General paper

Creep Fracture Analysis of W Strengthened High Cr Steel Weldment

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Abstract: Type IV cracking in heat affected zone (HAZ) of weldment is a problem for advanced high Cr ferritic steels. The present paper investigates the creep properties and microstructures of W strengthened P122 steel weldment at 923K. From the investigation of creep properties of simulated HAZ, it is clarified that heating up to around A_{C3} during welding minimized the grain size and creep strength. Most of the welded joint specimens were type IV fractured in fine-grained HAZ and resulted in shorter creep lives than those of the base metals. Electron beam welded joint with very narrow HAZ also showed the brittle type IV fracture due to the formation of creep voids and cracks. The growth of intergranular precipitates was faster for fine-grained HAZ. On the basis of experimental results, the FEM code that simulates type IV crack growth behavior has been developed. The vacancy diffusion under multi-axial stress condition in HAZ of weldment is analyzed.

Key words: 11Cr-0.4Mo-2W steel, Simulated HAZ, Welded joint, Type IV crack, Creep crack growth, FEM, Computational simulation

1. INTRODUCTION

In consideration of reduction of CO₂ emissions and energy saving for fossil power plants, the steam pressure and temperature conditions of boiler components is tending to increase. Several types of advanced 9-12%Cr ferritic boiler steels with high creep strength have been developed. The tempered martensitic P91 steel strengthened by Nb and V addition is being widely used for temperatures up to 873K. The 9-12%Cr steels strengthened by replacing Mo with W, P92 and P122, are being now performed for application to boiler components of ultra-supercritical (USC) power plants operating at around 898K [1]. The research and development of advanced heat resisting steels to be used at 923K is also being conducted [2].

For these advanced high Cr steels with high creep strength, the creep damages and cracks initiated in HAZ, which is called type IV crack, decrease the creep life of weldment at higher temperatures [3-6]. The type IV creep fracture in HAZ would be a limiting factor for realizing the USC boilers to be operated at higher temperatures. Little has been studied concerning creep fracture properties of advanced high Cr steel weldment and the mechanism for a drop of creep strength in weldment is not fully understood.

In the present study, creep properties and microstructures of simulated HAZ and welded joints for W strengthened P122 steel were investigated in order to clarify the mechanisms responsible for type IV fracture. Based on the experimental results, computational simulation for type IV crack growth behavior has been conducted. The vacancy diffusion under multi-axial stress condition in HAZ of weldment is analyzed.

2. EXPERIMENTAL PROCEDURE

2.1. Materials

The material investigated is 11Cr-0.4Mo-2W-CuVNb steel (ASME P122) plate of 27mm thickness. The present steel was normalized at 1323K for 100 minutes and tempered at 1043K for 6 hours. The A_{C1} and A_{C3} transformation temperatures are 1093K and 1193K respectively. The average grain size of the base metal is about 35 μ m.

2.2. Heat Treatments for Simulated HAZ

The creep deformation data for HAZ microstructures were obtained from creep tests of simulated HAZ specimens. The simulated HAZ specimens were produced by two procedures shown in Fig. 1. One is furnace heating and air-cooling and the other is rapid heating and gas cooling by weld simulator (Gleeble). The peak temperatures were varied from 1073 to 1473K. The post weld heat treatment (PWHT) condition is 1013K for 260 min. The creep tests for simulated HAZ were conducted using round-bar specimens of 6mm in diameter and 30mm in gage length at 923K. The creep tests for weld metal and base metal were also conducted using the same specimens.

2.3. Test Procedure for Welded Joint

The plates were welded using three different techniques, i.e. multi-layer gas tungsten arc welding (GTAW) with single bevel groove, GTAW with single U groove and electron beam welding (EBW). A filler wire developed indigenously [7] was used for GTA welding. PWHT condition for welded joint is the same for simulated HAZ specimen. Figure 2 shows the cross sectional

Fig.1. Simulated HAZ heat treatment.

Fig.2. Cross-sectional view of welded joints.

view of welded joints. Width of HAZ was about 2.7mm for GTAW with single bevel groove, 3.5mm for U groove and 0.5mm for EBW joints. The smooth plate specimens as shown in Fig. 3(a) were machined out from these welded joints, and subjected to creep tests at 923K.

The creep crack growth tests were conducted for welded joint, base metal and simulated HAZ specimens using CT specimen as shown in Fig. 3(b). The notch tip of welded joint CT specimen was located in HAZ. The fatigue pre-crack of 3mm and 25% side-groove in thickness was induced. Crack length was measured by D.C. electrical potential method.

