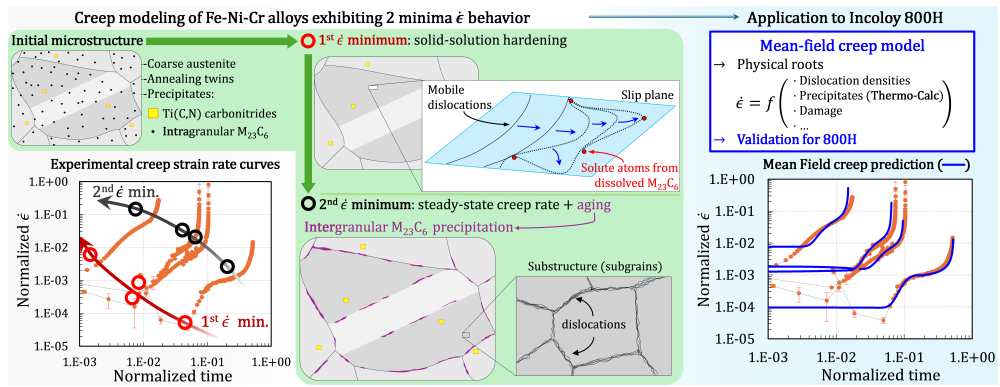




Mean-field prediction of transient solid-solution hardening creep deformation response in Fe-Ni-Cr alloy

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Abstract

High-temperature creep deformation is often accompanied by microstructure evolution. However, aspects affecting the creep properties of materials such as substructure formation and kinetics of primary and secondary precipitates are

not addressed by conventional models. In this work, a microstructure-dependent mean-field model is adapted to predict the non-classical creep deformation response of materials exhibiting solid-solution hardening regimes. Numerical novelties include an improved stress-sensitive apparent activation volume for the dislocation glide velocity, and a diffusion-based solid-solution stress formulation through the internal stresses of the material. An Incoloy 800H experimental campaign to identify and validate the model combining creep tests and microstructural characterization techniques was performed. Additionally, ex-situ thermokinetic simulations on Thermo-Calc[®] PRISMA module were conducted to recover the physical evolution of precipitates during creep. The numerical predictions are in good agreement with the experimental results for low-stress and high-temperature loadings. The loss of predictability observed at high-stress levels is attributed to a creep mechanism transition. Exact stress and temperature ranges cannot be openly disclosed for confidentiality reasons.

Keywords: High temperature creep, Solid-solution hardening, Particle strengthening, Cr₂₃C₆ thermokinetics, Incoloy 800H

1 Introduction

The continuous improvement in the efficiency and productivity of industrial processes in sectors such as power generation, metallurgy and aerospace [1] often relies on new technologies where higher operational temperatures are needed. In this context, resistance to the creeping viscous deformation observed in metallic components subjected to a combination of high-temperature and a constant mechanical load remains a major design aspect [2]. To satisfy these new requirements, researchers may explore the suitability of commercially available alloys [3, 4], develop new alloys [5, 6], or develop heat treatments aimed at enhancing the mechanical properties of commercially available alloys [7, 8]. In any case, accurate assessment of the creep performance during testing and service life is challenging. New creep testing methods such as indentation [9], miniaturized specimens [10, 11] or accelerated creep tests through the stepped iso-stress method [12] have been explored to reduce the costs of material characterization. Nevertheless, uniaxial tensile creep tests remain the most popular creep characterization technique. They consist of the application of a constant uniaxial load under high constant temperature (following either the ISO-204 [13] or ASTM E139 [14] standards). Conventional creep characterization campaigns covering wide stress and temperature ranges are, however, lengthy and costly. In this context, the development of reliable inter- and extrapolation methods that minimize experimental input to predict creep deformation and service life is a scientific challenge shared among industrial and academic researchers. Despite their popularity, conventional stress, temperature, and [time-to-rupture](#) (t_r) extrapolation methods such as the Larson-Miller parameter or the Manson-Haferd parameter assume a single creep mechanism and a constant microstructure per master curve. In general, simpler models fail to recover the creep behavior of materials that exhibit creep mechanism transition or microstructure evolution.

Under sufficiently high temperature and mid-to-low stress regimes, some austenitic steels exhibit complex creep behavior characterized by two consecutive local minima in the creep strain rate history. The second creep rate minimum corresponds to the conventional steady-state creep stage, whereas the first peak is often a consequence of complex microstructural transformations related to precipitate kinetics. For instance, [15] concluded that the non-classical regime observed in 2.25Cr-1Mo steel was the consequence of an Orowan-type particle-strengthening mechanism involving MX precipitates. More recently, [16] investigated the non-classical response of 18Cr-9Ni-3Cu-Nb-N at 873 K and 280 MPa by means of ex-situ microstructural characterization of material samples at different creep stages. The authors attributed the first local minimum to an intragranular NbX precipitate strengthening phenomenon, whereas the second minimum was ascribed to a grain boundary precipitate strengthening effect driven by the intergranular precipitation of Cr_{23}C_6 carbides associated with conventional aging.

