

A Unified Multiscale Ductility Exhaustion Based Approach to Predict Uniaxial, Multiaxial Creep Rupture and Crack Growth

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1. Abstract

Numerical and analytical methods for predicting uniaxial damage have largely depended on the constituent components of the stress/strain measured data which have inherent scatter. Models developed for this purpose have also attempted, with some degree of success, to address the fundamental issues of failure mechanisms within a multiaxial stress state context. This paper presents a new analytical/empirical/a postpriori unifying approach to predict creep damage and rupture under uniaxial/multiaxial and crack growth conditions by deriving a multiscale based constraint criterion. Essentially, the model links the global constraint due to geometry in a globally isotropic materials with a microstructural constraint arising from creep diffusional processes occurring in a sub-grain anisotropic microstructure. Furthermore, it is shown that the model is consistent with the established NSW crack growth model [1-3] which is routinely used to determine the plane stress/strain bounds for cracking rates in fracture mechanics geometries and cracked components. The concept assumes that at very short times an initial upper shelf material tensile strength and global plasticity and power law creep control creep damage failure and sub grain multiaxial axial stress state dependent failure strain dominates the long term diffusion/dislocation controlled creep response. It is established that the material yield strength in the short term and a measure of creep failure strain at the creep secondary/tertiary transition region described at the limits by the Monkman-Grant failure strain [4], are the important variables in both the uniaxial and multiaxial failure processes. For verification creep constitutive properties from long term data from uniaxial and multiaxial and crack growth tests on Grade P91/92 martensitic steels from various databases [5,6], are used to establish the procedure.

Keywords: uniaxial, multiaxial, creep, damage, cracks, fracture mechanics, constraint, ductility, multiscale

1.1. Nomenclature

MG	Monkman-Grant strain
MSF	Multiaxial Strain Factor
NSW	Nikbin, Smith, Webster Model
σ_l, σ_m	principal, mean stresses
$\sigma_e \equiv \bar{\sigma}$	von Mises stress (MPa)
σ_{eff}	effective stress (MPa)
t, t_r, t_i	time, time to rupture (hours), time to crack initiation
σ_t	failure stress at time t
$h = (\sigma_m / \sigma_e)$	constraint parameter
$h_o = 0.33$	value of h at plane stress
$h_o^* = h_o / h$	normalised h from plane stress h_o
α	multiaxial stress state parameter
A, n	The Norton's creep constant and creep index
$e^{-Q/RT}$	the temperature activation terms
β, B, b', A', A''	Material constants in the relevant equations
ϵ_{MG}	MG failure strain, (/1)
$\epsilon_f, \epsilon_{RA}$	Elongation (EL) failure strain and reduction in area (RA), (/1)
ϵ_f^*, ϵ_f	multiaxial failure strain and the uniaxial failure strain
ϵ_{fMG}^* and ϵ_{fMG}	MG multiaxial failure strain and the MG uniaxial failure strain
ϵ_o^* (MSF)	Multiaxial Strain Factor, $\epsilon_o^* = (\epsilon_{fMG}^* / \epsilon_{fMG})$
$a, \Delta a, da$	crack length (mm), crack initiation length, crack length increment
$\dot{a} = da/dt$	crack growth rate (mm/h)
$\dot{a}_i, \dot{a}_s$	Initial and steady state crack growth rate (mm/h)
$B,$	sample thickness
U^*	the potential energy rate
C^*	creep crack growth rate parameter integral (MJ/m ² h)
I_n, r_C, A_{rc}	stress exponent, creep process zone, Creep
$\phi = n/(n + 1)$	creep crack growth rate constant as a function of creep index
D	creep crack growth rate constant
dr	element ahead of the crack tip
MN/UB/LB	Mean/Upper/Lower Bounds

2. Background to Creep Damage Modelling

Life cycle for validated material characterisation used in high temperature plant could take upto 10-15 years. Understanding damage and quantifying it to predict creep rupture and crack initiation and growth from mainly accelerated tests has been the subject of many years of research [9-14]. This has been with the chief aim of reducing the production and verification time cycle of new alloys. There are numerous creep rupture ductility models which suggest mechanism change and base their assessment on parametric power law, multi-parameter methods, exponential fits and theta projection [15-17]. They are all used to some degree of success in uniaxial rupture predictions but they are not appropriate for use under multiaxial conditions. The effects of multiaxiality testing and modelling using notched bars have also been developed in various ways to some degree of success [18-28]. For example, one approach to assess constraint [20-23] is the use of a multiaxial stress state parameter, α , deriving an effective rupture stress, σ_{eff} , from the bounds derived from principal and von-Mises stresses, σ_I and σ_e , giving

$$\sigma_{eff} = \alpha\sigma_I + 1 - \alpha\sigma_e \quad (1)$$

In a simple form when $\alpha=0$ the failure is σ_e controlled and when $\alpha=1$ the failure is σ_I controlled. These models have been used to predict multiaxial failures in notched bars and cracked components. This approach highlights the importance of deriving a constraint term where creep cracking is concerned.

There are also available two categories of models that deal with multiaxial creep damage which are in turn analytical and numerical. In practice computationally intensive elastic/plastic/creep analysis methods are needed to derive useful results. One approach is the remaining multiaxial ductility based models which relate stress state described by a constraint parameter $h = (\sigma_m / \sigma_e)$ to multiaxial ductility [29,30] and the second approach is the continuum damage modelling (CDM) [31-38]. The latter uses different types of constraint based arguments including Eqn. (1) to determine the effects of multiaxiality under creep conditions.

Reviews of creep CDM models [13,14] highlights in depth the various isotropic damage models and their corresponding microstructure damage mechanisms. In most cases they need to establish complex constitutive relationships and define a larger number of variables to perform the predictions. Furthermore, using the CDM based models that need to derive α and h from notch bar tests to predict multiaxial rupture over long term, use numerically derived skeletal stresses in their analysis [22]. These skeletal stresses are at best numerical approximations of the notch region stress state, normalised against the creep index n . Therefore, they are only an approximate representation of the controlling stresses at the notch throat. Thus the CDM predictions from such approximations, plus the fact that upto seven materials variables may need to be determine in the analysis [14] from experimental data make their use limited to a qualitative understanding of the problem. The assessment using CDM are likely to be further diluted and unrealistic for very long test times assessment where little data will exist. In effect none of these models [31-38], have shown the ability to consistently predict transferable and validated failure life for the long term failure behaviour under multiaxial conditions unless their variables are specifically tailored to suite the circumstances.

The CDM models are, therefore, difficult and in many cases impractical for implementing in an industrial code of practice. It should be noted also that creep data, especially for large datasets have substantial scatter. This level of scatter and methods to reduce it by improving testing, measurement and analysis have always been an important subject for research [39,40]. Therefore, when case specific numerical modelling and data fitting and validations are needed to perform predictions it will be unlikely that the models will be able to sufficiently derive accurate variables to correctly predict long term failures in components.

In order to have both a consistent and robust approach to creep damage, rupture and crack growth prediction a simple and pragmatic model is needed, using only the appropriate and important variables, to address these predictive issues. This paper presents and verifies a multiaxial failure strain based constraint method that is simple in its application and can predict the whole range of creep damage and cracking under both uniaxial and multi-axial stress state.

3. Approach to Quantifying Creep Damage

Although numerous authors have suggested that there are distinct creep failure mechanism differences at various test times and stress levels, in principal, it can be said that the mechanism for creep damage mechanism is essentially, within a limited temperature range, due creep diffusional processes. This process should not change for long or short term test times [9]. However, at high loads plastic deformation and power law creep accompanies and overrides the diffusional failure mode that is characteristic for creep damage process observed in metals. It should also be noted that at longer test times although the creep damage mechanism does not change the degradation of material microstructural characteristics can further affect the local stress distribution by the diffusional processes that continue independent of the stress state. Damage in polycrystalline materials can be affected by grain boundary voids, grain orientation, precipitates and other complex sub-grain interactions and material degradation [9-11]. This process, in most cases, give rise in uniaxial tests to reduced failure ductility with a reduction in the applied stress and strain rate.

This can be explained qualitatively by means of metallurgy, void density and distribution and by identifying the alloying elements that help reduce ductility. These processes effectively give rise to reduced failure ductility at longer times in uniaxial creep tests [41,42] based on alloying impurities and heat treatment. This pseudo long term embrittlement may also be described mechanistically in terms of local sub-grain multiaxial stress state's non-uniform relaxation that develop as a result of diffusion occurring around voids, grain boundaries and other alloying elements or impurities which nucleate and join up under locally induced constrained conditions.

3.1. Creep Curve

Historically the different stages of primary, secondary and tertiary creep have been intensely modelled and categorised in relation to reducing failure ductilities. The primary and the secondary as the transient period after initial load and the steady state creep regimes occur with no measurable reduction in net-section which allow constrained creep strains. The secondary regime usually dominates creep life at low stresses and long times where diffusional creep operate. The tertiary region is in effect immaterial to the damage modelling process as it is a period where multiaxiality due to necking and the damaged microstructure reducing the effective net section finally allows the rupture of the specimen. Thus the critical failure strain is reached when the combined primary and secondary strains, accumulate to allow a transition

to the tertiary region. The primary becomes third order in long term tests where the minimum strain rate prevails allowing uniform damage to develop in the microstructure. This cut-off strain is generally called the Monkman-Grant (MG) [4] failure strain.

Creep damage development in this period, as mentioned above, is a combination of void initiation and growth due to diffusion, grain boundary sliding, and a multiplicity of other metallurgical factors and microstructure degradation and embrittlement [9-11]. All these factors are further complicated by the complex compositions and alloying elements and stress relaxation at the sub-grain level of microstructures that will have different material properties depending on their crystal orientations. On this basis the idealised MG strain in the long term tests is in effect an intrinsic measure of the micro strains and damage that accumulate with time. This concept is supported in numerous previous work in which it is shown, for example, that constrained cavity growth can occur a wide range of typical testing and service conditions [9-11]. For the case of void growth, for example, it has been shown that the number of cavities (per unit grain boundary area) is proportional to the creep strain, [10] suggesting that the number of cavities nucleated per unit time and unit grain boundary area, is proportional to the globally measured strain rate with the factor of proportionality depending strongly on the homogeneity of the materials [9].

The extent of the reduction in strain versus strain rate, shown schematically in Figure 1, can profoundly affect failure lives under uniaxial and even more under multiaxial stress states. This strain is a measurable parameter that can be conveniently linked to an increase in the local sub-grain constraint [29,30] arising from void growth, stress concentrations and any other microstructural anomalies described above. Although final failure strains from tests are usually measured and used for life predictions the appropriate strain and strain rate in these models are the measures of contained damage derived from MG strains. This local sub-grain constraint can also be idealised and quantified numerically, for example, using crystal plasticity [43,44] or grain-boundary damage modelling with random grain and grain boundary creep properties [7] which allows for the development of intergranular cracks. It is evident from the results that whilst global redistribution of stress can take place there are local regions of high constraint, usually but not always found at grain boundaries, which reflect the inhomogeneous nature that exist in a real microstructure. Even though the models are simple and idealised version of the void and damage development within a microstructure, both the analytical models [29,30] and the numerical ones [7,43,44] can contribute to the physical understanding of creep damage and quantify the relationship between constraint and creep ductility.

