Some Aspects of Mechanisms and Modelling of Creep Behaviour of 2.25Cr–1Mo Steel

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(Received on February 2, 1998; accepted in final form on April 14, 1998)

The present work attempts to establish experimentally the dominant mechanism based model for describing the shape of creep curve at any arbitrary stress and temperature. Creep data were collected from literature and some were generated experimentally under different microstructural conditions of 2.25Cr–1Mo steel. Influence of thermal aging and pre-strain on the shape of creep curve have also been studied.

The analysis of creep data clearly revealed that irrespective of initial microstructures, thermal aging significantly increases the tendency to soften whereas pre-strain results in marginal strengthening. Softening due to carbide coarsening was thus established as the dominant mechanism of creep in 2.25Cr–1Mo steel.

A numerical procedure was developed to extract the parameters of the model, describing the kinetics of carbide coarsening directly from creep curves. A fairly accurate prediction of creep strain up to 5% was obtained using this approach. The kinetics of carbide coarsening was found to depend on the applied stress.

KEY WORDS: 2.25Cr–1Mo steel; mechanism based model; thermal aging; pre-strain; carbide coarsening kinetics; microstructures; softening; stress dependence; temperature dependence; creep curves.

1. Introduction

Low alloy Cr–Mo steels are widely used for high temperature applications in power plants, oil refineries and chemical industries for pipings, superheaters and pressure vessels because of their excellent creep and oxidation resistance. The effect of chromium in ferritic creep resistance steel is complex. By itself chromium gives some enhancement of creep strength, although increasing chromium content in low carbon grades does not increase resistance to deformation at elevated temperature.¹⁾ It is reported that in a 0.5% Mo steel, the presence of chromium up to 2% does not increase the creep resistance, but beyond 2% significant reduction in creep strength has been reported. In contrast, for 1% Mo steel, the optimum creep strength occurs with about 2.25% Cr.²⁾ Therefore, amongst all grades of Cr–Mo steels, 2.25Cr–1Mo steel is most commonly used for a wide range of high temperature applications in both thermal and nuclear power plants.

The conventional approach for the evaluation of high temperature properties of this steel is mainly based on uniaxial stress rupture tests at different stresses and temperatures. The data so generated are analysed using combined time–temperature parameters to estimate long term rupture strength. Some of the frequently used and most popular combined time–temperature parameters are Larson–Miller Parameter, Sherby–Dorn Parameter and Manson–Haferd Parameter. A typical master rupture plot, which describes rupture stress as a function of the Manson–Haferd Parameter of this steel usually

exhibits a wide scatter.³⁾ Life assessment of engineering components using this approach depends on only one point on the entire creep curve. This may not be sufficient keeping in view the ever growing use of Finite Element Method (FEM) in the design of high temperature components. In spite of such an apprehension, conventional design practice for high temperature components is often based on parametric extrapolation of short term stress rupture data primarily because of the following reasons: (a) such tests are easy to perform and less expensive; (b) availability of a fairly large volume of stress rupture database in comparison to the time dependent creep deformation database in the literature; and (c) most design criteria of engineering components are based on creep rupture and not on the accumulation of creep strain.

Increasing use of computer based design technique to conserve both material and energy, demands a more precise description of the entire creep to allow reliable interpolation or extrapolation. The methods being followed to describe the shape of creep curve can be broadly classified as empirical approach and mechanism based approach. The empirical approach uses various creep strain models^{4–8)} to describe creep strain as a function of time, the model parameters being the functions of stress and temperature. Amongst these θ -projection popularised by Evans and Wilshire⁸⁾ has received considerable attention over the last two decades. Following this approach the creep of any material can be described as

$$\varepsilon = \theta_1[1 - \exp(-\theta_2 t)] + \theta_3[\exp(\theta_4 t) - 1] \quad \dots\dots\dots(1)$$

where ε is the time dependent creep strain and t is the time. The parameters θ_1 , θ_2 define primary or decaying components and θ_3 , θ_4 define tertiary or accelerating components of the creep curve. Standard computer softwares are used to extract these parameters from experimental strain (ε)-time (t) data obtained over a range of stress (σ) and temperature (T). It is necessary to express these parameters as a function of stress and temperature so as to predict creep strain at any arbitrary test condition. The function commonly used to represent the parameter θ_i is given by

$$\log \theta_i = a_i + b_i T + c_i \sigma + d_i \sigma T \quad \dots\dots\dots(2)$$

where $i=1, 2, 3, 4$ and a_i , b_i , c_i , d_i are constants.

Obviously sixteen constants are required to be estimated to predict the complete shape of the creep curve. The approach, being empirical only has a limited extrapolation power beyond the domain of available test data. Besides, a large volume of creep strain-time database would be required to estimate the above constants so that reliable predictions could be made at any arbitrary stress and temperature.

