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THE POSTIRRADIATION TENSILE PROPERTIES AND MICROSTRUCTURE OF
SEVERAL VANADIUM ALLOYS*

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ABSTRACT: Tensile specimens of V-15Cr-5Ti, Vanstar-7, V-3Ti-1Si, and V-20Ti were irradiated at 420°C in FFTF-MOTA to a damage level of 82 dpa. Helium was preimplanted to levels up to 480 appm in selected specimens using a modified tritium trick. Irradiation hardening was the dominant effect influencing the postirradiation tensile properties, and it markedly increased the yield strength and reduced the total elongation. The V-15Cr-5Ti alloy was very sensitive to helium embrittlement, but Vanstar-7 and V-3Ti-1Si were only slightly affected. Without helium, negligible swelling was found in V-15Cr-5Ti and Vanstar-7, and only small values of swelling (<1%) were measured in V-3Ti-1Si and V-20Ti. Preimplanted helium increased swelling in V-3Ti-1Si by increasing cavity nucleation.

KEY WORDS: Vanadium alloys, neutron irradiation, irradiation hardening, helium embrittlement, void swelling.

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INTRODUCTION

At the present time, vanadium alloys are among the candidate materials being considered for the first wall of a fusion reactor. One of the primary reasons for considering vanadium alloys for this application is their inherent low residual radioactivity following irradiation. This property impacts on maintainence, safety, waste management, and, very importantly, public acceptance of fusion reactors for power generation [1]. Vanadium alloys also have good mechanical properties in the design temperature range (450 to 800°C) [2], good compatibility with lithium [2], and relatively high thermal conductivity and low thermal expansion properties (for lower thermal stresses). Over the past few years, a number of the more promising vanadium alloys developed for fuel claddings on the breeder reactor program have been screened for possible fusion applications. The alloys include: V-15Cr-5Ti, Vanstar-7, V-3Ti-1Si, and V-20Ti. Specimens of these alloys are being irradiated in the Fast Flux Test Facility (FFTF), Materials Open Test Assembly (MOTA) experiment at 420, 520, and 600°C. The specimens were encapsulated in ^7Li to protect them from picking up interstitial contamination at these elevated temperatures. Since the neutron energies available in FFTF do not produce the amounts of helium expected under fusion conditions (about 5 appm/dpa), helium is added before irradiation by

doping selected specimens with ^3He using a modified tritium trick [3]. A few specimens were also included that were injected with ^4He using an accelerator.

This paper presents the results of an irradiation experiment in which a damage level of 82 dpa was reached. The results include those for: postirradiation tensile tests, scanning electron microscopy (SEM) examinations of fracture surfaces, and transmission electron microscopy (TEM) examinations of the irradiated microstructures. These results are discussed and the alloys are evaluated with regard to their resistance to irradiation hardening, helium embrittlement, and swelling.

EXPERIMENTAL PROCEDURE

Three vanadium alloys were studied: V-15Cr-5Ti, Vanstar-7, and V-3Ti-1Si. A fourth alloy, V-20Ti, was evaluated for cavity swelling only. The source, chemistry, and final heat treatment for each alloy are given in Table 1. Miniature tensile specimens (Fig. 1) were machined from 0.76-mm-thick sheet of V-15Cr-5Ti and Vanstar-7, and from 0.51-mm-thick sheet of V-3Ti-1Si. Disks, 3-mm in diameter, were punched from 0.25-mm-thick sheets of all four vanadium alloys and used for transmission electron microscopy.

Selected disks and tensile specimens were doped with ^3He using a modified "tritium trick" technique [3]. In this procedure, the specimens were held under a tritium pressure of 53 kPa (400 torr) at 400°C while the decay process took place. The levels of ^3He ranged from 23 to 480 appm. Several V-15Cr-5Ti and V-3Ti-1Si disks were injected with ^4He at 420°C , using an accelerator, in order to compare the resultant helium bubble distributions and subsequent swelling with those in the ^3He -doped specimens. The ^4He levels were either 10 or 100 appm.