In multiaxial conditions found for example in notched bar tests and fracture mechanics samples, there is an increase in the localisation of damage due to a global geometric constraint arising from a stress concentration or the crack tip. In such cases the global constraint, which is time independent unless there is a geometrical or crack size change, acts as merely a stress multiplier containing damage in the important region of the elastic/plastic/creep process zone. However, the metallurgically present local triaxiality at the sub-grain level will still drive the time-dependent failure response similar to the uniaxial condition except that it is contained within a small process zone. On the above basis this paper proposes a unifying multi-scale, remaining strain based empirical/analytical model to predict creep damage and rupture over the wide spectrum of stress states and constraint levels.

4. Constrained Void Growth Models

A review the state-of-the-art of models and their relationship between stress level and stress state [13,14] suggest that creep ductility involves the understanding of void growth and coalescence mechanisms [9-11] under a hydrostatic stress state. Also plastic and power law creep mechanisms for failure under a multiaxial stress state play an important part in both the analytical and empirical modelling. A simplified schematic of the complex diffusional processes at the sub-grain level in shown in Figure 1. Most alloys exhibit an UB/LB shelf strain levels joined by a transient region, as shown in Figure 1(a), that span a wide range of strain rates.

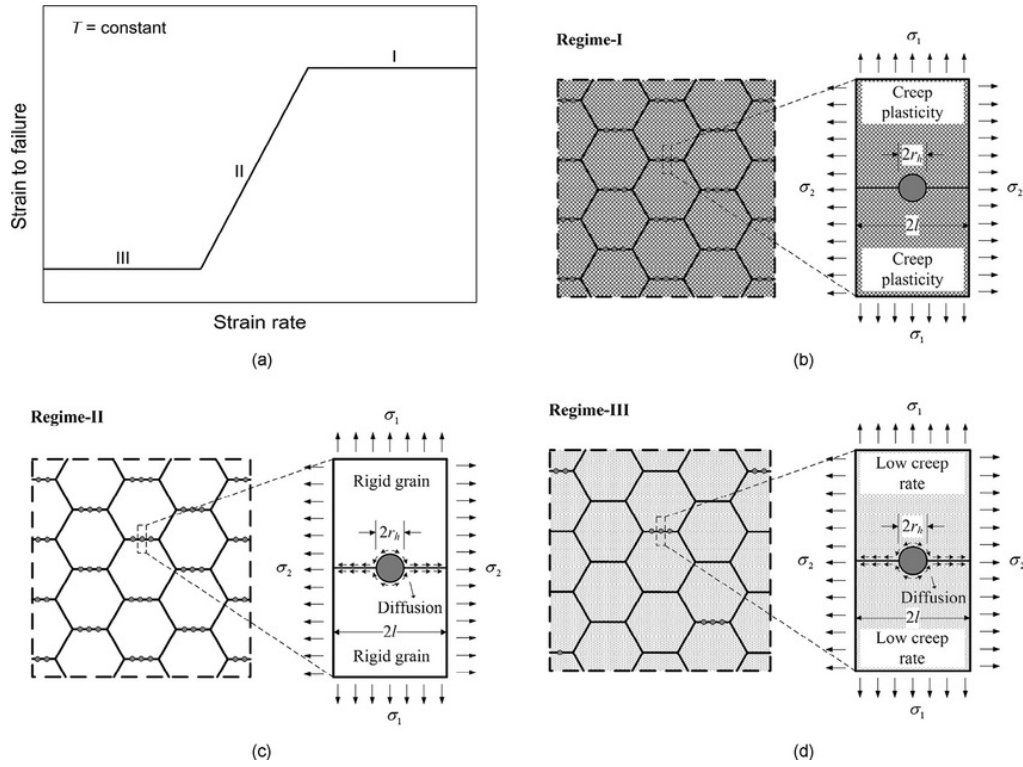


Figure 1: schematic representations of the effect of failure strain sensitivity to stress [20] showing: (a) UB/LB shelf strain over a strain rate range, (b) Regime-I: viscoplastic controlled cavity growth; (c) Regime-II: creep diffusion controlled cavity growth; (d) Regime-III: constrained diffusion cavity growth.

The variations in the different model predictions [27-29] shown in Figure 2 are due to the various assumptions, input properties and approximations adopted in the models. Specifically these void growth models derive the multiaxial/uniaxial failure strain ratio given as $\varepsilon_o^* = \varepsilon_f^*/\varepsilon_f$ with respect to h ($\sigma_m/\bar{\sigma}$), have been widely used to describe creep damage and crack growth [1-8] analytically and [4,8] numerically. The relationships presented in Figure 2 are therefore based on the following equations

$$\frac{\varepsilon_f^*}{\varepsilon_f} = \text{Sinh} \left[\frac{2(n-1/2)}{3(n+1/2)} \right] / \text{Sinh} \left[\frac{2(n-1/2)}{(n+1/2)} \frac{\sigma_m}{\bar{\sigma}} \right] \quad (2)$$

$$\frac{\varepsilon_f^*}{\varepsilon_f} = 1.65 \exp \left(-\frac{3}{2} \frac{\sigma_m}{\bar{\sigma}} \right) \quad (3)$$

where n is the creep index in the Norton's creep law, ε_f^* multiaxial failure strain and ε_f the uniaxial failure strain and multiaxial strain factor (MSF) is given as $\varepsilon_o^* = \varepsilon_f^* / \varepsilon_f$. The measure of ε_f has been in some cases loosely assumed to be the rupture strain ε_f or reduction in area, ε_{RA} , making ε_o^* more conservative. Strictly, the MG strain, ε_{fMG} , is the correct measure of the failure strain in these models since only constrained damage occurs during the useful life prior the tertiary state.

Also since the value of creep index n in Norton's law lies between 5 to 15 for most engineering materials n in the full Cocks and Ashby equation can be discounted allowing a lower-bound approximation of the relationship [7] given as

$$\varepsilon_o^* = \text{Sinh}[0.5] / \text{Sinh}[2h] \quad (4)$$

and conversely

$$h = 0.5 \ln \left(\sinh(0.5) / \varepsilon_o^* + \sqrt{(\sinh(0.5) / \varepsilon_o^*)^2 + 1} \right) \quad (5)$$

The development of these constraint multiaxial ductility models are based on the idealised growth of a single void at the microstructural level either at the grain boundaries or other micro discontinuities under a hydrostatic stress state. These equations present simply an inverse relationship of growth under hydrostatic stresses which relate a multiaxial strain factor described above as a function of a constraint level quantified by the constraint parameter $h = (\sigma_m / \bar{\sigma})$. The models, therefore, effectively predict that an increasing local h at the microscale produces a localization of creep damage allowing the material to fail at lower strains in a pseudo-brittle manner.

Conveniently, the models have been used to derive the reduction in failure strains where geometric constraint, such as in notched bars and fracture mechanics specimens, control failure [22,23]. In fact, given the sub-grain idealised approach to the development of voids in these models they could also strictly describe the drop in failure strain observed under uniaxial tests measured as MG strains. From the discussions above, also, there is also substantial justification for applying these equations to the uniaxial behaviour where local time-dependent triaxiality at low loads will dominate creep damage.

From the void growth equations shown above the local metallurgical and the global geometrically imposed constraints can reasonably be quantified by the constraint term $h = \sigma_m / \sigma_e$. Furthermore, the value of h can be determined both analytically and numerically at the global and the local level and can then be used in the above models that relate the multiaxial

failure strain factor (MSF) ε_o^* as a function of the MG uniaxial failure ductility to the constraint factor h in general terms

$$\varepsilon_o^* = \left(\frac{\varepsilon_{fMG}^*}{\varepsilon_{fMG}} \right) = f \left(\frac{\sigma_m}{\sigma_e} \right) = f(h) \quad (6)$$

where ε_{fMG}^* and ε_{fMG} in void growth models discussed above are the MG failure strains and appropriately described at different temperatures by

$$\varepsilon_{fMG} = t_r A \sigma^n e^{-Q/RT} \quad (7)$$

where t_r is the time to rupture, A and n are Norton's creep constants, $e^{-Q/RT}$ the temperature activation terms. Intrinsically ε_{fMG} is a measure of the local strains produced by the creep diffusional processes occurring uniformly at the microstructural/grain-boundary levels prior any necking. Therefore, the MG strains could be said to be a direct measure of the extent of damage developing locally at the microstructural level and in a uniform manner throughout the sample.

Figure 2 shows the range of normalised multiaxial ductility ε_o^* versus h for a number of models and creep indices n . Under extreme constrained conditions when cracks dominate ε_o^* can be upto 1/30 [1-3] to cover the full range of plane stress to plane strain conditions. However, in most engineering components the reduction ductility with constraint would be in the range $0.1 \leq \varepsilon_o^* \leq 1$. Where plasticity dominates at short times under plane stress the MSF is given as $\varepsilon_o^* = 1$ from which it can be derived that the global $h_o = 0.33$. Figure 3 show the inverse plot of the MN/UB/LB of Figure 2 but with h normalised by h_o . This normalisation in this instance is a useful approach as it allows the variation of the two variables to be compared in terms of their rate of change. Therefore by correlating $h_o^* = h_o/h$ with ε_o^* shown in Figure 3 the normalized relationship for ε_o^* to be relatively insensitive to h_o^* . From Figure 3 a simple relationship for ε_o^* as function of h_o^* can be described simply by

$$\varepsilon_o^* = B' h_o^* - b' \quad (8)$$

Where B' and b' are material constants and affected by the experimental unknowns and other approximations made by the model and ultimately affected by creep ductility in addition of the variability in the derivation of h . To simplify this further, as B' is near unity and $b' \ll 1$ as seen in Figure 3 for the wide range of steels under consideration and in the range $1 > \varepsilon_o^* > 0.1$, it can be argued that

$$\varepsilon_o^* \approx h_o^* \quad (9)$$

This simple approximation which suggests parity between the two normalised variables, is still consistent with the inverse relation of constraint and remaining multiaxial ductility derived analytically by a number of models shown Figure 2. This is later used as the basis of a damage/constraint factor to predict stress/rupture plots in uniaxial and multiaxial specimens.

