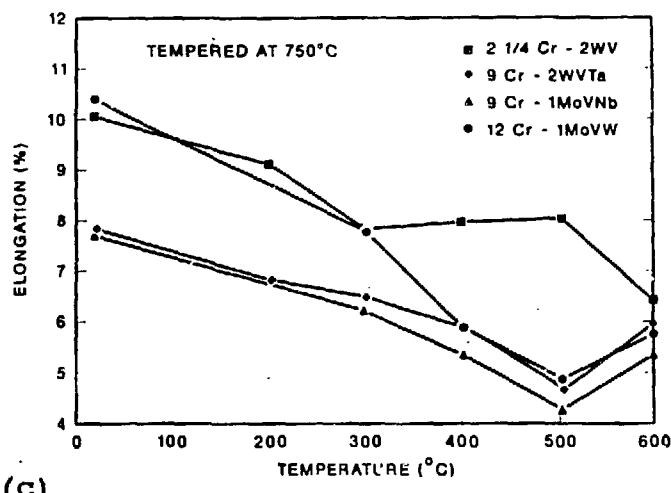
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(a)



(b)



(c)

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