The Effect of Grain Boundary Cavities on the Tertiary Creep Behavior and Rupture Life of 1.25Cr–0.5Mo Steel Welds

Shimpei FUJIBAYASHI

Engineering Design & Technical Development, Idemitsu Engineering Co., Ltd., Makuhari Technogarden B-23F, Nakase 1-Chome, Chiba 261-8501 Japan.

(Received on February 3, 2004; accepted in final form on April 23, 2004)

In general, the ultimate creep failure mode of welds fabricated from low alloy ferritic steels can be considered to be Type IV cracking resultant from coalescence of grain boundary damage. Thus, observation of grain boundary feature using a replication technique has been one of the most common practices for the remnant life assessment of welds operated at high temperatures. The experimental results in the present work and those found in previous works, however, prove that the feature of grain boundary damage highly depends upon testing conditions. Therefore, the relationship between the extent of damage and remaining life would not necessarily be effective for the practical use if it were obtained under the extremely accelerated laboratory conditions.

In contrast, strain rate measurement predicted the rupture life with reasonable accuracy independently of the susceptibility to grain boundary cavitation. The increase in the strain rate due to the presence of creep cavities was not observed in the experiments using heat treated specimens, which had the microstructure expected at the Intercritical HAZ in whole gauge length. This fact suggests the effectiveness of strain rate measurement for the life assessment and the necessity of reconsidering the physical meaning of grain boundary damage.

KEY WORDS: Type IV cracking; low alloy ferritic steel; continuum damage mechanics; the \( \Theta \) projection technique; The \( \Omega \) method.

1. Introduction

For the high temperature components fabricated from low alloy ferritic or tempered martensitic steels, Type IV cracking taking place at the Intercritical HAZ (ICZ) or Fine Grained HAZ (FGZ) is considered to be the terminal failure mode when the life is dominated by creep. In spite of the importance of this phenomenon, its mechanism and influential factors accelerating the damage have not been fully understood. The feature of Type IV damage commonly observed is the occurrence of grain boundary cavitation. Thus, the techniques to correlate the extent of cavitation with remnant life have been developed. In order to derive the quantitative answer from observation of grain boundary cavities, which grow in terms of number and volume, the Continuum Damage Mechanics (C.D.M.) based upon the Kachanov–Rabotnov concept has been widely applied. And equations to describe the creep behavior of materials, whose damage continuously progresses associated with the creep process, have been proposed. In these methodologies, it is assumed that the stress will be redistributed to the undamaged region by the presence of cavities at neighboring grain boundaries and then it is exposed to higher stress. In the present work, however, no appreciable increase in strain rate associated with cavitation has been found. Time to rupture for all the homogeneous specimens including the simulated ICZ specimens, which showed high susceptibility to grain boundary damage and failed in an intergranular manner, was well correlated with \( \Theta_p \), which is one of the parameters for the \( \Theta \) projection technique developed to describe the whole curve by a single equation. This experimental result suggests the effectiveness of strain rate measurement for the condition assessment of welds. At the same time, the further discussion on the physical meaning of grain boundary damage should be provoked. For this purpose, the influence of grain boundary damage on the strain rate at the tertiary creep stage shall be addressed in the present paper.

2. Experimental Procedure

2.1. Materials

For the current work, the circumferential butt weldment with a diameter of 660 mm and nominal thickness of 22 mm was removed from the inlet piping to a catalytic reformer in a refining plant. The weldment, which was composed of a forged flange and a pipe fabricated from plates, had been operated approximately at 500°C for twenty-three years. The service induced damage was found at the Coarse Grained HAZ (CGZ) and ICZ. The damage at the CGZ, which developed to the level of micro-cracking or oriented cavitation, was quite localized. The service-induced dam-
age existing at the ICZ was rather minor, diagnosed as isolated cavities by the classification by Neubauer et al. The relevant code for each parent material is ASTM A182 F11 for a flange and ASTM A 387 Gr.C for a pipe respectively. The welding was made using the matching electrodes for Shield Metal Arc Weld (SMAW) and welding preparation was single V. Post Weld Heat Treatment (PWHT) was performed at 700°C for 2 h. Chemical compositions and hardness of both parent materials and weld metal are shown in Table 1.