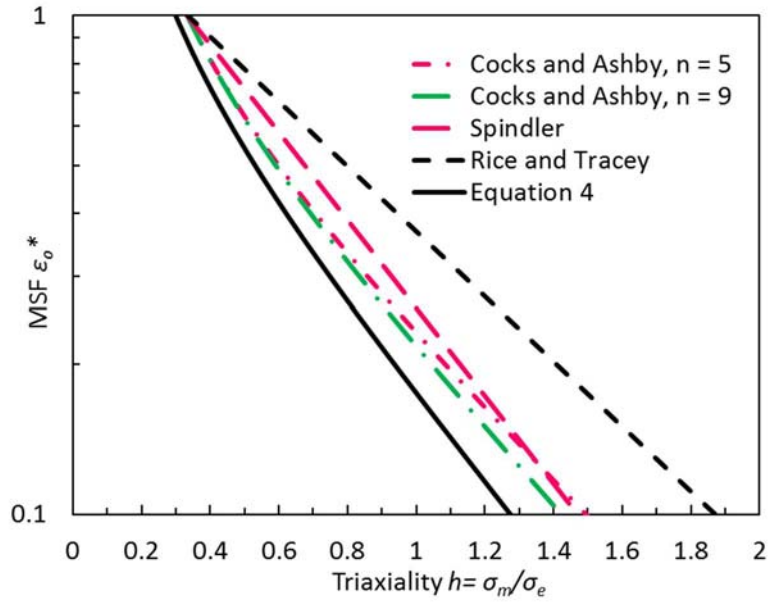


Figure 2: range of ε_0^* versus h for a number of models and creep properties

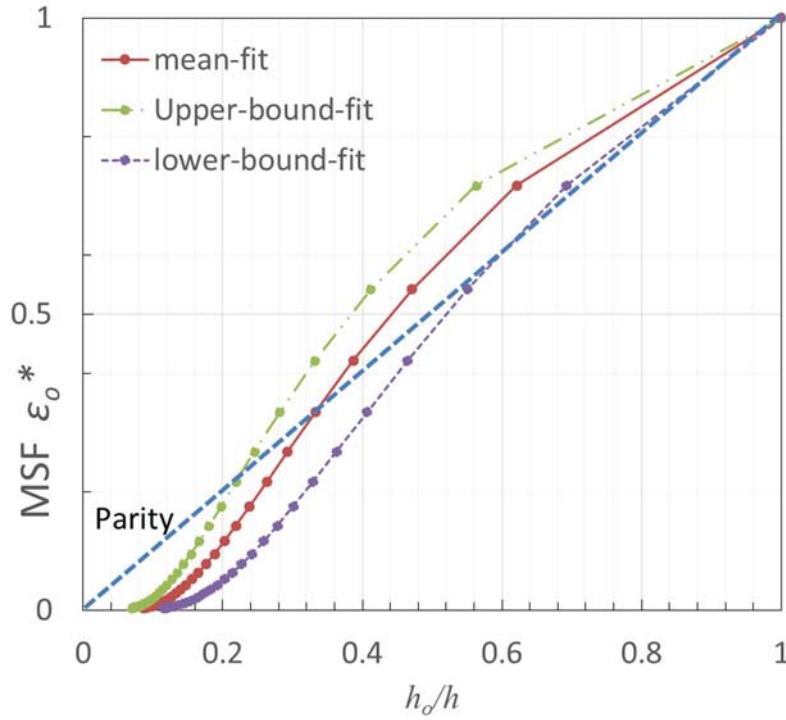


Figure 3: the normalised ductility versus ($h_o^* = h_o/h$) taken from the range of models in Figure 2 showing the 1:1 parity between the variables

5. Uniaxial Fracture Profiles

In support of the above arguments the fracture profiles in Figure 4 and Figure 5 show for P91/92, effectively a general trend towards reduced necking (highlighted by the circles drawn on the micrographs). From this the results show a reduction in failure strain with a reduction in stress ranging in failure durations of between 4000 and 20000 hours. At low loads and longer times, the reduction in ductility could reach a lower shelf with intergranular failure and virtually no necking present (as seen in Figure 5b). The implications of this is that any damage model

that attempts to predict damage and rupture needs to take into account the relationship between ductility and constraint shown in Figure 2.

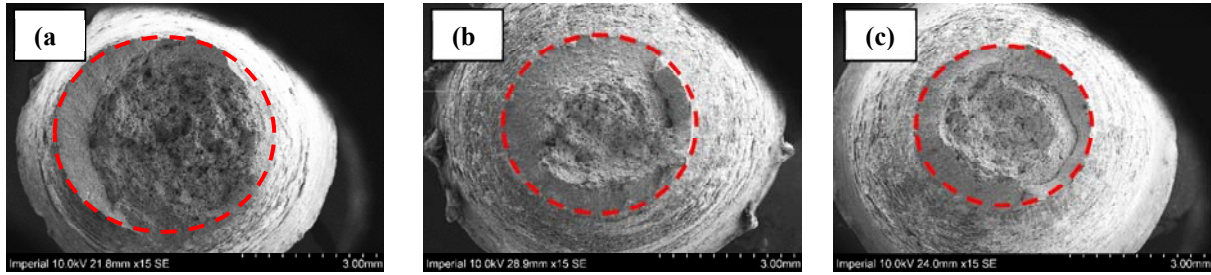


Figure 4: Fracture surface P92 at 650 °C at loads (a) 160MPa, (b) 170MPa and (c) 180MPa
[Internal data]



Figure 5: Fracture surface of Grade 91 crept specimen after (a) 100 hours at 550°C and (b) 20,014 hours at 650°C [45]

6. Development of a Strain-Based Local Constraint Criteria for Creep Rupture

A large uniaxial database [5-6], for P91/P92 over a range of temperatures and test time ($>100,000$ h), is used to develop and validate a local model which considers the presence of metallurgical constraint at the sub-grain level controlling creep damage. Clearly there will be a wide measure of scatter and a level of uncertainty in the data due to the wide range of the database, number of tests, test temperatures, different batches of materials and tests performed at different laboratories. However validation of a model under such conditions could give more support to its industrial applicability. The model, at its basis, considers two critical material properties. These are the very short term creep strength equivalent to the material tensile strength and at the long term the appropriate uniaxial failure strain that can best be described by the Monkman Grant failure strain relationship described above. From the available creep constitutive relations a combined geometric and microstructural constraint remaining ductility approach is arrived at which unifies the creep uniaxial, multiaxial and crack growth failure processes for very long test times.

Figure 6 and Figure 7 show for the P91 and P92 steels the rupture stress σ_r normalised by the extrapolated short term ($t \approx 0$) upper bound rupture stress σ_{t0} versus time highlighting a temperature dependence. At the one extreme, at time $t \approx 0$, the rupture stress σ_{t0} tends towards an upper shelf limiting stress which for these steels is found to be equivalent to the yield stress (σ_y) at any particular temperature. At the other extreme the rupture stress for times $>100,000$ hours show a temperature dependence that can be dealt with by an activation energy term as in Eq. (7).

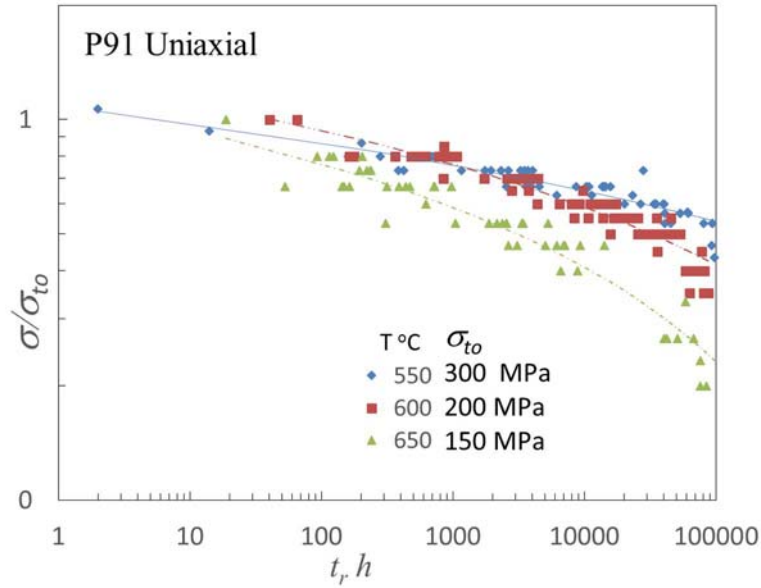


Figure 6: normalised stress rupture plot for P91 at different temperatures

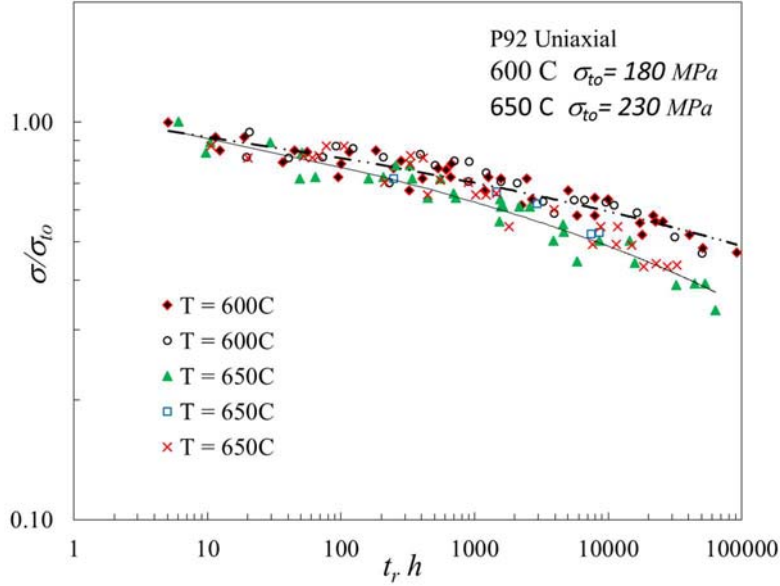


Figure 7: normalised stress rupture plot for P92 at different temperatures

in Figure 8 and Figure 9 for P91 and P92 respectively show the MG failure ductility versus rupture time have an upper limit of 0.06 or 6%. In these cases there seems to be very limited temperature dependence within the range of the scatter of data. In Figure 10 and Figure 11 normalising the failure strains by the upper-bound MG failure strain of 0.06 for both P91 and P92 in the manner of Eq. (6) and cross plotting against the data for the normalised stress rupture versus times in Figure 6 and Figure 7 can give a generalised relationship between ε_o^* and σ_t/σ_{t0} . The relationship can be conveniently be presented as

$$\frac{\sigma_t}{\sigma_{t0}} = \varepsilon_o^{*\beta} \quad (10)$$

Where β is a material constant related to creep strain and temperature. Over the range $1 \leq \varepsilon_o^* \geq 0.1$ the stress to failure can therefore be defined as $\sigma = f(\varepsilon_o^*)$ where $\varepsilon_o^* = f(h)$ as given by Eqn. (6) and derived from the MG strains.

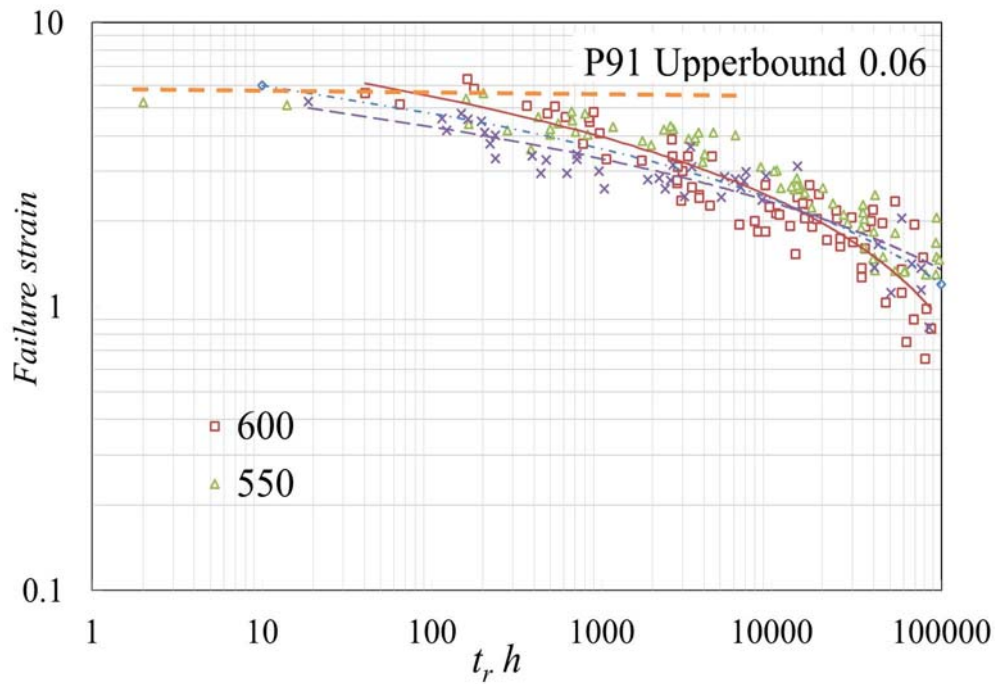


Figure 8: MG ductility versus time to rupture for a range of P91 steels at different temperatures.

