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Effect of heat treatments on tensile properties of F82H steel irradiated by neutrons

E. Wakai [a,*], M. Ando [a], T. Sawai [a], H. Tanigawa [a], T. Taguchi [a], R.E. Stoller [b], T. Yamamoto [a], Y. Kato [a], F. Takada [a]

[a] *Japan Atomic Energy Agency, 2-4 Shirakata, Tokai-mura, Ibaraki-ken 319-1195, Japan*
[b] *Oak Ridge National Laboratory, Oak Ridge, TN, USA*

Abstract

The dependence of irradiation hardening of F82H steel on tempering conditions was examined by JMTR and HFIR irradiation experiments. The JMTR and HFIR irradiations were performed to 2 dpa at 250 °C and 5 dpa at 300 °C, respectively. Irradiation hardening measured at 25–400 °C depended strongly on tempering conditions, and it basically increased with temperature and time of tempering. The increment of yield stress due to irradiation was negligible at 500 °C in the specimens tempered at 750 °C for 0.5–10 h, but was measured to be 130–200 MPa in the specimens tempered at 780 and 800 °C for 0.5 h. Irradiation hardening of F82H that was heat treated to produce a weld-like material was also examine. The weld-like material was annealed at 800, 860 and 920 °C for 0.5 h, and then continuously annealed at 700 °C. The level of irradiation hardening depended on the heat treatment conditions. From these results, it is suggested that the control of heat treatments such as tempering is useful to resist irradiation hardening and embrittlement.

1. Introduction

Reduced-activation ferritic/martensitic steels are candidate materials for the blanket structure of fusion reactors. Irradiation hardening of 8–9%Cr martensitic steels irradiated by neutrons occurs mainly at irradiation temperatures lower than about 400 °C, and it tends to increase with decreasing irradiation temperature down to around 250 °C. The shift of DBTT also increases with decreasing irradiation temperature, and the shift increase is large for irradiation at 250 °C. Several researchers [1–7] reported that the increase of yield strength and the shift of DBTT were different in Fe–9Cr alloy and several martensitic steels such as F82H, JLF-1, JLF-1B, ORNL 9Cr-2WVTa, OPTIFER Ia, II, MANET II and Mod.9Cr-1Mo, which had different concentrations of some elements and were tempered at different temperatures. The effects of the normalizing and tempering on tensile and impact behavior in martensitic steels before irradiation were reported by Schafer [8] and Gondi et al. [9]. The mechanisms controlling the relationship between the changes of yield strength and shift of DBTT due to irradiation in these martensitic steels are not clear, and it is necessary to determine the effects of heat treatment on irradiation hardening and embrittlement [10–12]. The primary purpose of this study is to examine

* Corresponding author. Tel.: +81 29 282 6563; fax: +81 29 282 5922.
E-mail address: wakai.eiichi@jaea.go.jp (E. Wakai).

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the dependence of irradiation hardening of F82H steel on tempering conditions as a function of test temperature.

2. Experimental

The chemical composition of IEA-F82H steel used in this study is shown in Table 1, and the IEA-F82H steel was normalized at 1040 °C for 38 min and tempered at 750 °C for 1 h. For the JMTR experiment, a second heat treatment was performed on the F82H steel, which was normalized at 1040 °C for 0.5 h and tempered at temperatures of 750, 780 and 800 °C for 0.5 h. The tempering time at 750 °C was varied between 0.5 and 10 h. SS-3 tensile specimens were prepared from the normalized and tempered F82H steel. Chemical extraction of precipitate residues was performed on some specimens using filters of 1 μm, 0.2 μm and 0.05 μm pore size before irradiation. XRD analyses were performed on the precipitate residue samples.