In contrast, the mechanism based approach, proposed by the High Temperature Materials Group⁹⁻¹¹⁾ of NPL, UK, using the principles of damage mechanics and physics of deformation, does not suffer from such limitations. The present status of the approach has been reviewed by Dyson and McLean.¹²⁾ The representation of creep behaviour of complex engineering alloys must consider the influence of tertiary creep.¹³⁾ This is primarily because tertiary creep constitutes the major fraction which is more than 95% of the total creep life. 2.25Cr-1Mo steel also exhibits similar creep behaviour.^{14,15)} The damages contributing to tertiary creep in high temperature alloys have been identified by Ashby and Dyson.¹⁰⁾ They classified them into three categories. These are: (a) microstructural (*e.g.* precipitate size and dislocation density) changes responsible for loss of strength of the material (ω_1); (b) loss of external section leading to a change in stress in constant load creep condition (ω_2) and (c) loss of internal section due to cavitation or cracking at grain boundaries (ω_3).

The above damage parameters ω_1 , ω_2 and ω_3 can be defined in number of ways.^{16,17)} When the accumulation of creep strain responsible for continuously increasing creep rate is due to an increase in mobile dislocation density, the effect of ω_1 on ε can be represented either as a linear or as an exponential function of creep strain. In cases where coarsening of precipitates is responsible for the loss of strength, ω_1 should be defined in terms of initial and current threshold stress. When the damage (ω_2) due to loss of external section during constant load creep test is defined as a linear function of strain and the damage (ω_3) due to loss of internal section is defined as a fraction of cross sectional area damaged due to cavitation or cracking, it can be shown that the creep rate is an exponential function of ω_2 and ω_3 .¹⁸⁾ The exact form of equations relating creep strain and damages have been established based on the current understanding

of operative creep mechanism in engineering materials. For example in Superalloys^{11,18)} where strain softening due to increasing mobile dislocation density is the dominant mechanism of deformation, the creep strain accumulation has been represented as

$$\begin{aligned} \dot{\varepsilon} &= \dot{\varepsilon}_i \exp(\omega_1 + \omega_2 + \omega_3) \\ \dot{\omega}_1 &= C_1 \dot{\varepsilon} \\ \dot{\omega}_2 &= C_2 \dot{\varepsilon} \\ \dot{\omega}_3 &= C_3 \dot{\varepsilon} \end{aligned} \quad \dots\dots\dots(3)$$

where $\dot{\varepsilon}_i$, C_1 , C_2 and C_3 are the functions of stress and temperature.

Similarly in cases where precipitate coarsening is the main mechanism of softening, accumulation of creep strain¹⁹⁾ is given by the following:

$$\begin{aligned} \dot{\varepsilon} &= \dot{\varepsilon}_i (1 + \omega_1)^n \exp(\omega_2 + \omega_3) \\ \dot{\omega}_1 &= a(1 - b\omega_1)^4 \\ \dot{\omega}_2 &= C_2 \dot{\varepsilon} \\ \dot{\omega}_3 &= C_3 \dot{\varepsilon} \end{aligned} \quad \dots\dots\dots(4)$$

Table 1. Chemical composition of the steel and its ASTM specification. (wt%)

Steel	C	Mn	Si	Cr	Mo	S	P
Present steel	0.14	0.46	0.28	2.33	0.98	0.01	0.03
ASTM	0.15	0.30–	0.50	1.90–	0.87–	0.05	0.03
A213T22	max	0.60	max	2.60	1.13	max	max

Fig. 1. Microstructures of 2.25Cr-1Mo steel.
(a) Dispersed ferritic grains in tempered bainitic matrix,
(b) Fully tempered bainite and
(c) Dispersed carbides in ferritic matrix.

where a and b are the stress and temperature dependent parameters representing the kinetics of precipitates coarsening and n is the stress exponent.

The present work attempts to establish experimentally the dominant mechanism of creep in 2.25Cr-1Mo steel in order to find out which of the above mechanism based models best describe the shape of its creep curve at any arbitrary stress and temperature.

2. Experimental Procedure

2.1. Material

Forged bars of 25 mm square section made from a commercial grade of 2.25Cr-1Mo steel were selected for this investigation. The chemical composition of the steel along with permissible range as per specification ASTM A213 T22 are given in Table 1.

2.2. Heat-treatments and Microstructures

The bars were heat-treated by selecting the following normalizing and tempering cycles in order to develop a range of microstructures in the steel:

(a) normalising at 920°C for 1 h followed by tempering at 730°C for 3 h.

(b) normalising at 990°C for 1 h followed by forced air cooling and tempering at 730°C for 3 h.