Before irradiation, tensile specimens and disks were sealed (by welding) in 9.5 mm OD x 8.3 mm ID x 38.1 mm long TZM subcapsules containing ^7Li . During the irradiation at elevated temperatures, the lithium protects the vanadium alloys from picking up interstitial contaminants such as C, N, and O. Additional specimens were encapsulated and used as thermally aged control specimens. This encapsulation work was provided by the Westinghouse Hanford Corporation [4]. The subcapsules were irradiated in the FFTF, Materials Open Test Assembly (MOTA) experiment at 420, 520, and 600°C to a fluence of approximately $1.4 \times 10^{23} \text{ n/m}^2$, which produced a damage level of about 82 dpa in the vanadium alloys. Unfortunately, the subcapsules at 520 and 600°C experienced temperature excursions above the desired temperatures during the irradiation. Due to the

uncertainties in interpretation resulting from these excursions, these specimens were not examined.

Following the irradiation, the subcapsules irradiated at 420°C were cut open and the ^7Li was dissolved using liquid ammonia. The specimens were further cleaned in successive baths consisting of: a mixture of ethyl alcohol and water (1:1, by volume), water, and ethyl alcohol. Both the irradiated and thermally aged control specimens were tested in a high vacuum furnace facility mated to a tabletop tensile-testing machine [5]. The pressure during testing was always less than 10^{-4} Pa (10^{-6} torr). The crosshead speed was 8.5×10^{-6} m/s which produced a strain rate of $1.1 \times 10^{-3} \text{ s}^{-1}$.

After testing, the fracture surfaces of selected specimens were examined using a hot cell SEM facility [6]. Irradiated TEM disks were dual jetpolished at -30°C in a solution of 14 vol % conc. H_2SO_4 in methanol. Cavity swelling was determined for the polished foils by measuring and counting the cavities observed in TEM micrographs with a Zeiss particle size analyzer. Foil thickness was measured using stereo techniques. A final value for cavity swelling (V/V_0) was calculated as:

$$V/V_0 = V / (V_f - V) ,$$

where V is the total cavity volume, V_0 is the original metal volume, and V_f is the final volume of the foil, as described in ASTM E521 [7].

RESULTS

Postirradiation Tensile Tests

The results of the tensile tests for the irradiated and thermally-aged control specimens are given in Table 2. Note that no yield stress could be determined for specimens possessing little or no ductility (total elongation), and only an ultimate tensile strength is listed. However, if a specimen exhibited even small amounts of plastic strain, a yield strength was measured. From these data, two important irradiation effects can be explored: irradiation hardening and helium embrittlement.

Irradiation hardening can be measured by the increase in the yield strength, as shown for three vanadium alloys in Fig. 2. The yield strength of specimens without helium is plotted against the damage level so that the most recent results, at 82 dpa, can be compared with those obtained from previous MOTA experiments at lower damage levels. It is seen that neutron irradiation essentially doubles the yield strength of all three alloys, after which saturation occurs. The damage level needed to cause saturation of the hardening

depends on the alloy. Apparent saturation occurred in the V-3Ti-1Si alloy after less than 10 dpa, but approximately 40 dpa was required for V-15Cr-5Ti and Vanstar-7. Note that for the two highest dpa values for the V-15Cr-5Ti, the ultimate strengths were plotted as yield strengths, even though true yield values were never reached due to premature failure of the specimens. By referring to Table 1, the reader can verify that at 82 dpa the presence of helium usually produced further increases in yield strength for the Vanstar-7 and V-3Ti-1Si.

Along with the increases in yield strength due to irradiation shown in Fig. 2, one would expect concurrent losses in ductility. Such is the case, as demonstrated clearly in Fig 3 for the three irradiated vanadium alloys without helium. Again, the most recent data at 82 dpa is shown in context with earlier data at lower damage levels. At the highest damage level, Vanstar-7 exhibited the least loss of ductility or total elongation, followed by V-3Ti-1Si. The V-15Cr-5Ti had zero elongation, which is not surprising since the alloy had already exhibited a zero value after only 10 dpa.