2.2. Specimen Preparation and Testing

To simulate the microstructure at the ICZ, as-received parent materials were heated to 850°C with six minutes and held at this temperature for five minutes, then cooled in air. This heat cycle was applied twice prior to the simulated PWHT at 700°C for 2h. Hardness of the simulated ICZ is slightly higher than the original. $H_v$ of the simulated ICZ for flange and pipe is 161 and 145 respectively. Creep tests for parent materials and weld metal were performed using a standardized cylindrical specimen with 6 mm diameter in gauge length. A specimen for the simulated ICZ had the same diameter but shorter gauge length, which was 20 mm, due to the size limitation of the heating device in a high frequency current furnace. The cross-weld creep behavior was examined using a round bar specimen with helical vee notches applied to all the microstructures constituting the weldment. The geometry of a specimen is shown in Fig. 1. Earlier experimental works using a square cross-weld specimen were not necessarily successful since the ultimate failure mostly took place at the parent material on the pipe side in a ductile transgranular manner despite the growth of cavitation damage at the ICZ on the flange side. The longest testing time for a square cross-weld specimen was 2700 h. In order to generate Type IV cracking within reasonable testing duration, the specimen subjected to higher triaxiality has been employed. And it successfully generated Type IV cracking. The shortest time to Type IV failure was only 601 h. The purpose of imposing the constraint on the whole gauge length is twofold. Firstly, it was postulated that the increase in triaxiality would accelerate Type IV damage. In fact, significant cavitation inside the wall thickness was observed in square specimens which failed at the base metal on the pipe side. In contrast, less constrained external surface was little damaged in terms of grain boundary cavitation. In addition, it was considered that the presence of notches would increase the rupture life of pipe parent due to reduction in the equivalent stress as found by the Finite Element Analysis (FEA).

All the creep tests were conducted in air with a constant load machine. Strain measurement for a spirally notched specimen during a test was not made. Grain boundary damage of a spirally notched specimen was examined by observing a sectioned specimen using optical microscopy and Scanning Electron Microscopy (SEM). A couple of spirally notched specimens were sectioned prior to ruptures.

3. Results

3.1. Creep Properties of Homogeneous Microstructures and Cross-weld Specimens

Testing results on rupture and creep properties for parent materials, weld metal and simulated ICZ are tabulated in Table 2. Rupture properties of spirally notched cross-weld specimens are shown in Table 3. In Fig. 2, time to rupture of homogeneous uniaxial specimens and spirally notched specimens are shown using Larson–Miller Parameter ($C/H^{1100}$). For the reference, the mean creep strength and the lower boundary defined by the mean value $\bar{C}/H^{1100}$ derived by the data base for normalized and tempered 1.25Cr–0.5Mo steel by NIMS is shown. Stress for a spirally notched specimen is nominal one calculated by the testing load and effective cross-sectional area. Though operating duration well exceeded the designing life of 100 000 h, creep strength of all the specimens is higher than the lower boundary of virgin materials when testing stresses are 80 MPa and lower. Although the flange parent has higher creep resistant than pipe parent, the ICZ generated on the flange side shows higher susceptibility to Type IV damage. All the Type IV failures have taken place on the flange side. In Fig. 3, the cross-section of a spirally notched specimen failed at the ICZ is shown. It is observable that the final crack path preferentially runs along the outer edge of the HAZ on the flange side. The higher susceptibility to grain boundary damage on the flange side was also found in the simulated ICZ specimens. In the case of simulated ICZ specimens produced from pipe parent, no significant difference in

![Table 1. Hardness and chemical compositions (in wt%) of parent materials and weld metal.](image-url)

|   | Pipe | Flange | Weld metal |
|---|------|--------|------------|
| $H_v$ | 139 | 145 | 187 |
| C | 0.1 | 0.12 | 0.066 |
| Si | 0.65 | 0.43 | 0.43 |
| Mn | 0.47 | 0.51 | 0.61 |
| P | 0.012 | 0.015 | 0.009 |
| S | 0.004 | 0.015 | 0.009 |
| Cr | 1.21 | 1.31 | 1.37 |
| Mo | 0.45 | 0.54 | 0.57 |
| Cu | 0.005 | 0.11 | 0.092 |
| As | 0.012 | 0.029 | 0.019 |
| Sn | 0.002 | 0.023 | 0.012 |
| Sb | 0.001 | 0.003 | 0.010 |
| O | 0.004 | 0.015 | 0.031 |

