Technical Issues in Fusion Welding of Reduced Activation Ferritic/Martensitic Steels for Nuclear Fusion Reactors

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Abstract

Reduced activation ferritic/martensitic steels (RAFM), which are the modified variants of 9Cr-1Mo heat resisting steels, have been recognized as candidate structural materials for the blanket module of nuclear fusion reactors. Since the blanket is exposed to a high flux of neutrons and complex periods of fluctuating mechanical and thermal stresses, evaluation of the high-temperature mechanical properties of both the RAFM steel and welds, and understanding of their degradation behaviors, are needed to provide feedback for the development of better RAFM steels and welding processes. In this review article, first, several welding processes used to fabricate the blanket modules are introduced. Microstructural evolution during fusion welding of the RAFM steel and its subsequent degradation during creep and/or irradiation are reviewed. The thermal stability of the RAFM steel originates from solid solution strengthening, sub-grain hardening and precipitation hardening. The degradation of the RAFM steel is discussed in terms of matrix recovery associated with lath widening and reduction of dislocation density, coarsening of precipitates and the formation of Laves and Z phases. This work helps provide insight on how to mitigate the microstructural degradation, and design optimal welding technology.

Key Words : Nuclear fusion, RAFM, Welding, Microstructure, Degradation

1. Introduction

Reduced activation ferritic/martensitic steels (RAFM) have been considered as the candidate structural materials for blanket module of nuclear fusion reactors¹⁻³⁾. The RAFM steel was derived from variants of modified 9Cr-1Mo heat resisting steel. The RAFM steel involved the replacement of Mo in the modified 9Cr-1Mo steel by W (1-2 wt.%) and/or V (0.2 wt.%)^{3,4)}. Nb was substituted by Ta (0.02-0.18 wt.%). Since the blanket is exposed to high flux of neutron and complex periods of fluctuating mechanical and thermal stresses, the RAFM steels are expected to experience stress/irradiation-induced non-equilibrium phases formation as well as coarsening of precipitates because of its high contents of Cr, W, V, Ta, which can affect high-temperature properties. Accordingly, the chemical composition optimization

has been extensively studied to assure the excellent properties of RAFM steels based on the research programs in European Union, USA, Japan, China, India and Korea²⁻⁹⁾.

The blanket is an important device trapping the core plasma under extremely high heat load and neutron irradiation environment. This blanket provides multiple-functions to convert kinetic energy of neutrons generated in the fusion reaction into heat energy (heat exchange), to shield neutrons, and to multiply tritium from lithium in the coolant¹⁻³⁾. The blanket is composed of about 400 to 500 boxes to form an outer wall which is in primary contact with the plasma. The blankets should be fabricated as final parts by assembling materials that perform different functions. Therefore, welding and joining techniques of homogeneous or dissimilar materials are considered as core technologies for blanket manufacturing. The first wall in contact with

Category	Position	Welding method	Joint type	Thickness (mm)	Length (m)
Water leak accident	First wall-side wall Side wall-back wall	EB	Corner joint	45	17
Disruption	Back wall-back wall Key-back wall	EB (Not identified)	Butt joint	90 (Not identified)	1.6 1.4
Water pressure against vacuum	First wall header Side wall header Back wall header Pipes connection from TBM to back shield	TIG	Butt joint	7 to 20	19
		(Not identified)	Butt joint		
Water pressure against 0.1MPa He purge gas	Inner headers Tube connections	TIG	Corner joint	5 to 10	11
Seal against purge gas	Membrane panel and header	Fiber laser	Butt joint	1.5	65

Fig. 1 Japanese WCSB blanket system for ITER and applied welding and joining processes²⁾

the plasma is fabricated by joining protective materials such as beryllium (Be) or tungsten (W). Other parts of the blanket is fabricated by using fusion welding of RAFM steel. Fig. 1 shows the “Water-Cooled Solid-Breeding (WCSB)” blanket system considered as the Test Blanket Module (TBM) for ITER in Japan and the welding methods actually applied. In order to welding the RAFM steel, TIG (Tungsten Inert gas)¹⁰⁻¹⁷⁾, EB (Electron Beam) welding¹⁴⁻¹⁸⁾ and HIP (Hot Isostatic Pressing) bonding are most frequently performed¹⁹⁻²³⁾.

