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A physically-based creep damage model for effects of different precipitate types

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Abstract

The development of a new precipitate coarsening continuum damage mechanics (CDM) model to simulate the multi-precipitate strengthening mechanisms present in 9Cr steels under high temperature creep deformation is presented here. The key strengthening and degradation associated with the different coarsening kinematics and volume fractions associated with $M_23C_6$ and MX precipitates in 9Cr steels are simulated within a CDM framework for the first time. The new CDM creep model is implemented in a uniaxial code and successfully applied to 9Cr steels across a range of temperatures via physically-based steady-state creep constants. The role of increasing Al content on the high temperature creep behaviour of 9Cr steels is simulated via varying the volume fraction of MX carbonitrides. The results highlight (i) the important role of MX carbonitrides on creep strength of 9Cr steels and (ii) the requirement to simulate steady-state creep behaviour in 9Cr steels from a physical basis.

1 Introduction

There is a requirement for higher firing temperatures and pressures in fossil fuel burning power plants to increase plant efficiency, decrease CO₂ emissions and reduce the overall
environmental cost of utilizing fossil fuels [1]. However, elevated temperatures and pressures cause increased microstructural creep deformation, potentially leading to significant reductions in power plant component life. In some cases, materials are required to withstand temperatures in excess of 650 °C [2] and pressures as high as 35 MPa in ultra-supercritical (USC) plants. These more severe loading conditions have resulted in an increased rate of unexpected component failures [3]. Specifically, Type IV cracking, occurring in the intercritical heat affected zone (IC-HAZ) region of the welded connections, has been identified as the primary mode of failure for power plant piping systems. The microstructural transformations which take place during the welding process have been highlighted as key drivers of crack initiation in the IC-HAZ region and subsequent propagation to failure [4]. Thus, there is a requirement to account for the effect of microstructure evolution on the creep performance of candidate materials, such as 9Cr steels.

The high creep strength of 9Cr steels is due to their precipitate and solute strengthened hierarchical (martensitic-ferritic) microstructure. This hierarchical microstructure consists of prior austenite grain boundaries, subdivided into packets and blocks. The blocks are further separated into low-angle boundary regions of martensitic laths, delineated by regions of high dislocation density. This hierarchical microstructure is illustrated schematically in Figure 1. Solid solution strengthening exists primarily in the form of Mo and W solutes and precipitate strengthening is due to MX and M_{23}C_6 particles. Fine MX carbonitrides, of type VN and NbC, are dispersed throughout the microstructure, with M_{23}C_6 carbides primarily distributed along boundaries (M representing Fe, Cr, Mo or W). It is the presence of these precipitates which provides the key strengthening mechanism for 9Cr steels operating in the dislocational creep regime (intermediate to high stresses), via a dislocation pinning mechanism as highlighted by the work of Panait et al. [5,6]. To date, numerous creep studies have been
performed on 9Cr steels [5-9], across a range of temperatures and stresses. Due to the complex nature of the microstructure in 9Cr steels, substantial microstructural evolution occurs during creep deformation. Under relatively short-term creep behaviour, the martensitic laths are replaced by more equi-axed subgrain structures which coarsen due to inelastic deformation [10,11]. The dislocation density also decreases substantially [12,13]. For lower stresses and longer creep test durations, subgrain transformation and coarsening occurs at a slower rate in conjunction with the formation of new precipitate phases, namely (i) Fe$_2$Mo and Fe$_2$W Laves phase particles at the expense of Mo and W solute strengthening [5,14] and (ii) Z-phase particles in place of MX carbonitrides [6]. During creep of precipitate strengthened alloys, coarsening of precipitates also occurs, due to (i) thermal aging via the Ostwald ripening mechanism and (ii) strain-induced particle coarsening, as reported in the work of Taneike et al. [15]. In a 12Cr steel, Taneike et al. measured significant precipitate diameter differences between the grip and gauge sections of creep tested specimens. These differences were attributed to enhanced dislocation diffusion taking place in the gauge length of the specimens, driven by strain accumulation. Hattestrand and Andrén [16] have performed similar studies to Taneike et al. for 9Cr steels and found a significant effect of strain-induced coarsening on M$_{23}$C$_6$ carbides.