Providing a balanced cost/weight ratio and good creep and corrosion resistance, alloys such as 800H, 800Rk and 800HT have experienced high popularity given their admissibility for the manufacturing of critical nuclear power plant components [17–19]. Investigations seeking to ascertain its applicability for temperatures roughly above 850°C were, however, limited by non-classical creep behavior. The first local creep strain rate minimum has been attributed to a [solid-solution hardening \(SSH\)](#) phenomenon linked to the incomplete precipitation of Cr_{23}C_6 carbides during [water quenching \(WQ\)](#) [20, 21], whereas the second minimum is consistent with a conventional steady-state creep rate; a delicate balance between dislocation production and annihilation. Incoloy 800H is hereafter chosen as the case study given its timeless industrial relevance and the modeling challenge it represents.

Accurate prediction of complex creep regimes involving microstructure evolution requires a numerical model in which the stress-strain dependency on dislocation and precipitate interactions is established. In this context, dislocation density-based [mean-field \(MF\)](#) models are a particularly prominent technique. They compute the microstructure dependent stress-strain mechanical response of materials by solving an equation system consisting of two main building blocks. The first is a mechanistic hardening model, such as those by E. Orowan [22] and the Bailey-Hirsh law [23], where the critical resolved shear strain and critical resolved shear stress dependency on the mean dislocation density (ρ) are established. The second block consists of the exploitation of ρ as a state variable. To express microstructure evolution within the equation system, variational ρ equation terms representing positive (e.g., dislocation production and multiplication) and negative (e.g., annihilation and immobilization) contributions are introduced. Altogether, this semi-physical framework provides ρ -based [MF](#) models with the ability to predict numerous microstructure-dependent inelastic phenomena. For instance, [24] achieved physical prediction of stress, strain and temperature assisted dynamic recrystallization by establishing a mean stored dislocation density threshold in addition to equations related to grain nucleation and growth. Such an innovative technique was used as a complement to cellular automaton simulations [25] or to predict the continuous inelastic deformation response and/or

flow stress of materials during hot forming processes [26, 27]. Multiscale approaches incorporating ρ formalisms into crystal plasticity models have also been explored [28].

In recent decades, creep-oriented ρ -based MF models have become increasingly popular among the scientific and industrial community because of the physically reliable creep deformation and service life predictions they provide. MF models developed during this period can be classified as static or continuum. Static creep models describe the stress and temperature dependency of a single feature (e.g., the **steady-state creep strain rate** ($\dot{\epsilon}_{ss}$)) for a steady microstructure [29], whereas continuum creep models describe the time-dependent creep deformation history for an evolving microstructure [30]. In view of the microstructural instabilities addressed in this work on the high-temperature creep of 800H alloy, the latter technique was chosen.

The creep model developed in this work follows the continuum MF modeling framework established in 1990 by N.M. Ghoniem et al. [31], where dependency to multiple mean dislocation density forests in combination with an evolving material substructure is introduced into a differential equation system consisting of 5 partial time-derivatives. Expressions describing ρ mobility and subgrain evolution are adopted from the analytical studies by H.P. Hirth and J. Lothe (lately revisited by P.M. Anderson in 2017 [32]), whereas terms representing dislocation production, annihilation and interaction with other microstructural features are largely inherited from the works by R. Sandström (a lifetime of creep modeling was recently summarized in [33]). The comprehensive MF model proposed by Ghoniem et al. [31] stands out as a versatile numerical framework capable of adapting to multiple materials, microstructures and problem types.

For instance, in 2012, M. Basirat et al. [34] achieved prediction of the full creep deformation response of 9Cr-1Mo steel by neglecting the subgrain evolution state variable, but introducing a climb-stress based SSH term, and a creep **continuum damage mechanics** (CDM) formulation addressing precipitate, cavitation, and lath depletion damage evolution. Later, S.D. Yadav et al. published on the application of creep MF models adapted to predict the creep behavior of a new 12% Cr [35] and a P92 [36] steel in 2016 and 2018 respectively. In addition to a CDM addressing precipitate coarsening and cavitation damage evolution, the authors compute the precipitate evolution through Mat-Calc [37] thermo-kinetic simulations. Noteworthy time-dependent ρ -based MF model applications to alloys such as new G115 steel [38], RAFM steel [39], austenitic stainless steel 304HCu [30], IN718 superalloy [40], and steel 316LN [41] were published during the 2020s. In particular, the successful applications by F. Riedlsperger et al. to steel alloys P91 [42], P92 [43], and Ni-alloy 617 [44] have led the research team to consolidate a proper MF software called CreeSo [45]. A recent exploitation of the new CreeSo development environment by L. Witzmann et al. [4] showcases good numerical creep predictions for a 10% Cr steel in addition to remarkable mean dislocation density validation.

The present work deals with the development, implementation and validation of a creep-oriented ρ -based MF model belonging to the broad Ghoniem family introduced here above. It is adapted to account for the microstructure-dependent non-classical deformation response experimentally observed for 800H during low-stress high-temperature loadings [3, 46–48]. The creep MF model is presented in section 2. An experimental campaign aimed at studying the non-classical creep deformation

response and microstructure evolution of alloy 800H is described in section 3. A brief summary of the parameter identification procedure, as well as the physical parameters used for this alloy, is described in section 4. The numerical validation of the model and the numerical creep and damage prediction results are summarized in section 5. A brief discussion addressing the main contributions presented in this work is provided in section 6. Finally, section 7 contains the conclusions and prospects.

