Creep Behavior of 8Cr2WVTa Martensitic Steel Designed for Fusion DEMO Reactor*

(An Assessment on Helium Embrittlement Resistance)

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Mechanical response against transmutational helium production, alternatively susceptibility to helium embrittlement, in a nuclear fusion reactor was examined on 8Cr2WVTa martensitic steel, a prominent structural candidate for advanced fusion systems. In order to simulate DEMO (demonstrative) reactor environments, helium was implanted into the material at 823 K with concentrations up to 1 000 appmHe utilizing an α -beam from a cyclotron. Creep rupture properties were subsequently determined at the same temperature and were compared with those of the material without helium. It has been proved that helium caused no meaningful deterioration in terms of both the creep lifetime and rupture elongation. Furthermore, failure occurred completely in a transgranular and ductile manner even after high concentration helium introduction and there was no symptom of grain boundary decohesion which very often arises in helium bearing materials. These facts would mirror preferable resistance of this steel toward helium embrittlement.

Key Words: Creep, Iron and Steel, Helium Embrittlement, Low Activation Material, Nuclear Fusion, Radiation Effect, Structural Material

1. Introduction

Material property degradation which is brought about as a consequence of high energy neutron bombardment has been commonly recognized as a matter of the greatest concern in the field of nuclear fusion technology. The problem becomes far more serious for structural materials of near-plasma components, like first wall/blanket structurals, of a prototype reactor and the beyond, in comparison with those for an experimental reactor, because much higher doses are expected in the former. At the same time, such heavy irradiation by fusion neutrons incites radioactivation of materials. The suppression of the radioactivity has been hence considered as another critical issue to solve, from the standpoint of radio-

active waste management, environmental safety and public acceptance to the utilization of fusion power. A lot of investigations have been accordingly conducted for these 15 years so as to contrive radiation-resistant structural materials with low activation characteristics and some attractive candidates are presently proposed.

Among these materials, the highest preference is placed at low activation martensitic steels, in which faster decay of induced-activity is attained by replacing usual additions of large activation (Mo, Nb) with more profitable elements (W, V, Ta), especially in European and Japanese programs ongoing, mainly due to their superior industrial maturity⁽¹⁾. Besides, these steels exhibit certain important advantages as structural materials for the fusion reactor, e.g., large thermal conductivity, small thermal expansion coefficient, low sensibility to void swelling and irradiation creep, etc. In addition, they are considered to be tolerable against helium embrittlement⁽²⁾, which is a kind of grain boundary embrittlement and is believed to be induced by grain boundary segregation of nuclear-transmutational helium generated through (n, α) reac-

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Table 1 Chemical analysis of F82H steels used in this experiment (wt. %)

Material	C	Cr	W	V	Ta	Si	Mn	P	S	Fe
F82H-IEA	0.09	7.82	1.98	0.19	0.04	0.07	0.1	0.003	0.001	bal.
F82H-PreIEA	0.10	7.46	2.1	0.18	0.03	0.09	0.07	0.003	0.003	bal.

tions between fusion neutrons and constituent atoms of materials. This hypothesis has been, however, deduced from rather fast experiments such as tensile tests and not been adequately certified with long time inspections, whereas mechanical degradation by helium is known to arise more markedly in the latter cases⁽³⁾⁻⁽⁵⁾. It is therefore necessary to examine for the further progress in fusion material development whether the favorable helium-embrittlement resistance of low activation martensitic steels quoted above will survive over long duration or not.

For this purpose, series of creep tests were carried out on a representative low activation martensitic steel containing helium. Homogeneous α -implantation at an ion accelerator was employed to simulate neutronic production of helium in the fusion reactor.

2. Materials and Methods

The materials studied were composed of two different large (5 tons) heats of F82H steel, namely IEA version⁽⁶⁾ and pre-IEA version⁽⁷⁾ (hereafter they will be denoted as F82H-IEA and F82H-PreIEA, respectively). The former was originally prepared for an international collaboration organized by IEA (International Energy Agency). The chemical compositions of each heats are given in Table 1. The both materials were provided by NKK Corporation as bulk stocks in normalized (1 313 K, 2.4 ks) and tempered (1 023 K, 3.6 ks for F82H-IEA and 1 013 K, 7.2 ks for F82H-PreIEA) conditions. These treatments yielded fully martensitic structure with austenitic grain sizes of 113 μm for F82H-IEA and 24 μm for F82H-PreIEA. Creep test pieces illustrated in Fig. 1 were obtained from those stocks after spark-cutting and mechanical grinding. The shallow penetration depth of α -particles mentioned below is responsible for the unusual slight thickness of the samples.