Helium embrittlement also affects specimen ductility in an adverse way and may lower the total elongation even further as shown in Fig. 4. In this figure, total elongation is plotted as a function of helium level in the

different vanadium alloys. Results for the unirradiated, thermally aged controls as well as the starting material are given. The data for V-15Cr-5Ti (Fig. 4a) show that both the starting material and aged controls behave similarly and that both were very sensitive to the amounts of helium implanted by the tritium trick. At a level of 300 appm, the material was completely embrittled. With no helium present, the irradiated material was also completely embrittled due to irradiation hardening [8]. An intermediate point at 80 appm was found to fail due to a mixture of both irradiation hardening (cleavage fracture) and helium embrittlement (intergranular fracture) [8]. Helium had very little effect on the ductility of the Vanstar-7 alloy (Fig. 4b) in both the starting and aged conditions, and only a minor effect in the irradiated specimens. However, irradiation hardening reduced the elongation of Vanstar specimens (without helium) from above 20%, down to about 6%. The results for V-3Ti-1Si (Fig. 4c) show that both the starting and aged material were only slightly affected by helium. It is also seen that the aged controls were consistently less ductile than the starting material. Preliminary examinations of both the starting and aged material failed to reveal any obvious reason for the difference in behavior of these two specimen conditions of V-3Ti-1Si. As in the case of the Vanstar-7, the ductility of the irradiated V-3Ti-1Si was not strongly influenced by the presence of helium, at least for levels below 500 appm.

Therefore, except for V-15Cr-5Ti, helium caused substantially less embrittlement than irradiation hardening.

Fractographic Examination

The fracture surfaces of the tested specimens were examined by SEM to aid in the determination of the failure mechanism. The fractographs in Fig. 5 characterize the behavior for each alloy by showing the fracture surfaces for specimens irradiated to 82 dpa. Those in the left column were taken of specimens without helium while those on the right represent specimens with the highest helium level for that particular alloy. Without helium, the V-15Cr-5Ti failed by cleavage (Fig. 5a) which is the mode of failure expected in a specimen that was hardened by radiation. With 300 appm ^3He (Fig. 5b), the dominant fracture mode is still cleavage, but there are also areas of intergranular fracture caused by the separation of helium-weakened grain boundaries. At the same helium level without irradiation, the V-15Cr-5Ti specimens will fail (with no elongation) in a completely intergranular fashion [1]. Therefore, it is clear that either irradiation or helium are capable of totally embrittling the V-15Cr-5Ti alloy. However, under the present range of conditions, irradiation (hardening) appears to be the dominant mechanism controlling the mode of failure.

The Vanstar-7 alloy exhibited substantial hardening due to the irradiation to 82 dpa (Fig. 2) but at the same time retained considerable total elongation (Figs. 3 and 4). This behavior is reflected in the fracture surfaces in specimens with and without helium, as shown in Figs. 5c and 5d, respectively. Both show the classic "ductile dimples" characteristic of a ductile fracture mode. Nearly the same results were observed for the V-3Ti-1Si fracture surfaces in Figs. 5e & 5f, except that the surfaces did not quite develop the classic ductile appearance found in the Vanstar-7.

The fracture surfaces of several thermally aged control specimens of V-3Ti-1Si were also examined because these specimens had somewhat less ductility than the starting material with comparable helium contents (see Fig. 4c). Figures 6a and 6b are fractographs of this alloy without helium and with 480 appm ^3He respectively. The aged control, without helium, had a fracture surface characteristic of ductile fracture (Fig. 6a), but the one with helium failed by intergranular separation (Fig. 6b). This latter result is surprising because the same alloy with the same helium level exhibited a ductile fracture after 82 dpa (Fig. 5f). These apparently contradictory observations will be addressed later in the discussion section.