2 Mean-field creep model

The creep-oriented MF model implemented in this work establishes the creep strain rate ($\dot{\epsilon}$) dependency on 3 field state variables: the mean mobile (ρ_m), static (ρ_s), and boundary (ρ_b) dislocation density forests. The corresponding non-linear equation system is presented in Equation (1). The labels throughout Equation (1b) and Equation (1c) are assigned to facilitate the identification of dislocation interaction phenomena later on. In particular, "DWs" represents dislocation walls, whereas "SR" and "DR" represent static and dynamic recovery, respectively.

$$\dot{\epsilon} = \frac{b\rho_m v_{\text{eff}}}{m_T(1 - D_c)} \quad (1a)$$

$$\begin{aligned} \dot{\rho}_m = & \underbrace{v_g \rho_m (\rho_m^{1/2} + \rho_s^{1/2})}_{\text{Frank-Read source}} + \underbrace{v_g \frac{\beta \rho_s R_{\text{DW}}}{h_{\text{DW}}^2}}_{\text{I}} + \underbrace{v_g \frac{\rho_s}{2r_i}}_{\text{II}} - \underbrace{v_g \frac{\rho_m}{2R_{\text{DW}}}}_{\text{III}} - \underbrace{v_g \frac{\rho_m}{\lambda_{\text{eq}}}}_{\text{IV}} - \underbrace{v_g \frac{\rho_m}{h_{\text{DN}}}}_{\text{V}} \\ & - \underbrace{v_g \delta_{\text{anh}} (\rho_m^2 + \rho_m \rho_s)}_{\text{VI (DR)}} + \underbrace{v_c \frac{\rho_s}{\lambda_{\text{eq}}}}_{\text{VII}} + \underbrace{v_c \frac{\rho_s f_v b}{r_i}}_{\text{VIII}} - \underbrace{8v_c \rho_m^{3/2}}_{\text{IX (SR)}} \end{aligned} \quad (1b)$$

$$\begin{aligned} \dot{\rho}_s = & \underbrace{v_g \frac{\rho_m}{2R_{\text{DW}}}}_{\text{III}} + \underbrace{v_g \frac{\rho_m}{\lambda_{\text{eq}}}}_{\text{IV}} + \underbrace{v_g \frac{\rho_m}{h_{\text{DN}}}}_{\text{V}} - \underbrace{v_g \frac{\rho_s}{2r_i}}_{\text{II}} - \underbrace{v_g \delta_{\text{anh}} \rho_m \rho_s}_{\text{DR}} \\ & - \underbrace{v_c \frac{\rho_s}{\lambda_{\text{eq}}}}_{\text{VII}} - \underbrace{v_c \frac{\rho_s f_v b}{r_i}}_{\text{VIII}} - \underbrace{v_c \frac{\rho_s}{h_{\text{DW}}}}_{\text{SR at DWs}} \end{aligned} \quad (1c)$$

$$\dot{\rho}_b = \underbrace{v_c \left[8(1 - 2\xi) \frac{\rho_s}{h_{\text{DW}}} \right]}_{\star} \quad (1d)$$

Each dislocation density rate equation in Equation (1) incorporates mathematical terms representing physical effects of multiplication, annihilation and transformation of dislocation density forests embedded in an evolving microstructure. These are

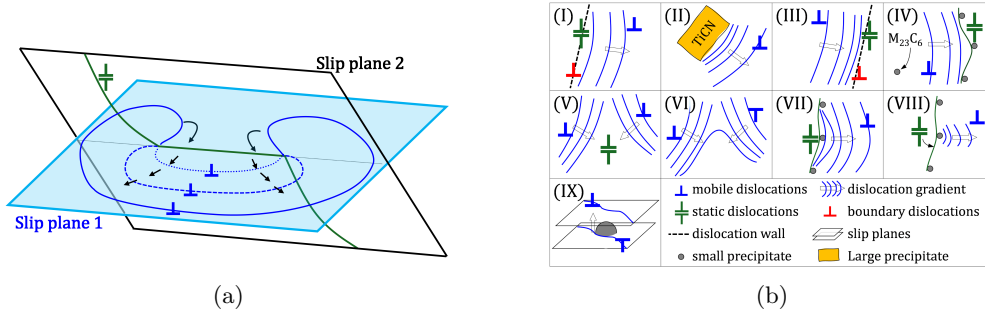


Fig. 1: Graphical representation of the multiple dislocation-precipitate interactions included in our MF model. (a) Mobile dislocation production following the Frank-Read source theory [49]. The mathematical expression in Equation (1b) incorporating dependency to ρ_m and ρ_s is recovered from [44]. (b) Illustrations of multiple creep-related microstructural phenomena according to the Roman numerical labeling system in Equation (1).

illustrated in Fig. 1b according to their roman numerical labels assigned throughout Equation (1b) and Equation (1c).

