Three-Dimensional Numerical Simulation of Creep Crack Growth Behavior for 316H Steel Using a Stress-Dependent Model

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Based on detailed three-dimensional numerical simulation, creep crack growth behavior of C(T) specimen with different thicknesses of 316H steel was predicted using a stress-dependent creep ductility and strain rate model. Three regions were observed in the relation of creep crack growth rate versus fracture parameter C∗. The C(T) specimen with higher thickness exhibits higher CCG rate. The turning point 1 location from low C∗ region to transition C∗ region increases with increasing thickness, while that of turning point 2 seems to be independent of specimen thickness. Based on the finite element results, constraint-dependent turning point 1 location and creep crack growth rate equations were fitted. More accurate and realistic life assessment may be made when the stress-dependent model and the constraint effect were considered for creep life assessments of high-temperature components subjecting to a low applied load.

1. Introduction

The influence of constraint level on creep crack initiation (CCI) and creep crack growth (CCG) plays an important role and has been considered a vital issue in high-temperature component life assessments. Because the constraint can dramatically alter fracture behavior of materials and structures, it is indispensable to quantify constraint accurately and to incorporate constraint effects in structural integrity assessments of high-temperature components. For this purpose, constraint parameters, such as Q [1–3], Tz [4], R [5], R∗ [6], and Ac [7], have been proposed leading to two-parameter approach C∗-Q, C∗-Tz, C∗-R, C∗-R∗, and C∗-Ac and three-parameter approach C(t)-Tz-Q to describe the creep-tip stress and strain rate fields.

Many experimental and theoretical assessments have shown that the constraint can affect creep crack growth (CCG) rate. The CCG rates increase with the increase of crack depth [8] and specimen thickness [9–11]. The investigation of specimen type effect on CCG rates demonstrated that C(T) specimen with high constraint has the fastest CCG rate and center cracked panels (CCP) with low constraint have the lowest CCG rate [3, 12]. It also has been indicated that the CCG rate examined in C(T) specimens is significantly faster than that of M(T) specimens at a given C∗ value for various steels [12, 13].

Significant work has been done in recent decades to test and predict the creep crack growth behavior of type 316H stainless steel. Creep ductility exhaustion model has been extensively employed to predict CCG rate [14–17], and uniaxial creep ductility is usually assumed to be a constant for a given temperature. However, a lot of experimental results and analyses have shown that the creep fracture mechanism depends on stress levels, which leads to the stress dependence of the creep ductility of materials [18–20]. To obtain an accurate prediction of CCG rate under a wide range of stress level, the stress-dependent ductility and strain rate model need to be employed in finite element (FE) simulation.

In recent works [20, 21], the CCG behavior of 316H steel was predicted using the stress-dependent creep ductility and strain rate model. Two-dimensional finite element analyses were conducted for C(T) [20] and CS(T), SEN(B), SEN(T), DEN(T), and M(T) specimens [21], and good agreements have been found between the predicted CCG rates and
available experimental data. Knowing that the CCG rates in steels generally increase with the increasing specimen thickness [9–11], investigation of the out-of-plane constraint effect induced by specimen thickness on CCG behavior of 316H steel using stress-dependent creep ductility and strain rate model has not been examined. Therefore, in this work, three-dimensional finite element simulations based on stress-dependent model have been conducted for C(T) specimen with different thicknesses to account for the effect of out-of-plane constraint on CCG behavior of 316H steel.

2. Finite Element Models and Numerical Procedures

2.1. Material Properties. The elastic-plastic behavior of 316H at 550°C is shown by the true stress-strain curve in Figure 1 [22]. Young's modulus E and yield stress σ_y (which is often taken as 0.2% proof stress σ_{0.2}) at 550°C are 140 GPa and 170 MPa, respectively. Poisson's ratio ν is assumed as 0.3, which is the same as that in literature [16–18, 23]. In general, Poisson's ratio changes very little with temperature for high-temperature steels [24, 25]. A stress-dependent average creep strain rate over a wide range of stresses has been observed for many engineering steels and alloys, which is determined by different creep deformation mechanisms [20, 26–31]. Two-regime Norton creep model has been proposed to characterize the different creep deformation behaviors, where the average creep strain rate is correlated by normalized stress σ/σ_{0.2} in recent studies [19–21, 26, 32–35]. For given stress and temperature, the average creep strain rate is defined as the ratio of creep ductility ε_f to rupture time t_r. The stress dependency of the average creep strain rate \( \dot{\varepsilon}_A \) can be described using a two-regime Norton power law relationship:

\[
\dot{\varepsilon}_A = \begin{cases} 
A_1 \sigma^{n_A}, & \sigma/\sigma_{0.2} < 1.035, \\
A_2 \sigma^{n_A}, & \sigma/\sigma_{0.2} \geq 1.035,
\end{cases}
\]

where \( \dot{\varepsilon}_A \) is in h^{-1} and \( A_A \) is in MPa^{-1}h^{-1}. The stress-dependent average creep strain rate for the 316H material at 550°C is illustrated in Figure 2 [20], and power law coefficients and exponents for low stress (σ/σ_{0.2} < 1.035) and high stress (σ/σ_{0.2} ≥ 1.035) are listed in Table 1 [20].