In this review paper, the microstructure evolution during welding of RAFM steels is addressed. Their high-temperature properties and irradiation resistances of RAFM welds are discussed in terms of their microstructure degradation. In particular, it is intended to provide an insight how to develop the welding technology for improving the welds performance. This review paper consists of most of the information presented in the previous column²⁴⁾.

2. Microstructure Characteristics of RAFM Steel Weld and Type IV Issue

RAFM steels are subjected to normalizing and tempering treatment to achieve the stable tempered martensitic microstructure. The microstructure shows the several elements such as prior austenite grain boundaries (PAGBs), blocks, laths, subgrains, packets, carbide pre-

cipitates, and other precipitates. The prior austenite grain boundaries (PAGBs) and lath boundaries are decorated with fine $M_{23}C_6$ in different morphologies²⁵⁾. Coarse $M_{23}C_6$ precipitates are formed along the PAGBs, and lath boundaries while fine $M_{23}C_6$ are inside the intra-lath region. Fine nano-sized MX precipitates are distributed uniformly inside the intra-lath structure. Fig. 2 shows typical microstructures of RAFM steels. The boundaries of the tempered martensitic microstructure can be delineated by EBSD enabling to determine the different boundary misorientation angles. The fraction of block boundaries and block width, which are delineated by red and green lines, can be quantified. TEM micrograph reveals a lath structure with a high density of dislocations, and decorated by small precipitates.

As shown in Fig. 3, Pandey et al.²⁵⁾ illustrated the effect of welding process on metallurgy of HAZ in P91 weldments, which are almost same as those of RAFM welds. The different microstructure of HAZ is obtained for different welding process. The microstructure evolution in HAZ is mainly influenced by the heating and cooling rate during weld thermal cycle, peak temperature experienced (T_p), dwell time, number of welding passes and finally subsequent post weld heating (PWH) and post weld heat treatment (PWHT). High heating rate during the weld thermal cycle resulted the shifting of transformation temperature, recrystallizing temper-

Fig. 2 Typical microstructures of RAFM steel: (a) SEM micrograph and corresponding boundary angle map by EBSD, (c) TEM micrograph showing lath structure³⁾

Sub-zones	Temperature experienced	Transformation reactions	Microstructure characteristics	Mechanical properties
CGHAZ	$T_m > T_p > T_{\gamma\delta}$	$\gamma + \delta \rightarrow M + \delta$	Untempered martensite, undissolved NbX, dissolved $M_{23}C_6$	Poor toughness and high strength
	$T_{\gamma\delta} > T_p > A_{c3}$	Coarse $\gamma \rightarrow M$	Untempered martensite, undissolved NbX, dissolved $M_{23}C_6$	Poor toughness and high strength
FGHAZ	$T_{\gamma\delta} > T_p > A_{c3}$	Fine $\gamma \rightarrow M$	Untempered martensite, dissolved fine $M_{23}C_6$, undissolved MX and coarse $M_{23}C_6$	Adequate toughness and strength
IC-HAZ	$A_{c3} > T_p > A_{c1}$	$\gamma \rightarrow M + OM$	Untempered martensite, tempered martensite, dissolved very fine $M_{23}C_6$, undissolved MX and coarse $M_{23}C_6$	High toughness and poor strength
Over-tempered base zone	$T_p < A_{c1}$	OM	Tempered martensite, Undissolved MX and coarse $M_{23}C_6$	High toughness and poor strength

Where, T_p =temperature experienced during welding; T_m = melting point of steel; $T_{\gamma\delta}$ =temperature at which γ to δ transformation completed during heating; A_{c3} =upper critical temperature at which α to γ transformation completed on heating; A_{c1} =lower critical temperature at which α to γ transformation started on heating; γ =austenite; δ =ferrite; M=martensite; OM=overtempered martensite.