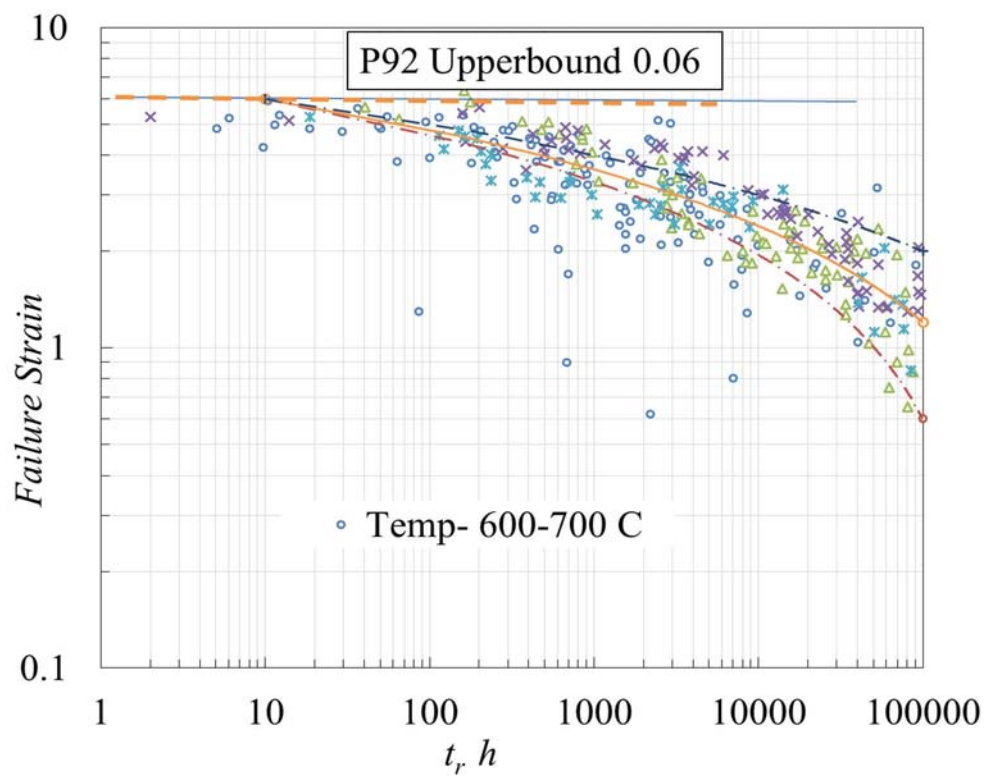


Figure 9: MG ductility versus time to rupture for a range of P92 steels for various temperatures.

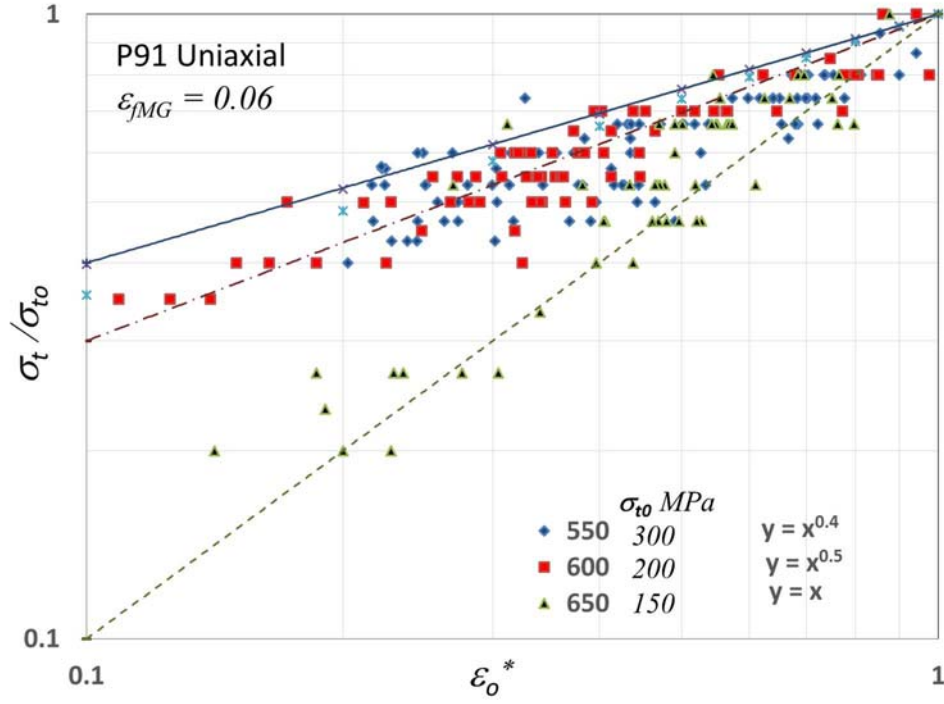


Figure 10: normalised rupture stress versus MG failure strains for P91 at different temperatures

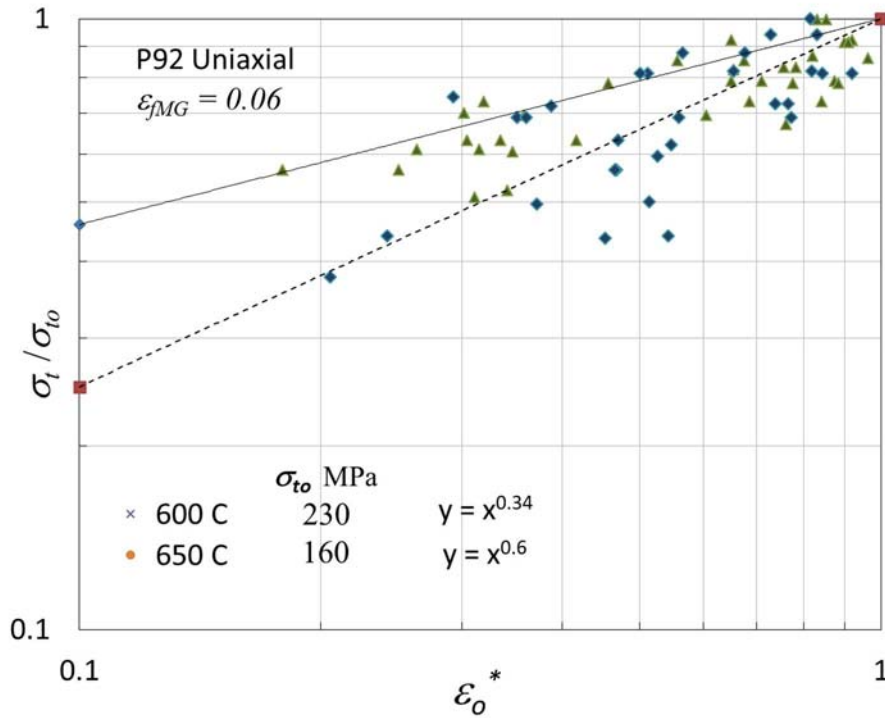


Figure 11: normalised rupture stress versus MG failure strains for P92 at different temperatures

in Eqn. (6) a relationship for the predicted rupture stress σ_t for time t is derived as a simple normalised form, appropriate in the practical range of $1 > \epsilon_o^* > 0.1$, giving

$$\sigma_t = \sigma_{to} \left(\frac{h_o}{h} \right)^\beta \quad (11)$$

where β is near a value of unity and for the case of uniaxial specimens, depending on the material creep ductility, temperature or the variations in the models in Eqn. (11) and shown in Figure 2 could vary between 0.3 to 1.

Experimental characterisation of MG creep failure strains with time in uniaxial plotted in Figure 8 and Figure 9 show substantial scatter and little sensitivity to temperature and material variation. Results for a number of P91/P92 steels from different batches and in X-weld and HAZ conditions show similar levels of scatter which is not convenient for detailed analysis. Clearly if standardised tests on (ductile) and (brittle) material were conducted under controlled testing conditions a difference in creep strain with time will be highlighted in line with the present model predictions.

7. Analysis of MG Failure Strains for Uniaxial and Notched Bars

Figure 12 show the failure times as a function of normalised MG failure strains for various P91/92 steels at different temperatures for as received conditions in uniaxial tests. It is clear that there is a wide scatter of data but no clear difference between the datasets. Since it is difficult to show individual trends for these datasets an assessment can be carried out based on the ME/ UB/LB at this stage to identify the sensitivity of the variables to the predictions.

In addition, for notch bar multiaxial type tests, shown schematically in Figure 13, which are tested in accordance to a testing standard [22] a similar approach for failure strain may be taken. The code of practice includes in its analysis the appropriate skeletal stresses and the mode of measuring the notch root strains. In effect the constraint due to the notch is time dependent but the root notch failure strain is still time dependent suggesting that a localised sub-grain constraint still applies. For these notch bars it is justifiable to compare the measured local strains at the notch throat as creep strains and damage mainly occur in the notch region. Clearly there will be a degree of inaccuracy in the way the local strains are normalised. However for the purpose of this paper only available literature data can be utilised. More detailed comparison between notch acuity and material pedigree is only possible when a careful testing programme of batches of alloys under controlled purity and fabrication is performed. However, it should also be noted that, within the range of data scatter, the data is relatively insensitive to the notch diameter and batch to batch variation hence the reason for not presenting a detail of each testing programme.

It should also be noted that the tertiary region of creep strains is substantially suppressed for notched tests especially at long times. Figure 14 shows the only available estimated MG strains data for notched bars test of a P92 steels at 600 and 650 °C [5,24-28]. Similar for the data shown in Figure 10 and Figure 11 where the normalised MSF, ε_o^* , is plotted against the normalised rupture stress in uniaxial tests Figure 15 shows the relationship constructed for notched bars. In this case the failure strains are both a function of the geometric constraint which is time independent and the microstructural constraint which as discussed for the uniaxial cases are dependent of diffusion creep process at long terms.