Microstructural Examination

Important microstructural features of the irradiated vanadium alloys include damage structure, helium bubbles, and cavities. The damage structures of the three alloys and also V-20Ti were generally similar. A typical example of such a structure is shown in Fig. 7 for the Vanstar-7 after 82 dpa. The damage structure is a complicated mixture of small dislocation loops and dislocation segments. No attempt was made in this experiment to quantify either the nature or numbers of the various defects. Qualitatively, it can be seen that the defect density is high, which would be expected in light of the substantial yield strength increase observed for that alloy (see Fig. 2).

Once an alloy containing helium is irradiated, the primary locations where helium bubbles may be observed are the grain boundaries or particle/matrix interfaces. The bubbles in the matrix absorb vacancies and become "cavities." Micrographs of grain boundary bubbles in V-15Cr-5Ti, after helium doping and then after irradiation, are shown in Fig. 8. After the tritium trick is performed, a network of approximately 5-nm-diam bubbles can be found on the grain boundaries (Fig. 8a) and carbide particle surfaces. After neutron irradiation to 82 dpa, the networks

were somewhat more diffuse (Fig. 8b), but the bubbles were still about the same size. Therefore, it appears that the helium distribution in the grain boundaries has not changed markedly during the irradiation at 420°C.

Cavities formed in the grain matrices of a number of vanadium alloys and caused swelling. The results of the swelling measurements are presented in Table 3. Without any implanted helium, the V-15Cr-5Ti exhibited zero swelling after the irradiation to 82 dpa. Even with 300 appm ^3He (tritium trick), or 10 and 100 appm ^4He (accelerator-injected), swelling was less than 0.01%. Therefore, V-15Cr-5Ti should be considered a very low-swelling alloy. The Vanstar-7 alloy was also very low swelling, both with and without implanted helium. By comparison, both the V-3Ti-1Si and V-20Ti displayed swelling values >0.1%, but were still quite low compared to many austenitic stainless steels [9].

The swelling in V-3Ti-1Si was observed to increase with helium content, as listed in Table 3 and shown in the micrographs in Fig. 9. The micrographs were taken of specimens having 0, 23, 135 and 480 appm of implanted ^3He . It is clear that swelling increased in an orderly fashion with increasing helium level. The actual swelling values for these same four specimens were plotted as a function of helium level as shown in Fig. 10a. As more helium was added to the specimen, swelling increased above the level for zero

helium. At first, the increase is rapid but then saturated as the helium level was increased to 480 appm. Note also that a specimen implanted with ^4He had less swelling than would be expected for one with a similar level of ^3He implanted using the tritium trick. However, since the value for this one specimen is within the expected scatter and the accelerator-injected helium levels were not confirmed by analysis, one hesitates to make any distinction between swelling in ^3He - and ^4He -implanted specimens. The average cavity diameters and cavity number densities are shown as a function of helium level for the same V-3Ti-1Si specimens in Fig. 10b. In going from 0 to 480 appm ^3He , the average cavity diameter for the four specimens drops from only about 16.5 to 12 nm, but the number density of cavities increases an order of magnitude. Therefore, increasing the helium level appears to increase the nucleation of cavities in this alloy while slightly decreasing their average size.

Let us now compare the irradiated microstructures in vanadium alloy specimens that have the helium implanted by two different techniques (Fig. 11). A specimen of V-3Ti-1Si doped with 135 appm ^3He at 400°C (Fig. 11a) is contrasted with one injected with 100 appm ^4He at 420°C (Fig. 11b). Both specimens have a higher concentration of cavities near their respective grain boundaries, but the ^3He -doped specimen has substantially more cavities throughout the microstructure and, consequently, higher swelling.

