PREDICTION OF LONG-TERM CREEP-FATIGUE LIFE OF STAINLESS STEEL WELDMENT BASED ON MICROSTRUCTURE DEGRADATION

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Abstract: This paper describes a newly developed analytical method for the evaluation of creep-fatigue strength for stainless weld metal. Based on the observation that creep-fatigue crack initiated adjacent to the interface of α-phase/δ-ferrite and matrix, a mechanical model, which allowed the evaluation of micro stress/strain concentration adjacent to the interface, was developed. Fatigue and creep damages were evaluated, using the model which described the microstructure after long time exposure to high temperatures. It is concluded that one major scope of this model is to predict analytically long term creep/fatigue life of stainless steel weld metals, the microstructure of which has been degraded as a result of in-service high temperature exposure.

Key words: Stainless steel weld metal, Degradation of microstructure, Creep-fatigue

1. INTRODUCTION

Construction cost of future large scale fast breeder reactors (FBRs) can be significantly reduced by adopting weldments instead of expensive forged parts in places where severe creep-fatigue damage is expected, for example, in reactor vessels which are subjected to thermal transients caused by varying sodium surface. Therefore, creep-fatigue evaluation method, which can accurately predict the long term strength of weldments, becomes necessary. Creep-fatigue strength of weldments is reduced being compared with parent metal as the results of:

1) Stress/strain concentration increased by the difference of inelastic response among parent material, heat affected zone (HAZ) and weld metal.
2) Stress/strain concentration enhanced by the geometrical discontinuity at the root of weld bead.
3) Degradation of weld metal and HAZ due to long term exposure to high temperature environment and stresses.

Pressure vessels undergo small plastic strain, because they are subject to weak thermal loads. If attention is given to the pressure vessels, the effect of weld bead should not be considered. Therefore, the dominant factor of the strength reduction will be the degradation of weld metal and HAZ due to long term high temperature service.

Previous investigation[1-5] has shown that creep-fatigue strength of weld metal is lower than that of parent material, and that the linear summation rule, which has been successfully applied to the parent material, gives too conservative prediction for weld metal creep-fatigue life. This implies that the degradation of weld metal should be accounted for in the procedure of weld metal creep-fatigue life prediction. Therefore the problem lies in the way to predict weld metal creep-fatigue strength, particularly after long term service of FBR plants, all while paying attention to the degradation of the weld metal. This paper suggests a new analytical evaluation method of long term creep-fatigue life of stainless steel weld metal (type 308 weld metal), that takes into account the degradation of weld metal microstructure. This method analytically evaluates creep-fatigue life that corresponds to the degraded microstructures. Therefore, the major scope of this method lies in significantly improving the reliability of the long term creep-fatigue life evaluation.

2. CREEP-FATIGUE STRENGTH REDUCTION DUE TO THE DEGRADATION OF MICROSTRUCTURE

2.1. The Problem of Weld Metal Creep-Fatigue Life Prediction

The material investigated is type 308 weld metal, which is used for SUS304 as parent material. It has been shown that, creep-fatigue strength of SUS304 is reasonably predicted by the linear damage summation rule. Nevertheless, the same method gives too unconservative prediction, by the order of one for type 308 weld metal
The linear damage summation rule evaluates creep-fatigue life, based on fatigue and creep properties of the material. This implies that creep-fatigue life of weld metal cannot be reasonably predicted based on simple fatigue and creep properties, and that microstructure degradation, which leads to creep-fatigue life reduction, thus needs to be taken into account.

2.2. Transformation of Microstructure

Type 308 weld metal, like other stainless steel weld metals, consists of austenitic phase and a few volume percent of δ-ferrite phase to prevent weld cracking. The δ-ferrite phase transforms during high temperature service. At 550 °C, (a temperature close to the operating temperature of FBRs), carbides are formed into the δ-ferrite due to short term aging, and they transforms into austenite phase. As the result, the content of alloying elements of remaining ferrite phase is much lower than the original one. Longer aging makes carbides transform into σ-phase. Therefore, the final structure is reduced to austenite structure containing σ and the remaining ferrite phases. The precipitation and transformation are accelerated by stress and strain. The rate of σ-phase precipitation is higher in creep-fatigue condition, and takes the minimum for aged specimen without any loading. For a detailed description of transformation of δ ferrite, see reference[6]. Typical micrographs for structure evolution in type 308 weld metal are shown in Fig.2.

