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The importance of creep strain in linking together the Wilshire equations for minimum creep rates and times to various strains (including the rupture strain): An illustration using 1CrMoV rotor steel

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Abstract This paper highlights the observation that the Wilshire equations for failure times and times to various strains, as reported in the original literature, may not be the most appropriate ones for all materials – including the one selected in this study. Further, such appropriateness can be determined by looking at the consistencies between the parameter estimates obtained using minimum creep rates in comparison to using failure times. It is shown, using 1CrMoV steel as an illustration, that parameter consistency can be achieved by generalising the Monkman – Grant relation so that it contains a temperature correction. Indeed, the ability of the Wilshire equations to produce meaningful physical parameters, such as the activation energy, is shown to be highly dependent upon a valid specification for the Monkman-Grant relation. It is shown that variations in the measured values for some of the Wilshire parameters (w and k_3) with strain indicate that the causes of deformation are different at different strains and different stresses. Finally, the measured variations in the parameters of the Monkman- Grant relation with strain enable accurate interpolated and extrapolated creep curves to be calculated for any test condition.

Keywords: Wilshire equations, creep, strain, prediction

Introduction

The UK faces a looming energy gap with around 20% of its generating capacity due for closure in the next 10 to 15 years as a result of plant age and new European legislation on environmental protection and safety at work. A number of solutions exist for this problem including the use of new materials so that new plants can operate at higher temperatures, new technologies related to carbon capture and gasification, development of renewable resources and less obviously the use of accurate models for predicting creep life. Such models could speed up the time required for new experimental alloys to be considered as safe for use over the design lives proposed for new A-USC or CCGT power plants.

Without parametric, numerical and computational methods for the accurate extrapolation of short-term property measurements (obtained from 1 or less years of testing), reliance must be placed on very protracted and expensive test programmes lasting 12 - 15 years to determine how long new materials will last at the operating conditions proposed for future A-USC power plants. For example, much of the research carried out under the COST [1] programme has involved carrying out tests lasting over 85,000 hours for multiple batches of the new steel alloys. It is therefore of little surprise to note that a reduction in this 12 - 15 year materials development cycle has therefore been defined as the No.1 priority in the 2007 UK Energy Materials – Strategic Research [2]. Such a reduction is quite urgent within the UK given the number of power plants that are going to need replacing within the current 15 year development cycle.

An important step in achieving reliable life time predictions from short term data is the recent arrival in the literature of the Wilshire equations [3]. It has been shown, through applications of these equation to numerous metal alloys used in power generation [4-7], that very accurate long term predictions can be made from tests of durations not exceeding 5,000 hours. The aims of this paper are to demonstrate that

i. The Wilshire equations for failure times and times to various strains, as reported in the original literature, may not be the most appropriate ones for all materials – including the one selected in this study. Further, such appropriateness can be determined by looking at the consistencies between the parameter estimates obtained using minimum creep rates in comparison to using failure times. Achieving an appropriate specification can be achieved through a generalisation of the Monkman – Grant relation [8] that contains a temperature correction. Indeed, for the material selected in this study, the ability of the Wilshire equations to produce meaningful physical parameters, such as the activation energy, is highly dependent upon such a generalisation.

ii. Variations in the Wilshire parameters with strain can provide information on the mechanisms generating this strain. Without detailed micro - structural analysis however, such information should be seen as speculative and suggestive of future research.

Experimental data

The batch of material used for the present investigation represents the lower bound creep strength properties anticipated for 1CrMoV rotor steels. (When looking at multi batch data sets on this material – see for example that published by the National Institute for Materials Science (NIMS) [15] – this data batch has creep lives consistent with NIMS batches that fail the quickest at all the stress/temperature combinations studied). The chemical composition of this batch of material (in wt %) was determined as 0.27%C, 0.22%Si, 0.77%Mn, 0.008%S, 0.015%P, 0.97%Cr, 0.76%Ni, 0.85%Mo, 0.39%V, 0.125%Cu, 0.008%Al and 0.017%Sn. Following oil quenching from 1238K and tempering at 973K, the material had a tensile strength of 741 MPa, elongation of 17%, reduction in area of 55% and a 0.2% proof stress of 618 MPa. This material had a tensile strength of 560 MPa, 505 MPa and 445 MPa at 783K, 823K and 863K respectively. In addition the 0.2% proof stress for this material was 460 MPa, 425 MPa and 360 MPa at 783K, 823K and 863K respectively

Eighteen test pieces, with a gauge length of 25.4mm and a diameter of 3.8mm, were tested in tension over a range of stresses at 783K, 823K and 863K using high precision constant-stress machines [9]. At 783K, six specimens were placed on test over the stress range 425 MPa to 290 MPa, at 823K seven specimens were placed on test over the stress range 335 MPa to 230 MPa and at 863K six specimens were tested over the stress range 250 MPa to 165 MPa. Up to 400 creep strain/time readings were taken during each of these tests. This data set was first published by Evans et. al. [10]. In addition to this **accelerated** test data, some long-term property data was supplied independently by an industrial consortium involving GEC-Alsthom, Babcocks Energy, National Power, PowerGen and Nuclear Electric. These long-term properties came from the same batch of material used in the accelerated test programme described above but for specimens with gauge lengths of 125mm and diameters of 14mm that were subjected to tests on high sensitivity constant-load tensile creep machines. This longer term data was only available at a temperature of 823K.

Minimum creep rate analysis

Power law expressions

Over the last half century, the creep and creep rupture properties of metals and alloys have been analysed through the dependency of the minimum creep rate $\dot{\varepsilon}_m$ on stress and temperature, usually using a multiplicative combination of the power law and Arrhenius equations [11,12] which can be combined into the form

$$\dot{\varepsilon}_m = A\sigma^n \exp\left(-Q_c/RT\right) \qquad (1)$$

where R = 8.314$Jmol^{-1}K$, T is the absolute temperature and σ is the stress. As generally found for metals and alloys, the parameter (A), the stress exponent (n) and the apparent activation energy for creep (Q_c) vary as the test conditions are altered. Thus, for 1Cr1Mo1V steel, Fig.1a reveals that a decrease from $n \cong 12$ to $n \cong 9.5$ occurs with increasing temperature at stresses above 170 MPa, whilst a decrease from $n \cong 11$ to $n \cong 5$ with decreasing stress at an unchanging temperature of 823K. Q_c ranged from around 550 to 460 $kJmol^{-1}$ over the stress/temperature conditions covered by the accelerated test data. This makes the prediction of long term creep properties from short term data set impossible using Eq. 1.

The original Wilshire equation

An alternative extrapolation method, termed the Wilshire equations [3], avoids these unpredictable n value variations, while still normalising σ through the ultimate tensile strength, σ_{TS}. In this case, stress and temperature dependencies of $\dot{\varepsilon}_m$ are described as

$$\left(\sigma/\sigma_{TS}\right)=\exp\left\{-k_2\left[\dot{\varepsilon}_m.\exp(Q_c^*/RT)\right]^v\right\} \tag{2}$$

where k_2 and v are unknown model parameters. This equation provides a sigmoidal data presentations such that $\dot{\varepsilon}_m \rightarrow \infty$ as $(\sigma/\sigma_{TS}) \rightarrow 1$, whereas $\dot{\varepsilon}_m \rightarrow 0$ as $(\sigma/\sigma_{TS}) \rightarrow 0$. Over the last 6 years, this Wilshire equation has been applied to many power generating materials where it has been shown to be in very good agreement with the long term experimental data on these materials. For example, the reader is referred to Wilshire and Whittaker [5] for an application toGrade 22 (2.25Cr-1Mo) steels; Wilshire and Scharning [6] for an application to 9-12% Chromium Steels;, Abdullah et. al. [7] for an application to the aerospace alloy Titanium 834 and Whittaker et. al. [13] for an application to Type 316H Stainless Steel.