DISCUSSION

Irradiation Hardening

Irradiation hardening occurs when energetic neutrons create dislocations, cavities, and/or precipitate particles in the alloy microstructure which restrict the motion of dislocations during plastic deformation. A dense network of such dislocations (Fig. 7) and, in some alloys, cavities (e.g. Fig 10) were produced by the neutron irradiation. These defects essentially doubled the yield strength of each alloy as shown in Fig. 2. The curves of yield strength versus damage level (Fig. 2) also indicate that irradiation hardening had saturated for the V-15Cr-5Ti and Vanstar-7 alloys by approximately 40 dpa. This agrees very well with the yield stress increases predicted by Loomis using microstructural data obtained in Ni^{++} ion irradiations on the same heat of V-15Cr-5Ti [10]. The V-3Ti-1Si alloy reached saturation at a damage level of about 10 dpa. The large increases in yield strengths have a strong influence on the ductilities of the alloys as shown in Fig. 4. In comparing values of total elongation along the ordinate for a zero helium level (see Fig. 3) one finds that significant decreases occur for each alloy due to irradiation hardening.

However, the presence of helium is much less detrimental, as shown by only a slight downward slope to the elongation curves with increasing helium level. Even the V-15Cr-5Ti specimens, that can be fully embrittled with 300 appm ^3He without radiation, had predominantly cleavage fracture (Fig. 5b) at that helium level with 82 dpa. This is added evidence of the dominant role played by irradiation hardening in the irradiation of vanadium alloys. Vanadium alloy development programs should favor alloys that resist irradiation embrittlement in a manner similar to Vanstar-7. In spite of considerable increases in yield strength after irradiation (Fig. 2), Vanstar-7 specimens retained reasonable ductility (Fig. 3) and exhibited ductile fracture modes (Figs. 5c and 5d). The V-3Ti-1Si alloy retained values of total elongation that were nearly as high as Vanstar-7 (Fig. 3) and also displayed some characteristic ductile features on its fracture surfaces (Figs. 5e and 5f).

Helium Embrittlement

Except for V-15Cr-5Ti, implanted helium did not appear to have a strong effect on specimen ductility, either with or without radiation (Fig. 4). The unirradiated starting material of V-15Cr-5Ti started to suffer losses in total elongation at helium levels above approximately 100 appm and was totally embrittled with 300 appm (Fig. 4a). These latter specimens failed by intergranular separation due to

helium bubbles on the grain boundaries [3]. The irradiation to 82 dpa did not appear to change the distribution of helium bubbles in the grain boundaries very much (Fig. 8), and one might expect the fracture mode to be entirely intergranular. However, this was not the case, as shown in Fig. 5b where the fracture surface was observed to be mostly cleavage. Therefore, the dominant mechanism for embrittlement, even in the helium-sensitive V-15Cr-5Ti alloy, appears to be irradiation hardening. On the other hand, while the helium bubble distributions in Fig. 8 look very similar they may, in fact, be different. The actual quantities of helium in the grain boundaries would be expected to be less than before irradiation due to the burnup of the ^3He caused by neutron irradiation. Greenwood [11] has calculated that after 100 dpa, about 30% of the ^3He will be converted back to ^3H which is free to diffuse out of the specimen. Thus, a reduction of helium in the microstructure, and, especially the grain boundaries, might reduce helium embrittlement. Some of the tritium could have migrated back into the specimen (to form ^3He again), but most would probably form bubbles and escape to the helium gas space in the subcapsule. The burnup of ^3He may also explain why the irradiated V-3Ti-1Si specimen with 480 appm ^3He exhibited a ductile fracture surface (Fig. 5f), while the aged control with the same helium content was intergranular (Fig. 6b). The burnup of ^3He is a disadvantage of using the tritium trick to preimplant

helium, but there are many advantages to the technique that justify its use. Some of the pros and cons of using the tritium trick for implanting helium in vanadium alloys are discussed in ref. [8].