2.2. Finite Element Models. Three-dimensional (3D) FE analyses have been conducted on plain sided C(T) specimens with different thicknesses. The geometry of the C(T) specimen is shown in Figure 3. Here, \( a_0 \) is the initial crack depth, \( B \) is the specimen thickness, \( W \) is the specimen width, and \( H \) is the distance between the clamps. The specimen width \( W \) is fixed at 50 mm, and \( H \) equals 0.71 W. The initial crack length \( a_0/W \) is set as 0.5, and four thicknesses \( B/W = 1/8, 1/4, 1/2, \) and 1 are adopted to investigate the effect of specimen thickness on CCG behavior using stress-dependent model.

Due to symmetry in geometry, only a quarter of C(T) specimen was built. The symmetry boundary condition is
applied on the un-cracked ligament and middle plane \((z/B = 0)\) of the specimen. The load is applied to the loading hole as a distributed load. The typical 3D finite element model constructed for \(C(T)\) specimen is illustrated in Figure 4(a), and the local mesh distribution around the crack tip is shown in Figure 4(b). The FE simulations were carried out using ABAQUS code [36] with eight-node linear reduced integration element (C3D8R). The element size of 100 \(\mu m\) was chosen in the crack growth zone, which is similar to the average grain sizes of 316H stainless steel [17].

2.3. Creep Damage Model and Creep Crack Growth Simulation. The creep damage accumulation ahead of the crack tip based on creep ductility exhaustion approach has been widely used in CCG simulations [14–17]. The accumulated creep damage parameter \(\omega\) is defined as follows:

\[
\begin{align*}
\omega &= \int_{t_0}^{t} \dot{\omega} dt = \int_{t_0}^{t} \frac{\dot{\varepsilon}_c}{\varepsilon_f'} dt, \\
&= \int_{t_0}^{t} \frac{\dot{\varepsilon}_c}{\varepsilon_f'} dt, \\
&= \int_{t_0}^{t} \frac{\dot{\varepsilon}_c}{\varepsilon_f'} dt,
\end{align*}
\]

(2)

where \(\dot{\omega}\) is the creep damage rate, \(\dot{\varepsilon}_c\) is equivalent creep strain rate, and \(\varepsilon_f'\) is the multiaxial creep ductility. The damage parameter \(\omega\) equals 0 at time \(t = 0\) and 1 when a failure occurs. When the value of \(\omega\) at a given element reaches 1, it is considered fully damaged and its load-carrying capacity is reduced to a value of close to zero. In this paper, the elastic modulus is reduced to 1 MPa to simulate the loss of load-carrying capacity which has been used in recent works [14, 17, 20, 21, 37–39].

Many models have been proposed to estimate the multiaxial creep ductility \(\varepsilon_f'\) from uniaxial creep ductility \(\varepsilon_f\). The one which has been used in this study is the Cocks and Ashby void growth model [40] which is shown as

\[
\frac{\varepsilon_f'}{\varepsilon_f} = \frac{\sinh[2/3(n - 0.5/n + 0.5)]}{\sinh[2(n - 0.5/n + 0.5)]} \frac{\sigma_m}{\sigma_e},
\]

(3)

where \(n\) is the creep exponent for power law creep and \(\sigma_m/\sigma_e\) is the ratio between mean (hydrostatic) stress and equivalent (von Mises) stress which is known as stress triaxiality.

A lot of experimental results and analyses have shown that the creep fracture mechanism depends on stress level, which leads to the stress-dependent creep failure strain (creep ductility) [20, 33, 34, 41, 42]. At high stress, the creep fracture progress is dominated by plasticity controlled void growth [43, 44]. The cavity growth rate is linear with creep strain rate and inversely proportional to creep rupture time, resulting in a constant higher creep ductility [41]. At low stress, the fracture mechanism is constrained diffusion cavity growth. The local deformation due to the growth of inter-granular voids exceeds the deformation rate of the surrounding material. By virtue of redistributed stresses, the cavitating boundaries transverse to the maximum principal stress unload until the local strain rate equals the remote strain rate caused by the applied stress [45]. The cavity growth rate is similar to that at high stress, resulting in a constant lower creep ductility [41]. At a narrow transition stress region, the fracture mechanism is diffusion controlled cavity growth and the cavity growth is linear with stress, giving the creep ductility as a function of stress [41]. This stress-dependent creep ductility has been identified for 316H steel [20], Cr-Mo-V steel [35], and T92, P92, T122, and P122 steels [46–48].

The stress-dependent creep ductility normalized at the lower shelf (LS), the upper shelf (US), and stress-dependent transition region in between is shown in Figure 5 [20]. A large ductility test data scatter collected from NIMS [49] and EDF Energy [50] has been found for 316H due to various material casts and batch differences. An average value of ductility has been calculated at high normalized stress shown in Figure 5. However, the ductility test data at low stress are not available. An estimated creep ductility at low stress has been obtained by assuming the same ductility difference between US ductility and LS ductility at 650°C and 550°C [20]. Then, the stress-dependent region between high stress and low stress was generated and illustrated by dashed lines in Figure 5. The stress-dependent creep ductility has been used to predict creep crack growth and creep crack initiation behaviors. A good agreement has been obtained for both short- and long-term creep crack growth test data [20, 21] and creep crack initiation times [32] for different specimen