The SS-3 sheet tensile specimens were 0.76 mm thick with a gage length of 7.6 mm. Irradiation was carried out in the Japan Materials Test Reactor (JMTR) to a displacement damage value of 1.9 dpa, nominally at 250 °C. After the JMTR irradiation, tensile testing was carried out at 25, 250, 400 and 500 °C at a strain rate of 4.4×10^{-4}/s. For the high flux isotope reactor (HFIR) irradiation, additional heat treatments of IEA-F82H steel were performed at 800, 860 and 920 °C for 0.5 h, and then continuously annealed at 700 °C for 10 h. Material in these conditions are denoted as Mod 1-A, Mod 1-B and Mod 1-C, respectively. The HFIR experiments were performed in the target region using a so-called Rabbit capsule to a displacement damage of 5 dpa at 300 °C. The tensile tests after the HFIR irradiation were performed at 25 °C at strain rates of 1.1×10^{-3}/s and 1.1×10^{-2}/s.

3. Results

Fig. 1(a) and (b) show stress–strain curves of F82H steel specimens irradiated at 250 °C to about 2 dpa in JMTR. The specimens were tempered at 750 °C for 1 h or 780 °C for 0.5 h before irradiation. Yield stress (YS) decreased with increasing test temperature, but the change of YS between 250 °C and 400 °C was relatively small. The yield stress of the specimen tempered at 780 °C after the irradiation has a larger value than that tempered at 750 °C for 1 h. Fig. 2(a)–(d) show fracture surfaces of the irradiated specimens tempered at 750 °C for 1 h and 780 °C for 0.5 h after tensile tests performed at 400 °C and 500 °C. Small dimples on the fracture surfaces were observed in the specimens tempered at 750 °C for 1 h as shown in Fig. 2(a) and (b), and

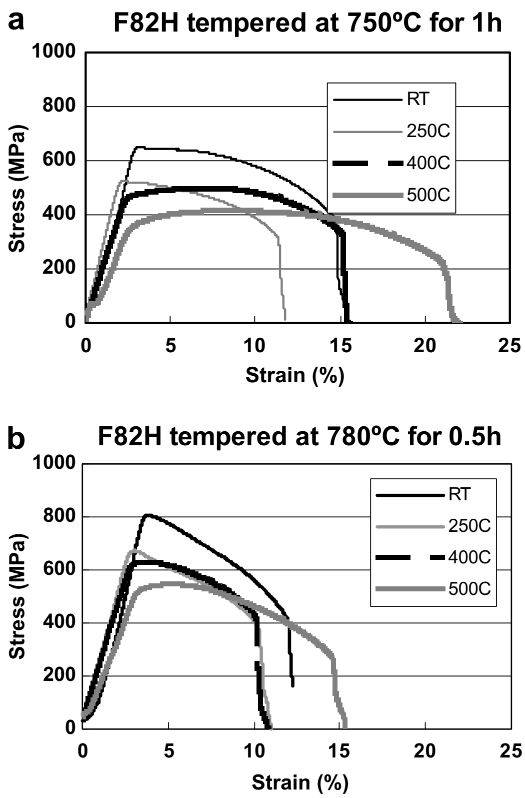

Fig. 1. Stress–strain curves of F82H steel irradiated at 250 °C to about 2 dpa in JMTR. The specimens were tempered at 750 °C for 1 h (a) and 780 °C for 0.5 h (b). The tensile data measured at 25 and 250 °C were published previously [10,12].

Table 1
Chemical compositions of IEA-heat F82H steel used in this study (wt%)

Element	C	Sol. Al	Si	Mn	P	Si	V	Ti	Cr	Ni
(%)	0.09	0.001	0.07	0.1	0.003	0.001	0.19	0.004	7.82	0.02