A brief description of the mathematical formula for each mean dislocation density type illustrated in Fig. 1b is offered below.

- I : emission of ρ_m at DWs. In the formula, R_{DW} is the mean radius of DWs [31]. The proportion of ρ_s dislocations participating in this depinning process is calculated as $\beta\rho_s$, where $\beta = 0.0375$ [31]. h_{DW} is the dislocation spacing ratio at the DW level and is calculated in Equation (2) [42].
- II : fraction of ρ_m produced at non-coherent precipitate-matrix boundaries that act as dislocation sources. r_i represents the mean radius of these precipitates. [44].
- III : fraction of ρ_m absorbed at DWs during creep deformation.[31].
- IV : fraction of ρ_m that become immobilized (ρ_s) upon finding a precipitate in their gliding path. λ_{eq} represents the mean particle spacing ratio of the precipitates participating in the process.[44].
- V : probability of ρ_m becoming immobilized upon encountering an equal-sign dislocation in their gliding path. The average mean dislocation spacing ratio (h_{DN}) is calculated by solving Equation (3).[44].
- VI : ρ_m annihilation probability following the encounter of two ρ_m of equal sign. δ_{anh} is the length parameter for spontaneous dislocation annihilation [31].
- VII : fraction of ρ_m produced from climbing [44].
- VIII : short ρ_m dislocation segments produced at small precipitates. In the formula, proposed by [44], f_v is the phase fraction of the precipitates participating in the process, and b is the Burgers' vector.
- IX : probability of dislocation annihilation during climbing [31].

★ : production of ρ_b from the static population. The subtracted quantity 2ξ ($\xi \ll 1$) represents the proportion of ρ_s annihilated in the process. After [31].

$$h_{\text{DW}} = [(\rho_s + \rho_b)R_{\text{DW}}]^{-1} \quad (2)$$

$$h_{\text{DN}} = 0.8 \cdot (\rho_m + \rho_s)^{-1/2} \quad (3)$$

Our MF creep model presented in Equation (1) is at its core a wisely chosen assembly of previous scientific works. Our contributions (namely the improvements enabling the prediction of the SSH regime, the enhanced stress sensitivity of 800H and the addition of a virtual inelastic work-based damage initiation criterion) are all introduced within Equation (1a) of our system: the Orowan equation. It describes the creep strain rate ($\dot{\epsilon}$) dependency on the field state variables, with m_T being the Taylor factor, D_c the equivalent creep damage, and v_{eff} the mean effective velocity of ρ_m . The latter is calculated as the geometrical summation of both glide (v_g) and climb (v_c) velocities [43], as shown in Equation (4).

$$v_{\text{eff}} = \left[\frac{1}{v_c} + \frac{\pi}{2} \sum_i (r_i^3 N_i) \frac{1}{v_g} \right]^{-1} \quad (4)$$

The equations describing v_g and v_c are shown in Equation (5) and Equation (6) respectively. They follow a thermodynamical interpretation that involves thermal and mechanical activation energy dependence [32]. In Equation (5), K_g is the pre-exponential factor [42], Q_g is the thermal activation energy for dislocation glide, k_B is the Boltzmann constant, T is the temperature in K, V_r is the apparent activation volume for dislocation glide, and σ_i and σ_a are the internal and applied (true) stresses, respectively. For Equation (6), Ω_{FCC} is the atomic size of the face centered cube (FCC) lattice per atom, and K_{cl} and K_{cp} are pre-exponential factors representing the lattice and pipe diffusion contributions respectively (see [42]).

$$v_g = 2K_g \exp\left(\frac{-Q_g}{k_B T}\right) \exp\left(\frac{-\sigma_i V_r}{k_B T}\right) \sinh\left(\frac{\sigma_a V_r}{k_B T}\right) \quad (5)$$

$$v_c = [K_{\text{cl}} + K_{\text{cp}}] \exp\left(\frac{-\sigma_i \Omega_{\text{FCC}}}{k_B T}\right) \sinh\left(\frac{\sigma_a \Omega_{\text{FCC}}}{k_B T}\right) \quad (6)$$

σ_a is calculated for a perfectly uniaxial deformation state and volume conservation as Equation (7a), where σ_0 is the initial (engineering) stress, ν is the temperature-dependent Poisson ratio and ϵ is the current plastic (creep) strain calculated in the model. σ_i refers to the internal mechanical stresses preventing free dislocation movement, a mechanical threshold that reduces the material ductility. In this work, Equation (7b) is calculated as the sum of the dislocation pile-up stress (σ_T) and SSH (σ_{ss}) contributions. σ_T is computed by solving the Bailey-Hirsch law [23] (more commonly referred to as the "Taylor stress") Equation (7c), where α is a material constant,

G is the shear modulus, and dependency on the square root of ρ_m is established for FCC alloys in [50].