![Figure 4: Three-dimensional finite element model of (a) C(T) specimen and (b) local mesh around the crack tip.](image-url)
The creep ductility values for a wide stress range have been fitted and expressed in

\[
\varepsilon_f = \begin{cases} 
0.9\% , & \frac{\sigma}{\sigma_{0.2}} \leq 0.84 , \\
1.03 \times 10^{-15} (\sigma / \sigma_{0.2})^{0.01} , & 0.84 < \frac{\sigma}{\sigma_{0.2}} < 1.32 , \\
13.6\% , & \frac{\sigma}{\sigma_{0.2}} \geq 1.32 .
\end{cases}
\]  

(4)

The CCG simulations were performed using ABAQUS [36] code with a user subroutine USDFLD. The stress-dependent average creep strain rate and creep ductility illustrated in Figures 2 and 5 and given in Table 1 and equation (4) are implemented in the subroutine to calculate the creep damage at each time interval. When an element is fully creep damaged, the crack advances by reducing the load-carrying capability of that element until numerical convergence occurs. For each specimen, the range of initial stress intensity factor \( K_{in} \) for CCG simulations is in the region between 3.46 MPa m\(^{1/2}\) to 21.19 MPa m\(^{1/2}\). The creep crack growth length was estimated by numbers of fully damaged elements. The \( C^* \) value was calculated by using the equation in ASTM E1457 [51]:

\[
C^* = \frac{F \dot{\varepsilon}_c}{B_n (W - a)} H_1 \eta,
\]  

(5)

where \( F \) is the applied load, \( \dot{\varepsilon}_c \) is the creep load line displacement rate, \( B_n \) is the specimen net thickness, \( W \) is the specimen width, and \( a \) is the current crack length. In equation (5), \( H_1 \) and \( \eta \) are nondimensional geometry-dependent parameters and the solutions of which can be found in ASTM E 1457 [51]. The criteria in ASTM E 1457 [51] were applied to ensure that \( C^* \) is a valid parameter to describe the CCG rate.

### 2.4. NSW Creep Crack Growth Model.

The well-known NSW model proposed by Nikbin, Smith, and Webster [52, 53] has been widely used to predict the steady-state CCG rate from uniaxial creep data. The NSW model is based on a creep ductility exhaustion approach, and the CCG is predicted to occur when a critical level of damage is attained at a characteristic distance, \( r_0 \), ahead of the crack tip. The NSW CCG model provides different prediction lines for plane stress (PS) and plain strain (PE) conditions due to its stress state dependence. Under steady-state creep condition, the CCG rate can be predicted by the NSW model denoted as \( \dot{a}_{NSW} \) and expressed as follows:

\[
\dot{a}_{NSW} = \frac{(n + 1) \varepsilon'_f (A \varepsilon_c)^{1/n + 1} (C^* \eta_n)^{n/30}}{L_n},
\]  

(6)

where \( \varepsilon'_f \) is the multiaxial creep ductility and \( A, n \) are power law creep coefficient and exponent. It is suggested that \( \varepsilon'_f \) in (6) is taken as the uniaxial creep ductility \( \varepsilon_f \) for plane stress condition and \( \varepsilon_f / 30 \) for plane strain condition [54]. The creep process zone distance \( r_0 \) in NSW model is usually taken as the average material grain size, thus is taken as 0.1 mm in this study [17]. Note that the \( \dot{a}_{NSW} \) is not sensitive to this value [3, 52, 53, 55]. The dimensionless constant \( L_n \) can be estimated by equations in [56] for PS and PE conditions.

### 3. Validation of the Simulation Method

To investigate CCG behavior of \( C(T) \) specimen with different thicknesses in 316H steel, the simulation method based on a stress-dependent model should be validated firstly. Comparisons between available CCG test data and FE results for short-term CCG test specimen (ITC(T) CT1 [17, 57]) and long-term CCG test specimen (CT1B and CT20 [57]) in literature were made. The short-term [12, 17, 58, 59] and long-term creep test band [60, 61] in Figure 6 is gathered from extensive \( C(T) \) specimens creep crack growth test data of 316H steel at 550°C. Note that the scatter band is so wide up to a factor of around 10 for short-term test data [17]. The creep constraint effect plays a key role of leading such a scatter band. Therefore, it is quite necessary to consider creep constraint effect in structural integrity assessments and give more accurate remaining life predictions of structures or components.

The comparison of CCG rate between FE results and creep test data is shown in Figure 6. The FE results were calculated based on a stress-dependent model. It can be seen in Figure 6(a) that both FE results and short-term creep test data are within the short-term test data band. At the beginning of CCG, the FE result corresponds well with test data. As the crack advances, the CCG rate of FE result is slightly lower than that of test data. Due to high applied load in short-term CCG test, large plasticity occurs at the crack tip. In FE analysis, the crack growth may be dominated by high stress constitutive equation and upper shelf creep ductility (\( \varepsilon_f = 13.6\% \)), which is higher than the test data of creep ductility (\( \varepsilon_f = 6\% - 9.3\% \) [17]). Higher creep ductility leads to lower CCG rate. Similarity, for long-term CCG, the crack growth simulation was dominated by the low stress constitutive equation and lower shelf creep ductility (\( \varepsilon_f = 0.9\% \)), which shows higher CCG rate than test data.
Note that the CCG rate of FE result lies above the long-term test band. In general, the CCG results of FE simulations based on the stress-dependent model are consistent with that of test data. Therefore, this simulation method can be used to investigate the effect of specimen thickness on CCG behavior of 316H.