Fig. 3 Schematic of the sub-zones of the HAZ of 9Cr steel weldments and corresponding transformation reaction²⁵⁾

ature and solution temperature on higher side. One single pass produces HAZ which can be divided based on peak temperature experienced (T_p): Coarse Grained HAZ (CGHAZ, $T_m > T_p \gg A_{c3}$), Fine Grained HAZ (FGHAZ, $T_{\gamma\delta} > T_p > A_{c3}$), InterCritical HAZ (ICHAZ or partially transformed HAZ, $A_{c3} > T_p > A_{c1}$), Over-Tempered HAZ (OTHAZ, $T_p < A_{c1}$).

In the welds of RAFM steels, FGHAZ and ICHAZ are highly susceptible to Type IV cracking. In the case of FGHAZ, the low peak temperature (just above A_{c3}) may lead transformation of tempered martensite into austenite with partial dissolution of the precipitates. The undissolved precipitates can be coarsened and limit the growth of austenite by pinning the austenite grain boundaries. Fig. 4 shows the microstructure of FGHAZ in P92 welds produced by shielded metal arc welding (SMAW) process. Shin et al.²⁶⁾ reported that PAG size within the FGHAZ was much smaller ($\sim 10 \mu\text{m}$), compared to the base metal ($\sim 25 \mu\text{m}$). In their work, few precipitates were observed at the fine PAG boundary, while a lot of large carbides were distributed at the ghost PAG boundary (Fig. 4(a)). This is caused by seg-

regation of $M_{23}C_6$ forming elements (Cr, W, and C atoms) during welding at the ghost PAG and block boundaries, which were former boundaries in the base metal. As a result, $M_{23}C_6$ precipitated at the ghost PAG and block boundaries instead of newly formed fine PAG boundaries during PWHT. Pandey et al.²⁵⁾ illustrated microstructure evolution occurred in FGHAZ during welding followed by PWHT (Fig. 4(c)).

Additionally, tendency of sub-grain formation was recognized in the FGHAZ (Fig. 4(b)). Reduced dislocation density may be also an indicative of recovery occurred in the FGHAZ during PWHT. The FGHAZ is characterized with low hardness and moderate toughness because of weak solid solution strengthening compared to CGHAZ²⁷⁾. Tsukamoto et al.²⁸⁾ addressed this observation. They investigated systematically the cause of type IV failure, and claimed that lack of precipitation strengthening at PAG and block boundaries of FGHAZ is the most dominant factor to cause type IV failure^{26,28,29)}.

In the case of ICHAZ, partial dissolution of precipitates and coarsening of precipitates both are observed simultaneously. After the cooling, ICHAZ shows a

Fig. 4 Microstructure of FGHAZ in P92 welds produced by SMAW process²⁶: (a) SEM micrograph and (b) TEM micrograph. Schematic of microstructure evolution in FGHAZ is shown in (c)²⁵

Fig. 5 Microstructures of various HAZ in RA FM steel weld produced by TIG process³⁰

complex structure of untempered lath martensite, retained over-tempered martensite, newly-formed PAGs, coarse tempered PAGs with newly formed fine precipitates and existing coarse precipitates along the new PAGs. Among them, the local area containing the retained over-tempered martensite within ICHAZ, should be subjected to one more tempering during PWHT. Hence, polygonal subgrains with lower density of dislocations are well-developed due to recovery, and more

coarsened precipitates resulted from double tempering are present in ICHAZ. This microstructure evolution makes ICHAZ have the lowest hardness. Hong et al.³⁰ investigated microstructure and hardness of RA FM steel weld produced by TIG. PWHT was conducted at 760 °C for 2hr. Fig. 5 shows HAZ microstructures of RA FM steel weld. It is clear that ICHAZ showed larger size of precipitates, wider lath width and smaller lath fraction, compared to CGHAZ or FGHAZ. These met-

Fig. 6 Microhardness distribution traversing RAJM steel weld produced by TIG process³⁰⁾

allurgical factors lead to the softest hardness in ICHAZ (Fig. 6).