The effect of aluminium on the creep life of 9Cr steels was investigated by Magnusson and Sandström [17], who demonstrated weakening of the material, employing thermodynamic simulations to demonstrate the formation of Al nitrides, in place of the more thermodynamically stable VN carbonitrides. Al nitride (AlN) formation reduced the rupture time by up to a factor of 6 for an aluminium content increase of <0.01 wt.% to 0.028 wt.%.

This highlights the important role of MX carbonitrides on the creep strength of 9Cr steels and
the requirement to include the parameters defining MX precipitates as key variables for constitutive modelling of such materials.

A wide range of computational models exist for simulating creep deformation in candidate materials, including the Norton power law model [18] and the hyperbolic sine modelling methodology of Hayhurst and co-workers [19,20]. More physically-based models have been developed by Magnusson and Sandström [21], Spigarelli [22] and Yadav et al. [23], for example, to capture the key microstructural evolutions in 9Cr steels during primary creep. The continuum damage mechanics (CDM) modelling methodology of Hayhurst and co-workers has been successfully applied to 9Cr alloys in the work of Hyde et al. [8], Christopher et al. [24] and Nandi et al. [25]. However, these works have only considered the role of M$_{23}$C$_6$ carbide coarsening on the creep strength of 9Cr steels, and omitted the effect of MX carbonitrides on microstructural degradation. Although the approach of Oruganti et al. [26] does explicitly take into account the role of the MX precipitates in place of the M$_{23}$C$_6$ carbides, a unified continuum damage mechanics approach to modelling multiple precipitate type strengthening in 9Cr steels has not been developed to date.

Thus, the primary focus of this paper is to extend the modelling methodology of Hayhurst et al. [19,20] to account for the multi-precipitate mechanisms present in 9Cr steels via the development of a new precipitate coarsening damage model. This new model will capture the effect of different coarsening kinematics, inter-particle spacing and volume fraction on the creep life of multi-precipitate type strengthened alloys. The material model is successfully applied to 9Cr steels across a range of temperatures via the inclusion of temperature-independent steady-state creep constants. The role of Al content on creep performance is also simulated via a decreasing MX precipitate volume fraction, a variable which is identified in the present work as a critical value for creep life assessment of 9Cr steels.
2 Methodology

2.1 Hyperbolic sine creep model

The behaviour of 9Cr steels under creep deformation is simulated using the three state variable CDM model of Hayhurst and co-workers [19,20], modified here to include temperature-independent creep constants and, more specifically, to account for multiple precipitate types, as present in 9Cr steels. The uniaxial form of the hyperbolic sine creep flow rule proposed is:

\[
\dot{\varepsilon}_c = \dot{\varepsilon}_0 \exp \left( -\frac{\Delta F}{k_B T} \right) \sinh \left( \frac{\sigma(1-H)}{\sigma_0 (1-D_p)(1-D_{cr})} \right),
\]

here \( \dot{\varepsilon}_0 \) is the pre-exponential creep constant, \( \Delta F \) is Helmholtz free energy, \( k_B \) is Boltzmann constant, \( T \) is absolute temperature, \( \sigma \) is applied stress and \( H, D_p \) and \( D_{cr} \) are the CDM state variables related to primary hardening, precipitate coarsening and intergranular cavitation, respectively. The creep constant, \( \sigma_0 \), is a stress quantity related to the key strengthening mechanisms in the material and defined as [27]:

\[
\sigma_0 = \frac{M k_B T}{b^2 \lambda_{ob}},
\]

where \( M \) is the Taylor factor, \( b \) is the magnitude of the Burgers vector and \( \lambda_{ob} \) is the mean spacing of obstacles. In the hierarchical precipitate-strengthened microstructure of 9Cr steels, the primary obstacles are (i) MX and M_{23}C_{6} precipitates, (ii) high and low angle grain boundaries and (iii) dislocations, as illustrated schematically in Figure 1.

