Creep Behavior of 2.25Cr–1Mo Steel Shield Metal Arc Weldment

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In order to guarantee the safety of high temperature components and prevent unscheduled outage, remnant life assessment for creep is quite important. The experiences in actual components reveal that almost all the problems relating to creep are generated at the weldment, which contains microstructures with different creep properties from those presumed in the parent material. Nowadays, it is generally accepted that a terminal failure mode of low alloy ferritic steels or tempered martensitic steels become Type IV, taking place at the Intercritical HAZ or Fine Grained HAZ. However, in the present work using 2.25Cr–1Mo steel welds fabricated with basic coated Shield Metal Arc Weld (SMAW) consumables, the most prevalent failure mode was Type I in the center of weld metal rather than Type IV. And this tendency was more pronounced at low stresses, suggesting the possibility of Type I cracking in actual components. It was concluded that poor creep strength of 2.25Cr–1Mo weld metal was ascribed to its fully bainitic microstructure in which carbides evolution was accelerated and $M_{23}C_6$ depletion took place earlier than base metal.

KEY WORDS: creep; weldment; 2.25Cr–1Mo steel; weld metal; bainite.

1. Introduction

Nowadays, most refiners are trying to extend the interval of turnarounds in order to increase the competitiveness of their plants. Therefore, more accurate life assessment techniques based upon profound knowledge on the material's reaction to an operational environment are required to guarantee the soundness of a component. It is pretty much so for old high temperature components whose operating duration exceed their design life of 100 000 h. In the case of high temperature refining components, the life determining damage mechanism will be creep when so called dry corrosion such as sulfidation, High Temperature Hydrogen Attack (HTHA), metal dusting and so on are prevented by careful design, operation and maintenance. One of the problems in assessing component's integrity against creep is that this phenomenon is not necessarily associated with apparent symptoms such as reduction in wall thickness or surface cracking until the final stage of catastrophic failure. The presence of weldment, which is inevitable for most components operated in actual plants, makes the remnant life assessment more difficult to perform. However, it is the weldment where the ultimate creep failure often takes place like other mechanical problems. The weldment is composed of parent material, weld metal and Heat Affected Zone (HAZ) produced by heat input during welding. Creep properties of weld metal and HAZ could be significantly different from those of parent material. Shuller *et al.*¹⁾ categorized the creep damage found in the weldment of low alloy ferritic steels into four types based upon the location failed. The failure at parent material is not involved in this classification

since it is quite unlikely. Among them, Type IV cracking at the Intercritical HAZ (ICZ) or Fine Grained HAZ (FGZ) has been considered the most likely terminal failure mode.²⁾

However, in the present work on creep behavior of 2.25Cr–1Mo steel welds, Type I failure in the center of weld metal fabricated with a Shield Metal Arc Weld (SMAW) technique was more prevalent than Type IV. And the transition of failure mode from Type IV to Type I was observed with decrease in testing stresses, suggesting the possibility of Type I failure in actual components. In the followings, features of damage and creep properties of 2.25Cr–1Mo steel weldment shall be discussed.

2. Experimental Details

2.1. Materials

Two types of parent materials, a commercial plate and specially produced one doped with P, Sn, As and Sb were prepared. Chemical compositions of parent materials and weld metal are shown in **Table 1**. *J* factor and $\bar{X}$, which have been utilized to assess the susceptibility to temper embrittlement caused by enrichment of impurities, are also shown. Oxygen content in weld metal, which represents concentration of oxide inclusions, lies in the expected level when basic coated Shield Metal Arc Welding (SMAW) electrodes are employed.³⁾

The same heat treatment prior to welding was given to both plates. Normalizing was performed at 900°C for 34 min and then tempering at 730°C for 53 min was given. Microstructure of commercial and high residual parent is

Table 1. Chemical compositions of parent materials and weld metal (wt%).