Element	Cu	Nb	Ta	W	B	O	N
(%)	0.01	0.002	0.04	1.98	0.0002	–	0.007

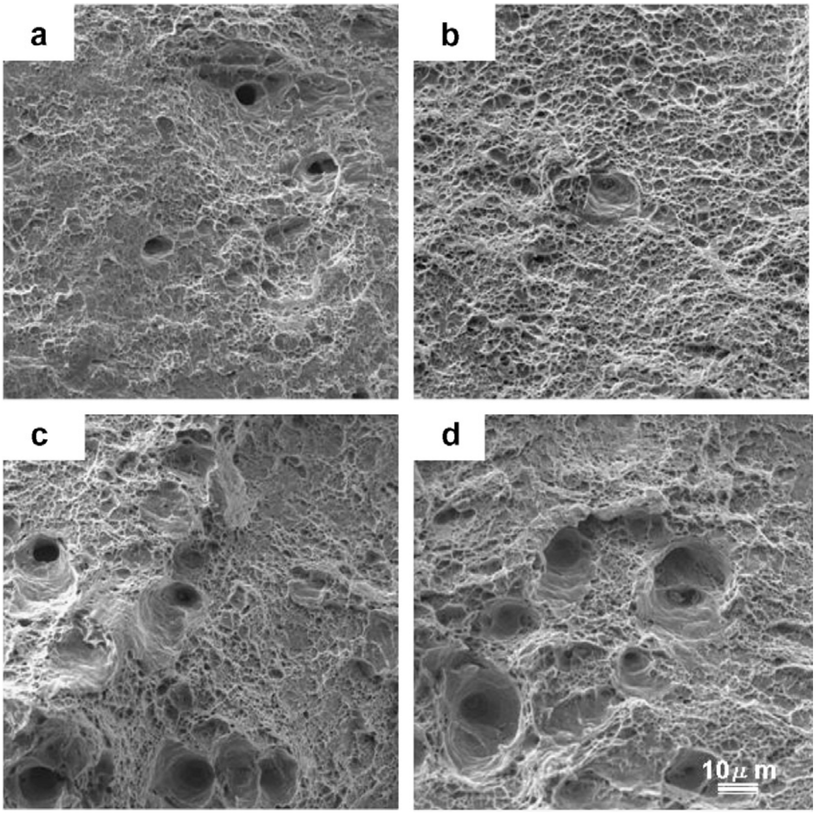

Fig. 2. Fracture surfaces after tensile tests performed at (a) 400 °C and (b) 500 °C in F82H tempered at 750 °C for 1 h and (c) 400 °C and (d) 500 °C in F82H tempered at 780 °C for 0.5 h.

larger ones were formed in the specimens tempered at 780 °C for 0.5 h as given in Fig. 2(c) and (d). The number of dimples with larger size decreased with increasing test temperature. Fig. 3 shows the mass of extracted precipitate residue for the unirradiated specimens which were tempered at 750 °C for 1 h and 780 °C for 0.5 h. The mass residue obtained by the 1 μm filter is interpreted as the value corresponding to the mass of large precipitate. The result indicates that the specimen tempered at 780 °C for 0.5 h has larger precipitates than that tempered at 750 °C for 1 h. The total amount of precipitates formed in the specimen tempered at 780 °C for 0.5 h is larger than that at 750 °C for 1 h. From these results, it may be suggested that the larger dimples as given in Fig. 2(c) and (d) are caused by the formation of the larger precipitates. The XRD analyses on the residue revealed that the majority of precipitates were $M_{23}C_6$ carbides for the unirradiated specimens.

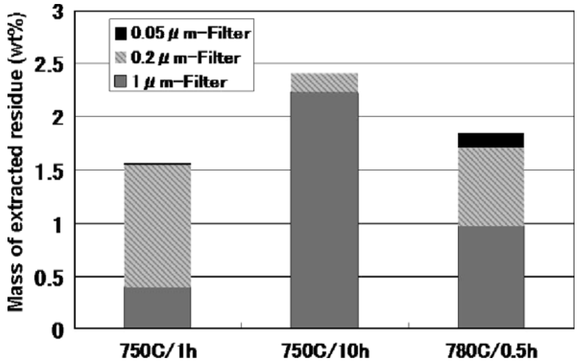

Fig. 3. Mass of extracted precipitate residue extracted with 1 μm filter, 0.2 μm filter and 0.05 μm filter for the unirradiated F82H steel tempered at 750 °C for 1 h (a), 750 °C for 1 h (b) and 780 °C for 0.5 h (c).

Fig. 4(a) and (b) show the YS of F82H tested at 25 and 500 °C before and after the irradiation. The irradiation hardening obtained in the test at 25 °C for the specimens tempered at 780 °C was larger than those tempered at 750 °C. An increase of YS due to irradiation in the tests at 500 °C was detected in F82H steel tempered at 780 °C for 0.5 h, but not in the specimen tempered at 750 °C for 1 h.