3. MECHANISM OF MICRO CRACK INITIATION AND ITS MODELING

3.1. Mechanism of Micro-Crack Initiation

Microscopic observation reveals that micro-cracks initiate near the interface between σ or the remaining ferrite and austenite phases, as is shown in Fig.2. Creep-fatigue failure of weld metal is caused by the propagation and the coalescence of these micro-cracks. Cracks propagate and coalesce in austenite phase. This leads to the final failure of weld metal. As δ-ferrite exists separately within the austenite phase, failure of δ-ferrite itself or on the interfaces does not necessarily mean macroscopic failure of weld metal. The dominant crack is considered to propagate along austenite grain boundaries. Micro-cracks initiate adjacent to the interfaces of σ or the remaining ferrite and austenite phases, as the result of significant differences of deformation behavior between the two phases; that is, σ-phase is so hard as a rigid body, while the remaining ferrite and austenite phases behave like ferritic and austenitic steels with similar chemical compositions. This difference leads to microscopic stress/strain concentration adjacent to the interfaces.
3.2. Modeling of Micro-Crack Initiation

3.2.1. Model

We propose a new model based on the micro crack initiation mechanism described in the previous section. This model enables us to simulate the microscopic stress/strain concentration by describing the weld metal microstructure as a combination of three elements with different properties. We employed two models; simplified and detailed three-dimensional models. In both models, original ferrite phases were modeled as infinite long bars located periodically in the austenite phase. The former one consists of \( \sigma \) and austenite phases, using the assumption that the entire \( \delta \)-ferrite phase transforms into the \( \sigma \)-phase.

The interval of the \( \delta \)-ferrite phase is taken as 10nm and the diameter of \( \sigma \)-phase is assumed as 1.5, 2.5 and 3.5nm. The ferrite number corresponding to 2.5nm diameter is 5. The detailed model takes into account the remaining \( \delta \)-ferrite, along with the two above phases. In the detailed three-dimensional model, the sphere shaped \( \sigma \)-phase is located periodically in the ferrite phase with bar shape. The interval of the \( \delta \)-ferrite phases is taken as 10nm, and the diameter of \( \delta \)-ferrite and \( \sigma \)-phase is assumed as 2.5nm. Although the original ferrite phase gradually transforms with time, for the sake of simplicity, we simulated the microscopic stress/strain concentration for the structure, corresponding to the final stage of transformation.

3.2.2. Parameters

For the simulation by using the model, deformation properties, such as cyclic stress-strain curves, creep strain curves, fatigue strength and creep strength, as well as material parameters, were necessary to be propounded for the three phases, austenitic, \( \delta \)-ferrite and \( \sigma \)-phases. Since the mechanical quantities are very difficult to be evaluated for these phases, we assumed that the deformation of \( \sigma \)-phase was identical to the one of a rigid body. The deformation and strength of ferrite and austenite phases were regarded to be same as those of ferritic and austenitic steels with similar chemical compositions. We also assumed that the Cr content, in the remaining \( \delta \)-ferrite in the detailed three-dimensional model, was approximately 9%, as the after transformation, and that the corresponding deformation was equivalent to that of modified 9Cr-1Mo steel. Stress-strain curves for SUS304 steel were employed as the data for the austenite. In addition, cyclic behavior of austenite was modified, so that the simulated cyclic stress-strain curve of weld metal coincided with the measured one.

To consider the dependence of mechanical properties on heat to heat variation of weld metals, two curves were assumed: one corresponding to the upper limit of measured values shown as case A in Fig.4, and the other corresponding to the lower limit shown as case B in Fig.4.