	Commercial cast	High residual cast	Weld metal
C	0.15	0.11	0.12
Si	0.17	0.30	0.46
Mn	0.58	0.52	0.78
P	0.011	0.03	0.008
S	0.003	0.03	0.003
Cr	2.35	2.28	2.25
Mo	1.01	1.01	0.99
Cu	0.02	0.21	0.03
As	0.001	0.021	0.0034
Sn	0.0004	0.039	0.0033
Sb	0.0004	0.054	0.0004
O	0.0028	0.0016	0.0247
J factor ⁽¹⁾	85.5	566	140
X-bar ⁽²⁾	11.5	74.7	9.86

*1) $J = (Si+Mn)(P+Sn) \times 10^4$ (in wt%)

*2) X-bar = $(10P+5Sb+4Sn+As)/100$ (in ppm)

Fig. 1. Optical microstructure of commercial parent.

Fig. 2. Optical microstructure of high residual parent.

shown in **Figs. 1** and **2**, respectively. Parent material of a commercial cast is mainly composed of bainite associated with a small amount of proeutectoid ferrite around Prior Austenite Grain boundary (PAGB). As shown in **Fig. 2**, a volume fraction of ferrite in a high residual cast is significantly higher compared with a commercial cast.

2.2. Welding Procedure and Hardness Distribution

To prepare cross-weld specimens, commercial SMAW

Table 2. Welding condition.

Weld preparation	50° V with back chipping
Number of pass	23
Electrode diameter (mm)	5
Welding current (A)	220/260
Voltage (V)	20/28
Travel speed (mm/min)	120/260
Preheating temperature (°C)	150/250
Interpass temperature (°C)	150/250
PWHT (°C)	690±15 x 8hrs

Fig. 3. Optical microstructure of weld metal.

consumables, which were equivalent to AWS E9016B, were applied. Conditions for multi-pass welding for both plates with 32 mm thickness are shown in **Table 2**. Microstructure observed in the center of weld metal is predominantly bainitic as shown in **Fig. 3**. Randomly distributed black dots in this figure are non-metallic inclusions. They tended to be concentrated in the central part of weld deposits. No apparent cusp-like feature of a weld deposit existed in gauge length of creep and tensile specimens. Hardness distribution of commercial/high residual welds measured after PWHT is shown in **Fig. 4**. The highest hardness is observed at weld metal, ranging from Hv 187 to Hv 204. The averaged hardness of commercial parent, 185, is about Hv 20 higher than that of high residual parent. The ICZs of both materials show slightly lower values compared with parent materials.

2.3. Specimens and Testing

Combinations of constituting the weldment were three-fold, namely, commercial/commercial, high residual/high residual and commercial/high residual. In the creep tests for welds, a cylindrical standardized specimen with a 10 mm diameter and 12×12 mm² square specimen with 90 mm of gauge length were employed. Creep properties of parent materials were examined by the former type. An area ratio of weld metal in gauge length was approximately 30% for a cylindrical specimen and 17% for a square specimen respectively. The surface of a square cross-weld specimen was examined at interruptions during creep tests using a replication technique. All the creep tests were conducted in air with constant load. Grain boundary damage was observed by optical microscopy and Scanning Electron Microscopy (SEM). To analyze the precipitates, carbides and non-metallic inclusions, Transmission Electron

Fig. 4. Hardness distribution of the weldment.

Table 3. Tensile properties of parent materials and welds.

Specimen type	Temp. (°C)	Yield stress (MPa)	Tensile strength (MPa)	El. (%)	RA (%)	Failure position
Commercial cast base metal	RT	514	664	29.0	78.1	-
Commercial cast base metal	550	394	467	22.4	85.9	-
Commercial/commercial welds	RT	461	617	24.4	74.2	Base metal
Commercial/commercial welds	550	334	395	24.2	85.2	Base metal
High residual cast base metal	RT	343	550	31.4	71.3	-
High residual cast base metal	550	247	393	25.3	75.0	-
High residual/high residual welds	RT	330	544	20.3	62.6	Base metal
High residual/high residual welds	550	223	348	19.7	68.6	Base metal

Microscopy (TEM) and Energy Dispersive X-ray spectroscopy (EDX) were utilized in the observation of carbon extraction replicas and sectioned specimens.

3. Results

3.1. Tensile Properties

Tensile properties of parent materials and cross-weld specimens at room temperature and 550°C are shown in Table 3. Both 0.2% proof strength and ultimate tensile strength of a commercial cast are higher than those of a high residual cast. In the tensile tests for cross-weld specimens, all the failures took place at the high residual parent material in a ductile transgranular manner.