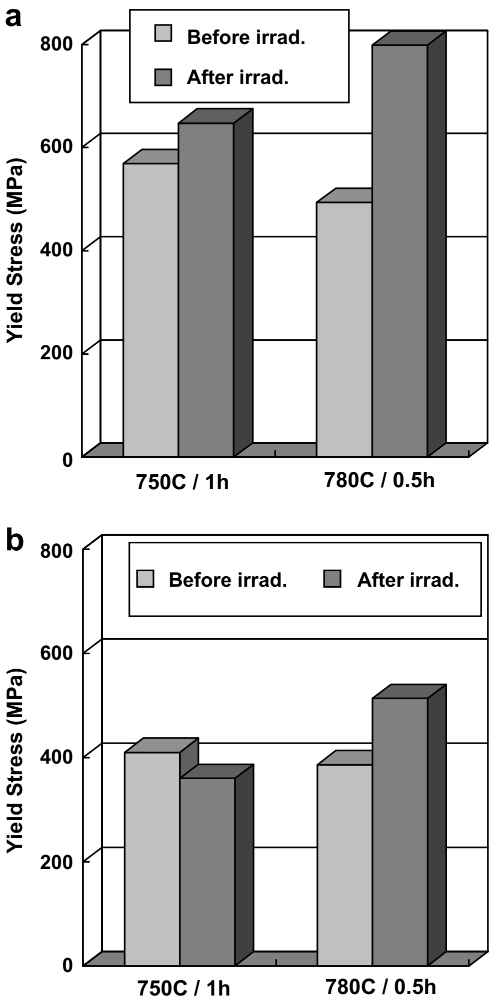

Fig. 4. Yield stress before and after the irradiation of the F82H steel tempered at 750 °C and 780 °C and tested at 25 °C (a) and 500 °C (b).

Figs. 5 and 6 show YS and elongation, respectively, measured at 25, 250, 400 and 500 °C in F82H steel tempered at several different conditions. The irradiation hardening measured at 25 °C in the specimens tempered at 750 °C for 0.5–10 h was about 100–240 MPa, and that tempered at 780 °C and 800 °C for 0.5 h was about 300 MPa. In the tests at 500 °C, there was essentially no irradiation hardening of the specimens tempered at 750 °C, except for the specimens tempered for 10 h, but irradiation hardening of the specimens tempered at 780 and 800 °C was about 130–200 MPa. The irradiation hardening in the test at 400 °C was observed in all specimens, except for the specimens tempered at 750 °C for 0.5 and 1 h, and it was detected in all specimens in case of the tests performed at 25 and 250 °C. Irradiation hardening tended to increase with increasing time and temperature of tempering. The dependence of elongation on tempering conditions increased with increasing test temperature as shown in Fig. 6. Uniform elongation of the specimens measured at 25 °C and 500 °C varied in a range of about 0.3–0.4% and 2–5%, respectively. The total elongation of the specimens measured at 25 °C and 500 °C were in the range of about 10–13% and 12–20%, respectively.

4. Discussion

4.1. Effect of tempering condition on irradiation hardening

Irradiation hardening can be evaluated by Orowan's theory for a thermal bowing of dislocations around obstacles on a slip plane [13,14] as described follows:

$$\Delta\sigma = M\alpha\mu b(Nd)^{1/2}, \tag{1}$$

where M, α, μ, b, N and d are factors converting critical resolved shear stress in a crystal to the uniaxial

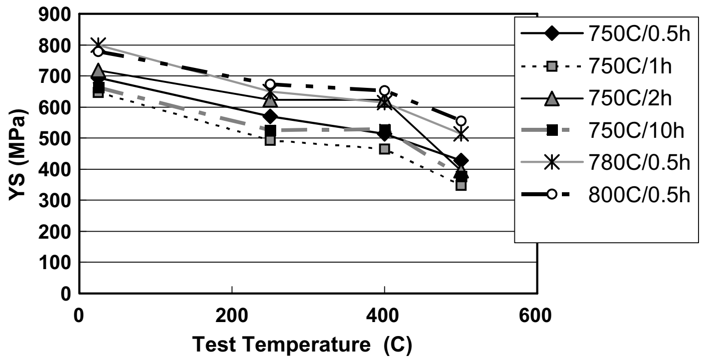

Fig. 5. Yield stress measured at 25, 250, 400 and 500 °C in the irradiated F82H steel tempered at different conditions.