Swelling

All of the vanadium alloys investigated were, in general, low swelling. At 420°C, even the higher swelling V-3Ti-1Si and V-20Ti are very competitive with advanced austenitic [9,12] or ferritic [12] stainless steels. The V-15Cr-5Ti alloy is especially low swelling (Table 3), which agrees with the high damage level Ni^{++} -ion irradiations of Loomis [10] on the same heat of material. It also agrees with the EBR-II irradiations to 30 dpa by Grossbeck and Horak [13] using a different heat of V-15Cr-5Ti. Vanstar-7 showed low swelling but exhibited moderately high swelling of 6% after irradiation in FFTF at 520°C to 40 dpa [8]. The V-3Ti-1Si alloy, which exhibited low swelling up to 40 dpa [8], now shows noticeable swelling at 82 dpa (Fig. 9). The "cavities" (voids partially filled with helium and possibly other gases) in Fig. 9 are not equilibrium bubbles because the cavities are much larger than the equilibrium helium bubbles found in the grain boundaries (see Fig. 11a). Helium in the matrices increased swelling by providing more cavity nuclei [14], which have now probably grown above their critical radius and will continue to grow by the absorption

of vacancies. The cavity size in the specimens with higher helium levels was slightly lower than those with little or no helium (Fig. 10b) because of greater competition for the existing vacancies in the high helium case (i.e., because of the higher cavity density).

Only one specimen of V-20Ti was analyzed in this experiment; it was found to exhibit about 1% swelling. Previous EBR-II irradiation experiments using this alloy observed negligible swelling at 400°C and 17 dpa [15], and about 0.1% at 440°C and 45 dpa [16].

Although many of the results for the vanadium alloys are encouraging, it should be evident that the selection of any one of the present vanadium alloys for a near-term nuclear application might result in serious trade-offs in performance. Further alloy development is needed to produce a vanadium alloy that will demonstrate adequate performance with regard to all of the important radiation effects just discussed, and at the same time maintain the compatibility, mechanical, and thermal properties required for a fusion reactor first wall.

CONCLUSIONS

(1) Irradiation hardening of the vanadium alloys saturates by 40 dpa and, at 82 dpa, is still the dominant factor controlling their postirradiation tensile properties.

(2) Of the alloys tested, Vanstar-7 had the best resistance to irradiation hardening followed, respectively, by V-3Ti-1Si and V-15Cr-5Ti.

(3) Preimplanted ^3He formed bubbles on the grain boundaries in vanadium alloys and caused embrittlement. However, the effect of helium on the postirradiation ductility was considerably less than that produced by irradiation hardening.

(4) The V-15Cr-5Ti alloy was quite sensitive to helium implanted by the tritium trick and was totally embrittled by 300 appm; Vanstar-7 and V-3Ti-1Si were only slightly affected.

(5) The V-15Cr-5Ti and Vanstar-7 alloys exhibited negligible swelling after irradiation at 420°C to 82 dpa, and low swelling (<1%) was measured in V-3Ti-1Si and V-20Ti. Preimplanted helium increased swelling in V-3Ti-1Si by increasing cavity nucleation.

(6) More alloy development is needed to produce a

vanadium alloy that will simultaneously resist radiation hardening, helium embrittlement, and swelling.

