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Recent progress of R&D activities on reduced activation ferritic/martensitic steels

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ABSTRACT

Several types of reduced activation ferritic/martensitic (RAFM) steel have been developed over the past 30 years in China, Europe, India, Japan, Russia and the USA for application in ITER test blanket modules (TBMs) and future fusion DEMO and power reactors. The progress has been particularly important during the past few years with evaluation of mechanical properties of these steels before and after irradiation and in contact with different cooling media. This paper presents recent RAFM steel results obtained in ITER partner countries in relation to different TBM and DEMO options.

1. Introduction

Reduced activation ferritic/martensitic (RAFM) steel is considered the reference structural material for future fusion power reactors, due to its technological maturity, i.e. advanced fabrication routes, welding technology and general industrial experience [1,2]. It is also chosen as the main structural material for most ITER test blanket modules (TBMs), which are under development by ITER participants. Hence, RAFM steels are being widely developed in Europe, Japan, Russia, China, the USA and India. Specifically, F82H is being developed by the Japan Atomic Energy Agency (JAEA), JLF-1 by Japanese universities and the National Institute for Fusion Science (NIFS), Eurofer97 is being developed in Europe within the framework of the European Fusion Development Agreement (EFDA) program and Fusion for Energy (F4E) program, China low activation martensitic (CLAM) steel is being developed by the Institute of Nuclear Energy Safety Technology (INEST), Chinese Academy of Sciences (CAS) and 9Cr-2WVTa is being developed by Oak Ridge National Laboratory (ORNL). The specific chemical compositions of the RAFM steels are listed in Table 1 [1–4]. Recent progress has mainly focused on fabrication techniques, mechanical properties, manufacturing of TBMs, the effects of neutron irradiation, compatibility experiments and development of coatings. These are presented in detail in the following sections.

2. Recent progress

2.1. Fabrication of RAFMs

Development of fabrication techniques has mainly focused on purification to reduce neutron-induced activation and on large-scale smelting. The impurities are difficult to control to low levels; they mainly depend on the raw materials and smelting process. The effect of fabrication processes on microstructure and mechanical properties is also being studied.

2.1.1. Production of RAFMs

In 2007, a 5-ton heat of F82H was produced by JAEA using vacuum induction melting (VIM) followed by electro-slag remelting (ESR) [5]. An 11-ton heat of Eurofer97-3 was produced in 2009 fol-

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Table 1
Specific chemical compositions of the RAFM steels (wt%).

Element	CLAM	Eurofer97	F82H	JLF-1	9Cr-2WVTa
Fe	Bal.	Bal.	Bal.	Bal.	Bal.
Cr	8.8–9.2	8.5–9.5	7.5–8.5	9.0	8.5–9.0
W	1.3–1.7	1.0–1.2	1.8–2.2	2.0	2.0
V	0.15–0.25	0.15–0.25	0.15–0.25	0.19	0.25
Ta	0.10–0.20	0.10–0.14	0.01–0.06	0.07	0.07
Mn	0.35–0.55	0.20–0.60	0.05–0.20	0.45	0.45
C	0.080–0.12	0.09–0.12	0.08–0.12	0.1	0.1

lowing two other industrial heats, which was melted in a vacuum induction furnace and remelted in a vacuum arc device [6].

In China, a 1.2-ton heat of CLAM steel (HEAT 0912) was prepared by INEST in collaboration with other institutes and companies in China in 2009 with different product forms (bars, plates and tubes) for property testing and studies on fabrication techniques for the TBM [4]. Following on from the combined experience of smelting many small heats- a 4.5-ton batch of CLAM steel (HEAT 1105) was produced with a vacuum induction furnace and vacuum consumable electrode arc furnace in 2011. This was then hot-forged at 1423 K and rolled into different product forms for the fabrication of TBM.

2.1.2. Effect of smelting processes on precipitation behavior

The effect of ESR on the precipitation behavior has been studied [7]. Besides MX particles within the matrix, there were also the rod M_7C_3 carbides and spherical $M_{23}C_6$ carbides existing along lath and grain boundaries in steel melted by VIM. M_7C_3 was precipitated in zones with low W concentration due to macro-segregation of W in the tempered steel, while the rod M_7C_3 carbides were not detected in the steel produced by VIM + ESR due to the improved macro-segregation of W.