4. Results

4.1. Creep Crack Growth Rate for a Wide Range of Load Levels. Figure 7 shows the typical creep damage distribution (CCG profile) of plane-sided C(T) specimen along the thickness direction. The parameter $z$ is the distance from middle plane to surface plane, which means the middle plane and surface plane are denoted as $z/B = 0$ and $z/B = 0.5$, respectively. It can be seen in Figure 7 that the largest CCG length is observed at the middle plane $z/B = 0$ while the smallest one is clearly seen at surface plane $z/B = 0.5$. This is caused by different $C^*$ and constraint levels through the specimen thickness. The specimen region with higher crack tip fracture parameter $C^*$ and constraint level has higher CCG rate. The distribution of crack tip fracture parameter $C^*$ along thickness direction has been given with the highest $C^*$ value at mid-thickness plane and the lowest $C^*$ value at the surface plane [62]. The constraint level decreases as $z/B$ increases [23], leading to the highest constraint level at the middle plane ($z/B = 0$) and the lowest constraint level at the surface plane ($z/B = 0.5$).

Figure 8 shows the creep crack growth rate versus $C^*$ from about $1 \times 10^{-8}$ MPa m/h to $1 \times 10^{-2}$ MPa m/h. It can be seen that similar creep crack growth trends are obtained for C(T) specimen with $B = 12.5$ and $B = 25$. The whole $C^*$ region can be divided into low $C^*$ region, transition $C^*$ region (between point 1 and point 2), and high $C^*$ region. This is consistent with the experimental observation of 316H steel.

Figure 6: Comparison of creep crack growth between FE results and (a) short-term [17, 57] and (b) long-term creep test data [57].

Figure 7: Typical creep damage distribution (creep crack growth profile) of plane-sided C(T) specimen along thickness direction.
and Cr-Mo-V steel [10]. The simulation results of Cr-Mo-V steel in previous works also show such a trend [35, 37, 38, 63]. The CCG rate in low $C^*$ region is higher than that of extrapolation from high $C^*$ region, which means nonconservative prediction will be made if the extrapolated CCG rate from the lab test data is used for life prediction of practical components operated at low load level. Thus, accurate CCG prediction should be made by using actual CCG rate data at low $C^*$ region. The CCG behavior of $C(T)$ specimen with $B = 6.25$ mm and $B = 50$ mm is similar to that in Figure 8.

4.2. Effect of Specimen Thickness on CCG Behavior

Figure 9 shows the creep crack growth behavior of $C(T)$ specimen with different specimen thicknesses for a wide range of $C^*$ levels. The prediction lines calculated using the constant lower shelf (LS) creep ductility and average creep strain rate exponent at low stress level for PE and PS conditions are denoted as NSW-PE-LS and NSW-PS-LS, respectively. In addition, the NSW prediction lines corresponding to the upper shelf (US) creep ductility and average creep strain rate exponent at high stress level for PE and PS conditions are denoted as NSW-PE-US and NSW-PS-US, respectively.

It can be seen that different CCG rates were obtained at different load levels $C^*$. At both low $C^*$ and transition $C^*$ regions, the CCG rate increases with specimen thickness, which is the same as test data in the literature [9, 10]. The location of turning point 1 from low $C^*$ to transition $C^*$ increases with specimen thickness. For turning point 2 between transition $C^*$ region and high $C^*$ region, the effect of thickness on location variation is less conspicuous than that of point 1. At high $C^*$ region, with increasing load level $C^*$, the CCG rate initially increases with thickness and then slightly decreases after point 3, which was observed in creep test of Cr-Mo-V steel [10]. This is caused by a loss of
constraint at high $C^*$ levels where large plastic deformation occurs \cite{61, 64}. Note that the NSW CCG prediction is strongly dependent on creep ductility and thus higher CCG rates are predicted using NSW PE prediction lines as much lower creep ductility is employed in NSW models for the PE condition than that for PS condition. For short-term CCG data at high $C^*$ region, the NSW-PS-US prediction line falls close to the PE results, while more conservative results are predicted using NSW-PE-LS line for long-term CCG data at low $C^*$ region.

### 5. Constraint-Dependent CCG Rate Formula

Recently, a new load-independent constraint parameter $R^*$ has been proposed based on crack tip stress field by authors \cite{6, 65}, which can fully characterize out-of-plane constraint when $B/W \leq 1$ \cite{23}. In this paper, the constraint parameter $R^*$ is used to establish a constraint-dependent CCG rate formula. The method for establishing the correlation of CCG rate with creep constraint parameters has been given in detail in the recent work of authors \cite{6, 7}. The load-independent creep constraint parameter $R^*$ at steady-state creep has been investigated and expressed as follows \cite{6, 65}:

$$R^* = \sigma_{22} \frac{(C_1)^{-1/n+1} - \sigma_{22,C(T)} (C_3)^{-1/n+1}}{AL}$$

$$\text{atr} = 0.2 \text{mm}, \theta = 0, \frac{t}{t_{\text{red}}} = 1,$$  \hspace{1cm} (7)

where $\sigma_{22}$ and $C_1$ are the opening stress and $C^*$ value in the under-evaluated cracked specimen or component, respectively, and $\sigma_{22,C(T)}$ and $C_3$ are the opening stress and $C^*$ value of the standard $C(T)$ specimen in PE condition for obtaining reference stress field, respectively.