Primary creep in 9Cr steels is related to hardening and microstructural evolution associated with the key strengthening mechanisms, namely martensitic lath and subgrain strengthening,
dislocation hardening and back-stresses associated with dislocation pinning at MX carbonitrides and M$_{23}$C$_6$ carbides. The time derivative of primary hardening, $H$, is defined here, using the model of Hayhurst et al. [19,20], as:

$$
\dot{H} = \frac{h}{\sigma} \left(1 - \frac{H}{H^*}\right) \dot{\varepsilon}_{cr},
$$

(3)

where $h$ is a hardening modulus and $H^*$ is the saturated value of primary hardening, $H$. The creep damage term, $D_{cr}$, accounts for the formation and growth of creep cavities via the mechanism of vacancy diffusion (as illustrated schematically in Figure 2a) in the material. The time derivative of cavitation, $D_{cr}$, is defined here as:

$$
\dot{D}_{cr} = \bar{C} \dot{\varepsilon}_{cr},
$$

(4)

where $\bar{C}$ is a temperature-dependent constant related to creep ductility, $\varepsilon_f$. Vacancy diffusion through the crystal lattice to grain boundaries results in the individual vacancies coalescing and nucleating a cavity as illustrated schematically in Figure 2b. Further growth of the cavity due to strain accumulation is included in the model via Equation 4 above as per the work of Cane [29].

2.2 *Multiple precipitate type CDM term*

Unlike other alloys, such as CrMoV alloys, for example, 9Cr steels have multiple precipitate types contributing to their high creep strength. In the initial microstructure, MX precipitates of type VN and NbC are distributed throughout the microstructure and M$_{23}$C$_6$ carbides, dominated by Cr$_{23}$C$_6$ carbides, are dispersed along boundaries. Thus, to account for the effects of coarsening related evolution of MX and M$_{23}$C$_6$ precipitates, it is necessary to modify the carbide coarsening precipitate CDM model of Hayhurst and co-workers [19,20],
to incorporate volume fraction and precipitate diameter as key variables and to extend to two precipitate types (MX and M$_{23}$C$_6$), as described below.

Figure 3 is a schematic illustration of the mechanism of precipitate coarsening, in terms of the formation of Orowan loops around the precipitate and increased inter-particle spacing, reducing the effect of dislocation pinning and loop formation. Figure 3a shows the material condition with a fine dispersion of precipitates prior to creep exposure. Figure 3b shows the material post creep exposure where the pinning effect has been reduced due to the coarsening of pinning precipitates.

The precipitate damage term is defined as a function of the ratio of the mean inter-particle spacing for the initial, $\lambda_0$, and current, $\lambda$, microstructure:

$$D_p = 1 - \frac{\lambda_0}{\lambda},$$

(5)

with the mean inter-particle spacing defined, using the approach of Magnusson and Sandström [21], as:

$$\frac{1}{\lambda^2} = \frac{1}{\lambda_c^2} + \frac{1}{\lambda_m^2},$$

(6)

where $\lambda_k$ is the mean inter-particle spacing of a type k precipitate and the subscripts 'c' and 'm' denote M$_{23}$C$_6$ and MX precipitates, respectively. The mean spacing of precipitates is a function of the diameter and volume fraction of the precipitate, such that:

$$\lambda_k = d_k \left[ \left( \frac{\pi}{6 f_k} \right)^{\frac{1}{3}} \right]^{-1},$$

(7)
where $d_k$ and $f_k$ are precipitate diameter and volume fraction, respectively. The evolution of the mean precipitate diameter is defined via the Ostwald ripening process shown schematically in Figure 3. However, to account for strain-induced particle coarsening, the model of Taneike et al. [15] is used to define the mean precipitate radius:

$$d_k^3 - d_{k,0}^3 = K_k t + \phi_k \epsilon, \quad (8)$$

where $d_{k,0}$ is the initial precipitate diameter, $K_k$ is the thermal coarsening rate and $\phi_k$ is the strain-induced precipitate coarsening constant. Thus, the precipitate damage term of Equation (5) becomes:

$$D_p = 1 - \frac{d_{c,0}^2 d_{m,0}^2}{d_c^2 d_m^2} \left( \frac{d_c^2 B_c + d_m^2 B_m}{d_{c,0}^2 B_c + d_{m,0}^2 B_m} \right)^2, \quad (9)$$

where

$$B_k = \left[ \left( \frac{\pi}{6f_k} \right)^{\frac{1}{3}} \right]^2 - 1. \quad (10)$$