3.2.3. Method of analysis

As the first step of our study, we assume that continuum damage mechanics is applicable to the present model. Due to the symmetricity of the model, a unit cell is employed for numerical calculation as is depicted in Fig.5. Observation of micro-crack propagation suggests that the failure of weld metal is caused by the failure of austenite phase and that the failure of the interface alone does not necessarily mean macroscopic failure[7,8], which follows the unnecessity of the modeling of the interface. A Finite element mesh of the detailed three-dimensional model is illustrated in Fig.5.
4. MICROSCOPIC STRESS/STRAIN CONCENTRATION

4.1. Comparison of Simplified and Detailed Models

The same magnitude of macroscopic strain was imposed to both the detailed and the simplified model in the direction perpendicular to the δ-ferrite and σ-phases. Microscopic stress and strain concentrations were compared between the two models as shown in Fig.6. The detailed model tends to give higher value of stress/strain concentration by 10 to 15%, regardless of macroscopic strain range. Considering that both models give similar results from a qualitative point of view and that a tremendous time is required to perform a parametric analysis, however, we only used the simplified model in the following analysis.

4.2. Microscopic Stress/Strain Concentration

An example of microscopic stress/strain concentration analysis using the simplified model is given in Fig.7. This figure indicates contours of stress and strain components in the direction of loading adjacent to the σ-phase, when a macroscopic strain of 0.5% is imposed perpendicularly to the σ-phase. Maximum strain was observed in 45 degree direction from the loading axis, and the maximum stress concentration was observed above the σ-phase.

5. EVALUATION OF MICROSCOPIC CREEP-FATIGUE DAMAGE

5.1. Definition of Microscopic Creep-Fatigue Damage

Based on the microscopic stress/strain concentration analyzed in the previous section, creep-fatigue damage due to life fraction summation[9] was evaluated. As multiaxial stress state arises adjacent to the σ-phase, it is necessary to define stress and strain components that contribute to creep-fatigue damage. In this study, considering the fact that micro-cracks initiate at the interface of σ-phase and austenite[6], the authors assumed that stress/strain components perpendicular to the interface are responsible for creep-fatigue damage. The distribution of stress/strain components perpendicular to the interface along the path C (see Fig.3(b)) is shown in Fig.8. Both maximum stress and strain are observed at the austenitic phase adjacent to the interface. Figure 8 represents the comparison of stress/strain concentration under various diameters of σ-phase. Strain concentration increases with diameter, while stress concentration seems independent from this parameter.
Microscopic fatigue and creep damages per cycle are defined by

\[
d_{\text{f.micro}} = \frac{1}{N_f(\Delta \varepsilon_{\text{micro}})} , \quad d_{\text{c.micro}} = \int_{\varepsilon_f} t_s(\sigma_{\text{micro}}) \, dt,
\]

where \( \Delta \varepsilon_{\text{micro}} \), \( \sigma_{\text{micro}} \), \( N_f \) and \( t_s \) are microscopic strain range and stress, fatigue and creep lives, respectively.

Furthermore, the criterion of creep-fatigue failure was assumed to be the same as that of macroscopic failure, that is, the Campbell type diagram [10] which crosses \((D_{\text{f.micro}}, D_{\text{c.micro}}) = (0.3, 0.3)\) was used.

Macroscopic creep-fatigue strength was predicted based on creep and fatigue damages which are obtained by Eq.(1); the creep-fatigue failure criteria is described as follows.

\[
D_{\text{f.micro}} + D_{\text{c.micro}} < D,
\]

where \( D_{\text{f.micro}} \) and \( D_{\text{c.micro}} \) are equal to \( d_{\text{f.micro}} \) multiplied by \( N_f \) and \( d_{\text{c.micro}} \) multiplied by \( N_f \), respectively. \( N_f \) is the creep-fatigue life. Damage limit \( D \) follows the Campbell type equation. The authors assumed that this method evaluated the crack initiation in the austenitic phase adjacent to the interface of austenite and \( \sigma \)-phases.