$$\sigma_a = \sigma_0 (1 + \nu\epsilon)^{-2} \quad (7a)$$

$$\sigma_i = \sigma_T + \sigma_{ss} \quad (7b)$$

$$\sigma_T = \alpha m_T G b \sqrt{\rho_m} \quad (7c)$$

2.1 Solid-solution hardening

In this work, the SSH contribution to the inelastic deformation response is computed as a stress effect. σ_{ss} is calculated as the product of the strengthening function $f_{ss} = f_{ss0} \exp(\dots)$ and the solute atom concentration $\mathcal{X}_i^{2/3}$, where i represents the atom type found in solution, and the exponent 2/3 is optimized for FCC materials [51]. The Arrhenius expression governing f_{ss} in Equation (8) follows the interpretations of the kink-pair formation (KPF) model [52], inasmuch as it reduces the σ_{ss} calculation of the breakaway probability for a single dislocation segment. The form of f_{ss} in Equation (8) follows the proposal by [53, 54]. With only minor modifications, their model has seen successful applications for Inconel 718 in annealed [55] and "without-precipitate" configurations [27]. In Equation (8), C_1 is a fitting parameter, ΔE_b is the energy barrier for dislocation breakaway, σ_y is the temperature-dependent yield stress, and m is the hardening exponent. Our set of equations, guided by experimental observations, is described below.

$$\sigma_{ss} = f_{ss} \mathcal{X}_i^{\frac{2}{3}} = f_{ss0} \exp \left[\frac{-1}{C_1} \frac{k_B T}{\Delta E_b} \ln \left| \left(\frac{\sigma_a}{\sigma_y} \right)^m \right| \right] \mathcal{X}_i^{\frac{2}{3}} \quad (8)$$

The pre-exponential factor f_{ss0} is calculated in Equation (9), where ϵ_{eq} is the equivalent misfit parameter.

$$f_{ss0} = \frac{1}{\sqrt{2}} M_T G (\epsilon_{eq})^{\frac{4}{3}} \quad (9)$$

ϵ_{eq} is calculated in Equation (10) as the geometrical equivalent of the contributions of both the shear-modulus misfit η_i and the solute atom size misfit δ_i [56], where a is the solvent unit cell parameter; an experimental constant [57].

$$\epsilon_{eq} = \left[\left(\frac{\eta_i}{1 + 0.5|\eta_i|} \right)^2 + a^2 \delta_i^2 \right]^{1/2} \quad (10)$$

In Equation (8), ΔE_b represents the threshold energy barrier necessary for dislocations to surmount solute atoms. Equation (11a) is an extension of the one used in [55] and [27], where w/b is the mean atom travel distance ratio. The additional $\phi(T)$ function in Equation (11b) is introduced to recover the temperature dependency observed during model optimization for alloy 800H, where R is the gas constant and φ_1 and φ_2 are fitting parameters.

$$\Delta E_b(T) = \phi(T) M_T G b^3 \left(\frac{w}{b}\right)^{\frac{3}{2}} \sqrt{\frac{\sigma_a}{M_T G}} \quad (11a)$$

$$\phi(T) = \varphi_1 \exp\left(\frac{-\varphi_2}{RT}\right) \quad (11b)$$

The dissolution of submicron-sized Cr_{23}C_6 precipitates that nucleate at dislocations during **WQ** (see for instance [46, 58]) leads to the production of solid-solution Cr and C. A quick calculation involving the proportion and size differences between the two elements (see Supplementary material) suggests that Cr atoms are responsible for more than 98% of the total volume fraction of the particles produced from the dissolution of these carbides. Therefore, further σ_{ss} calculations involve exclusively Cr atoms. Inspired by the analytical works of [59], the form of the stress-assisted diffusion model of Cr atoms $\mathcal{X}_{\text{Cr}}$ is presented in Equation (12), where d_{AGS} is the average grain size (diameter), D_{eff} is the effective diffusion coefficient for Cr diffusion during **SSH**, t is the time, b_5 is a fitting parameter, and $\mathcal{X}_{\text{Cr},0}$ is the maximum Cr solute atom concentration participating in the **SSH** process. The latter is a temperature-dependent physical quantity recovered from ex-situ thermokinetic simulations (see section 4.3).

$$\mathcal{X}_{\text{Cr}} = \frac{\mathcal{X}_{\text{Cr},0}}{2} \left[1 + \operatorname{erf}\left(\frac{d_{\text{AGS}}}{2\sqrt{D_{\text{eff}} t}} + b_5\right) \right] \quad (12)$$

The temperature and stress dependency of D_{eff} is established in Equation (13a) by numerical optimization. D_{ss}^0 is a pre-exponential factor, Q_{eff} is the activation energy, and b_4 is a material constant. The temperature dependency of Q_{eff} is described by the Arrhenius exponential function shown in Equation (13b), where b_1 , b_2 and b_3 are material parameters.