Under the assumption that the formation of Z-phase particles, at the expense of MX carbonitrides, is omitted from the present study and the loss of $M_23C_6$ carbides via processes such as Cr depletion during oxidation does not occur, it is assumed that the volume fraction of precipitates remains constant during the Ostwald ripening process. Thus, using Equation (8), the derivative of Equation (9) with respect to time gives the precipitate damage rate as:

$$\dot{D}_p = \frac{1}{3} \frac{d_{c,0}^2 d_{m,0}^2}{d_c^2 d_m^2} \left( \frac{B_c}{d_m^3} \right)^{\frac{1}{3}} \left( K_m + \phi_m \dot{\epsilon} \right) + \frac{B_m}{d_c^3} \left( K_c + \phi_c \dot{\epsilon} \right) \frac{1}{1 - D_p}, \quad (11)$$
where it can be shown that Equation (11) reduces to the more familiar model of Hayhurst and co-workers [19,20] (e.g. $\dot{D}_p = (K/3d_0^3)(1 - D_p)^4$ for $\phi = 0$), for a single precipitate type.

3 Results

3.1 Model calibration

The CDM creep model is applied here to two Gr. 91 alloys, (i) conventional P91 alloy and (ii) Bar 257, a P91 alloy with increased Al content. The nominal compositions of the various materials investigated in this study are presented in Table 1. The required material parameters are identified using a combination of creep and thermal aging data, in conjunction with the results of published microstructural data.

The temperature-dependent Young's modulus is identified from the elastic region of monotonic tensile test data [30] and presented in Table 2. The magnitude of Burgers vector, $b$, and the Taylor factor, $M$, have values of 0.248 nm and 2.9 respectively for body centre cubic materials. The Helmholtz free energy is calculated as $5.978 \times 10^{-19}$ J from [26].

The values for the initial precipitate diameters, $d_{m,0}$ and $d_{c,0}$, are taken from the data of Hald and Korcakova [31]. The volume fractions of the precipitates are obtained from Magnusson and Sandström [17], with a reduced value of $f_m$ for Bar 257, consistent with the formation of thermally unstable AlN nitrides at the expense of the MX precipitates. The thermal coarsening rates, $K_m$ and $K_c$, are identified via a least squares optimisation procedure, fitting Equation (8) to the experimental data of Hald and Korcakova [31] for P91 and P92 alloys thermally aged at temperatures of 600 °C and 650 °C, individually for both $M_{23}C_6$ and MX precipitates. A comparison with the experimental data is presented in Figure 4. The P92 data in Figure 4b is assumed to be applicable for the identification of $K_m$, as the MX precipitates
are of similar composition in both materials [31]. The identified coarsening rates are shown in Table 2, where the 625 °C values have been obtained by linear interpolation.

The constants, $\dot{\varepsilon}_0$ and $\lambda_{ob}$, associated with the steady state creep behaviour of the candidate materials, are identified from minimum creep strain-rate (MSR) data. Using Equation (1), the MSR can be defined as, $MSR = \dot{\varepsilon}_0 \exp(-\Delta F/k_B T) \sinh(\sigma/\sigma_0)$, representative values of $\dot{\varepsilon}_0$ and $\lambda_{ob}$ for P91 are identified using experimental data at temperatures of 600 °C and 650 °C, as presented in Figure 5. The mean obstacle spacing, $\lambda_{ob}$, is related to the mean spacing of (i) precipitates ($\lambda_c$ and $\lambda_m$), (ii) grain boundaries, $d_g$, (iii) laths (or subgrains), $w$, and (iv) dislocations, $s$. Using an approach similar to that of Roters et al. [32], $\lambda_{ob}$ is thus:

$$\frac{1}{\lambda_{ob}} = \frac{\lambda_c^2 + \lambda_m^2}{\lambda_c^2 \lambda_m^2} + \frac{1}{w} + \frac{1}{d_g} + \frac{1}{s}, \quad (12)$$

where the spacing of dislocations, $s$, is inversely proportional to the square root of the dislocation density, $\rho$. The $\lambda_{ob}$ value of 31 nm identified, via the MSR data of Figure 5, is consistent with measured microstructural values of $w$ in the range of 0.3 to 1.5 µm [8,9,34], $d_g$ on the order of 4 µm [34] and a dislocation density of $1.6\times10^{14}$ to $7.5\times10^{14}$ m$^{-2}$ [12,34] for P91 alloys, via Equation (12). The value of $\lambda_{ob}$ for Bar 257 is then identified retrospectively using Equation (12) in conjunction with a modified value of $\lambda_m$ for Bar 257.