5.2. Microscopic Creep-Fatigue Damage

An example of creep-fatigue damage evaluation based on the above method is given in Fig.9. This figure shows the points where creep-fatigue damage reaches the limit value of \( D \) according to the simplified model, which means failure is predicted to be 0.5 macroscopic strain range and one hour hold time using the loading condition. Microscopic failure is found to occur above the \( \sigma \)-phase. This means that micro-cracks initiate as...
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5.3. Prediction of Macroscopic Strength

5.3.1. Anisotropy of microstructure of weld metal

In the problems analyzed so far, the load direction was perpendicular to the \( \sigma \) or the original ferrite phase, which led to the most severe stress/strain concentration. As weld metals show high micro structural anisotropy, we have to define the direction of load when we analyze macroscopic strength. Microstructures of weld metal observed on the cross section of a weld metal specimen are seen in Fig.10. It is clear that the direction of \( \delta \)-ferrite (the direction of dendrite structure) is perpendicular to the cross-sectional plane near the interface of parent material and weld metal. But it is almost horizontal near the fracture surface (failure is observed where the \( \delta \)-ferrite is perpendicular to the load) as seen in the figure. Therefore, it can be said that, irrespective of the direction of the load to weldment, some \( \delta \)-ferrites that are always found perpendicular to the load. Those \( \delta \)-ferrites are the origin to initiate micro-cracks. Thus, macroscopic strength can be evaluated, based on the result obtained under the assumption that load is perpendicular to \( \delta \)-ferrite.

5.3.2. Predicted macroscopic strength

The result of creep fatigue life prediction of type 308 weld metal using the model proposed in the previous sections is summarized in Fig.11. The predicted creep-fatigue life depends on both the stress-strain responses of the austenitic phase and the diameter of \( \sigma \)-phase. Firstly, if we assume that the stress-strain response of the austenite is the same as that of SUS304 steel and that the stress at the beginning of strain hold time can be estimated by the properties of SUS304, creep-fatigue life of weld metal is predicted to be shorter than the parent material by the order of two, regardless of the strain range. The results are shown in Fig.11 by three prediction lines, which are indicated as ‘analysis-1’. Creep-fatigue life is shorter with increasing diameter of \( \sigma \)-phase. The dependence of creep-fatigue life on the diameter of the \( \sigma \)-phase becomes more dominant in the larger strain range.

If we modify the stress-strain response of austenitic phase so that the predicted macroscopic stress-strain one coincides with the measured value of weld metal, the accuracy of creep-fatigue life prediction is improved at higher strain ranges around 0.5%. Here, the stress-strain curves applied to ‘analysis-2’ in Fig.11 are correspond to those for Cases A and B in Fig.4. For analysis-2, the predicted creep-fatigue life is shorter than that of the parent material by the order of one. In Fig. 11, creep-fatigue lives with one hour strain hold time correspond...
ing to three kinds of type 308 weld metals are plotted. The comparison of experimental data with the predicted life, shows that, if we assume the stress-strain curve of the austenitic phase to be same as that of SUS304, the predicted life tends to be too conservative at higher strain ranges. But assuming the stress-strain curve of the austenitic phase to be as Case A and Case B in Fig.4, it is possible to predict creep-fatigue life of weld metal with reasonable accuracy, at least at higher strain ranges. It follows that the experimental data with longer term at lower strain ranges are necessary to be obtained, and that the validity of the proposed model is more carefully evaluated.

6. CONCLUDING REMARKS

A new creep-fatigue life evaluation model taking into account microstructure degradation caused by long term high temperature service was developed for type 308 weld metal. The following observations were made.

1. A three-dimensional finite element model that evaluated microscopic stress/strain concentration adjacent to the \(\sigma\) -phase was developed.
2. Based on the stress/strain concentration, microscopic fatigue damage and creep damage were evaluated and creep-fatigue life of degraded material was predicted.
3. Predicted life agreed well with experimental data in relatively short term region; it was concluded that this model can also predict the long term creep-fatigue life with reasonable accuracy. This should be verified by obtaining long term experimental data.

REFERENCES