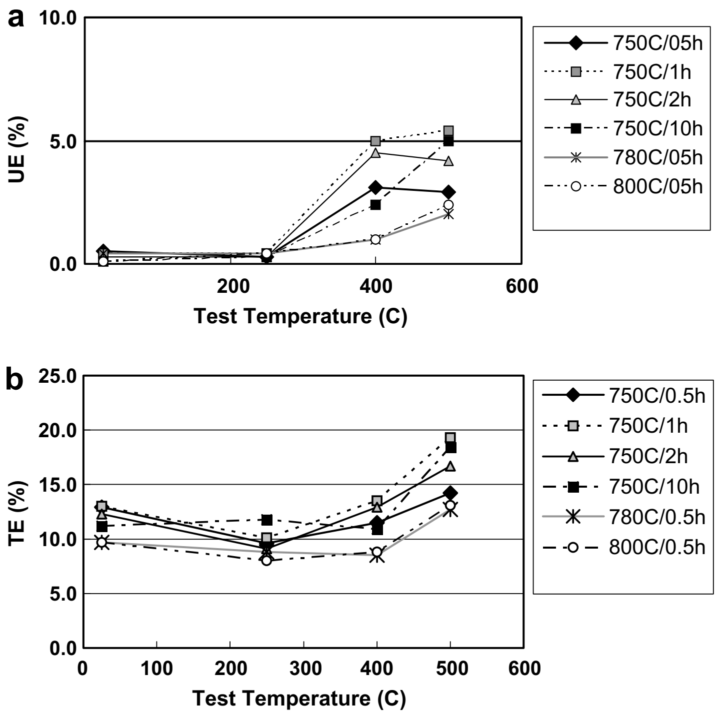

Fig. 6. Uniform elongation (a) and total elongation (b) measured at 25, 250, 400 and 500 °C in the irradiated F82H steel tempered at different conditions.

yield stress in a random crystal, the barrier strength of obstacle, the shear modulus of matrix, the Burger's vector of moving dislocation, the number density of obstacle, respectively. From Eq. (1), it can be determined that irradiation hardening is primarily dependent on the number density and size of defect clusters.

The present results indicate that irradiation hardening and elongation behavior of RAF's are very sensitive to tempering conditions. The matrix concentration of solute atoms such as carbon, and the number densities and the size of carbides would strongly depend on tempering conditions. In the specimens tempered at higher temperature and for longer times, the amount of larger carbides increases as shown in Fig. 3, and carbon concentration in the matrix will decrease. In this situation, the mobility of interstitial atoms may increase and the growth rate of dislocation loops will increase. Therefore, the irradiation hardening may be enhanced by the increase of cluster size, depending on tempering conditions. Similar results for the relation between precipitates and irradiation hardening were reported in other studies of F82H and ORNL-9Cr steels [15–17].

4.2. Effect of heat treatments related to welding joints on irradiation hardening

Welding is an indispensable procedure for the fabrication of fusion reactor core components, and the microstructures and hardness of both the weld metal and the heat-affected zone (HAZ) are different from that of the base metal [18]. F82H steel is quench-hardenable so that the cooling of weld metal and surrounding HAZ from the austenite region easily makes a fully martensitic structure, which is harder than the base metal. In the rest of the HAZ where the temperature was below A_{c3} during welding, the welding heat just anneals the existing martensite. In order to understand the effect of the welding heat on irradiation hardening of weld metal and HAZ, tensile properties of the specimens annealed at temperatures of 800, 860 and 920 °C for 0.5 h and then continuously annealed at 700 °C were examined. The results are shown in Fig. 7, which includes the YS and ΔYS for four F82H variants tested at 25 °C, before and after the HFIR irradiation at 300 °C to about 5 dpa. The Mod 1-A, -B and -C F82H variants had a larger ΔYS than that of IEA standard heat-treated F82H