![Fig. 1. Spirally notched cross-weld specimen.](image-url)
terms of creep strength and rupture manner was observable compared with the base metal. In contrast, the simulated specimens fabricated from a forged flange showed decrease in rupture life and intergranular failures took place. Therefore, it is presumed that the susceptibility to Type IV damage largely depends on the characteristics of base metal prior to welding. Fujibayashi et al. 7,9) concluded that the higher susceptibility at the flange ICZ was attributable to two inherent characteristics owned by flange parent, namely, the morphology of the primary carbide (M3C) and higher impurities. Densely populated grain boundary cementite due to the forging process, which was water quench followed by tempering, would generate continuously distributed fine grains during the \(\alpha'\gamma\) transformation by welding. As discussed below, it was observed at the ICZ for both of flange and pipe that grain boundaries around fine grains were preferential cavity nucleus sites. In addition, the enrichment of Sb and Sn was found at grain boundaries within the flange ICZ by the Auger Electron Spectroscopy (AES) analysis using an interrupted creep specimen.7) These tramp elements should accelerate the cavitation by embrittling the grain boundaries.

### Table 2. Experimental results of creep rupture properties, \(\Theta\), \(\dot{\epsilon}\), and \(\Omega\).

| Specimen reference | Temp. (°C) | Stress (MPa) | \(t_b\) (h) | El. (µm) | \(\Theta_1\) (h) | \(\Theta_2\) (h) | \(\Theta_3\) (h) | \(\dot{\epsilon}_1\) (h) | \(\dot{\epsilon}_2\) (h) | \(\dot{\epsilon}_3\) (h) | \(\Omega\)  |
|-------------------|-----------|-------------|---------|-------|-------------|-------------|-------------|-------------|-------------|-------------|-------|
| PBM1              | 670       | 30          | 834     | 41    | 6.88 x 10^-5 | 1.15 x 10^-5 | 8.02 x 10^-5 | 9.35 x 10^-5 | 3.55 x 10^-5 | 3.49        |
| PBM2              | 670       | 40          | 323     | 40    | 1.47 x 10^-3 | 1.29 x 10^-3 | 1.96 x 10^-3 | 1.42 x 10^-3 | 1.46 x 10^-4 | 21.4        |
| PBM3              | 650       | 60          | 130     | 57    | 7.6 x 10^-5  | 5.13 x 10^-5 | 1.01 x 10^-2 | 2.51 x 10^-2 | 9.58 x 10^-4 | 10.5        |
| PBM4              | 650       | 60          | 110     | 81    | 6.02 x 10^-5 | 2.94 x 10^-5 | 6.41 x 10^-5 | 3.79 x 10^-5 | 9.06 x 10^-5 | 9.74        |
| PBM5              | 630       | 60          | 418     | 65    | 3.58 x 10^-5 | 2.90 x 10^-5 | 8.20 x 10^-5 | 3.98 x 10^-5 | 1.40 x 10^-4 | 16.0        |
| PBM6              | 600       | 60          | 2730    | 13    | 1.53 x 10^-3 | 3.22 x 10^-3 | 1.07 x 10^-2 | 1.06 x 10^-2 | 1.87 x 10^-2 | 20.1        |
| PBM7              | 600       | 60          | 1974    | 28    | 2.26 x 10^-3 | 8.96 x 10^-4 | 9.03 x 10^-4 | 1.42 x 10^-3 | 2.25 x 10^-3 | 23.5        |
| PBM8              | 610       | 80          | 199     | 63    | 1.89 x 10^-3 | 3.66 x 10^-3 | 1.81 x 10^-3 | 1.43 x 10^-3 | 5.21 x 10^-3 | 9.00        |
| PBM9              | 610       | 100         | 25      | 70    | 4.08 x 10^-3 | 3.01 x 10^-3 | 1.16 x 10^-3 | 1.37 x 10^-3 | 4.25 x 10^-3 | 8.11        |
| PBM10             | 600       | 100         | 57      | 65    | 4.75 x 10^-3 | 1.33 x 10^-3 | 6.66 x 10^-3 | 8.18 x 10^-3 | 2.12 x 10^-3 | 8.50        |