Finally, the primary creep constants $h$ and $H^r$, intergranular cavitation constant, $\overline{C}$, and the strain-induced precipitate coarsening constants, $\phi_k$, were identified using a least squares approach to minimise the difference between the model and experimental data. The optimised set of material parameters are presented in Table 2 and Table 3. From the work of Mustata and Hayhurst [35], the creep constant $\overline{C}$ is related to creep ductility via the expression:
\[
\overline{C} = \frac{k_N}{\varepsilon_i},
\]  

(13)

where \(k_N\) is a constant with a value of 1/3 for uniaxial loading. Using Equation (13), the optimised values of \(\overline{C}\) lead to creep ductility values of between 3.9 % and 13.3 % for temperatures increasing from 600 °C to 650 °C, values similar to the data of Klueh [36]. Furthermore, the low value of \(\phi_m\) is justified by the negligible increase in MX diameter measured by Hattestrand and Andrén [16] in crept specimens, compared to thermal aging data.

3.2 Modelling creep deformation in 9Cr steels

The ability of the proposed CDM model to predict creep deformation in conventional P91 steel (ASTM A335/ASME SA335), is presented in this section. Although data is obtained from different sources, potentially with small variations in composition and heat treatment, the alloys investigated fit within the composition range for Gr. 91 steel presented in ASME specifications [37]. Hence, these Gr. 91 alloys will be collectively denoted as ‘conventional P91’.

Figure 6a presents a comparison of the predicted creep response against the measured data of Orlová et al. [9] at 600 °C for three different stresses. The predictions correlate closely with the experimental data for the intermediate stresses, viz. 110 MPa and 125 MPa, and reasonably closely for the higher stress, 175 MPa. A comparison of the predicted and measured creep rates is presented in Figure 6b for a stress of 125 MPa at 600 °C, showing very close correlation.
The comparison of predicted and measured responses at 625 °C is presented in Figure 7, showing close agreement of the model with experimental data. Similar results are obtained for two different applied stresses at a temperature of 650 °C, as illustrated in Figure 8. Figure 9 highlights the ability of the proposed model to predict creep rupture life via comparison with stress rupture data from the NIMS data sheet [38]. The model is shown to successfully capture the effect of temperature on the creep life of conventional P91; this is attributed to the inclusion of temperature-dependence in the steady-state creep constants.

3.3 Simulating the effect of Al content on the creep performance of 9Cr steels

The effect of increased Al content in 9Cr steels causes a reduced volume fraction of the finely dispersed MX precipitates, leading to a reduced creep strength and, hence, decreased rupture time of the material [17]. Bar 257 is a P91 alloy with a high Al content, on the order of 0.028 wt.% [17], which has attracted some attention due the detrimental effect of the increased Al content on rupture life. In the present model, the effect of increasing Al content is simulated by reducing the volume fraction of MX precipitates, $f_m$. This leads to a concomitant reduction in the mean inter-particle spacing of MX precipitates (via Equation (7)) and hence, a decrease in mean obstacle spacing, $\lambda_{ob}$, via Equation (12). The subsequent predicted effect on MSR is presented in Figure 10 for a temperature of 650 °C. Bar 257 is seen to exhibit a higher MSR for stresses above ~20 MPa, with the effect predicted to increase with increasing stress. This reduced effect in the (lower stress) diffusional creep regime can be attributed to the reduced effect of MX precipitates in diffusion-based creep.

Figure 11 presents a comparison between the predicted and measured creep strain response in a Bar 257 alloy at 650 °C, for 70 MPa and 82 MPa, using the physically-based reduction in $f_m$. The key outcome here is the demonstrated ability to predict rupture life, based on
composition induced alterations to the microstructure. Figure 12 shows the predicted and measured effect of increasing Al content in Gr. 91 alloys for 70 MPa at 650 °C. A creep life reduction factor of ~5.8 is predicted for Bar 257, which is consistent with the test data. The proposed model is thus shown to be capable of predicting this detrimental effect of increasing Al content, highlighting the benefit of the physically-based precipitate modelling methodology. The predicted effect on stress-life rupture curves for Bar 257 and P91, at both 600 °C and 650 °C, is shown in Figure 13. Bar 257 is predicted to have a significantly lower creep life compared to conventional P91 across the full range of stress examined and for both temperatures.