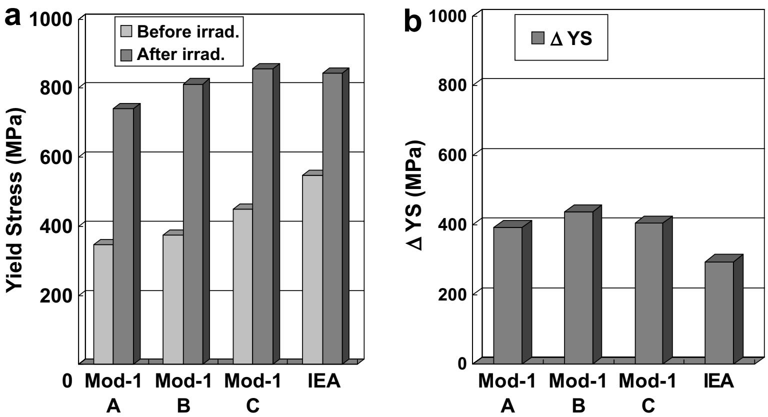

Fig. 7. (a)Yield stress and (b) ΔYS of F82H tested at 25 °C, before and after the HFIR irradiation at 300 °C to about 5 dpa.

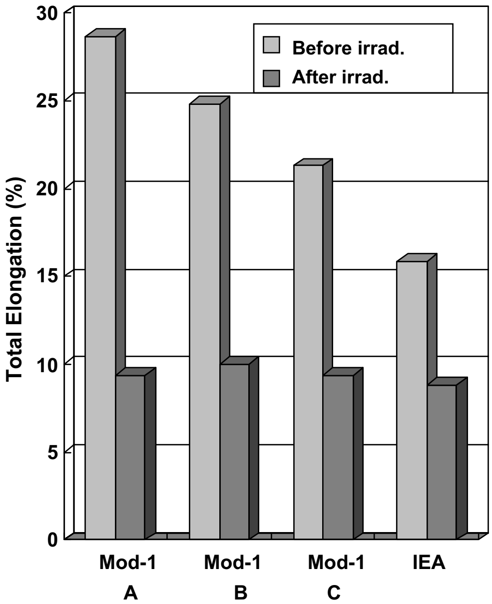

Fig. 8. Total elongation of F82H tested at 25 °C, before and after the HFIR irradiation at 300 °C to about 5 dpa.

steel, and the elongation of the Mod-1 groups after irradiation was slightly larger than that of IEA F82H steel, Fig. 8. These results also show that irradiation hardening depends on the initial heat treatment and microstructure of martensitic steels, and that control of the initial microstructure due to heat treatments is also important for resisting irradiation hardening.

5. Summary

The optimum heat treatment of reduced-activation ferritic steels (RAF's) is required to improve resistance to irradiation hardening and embrittlement, and effects of tempering conditions on tensile properties have been examined in F82H steel irradiated by neutrons. Irradiations were performed at 250 °C in the JMTR to 2 dpa and at 300 °C in the HFIR to 5 dpa. After irradiation, tensile tests were performed at temperatures from 25 to 500 °C.

In the JMTR experiments, irradiation hardening measured at 25 °C in the specimens tempered at 750 °C for 0.5–10 h was about 100–240 MPa, and that tempered at 780 and 800 °C for 0.5 h was about 300 MPa. The tempering condition for the minimum irradiation hardening was 750 °C for 1 h. In the tests at 500 °C, the irradiation hardening of the specimens tempered at 750 °C was almost lost, but that of the specimens tempered at 780 and 800 °C was about 130–200 MPa. These results may indicate that defect clusters formed in the specimens tempered at higher temperatures are relatively stable even at 500 °C.

In the HFIR experiments, the tensile properties of the weld-like heat-treated F82H steel which was tempered at a lower temperature of 700 °C for 10 h after the heat treatments at 800, 860 and 920 °C for 0.5 h were also examined. Irradiation hardening of these weld-like variants was about 400 MPa in three specimens measured at 25 °C, which was larger than that of IEA F82H given the standard heat treatment. However, the elongation of these specimens was slightly better than that of IEA F82H-std.

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Further reading

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