Batch to batch variability and compositional differences must exist in the normalised measures of MG strains versus time. However, within the scatter it is very difficult to differentiate between them. Hence using Figure 12 and Figure 14 the general relationship between failure times and normalised MSF, ε_o^* , for both uniaxial and notched bars can be given by the relationship

$$t = A' \exp(A''(\varepsilon_o^*)) \quad (12)$$

Where A' and A'' are material constants. Table 1 shows the MN/UB/LB values for the constants in Eqn. (12) for the various steels. Following the proposed constitutive relationships in Eqns.(11) and (12) it may therefore be possible to make appropriate and conservative predictions for the stress/rupture behaviour for these steels at different material conditions and test temperatures.

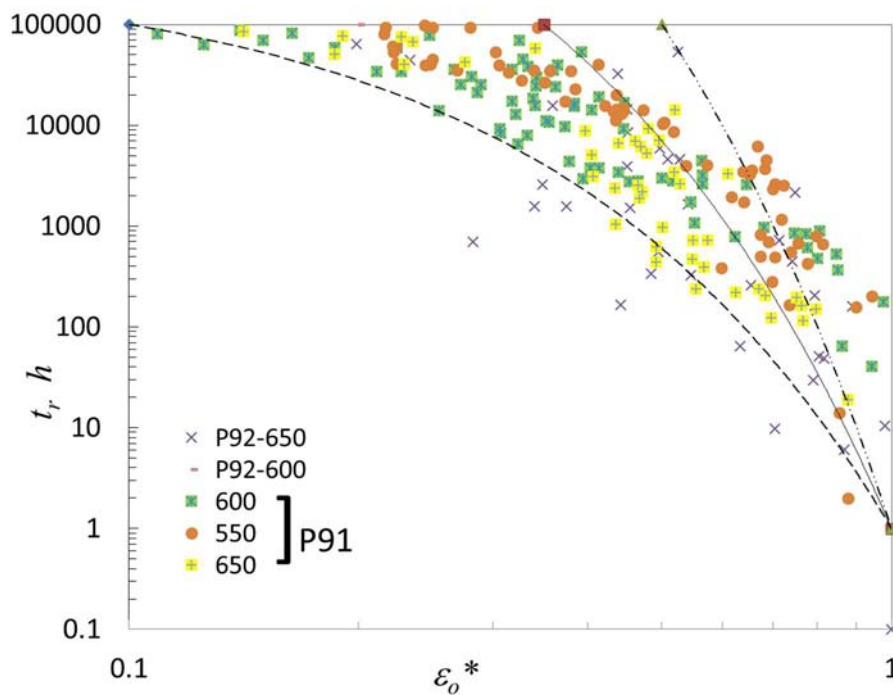


Figure 12: Normalised MG strains versus rupture for a parent P91 and P92 for a range of temperatures, bounded by exponential, MN/UB/LB fits

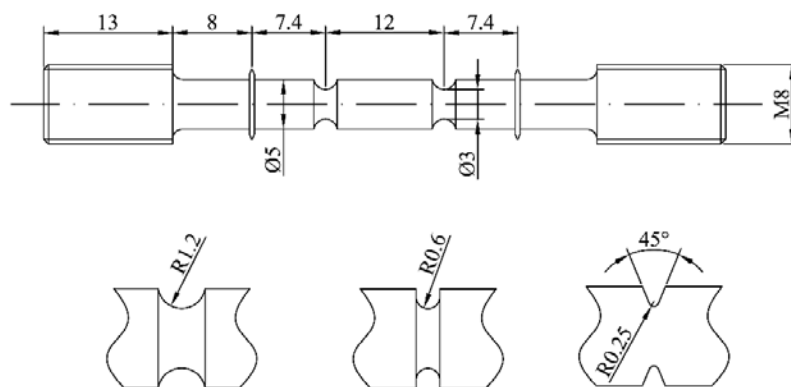


Figure 13: Dimension and size of the employed notch bar specimen [22].

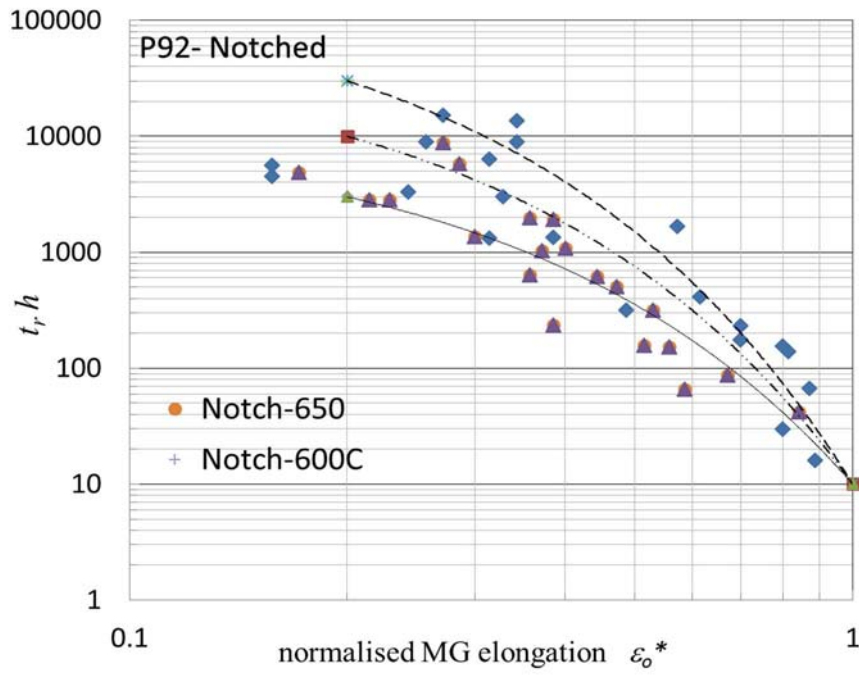


Figure 14: Normalised ductility MG ductility for P92 Notched bar versus time bounded by exponential, MN/UB/LB fits

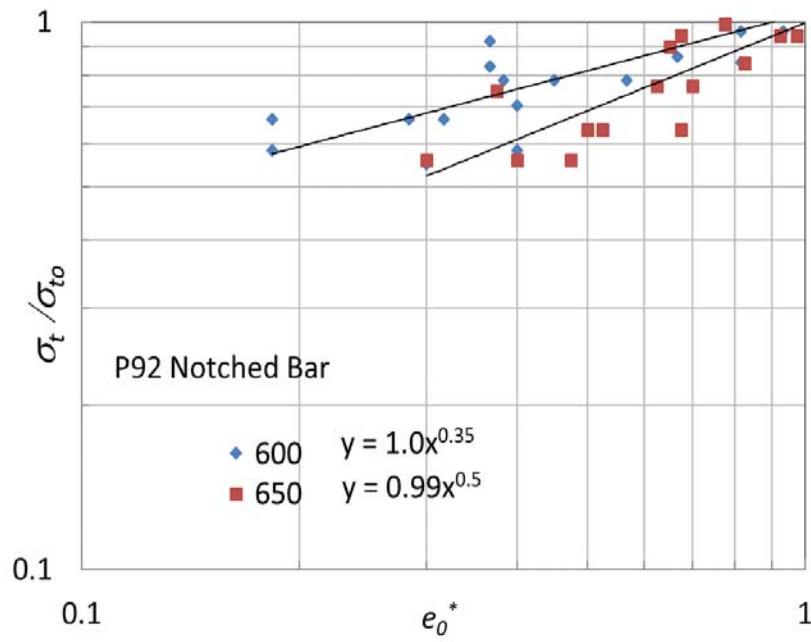


Figure 15: normalised stress versus MG failure strains for Notched P92 [5] steels

8. Application of the Model

In this section the model presented above in Eqns.(11), (12) will be used to predict large P91/P92 datasets. Table 1 shows a list of variables that are used selectively in Eqns.(11), (12) make the predictions. Given the level of scatter in most of the data a sensitivity is carried out in this way to identify the prediction bounds.

8.1. Predictions for Uniaxial and Notched Bar Creep Rupture times

The creep damage/strain based constraint model described above serves an important purpose in developing a uniform method in predicting creep damage under uniaxial and multiaxial stress states and ultimately for creep crack initiation and growth. The microstructurally controlled local constraint determines the drop in failure strain in uniaxial models and the globally based constraint calculations determine the notched bar stress state and stress level at a contained elastic/plastic/creep process zone. Therefore, as discussed earlier the global h acts only as a stress multiplier and the local h dominates the damage process zone over the loading period.

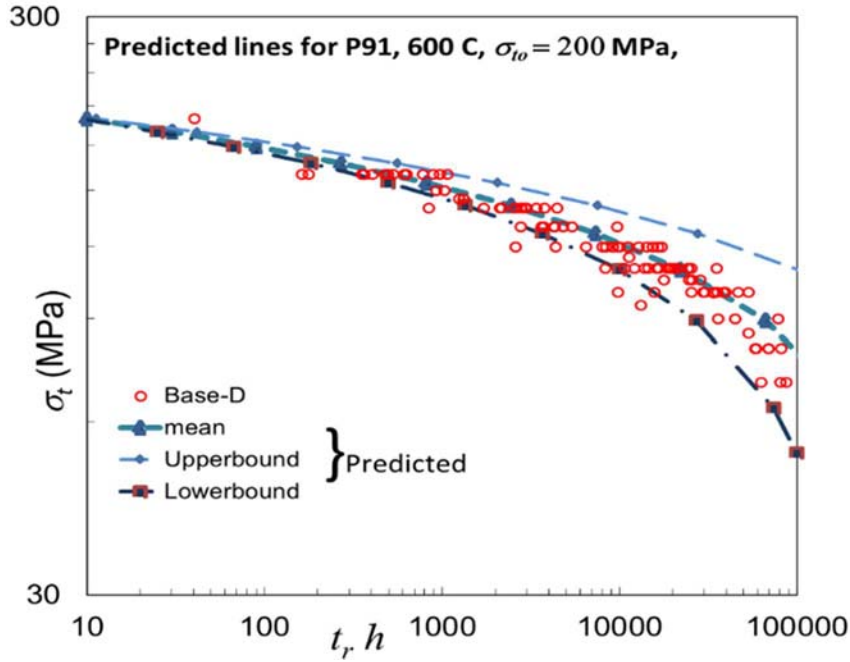


Figure 16: Predicted rupture times for base P91 steels [5,6] at 600 °C using Eqns.(11), (12) with the values in Table 1 showing sensitivity to the bounds in Figure 12

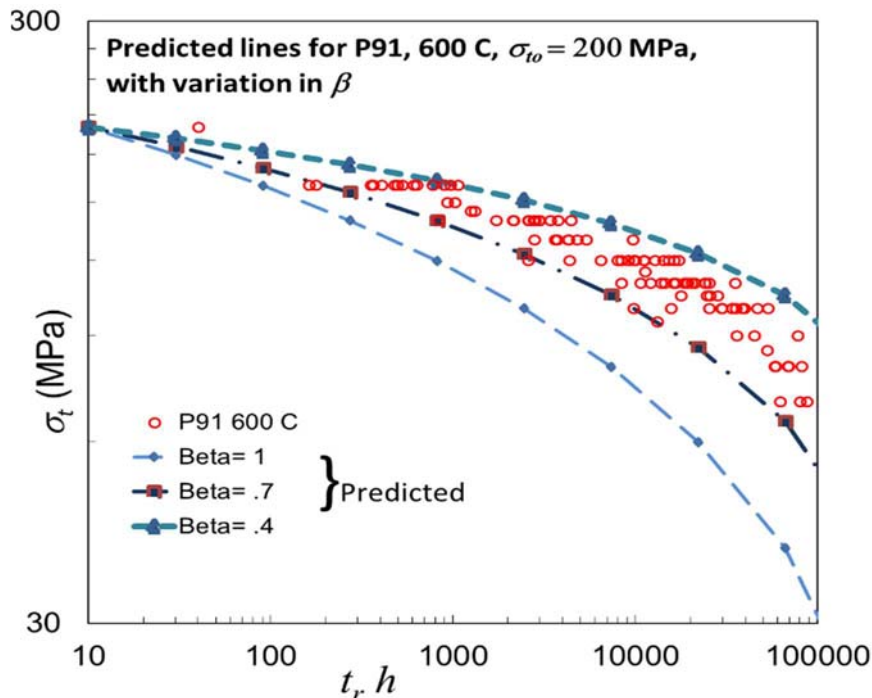


Figure 17: Predicted rupture times for P91 [5,6] at 600 °C using Eqns.(11), (12) with the input values in Table 1 showing sensitivity to β Eqn. (12)

Predictions and sensitivity analyses using Eqns.(11), (12) for the various P91/P92 steels at various temperatures are shown in to Figure 21 It should be noted that the available data for the material characterisation needed to perform the predictions are not always data of known quality and in some cases the appropriate strain measurements may not have always been available. For this reason, there is substantial scatter involved. The approach taken therefore is to look at the MN/UB/LB for material characterisations and carry out a sensitivity analysis.