For 3D practical analyses, an average value of constraint parameter along crack front is usually used to characterize constraint in addition to that at mid-plane \cite{65–67}. Therefore, an average value of the parameter $R^*$ along specimen thickness, denoted as $R^*_{\text{avg}}$, was used to correlate with CCG rate. In this paper, 10 elements have been meshed along 3D specimen thickness. The $R^*$ value at every thickness layer from the mid-plane ($z/B = 0$) to the surface plane ($z/B = 0.5$) was obtained by (7). Then, the average value $R^*_{\text{avg}}$ can be calculated as follows \cite{6}:

$$R^*_{\text{avg}} = \frac{1}{B/2} \int_0^{B/2} R^* (z) \, dz,$$  \hspace{1cm} (8)

where $z$ is the distance from middle plane along specimen thickness.

Figure 10 shows the relation between the turning point locations in Figure 9 and the constraint parameter $R^*_{\text{avg}}$. As actual structures or components are subject to low load in practical use, only the locations of turning points 1 and 2 are shown in Figure 10. A linear relation exists between the $v$ value and constraint parameter $R^*_{\text{avg}}$ for turning point 1, while the $C^*$ values for turning point 2 are constant, which is constraint independent. An approximate $C^*$ value of $5 \times 10^{-5}$ MPa m/h was observed for point 2. If the formula between turning point 1 location and constraint parameter $R^*_{\text{avg}}$ was fitted, it can be predicted for specimens with other thicknesses and then the size of the transition region can be obtained. The fitted equation between turning point 1 and constraint parameter $R^*_{\text{avg}}$ is expressed as follows:

$$C^* = 2 \times 10^{-5} \left(1 - R^*_{\text{avg}}\right)^{-3.5}.$$  \hspace{1cm} (9)

Figure 11 shows the relation between CCG rate ratio $\dot{a}/\dot{a}_0$ and constraint parameter $R^*_{\text{avg}}$. It can be seen that a linear relation exists between $\dot{a}/\dot{a}_0$ and $(1 - R^*_{\text{avg}})$ on a log-log scale. The fitting formula is shown as follows:

$$\frac{\dot{a}}{\dot{a}_0} = 0.7 \left(1 - R^*_{\text{avg}}\right)^{-0.7},$$

$$\frac{\dot{a}}{\dot{a}_0} = 1.0 \left(1 - R^*_{\text{avg}}\right)^{-1.3},$$  \hspace{1cm} (10)
where $a_0$ is the CCG rate of standard reference C(T) specimen with $W = 50$ mm and $a/W = 0.5$ in PE condition. It has been validated that the relation is load-independent in a specific $C^*$ region in previous work [7]. This means the CCG rate can be predicted under different $C^*$ levels within a certain $C^*$ region.

The formula above may be used to predict CCG rate data for various specimens with different thicknesses and creep life of high-temperature components accounting for constraint effect. As long as the constraint parameter $R_{avg}^{\ast}$ is obtained at a certain load level $C^*$ using FE analysis, then the CCG behavior can be predicted by (10). To validate the formula above, a new $C(T)$ specimen with $a/W = 0.5$ and $B = 20$ mm is modeled. A comparison between the CCG rates calculated based on FE results obtained by stress-dependent model and that predicted by (10) with constraint parameter $R_{avg}^{\ast}$ was made and is shown in Figure 12. The NSW model prediction lines are also included. It shows a good agreement of CCG rate data between those two methods at low $C^*$ region and transition region, while the NSW PE prediction lines provide a more conservative CCG rate for a wide range of load level. In addition, the prediction of point 1 location by equation (8) ($C^* = 3.55 \times 10^{-6}$ MPa m/h) is very close to that obtained by FE simulation based on stress-dependent model. It means the CCG rate and location of the turning point at low $C^*$ region can be successfully predicted by constraint-dependent equation. Therefore, more accurate CCG data and reasonable life assessment can be made for high-temperature components subjecting to a low applied load.

The whole $C^*$ region of CCG rate can be divided into low $C^*$ region, transition $C^*$ region, and high $C^*$ region. The CCG rate in low $C^*$ region is higher than that of extrapolation from high $C^*$ region, which means nonconservative prediction will be made if the extrapolation CCG rate from the laboratory test data is used for life prediction of practical components operated at low load level.

At both low $C^*$ and transition $C^*$ regions, the CCG rate increases with specimen thickness, while a shift from increase to decrease with increasing thickness appears at high $C^*$ region. The $C^*$ values at the location of turning point 1 increase with specimen thickness. The variation of $C^*$ values at turning point 2 is less conspicuous than that of point 1, leading to the smaller size of transition region with increasing thickness.

Based on constraint parameter $R_{avg}^{\ast}$, the constraint-dependent equations of turning point 1 location and CCG rate at low $C^*$ region and transition $C^*$ region were fitted. By using the equations, the $C^*$ value at turning point 1 from low $C^*$ region to transition region may be predicted and the transition region size can be approximately obtained. The CCG rate can be easily predicted by using the constraint-dependent CCG rate formula at both low $C^*$ region and transition $C^*$ region.