3. RESULTS AND DISCUSSION

3.1. Creep Properties of Simulated HAZ

Type IV creep crack initiates in the inter-critical zone of HAZ heated from A_{C1} to A_{C3} temperature during welding. In order to clarify the relation between creep properties and HAZ microstructures and to obtain creep deformation data of HAZ, the creep tests on simulated

(a) creep test specimen

(b) CT specimen

Fig.3. Test specimens of welded joints.

Fig.4. Changes in creep rupture time of simulated HAZ specimens as a function of peak-temperature.

HAZ specimens were conducted. The relations between creep rupture times and peak-temperatures during simulated HAZ heat treatment are shown in Fig.4. The creep rupture time shows a minimum value for the specimens heated up around A_{C3} temperature. The microstructure of simulated HAZ specimen heated up to A_{C3} is characterized by fine-grained structure without lath martensite [8], and its grain size is smaller than $10\mu\text{m}$. The creep rupture time for the specimen heated around A_{C3} was about one-fifth of that for base metal at all stress levels.

The difference in creep strength of fine-grained HAZ between simulated by furnace heating and simulated by gleeble weld simulator is not significant as shown in Fig. 4, so that the effect of heating rate and cooling rate of simulated HAZ heat treatment on creep rupture time is

Table 1. Mechanical properties.

	$E(\text{GPa})$	$A(\text{MPa}\cdot\text{h}^{-n})$	n
Weld metal	106	2.45×10^{-36}	14.8
Coarse HAZ	99	4.86×10^{-22}	8.4
Fine HAZ	77	4.58×10^{-20}	7.7
Base metal	103	1.35×10^{-25}	10.0

comparatively small at 923K. The obtained mechanical properties of Young's modulus and Norton's constant for weld metal, simulated coarse-grained HAZ, simulated fine-grained HAZ and base metal at 923K are shown in Table 1.

3.2. Creep Strength and Microstructures of Welded Joints

Figure 5 shows the relations between stresses vs. creep rupture times of smooth plate specimens for welded joints and of round bar specimens for weld metal, base metal and simulated fine-grained HAZ at 923K. Creep rupture time is longer in the order of weld metal, base metal, EBW joint, GTAW joint with U-groove, GTAW with single bevel groove and simulated fine-grained HAZ. Welded specimens were all fractured in HAZ (type IV fractured) except for one EBW joint tested at 110MPa that was fractured in base metal. Further the rupture times of type IV fractured joints were lower than that of base metal.

The profiles of the type IV creep fractured specimens are shown in Fig. 6. Specimens were fractured in the fine-grained HAZ and many creep voids on grain boundaries could be observed. The GTAW joint with single bevel groove was fractured in HAZ of higher bevel angle by the shear deformation. The GTAW joint with U-groove that has a little wider HAZ deformed largely in HAZ and crack initiated inside the specimen. In EBW joint with narrow HAZ width, the very brittle type IV fracture was occurred. Creep rupture times of the EBW joints were about twice longer than that of the GTAW joints. This result confirms that decreasing the bevel angle and width of HAZ is effective to prolong the creep life [4]. However, the brittle type IV fracture would occur even in the EBW joints with very narrow HAZ for long-term services at higher temperatures.

Figure 7 shows the profile of creep crack and microstructures of EBW joints crept for 7300h at 60MPa. Creep crack initiated near the surface of joints and propagated towards the center of thickness along the fine-grained HAZ region that would be heated to A_{c3} temperature during welding. In the SEM micrographs, the fine transgranular precipitates, which would contribute to creep strength, were observed on the lath interface in the base metal. In the fine-grained HAZ, the coarse precipitates of $M_{23}C_6$ and Laves phase (Fe_2W) on grain boundaries were observed, while the fine transgranular precipitates were not observed. The growth rate of intergranular precipitates during creep is faster for fine-

Fig.5. Creep rupture times of welded joints.

Fig.6. Profiles of the creep fractured specimens of GTAW and EBW joints

grained HAZ. Further the recovery of dislocation structures is also faster in fine grains [8]. These differences of microstructures will be important in the degradation of creep strength of welded joints.

Creep voids were observed on the interface between precipitates and matrix ahead of the crack tip. The type IV crack grew accompanied by creep voids as shown in Fig. 7. Creep deformation of fine-grained HAZ with lower creep strength is mechanically constrained by weld metal and base metal with higher creep strength. Vacancy diffusion, void formation and growth is accelerated under multi-axial condition for welded joint specimens [9]. The creep crack growth is also accelerated by void formation ahead of the crack tip [10,11]. In order to evaluate the vacancy diffusion under multi-axial stress condition, we have conducted the following computational analysis.