### Table 3. Rupture properties of spirally notched cross-weld specimens.

| Temperature (°C) | Stress (MPa) | Time to rupture (h) | Elongation (%) | Failure location | Failure mode          |
|------------------|--------------|---------------------|----------------|-----------------|----------------------|
| 670              | 30           | 1,151               | 13             | Flange ICZ      | Intergranular        |
| 670              | 40           | 601                 | 13             | Flange ICZ      | Intergranular        |
| 650              | 40           | 1,311               | 9.9            | Flange ICZ      | Intergranular        |
| 650              | 60           | 348                 | 10             | Pipe parent     | Transgranular        |
| 630              | 40           | 3,282               | 3.8            | Flange ICZ      | Intergranular        |
| 630              | 60           | 913                 | 14             | Flange ICZ      | Intergranular        |
| 610              | 60           | 1,167               | 11             | Flange ICZ      | Intergranular        |
| 600              | 60           | 3,491               | 6.6            | Flange ICZ      | Intergranular        |
| 600              | 100          | 399                 | 15             | Pipe parent     | Transgranular        |

**ICZ: Intercritical HAZ (Intercritical Region)**
3.2. Morphology of Grain Boundary Damage

The morphology of grain boundary damage adjacent to the final crack path on the flange side observed by SEM is shown in Fig. 4. Heterogeneous damage distribution, which is preferential cracking around fine grains that are smaller than 10 μm, is observable. The damage level at the pipe ICZ is significantly milder than that on the flange side as shown in Fig. 5. Unlike the ICZ on the flange side, fine grains are forming colonies and isolated by larger grains. It can be considered that discontinuous nature of fine grains impart higher resistance to Type IV cracking to the pipe ICZ. In Fig. 6, the morphology of grain boundary damage observed in the simulated ICZ generated from flange parent is shown. Despite the brittle feature of the final fracture, it showed relatively good ductility, 28% in this case. Like the actual welds, preferential cavity formation sites are grain boundaries around fine grains as shown in Fig. 7. Low temperatures and low stresses are favored to promote the grain boundary cavitation. The ductile feature of the final crack path of a simulated flange ICZ specimen tested at 610°C and 80 MPa is shown in Fig. 8. Transition of failure mode, from intergranular to transgranular, was also observed by increasing testing tempera-
Kimmins et al. claimed that Type IV cracking was caused by grain boundary sliding. The occurrence of grain boundary sliding has been also suggested directly and indirectly in the present work. In Fig. 9, grain boundary damage found by SEM observation at the ICZ on the pipe side, which shows lower sensitivity to Type IV cracking, are shown. Although grain boundary damage observed by optical microscopy is very isolated as shown in Fig. 5, wedge-like cavities at triple points and a single grain facet crack which is not perpendicular to stress direction, can be found by SEM. These damage morphologies directly suggest that they were generated by grain boundary sliding. In addition, the commonly observed tendency of Type IV cracking, which is that lower stresses and lower temperatures are preferred to promote the damage, should be ascribed to lower creep exponent and lower activation energy for grain boundary sliding.  

### 3.3. Creep Behavior of Microstructures Constituting the Welds

Kusima et al. found the good relationship between the constant, which was equivalent to \( Q_4 \) in Eq. (1), for the modified \( \Theta \) projection technique proposed by Maruyama et al. and time to rupture. The same result was obtained in the experimental works using new and service-exposed materials, by Fujibayashi et al. In the \( \Theta \) projection technique, a whole creep curve is given by Eq. (1).

\[
\varepsilon = \varepsilon_p + \varepsilon_t = \Theta_1 (1 - \exp(-\Theta_2 t)) + \Theta_3 (\exp(\Theta_4 t) - 1) \ldots (1)
\]

where \( \varepsilon_p \) is the strain at the primary stage, \( \varepsilon_t \) is the strain at the tertiary stage, \( \Theta_1, \Theta_2, \Theta_3 \) and \( \Theta_4 \) are constants depending on stress and temperature.