Estimation of the Wilshire equation for minimum creep rates

Evans [14] recently proposed an estimation strategy based on the ordinary least squares technique for estimating values for v and k_2. Eq. 2 can be written as

$$\ln(\dot{\varepsilon}_m)=-\frac{\ln(k_2)}{v}+\frac{1}{v}\sigma^*-Q_c^*(1/RT) \tag{3a}$$

where $\sigma^* = \ln\{-\ln(\sigma/\sigma_{TS})\}$. A linear least squares regression of $\ln(\dot{\varepsilon}_m)$ on σ^* and 1/RT will yield an estimate of 1/v (and thus v), $-\ln(k_2)/v$ (and thus k_2 given the estimate for v) and $-Q^*_c$ (and thus Q^*_c). These estimates minimise the residual sum of squares defined as the squared difference between all the $\ln(\dot{\varepsilon}_m)$ and all the corresponding predicted values given by the right hand side of Eq. 3a. It was first noted by Wilshire and Scharning [4], using multi batch data on this material (obtained from the National Institute for Materials Science (NIMS) [15]), that the values for v and k_2 appear to change at a specific value for the normalised stress. This estimation procedure therefore needs to jointly estimate these parameters, together with the point at which their values change. This can be achieved using a dummy variable D

$$\ln(\dot{\varepsilon}_m)=-\frac{\ln(k_2)}{v}+\frac{1}{v}\sigma^*-Q_c^*(1/RT)+\Delta[\sigma^*-\sigma^*_{kink}]D \tag{3b}$$

where $\sigma^*_{(kink)}$ is the value for σ^* at which the above described discontinuity occurs, i.e. at which the values for v and k_2 change. D is a dummy variable such that D = 0 when $\sigma^* \geq \sigma^*_{(kink)}$ and D = 1 otherwise. Δ is a further parameter to be estimated. Thus a simple grid search is conducted where by the parameters in Eq. 3b are estimated for all values of $\sigma^*_{(kink)}$ in the range defined by the maximum and minimum values for σ^*. For each value of $\sigma^*_{(kink)}$, Eq. 3b will have a different residual sum of squares associated with it. The estimated values for v, Q^*_c, k_2, Δ and $\sigma^*_{(kink)}$ are then those that produce the smallest residual sum of squares. Eq. 3b implies that below $\sigma^*_{(kink)}$, 1/v changes to 1/v + Δ and $-\ln(k_2)/v$ will change to $-\ln(k_2)/v - \Delta\sigma^*_{(kink)}$ - hence allowing k_2 and v to change at some specific value for the normalised stress.

Using this approach, the following parameter estimates were obtained when Eq. 3b was applied to the accelerated test data shown in Fig.1a (the longer term data points shown as unfilled squares were not used for parameter estimation)

$$\ln(\dot{\varepsilon}_m) = 21.179 - 7.598\sigma^* - 283{,}471(1/RT) + 1.849[\sigma^* - -0.58]D \qquad (3c)$$

These values imply an activation energy of around 284 $kJmol^{-1}$- which is very similar to the values used by Wilshire and Scharning [4] (300 $kJmol^{-1}$) in their study of the NIMS data on this material. (Indeed the 95% confidence interval for Q^*_c in this study is 254 to 314 $kJmol^{-1}$ which contains the values used by Wilshire and Scharning).

Next the values for k_2 and v change at around $\sigma^*_{(kink)} = -0.58$, which implies a change above and below a normalised stress of 0.571. So at 783K this kink occurs at a stress of 320 MPa, and 823K this kink occurs at 290 MPa and at 863K at a stress of 255 MPa. What is remarkable about these three stresses is that they all correspond to about 80% of the 0.2% proof stresses at these three temperatures – which approximately speaking corresponds to the yield stresses at these temperatures. Wilshire and Scharning [4] gave this a rather neat physical interpretation – namely that above the yield stress, creep deformation occurs by the movement of dislocations which are generated within the grains. Below the yield stress dislocation movement occurs only at the grain boundaries or by the movement of pre-existing dislocations.

Below a normalised stress on 0.571, the value for v is estimated at -1/7.598 = - 0.132, whilst above this normalised stress it is estimated at 1/(-7.598 + 1.849) = -0.174. Below a normalised stress on 0.571, the value for k_2 is estimated at exp[-0.132(-21.179)] = 16.239, whilst above this normalised stress it is estimated at exp{-0.174[(-21.179)+1.849(0.58)]} = 47.975. That is,

$$(\sigma/\sigma_{TS}) = \exp\left\{-16.239[\dot{\varepsilon}_m.\exp(284{,}000/RT)]^{-0.132}\right\} \text{ when } (\sigma/\sigma_{TS}) \leq 0.571$$

$$(\sigma/\sigma_{TS}) = \exp\left\{-47.975[\dot{\varepsilon}_m.\exp(284{,}000/RT)]^{-0.174}\right\} \text{ when } (\sigma/\sigma_{TS}) > 0.571$$

which are shown in Fig.1b as the solid kinked line. Recall that this kinked line was estimated from only that data shown as solid symbols in this figure – i.e. the accelerated test data. When this line is extrapolated out to lower normalised stresses (the dashed line in Fig. 1b) it predicts the longer term data extremely well. This is quite a remarkable achievement when it is realised that at 823K the lowest test stress in the accelerated data set was 230 MPa, which is a stress in excess of 10 times that which this material would typically be subjected to at a power plant. The lowest stress associated with the longer term data in Fig. 1b is just 70 MPa – which is only some 2-3 times the stress at which this material would typically be subjected to at a power plant.

Failure time analysis

The Wilshire equation

Wilshire and Battenbough [4] proposed a very similar expression to Eq. 2 for the stress and temperature dependencies of t_f

$$(\sigma/\sigma_{TS}) = \exp\left\{-k_1\left[t_f.\exp(-Q_c^*/RT)\right]^u\right\} \tag{4a}$$

To link this Wilshire expression to that for the minimum creep rate in Eq. 2, use must be made of the Monkman and Grant [8] relation which is an empirical relationship that exist between the time to failure and the minimum creep rate. This relationship is often expressed in the form

$$\dot{\varepsilon}_m t_f = M \tag{4b}$$

where M is a material specific constant. Essentially, the value for M measures what the strain at rupture would have been had the material deformed at the minimum creep rate over its whole life. Monkman and Grant believed M to be independent of the test conditions.

Rearranging Eq. 4b for $\dot{\varepsilon}_m$ and substituting the resulting expression into Eq. 2 gives

$$(\sigma/\sigma_{TS}) = \exp\left\{-k_2 M^v\left[t_f.\exp(-Q_c^*/RT)\right]^{-v}\right\} \tag{4c}$$

Eq. 2 and Eq. 4b imply that the values for k_1 and u in Eq. 4a should equal

$$u = -v \quad ; \quad k_1 = k_2 M^v \tag{4d}$$

The main aim of this paper is to highlight the fact that the form of Eq. 4a may not be the most appropriate one for all materials – including the one selected for this study. Further, the appropriateness of Eq. 4a can be determined by looking at the consistencies between the estimates made for the parameters in Eq. 2 and Eq. 4a.