(c) normalising and tempering cycles as in (b) followed by further aging at 650°C for 190 h.

The microstructures thus developed in the steel following the above heat-treatment cycles are (a) dispersion of ferritic grains in tempered bainitic matrix; (b) fully tempered bainite and (c) dispersed carbides in ferritic matrix respectively. These are presented in Figs. 1(a) to 1(c).

2.3. Creep Tests

Constant load creep tests were performed in air on specimens having the above microstructures in Mayes creep testing machines at different temperatures. The specimens were made from the heat-treated bars to measure creep strain till rupture.¹⁴⁾ A few specimens were prestrained at room temperature (25°C) following constant displacement rate tensile testing upto a pre-determined strain level. The initial stress levels at each

Fig. 2. Creep curves of 2.25Cr-1Mo steel having different initial microstructure: (a) 550°C, 150 MPa and (b) 600°C, 80 MPa.

Fig. 3. TEM micrographs of fully tempered bainitic 2.25Cr-1Mo steel.

(a) Before creep exposure and
(b) After creep exposure at 550°C, 150 MPa.

temperature in the range of 500 to 600°C were so selected as to obtain rupture within a reasonable span of time. During the entire span of creep tests the temperatures were maintained within $\pm 2^\circ\text{C}$. The creep strain was monitored till rupture using LVDT with an accuracy of $\pm 0.001\%$. Selected creep tests were additionally performed on the thermally aged and pre-strained specimens to establish the dominant mechanism of creep deformation in this steel.

3. Results and Discussion

3.1. Dominant Mechanism of Creep

Unlike pure metals or stable single phase alloys, Cr-Mo steel exhibits continuously increasing creep rate over most part of its life. In order to study the dependence of such creep rate on initial microstructures, stress and temperature, extensive creep data of 2.25Cr-1Mo steel have been generated. A few typical creep curves of the steel, having different initial microstructures, at (a) 550°C, 150 MPa and (b) 600°C, 80 MPa are shown in Figs. 2(a), 2(b). The shape of the curves clearly demonstrates the evidence of tertiary creep *i.e.* continuously increasing creep rate over a major fraction of life, irrespective of initial microstructures and test parameters.

Microstructural examination of the steel, before and after creep exposure, using TEM, as shown in Figs. 3(a), 3(b) indicate that the carbides initially present in the fully tempered bainitic structure coarsen significantly during creep. Softening due to coarsening of carbides may, therefore, appear to be the obvious mechanism respon-

Fig. 4. Influence of thermal exposure and pre-strain on the shape of creep curves of fully tempered bainitic 2.25Cr-1Mo steel.

Fig. 5. Influence of pre-strain on the shape of creep curve of thermally exposed 2.25Cr-1Mo steel having ferrite-carbide structure.

sible for continuously increasing creep rate. However, in case of superalloys although particle coarsening is evident from microstructural examination, softening due to increasing mobile dislocation density was identified as the dominant mechanism for tertiary creep.^{19,20)} The need for conducting selected experiments was thus felt to study the influence of thermal exposure and pre-strain on the shape of creep curve to establish whether time softening or strain softening is the dominant mechanism of creep.

The influence of thermal exposure on the shape of creep curve of steel having fully tempered bainitic structure is shown in Fig. 4. The rate of accumulation of creep strain after thermal exposure when the microstructure transforms to dispersed carbides in ferritic matrix is significantly higher. The influence of pre-strain on the shape of creep curve is also shown in Fig. 4 for the purpose of comparison. Besides, the influence of pre-strain on the shape of creep curve of thermally exposed steel having ferrite-carbide structure have been studied and presented in Fig. 5. It is evident from the results that irrespective of initial microstructure of the steel, pre-strain improves the resistance to creep deformation to some extent. Hence overall softening in the steel due to pre-strain is almost insignificant. Carbide

coarsening on the other hand has a greater influence on the tertiary creep behaviour of this steel. Overall softening due to carbide coarsening thus appears to be the dominant mechanism of creep.

3.2. Mechanism Based Creep Strain Prediction

Having experimentally established the overall softening due to carbide coarsening to be the dominant mechanism of tertiary creep in 2.25Cr-1Mo steel, a mechanism based approach for creep strain prediction has been developed. In case of 2.25Cr-1Mo steel a mechanism based model as represented by equation set (4) could be used for predicting creep behaviour. Since microstructural examination did not reveal any sign of cavitation even after creep exposure of thermally aged and prestrained specimens, the contribution of cavitation towards strain accumulation can be neglected. In case of constant load creep test the change in specimen cross-section contributes to strain accumulation. Considering all these effects the equation set (4) can be reduced to a set of two coupled differential equations to represent the creep behaviour of 2.25Cr-1Mo steel.