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Figure Captions

- Fig. 1 - SS-3 Miniature tensile specimen
- Fig. 2 - Yield strength of vanadium alloys after irradiation in FFTF at 420°C as a function of damage level. The alloys do not contain implanted helium.
- Fig. 3 - Total elongation of vanadium alloys after irradiation in FFTF at 420°C as a function of damage level. The alloys do not contain implanted helium.
- Fig. 4 - The effect of preimplanted helium on the total elongation of starting material, unirradiated thermally aged controls, and irradiated specimens: (a) V-15Cr-5Ti, (b) Vanstar-7, and (c) V-3Ti-1Si.
- Fig. 5 - SEM fractographs of irradiated vanadium alloys: (a) V-15Cr-5Ti, 0 appm ^3He , (b) V-15Cr-5Ti, 300 appm ^3He , (c) Vanstar-7, 0 appm ^3He , (d) Vanstar-7, 150 appm ^3He , (e) V-3Ti-1Si, 0 appm ^3He , (f) V-3Ti-1Si, 480 appm ^3He .
- Fig. 6 - SEM fractographs of thermally aged controls: (a) V-3Ti-1Si, 0 appm ^3He , (b) V-3Ti-1Si, 480 appm ^3He .
- Fig. 7 - Irradiation-produced defects in Vanstar-7 after 82 dpa at 420°C. The specimen contained no preimplanted helium.
- Fig. 8 - Grain boundary helium bubbles in V-15Cr-5Ti (300 appm level): (a) Before irradiation, and (b) after irradiation to 82 dpa in FFTF.
- Fig. 9 - Matrix cavities in V-3Ti-1Si with (a) 0, (b) 23, (c) 135, and (d) 480 appm of preimplanted ^3He , after irradiation in FFTF at 420°C to 82 dpa.
- Fig. 10 - (a) Cavity swelling, and (b) average cavity diameter and cavity number density, in V-3Ti-1Si irradiated to 82 dpa in FFTF, as a function of the preimplanted helium level.
- Fig. 11 - Microstructure of V-3Ti-1Si with (a) 135 appm ^3He , and (b) 100 appm accelerator-injected ^4He , after irradiation in FFTF at 420°C to 82 dpa.

Table 1. Vanadium alloy data

Alloy	Heat	Composition, wt %								Final Heat Treatment
		Cr	Ti	Fe	Zr	Si	C	O	N	
V-15Cr-5Ti ^a	CAM-834-3	14.5	6.2				0.032	0.031	0.046	1 h @ 1200°
VANSTAR-7 ^a	CAM-837-7	9.7		3.4	1.3		0.064	0.028	0.052	1 h @ 1350°
V-3Ti-1Si ^b	11153		3.4	0.04		1.28	0.045	0.091	0.026	1 h @ 1050°
V-20Ti ^a	CAM-832		20.3				0.020	0.039	0.044	1 h @ 1100°

^aSource: Westinghouse Electric Corporation.

^bSource: KFK, Karlsruhe, West Germany (Dr. D. Kaletta).

Table 2. Vanadium Alloy Tensile Data — Unirradiated Controls and Specimens Irradiated to 90 dpa

Unirradiated Controls (Aged 420°C for 730 days)						Irradiated (90 dpa — 552 EFPD ^a)					
Specimen No.	³ He Level (appm)	Strength, MPa		Elongation, %		Specimen No.	³ He Level (appm)	Strength, MPa		Elongation, %	
		Yield	Ulti- mate	Uniform	Total			Yield	Ulti- mate	Uniform	Total
<u>V-15Cr-5Ti</u>						<u>V-15Cr-5Ti</u>					
RA155	0	417	541	14.1	24.6	RA109	0		725		
RA51	0	397	556	12.4	20.3	RA179	0		721		
RA70	0	379	550	13.5	22.3	RA80	14		754		
RA142	80	375	539	12.6	21.7	RA52	80		577		
RA173	80	397	556	12.4	20.3	RA168	300		465		
RA167	300		359			RA32	300		465		
RA12	300	373	401		0.5	<u>VANSTAR-7</u>					
<u>VANSTAR-7</u>						QA72	0	839	849	0.5	6.2
QA99	0	302	459	10.6	22.5	QA35	42	844	844	0.3	3.3
QA72	0	311	465	12.3	23.8	QA13	42	867	881	0.6	5.2
QA117	0	271	434	10.3	19.6	QA46	150	865	890	0.3	0.6
QA05	42	312	447	13.8	22.2	QA51	150	894	897	0.5	3.3
QA42	150	330	471	13.3	22.5	<u>V-3Ti-1Si</u>					
<u>V-3Ti-1Si</u>						RB25	0	790	874	3.2	4.6
RB36	0	313	461	8.3	12.6	RB58	135	968	1023	1.5	2.0
RB60	135	456	554	9.2	14.6	RB02	135	823	904	3.1	5.6
RB49	480	464	597	8.6	10.6	RB18	480	963	1022	2.0	2.2

^aEffective full power days.