Fig.7 Profile and microstructures of creep cracked EBW joint.

3.3. Computational Simulation for Type IV Creep Crack Growth

Based on the observation in Fig. 7, we attempted to simulate type IV creep crack growth behavior taking the diffusive growth of voids into account. At first, creep crack growth is computed by the following method. The creep crack growth could be characterized by critical strain criteria ahead of the crack tip [11,12]. The creep strain distribution of CT specimen model shown in Fig. 8 was calculated by FEM using Norton's creep constitutive equation shown in Table 1. When the equivalent creep strain ahead of the crack tip reached to the critical value, the coordinate of the crack tip node was moved according to the method proposed by Hsu et al. [13]. The critical strain value can be assumed as the creep ductility of round bar specimen for plane stress condition. The C^* line integral for the path shown in Fig. 8 was calculated for each time step.

An example of comparison for computational and experimental relations of creep crack growth rate da/dt vs. C^* parameter is shown in Fig. 9. The symbols indicate the relations between experimental da/dt vs. experimental C^* obtained using displacement rate for base metal of the present steel. The dashed line shows the relations between computational da/dt vs. C^* integral calculated by FEM. The crack growth rate could be predicted from Norton's creep rule and creep fracture strain, because creep elongation is about 20% for the present steel. If we use half value of fracture strain for fracture criteria, the twice value of da/dt is obtained as shown in Fig. 9. These computational results coincide with the experimental ones that da/dt is inversely proportional to creep ductility [12].

Three-dimensional FE model and equivalent creep strain contour is shown in Fig. 10. In this case, creep crack grows faster in the center of specimen thickness than in specimen surface due to the three-dimensional constraint effect [12,14]. Because the creep fracture strain is influenced by the stress multi-axiality, the smaller fracture strain value than that used in plane stress analysis

Fig.8. 2D analytical model for creep crack growth.

Fig.9. Comparison of experimental and computational relation between da/dt vs. C^* .

should be used in 3D analysis. The effect of specimen thickness on creep fracture strain value, however, is not clear, so we indicate the computed examples obtained by assuming that the critical strain is 10% in Fig. 11. The dashed curves show the computed crack length in the

Fig.10. 3D model and equivalent creep strain contour.

Fig.11. Creep crack growth behavior of welded joint of P122 steel.

center of specimen thickness, which are obtained using creep parameters for fine-grained HAZ and base metal in Table 1. Although the tendency of crack growth life for simulated HAZ, GTAW joint and base metal can be simulated in Fig. 11, the effect of critical strain value under multi-axial condition and also the effect of FE mesh configuration should be further investigated.

In order to evaluate the vacancy diffusion under multi-axial stress condition, we conducted the following calculations. The vacancy diffusion equation under multi-axial condition is given as follows [15]:

$$\frac{\partial C}{\partial t} = D \nabla \left(\nabla C + \frac{C}{RT} \nabla v \right), \quad (1)$$

$$v = -\sigma_p \Delta V, \quad (2)$$

where, C is the vacancy concentration, D is the diffusion coefficient, R is the gas constant, T is the absolute temperature, σ_p is the hydrostatic stress and ΔV is the volume changes by vacancy diffusion. In this equation the vacancy diffusion is controlled by hydrostatic stress

gradient. In order to simulate realistic vacancy diffusion and concentration, Eq.(3) in which weight coefficient α_1 and α_2 are added is adopted [16,17]

$$\frac{\partial C}{\partial t} = \alpha_1 D \nabla^2 C + \alpha_2 \frac{D}{RT} \nabla C \nabla v. \quad (3)$$

The vacancy concentration around the crack tip is calculated by using FEM according to Eq.(3). The changes of vacancy concentration ahead of the crack tip for GTAW joint model during creep calculated by using $\alpha_2/\alpha_1=300$, $\Delta V=2.0 \times 10^{-6} (\text{m}^3/\text{mol})$ and $D=1.5 \times 10^{-9}$, $1.5 \times 10^{-10} (\text{m}^2/\text{s})$ are shown in Fig. 12. The vacancy concentration increases to higher level in the center of specimen where multi-axiality is higher than that in specimen surface, when $1.5 \times 10^{-9} (\text{m}^2/\text{s})$ is used as diffusion coefficient. This result is consistent with the experimental results that creep voids were often observed in the center of specimen thickness.

Fig.12. Changes of vacancy concentration ahead of the crack tip for welded joint during creep.