3.2. Creep Rupture Properties

Testing results obtained by cylindrical specimens and square specimens are shown in Tables 4 and 5, respectively. Time to rupture for parent materials and cross-weld specimens is plotted using Larson–Miller Parameter (*LMP*) in Fig. 5. In the present work, the constant *C* has been assumed to be 20. No remarkable difference in rupture life between a cylindrical and square cross-weld specimen can be found. For the reference, the mean rupture strength and lower boundary derived from the NIMS data base⁴⁾ for normalized and tempered 2.25Cr–1Mo plates are also shown. To derive these averaged and minimum values, the second power polynomial equations were employed. The curve for lower boundary strength was obtained by correlating Time–Temperature parameters with stresses that were 20% lower

than those in testing.

The parent material of a commercial cast showed higher creep strength than the average. On the other hand, the creep strength for parent material containing higher tramp elements, which failed transgranularly, was lower than the mean values. Higher creep strength of a commercial cast is consistent with tensile properties at room temperature and 550°C. However, in spite of higher strength in tensile tests, the most predominant failure mode for cross-weld specimens was Type I taking place in the center of weld metal. Twenty-four out of thirty-three cross-weld specimens failed in this mode. Another failure mode was Type IV at the outer-edge of HAZ on the side of a high residual cast. No parent material failure has taken place. Creep curves tested at 600°C and 100 MPa using cylindrical specimens for commercial parent, high residual parent, commercial/commercial welds and high residual/high residual welds are shown in Fig. 6. Due to the presence of weld metal, higher strain of weldment, especially in a commercial/commercial weldment failed in a Type I manner, is observable. The measured minimum creep rate of the commercial/commercial weldment became approximately four times higher than that of a commercial parent. A high residual/high residual weldment revealing slightly accelerated deformation failed in a Type IV manner. Considering the area ratio of weld metal (30%), it can be considered that the minimum creep rate of weld metal is an order of magnitude higher than that of a commercial parent. It has been demonstrated by the Finite Element Analysis (FEA) that the soft weld with lower creep strength than base metal can cause strain accel-

Table 4. Creep test results for cylindrical specimens.

Specimen ID	Temp (°C)	Stress (MPa)	t_r (h)	El. (%)	R.A. (%)	LMP at failure *1) (x10 ³)	Failure mode
CBM1 *2)	600	100	1,685	42	91	20.28	-
CBM2	625	80	823	53	81	20.58	-
CBM3	650	50	1,788	42	95	21.46	-
CBM4	650	60	752	55	95	21.11	-
CBM5	680	40	607	56	95	21.71	-
HRBM1 *2)	600	100	424	33	85	19.75	-
HRBM2	625	80	353	40	90	20.25	-
HRBM3	650	50	767	44	93	21.12	-
HRBM4	650	60	472	60	92	20.93	-
HRBM5	680	40	325	99	96	21.45	-
C/CW1 *2)	600	100	490	23	89	19.81	Type I
C/CW2	625	80	292	26	92	20.17	Type I
C/CW3	650	50	436	31	96	20.90	Type I
C/HRW1 *2)	600	80	946	16	37	20.06	Type IV *3)
C/HRW2	600	100	355	19	62	19.69	Type IV *3)
C/HRW3	625	60	910	30	92	20.62	Type I
C/HRW4	625	80	291	29	66	20.17	Type IV *3)
C/HRW5	650	50	485	38	95	20.94	Type I
HR/HRW1 *2)	600	80	970	21	46	20.07	Type IV
HR/HRW2	600	100	337	23	56	19.67	Type IV
HR/HRW3	625	60	825	35	92	20.58	Type I
HR/HRW4	625	80	238	26	56	20.09	Type IV
HR/HRW5	650	50	419	36	94	20.88	Type I

*1) : The constant C is 20.

*2): CBM is the base metal of the commercial cast, HRBM is the base metal of the high residual cast, C/CW is the commercial / commercial cast welds, C/HRW is the commercial / high residual cast welds and HR/HR is the high residual / high residual cast welds.

*3): Type IV failures took place on the high residual cast side.

Table 5. Creep test results for square cross-weld specimens.