Hence, strain rate is given by Eq. (2).

\[
\dot{\varepsilon} = \Theta_1 \Theta_2 \exp(-\Theta_2 t) + \Theta_3 \Theta_4 (\exp(\Theta_4 t) - 1) \ldots \ldots \ldots \ldots (2)
\]

Therefore, \( \Theta_4 \) becomes the slope of strain rate versus strain plotting at the tertiary stage. In Fig. 10, the relationship between time to rupture and \( \Theta_4 \) is shown. Good correlation, which enables the life assessment with the accuracy of the factor of 2, is also observed in the simulated ICZ specimens failed with a brittle intergranular manner, suggesting that grain boundary cavitation would not significantly accelerate the tertiary strain rate. As a matter of fact, the relationship between life consumption and strain rate for the specimens with similar rupture lives, which were approxi-
mately 1200 h, became almost identical independently of failure mode when $t/t_r$ exceeds 0.5 as shown in Fig. 11. Furthermore, by measuring the Slope of Strain Rate Versus Strain (SSRVS) for a period between $t/t_r=0.6$ and $t/t_r=0.8$, which should be equivalent to $Q^4$ as described above, the rupture life of homogeneous specimens shown in Fig. 10 can be predicted with the reasonable accuracy as shown in Fig. 12.

4. Discussion

4.1. Quantification of Grain Boundary Damage

For the condition assessment of the low alloy ferritic welds, direct observation of the grain boundary damage with a replication technique is an indispensable item during turnarounds. A couple of methodologies to correlate the significance of grain boundary damage with the remnant life have been proposed.

In the case of the damage taking place at the CGZ with fully bainitic microstructure, the concept of ‘A’ parameter was developed at now defunct Central Electric Generating Board (CEGB). However, it is generally accepted that the major driving force to cause the damage at the CGZ is residual stress at weldment and the possibility leading to the ultimate failure would be low if the component were designed and operated properly. Because the residual stress at the welds would be relaxed associated with high temperature exposure. Therefore, the application of ‘A’ parameter to the actual component often results in too pessimistic assessment. As pointed out by Price et al., it should be the Type IV damage that can be a real threat to the integrity of weldment of actual equipment operated for long hours. Thus, the methodology to estimate the remaining life against Type IV cracking is important to guarantee the safe operation for users of high temperature plants. To assess the effect of grain boundary damage upon the integrity of the cavitated component, the damage mechanics introduced by Kachanov–Rabotnov has been applied. ‘A’ parameter de-
scribed above is also based upon this concept. By the simplest way introducing the Kachanov–Rabotnov damage mechanics into the constitutional equation for the creep, the strain rate at the stage following the primary creep can be described below when the damage is defined as \( w \).

\[
\dot{\varepsilon} = B \sigma^n (1 - \omega)^n \quad \text{(3)}
\]

where \( B \) is the material constant, \( n \) is the creep exponent and the \( \omega \) is the number representing the extent of damage \((0 \leq \omega \leq 1)\).

Equation (3) can be substituted by Eq. (4) on the assumption that the increasing rate of \( \omega \) is proportional to the strain rate.\(^{17}\)

\[
(1 - \dot{t}/t_r) = (1 - \dot{\varepsilon}/\varepsilon)^A \quad \text{(4)}
\]

where \( t_r \) is the time to rupture, \( \dot{\varepsilon} \) is the strain rate at time \( t \), \( \varepsilon \) is the fracture strain and \( A \) is the tertiary ductility ratio given by the following equation.

\[
A = \varepsilon_\infty / (\dot{\varepsilon}_\infty t_r) \quad \text{(5)}
\]

where \( \varepsilon_\infty \) is the minimum creep rate.