Estimation of the Wilshire equation given by Eq. 4a

Eq. 4a can again be linearized as follows

$$\ln(t_f) = \frac{-\ln(k_1)}{u} + \frac{1}{u}\sigma^* + Q_c^*(1/RT) \tag{5a}$$

and allowing for a break in the relation

$$\ln(t_f) = \frac{-\ln(k_1)}{u} + \frac{1}{u}\sigma^* + Q_c^*(1/RT) + \Delta[\sigma^* - \sigma_{kink}^*]D \tag{5b}$$

When Eq. 5b was estimated by applying least squares to the accelerated test data **only** the following estimates were obtained

$$\ln(t_f) = -18.949 + 7.524\sigma^* + 248{,}203(1/RT) - 2.144[\sigma^* - -0.58]D \tag{5c}$$

Eq. 5a together with the estimates shown in Eq. 5c are shown visually in Fig.2. As should be expected, the kink point associated with the failure times is the same as that associated with the minimum creep rates – i.e. at a normalised stress of 0.571. Similar consistencies also exist in some the estimates of the other parameters in the Wilshire equations. For example, below a normalised stress on 0.571, the value for u is estimated at 1/7.524 = 0.133, whilst above this normalised stress it is estimated at 1/(7.524- 2.144) = 0.186. But recall that Eq. 4d suggests that u should equal -v. So below a normalised stress on 0.571, the data is in good agreement with Eq. 4d and in broad agreement above the normalised stress.

Next consider the estimated activation energy. When using minimum creep rate data this is estimated at approximately 284 $kJmol^{-1}$ (see Eq. 3c), but when using failure time data this is estimated at approximately 248 $kJmol^{-1}$ (see Eq. 5c). This reflects itself in Fig. 2, where the extrapolations to higher failure times are not as good as the extrapolations made for the minimum creep rates in Fig. 1b.

Next consider the value for k_1 in Eq. 4a. Below a normalised stress of 0.571, the value for k_1 is estimated from Eq. 5c to be exp[0.133(18.949)] = 12.411. Above a normalised stress of 0.571, the value for k_1 is estimated from Eq. 5c to be exp[0.133(18.949)] = 42.674. However, Eq. 4d states that k_1 should equal k_2M^v. Further, Fig. 3a reveals M = 0.14 (or 14%). So, below a normalised stress of 0.571 k_2M^v = 21.03, and above this normalised stress k_2M^v = 67.544. These are very different from the direct estimates derived from Eq. 5c.

Clearly then, the Wilshire equation for times to failure (whose original from is shown in Eq. 4a) is not the most appropriate specification at least for this material.

A generalisation

A clue to a more appropriate specification of Eq. 4a for this material is evident in Fig. 3a, where the exponent on the minimum creep rate in the Monkman – Grant relation is less than unity. That is Eq. 4b should more generally be written as

$$\dot{\varepsilon}_m^{\rho} t_f = M \tag{6a}$$

where M and ρ are material specific constants. Rearranging Eq. 6a for $\dot{\varepsilon}_m$ and substituting the resulting expression into Eq. 2 gives

$$(\sigma/\sigma_{TS}) = \exp\left\{-k_2 M^{v/\rho}\left[t_f.\exp(-\rho Q_c^*/RT)\right]^{-v/\rho}\right\} \tag{6b}$$

In terms of the original Wilshire expression, it must follow that in Eq. 4a the value for k_1 and u should equal

$$u = -v/\rho \quad ; \quad k_1 = k_2 M^{v/\rho} \tag{6c}$$

and Q^*_c in Eq. 4a should be ρQ^*_c.

With $\rho = 0.945$ in Fig. 3a, the estimate for the activation energy obtained from Eq. 5c should be revised up to 248,203/0.945, or approximately 263 $kJmol^{-1}$. But this is still well below the estimate of 284 $kJmol^{-1}$, obtained in Eq. 3c using the minimum creep rate data. Further, Eq. 6c states that k_1 should equal $k_2M^{v/\rho}$. So below a normalised stress of 0.571, $k_2M^{v/\rho} = 21.354$, and above this normalised stress $k_2M^{v/\rho} = 68.903$. These are still very different from the direct estimates shown in Eq. 5c.

Consistency between the parameter estimates shown in Eqs. 3c and 5c is achieved by realising that for this materials the parameter M in the Monkman – Grant relation is temperature dependent. This possible temperature dependency of M has been known for a considerable length of time, but the strain dependency is not often stated in the literature. For example, Dunand et.al. [16], when looking at dispersion strengthened and particulate reinforced Aluminium, noted that a better fit to the experimental data can be obtained by introducing the strain at failure ε_f into Eq. 4b

$$\frac{\dot{\varepsilon}_{m} t_f}{\varepsilon_f} = M^{/}$$

This relation is shown in Fig. 3b, where the scatter is considerably reduced compared to that in Fig. 3a. (By doing so the value for ρ will be approximately unity). $M^{/}$ is now a measure of the proportion of a materials life that would be used up if it were to creep at a rate of $\dot{\varepsilon}_{\mathbf{m}}$ over its entire life (so again its unit are in %). Alternatively, this equation can be interpreted as saying that the average creep rate over the life of a material is proportional the minimum creep rate, with the coefficient of proportionality being $1/M^{/}$. However, in their analysis, Dunand et.al also found that the estimates made for $M^{/}$ (and thus M) differed depending on the test temperature.

A more general representation of the Monkman – Grant relation, is therefore

$$\dot{\varepsilon}_{m}^{\rho} t_f = Me^{-b\frac{1}{RT}} \tag{7a}$$

Rearranging Eq. 7a for $\dot{\varepsilon}_{m}$ and substituting the resulting expression into Eq. 2 gives

$$(\sigma/\sigma_{TS}) = \exp\left\{-k_2 M^{v/\rho}\left[t_f.\exp([b-\rho]Q_c^*/RT)\right]^{-v/\rho}\right\} \tag{7b}$$

In terms of the original Wilshire expression, it must follow that in Eq. 4a the value for k_1 and u should equal

$$u = -v/\rho \quad ; \quad k_1 = k_2M^{v/\rho} \tag{7c}$$

and Q^*_c in Eq. 4a should be $[b-\rho]Q^*_c$.

Estimation of Eq. 7a using ordinary least squares yielded

$$\dot{\varepsilon}_m^{0.967} t_f = 4.775 e^{-26540\frac{1}{RT}} \tag{7d}$$

With ρ = 0.967 and with b = 26,540 in Eq. 7d, the estimate for the activation energy obtained from Eq. 5c should be revised up to (248,203+26,540)/0.967 = 284,120 $Jmol^{-1}$. This is now in very good agreement with the estimate of 283,471 $Jmol^{-1}$, obtained from Eq. 3c using the minimum creep rate data.

Further, Eq. 7c states that k_1 should equal $k_2M^{v/\rho}$. So with M = 4.775 in Eq. 7d and below a normalised stress of 0.571, $k_2M^{v/\rho}$ = 16.239*4.775^(-0.132/0.967) = 13.126, and above this normalised stress $k_2M^{v/\rho}$. = 36.211. These are now much closer to the direct estimates derived from Eq. 5c – namely 12.411 and 42.674 above and below the normalised stress respectively. It therefore appears that for this material, a more suitable specification for Eq. 4a, is Eqs. 7b,c. However, for materials where M is constant and ρ =1, the original Wilshire failure time equation has no inconsistencies with the minimum creep rate equation and is perfectly valid.

Time to x% strain analysis

The appropriateness of the re specification of the Wilshire equation given by Eq. 7b for failure times in this material is reinforced by looking at the Wilshire equation for times to various strains.

The Wilshire equation

Wilshire and Batenbough [3] has proposed a very similar expression to Eq. 2 for the stress and temperature dependencies of t_ε

$$(\sigma/\sigma_{TS}) = \exp\left\{-k_3\left[t_\varepsilon.\exp(-Q_c^*/RT)\right]^w\right\} \tag{8}$$

where t_ε is the time to reach a strain of ε. This equation has been applied to various materials and more recently to predict the full creep curve shapes – see for example – [7] and [17].