Specimen ID	Temp. (°C)	Stress (MPa)	t_r (h)	El. (%)	LMP at failure *1) (x10 ³)	Failure mode
SC/CW1 *2)	625	60	814	30	20.57	Type I
SC/CW2	625	80	281	17	20.16	Type I
SC/CW3	650	40	802	22	21.14	Type I
SC/CW4	650	50	332	19	20.79	Type I
SC/CW5	650	60	249	22	20.67	Type I
SC/HRW1 *2)	600	70	2,066	21	20.35	Type IV *3)
SC/HRW2	625	50	1,778	21	20.88	Type I
SC/HRW3	625	60	662	21	20.49	Type I
SC/HRW4	625	80	282	23	20.16	Type I
SC/HRW5	650	40	826	24	21.15	Type I
SC/HRW6	650	50	419	22	20.88	Type I
SC/HRW7	650	60	264	28	20.70	Type I
SHR/HRW1 *2)	575	100	1,671	14	19.69	Type IV
SHR/HRW2	590	80	1,578	16	20.02	Type IV
SHR/HRW3	600	70	1,669	26	20.27	Type I
SHR/HRW4	625	60	670	25	20.50	Type I
SHR/HRW5	625	80	306	30	20.19	Type I
SHR/HRW6	650	40	826	24	21.15	Type I
SHR/HRW7	650	50	419	22	20.88	Type I
SHR/HRW8	650	60	264	28	20.70	Type I

*1) : The constant C is 20.

*2): SC/CW is the square commercial / high residual cast weldment, SC/HRW is the square commercial / high residual cast weldment and SHR/HRW is the square high residual / high residual cast weldment.

*3) : Type IV failure took place on the high residual cast side.

eration at the weld metal/ parent material interface (HAZ). For example, Nicol *et al.*⁵⁾ examined the strain rate distribution through the welds using bimaterial model (base metal and weld metal). It was found that the strain rate at the weld metal/parent interface increased with the creep mismatch in the case of a soft weld. In the current work, however, creep damage at the HAZ has been found only on the high residual side showing lesser difference in strain rate. This result

suggests that the susceptibility to Type IV damage owned by each ICZ would be more important than additional stress associated with the creep mismatch.

As shown in Fig. 5, time to Type IV failure lies above the lower boundary of base metal. On the other hand, the values of LMP associated with weld metal failure become smaller than the lower boundary for base metal at low stresses. The transition of failure mode, from Type IV to Type I with de-

Fig. 5. Larson–Miller Parameter–stress correlation for cylindrical and square specimens.

Fig. 6. Creep curves of parent material and cross-weld specimens.

crease in testing stresses, was also observed by Watanabe *et al.*⁶⁾ in the work on the creep behavior of 2.25Cr–1Mo steel weldment, which was prepared by Submerged Arc Welding (SAW). Time to failure for both types of failure in the present work, namely, the ultimate failure at weld metal and that at the ICZ, was well described by the power law relationship given by the following equation.

$$t_r = A \sigma^{-n} \exp(Q/RT) \quad \text{.....(1)}$$

or

$$P = \ln t_r - Q/RT = \ln A - n \ln \sigma \quad \text{.....(2)}$$

where, t_r is the time to failure, A is the constant, Q is the activation energy in kJ/mol, R is the universal gas constant and T is the absolute temperature in K.

By the regression analysis, A , Q and n determining the rupture life for each manner were derived. These values for cross-weld specimens failed in a Type I and Type IV manner are tabulated in Table 6. Time to rupture data for both failure modes, nine for Type IV and twenty-four for Type I, are compared with predicted values using Orr–Sharby–Dorn Parameter in Fig. 7. It was found that rupture life pre-

Table 6. Constants for the power law relationship on rupture life.

Failure Type	A	n	Q (kJ/mol)
Failure at weld metal	1.25×10^{-11}	3.39	340.7
Failure at the ICZ on the high residual cast	3.50×10^{-10}	4.68	357.0

diction for both types was achieved with the accuracy of a factor of two. It should be noted that extrapolated life derived by Eq. (1) becomes significantly shorter than that obtained by LMP. For example, the predicted time to Type I failure based upon the power law relationship is 170 000 h at 50 MPa and 540°C, which is the typical superheated steam temperature in conventional oil firing power generating plants. The equivalent value of LMP is 20.51×10^3 , which is remarkably smaller than experimental values in the present work, ranging from 20.79×10^3 to 20.94×10^3 . Indeed, Henry *et al.*⁷⁾ found the systematic decrease in LMP associated with decrease in temperatures at the iso-stress creep rupture tests for 2.25Cr–1Mo steel weld metal.