$$\dot{\epsilon} = \dot{\epsilon}_i (1 + \omega)^n \exp(n\epsilon) \dots\dots\dots (5)$$

$$\dot{\omega} = a(1 - b\omega)^4$$

It is, therefore, obvious that a set of four material parameters *viz.* initial minimum creep rate ($\dot{\epsilon}_i$); parameters *a* and *b* representing the extent of softening due to carbide coarsening and the stress exponent (*n*) are required to characterize the complete shape of a creep curve. While standard methods for estimation of the parameters $\dot{\epsilon}_i$ and *n* are available in literature, there was no mention as to how *a* and *b* could be estimated. Keeping this in view, a new method for estimating these parameters and establishing their stress and temperature dependence directly from creep curves^{14,21)} has been developed.

3.3. Analysis of Creep Curves

Creep curves of 2.25Cr-1Mo steel¹⁵⁾ over a range of stress and temperature were analysed using this method. Analysis of data²¹⁾ revealed that the estimation of eight material constants, shown in **Table 2**, representing the stress and temperature dependence of $\dot{\epsilon}_i$, *a* and *b* are sufficient to predict creep behaviour of this steel at any arbitrary stress and temperature.

Using these constants, creep curves have been predicted over a range of stress and temperature and com-

pared with experimental data so as to examine the validity of such a mechanism based approach. **Figure 6** shows the comparison of predicted creep curves at 550°C with experimental results reported by Wolf.¹⁵⁾ These reveal that although the nature of predicted curves are very similar to the experimental curve, the magnitude of predicted strain compare fairly well with experimental values up to about 5% strain. A broader comparison between the predicted and experimental time needed to

Fig. 6. Comparison of predicted creep curves with experimental data reported by Wolf.¹⁵⁾

Fig. 7. Comparison of predicted and experimental time to achieve 5% creep.

Fig. 8. Comparison of predicted creep curve at 550°C, 108 MPa with experimental data reported by Maruyama *et al.*²²⁾

Table 2. Estimated material constants for 2.25Cr-1Mo steel.

Constants	Magnitudes
<i>n</i>	4.23
<i>Q</i>	315 K J mol ⁻¹
<i>ε_{io}</i>	3.2 × 10 ⁻⁶ /h
<i>C</i>	6.52 × 10 ⁻⁶ MPa
<i>Q₁</i>	284.7 K J mol ⁻¹
<i>Q₂</i>	-107.03 K J mol ⁻¹
<i>K₀</i>	5.9 × 10 ⁻¹⁴ m ³ h ⁻¹
<i>m</i>	3.957

Fig. 9. Comparison of predicted creep curve at 600°C, 100 MPa with experimental data for different initial microstructures.

achieve 5% strain over a range of stress (20–220 MPa) and temperature (500–600°C), as shown in Fig. 7, indicates that the agreement is quite satisfactory keeping in view the wide scatter associated with creep data. The above comparison have been made with the set of strain-time plots from which the material constants were estimated. One may, therefore, argue that the prediction should be fairly close. Keeping this in view the predicted creep curve for a specific test condition, as shown in Fig. 8, has also been compared with experimental data selected from a different source.²²⁾ The experimental values are indeed very close to the theoretical predictions. Satisfactory prediction in this situation is certainly of greater significance.

Figure 9 presents typical comparison of predicted creep curve with the set of experimental data at 600°C, under different microstructural conditions of the steel. Here as well the predictions are found to lie close to the experimental data of the steel having ferrite-bainite microstructure. This is mainly because the material constants used for such predictions were estimated from the reported creep data¹⁵⁾ of the steel having similar microstructure. It is, therefore, necessary to obtain material constants of service exposed material so that the approach can be extended for estimation of their lives. Work in this direction is in progress.

4. Conclusions

(1) The coarsening of carbides has been experimentally established as the dominant mechanism of creep deformation in 2.25Cr–1Mo steel.

(2) A method has been developed which can estimate the parameters representing softening due to carbide coarsening and their stress-temperature dependence directly from the analysis of creep curve.

(3) A fairly accurate prediction of creep curve up to about 5% strain can be made following this method.

(4) Only eight material constants representing stress and temperature dependence of the parameters of carbide coarsening model can adequately simulate creep behaviour of 2.25Cr–1Mo steel at any arbitrary stress and

temperature.

(5) Capability of mechanism based prediction can however be further improved by using the material constants estimated from the creep curves of steel having similar microstructure.

Acknowledgments

The authors are grateful to Prof. P. Ramachandra Rao, Director, National Metallurgical Laboratory for his kind permission to publish this paper. They wish to thank Ms. B. K. Verma for her help in the preparation of the manuscript of this paper.

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