Bissell \textit{et al.}\(^{2}\) employed the cavity density, which was the number of cavities per unit area, to quantify the \( \omega \) in the life assessment of 2.25Cr–1Mo steel weldment. In their work, it was assumed that the cavity density \( N \) at the time of \( t \) was proportional to \( \varepsilon \) at \( t \). Hence, Eq. (4) can be rewritten as Eq. (6).

\[
(1 - \dot{t}/t_r) = (1 - N/N_f)^A \quad \text{(6)}
\]

where \( N_f \) is the cavity density at failure.

\( N_f \) in their work ranged from 9 000 to 12 000 (n/mm\(^2\)) and \( A \) of 10 was applied. The relationship between \( \dot{t}/t_r \) and \( N/N_f \) derived by Eq. (6) and testing results in the present work is shown in Fig. 13. For the reference, the calculated correlation by assuming \( A = 2.5 \), which has been applied to brittle microstructure like fully bainitic CGZ,\(^{15}\) is also plotted. As for the data in the present work, the relationship between \( \dot{t}/t_r \) and \( N/N_f \) differs dependently on testing conditions. Steep rise in \( N/N_f \) at the final stage of life, in which accelerated strain increase is expected, is not observable ex-
cept the one tested at 670°C and 40 MPa. In three cases out of four, the increase in N/Nt is presumed to be almost linear against t/tt. More extreme case was reported by Gooch et al.19) in the work using 0.5Cr–0.5Mo–0.25V steel. They found that the major difference in grain boundary damage between rupture and half life was cavity size rather than cavity density which ranged from 2000 to 4000. In the case of the work by Bissell et al., in which Eq. (6) and A of 10 was assumed, the abrupt increase in N/Nt should take place shortly before the rupture as shown in Fig. 13. The behavior of Type IV damage evolution for 0.5Cr–0.5Mo–0.25V steel reported by Walker et al.19) also varied from those described above. Though they didn’t examine the applicability of Eq. (6), the relationship between t/tt and N/Nt for the testing results at 60 MPa looks consistent with Eq. (6) when A of 2.5 is assumed.

Besides the difference in t/tt–N/Nt correlation, the value of Nt is significantly variable. For example, Nt observed in the current welds ranges from 1300 to 2900. In the work by Westwood20) using the new and service-exposed 1.25Cr–0.5Mo welds, Nt at the failure was less than 1000. In addition to possibility of cast to cast variation, Nt dependence on testing conditions were found in previous works. Walker et al.19) found the decrease in Nt by increasing testing stresses. And also, Fujibayashi21) reported the reduction in Nt by rising the test temperature. These phenomena can be explained when the role of grain boundary sliding is considered. Under the accelerated creep tests in a laboratory, the contribution of grain boundary sliding to total strain would be reduced.

Furthermore, considering the fact experienced both in actual plants and laboratories that the maximum cavitation damage is often found inside the wall thickness where a replication technique is not available, other techniques that are not dependent upon damage observation at grain boundaries should be introduced.

### 4.2. The Effect of Grain Boundary Damage on Creep Behavior

The Kachanov–Rabotnov model can be described as the following equation by integrating Eq. (3).15)

\[
\frac{1 - \alpha}{t} = \left(1 - \frac{t}{t^*}\right)^{\frac{A - 1}{A}} \quad \text{............(7)}
\]

By substituting Eq. (7) into Eq. (3), which is correlating the strain rate with the damage level of \(\alpha\), Eq. (8) is given.

\[
\dot{e} = B\dot{\epsilon}_0^*(1 - \frac{t}{t^*})^{\frac{1}{1 - \frac{A}{A}}} = \dot{\epsilon}_m(1 - \frac{t}{t^*})^{\frac{1}{1 - \frac{A}{A}}} \quad \text{............(8)}
\]

Thus, the increase in the strain rate associated with the damage becomes a function of a life fraction consumed and tertiary ductility ratio.