Estimation of the Wilshire equation given by Eq. 8

Eq. 8 can again be linearized as follows

$$\ln(t_\varepsilon) = \frac{-\ln(k_3)}{w} + \frac{1}{w}\sigma^* + Q_c^*(1/RT) \tag{9a}$$

and

$$\ln(t_\varepsilon) = \frac{-\ln(k_3)}{w} + \frac{1}{w}\sigma^* + Q_c^*(1/RT) + \Delta[\sigma^* - \sigma_{kink}^*]D \tag{9b}$$

Fig. 4 shows the results of estimating the parameters in Eq. 9b using least squares at various strains between zero and the rupture strain. Because the rupture strain differs with the test conditions, the strains shown in Fig. 4 are scaled to be in the range zero to unity by dividing the actual stains by the rupture strains. (Eq. 9b was applied to the accelerated test data **only).**

The first point of interest is that unlike in Fig. 1b and Fig. 2, where minimum creep rates and failure times are used, there is no noticeable kink in the relationship below a scaled strain of about 0.1 (i.e. below 10% of the rupture strain). That is, the values for k_3 and w are the same above and below the normalised stress of 0.571, suggesting that the dominant deformation mechanism leading to strains up to 10% of the rupture strain is the same at low and high stresses. Above a scaled strain of 0.1, a kink appears in the relation with two distinct values for w and k_3 above and below a normalised stress of 0.571. Further, as the strain increases, w falls in value below the normalised stress of 0.571 but increases in value above the normalised stress of 0.571.

When considering k_3, it appears that its value below a scaled strain of 0.1 is a continuation of the trend observed in the values for k_3 associated with low normalised stresses above a scaled strain of 0.1. One possible interpretation of this trend in k_3 is therefore that strains less than 10% of the failure strain are caused predominantly by dislocation movement at the grain boundaries and or by the movements of pre-existing dislocations irrespective of the stress level. Then strains above 10% are caused in the same way provided the normalised stress is below 0.571.When the normalised stress is above 0.571, strains in excess of 10% are predominantly caused by the movement of dislocations which are generated within the grains. This is consistent with the observed values for w in Fig 4a, where below a normalised stress of 0.571, the relative value for w is smaller so that smaller increases in stress are required to produce larger decreases in the times to strains above 10% of the rupture strain as dislocations are easier to move along the grain boundaries. It must be emphasised however, that without any additional information on microstructure or information about creep mechanisms, such a conclusion is speculative and could form the content of an important area for future work.

However, like the failure time equation, Eq. 8 is again mis-specified for this material. To be consistent with Eq. 7b as strains approach the rupture strain (and thus as t_ε approaches t_f), Eq. 8 must have the following form

$$(\sigma/\sigma_{TS}) = \exp\left\{-k_2 M_\varepsilon^{v/\rho_\varepsilon}\left[t_\varepsilon.\exp([b-\rho_\varepsilon]Q_c^*/RT)\right]^{-v/\rho_\varepsilon}\right\} \tag{10a}$$

so

$$w = -v/\rho_\varepsilon \qquad \text{and} \qquad k_3 = k_2 M_\varepsilon^{v/\rho_\varepsilon} \tag{10b}$$

and Q^*_c in Eq. 8 should be $[b-\rho_\varepsilon]Q^*_c$.

As ε approaches the rupture strain, t_ε approaches t_f and M_ε and ρ_ε approach M and ρ so that Eq. 10a and Eq. 7b then become equivalent. For this to be so it must follow that b, M_ε and ρ_ε are defined through

$$\dot{\varepsilon}_m^{\rho_\varepsilon} t_\varepsilon = M_\varepsilon e^{-b\frac{1}{RT}} \tag{10c}$$

Eq. 10c implies that for each strain there is a different version of the Monkman – Grant relation, with time to a given strain being linearly related to the minimum creep rate (on a log - log scale). M_ε is the strain that would be observed if creep had occurred at the minimum rate up to time t_ε. Note that ρ is subscripted by the strain to indicate it may, like M, be strain dependent. A partial insight into the validity of Eq. 10c is shown in Fig. 5 that plots the times to various strains against the minimum creep rate on a log- log scale. (No temperature adjustment is shown in this figure). ρ appears to remain fairly constant over the shown range of strains, but M changes quite dramatically with strain. The degree of fit is very good at all strains – as show by the R^2 values.

This complete variation in M_ε (and ρ_ε) with strain is shown in Fig. 6 for all strains up to the rupture strain. At low strains, ρ increases rapidly with strain, but remains fairly constant after a strain of about 10%. (Note how ρ tends to the values of the Monkman – Grant exponent as the rupture strain is approached). The variation in M with strain is more complex. It follows a sigmoidal S shaped pattern up to a strain of about 70 of the rupture strain and then decreases. Again note how M_ε tends to the value for M in Eq. 7d as the rupture strain is approached

Fig. 7 brings all this together to shows the consistency between all the modified Wilshire equations – namely Eq. (4a, 7b and 10a). In this Figure, the estimated values for k_3 in Fig. 4b for normalised stress below 0.571 is plotted alongside that predicted by Eq. 10b using the values for M_ε and ρ_ε shown in Fig. 6. As can be seen there is a broad agreement between the actual and predicted values for k_3.

Given the validity of Eq. 10a, the least squares estimate of the parameter in front of 1/RT in Eq. 9b is actually an estimate of $(b-\rho)Q^*_c$. Fig. 8 shows the values for Q^*_c obtained from iteratively estimating Eq. 9b (and adjusting for the above shown value for b and ρ in Fig. 6) at various different strains. The first point to note is that the activation energy converges in value (as the rupture strain approaches) to that shown in Eq. 3c using minimum creep rates (approximately 284 $kJmol^{-1}$). The activation energy also remains reasonably constant, at this value, down to strains of around 20% of the rupture strain. At very low strains the activation energy appears to fall slightly.

Creep curves

Eqs. 10 imply creep curves of a certain shaped that will also depend on the test conditions. These equations can be expressed as

$$\ln[M_\varepsilon] = -a[\sigma, T] - \frac{\rho_\varepsilon}{v}\ln[k_2] + \ln[t_\varepsilon] \quad \text{where} \quad a[\sigma, T] = \frac{1}{w}\sigma^* + Q^*_c\frac{1}{RT} \tag{11}$$

Fig. 6 reveals that M_ε and ρ_ε are exponential type functions of strain and using this figure, values for M_ε and ρ_ε can be obtained for all the different values for strain, and Eq. 11 then used to find the corresponding times to these strains, i.e. used to find the complete creep curve for a given stress and temperature combination. Such test conditions only influence the intercept term in Eq. 11, but a[σ,T] may vary with strain at the start of the creep curve as Fig. 8 reveals a small change the activation energy with low strains.

Figs. 9 shows how well Eq. 11 traces out the actual experimental creep curves. Fig. 9a show the full experimental creep curve obtained at 823K and 270 MPa. The figure also shows the creep curve predicted by Eq. 11 in conjunction with the values for M_ε and ρ_ε found in Fig. 6 and Q^*_c in Fig. 8. This is an example of an interpolated creep curve as the test condition is within the range of conditions used to estimate the parameters of Eq. 10a. Finally, Fig. 9b shows the early stages of the experimental creep curve obtained at 823K and 140 MPa (test not yet complete at this low stress at time of publication) as well as the curve predicted using Eq. 11 under these conditions. Despite an extrapolation from 230 MPa to 140 MPa, the agreement with the experimental data is very good.

Conclusion

It thus appears that the form of the three Wilshire equations, as reported in the original literature, are not the most suitable ones for this material. For the estimated parameters of the equations containing failure times and times to various strains to be consistent with the parameter estimates in the equation containing minimum creep rates, the re-specifications given by Eq. 7b and Eq. 10a are required – namely a relaxation of the Monkman – Grant relation to allow an exponent below unity and a temperature dependency of M. When this is done, the estimates made for the activation energies in all three equations are consistent in value and the estimates for u, w, k_1 and k_3 can all be related to the estimates for k_2 and v in Eq. 2. Finally, the variations in $M\varepsilon$ and p_ε (and to a lesser extent Q^*_c) with strain shown in Fig. 6 are such that accurate interpolated and extrapolated creep curves are derivable at any test condition. These types of prediction is especially useful when components must be designed for low strain tolerances.