The more general sensitivity of the model to increasing Al content on rupture life, via varying $f_m$ and $\lambda_{ob}$ independently, is presented in Figure 14. Figure 14a shows for $f_m$ of 0.1 %, the predicted rupture life is reduced by a factor of ~3 compared to conventional P91 steel. Figure 14b shows the effect of increasing Al content via the $\lambda_{ob}$ parameter and the effect of such changes on the predicted time to failure, with an increasing $\lambda_{ob}$ value predicted to result in substantial reductions to creep life. However, neither variation of $f_m$ nor $\lambda_{ob}$ alone can predict the measured effect of increasing Al content, as presented in Figure 10, highlighting the requirement for the more mechanistic approach presented here, including separating the effect of $M_{23}C_6$ and MX precipitates and defining the $\sigma_0$ creep constant from a physical basis.

4 Discussion

This paper presents a new CDM modelling methodology to predict microstructural degradation associated with different precipitate types in 9Cr steels. In particular, this new approach considers, for the first time, the key strengthening and degradation mechanisms of these different precipitate types in the well-established model of Hayhurst and co-workers.
The new methodology is able to predict the creep response of P91 alloys across a range of temperatures and stresses, as illustrated in Figure 6 to Figure 8, as well as rupture life as presented in Figure 9.

The inclusion of temperature-dependence in the model, in particular for the steady-state creep constants, is shown to be able to predict creep deformation, with only a small number of temperature-dependent variables required. However, the temperature-dependent variables $K_k$ and $C$ are determined using physical arguments and measured data, with the temperature-dependent Young’s modulus easily defined from monotonic tensile test data. The temperature-dependence of the $C$ variable is consistent with ductility values at 600 °C presented in Klueh [36]. The decreasing $C$ values with increasing temperature are also consistent with an increasing ductility with temperature. The final remaining temperature-dependent constant, $H^*$, is related to the back-stress in the material due to the key strengthening mechanisms of precipitates, laths and subgrains and immobile dislocations [21,22].

As illustrated in Figure 5, the use of a hyperbolic sine modelling methodology enables reliable extrapolation and interpolation of the MSR across a broad range of applied stresses. The relationship between microstructure and creep constants, presented here in Equation (2) and Equation (12), is key to the accurate prediction of the secondary creep response across a range of temperatures. This result also highlights the necessity to include microstructure in a modelling methodology for 9Cr steels to accurately predict creep behaviour.

Although the model predicts the creep response to a relatively high degree of accuracy in the 70 to 125 MPa stress regime, at higher stresses, the proposed modelling methodology appears to breakdown, over predicting the creep life at an applied stress of 175 MPa (see Figure 6a).
It is argued here that this is possibly due to a change in primary mechanism of deformation under different stress regimes, further highlighting the requirement to include a more physically-based primary creep modelling methodology. In particular, hardening associated with microstructural evolution and the key mechanisms of strengthening should be modelled using more physical models, such as proposed by Magnusson and Sandström [21], Spigarelli [22] or Barrett and co-workers [39]. It is proposed here that this modelling methodology should include the following key processes: (i) dislocation density evolution, (ii) martensitic lath to subgrain formation and subsequent subgrain coarsening, (iii) solid solution strengthening, and (iv) formation of new secondary phase precipitates. For instance, under higher stress regimes, the role of inelastic deformation-controlled subgrain formation and dislocation density evolution should be incorporated in the model. For lower applied stresses, the formation of Laves phase particles at the expense of solid solution strengthening mechanisms represents a key process for longer-term creep deformation [5,6], with precipitate deformation representing the key strengthening mechanism at intermediate stresses, as presented in the current work.

Figure 11 presents application of the proposed model to Bar 257, including a comparison with creep test data from the literature [8]. Although the key point of this comparison is to demonstrate the ability of the approach to predict the detrimental effect of MX carbonitride volume fraction on rupture life for Bar 257, as compared to P91, it is worth noting that the model underpredicts strain at short times for the higher stress (82 MPa) and slightly overpredicts strain at long times for the lower stress (70 MPa). These deviations may be attributed to the fact that the primary and creep damage (cavitation) models do not explicitly account for MX volume fraction. Future work will investigate the development of a more physically-based primary creep model, which explicitly accounts for evolution of dislocation
density (and hence effects of MX volume fraction, for example) and stress-dependent subgrain coarsening, to potentially give further improvements in accuracy for creep strain.