Fig. 7. Relationship between stress and Orr-Sharby-Dorn Parameter for Type IV and Type I failure.

3.3. Damage Morphology

3.3.1. Damage Evolution at Weld Metal

The morphology of the weld metal failure observed in the specimen tested for the longest duration (1 778 h) among those broken in a Type I manner is shown in Fig. 8. The ultimate failure takes place in a ductile manner, little grain boundary cavitation can be found. Cavity nucleation at non-metallic inclusions, which were identified as oxides from peaks of Si and Mn in the EDX spectrum, was not found, either. In the case of the service-exposed seam weld for steam piping in Gallatin Unit 2, where 75% through wall cracking at weld metal was found after 184 000 h of operation at 566°C, preferential cavitation at non-metallic inclusions was observed by Lundin *et al.*⁸⁾ The seam weld was fabricated by Submerged Arc Weld (SAW) using an acid flux and subcritically heat treated. Creep damage was located at the region adjacent to the fusion boundary and the cause of concentrated attack was ascribed to densely aligned oxide particles entrained from the acid flux. The reason for difference in coherency of oxides between the cracked seam weld and the present work is unknown. However, the dispersed feature and low density of oxide particles in the present weld metal suggest that they would not be of particular importance in the creep damage development.

Although Type I failure in the current work has not been associated with grain boundary cavities, brittle intergranular failure which is typically observed at the HAZ, would be the case due to decrease in ductility of weld metal. Wagner *et al.*⁹⁾ found the decreases in rupture ductility of 2.25Cr-1Mo weld metal associated with increase in testing hours. Reduction of Area in weld metal, fabricated with a SMAW technique and then subcritically heat treated, grew lower with testing time and values that were lower than 20% were obtained.

In the work by Watanabe *et al.*⁶⁾ discussed previously, they succeeded in generating the cavitation damage at the PAGB of bainitic microstructure in the weld metal, where recrystallization due to subsequent weld bead depositions took place, by the tests that were longer than 70 000 h. They examined the features of weld metal failures tested at 550°C and found the change of fracture mode. As with the

Fig. 8. Feature of transgranular failure at weld metal, tested at 625°C and 50 MPa for 1 778 h.

present work, transgranular failure took place in the short term tests, which were less than 3 000 h. By decreasing testing stresses, however, creep cavitation at PAGB was generated. They speculated that further decrease in stress and temperature would result in the ultimate failure at PAGB. It is noteworthy that the identical phenomenon, preferential creep damage at the grain boundaries of refined grains by multiple heat inputs during welding, has been found in the center of longitudinal seam welds for steam piping in actual plants.¹⁰⁾

3.3.2. Damage Evolution at HAZ

The grain boundary damage both at the ICZ and the Coarse Grained HAZ (CGZ) was found only on the high residual side. As for the cracking at the CGZ, it did not lead to the ultimate failure. One possible interpretation of higher susceptibility to Type IV and Type III damage in the high residual cast should be the grain boundary embrittlement due to segregation of impurities. Yamauchi *et al.*^{11,12)} compared creep properties of service-exposed 2.25Cr-1Mo steel containing a higher amount of impurities such as P, S, Cu, As, Sn and Sb with those of a new one with lower impurities. Lower ductility and higher creep rate were observed in the service-exposed material in the experiments using the simulated Fine Grained HAZ specimens.

Fig. 9. Creep damage at the unbroken ICZ, tested at 575°C and 100 MPa for 1 671 h.

Fujibayashi *et al.*¹³⁾ found the difference in the sensitivity to Type IV cracking in the service-exposed 1.25Cr–0.5Mo weldment, which was composed of a forged flange and normalized and tempered plate. Although parent material on the flange side containing a higher amount of tramp elements showed higher creep strength, Type IV failure took place only on the flange side. The significant enrichment of Sb and Sn on the grain boundaries located at the flange ICZ was found by Auger Electron Spectroscopy in an interrupted crept specimen.