Finally, whilst changes in the values of the parameters in the Wilshire equations with stress and strain can be related to the mechanisms of deformation, further detailed work would be required to collaborate any speculations made on this topic. Future work could also include studies into whether the reformulations of the Wilshire time and strain equations presented in this paper, are more appropriate than the original ones used in the literature to analyse low alloy and high chrome steels.

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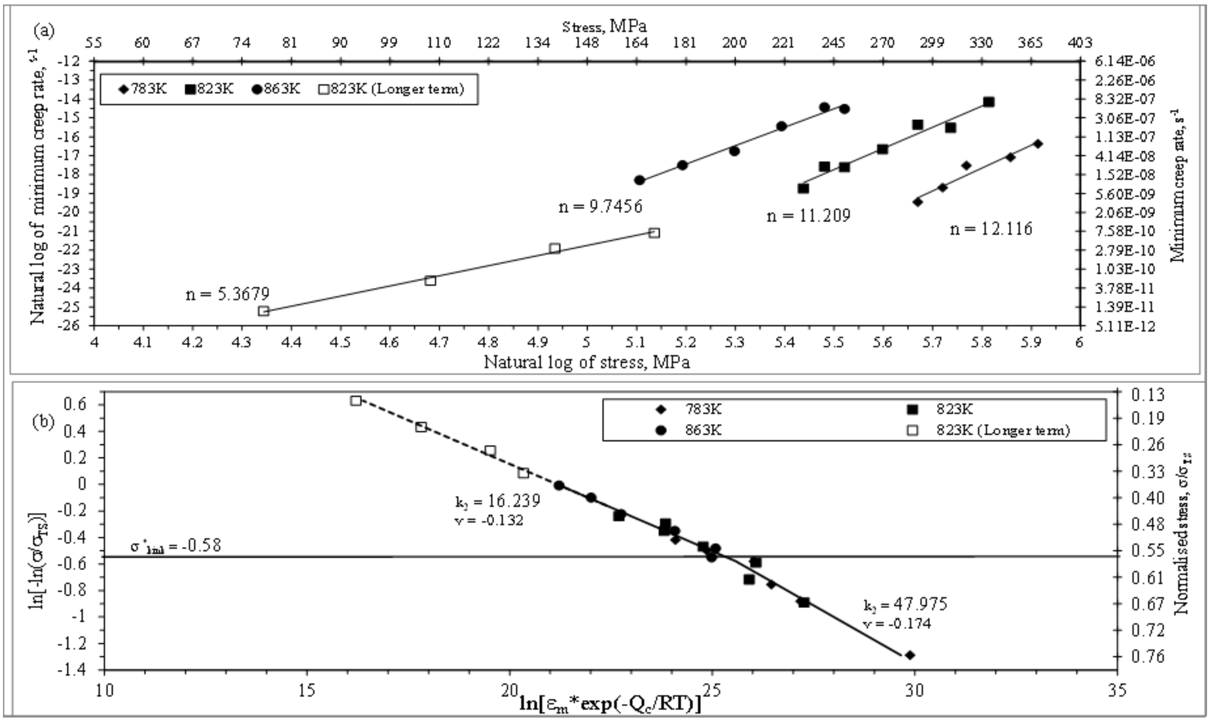

Fig. 1 Behaviour of minimum creep rates for 1Cr-1Mo-0.25V steel at 783 to 863 K. a dependence of ln[$\dot{\varepsilon}_m$] on log stress and temperature and b dependence of ln[$\dot{\varepsilon}_m$ exp(Q^*_c/RT)] on ln[-ln(σ/σ_{TS})] with Q^*_c = 284 kJ mol^{-1}in Eq. 2

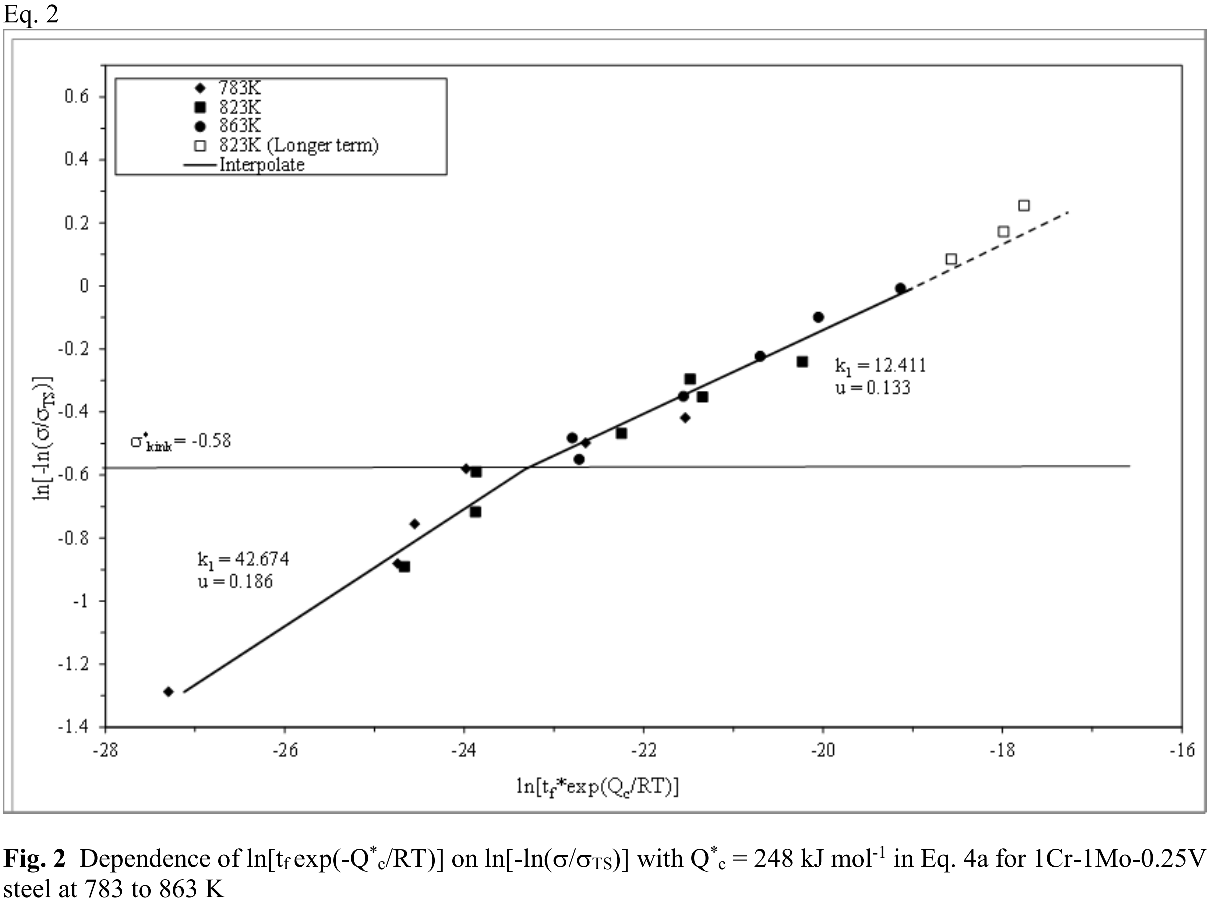

Fig. 2 Dependence of ln[t_f exp(-Q^*_c/RT)] on ln[-ln(σ/σ_{TS})] with Q^*_c = 248 kJ mol^{-1} in Eq. 4a for 1Cr-1Mo-0.25V steel at 783 to 863 K