The results highlight the requirement to account for chemical composition and heat treatment when evaluating the performance of candidate materials. It is widely known that increasing the Al content has a detrimental effect on the high temperature creep performance of 9Cr steels, with a creep life reduction factor of 5.8 presented in Figure 12. This comparison clearly highlights the key role of precipitates, and in particular, of MX precipitates on the creep performance of 9Cr steels. Hence, it is necessary to (i) ensure strict control of chemical composition and careful heat treatment procedures are in place during component manufacture to optimise precipitate strengthening and (ii) develop material models capable of capturing the key effects of such precipitates. In the present work, it is found that reducing the volume fraction of MX carbonitrides, consistent with the formation of AlN nitrides at the expense of the more thermally stable VN nitrides, only captures this effect to a certain extent. The effect of increased mean spacing of obstacles associated with increasing Al content (as per Equation (12) with a reduced volume fraction of MX precipitates) on the creep constant, $\sigma_0$, must also be included to accurately capture this effect.

5 Conclusions

This paper presents a continuum damage mechanics creep model for 9Cr steels, specifically designed to predict the key effects of precipitate-strengthening due to different precipitate types and microstructure-sensitivity of minimum creep strain-rate. The model is applied here to P91 steel for $M_{23}C_6$ and MX precipitates, but the method is more generally applicable to multiple precipitate types, and indeed other alloy systems. The key conclusions are:
• The three state variable continuum damage mechanics model with multiple precipitate types is capable of predicting the creep deformation of 9Cr steels, incorporating the key role of MX carbonitrides on creep strength.

• Thermal aging and strain-induced precipitate coarsening of MX and M\textsubscript{23}C\textsubscript{6} precipitates represent key strengthening mechanisms to resist creep deformation in 9Cr steels. MX strengthening is dominated by thermal aging, with strain-induced coarsening of M\textsubscript{23}C\textsubscript{6} carbides necessary to predict the microstructural degradation and creep deformation in 9Cr steels.

• The multi-precipitate continuum damage mechanics model is capable of predicting the detrimental effect of increasing aluminium content, and associated reduction of MX precipitate volume fraction, on creep deformation and rupture life of 9Cr steels.

• MX carbonitride volume fraction is identified as a critical parameter for creep strength and life assessment in 9Cr steels operating in the target temperature range (above 600 °C) and in intermediate to higher applied stress regimes (approximately in the 70 to 100 MPa range). Aluminium content should be minimised to approximately 0.01 wt.% to ensure the formation of an optimum volume fraction of MX carbonitrides for high creep strength.

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Figure 1: Hierarchical microstructure of 9Cr steels highlighting precipitate distributions.

Figure 2: Schematic of (a) vacancy diffusion to grain boundaries and (b) void formation through vacancy coalescence.
Figure 3: Schematic of precipitate coarsening with resulting dislocation bowing and loop formation in a material (a) prior to creep test and (b) post creep test.

Figure 4: Identification of thermal coarsening rates of (a) $M_2\text{C}_6$ and (b) MX precipitates in 9Cr steels from the measured data of Hald and Korcakova [29] at temperatures of 600 °C and 650 °C.
Figure 5: Identification of the steady-state creep constants from the MSR data of Haney et al. [40].

Figure 6: Comparison of predicted (a) creep strain response and (b) creep-rate response of conventional P91 at 600 °C with experimental data of Orlová et al. [9].
Figure 7: Comparison of predicted creep response with the experimental data of Hyde et al. [8] at 625 °C.

Figure 8: Predicted and experimentally observed creep response for conventional P91 steel at 650 °C. The 70 MPa data is from the work of Abd El-Azim et al. [41] and the 98 MPa data is taken from Maruyama et al. [42].
Figure 9: Comparison of P91 predicted stress rupture life with experimental data. The experimental data is taken from the NIMS data sheet [36].

Figure 10: Predicted effect of Al content on the minimum creep rate at a temperature of 650 °C.
Figure 11: Experimental [8] and model predicted creep curves for Bar 257 at 650 °C.