Figure 16 shows the predicted rupture times for base P91 steels at 600 °C using Eqns.(11), (12) with the values in Table 1 at constant σ_{t0} equivalent to the materials yield stress, showing sensitivity to the bounds of the data in Figure 12. Figure 17 shows the sensitivity to the variable β in Eqn (12) for the same P91 data at 600 °C from the previous figure. It is clear that sensitivity to this variable is high and small changes can bound the data. The increase in β increases conservatism and would reflect a reduction in creep toughness in the alloy. Figure 18 and Figure 19 for P91 and P92 at different temperatures shows the mean prediction lines for these alloys. Overall the correlation is good over a 100000 hours. Any further level of conservatism in the predictions can be made more confidently by increasing β . Given the level of scatter in the data it will be difficult to identify which batch of steel is any better than another. In order to make a comment on the creep damage resistance of a particular batch carefull tests of brittle and ductile material with known composition and fabrication pedigree would need to be carried out.

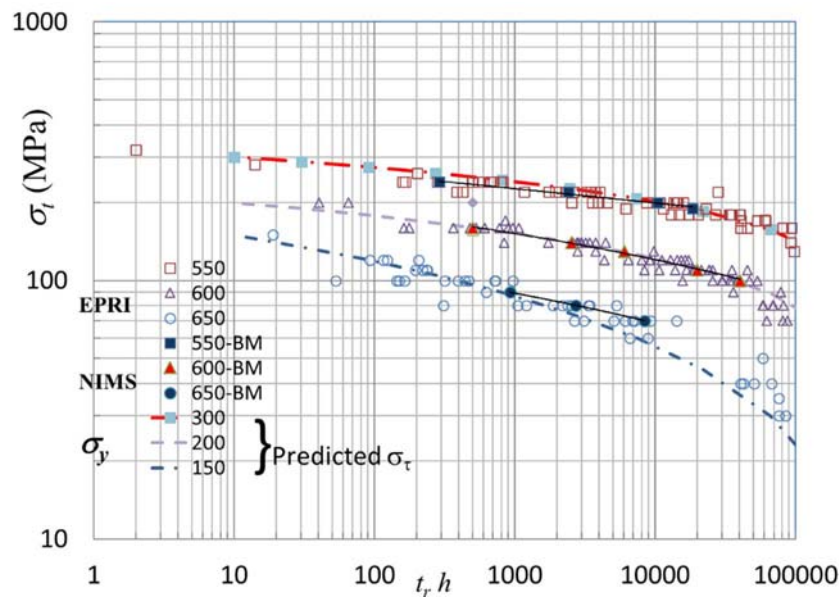


Figure 18: Predicted rupture times for various P91 batches [5,6] between 550-650 °C using Eqns.(11), (12) with the values in Table 1 using the mean fit shown in Figure 12

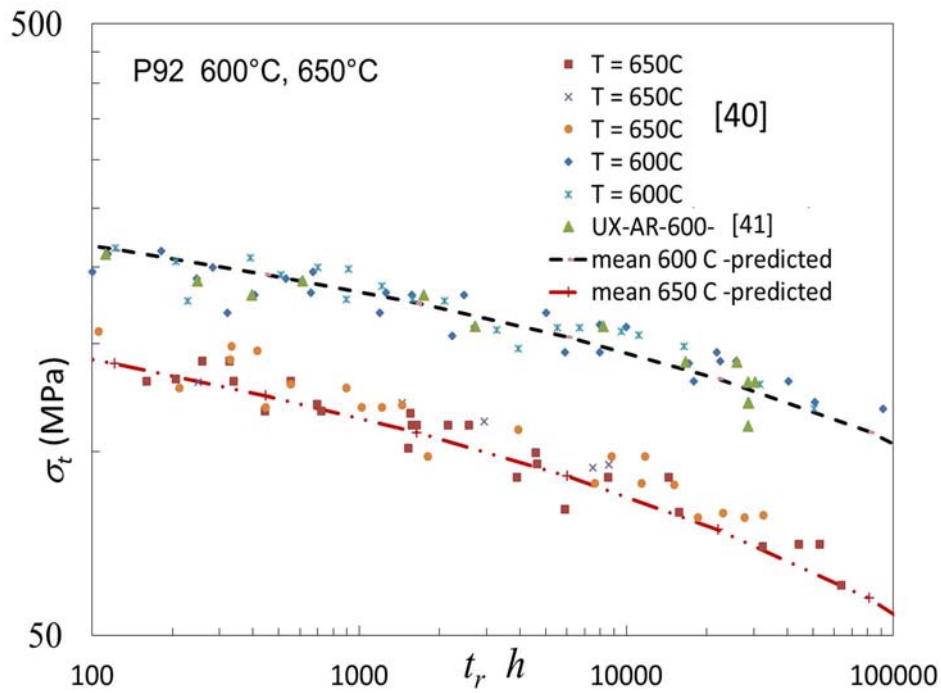


Figure 19: Predicted rupture times for various uniaxial P92 batches between 550-650 °C using Eqns.(11), (12) with the values in Table 1 using the mean fit.

Table 1: Material constants in Eqns.(11), (12)

Material	Bet fits	A'	A''	σ_{to} @550C	σ_{to} @600C	σ_{to} @650C	β
P91-parent	Mean (ME)	6E+05	-11	300	200	150	.4
X-weld	Upper (UB)	5E+06	-13	300	200	150	.7
HAZ	Lower (LB)	3E+05	-10	300	200	150	1.
P92 parent	Mean (ME)	4E+06	-13		250	170	.5
	Upper (UB)	2E+06	-14		250	170	.5
	Lower (LB)	1E+07	-11.5		250	170	.5
P92	Mean (ME)	5E+04	-8.5		400	300	.7
Notch	Upper (UB)	2E+05	-10		400	300	.7
	Lower (LB)	1E+04	-7		400	300	.7

Notched bar analysis was performed on the data from P92 notch tests at 600 and 650 °C []. The stress rupture for notched bars are usually described in terms of net-section stress and erroneously compared to uniaxial tests in terms of their level of notch strengthening. This is clearly misleading in that the presence of the notch effectively weakens the material due to constraint. The data are also plotted against skeletal principal stress and von-Mises stress which are in themselves approximations of a representative stress that may correlate the data. None of these correlating stresses can exactly represent the failure modes in these geometries but at best can be used to compare failure between different notch size and dimensions.

Table 1 lists the relevant material properties for Eqns.(11), (12) to derive the notch rupture predictions. For the case of notch bars the derivation for σ_{to} is when time $t \approx 0$ depending with which stress parameter the data is analysed with. Figure 20 shows the test data and the predictions for P92 Notch bars at 600 and 650 °C for which MG strains were available and

shown in Figure 14. By using the appropriate mean values in Table 1 derived from a net-section stress analysis for the notched bar it is clear from this figure that the accelerated failure due to the notch is well predicted. In fact the model suggests that accelerated testing due to the presence of the notch is due to a combination of the geometric constraint plus the microstructural constraint. When compared to uniaxial test the enhanced stress level due to the triaxiality dictates the level of damage in the creep process zone but the local constraint will drive the failure over time. In fact one reason why it is difficult to extend the lives for notched bars to the 100,000 h and beyond level is due to the increased constraint due to the two factors discussed. More detailed analysis and identification of the notch failure mode and strains at long terms is necessary in order to compare them with similar uniaxial tests with equivalent MG failure strain levels.

8.2. Predictions for Weld, X-Weld and CT specimens

As a further extension of the model's capabilities, data for P91 X-Weld, HAZ and Compact tension (CT) tests [3] were analysed in terms of time to rupture. For these cases no MG strain data were available and for the case of CT it is not possible to derive MG strains. X-Weld and HAZ material will behave in a more creep brittle manner hence the lower-bound values of P91 parent data in Table 1 for Eqns.(11), (12) were used for the predictions. Figure 21 show the predictions for P91 X-Weld, HAZ uniaxial tests. This figure highlights substantial scatter in the data which is typical of a pseudo brittle constrained failure where the growth and joining up of micro-cracks present at grain boundaries could shorten lives considerably.

For the CT geometry, also shown in Figure 21, a reference stress approach [46] at the initial crack length is usually used to compare total failure times with uniaxial tests. In these cases the σ_{Io} at $t \approx 0$ is derived from the calculated skeletal or reference stresses. Also the time to total rupture rather than the crack growth rate was used to correlate against the calculated σ_I . In the compact tension (CT) specimen data shown in Figure 21, it is at first clear from the scatter in the data that correlating the cracking data in this manner is not ideal. However as a comparison with uniaxial data it is useful to approach the problem in this manner. Based on these difficulties the predictions using the lower bound values in Table 1 can at best present a range of idealised predictions.

It is also evident in all the predictions shown in Figure 20 and Figure 21 for the notched and the CT tests that increase in geometric constraint generally reduces the ductility of the material allowing faster failure. Hence the CT with the highest constraint would show a faster failure times compared to the notch bar. In fact, tests under these conditions fail at shorter times and are used to as accelerated tests for batch to batch comparisons. In Figure 20 and Figure 21 for the notch bars and CT there is, therefore, a rapid drop in the prediction of the rupture stress at longer times. This effectively confirms the model's conservatism since it is clear that any reduction of ε_0^* to near below 0.1 means that crack dominant brittle failures would occur. In such cases a fracture mechanics approach would be the route for assessment. This is discussed in the next section.