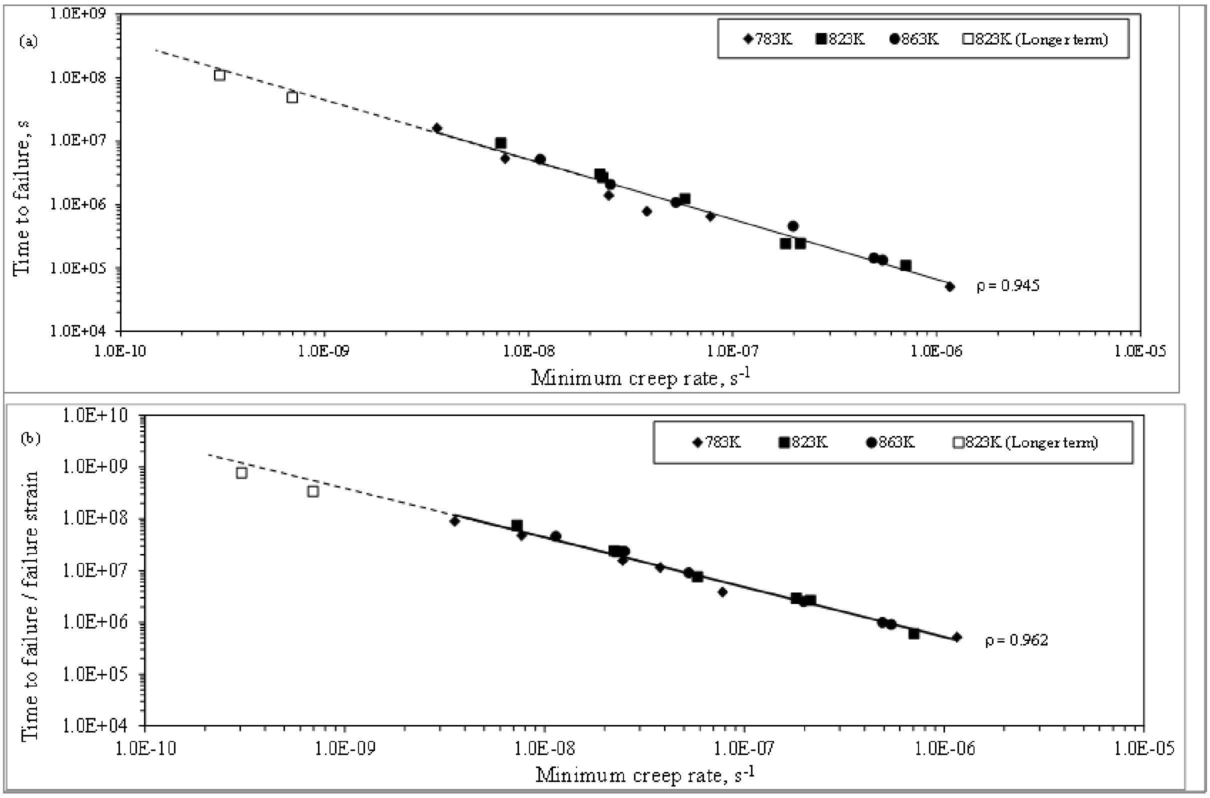

Fig. 3 Monkman – Grant relation. a the dependence of ln failure time on the ln minimum creep rate and b the dependence of ln failure time normalised by the rupture strain on ln minimum creep rate

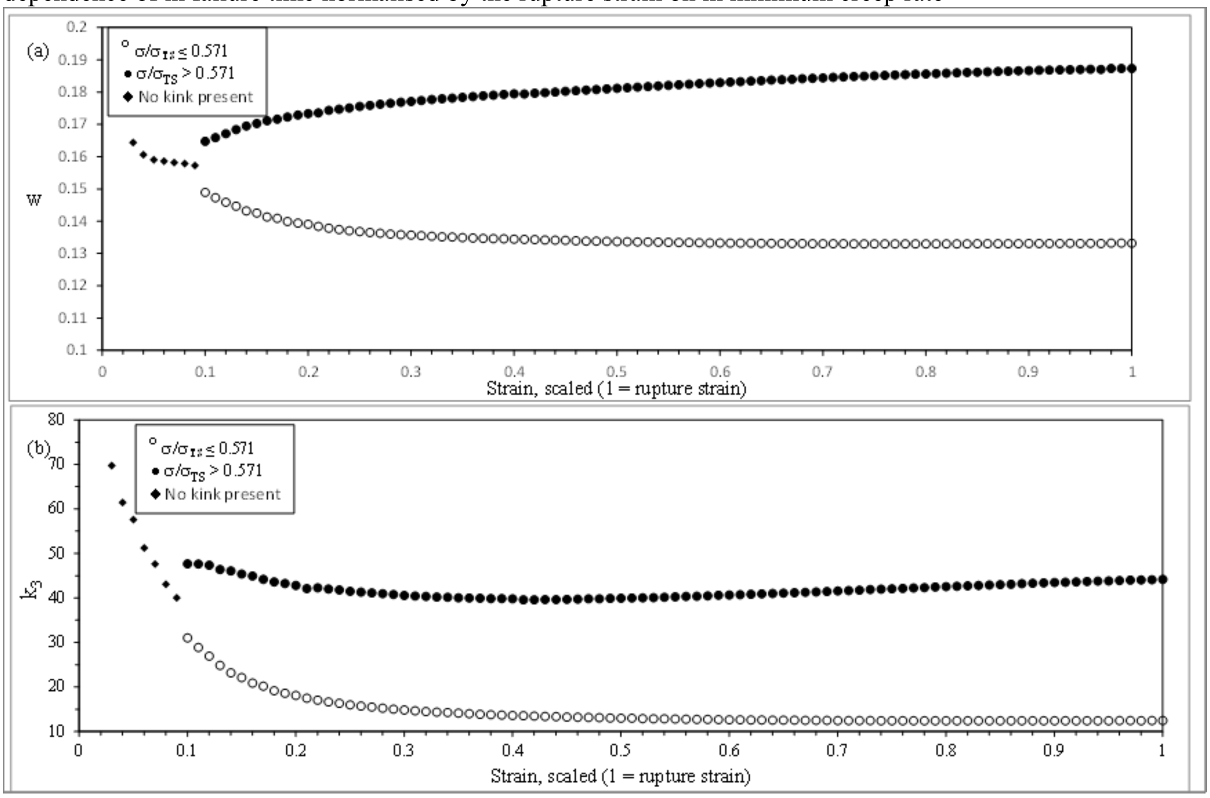

Fig. 4 The dependence of a w in Eq. 8 and b k_3 in Eq. 8 on strain (normalised by the rupture strain) and stress