Table 3. Cavity Swelling of Vanadium Alloys Irradiated
in FFTF to 82 dpa at 420°C

Alloy	Specimen No.	Level, appm		Swelling (%) $\frac{\Delta V}{V - \Delta V}$
		^3He	^4He	
V-15Cr-5Ti	RA49	0		0
V-15Cr-5Ti	RA31	300		<0.01
V-15Cr-5Ti	A1		10	<0.01
V-15Cr-5Ti	A7		100	<0.01
VANSTAR-7	QA99	0		<0.01
VANSTAR-7	QA70	150		<0.01
V-3Ti-1Si	H4		100	0.332
V-3Ti-1Si	80	0		0.257
V-3Ti-1Si	61	23		0.568
V-3Ti-1Si	46	135		0.768
V-3Ti-1Si	24	480		1.130
V-20Ti	WADO	0		0.955

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$W_1 = 1.52$ mm
 $W_2 = 0.025$ TO 0.038 mm
GREATER THAN W_1
DIMENSIONS IN MILLIMETERS

Fig. 1 - SS-3 Miniature tensile specimen

Fig. 2 - Yield strength of vanadium alloys after irradiation in FFTF at 420°C as a function of damage level. The alloys do not contain implanted helium.

Fig. 3 - Total elongation of vanadium alloys after irradiation in FFTF at 420°C as a function of damage level. The alloys do not contain implanted helium.

Fig. 4 - The effect of preimplanted helium on the total elongation of starting material, unirradiated thermally aged controls, and irradiated specimens: (a) V-15Cr-5Ti, (b) Vanstar-7, and (c) V-3Ti-1Si.

(a)

(b)

(c)

(d)

(e)

(f)

10 μ m

Fig. 5 - SEM fractographs of irradiated vanadium alloys:
 (a) V-15Cr-5Ti, 0 appm ^3He , (b) V-15Cr-5Ti, 300
 appm ^3He , (c) Vanstar-7, 0 appm ^3He , (d) Vanstar-7,
 150 appm ^3He , (e) V-3Ti-1Si, 0 appm ^3He ,
 (f) V-3Ti-1Si, 480 appm ^3He .

(a)

(b)

10 μm

Fig. 6 - SEM fractographs of thermally aged controls:
(a) V-3Ti-1Si, 0 appm ^3He , (b) V-3Ti-1Si, 480
appm ^3He .

Fig. 7 - Irradiation-produced defects in Vanstar-7 after 82 dpa at 420°C. The specimen contained no pre-implanted helium.

(a)

(b)

Fig. 8 - Grain boundary helium bubbles in V-15Cr-5Ti (300 appm level): (a) Before irradiation, and (b) after irradiation to 82 dpa in FFTF.

Fig. 9 - Matrix cavities in V-3Ti-1Si with (a) 0, (b) 23, (c) 135, and (d) 480 appm of preimplanted ^3He , after irradiation in FFTF at 420°C to 82 dpa.

Fig. 10 - (a) Cavity swelling, and (b) average cavity diameter and cavity number density, in V-3Ti-1Si irradiated to 82 dpa in FTRF, as a function of the preimplanted helium level.

(a)

(b)

Fig. 11 - Microstructure of V-3Ti-1Si with (a) 135 appm ^3He , and (b) 100 appm accelerator-injected ^4He , after irradiation in FFTF at 420°C to 82 dpa.

(a)

(b)

Fig. 11 - Microstructure of V-3Ti-1Si with (a) 135 appm ^3He , and (b) 100 appm accelerator-injected ^4He , after irradiation in FFTF at 420°C to 82 dpa.