RAJM steels (including 9-12Cr heat-resisting steels) is susceptible to Type IV premature failure during long-term creep. This is a challenging issue which is believed to be difficult to overcome. Cracks are defined on the basis of their location in weldments. Localized creep cavitations in soft zone lead to intergranular fracture during the long-term creep exposure at low stress. This is termed as Type IV cracking occurred near HAZ boundary (Fig. 7). The carbide coarsening and partial dissolution of precipitates in FGHAZ and ICHAZ make it relatively soft compared to CGHAZ and weld zone.

Compared to creep failure (or rupture) in weld metal (Type I rupture) or base metal, when Type IV failure occurs, the actual creep life is much shorter than expected (see significant life difference in Fig. 8). It is disadvantageous in terms of stable operation and maintenance of the fusion reactor. Many studies have been made to elucidate the cause of type IV failure: grain refinement of FGHAZ³²⁾, HAZ softening and coarsening of $M_{23}C_6$ ^{33,34)},

Fig. 8 Comparison of creep lives depending on the occurrence of Type IV failure³¹⁾

the formation of Laves phase³¹⁾ and distribution of precipitates^{28,29)} were proposed. Recently, Tsukamoto et al.²⁸⁾, Abe²⁹⁾ and Shin et al.²⁶⁾ claimed that lack of precipitation strengthening at PAG and block boundaries of FGHAZ could be the most dominant factor to cause type IV failure. As shown in Figs. 4(a) and (b), few precipitates were observed at the fine PAG boundaries of FGHAZ before creep. During creep at higher temperatures/lower stresses, creep damage would be accumulated more by diffusion process rather than by dislocation glide. Under this condition, most of strains could be localized at the grain boundaries by either strain concentration at the weak (or softest) grain boundaries or diffusion-mediated sliding of grain boundaries. This damage accumulation to grain boundaries could be the most significant at the FGHAZ which has the largest area of grain boundaries. Furthermore, this damage accumulation to the FGHAZ could be accelerated when the strengthening of grain boundaries by carbides is not facilitated²⁶⁾.

Therefore, in order to suppress the Type IV fracture of RAJM steels (including 9-12Cr heat-resisting steels), it is understood that a welding technology should be de-

Fig. 7 Microhardness traverse of 9Cr steel weld before and after creep, and Type IV cracking illustration³¹⁾

signed to reinforce carbide precipitation strengthening at grain boundaries (especially FGHAZ/ICHAZ). Several studies demonstrated that the Type IV fracture can be effectively suppressed, when PWHT includes normalizing followed by tempering^{28,29,35}. This full PWHT makes HAZ microstructures be similar to that of base metal.

3. Microstructure Degradation and Welding Technology Development

Tanigawa et al.³⁶ conducted room-temperature tensile tests of Japanese RAFM steel F82H. They investigated the influence of irradiation effect (300 °C, 5 dpa @ High Flux Isotope Reactor) on the tensile properties of base metal and TIG welded joints. When irradiated, the welded part and the base metal were hardened so that the tensile strength greatly increased and the elongation was significantly reduced. Even in the case of HAZ, the degree of hardening was low, indicating relatively low strength (Fig. 9).

As already mentioned, ICHAZ (indicated by zone 1 in Fig. 9(a)) shows the softest hardness since it contains larger size of precipitates, wider lath width and smaller lath fraction. It is normal that the ICHAZ is the tensile-fractured position since most strains are localized at the softest zone during deformation. Hence, we have to minimize the width of ICHAZ as much as possible to induce homogeneous deformation.

Fig. 11 shows the comparison of creep lives between base metal and TIG welded joint of F82H. In the creep temperature range of 500-600 °C, the welds show a creep life that is not significantly inferior to the base metal. This is because the TIG welded joints in this range were all ruptured at the base metal, and thus exhibits a creep life close to base metal. However, when the stress is lower than 100 MPa at 650 °C, it can be seen that the Type IV fracture is clear, thereby significantly reducing the creep life.