2.2. Mechanical properties and microstructure

2.2.1. Precipitation behaviors

Extraction residue analysis was conducted on F82H-BA07 heat, F82H-IEA heat, JLF-1 and CLAM [8]. $M_{23}C_6$ precipitates were coarsened in F82H-BA07 compared with the other steels in the as-normalized and tempered (as-NT) condition, because of the additional normalizing heat treatment. TaC precipitate formation was enhanced in JLF-1 and CLAM compared with F82H-BA07 and F82H-IEA in the as-NT condition, due to the higher Ta content. Laves phase was detected in F82H-IEA after aging above 823 K, where solid-solution W was significantly decreased. However, the amount of solid-solution W is not expected to change after 10,000 h at 873 K and after 1000 h at 923 K. F82H-IEA exhibited hardening after aging at both 673 K and 773 K for 100,000 h, whereas softening occurred at 873 K and 923 K. This behavior is similar to those of JLF-1 and CLAM. Hardening at lower temperatures could be explained by precipitation hardening from TaC, while softening was partly due to loss of solid solution hardening by W.

Precipitation behavior in F82H during blanket fabrication heat treatments, which simulated homogenizing, hot isostatic pressing (HIP) and temperature fluctuations of actual normalizing, was studied by Sakasegawa et al. [9]. Fine particles such as tantalum and vanadium carbides did not precipitate under simulated heat treatments higher than about 1263 K. They should precipitate on packet and block boundaries at lower temperatures such as the tempering temperature.

2.2.2. Fracture toughness

Conventional testing standards used to characterize the fracture resistance of metallic materials always use deeply cracked specimens in order to guarantee high crack tip constraint conditions and small-scale yielding levels. However, structural defects in real components are usually surface cracks that are generated in the course of fabrication. Therefore, predictions of fracture resistance based on standard specimens may be unduly conservative and pessimistic and also can greatly increase the operational and maintenance costs. Hence, the biparameteric fracture behavior with single-edge notched bend (SENB) specimens was assessed by Rodríguez et al. [10]. SENB specimens with shallow cracks always give much higher initiation J values and R-curves with a larger slope than standard deeply cracked specimens. Consequently, accurate failure predictions of real components can only be performed after the calculation of the constraint acting on the component and making use of the J-value characteristic to this particular constraint; otherwise pessimistic expectations would be predicted if the critical J value obtained using standard specimens is used.

2.2.3. Creep properties

Creep-rupture behavior at high temperature is one of the key issues for the application of RAFM steels in a fusion reactor. Fig. 1 shows that F82H has creep strength comparable to that of T91, which is used for high temperature components in power plants worldwide. On the other hand, Eurofer97 showed a bit lower creep strength compared with F82H at higher temperatures, which is probably due to the difference in W concentration [11].

2.2.4. Fatigue properties

In order to develop the fatigue life assessment methods based on the crack growth behavior in RAFM steels, the crack growth behavior under low-cycle fatigue in F82H was investigated by low-cycle fatigue tests at room temperature in air under a total strain in the range of 0.4–1.5% using an hourglass-type miniature fatigue specimen [12]. The relationship between the surface crack length and life fraction was described using one equation independent of the total strain range. Therefore, it might be possible for the fatigue life and residual life of F82H to be predicted using the surface crack length. The microcrack initiation life could be estimated using the total strain range if there is a one-to-one correspondence between the total strain range and the number of cycles to failure. The crack growth rate could be estimated using the total strain range and the surface crack length by introducing the concept of the normalized crack growth rate.

The effect of specimen shape on low-cycle fatigue life of miniature RAFM steels fatigue specimens was studied by Nogami et al. [13]. The differences among the four specimen types tested at relatively high strain ranges (0.8–3.0%) were nearly negligible. The fracture modes of these specimens were almost the same. The stress distribution was independent of the specimen shape above the total strain range of 0.8%, although a significant dependence of the peak stress on the specimen shape was observed under relatively low strain range conditions.

Fatigue life is generally composed of the microcrack initiation stage, crack propagation stage and final fracture stage. Therefore, evaluation of the fatigue behavior of RAFM steels is improved by evaluating not only the overall fatigue life but also the crack growth behavior in the individual stages. The low-cycle fatigue behavior of the miniature fatigue specimens of the RAFM steels was investigated by Nogami et al. [14]. Almost no difference in the microcrack initiation life due to the specimen shape or the total strain range was observed under high strain in the range above 0.8%. Almost no effect of the specimen shape on the normalized crack growth rate was observed under any of the strain range conditions in this work. The normalized crack growth rate was strongly dependent on the total strain range.