Instead of transmutational helium generation in the fusion reactor, helium was injected into specimens in a vacuum by means of bombardment with 20 MeV α -particles from the NIMS (National Institute for Materials Science) compact cyclotron. The projected range of the beam was computed to be 82.5 μm on SRIM 96 code⁽⁸⁾ and sufficed for the specimen thickness noted above. Figure 2 presents a schematic drawing of the irradiation setup. Before reaching the specimens, α -particles were passed a rotating energy

Thickness : 70 - 80 μm

Fig. 1 Dimensions of creep specimen (in mm)

Fig. 2 Schematic illustration of helium implantation arrangement

degrader wheel consisting of Al foils with 16 different thicknesses which were mounted on the circumference of the wheel. In this way the incident energy was varied between 0 and 20 MeV, resulting in uniform helium loading all over the specimen depth. Lateral homogeneity was achieved by beam wobbling in the other two directions. Specimen temperature during irradiation was measured with two chromel-alumel thermocouples spot-welded on a dummy plate located in the middle of the irradiated area. Temperature change through beam fluctuation was compensated with an infra-red lamp heater with rapid response. The amount of injected helium was assessed on the basis of measured target current.

Such helium implantation was performed at 823 K, the upper limit temperature desired for martensitic steels in advanced fusion systems, with dose rates of $1.1 - 1.6 \times 10^{-3}$ appmHe/s, that is nearly two orders of magnitude higher than in a typical prototype reactor⁽⁹⁾. The implanted helium levels were around 100 appmHe and 300 appmHe for F82H-IEA, and 1 000 appmHe for F82H-PreIEA, respectively, the latter of

Fig. 3 Creep testing machine

which roughly corresponds to helium accumulation at the time of blanket unit replacement⁽⁹⁾. The variation of temperature was kept within 6 K and the difference of readings between right and left side sensors was typically less than 15 K during each irradiation.

Subsequent to helium charging, creep rupture tests were run on at the same temperature also in a vacuum to the time period reaching around 4 Ms, on electro-mechanically controlled machines outlined in Fig. 3. In this apparatus the applied load is monitored with a load cell and a load deviation from the nominal value is regulated through automatic motion of a pull rod on the base of feedback signals. By using these computer controlled instruments, deviations from the stated temperature and stress throughout creep tests were dominated less than ± 2 K and ± 1 MPa, respectively. The creep elongation was measured in a high precision of few μm with two linear variable transformers each coupled with a rotary encoder. To distinguish helium effects, unimplanted reference samples that were subjected to exactly parallel metallurgical histories with those of implanted ones were similarly examined. After creep rupture, the fractured faces of both implanted and unimplanted samples were observed in SEM (scanning electron microscope) to decide failure mode and more ultimately to clarify the influence of helium on it.

3. Results and Discussions

Figure 4 typifies creep strain-time readings of a helium injected specimen and an unirradiated counterpart both applied with a nearly equal stress at every

Fig. 4 Comparison of creep curves of helium free and injected specimens for F82H-IEA and F82H-PreIEA tested at 823 K

implantation level. As seen in the figure, the creep rupture time was hardly affected by helium when it was injected to 100 and 1000 appmHe, while implantation of 300 appmHe seems to have brought on a creep life extension. The same was also shown in Fig. 5, in which standard creep rupture plots are given for all specimens tested. In cases of 100 and 1000 appmHe implantations, the creep lifetimes of a helium bearing sample and a corresponding unimplanted control which crept at a parallel stress normally accorded within the usual variation of creep rupture times in thin specimens⁽¹⁰⁾, and were hence considered to be equivalent. By contrast to this, roughly five times longer rupture life was found for 300 appmHe implanted samples as compared to unimplanted companions.

The reason of this somewhat singular strengthening at 300 appmHe may be, at least partly, attributed to helium bubble microstructure in the following way. Sagüés and associates⁽¹¹⁾ pointed out that different

Fig. 5 Variations of creep rupture strength against time to rupture for helium implanted specimens and non-implanted references of F82H-IEA (upper) and F82H-PreIEA (lower) tested at 823 K

from helium bubbles at grain boundaries, which are often regarded a main cause of helium embrittlement, small helium clusters and/or bubbles in the matrix would contribute to material strengthening through a mechanism of dislocation locking. Though microstructural observations have not done yet, the existence of such bubbles could be speculated in the present experiment according to our previous study on similar steels⁽¹²⁾. The size distribution of those bubbles is expected to become coarser, in other words, larger in size and lower in number density, with increasing helium content as reported by Batfalsky and Schroeder⁽¹³⁾. The delay of creep rupture might thereby occur at an intermediate helium concentration provided that the bubble dispersion is more beneficial, since the degree of the strengthening was estimated to be determined by the product of the bubble number density and the diameter⁽¹⁴⁾.