Figure 12: Comparison of model and experimental data for Bar 257 and conventional P91 steel at 650 °C and an applied stress of 70 MPa. The Bar 257 data is from Hyde and co-workers [8] and conventional P91 data from Abd El-Azim et al. [39]
Figure 13: Predicted creep rupture plot for P91 and Bar 257 alloys at (a) 600 °C and (b) 650 °C.

Figure 14. Predicted effect of (a) reducing the volume fraction of MX precipitates and (b) varying the mean spacing of obstacles, $\lambda_{ob}$, on time to failure in P91 alloys for an applied stress of 90 MPa at 625°C.
Table 1: Chemical composition of P91 steels in wt.%. The P91 steel composition ranges are obtained from Abd El-Azim et al. [39] Magnusson and Sandström [17] and Orlová et al. [9], with the Bar 257 data taken from Magnusson and Sandström [17]. Note that trace elements such as Cu and S are not presented. The balance is Fe.

<table>
<thead>
<tr>
<th>Material</th>
<th>Al (wt.%)</th>
<th>C (wt.%)</th>
<th>Cr (wt.%)</th>
<th>Mn (wt.%)</th>
<th>Mo (wt.%)</th>
<th>N (wt.%)</th>
<th>Nb (wt.%)</th>
<th>Ni (wt.%)</th>
<th>P (wt.%)</th>
<th>Si (wt.%)</th>
<th>V (wt.%)</th>
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<tbody>
<tr>
<td>P91</td>
<td>0.007 - 0.09</td>
<td>8.17 - 0.46</td>
<td>0.86 - 0.05</td>
<td>0.16 - 0.014</td>
<td>0.2 - 0.194</td>
<td>0.01 - 0.104</td>
<td>8.36 - 0.56</td>
<td>0.90 - 0.065</td>
<td>0.47 - 0.02</td>
<td>0.34 - 0.23</td>
<td>0.06 - 0.075</td>
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<tr>
<td>Bar 257</td>
<td>0.028</td>
<td>0.09</td>
<td>8.34</td>
<td>0.49</td>
<td>0.89</td>
<td>0.03</td>
<td>0.07</td>
<td>0.47</td>
<td>-</td>
<td>0.34</td>
<td>0.23</td>
</tr>
</tbody>
</table>

Table 2: Temperature-dependent material parameters.

<table>
<thead>
<tr>
<th>T (°C)</th>
<th>E (GPa)</th>
<th>H' (°)</th>
<th>$\bar{C}$ (°)</th>
<th>$K_c$ (mm²/hr)</th>
<th>$K_m$ (mm²/hr)</th>
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</thead>
<tbody>
<tr>
<td>600</td>
<td>144</td>
<td>0.45</td>
<td>8.5</td>
<td>1.0×10⁻¹⁵</td>
<td>2.35×10⁻¹⁸</td>
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<td>625</td>
<td>125</td>
<td>0.131</td>
<td>7.5</td>
<td>3.68×10⁻¹⁵</td>
<td>3.13×10⁻¹⁸</td>
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<tr>
<td>650</td>
<td>110</td>
<td>0.09</td>
<td>2.5</td>
<td>6.35×10⁻¹⁵</td>
<td>3.90×10⁻¹⁸</td>
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</tbody>
</table>

Table 3: Temperature-independent material parameters.

<table>
<thead>
<tr>
<th>Material</th>
<th>$\dot{\varepsilon}_0$ (1/hr)</th>
<th>$\Delta E$ (J)</th>
<th>h (MPa)</th>
<th>$\lambda_{sh}$ (nm)</th>
<th>$d_c$ (nm)</th>
<th>$d_m$ (nm)</th>
<th>$f_c$ (-)</th>
<th>$f_m$ (-)</th>
<th>$\phi_c$ (mm³)</th>
<th>$\phi_m$ (mm³)</th>
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<tbody>
<tr>
<td>P91</td>
<td>6×10¹³</td>
<td>5.978×10⁻¹⁹</td>
<td>0.07E</td>
<td>31</td>
<td>36</td>
<td>100</td>
<td>0.02</td>
<td>0.005</td>
<td>1.04×10⁻⁹</td>
<td>1.18×10⁻¹⁴</td>
</tr>
<tr>
<td>Bar 257</td>
<td>6×10¹³</td>
<td>5.978×10⁻¹⁹</td>
<td>0.07E</td>
<td>38</td>
<td>36</td>
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<td>0.001</td>
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