Fig.13. Example for simulation of creep crack growth taking the diffusion into account.

In order to reflect the vacancy calculation to crack growth analysis, here we assume the following computational procedure. When the vacancy concentration reaches to the critical value, which means the void initiation, the stiffness matrix $[K]$ around the crack tip is decreased as the vacancy concentration increased. We attempted to simulate creep crack growth by combining the critical strain criteria and vacancy diffusion criteria. Fig. 13 shows an example calculated by assuming that the critical vacancy concentration C/C_0 is 1.5. We could obtain the qualitative result that the crack initiation time and growth rate is accelerated by taking the void formation ahead of the crack tip into account. The calculating procedure concerning void formation and parameters including critical values for simulation should be further investigated.

4. CONCLUSIONS

To investigate the type IV fracture mechanism of W strengthened high Cr steel weldment, creep tests were conducted for simulated HAZ and welded joint specimens. The computational simulation for type IV fracture was also conducted. The results are summarized as follows;

- (1) Heating up to the Ac_3 temperature during welding minimized the grain size and creep strength. The creep rupture time of the fine-grained HAZ was about one-fifth of the base metal at 923K.
- (2) Most of welded joint specimens were type IV fractured in fine-grained HAZ at 923K. The creep life of EBW joint with narrow HAZ was about twice longer than the GTAW joints, however it showed brittle type IV fracture. The growth rate of intergranular precipitates of $M_{23}C_6$ and Laves phase during creep was faster in fine-grained HAZ than in base metal.
- (3) Based on the experimental results, FEM code that simulates type IV crack growth has been developed. The vacancy diffusion under multi-axial stress condition in HAZ of weldment is calculated. The effect of creep

ductility and void formation ahead of the crack tip on crack growth rate could be simulated.

REFERENCES

1. K. Muramatsu, Advanced Heat Resistant Steels for Power Generation, ed. by R. Viswanathan and J. Nutting, The University Press, Cambridge, (1999) 543.
2. F. Abe, H. Okada, S. Wanikawa, M. Tabuchi, T. Itagaki, K. Kimura, K. Yamaguchi and M. Igarashi, Proc. 7th Liege Conf. on Materials for Advanced Power Engineering, Liege (2002) 1397.
3. F. Takemasa, K. Itoh and A. Suzuki, Proc. 37th Symposium on Strength of Materials at High Temperatures, Kawasaki, (1999) 101 (in Japanese).
4. M. Hashimoto, T. Koyama, T. Sato and K. Tamura, J. Soc. Mater. Sci., Japan, **44** (1995) 11 (in Japanese).
5. E. Letofsky and H. Cerjak H, Modeling of Microstructural Evolution in Creep Resistant Materials, The University Press, Cambridge, (1999) 91.
6. Y. Hasegawa, M. Ohgami, Y. Okamoto, Proc. Advanced Heat Resisting Steels for Power Generation, San Sebastian, Spain, (1998) 581.
7. K. Ogawa, A. Iseda, Y. Sawaragi, S. Matsumoto and F. Masuyama, Sumitomo Metals, **47** (1995) 39 (in Japanese).
8. M. Tabuchi, T. Watanabe, K. Kubo, M. Matsui, J. Kinugawa and F. Abe, J. Soc. Mater. Sci., Japan, **50** (2001) 116 (in Japanese).
9. T. Ogata and M. Yaguchi, J. Soc. Mater. Sci., Japan, **47** (1998) 253 (in Japanese).
10. H. Riedel, Fracture at High Temperatures, Springer-Verlag, Berlin (1987) 272.
11. M. Tabuchi, K. Kubo and K. Yagi, J. Soc. Mater. Sci., Japan, **41** (1992) 1255 (in Japanese).
12. K. M. Nikbin, D. J. Smith and G. A. Webster, Trans. ASME, J. Eng. Mater. Tech., **108** (1986) 189.
13. T. R. Hsu and Z. H. Zhai, Eng. Fract. Mech., **20** (1984) 521.
14. M. Tabuchi, K. Kubo and K. Yagi, J. Soc. Mater. Sci., Japan, **46** (1997) 53 (in Japanese).
15. H.P. Leeuwen, Eng. Fract. Mech., **6** (1974) 141.
16. A. T. Yokobori, Jr., T. Nemoto, K. Sato and T. Yamada, Japan Soc. Mech. Eng., **59** (1993) 104 (in Japanese).
17. K. Kikuchi and Y. Kaji, Proc. 33rd Symposium on Strength of Materials at High Temperatures, Yokohama, (1995) 1 (in Japanese).