$$D_{\text{eff}}(T) = D_{\text{ss}}^0 \exp\left(\frac{-Q_{\text{eff},i}}{RT}\right) \exp\left(-b_4 \frac{\sigma_y}{\sigma_a}\right) \quad (13a)$$

$$Q_{\text{eff},i} = b_1 \left[1 + b_2 \exp(b_3/T) \right] \quad (13b)$$

2.2 Apparent representative volume

The stress-sensitivity of the model is largely determined by the apparent representative volume (V_r). Although similar creep-oriented **MF** models treat V_r as a fitting parameter varying within a reasonable physical range [44], a single V_r proved unsuitable for recovering the stress-sensitivity of 800H during creep. The independent magnitudes of V_r optimized to recover the $\dot{\epsilon}_{ss}$ across our experimental creep curves via inverse modeling depict an inversely proportional relationship to σ_a . The V_r dependency on σ_a was thereby recovered by the Arrhenius formula presented in Equation (14).

$$V_r(\sigma_a) = A_V \left[1 + B_V \exp(C_V \cdot \sigma_a) \right] \cdot b^3 \quad (14)$$

2.3 Creep damage

The addition of a **CDM** enables the prediction of the progressive loss of load carrying capacity and creep rupture time of the material [34, 35]. The creep damage denominator $(1 - D_c)$ in Equation (1a) is calculated by solving Equation (15), where D_i is the damage contribution of the i^{th} damage mechanism acting in our creep case study.

$$(1 - D_c) = \prod_i (1 - D_i) \quad (15)$$

The **MF** creep model presented in this work includes only two damage mechanisms: cavitation damage (D_{cav}) and precipitate coarsening damage (D_{ppt}). D_{cav} is defined by Equation (16). The temperature-dependent constant $A(T)$ in Equation (16a) is expressed as an Arrhenius-type function in Equation (16b), where A_1 , A_2 and A_3 are fitting parameters.

$$\dot{D}_{cav} = A(T) \dot{\epsilon} \quad (16a)$$

$$A(T) = A_1 \left[1 + A_2 \exp(T/A_3) \right] \quad (16b)$$

Precipitate damage (D_{ppt}) in 800H is associated with the massive precipitation and coarsening of $Cr_{23}C_6$ at non-coherent grain boundaries [58, 60, 61]. Equation (17a) recovers the D_{ppt} attributed to the j^{th} precipitate type following the particle-spacing ratio formulation by C.Ó. Murchú et al. [62], where $\lambda_{j,0}$ and λ_j are the initial and current average spacing ratios of the precipitate j respectively. The spacing ratio of a precipitate j is calculated by solving Equation (17b) and Equation (17c) as proposed by B. Sonderegger & E. Kozeschnik in [63]. In the case of multiple precipitates, the total D_{ppt} is calculated as in Equation (15).

$$D_{ppt,j} = 1 - \frac{\lambda_{j,0}}{\lambda_j} \quad (17a)$$

$$\lambda_j = \left[\frac{\ln(3)}{2\pi \sum_j (N_j \cdot r_j)} + (2r_A)^2 \right]^{1/2} - 2r_A \quad (17b)$$

$$r_A = \sqrt{\frac{2}{3}} \cdot \frac{\sum_j (N_j \cdot r_j^2)}{\sum_j (N_j \cdot r_j)} \quad (17c)$$

Additionally, our model incorporates a simplified (uniaxial) damage initiation criterion [64]. The accumulated plastic work formula is computed by Equation (18), where m_w is a fitting parameter. Damage is initiated once a user-defined critical threshold of virtual plastic work w_{crit} is reached.

$$w_s(\sigma, \epsilon) = \int_0^t \left(w_{iso}(\sigma, \epsilon) + w_{kin}(\sigma, \epsilon, \dot{\epsilon}) \right) dt \approx \sigma_a \epsilon^{1/m_w} \quad (18)$$

Table 1: Standard Fe-balanced stoichiometric composition of alloy 1.4959 (wt.%), after [65].

*: $0.46 \leq (\text{Al} + \text{Ti}) \leq 1.07$ wt.%

** : $30.09 \leq (\text{Ni} + \text{Co}) \leq 34.78$ wt.%

constituents	Ni **	Cr	C	Si	Mn	S	Ti *	Al *	Co **	Cu
min	29.4	19.0	0.05	0.32	0.50	0.001	0.21	0.19	0.028	0.006
max	34.45	21.5	0.10	0.74	1.34	0.017	0.59	0.58	0.76	0.070

3 Experimental campaign

Incoloy 800H material was acquired in as-received condition in the form of 20 mm thick sheets. A heat treatment **SA+WQ** consisting of **solution annealing (SA)** for 30 min at 1150°C followed by **WQ** preceded the hot rolling process. The standard chemical composition of the material is given in Table 1.

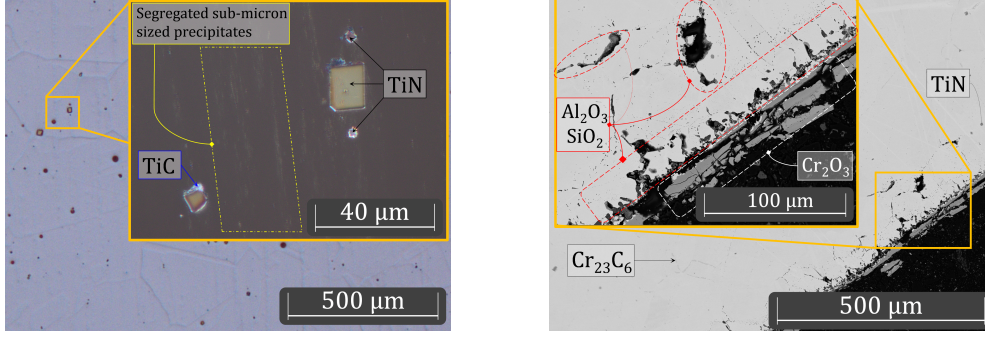
Material samples were extracted along the rolling direction of the sheets using a Charmilles Robofil **electric discharge machine (EDM)** model 310. These were destined to an experimental campaign combining mechanical creep tests and microstructural characterization techniques. A summary of the experimental results relevant for the application and validation of the **MF** creep model is given below.