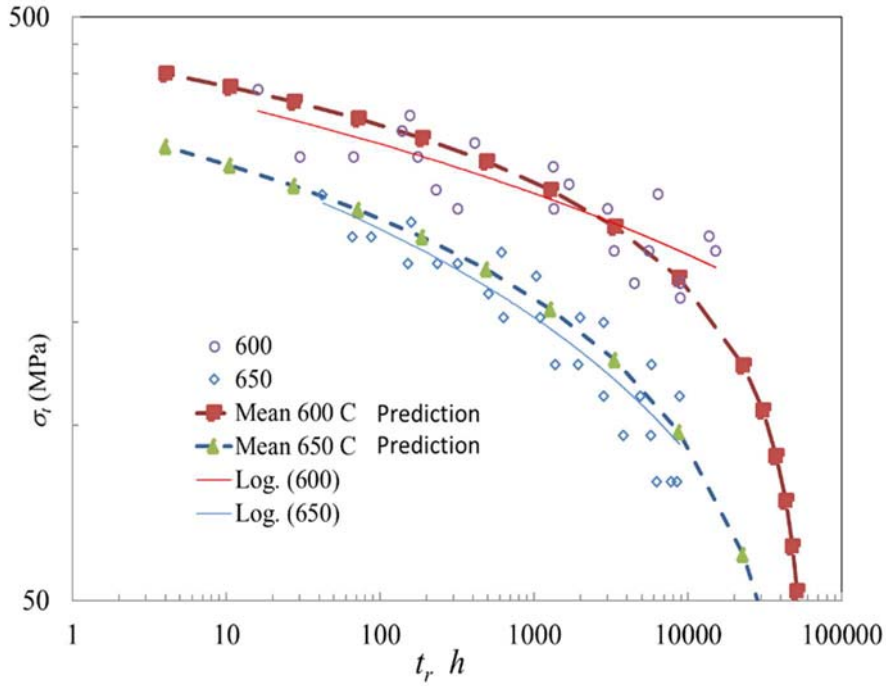


Figure 20: Predicted rupture times for notched P92 at 600 °C and 650 °C using Eqns.(11), (12) with the values in Table 1 using the mean fit shown in Figure 14

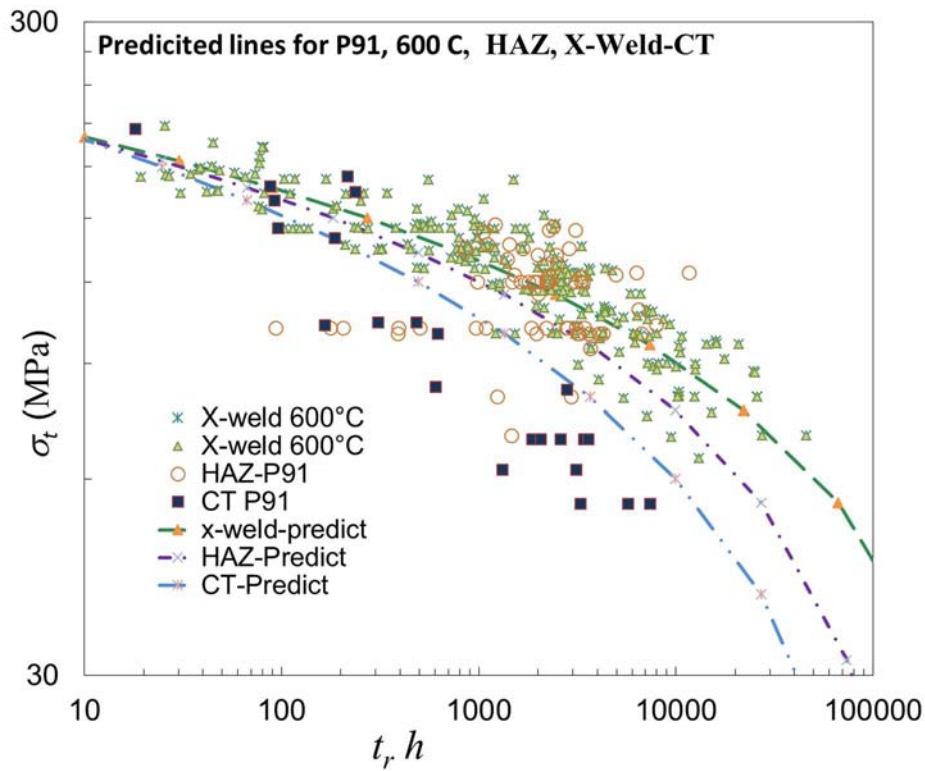


Figure 21: Predicted rupture times for P91 X-weld, HAZ and Compact Tension samples [28] using Eqns.(11), (12) using P91 lower-bound values shown in Table 1 and Figure 12

9. Extension of the Model to Creep Crack Initiaiton and Growth

As argued earlier, creep mechanism for damage development is the same irrespective of applied stress and geometry except that at high stresses visco-plasticity and power law creep plays an important part in the development and rupture of micro-voids whereas at lower

stresses the local sub-grain stress state could control creep induce failure. Therefore when the extreme case of creep crack growth under plane strain is taken into consideration it is possible to predict crack initiation and failure by crack growth using the same remaining multiaxial ductility based model discussed above. This approach has, for many years, been implemented in a well established remaining multiaxial ductility dependent model [1-3] called the NSW (Nikbin, Smith and Webster), described below. The model predicts a remaining ductility based constraint controlled UB/LB cracking response over the plane stress/strain limits.

9.1. NSW Crack Growth Rate Model

Under steady state creep conditions, a power law relationship can be inferred when the experimental creep crack growth rate is plotted against the creep fracture mechanics parameter, C^* shown as

$$\dot{a} = DC^{*\phi} \quad (13)$$

Where

$$C^* = -\frac{1}{B} \frac{dU^*}{da} \quad (14)$$

where a is the crack length. $\dot{a} = da/dt$ is crack growth rate, B is the thickness and U^* is the potential energy rate. The C^* integral as a creep crack growth correlating parameter identifies the appropriate crack tip stress distribution. This parameter is widely used for correlating creep crack growth under steady state creep conditions. It can be derived numerically analogous to J [3,46] or experimentally from the energy release rate principles [39].

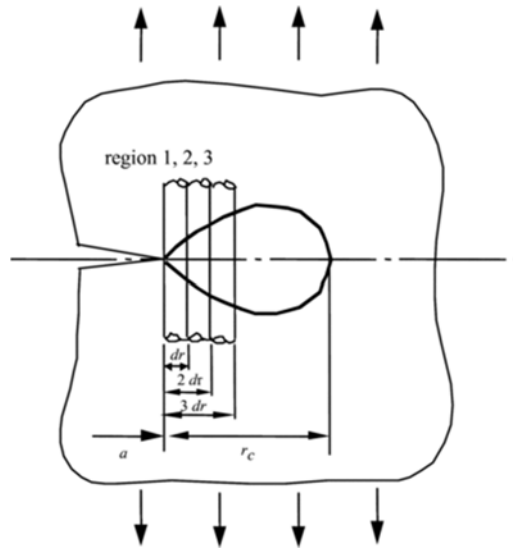


Figure 22: Schematic showing uniaxial bars with width dr with the r_c creep process zone ahead of the crack tip.

The NSW model [1-3] is a multiaxial ductility based approach which under steady state conditions, uses the uniaxial creep data to predict crack initiation and growth assuming a creep process zone. This process zone is shown schematically in Figure 22 where the individual creep bars of, length dr , as pseudo uniaxial samples under varying stress levels. In the same way as for the uniaxial and the notched bar tests the multiaxial failure ductility and constraint is controlled, by the development of voids, in the case of cracks C^* represents the crack tip stress

intensification field but local constraint h as modelled in Eqns. (2)- (4) controls the stress state. In this way, cracking rate can be derived by the integration of the individual failure rates of each dr uniaxial section. Using this approach, the NSW model can be used to predict cracking rate as an inverse function of the constraint related multiaxial ductility ε_o^* and C^* giving

$$\dot{a} = \frac{(n+1)}{\varepsilon_o^*} (A_{rc})^{1/(n+1)} \left[\frac{C^*}{I_n} \right]^{n/(n+1)} \quad (15)$$

where n is the power law stress exponent, I_n is a dimensionless stress state constant dependent on n , r_c is the creep process zone, whose size is relatively insensitive in this form due to the small fractional power. As a good simplification and when material properties are not available an approximate solution to Eqn. (15) for predicting cracking rate for most engineering materials [2] can be given in the form

$$\dot{a} = \frac{3}{\varepsilon_o^*} (C^*)^\phi \quad (16)$$

where $\dot{a}$ is crack growth rate in mm/h, $\phi = n/(n+1)$ and ε_o^* is chosen as the normalised multiaxial failure strain (MSF) which should ideally be derived from the MG ductility and for extreme conservatism from the failure ductility or the reduction in area. Also by taking Eqn. (9) into consideration and assuming MG failure controls the multiaxial ductility, as shown in Eqn. (6), then Eqn. (16) can be conveniently be presented as

$$\dot{a} = \frac{3}{h_o^*} (C^*)^{(\phi)} \quad (17)$$

From Eqn. (9) it has been shown that $\varepsilon_o^* = h_o^* = 1$ at plane stress and under plane strain $\varepsilon_o^* = h_o^* \leq 1/30$ where intergranular creep brittle crackling tends to occur at high levels of constraint. But in most cases for engineering alloys $\varepsilon_o^* = h_o^* \leq 1/10$ would bound the crack growth rate data.

9.2. NSW Crack Initiation Model

Further to the NSW crack growth rate model an expression for crack initiation time, t_i , can also be derived based on the attainment of a critical strain at a critical distance ahead of a stationary crack tip. Time t_i is defined as the time to achieve a measurable small amount of crack extension. If the minimum crack extension, da , that can be measured reliably is taken as $\Delta a = 200 \mu m$ [39,47] and that it is assumed the crack growth is a continuous process, which begins immediately on loading, then the initiation time t_i may be obtained from

$$t_i = \int_{a_i}^{a_i + \Delta a} \frac{da}{\dot{a}} \quad (18)$$

The integration in Eqn. (18) cannot be simply performed as the dependence of $\dot{a}$ on time is not generally available. However, estimates of initiation time may be obtained by assuming a constant crack growth rate over Δa . UB/LB estimates for t_i can be derived by substituting

appropriate estimates of initial cracking rate $\dot{a}_i$ or the steady state cracking rate $\dot{a}_s$ respectively [47], into Eqn. (15) giving

$$\frac{\Delta a \varepsilon_o^*}{3C^{*0.85}} \leq t_i \leq \frac{\Delta a(n+1)\varepsilon_o^*}{3C^{*0.85}} \quad (19)$$

Depending on the level of conservatism needed and the availability of good test data either limit can be used to predict initiation times t_i .

9.3. Predictions for crack growth rates using the NSW model

The crack growth rate predictions are highlighted in Figure 23 for P91 parent and HAZ tests and the crack initiation predictions are shown in Figure 24. From Eqn. (17) in Figure 23 the approximate NSW LB/UB plane/stress/strain predictions for crack growth rates are shown in the figure using 0.01 as LB failure strain when both MG and EL failure strains tend to converge and UB uniaxial failure ductility based on MG at 0.06 and EL failure strain of 0.3. It has been shown previously [1-3] that the UB strains using EL failure strain or RA reduction in area, very conservatively predict the cracking rates [1-3,50]. However, by using the MG failure strain criteria the level of conservatism can be reduced to a more acceptable level as shown in Figure 23.