For the comparison, the change in strain rate with life consumption derived by the \(\Omega\) method developed by Prager10) shall be examined. In this methodology, the creep strain is described by Eq. (9) and \(t\) becomes \(1/\dot{\epsilon}_0\Omega\).

\[
\dot{e} = -\frac{1}{\Omega} \ln(1 - \dot{\epsilon}_0 \Omega) \quad \text{............(9)}
\]

where \(\Omega\) is the slope of the strain–natural logarithm of strain rate plotting and \(\dot{\epsilon}_0\) is the initial strain rate given by the intersection of \(y\)-axis in this plotting. Prager defined the parameter \(\Omega\) is the representative of creep damage which is the summation of matrix softening, particle coarsening and cavitation.

\[
\dot{e} = \dot{\epsilon}_0/(1 - \dot{\epsilon}_0 \Omega) = \dot{\epsilon}_0/(1 - t/t^*) \quad \text{............(10)}
\]

In the case of the \(\Omega\) method, strain rate during the creep deformation is independent of ductility (tertiary ductility ratio) and simply determined by initial strain rate and life fraction consumed. The experimental data on the relationship between \(\dot{e}/\dot{\epsilon}_0\) and \(t/t^*\) for flange parent whose \(A\) ranges from 4.8 to 32 are compared with the prediction by Eq. (10) in Fig. 14. Except the early stage of creep, most data lie in the envelope of the factor of 2 scatter band predicted by the \(\Omega\) method.

In Fig. 15, the relationship between \(\dot{e}/\dot{\epsilon}_0\) and life consumption and for the simulated flange ICZ specimens failed in both manners, intergranular and transgranular, and those derived by Eqs. (8) and (10) are shown. \(A\) of the simulated flange ICZ ranges from 3.3 to 7.2. For the direct comparison of Eqs. (8) with (10), \(\dot{\epsilon}_m\) of Eq. (8) is replaced by \(\dot{\epsilon}_0\) in Figs. 14 and 15. Except the early stage, all the data show similar behavior to those predicted by the \(\Omega\) method or Eq. (8) when \(A\) of 10 assumed. Though it seems more likely...
that a less ductile material with the low \( A \) of 2.5 shows the steep increase in strain rate associated with cavitation, a calculated result becomes just opposite. Considering that the value of 10 for \( A \) fitted to the damage progress of the experiment by Bissell et al., in which significantly higher values of \( N_r \) were observed in comparison with those in the present work, the different conclusion on the effect of grain boundary cavitation could be possible. Namely, its effect would not result in large increase in strain rate and the extent of grain boundary damage would not become the absolute parameter to predict the remnant life.

4.3. Effect of Triaxial Stress State on Type IV Cracking

Due to the shortage of experimental results, the conclusion on the triaxiality effect upon damage evolution and life of welds has not been cleared. One possible interpretation should be that higher susceptibility to Type IV cracking under the triaxial stress state is caused by reduction in ductility. Manjonie\textsuperscript{22} correlated creep ductility with the stress state using the Davis Triaxial Factor, \( TFD \), given by the following equation.

\[
TFD = \left( \sigma_1 + \sigma_2 + \sigma_3 \right) \left[ \frac{(\sigma_1 - \sigma_2)^2 + (\sigma_1 - \sigma_3)^2 + (\sigma_2 - \sigma_3)^2}{2} \right]^{1/2}
\]

where \( \sigma_1, \sigma_2 \) and \( \sigma_3 \) are the principal stresses.

It was found that creep ductility consistently decreased with the increase in \( TFD \). In the present work, elongation of spirally notched specimens was rather low, especially when Type IV failures took place as shown in Table 3. Thus, the regions subjected to higher value of \( TFD \) like the inside the wall thickness or notch root would show lower ductility and result in higher grain boundary damage.

Other experimental results, however, do not necessarily support the effect of triaxiality. Namely, the creep life of an uniaxial specimen is well correlated with \( \Theta_1 \) or SSRVS, which should be a sole function of the equivalent stress. And the quantitative explanation on the difference in rupture life between a simulated flange ICZ and spirally notched specimen has not been derived. For example, time to rupture of the simulated flange ICZ tested at 650°C and 40 MPa was shorter than that of a spirally notched specimen tested at the same condition. However, simple comparison between the two might be misleading since the vulnerable ICZ in the actual welds could be off-loaded by the adjacent stronger microstructure like a soldered joint.\textsuperscript{10}

If the rupture life under higher triaxiality were significantly reduced, the reason should be ascribed to easier linkage of grain boundary damage. From the feature of damage adjacent to the final crack pass shown in Fig. 4, it can be presumed that the linkage of single grain facet cracks would be dominant rather than coalescence of cavities late in the life. Further works including FEA have been conducted to quantify the triaxial effect on Type IV cracking.