3.1 Microstructure observations

The material microstructure is studied via an Olympus BX60 **Light Optical Microscopy (LOM)** coupled with an OLYMPUS UC30 CCD digital camera and equipped with a motorized stage together with the OLYMPUS STREAM Motion® software. The material samples were first ground and polished down to 1 µm prior to super finishing with **Oxide polishing suspension (OPS)** down to 0.3 µm on a Struers Tegramin 30 automatic device. Complementary **LOM** techniques include chemical etching with glyceresia (2 parts glycerol + 4 parts hydrochloric acid (HCl) + 1 part nitric acid (**HNO₃**)) to highlight carbide-decorated **grain boundaries (GBs)**, and light polarization to highlight precipitates. **Scanning electron microscopy (SEM)** analyses were carried out using a Tescan Clara SEM associated with **Back-Scattered electron (BSE)** for the images and a 129 eV BRUKER **energy dispersive x-ray spectrometer (EDS)** with a silicon drift detector. The system was connected to ESPRIT® software for pre- and post-treatment of the data. The accelerating voltage for the SEM analyses was set at 15 kV, with a minimum scanning step size of the electron beam of 1 µm.

As-Received (AR) material:

The **AR** microstructure shows coarse austenitic grains with abundant annealing twins and small micron and submicron-sized precipitates scattered throughout the lattice. Low-density titanium carbo-nitrides (**Ti_x(C, N)**) appearing dark in **BSE** images [48] are differentiated into **TiN** and **TiC** via in-situ polarized **LOM** analyses, as shown in Fig. 2a. Owing to the resolution limitations of our **SEM** machine, submicron-sized precipitates (**TiN** and **M₂₃C₆**, observed via **transmission electron microscopy (TEM)** by [58, 66]) could not be identified. Nevertheless, their presence can be correlated to



(a) Qualitative identification of precipitates in AR material via in-situ polarized LOM.

(b) BSE image showing oxide penetration in 800H after 2 months of high-temperature creep testing in air.

Fig. 2: Microstructural characterization (a) AR and (b) creep-affected microstructures of Incoloy 800H.

the bright reflection of polarized light at segregation lines (i.e., the agglomeration of submicron-sized precipitates formed during the hot-rolling process of the material) observed during LOM, as shown in Fig. 2a.

Finally, digital post-treatment of LOM images was performed using the open-source software ImageJ [67]. The grain aspect ratio was measured at 1.3 with no particular geometrical orientation. An austenite grain size of $135.7 \pm 30 \mu\text{m}$ was calculated by applying the linear intersection method according to the standard ASME E112-13 [68].

Creep-affected material:

Material samples from specimens affected by up to 2 months under high-temperature creep in air were subjected to similar microstructural analyses. A progressively greater number of Cr-rich M_{23}C_6 precipitates nucleating at GBs were observed in the samples from longer creep tests, which is consistent with aging. No grain coarsening was detected. Fig. 2b showcases the loss of the characteristic Cr_2O_3 protective layer [1, 69] as well as oxide penetration in the form of nucleation and coarsening of Si and Al oxides. These observations are consistent with those presented in [48]. The precipitation of oxides during creep is attributed to high-temperature exposure to air under sea-level atmospheric pressure conditions. In contrast with the observations presented in [48, 58], the samples analyzed herein showed no evidence of nitridation. The absence of nitrides is explained mainly by the short duration of the high-temperature creep tests in air.

3.1.1 Creep behavior

A total of 11 mechanical creep tests were conducted on a Zwick/Roell (Z/R) Kappa 100 DS machine equipped with a high-inertia resistance furnace MP00850. Creep strain was measured at the uniaxial gauge length of the specimen by means of a high-precision

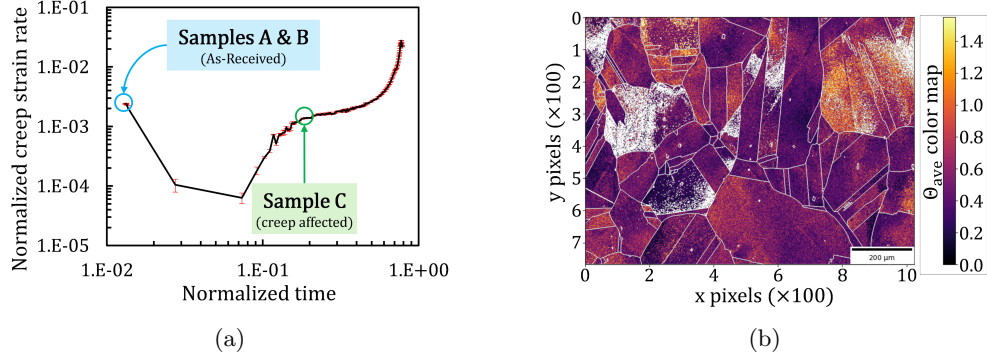


Fig. 3: (a) Normalized creep strain rate vs. time curve showing the source of samples A, B and C. Sample C is not extracted from a creep test interrupted after the $\dot{\epsilon}_{ss}$ was reached. (b) KAM map calculated from EBSD measurements for sample A (AR condition).