The observation of grain boundary damage at the interruption was made using square cross-weld specimens. However, satisfactory data correlating the life fraction consumed with an extent of cavitation was not derived because of dormant evolution of grain boundary damage till the final stage of failure. In **Fig. 9**, Type IV damage observed on the opposite side of the rupture for the failed high residual/high residual square specimen is shown. Although the life consumption in the unbroken region can be considered almost unity, grain boundary damage is quite localized. Localized cracking and absence of cavities in the surrounding region suggests that the final failure was caused by linkage of cracks rather than coalescence of grain boundary cavities. And cracks at the grain boundaries, which are inclined and almost parallel to the principal stress, reveal that the grain boundary sliding plays a major role in crack formation. Volumetric growth of grain boundary damage due to opening of the cracking can be seen at the grain boundary that is perpendicular to tensile direction. Under the current testing conditions, where the longest testing duration for weldment is 2 066 h, the change in cavity area ratio should be more representative of the remnant life in comparison with the increase in number of cavities.

3.4. Carbides Morphology

The morphology of carbides prior to the test found in the parent material of a high residual cast is shown in **Fig. 10**. The microstructure consists of ferrite containing intragranular needle-like M_2C and bainite associated with coarse carbides. Due to longer holding time for PWHT than normal practice (2 h per 1" of thickness), $M_{23}C_6$ carbides were found in a bainite region. Carbides morphology in commercial parent prior to the test is shown in **Fig. 11**. The feature

Fig. 10. Carbides morphology of high residual parent prior to the test.

Fig. 11. Carbides morphology of commercial parent prior to the test.

Fig. 12. Carbides morphology of weld metal prior to the test.

of carbide is rather uniform and $M_{23}C_6$ were found both inside the grains and grain boundaries. M_2C , which were much smaller than those found in high residual parent, were found around rod-like $M_{23}C_6$. In **Fig. 12**, carbides in weld metal found before the test are shown. Coarse carbides, shaped ellipsoidal or polygonal, and smaller ones around

them are observable. It was considered from the electron diffraction pattern that most of these small carbides were M_2C . In weld metal, M_6C was already generated before the test.

Carbides in high residual parent of the interrupted commercial/high residual welds, crept at 625°C and 60 MPa, are shown in Fig. 13. The interruption was made after 728 h of testing. Larson–Miller Parameter ($C=20$) at the interruption is 20.53×10^3 which gives the equivalent operating time of 178 800 h at 540°C. Although coarsening and spheroidizing of carbides in bainite and resultant generation of Precipitate Free Zone (PFZ) are observed, needle-like carbides in ferrite have increased their density at the core of grains. In a bainite region, M_6C was detected by electron diffraction pattern.

In the crept commercial parent, appreciable change in carbide spacing was not observed though decreasing in small M_2C and coarsening of some carbides were progressing. In the case of the crept weld metal, most of small M_2C are depleted as shown in Fig. 14. And the growth of coarse carbides at the expense of M_2C is observed.

4. Discussion

It is generally accepted that the high temperature ferritic

components eventually fail in a Type IV manner when operating duration becomes long. In the present work, however, it was Type I failure which was the most predominant mode in a high LMP region. Testing results suggest the possibility of Type I failures at low stresses that are expected in the actual components. Poor creep strength of weld metal fabricated with a SMAW technique at low stresses are also found in the previous works by Wagner *et al.*⁹⁾ and Leyda *et al.*¹⁴⁾ Creep rupture data for weld metal and weldment are shown together with those in the present work in Fig. 15. As for rupture data of weldments, only the results of weld metal failure are plotted. Klueh *et al.*¹⁵⁾ found that creep strength of weld metal depended upon its carbon content. In their experiments conducted at rather high stresses, 103 MPa at the lowest, creep strength in terms of time to rupture and the minimum creep rate increased with carbon content. To show the effect of carbon, data in Fig. 15 were divided into three groups which were low carbon ($C < 0.05$ wt%), medium carbon ($0.05 \text{ wt}\% \leq C \leq 0.1 \text{ wt}\%$) and high carbon ($C > 0.1 \text{ wt}\%$). All the weld metals and welds were given PWHT at the temperatures ranging from 691 to 732°C. LMP -stress plotting shows significant scatter especially at high stresses and no obvious relationship between carbon content and creep strength can be found. At the stress of 60 MPa and lower, all the data lie below the

Fig. 13. Carbides morphology of crept high residual parent, tested at 625°C and 60 MPa for 728 h.