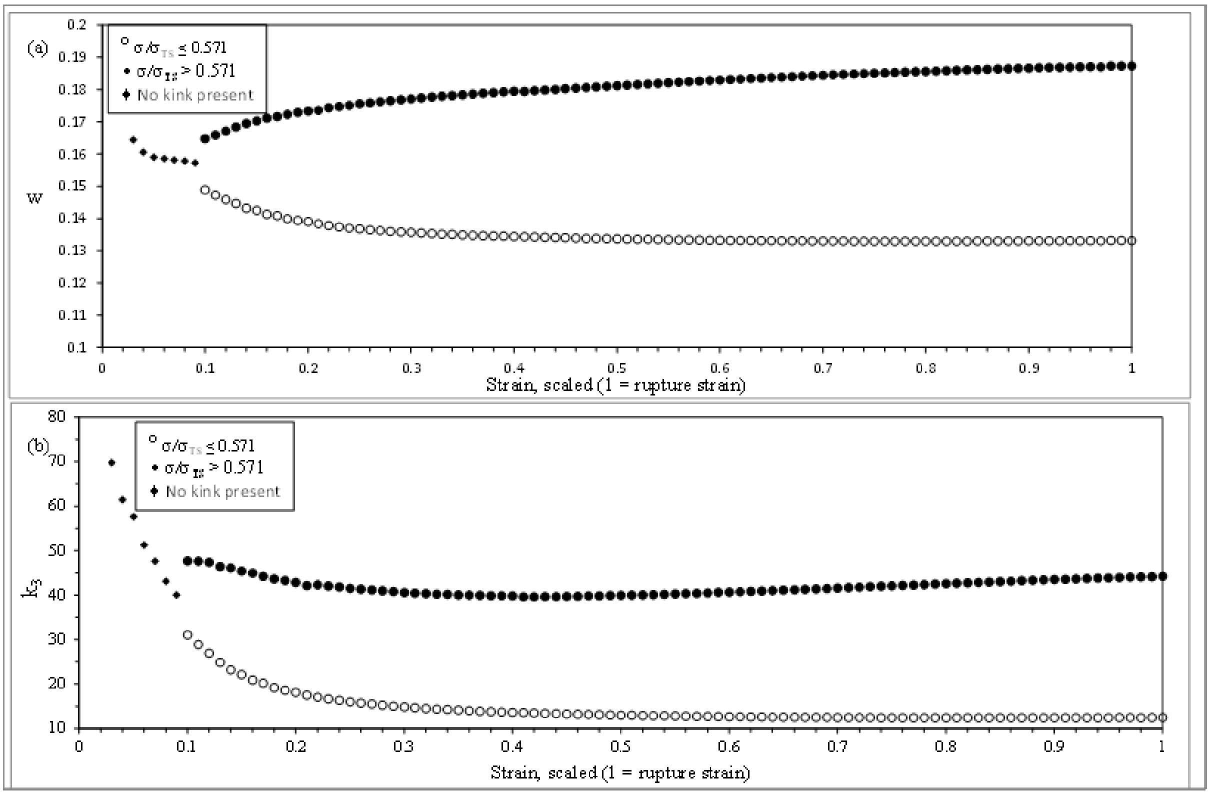

Fig. 5 The dependence of log times to various strains on the log minimum creep rates

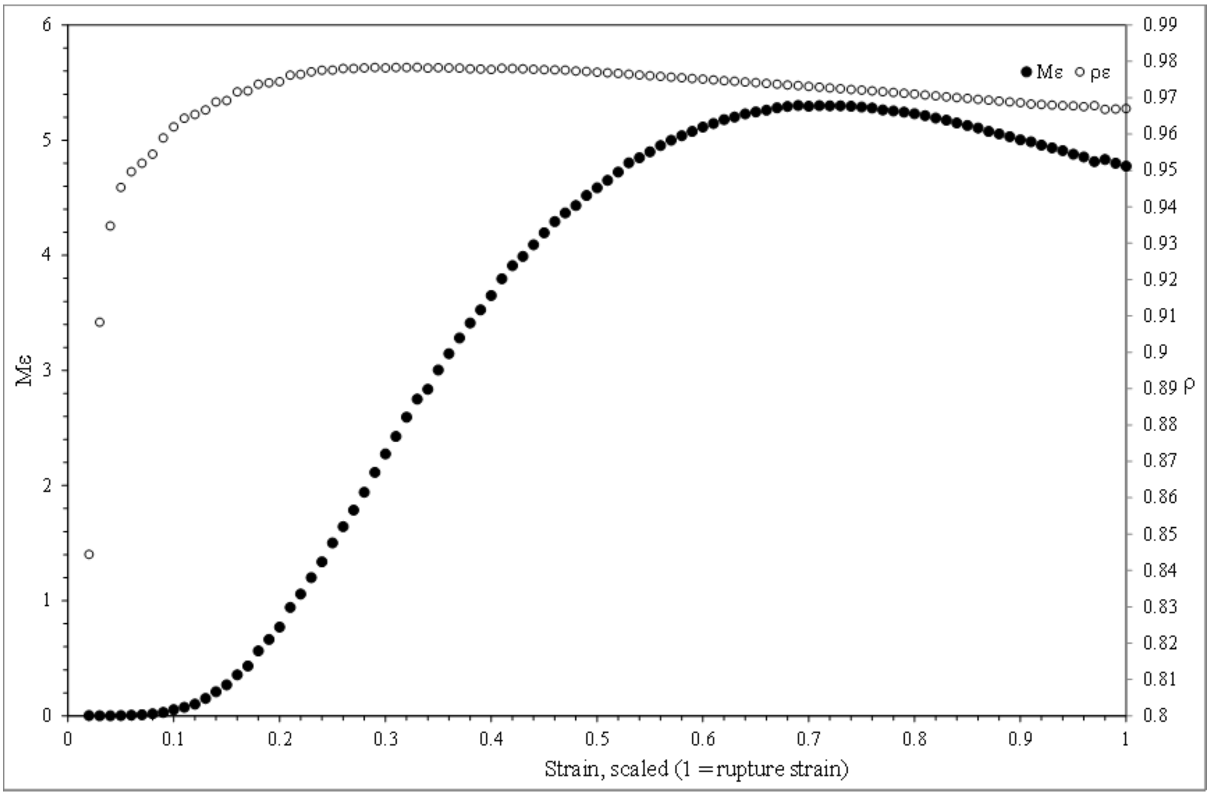

Fig. 6 The dependence of M_ε and ρ_ε in Eq. 10c on strain (normalised by the rupture strain)