For the P91 parent and HAZ data it seems that a transition exists between low to high C^* with cracking rates moving towards plane strain at low loads and long term crack growth rate tests which exhibit very low ductility and pseudo brittle intergranular failure. This is shown by the dashed line as the best fit to the data in Figure 23. Effectively at low C^* the cracking rate is at the upper-bound at or near a lower shelf ductility. This transition phenomenon has been observed both experimentally [31,51-52] and numerically [8,53]. Clearly with more accurate knowledge of the material's tensile and creep properties this present model is able to robustly predict failure at very low stresses at which components operate.

Figure 23, therefore, can be looked at as the reverse of the behaviour observed in uniaxial and multiaxial rupture behaviour. Therefore at extremely low loads, where there is an increase in both global constraint and cracking rate at plane strain, the creep response corresponds to the locally induced constraint in long term uniaxial tests. Conversely at high loads the failure is plane stress and more likely to be ductile rupture in both the uniaxial and cracked specimens. The latter short term tests have less relevance in an industrial life assessment context.

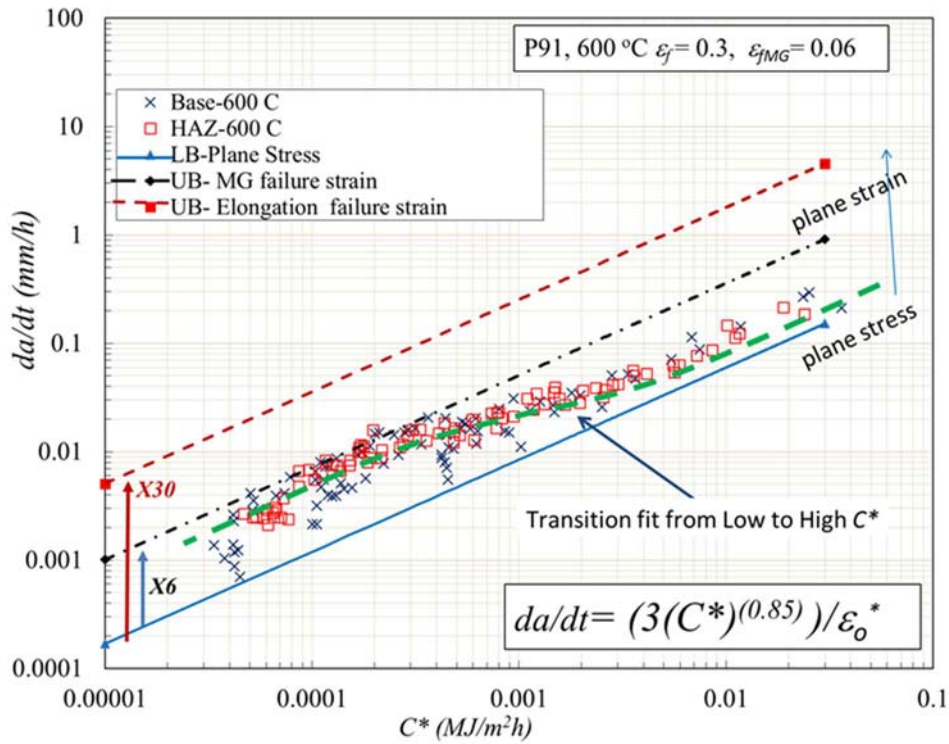


Figure 23: Approximate NSW LB/UB plane/stress/strain predictions for P91 base and HAZ crack growth between 580-600 °C [3] using 0.01 failure strain as lowerbound based on MG and tensile uniaxial elongation (EL) failure strain uppershelves of 0.06 and 0.3 respectively. The dashed curve is the best fit for the parent and HAZ showing a transition shift between low to high C^* .

9.4. Predictions for crack initiation times using the NSW model

For crack initiation the data for P91 parent and HAZ at 600 °C [3] is shown in Figure 24 where the predicted initiation times to crack initiation depth of $\Delta a = 200 \mu m$ are shown plotted against the C^* parameter. From Eqn. (19) the approximate NSW UB/LB plane/stress/strain predictions for crack initiation times are shown in the figure using, as for the cracking rate in Figure 23, 0.01 failure strain as lowerbound and uppershelve uniaxial failure ductility based on MG 0.06 and tensile uniaxial failure strain of 0.3. The dashed line is the best fit to the data showing a transition between low to high C^* very similar to Figure 23 for the crack growth rate. In Figure 24 as each experimental point is derived from one test as opposed to crack growth rates in Figure 23 where multiple points are derived from each there will be substantial scatter as seen. To highlight the detailed trend over a wide C^* range many more tests would be necessary to fully verify these trends especially at longer times. However it is clear that the model is sufficiently robust to highlight the differences between very long and short times in terms of their relative initiation times with respect to C^* .

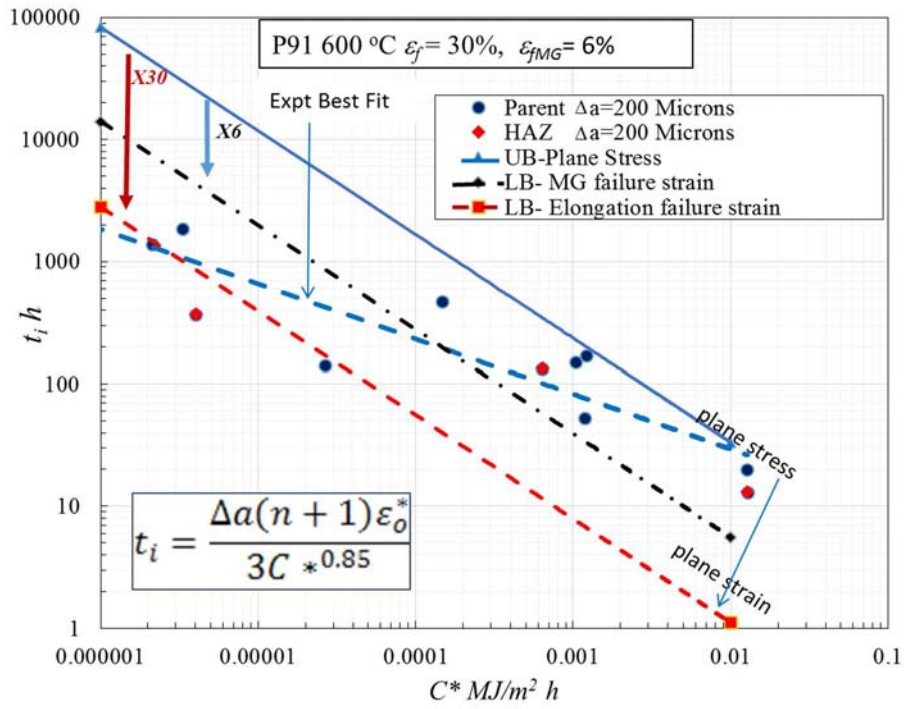


Figure 24: Approximate NSW UB/LB plane/stress/strain predictions for P91 base, and HAZ at 600 °C [3] using 0.01 failure strain as lowerbound based on MG and tensile uniaxial elongation (EL) failure strain uppershelves of 0.06 and 0.3 respectively. The dashed line is the best fit to the data showing a transition between low to high C^*

10. Discussion and Conclusions

A short review of available uniaxial creep and CDM predictive models, which are numerous and wide ranging in their applications, show the use of various approaches towards the effects constraint in order to predict multiaxial failure. The idealised approach to the models also suggest that although they are scientifically relevant and fundamental in their approach they are usually analytically complex, contain too many variables that need measurements or numerically intensive to make them universally acceptable under industrial applications. Also very few approaches are able to robustly apply their assessment to a wider defect analysis where substantial data scatter exist at varying time and lengths scales. The availability of a large database of long term uniaxial and notched P91/92 tests data presented in this paper, albeit with substantial scatter, has allowed a radically different and simplified analytical/empirical strain based approach to be taken with respect to quantifying time/stress dependent creep damage. The proposed method, based on a multiscale level of constraint, is shown to predict rupture in uniaxial, and notched bars as well as predict crack initiation and growth rates in fracture mechanics samples. This suggests that as a structural integrity defect assessment tool the model could unify the spectrum of creep failure modes in engineering components.

In uniaxial and notched bars by simply considering the initial fracture strength of the material and the time dependent creep toughness/ductility of the material it has been shown that it is possible to predict long term failure in uniaxial and multiaxial stress conditions. It is proposed that MG strains in polycrystalline materials are intrinsically linked to the local sub-grain creep damage due to constrained void growth, grain boundary sliding, inclusions and many other forms of microstructural degradation which is distributed in uniaxial samples and more locally at the notch root. Previously proposed void growth models are simplified to a linear inverse relationship between normalised MG failure strains and the sub-grain local constraint (h) which

develops at grain boundaries. Thus by simply linking multiaxial ductility inversely to an appropriate constraint factor h it is possible to derive an appropriate effective stress to rupture σ_{ir} versus time to rupture based on the local sub-grain constraint level. The model has been shown to successfully predict uniaxial rupture over a wide time range ($>100,000$ hours) and also notched bars which fail at substantially lower time frames of around 10,000h for the P91/P92 steels.

The model is further extended to predict failure under crack initiation and growth controlled stress state by directly linking it to the well-known NSW model [1-2] used to predict the UB/LB plane strain/stress crack growth rates in fracture mechanics geometries. By the same token as for the notched and uniaxial cases, a multiscale constraint argument affecting the creep process zone ahead of the crack controls the state of stress at the sub-grain level whilst the global stress intensity described by C^* , which contains the creep process zone, controls geometric constraint at the global level.

The model needs the material properties such as tensile yield stress and UTS, stress levels at and short term creep times plus a measure of the MG strain sensitivity to applied stress in order to perform the predictions in uniaxial and notched bars. For fracture mechanics geometries the uniaxial properties are also needed as well as the crack initiation and growth rate properties of the material batches. The stress sensitivity to creep ductility can be effectively derived from uniaxial tests of 5000-10,000 hours or possibly from accelerated notch bar tests which would need to be calibrated. At the very least, using this method, any improvement in the composition or material heat treatment can quantitatively be compared and conservative predictions of their rupture live to be made within a practical time frame.

Additional testing specifically focused on obtaining short term pedigree data to compare detailed MG data for creep brittle/ductile uniaxial and notched bar tests will help in improved validation of the model. In parallel, numerical approaches to verify these findings at the multiscale and sub-grain level have clearly shown that the local stress states do dictate the way voids or local microstructural anomalies and discontinuities develop into micro-damage and the subsequent linking to produce larger cracks and failure. The reduction in MG ductility at long terms and low applied stresses fully support this proposed strain based multiaxial constraint model. Given the level of inherent scatter in the data and fabrication and testing uncertainties that cannot be accounted for in such databases it is shown that the model is sufficiently simple and robust to be developed into an industrial code of practice. This model therefore would be a pragmatic approach to unify the creep damage development process in uniaxial, notched and cracked specimens based on the present analysis of data. Further work in verification and validation, especially with good quality pedigree data, is needed to identify the safe design levels for use in life assessment predictions.

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