Fig. 14. Carbides morphology of crept weld metal, tested at 625°C and 60 MPa for 728 h.

Fig. 15. Creep strength of 2.25Cr-1Mo weld metal fabricated with a SMAW technique.

Fig. 16. Comparison of creep strength of 2.25Cr-1Mo weld metal with that of 1.25Cr-0.5Mo weld metal.

lower boundary of base metal independently of the carbon level in weld metal.

It must be noted that Wagner *et al.*⁹⁾ and Leyda *et al.*¹⁴⁾ also examined creep properties of 2.25Cr-1Mo steel fabricated using other welding techniques, which were Electroslag and Submerged Arc weld, and they found no significant difference in creep properties associated with welding techniques.

As discussed in Sec. 3.4, carbides in weld metal more readily transform into equilibrium ones in comparison with base metal. The reason for progressive carbide evolution at the weld metal should be attributable to rapid cooling at welding and resultant fully bainitic microstructure containing little proeutectoid and acicular ferrite. In Fig. 16, LMP-stress correlation for weld metal fabricated from 1.25Cr-0.5Mo steel with a SMAW technique is plotted together with the present data for 2.25Cr-1Mo steel weldment. For the comparison, mean strength and lower boundary for normalized and tempered plates derived from NIMS data sheet¹⁶⁾ for normalized and tempered 1.25Cr-0.5Mo steel plates are also shown. The 1.25Cr-0.5Mo steel weld metal in this graph are unused one examined by Leyda *et al.*¹⁴⁾ and the service-exposed examined by Fujibayashi *et al.*¹³⁾ The service exposed weld metal had been in service for twenty-three years at approximately 500°C. Both welds were subcritically heat treated after multi-run welding. Carbon content for virgin and ex-service weld metal was 0.02 wt% and 0.066 wt% respectively. Creep strength of both weld metal are comparable not only to their base metal but also to the weld metal fabricated from 2.25Cr-1Mo steel. From the results shown in Fig. 16, apparent advantage of employing a 2.25Cr-1Mo steel weld consumable over a lower alloying consumable, in terms of creep strength, cannot be found. The microstructure of weld metal of the service-exposed 1.25Cr-0.5Mo steel, observed in the middle of wall-thickness is shown in Fig. 17. Unlike the weld metal fabricated from 2.25Cr-1Mo steel shown in Fig. 3, that for 1.25Cr-0.5Mo steel shows more columnar feature and has retained a higher amount of ferrite. Large difference in a ferrite content in weld deposit can be explained by the work of Evans *et al.*¹⁷⁾ They found that the microstruc-

Fig. 17. Microstructure of the service-exposed 1.25Cr-0.5Mo weld metal, operated at 500°C for twenty-three years.

ture of weld metal generated in Cr-Mo steels became a function of chromium and molybdenum content. The amount of proeutectoid ferrite decreased with increase in concentration of both elements and that of acicular ferrite peaked when Cr and Mo content was approximately 1% and 0.5% respectively.

It has been pointed out that the way of sequential transformation of carbides depends upon the original microstructure in which carbides are existing. Yukitoshi *et al.*¹⁸⁾ examined the influence of original microstructures of 2.25Cr-1Mo steel upon creep properties and change in carbide morphology during high temperature exposure. The material was given three types of heat treatment prior to tests, which were Water Quenched (WQ), Oil Quenched (OQ) and Furnace Cooled (FC). The microstructures of a WQ steel was martensite and that for a OQ was bainite plus a small amount of martensite. Equiaxed ferrite and pearlite were generated in a FC steel. As expected, the order of hardness of weld metal was WQ, OQ and FC. However, creep strength of a OQ steel was the highest at high stresses and that of a FC steel became the highest at 60 MPa and lower. They attributed to the reverse of creep strength to the

stability of needle-like M_2C . Significant difference in remained M_2C was observed among the three. M_2C existing in a WQ and OQ steel was depleted when LMP ($C=20$) exceeded 21×10^3 . On the other hand, M_2C in a FC steel had survived until 22×10^3 of LMP . In the case of parent material, a commercial cast containing a higher amount of bainite showed higher creep strength than a high residual cast. Therefore, the ideal content of bainite conferring the best creep properties to 2.25Cr–1Mo steel might exist.