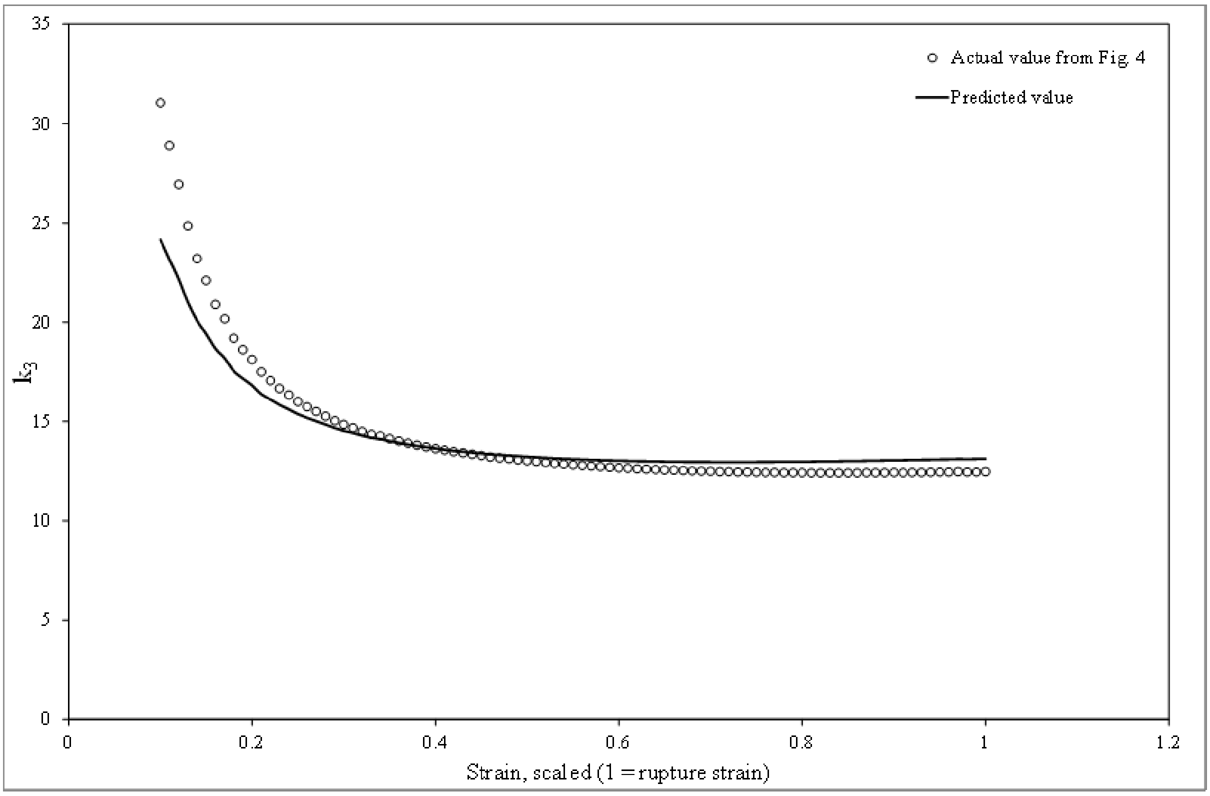

Fig. 7 The variation in k_3 with strain: obtained a from estimating Eq. 8 and b that predicted from Eq. 10c

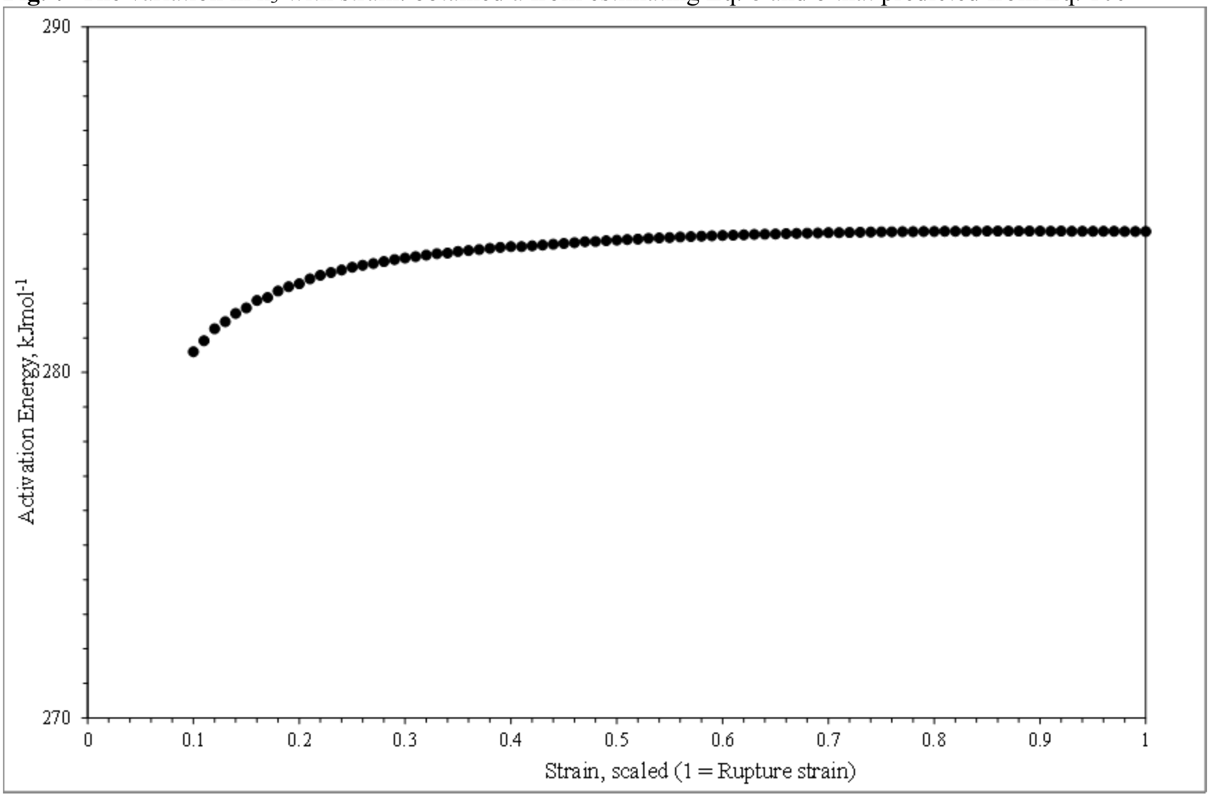

Fig. 8 Variation in the estimated activation energy with strain (normalised by the rupture strain)

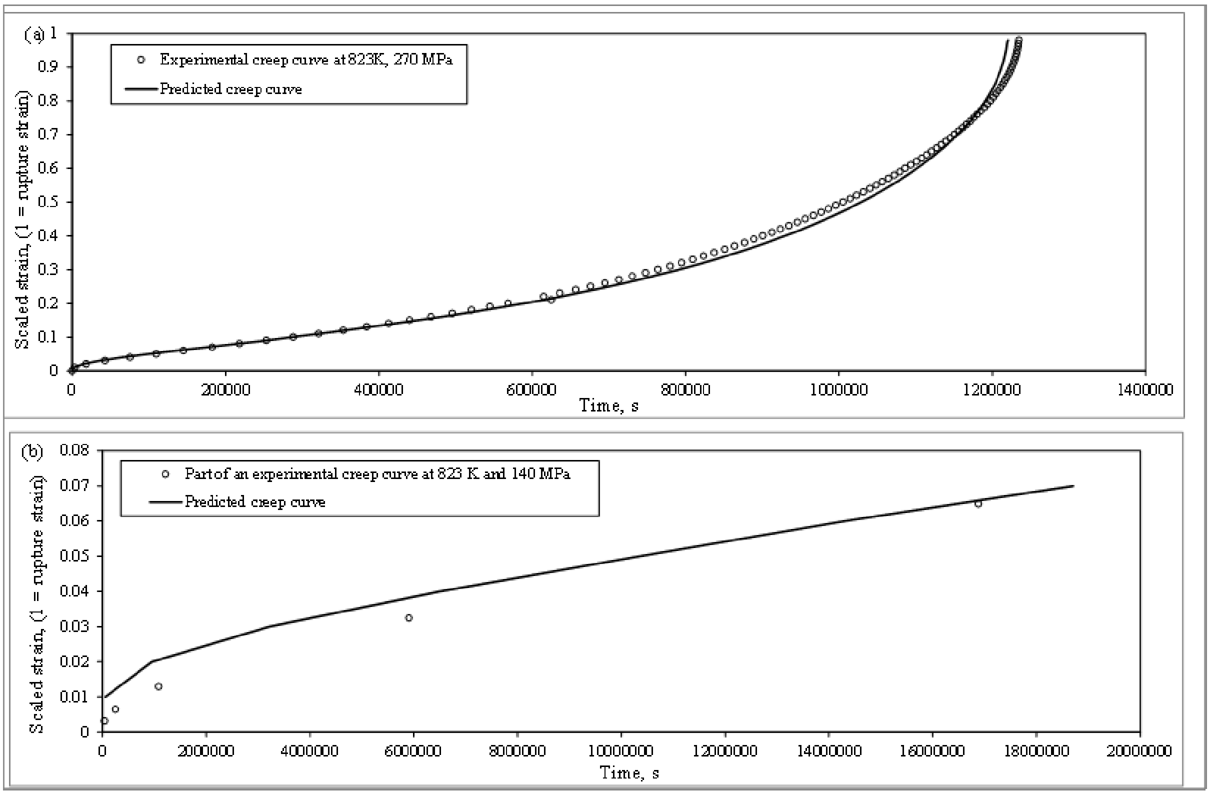

Fig. 9 Experimental and predicted (using Eq. 11) creep curves at a 823 K and 270MPa and b 823 K and 40 MPa