5. Summary

The findings in the present work upon the effect of grain boundary damage on the creep behavior are described below.

1. Cavity density could not necessarily be an absolute parameter to represent the creep damage of Intercritical HAZ (ICZ). Accelerated creep tests in laboratories result in fewer cavities due to reduction in relative contribution of grain boundary sliding to total strain.

2. No appreciable increase in strain rate associated with grain boundary damage was found in the experiments using the simulated ICZ specimens, suggesting that the presence of grain boundary cavities would not result in significant increase in the stress loaded on the area without the grain boundary cavity.

3. It was confirmed that strain rate measurement was also effective to quantify the creep life of materials with high susceptibility to creep cavitation.

4. Both the \( \Theta \) projection technique and the \( \Omega \) method would be promising techniques to assess the remaining life against Type IV cracking, if the localized strain rate at the critical area were accurately measured.
REFERENCES

1) I. J. Perrin and D. R. Hayhurst: *Int. J. Pressure Vessels Piping*, 76 (1999), 599.
2) A. M. Bissell, B. J. Cane and J. F. Delong: Proc. Conf. ASME Pressure Vessels and Piping, ASME, New York, (1988), 1.
3) M. Prager: ASME RPV-288, ASME, New York, (1994), 401.
4) R. W. Evans and B. Wildshire: Creep of Metals and Alloys, The Institute of Metals, London, (1985), 197.
5) B. Neubauer and U. Wadel: Advances in Life Prediction Methods, ASME, New York, (1994), 351.
6) S. Fujibayashi, T. Ohtsuka and T. Endo: *Tetsu-to-Hagané*, 88 (2002), 326.
7) S. Fujibayashi and T. Endo: *ISIJ Int.*, 42 (2002), 1309.
8) NRIM Creep Data Sheet, No.21B, NIMS, Tsukuba, (1994).
9) S. Fujibayashi and T. Endo: *ISIJ Int.*, 43 (2003), 790.
10) S. T. Kimmins, M. C. Coleman and D. J. Smith: Int. Conf. on Creep & Fracture of Engineering Materials and Structures, The Institute of Metals, London, (1993), 681.
11) H. Evans: Mechanics of Creep Fracture, Elsevier Applied Science, London, (1984), 9.
12) H. Kushima, K. Kimura, F. Abe and K. Maruyama: *CAMP-ISIJ*, 11 (1998), 416.
13) K. Maruyama and H. Oikawa: *Trans. Jpn. Inst. Met.*, 55 (1991), 1189.
14) S. Fujibayashi, M. Miura and K. Togashi: *ISIJ Int.*, 44 (2004), 919.
15) R. Viswanathan: Damage Mechanics and Life Assessment of High-Temperature Components, ASM International, Metals Park, OH, (1989), 219.
16) A. T. Price and J. A. Williams: Recent Advances in Creep and Fracture of Engineering Materials, Pineridge Press, Swansea, (1982), 265.
17) B. F. Dyson and D. McLean: *Met. Sci.*, 6 (1972), 220.
18) D. J. Gooch and S. T. Kimmins: Int. Conf. on Creep & Fracture of Engineering Materials and Structures, The Institute of Metals, London, (1987), 689.
19) N. S. Walker, D. J. Smith and S. T. Kimmins: Proc. Int. Conf. on Creep and Fatigue, IMechE, London, (1996), 341.
20) H. J. Westwood: Int. Symp. on Materials Aging and Component Life Extension, EMAS, Sheffield, (1995), 733.
21) S. Fujibayashi: *ISIJ Int.*, 43 (2003), 2054.
22) M. J. Manjoine: *Amer. Welding J. Res. Supp.*, 2 (1982), 50.