5. Summary

From the experimental works on 2.25Cr–1Mo steel weldment, following observations and conclusions were made.

(1) As with Type IV cracking, Type I failure of 2.25Cr–1Mo steel welds could be a serious threat to actual components operated at low stresses, independently of carbon content and welding techniques.

(2) Carbides evolution at the weld metal fabricated from 2.25Cr–1Mo steel was more progressive than parent material. A major contributor to creep strength, M_2C , was readily depleted in weld metal, resulting in poor creep resistance.

(3) Low creep strength of 2.25Cr–1Mo weld metal should be ascribed to fully bainitic microstructure in which carbides evolution can be accelerated.

(4) It was suggested that tramp elements, such as P, As, Sn and Sb could promote grain boundary damage at the ICZ from the higher susceptibility to Type IV cracking of a doped material.

(5) In the present work, in which relatively short term creep tests (2 066 h at the longest) were conducted, Type IV damage appeared late in life. And it was associated with volumetric growth of the cavities at the grain boundaries that were perpendicular to stress direction rather than apparent increase in cavity density.

(6) Time to rupture, both for Type IV and Type I failure,

was well expressed by the power law relationship.

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REFERENCES

- 1) H. J. Shuller, L. Hagn and A. Woitscheck: *Maschinenshaden*, **1** (1974), 1.
- 2) A. T. Price and J. A. Williams: Recent Advances in Creep and Fracture of Engineering Materials, Pineridge Press, Swansea, (1982), 265.
- 3) J. F. Langaster: Metallurgy of Welding, 6th Ed., Abington Publishing, Cambridge, (1999), 219.
- 4) NRIM Creep Data Sheet, No. 11B, NIMS, Tsukuba, (1997).
- 5) D. A. C. Nicol and J. A. Williams: *Res. Mech.*, **14** (1985), 197.
- 6) T. Watanabe, M. Yamazaki, H. Hongo, J. Kinugawa, T. Tanabe and Y. Momma: *J. Soc. Mater. Sci. Jpn.*, **48** (1999), 122.
- 7) J. F. Henry, F. V. Ellis and C. D. Lundin: WRC Bulletin 354, Welding Research Council, New York, (1990), 39.
- 8) C. D. Lundin, K. K. Khan, D. Yang, S. Hilton and W. Zielke: WRC Bulletin 354, Welding Research Council, New York, (1990), 1.
- 9) Properties of Steel Weldments for Elevated Temperature Pressure Containment Applications, ASME, New York, (1978), 103.
- 10) F. V. Ellis and R. Viswanathan: ASME PVP, 380, ASME, New York, (1998), 59.
- 11) M. Yamauchi, N. Nishimura, F. Masuyama and T. Sakurai: Proc. of 33rd Symp. on Strength of Materials at High Temperatures, The Society of Materials Science, Kyoto, (1995), 61.
- 12) M. Yamauchi, N. Nishimura, F. Masuyama and M. Sugazawa: Proc. of 33rd Symp. on Strength of Materials at High Temperatures, The Society of Materials Science, Kyoto, (1995), 65.
- 13) S. Fujibayashi and T. Endo: *ISIJ Int.*, **42** (2002), 1309.
- 14) Properties of Steel Weldments for Elevated Temperature Pressure Containment Applications, ASME, New York, (1978), 183.
- 15) R. J. Klueh and D. A. Canonico: *Weld. J.*, **55** (1976), 381.
- 16) NRIM Creep Data Sheet, No. 21B, NIMS, Tsukuba, (1994).
- 17) G. M. Evans and N. Bailey: Metallurgy of Basic Weld Metal, Abington Publishing, Cambridge, (1997), 161.
- 18) T. Yukitoshi and K. Nishida: *Tetsu-to-Hagane*, **59** (1973), 47.