The weld HAZ of RA FM steels has a detrimental effect on the blanket stability because of the uneven distribution of hardness, the lowest hardness and the pres-

Fig. 10 Strain localization at ICHAZ/OTHAZ during tensile deformation³⁶

Fig. 11 Comparison of creep lives between base metal and TIG welded joint of F82H³⁶

ence of FGHAZ which can cause Type IV failure during creep deformation. For this reason, it is essential to develop welding techniques to minimize the width of HAZ (especially FGHAZ) as much as possible. In the case of TIG welding, research and development are underway to control the heat input to avoid the generation of δ -ferrite in CGHAZ and to realize the optimum microstructure¹⁰. In addition, the development of filler metals for TIG welding is actively in progress¹¹. The RA FM steels with tempered martensite have high hardenability so that fresh martensite is inevitably produced

Fig. 9 (a) Macro-cross section of TIG welded joint of F82H. Comparison of (a) hardness traverse and (b) tensile properties at room temperature with regard to irradiation³⁶

Fig. 12 (a) Macro-section showing the TIG weld of Chinese RAFM steel CLAM, and (b) its hardness traverse with different PWHT conditions¹²⁾

in welds immediately after welding. Thus, PWHT treatment must be performed. For this reason, PWHT optimization has been investigated to remove hardness non-uniformity in the HAZ^{12,28,29,35)}.

Research into applying a high density of energy heat source, such as laser or electron beam welding, has been actively conducted. As shown in Fig. 13, it can be seen that the ductile brittle transition temperature (DBTT) curve of the EB weld is much better than the TIG weld.

Sawai et al.¹⁵⁾, who is affiliated with the Japan Atomic Energy Research Institute (JAERI), conducted a PWHT (720 °C/ 1hr/ air cooling) for TIG and EB beam welds using F82H. When optimizing the EB welding, it was

Fig. 13 Comparison of CVN impact toughness curves with temperature¹⁰⁾

Fig. 14 Macro-section showing (a) TIG weld and (b) EB weld, and their hardness traverses are present in (c) for TIG weld and (d) for EB weld¹⁵⁾

Fig. 15 Candidate welding processes of European RAFM steel Eurofer97 for ITER test TBM³⁷⁾

confirmed that the weld width was very narrow and the sharp hardness decrease in ICHAZ/OTHAZ was hardly found, as seen in Fig. 14.

For this reason, in Europe, in addition to TIG welding, laser welding, laser-MIG hybrid welding, and EB welding are actively being applied to fabricate TBM by using European RAFM steel Eurofer97 (note Fig. 15).

4. Summary

The R&D status of welding processes suitable for RAFM steels was reviewed in terms of microstructure/mechanical degradation in order to address the key technical issues, and finally to provide an insight how to develop the welding technology for improving the welds performance. Major conclusions derived are as follows:

1) Since the blanket is exposed to high flux of neutron and complex periods of fluctuating mechanical and thermal stresses, understanding of degradation behaviors of RAFM welds are needed to provide feedback for

better development of welding processes.

2) The degradation of the RAFM steel welds originates from matrix recovery associated with lath widening and reduction of dislocation density, coarsening of precipitates and formation of Laves and Z phases.

3) In the welds of RAFM steels, FGHAZ and ICHAZ are highly susceptible to Type IV cracking. Localized creep cavitations in soft zone lead to intergranular fracture during the long-term creep exposure at low stress.

4) For this reason, it is essential to develop welding techniques to minimize the width of HAZ (especially FGHAZ) as much as possible. A high density of energy heat source, such as laser or EB welding, has been actively conducted, and will be more applied to guarantee a higher safety of blanket modules.

5) PWHT optimization has been investigated to remove hardness non-uniformity in the HAZ. Enhanced HAZ strengthening approach will be designed to obtain excellent microstructures, which is a highly resistant to Type IV failure.

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