Comparison of CCG rate between FE simulation using stress-dependent model and calculation by constraint-dependent equations shows a good agreement with each other. More accurate and reasonable life assessment may be conducted for high-temperature components subjecting to a low applied load.

### 6. Conclusion

In this paper, three-dimensional finite element analyses have been conducted on $C(T)$ specimen with different thicknesses using stress-dependent creep ductility and strain rate model. Based on the result, the main conclusions are as follows.

The whole $C^*$ region of CCG rate can be divided into low $C^*$ region, transition $C^*$ region, and high $C^*$ region. The CCG rate in low $C^*$ region is higher than that of extrapolation from high $C^*$ region, which means nonconservative prediction will be made if the extrapolation CCG rate from the laboratory test data is used for life prediction of practical components operated at low load level.

At both low $C^*$ and transition $C^*$ regions, the CCG rate increases with specimen thickness, while a shift from increase to decrease with increasing thickness appears at high $C^*$ region. The $C^*$ values at the location of turning point 1 increase with specimen thickness. The variation of $C^*$ values at turning point 2 is less conspicuous than that of point 1, leading to the smaller size of transition region with increasing thickness.

Based on constraint parameter $R_{avg}^{\ast}$, the constraint-dependent equations of turning point 1 location and CCG rate at low $C^*$ region and transition $C^*$ region were fitted. By using the equations, the $C^*$ value at turning point 1 from low $C^*$ region to transition region may be predicted and the transition region size can be approximately obtained. The CCG rate can be easily predicted by using the constraint-dependent CCG rate formula at both low $C^*$ region and transition $C^*$ region.

Comparison of CCG rate between FE simulation using stress-dependent model and calculation by constraint-dependent equations shows a good agreement with each other. More accurate and reasonable life assessment may be conducted for high-temperature components subjecting to a low applied load.

### Nomenclature

- $a$: Crack depth (mm)
- $a_0$: Initial crack depth (mm)
- $a$: Creep crack growth rate (mm/h)
- $a_0$: Creep crack growth rate of the standard reference $C(T)$ specimen with $W = 50$ mm and $a/W = 0.5$ in plain strain condition (mm/h)
- $a_{NSW}$: Creep crack growth rate predicted by NSW model (mm/h)
- $A, A_c$: Coefficient in the power law creep strain rate expression (MPa$^{-3/2}$h$^{-1}$)
- $A_c$: Unified characterization parameter of in-plane and out-of-plane creep constraint (1)
- $B$: Specimen thickness (mm)
- $B_n$: Specimen net thickness (mm)
- $C^*$: $C^*$ integral analogous to the $J$ integral (MPa m/h)
- $C_{1}^*$: $C^*$ value in the under-evaluated cracked specimen or component (MPa m/h)
- $C_{2}^*$: $C^*$ value of the standard $C(T)$ specimen in plain strain condition (MPa m/h)
- $C(t)$: $C(t)$ integral at time $t$ (MPa m/h)
- $E$: Young’s modulus (GPa)
- $F$: Applied load (MPa)
- $H$: Distance between the clamps (mm)
- $H_1$: Geometric function to calculate $C^*$ from load line displacement rate (1)
$I_n$: Dimensionless function of $n$ in the Riedel-Rice stress field distribution (1)

$K_{in}$: Initial stress intensity factor (MPa m$^{1/2}$)

$n$, $n_A$: Power law creep stress exponent (1)

$Q$: Constraint parameter under elastic-plastic condition (1)

$T_z$: Out-of-plane constraint parameter (1)

$r_z$: Creep process zone distance (mm)

$R^*_{\text{avg}}$: Average value of $R^*$ along crack front (1)

$t_c$: Creep rupture time (hour)

$\omega$: Creep damage parameter (1)

$\omega$: Creep damage rate (h$^{-1}$)

$W$: Specimen width (mm)

$z$: Distance from middle plane to surface plane (mm)

$\varepsilon_f$: Uniaxial creep ductility (1)

$\varepsilon_f^*$: Multiaxial creep ductility (1)

$\dot{e}$: Creep strain rate (h$^{-1}$)

$\dot{e}_A$: Average creep strain rate (h$^{-1}$)

$\dot{e}_c$: Equivalent creep strain rate (h$^{-1}$)

$\dot{e}_0$: Creep strain rate at normalized stress (h$^{-1}$)

$\sigma$: Stress (MPa)

$\sigma_c$: von Mises equivalent stress (MPa)

$\sigma_m$: Mean hydrostatic stress (MPa)

$\sigma_y$: Yield stress (MPa)

$\sigma_0$: Normalized stress (MPa)

$\sigma_0^*$: 0.2% proof stress (MPa)

$\sigma_{22}$: Opening stress (MPa)

$\sigma$ Opening stress of $C(T)$ specimen under plane strain

$\Delta_{22,C(T)}$: (MPa)

$\nu$: Poisson’s ratio (1)

$\eta$: Geometry-dependent function relating $C^\ast$ to load and displacement measurements (1)

$\Delta$: Creep load line displacement rate (mm/h)

Abbreviations

3D: Three-dimensional
CCG: Creep crack growth
CCI: Creep crack initiation
CCP: Center cracked panel
C(T): Compact tension
CS(T): C-shaped cracked tension
DEN(T): Double edge-notched tensile
FE: Finite element
LS: Lower shelf
M(T): Middle tension
NSW: Nikbin, Smith, and Webster creep crack growth model

NSW-PE-LS: NSW model prediction lines calculated with upper shelf creep ductility and average creep strain rate exponent at high stress level for plane strain condition

NSW-PS-LS: NSW model prediction lines calculated with upper shelf creep ductility and average creep strain rate exponent at high stress level for plane strain condition

Data Availability

The data used to support the findings of this study are available from the corresponding author upon request.