Creep rupture data in Fig. 5 were analyzed by the method of linear regression based on a creep power law, $t_r \propto \sigma^{-n}$, and the results are summarized in Table 2. The stress exponents of implanted and unimplanted cases were quite comparable for individual helium concentrations. This fact hints that the

Table 2 Stress exponent (n) of creep power law fitting, $t_r \propto \sigma^{-n}$, for helium implanted and unimplanted F82H martensitic tested at 823 K

Material	He concentration	Non-implanted	Implanted
F82H-IEA	100 appm	17.9	19.9
	300 appm	34.4	34.6
F82H-PreIEA	1000 appm	18.6	18.1

Fig. 6 Variations of creep rupture elongation against time to rupture for helium implanted specimens and non-implanted references of F82H-IEA (upper) and F82H-PreIEA (lower) tested at 823 K

creep mechanism itself would not be altered by helium introduction.

Larger values of the rupture life for F82H-PreIEA observed in Fig. 5 by comparison with those of F82H-IEA could be related to its finer-grained microstructure.

The elongation at rupture is plotted in Fig. 6 as a function of creep rupture time. On the average, no significant difference between implanted samples and helium free controls was appreciated in each test series, as seen in Table 3. The scatterbands of data points fell within a commonly acceptable error range but were too large to identify any trend in respect of the rupture life. The rupture elongation thus suffered

Table 3 Average creep rupture elongation of helium implanted and unimplanted F82H martensitic steels tested at 823 K.

Material	He concentration	Non-implanted (%)	Implanted (%)
F82H-IEA	100 appm	4.4 ± 2.2	3.9 ± 1.3
	300 appm	3.5 ± 1.6	4.2 ± 2.0
F82H-PreIEA	1000 appm	9.1 ± 5.2	7.0 ± 2.9

$\pm$: 2 σ confidence limit

Fig. 7 SEM fractographs representing rupture modes of F82H Pre-IEA heat version creep-tested at 823 K, 235 MPa; (a), (b) helium free control ($t_r = 473.0$ ks); (c), (d) helium implanted specimen ($C_{He} = 1\,000$ appm, $t_r = 559.3$ ks)

no substantial deterioration exerted by helium, even when it was introduced up to 1 000 appmHe. The enhanced creep ductility of PreIEA heat seen in Table 3 would be again attributed to the difference in grain size.

Creep rupture data (the lifetime and rupture elongation) obtained here were appreciably smaller than those of bulk size samples⁽⁶⁾. This disagreement could be most probably explained by specimen size effects⁽¹⁵⁾, chiefly owing to early failure in the tertiary creep stage as discerned in Fig. 4. The unusually small thickness-width ratio of specimens employed in this experiment (see, Fig. 1) might be an origin of the premature fracture.

Figure 7 typifies rupture surfaces of specimens injected with 1 000 appmHe and their corresponding controls. Although there were some morphological varieties, viz., glide plane decohesion (Fig. 7(a) and (c)) and dimple fracture (Fig. 7(b) and (d)), failure modes of implanted and unimplanted materials were very similar in character, appearing the same fully transcrystalline and ductile fashion. Features of grain boundary embrittlement induced by helium were not perceived at all. This also holds for 100 and 300 appmHe implantations as shown by Fig. 7 in the

preceding paper⁽¹⁶⁾.

4. Conclusions

The effect of transmutional helium on long term mechanical behavior of a low activation martensitic steel, F82H, was assessed in creep rupture tests at 823 K after hot helium implantation ($T_{impl.} = 823$ K, $C_{He} = 100, 300$ and $1\,000$ appm). The stated material exhibited a fairly good resistance against high temperature helium embrittlement up to a helium content anticipated in the prototype nuclear fusion reactor. The principal results which support this conclusion are:

- (1) Helium introduction aroused no detrimental influence on both creep rupture time and elongation.
- (2) The fracture surface remained entirely intra-granular and ductile after helium implantation, not showing any sign of intergranular failure induced by helium.

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