Conflicts of Interest

The authors declare that they have no conflicts of interest.

References

[1] P. J. Budden and R. A. Ainsworth, “The effect of constraint on creep fracture assessments,” *International Journal of Fracture*, vol. 97, no. 1/4, pp. 237–247, 1999.

[2] A. D. Bettinson, N. P. O’Dowd, K. M. Nikbin, and G. A. Webster, “Two parameter characterization of crack tip fields under creep conditions,” *IUTAM Symposium on Creep in Structures*, pp. 95–104, Springer, Berlin, Germany, 2001.

[3] K. M. Nikbin, “Justification for meso-scale modelling in quantifying constraint during creep crack growth,” *Materials Science and Engineering A*, vol. 365, no. 1-2, pp. 107–113, 2004.

[4] M. J. Xiang, Z. B. Yu, and W. Guo, “Characterization of three-dimensional crack border fields in creeping solids creeping solids,” *International Journal of Solids and Structures*, vol. 48, no. 19, pp. 2695–2705, 2011.

[5] G. Z. Wang, X. L. Liu, F. Z. Xuan, and S. T. Tu, “Effect of constraint induced by crack depth on creep crack-tip stress field in CT specimens,” *International Journal of Solids and Structures*, vol. 47, no. 1, pp. 51–57, 2010.

[6] J. P. Tan, G. Z. Wang, S. T. Tu, and F. Z. Xuan, “Load-independent creep constraint parameter and its application,” *Engineering Fracture Mechanics*, vol. 116, pp. 41–57, 2014.

[7] H. S. Ma, G. Z. Wang, F. Z. Xuan, and S. T. Tu, “Unified characterization of in-plane and out-of-plane creep constraint based on crack-tip equivalent creep strain,” *Engineering Fracture Mechanics*, vol. 142, pp. 1–20, 2015.

[8] L. Zhao, H. Jing, L. Xu, Y. Han, and J. Xiu, “Evaluation of constraint effects on creep crack growth by experimental investigation and numerical simulation,” *Engineering Fracture Mechanics*, vol. 96, pp. 251–266, 2012.

[9] M. Tabuchi, K. Kubo, and K. Yagi, “Effect of specimen size on creep crack growth rate using ultra-large CT specimens for 1Cr-Mo-V steel,” *Engineering Fracture Mechanics*, vol. 40, no. 2, pp. 311–321, 1991.

[10] J. P. Tan, S. T. Tu, G. Z. Wang, and F. Z. Xuan, “Effect and mechanism of out-of-plane constraint on creep crack growth behavior of a Cr–Mo–V steel,” *Engineering Fracture Mechanics*, vol. 99, pp. 324–334, 2013.
A. Mehmanparast, C. M. Davies, G. A. Webster, and E. Hosseini, S. Holdsworth, and E. Mazza, “Stress regime-sensitivity of creep deformation and failure behaviour of modified 9Cr-1Mo steel,” Journal of Nuclear Materials, vol. 423, no. 1-3, pp. 110–119, 2012.

M. Altenbach, K. Naumenko, and Y. Gorash, “Creep analysis for a wide stress range based on stress relaxation experiments,” Engineering Plasticity And Its Applications From Nanoscale To Macroscale: (With CD-ROM), pp. 41–46, 2009.

H. Quintero and A. Mehmanparast, “Prediction of creep crack initiation behaviour in 316H stainless steel using stress dependent creep ductility,” International Journal of Solids and Structures, vol. 97-98, pp. 101–115, 2016.

M. Tabuchi, A. Yokobori, R. Sugiura, M. Yatomi, A. Fuji, and K. Kobayashi, “Results of a Japanese round robin program for creep crack growth using Gr. 92 steel welds,” Engineering Fracture Mechanics, vol. 77, no. 15, pp. 3066–3076, 2010.

E. Hosseini, S. Holdsworth, and E. Mazza, “The LICON methodology for predicting long-time uniaxial creep rupture strength of materials,” International Journal of Pressure Vessels and Piping, vol. 111-112, pp. 27–35, 2013.

J. W. Zhang, G. Z. Wang, F. Z. Xuan, and S. T. Tu, “The influence of stress-regime dependent creep model and ductility in the prediction of creep crack growth rate in Cr–Mo–V steel,” Materials & Design, vol. 65, no. 1980-2015, pp. 644–651, 2015.

D. Hibbitt, B. Karlsson, and P. Sorensen, ABAQUS v6.10, User’s Manuals, Providence, USA, 2011.

J. W. Zhang, G. Z. Wang, F. Z. Xuan, and S. T. Tu, “In-plane and out-of-plane constraint effects on creep crack growth rate in Cr-Mo-V steel for wide range of C,” Materials at High Temperatures, vol. 32, no. 5, pp. 512–523, 2014.

J. Zhang, G. Wang, F. Xuan, and S. Tu, “Prediction of creep crack growth behaviour in Cr–Mo–V steel specimens with different constraints for a wide range of C,” Engineering Fracture Mechanics, vol. 132, pp. 70–84, 2014.

J. Z. He, G. Z. Wang, S. T. Tu, and F. Z. Xuan, “Prediction of creep crack initiation in Cr–Mo–V steel specimens with different geometries,” Materials at High Temperatures, vol. 34, no. 1, pp. 87–96, 2016.

A. C. F. Cocks and M. F. Ashby, “Intergranular fracture during power-law creep under multiaxial stresses,” Metal Science, vol. 14, no. 8-9, pp. 395–402, 1980.

R. Hales, “The role of cavity growth mechanisms in determining creep-rupture under multiaxial stresses,” Fatigue and Fracture of Engineering Materials and Structures, vol. 17, no. 5, pp. 579–591, 1994.

M. W. Spindler, “The multiaxial creep ductility of austenitic stainless steels,” Fatigue and Fracture of Engineering Materials and Structures, vol. 27, no. 4, pp. 273–281, 2004.

J. R. Rice and D. M. Tracey, “On the ductile enlargement of voids in triaxial stress fields,” Journal of the Mechanics and Physics of Solids, vol. 17, no. 3, pp. 201–217, 1969.
[44] J. W. Hancock, “Creep cavitation without a vacancy flux,” Metal Science, vol. 10, no. 9, pp. 319–325, 1976.
[45] B. F. Dyson, “Constrained cavity growth, its use in quantifying recent creep fracture results,” Canadian Metallurgical Quarterly, vol. 18, no. 1, pp. 31–38, 1979.
[46] K. Kimura, K. Sawada, and H. Kushima, “Creep rupture ductility of creep strength enhanced ferritic steels,” Journal of Pressure Vessel Technology, vol. 134, no. 3, 2012.
[47] K. Kimura, K. Sawada, and H. Kushima, “Creep deformation, rupture strength, and rupture ductility of grades T/P92 steels,” Pressure Technology, vol. 407-40, pp. 193–201, 2014.
[48] J. F. Wen, S. T. Tu, F. Z. Xuan, X. W. Zhang, and X. L. Gao, “Effects of stress level and stress state on creep ductility: evaluation of different models,” Journal of Materials Science and Technology, vol. 32, no. 8, pp. 695–704, 2016.
[49] Nims, “NIMS creep data sheet,” pp. 58–59, 2011.
[50] M. Spindler, The Development of Improved Methods for the Calculation of Creep Damage in Type 316H Steel, EDF Energy Nuclear Generation Ltd, London, UK, 2003.
[51] Astm, Standard Test Method for Measurement of Creep Crack Growth Times and Rates in Metals, ASTM E1457, 2015.
[52] K. M. Nikbin, D. J. Smith, and G. A. Webster, “Prediction of creep crack growth from uniaxial creep data,” Proc R Soc London, Ser A, vol. 396, no. 1810, pp. 183–197, 1984.
[53] K. M. Nikbin, D. J. Smith, and G. A. Webster, “An engineering approach to the prediction of creep crack growth,” Journal of Engineering Materials and Technology, vol. 108, no. 2, pp. 186–191, 1986.
[54] M. Tan, N. Celard, K. Nikbin, and G. Webster, “Comparison of creep crack initiation and growth in four steels tested in HIDA,” International Journal of Pressure Vessels and Piping, vol. 78, no. 11-12, pp. 737–747, 2001.
[55] M. Yatomi, N. O’Dowd, K. Nikbin, and G. Webster, “Theoretical and numerical modelling of creep crack growth in a carbon–manganese steel,” Engineering Fracture Mechanics, vol. 73, no. 9, pp. 1158–1175, 2006.
[56] G. A. Webster and R. A. Ainsworth, High Temperature Component Life Assessment, Chapman and Hall, Springer, London, UK, 1994.
[57] A. Mehmanparast, C. M. Davies, D. W. Dean, and K. Nikbin, “Material pre-conditioning effects on the creep behaviour of 316H stainless steel,” International Journal of Pressure Vessels and Piping, vol. 108-109, pp. 88–93, 2013.
[58] A. D. Bettinson, The Influence of Constraint on the Creep Crack Growth of 316H Stainless Steel, Imperial College London (University of London), London, UK, 2002.
[59] C. M. Davies, Crack Initiation and Growth at Elevated Temperatures in Engineering Steels, Imperial College London (University of London), London, UK, 2006.
[60] D. Dean and D. Gladwin, “Creep crack growth behaviour of Type 316H steels and proposed modifications to standard testing and analysis methods,” International Journal of Pressure Vessels and Piping, vol. 84, no. 6, pp. 378–395, 2007.
[61] C. M. Davies, D. W. Dean, M. Yatomi, and K. M. Nikbin, “The influence of test duration and geometry on the creep crack initiation and growth behaviour of 316H steel,” Materials Science and Engineering A, vol. 510-511, no. 0, pp. 202–206, 2009.
[62] P. Sun, G. Wang, F. Xuan, S. Tu, and Z. Wang, “Three-dimensional numerical analysis of out-of-plane creep crack-tip constraint in compact tension specimens,” International Journal of Pressure Vessels and Piping, vol. 96-97, pp. 78–89, 2012.