ceramic extensometer. To enable the SSH-driven minimum creep strain rate regime to take place, standard creep test (SCT)s experiments were primarily conducted under a combination of low-stress and high temperature loadings. In addition, the loading application methodology instructed in ISO-204 [13] was modified by decreasing the pre-load temperature holding time from 60 to 15 minutes [60]. Although specific stress-temperature regimes cannot be openly disclosed, they range between 10–80 MPa and 700–1050°C.

3.1.2 Dislocation density estimation

Experimental characterization of the geometrically necessary dislocation density (ρ_{GND}) was conducted on material samples under both AR (samples A & B) and creep-affected (sample C) conditions. In particular, sample C was obtained from a single creep test interrupted after reaching the $\dot{\epsilon}_{ss}$ (Fig. 3a). To calculate the ρ_{GND} , a kernel average misorientation (KAM) map was first constructed on the basis of the Nye-tensor approximation methodology for local misorientation of neighboring points detected by Electron backscattered diffraction (EBSD) (Fig. 3b). The ρ_{GND} is then estimated from the slope of the linear fit of the accumulated frequency of angular local misorientation (Θ_{ave}) as a function of the step size (length scale defined by the beam size definition and grid during EBSD) [70]. The ρ_{GND} values estimated for the three samples along with the corresponding misorientation gradient, background noise and R^2 error are given in Table 2. The grain map, KAM map and Θ_{ave} identification for specimens A, B and C are provided in the supplementary material file.

Table 2: Summary of results of the EBSD-based ρ_{GND} measurements (Fig. 3) for samples A, B (AR condition) and C (creep test interrupted at $\dot{\epsilon}_{ss}$).

	sample A	sample B	sample C
Estimated ρ_{GND} (m^{-2})	2.33×10^{11}	1.71×10^{11}	2.19×10^{12}
$\Theta_{\text{ave}} \times 10^{-6}$ ($\text{deg.} \cdot \text{m}^{-1}$)	0.0032	0.0025	0.0318
Background noise (deg.)	0.5651	0.5689	0.5802
R ² -error	0.9913	0.9918	0.9924

4 Input parameters

4.1 Initial state variables

The boundary conditions necessary to solve the non-linear equation system of our MF model are hereafter defined as the vector $W_0 = [\epsilon_0 \quad \rho_{m,0} \quad \rho_{s,0} \quad \rho_{b,0}]$, where ϵ_0 is the initial creep strain ($\epsilon_0 = 0.0$) and $\rho_{m,0}$, $\rho_{s,0}$ and $\rho_{b,0}$ are the initial mobile, static and boundary dislocation densities, respectively. $\rho_{s,0}$ is approximated as the 10% of $\rho_{m,0}$, as seen in [44], whereas $\rho_{b,0}$ is calculated in Equation (19) as a function of the R_{DW} and the mean DW misorientation angle [71].

$$\rho_{b,0} = \frac{3\Theta_{\text{DW}}}{2bR_{\text{DW}}} \quad (19)$$

Finally, a constant magnitude of $R_{\text{DW}}=250$ nm is used in this work. This value is estimated from the numerous TEM experimental observations of the AR hot-rolled microstructure of 800H alloy [72–74].

4.2 Physical and fitting parameters

4.2.1 Diffusion coefficients

Diffusion in solids is modeled as a thermally-activated process by means of the Arrhenius equation shown in Equation (20a), where D_i^0 and Q_i are the pre-exponential factor and activation energy for diffusion in a medium i , where $i = \text{l}$ (lattice), GB, or p (pipe). Following a similar approach to that proposed by [75], the parameters D_i^0 and Q_i to calculate D_{l} and D_{GB} are computed on the basis of the weighted average of those measured for Fe, Ni and Cr in an 800-type alloy. The weighted average function is shown in Equation (20b), where x is the property under evaluation and w_i are the relative molar mass of the alloy per element ($w_{\text{Fe}} = 0.335$, $w_{\text{Ni}} = 0.353$ and $w_{\text{Cr}} = 0.312$). A summary of the parameters used to calculate D_{l} and D_{GB} following this approach is given in Table 3.

$$D_i(T) = D_i^0 \exp\left(\frac{-Q_i}{RT}\right) \quad (20a)$$

$$x_{\text{eq},800\text{H}} = x_{\text{Fe}} \cdot w_{\text{Fe}} + x_{\text{Ni}} \cdot w_{\text{Ni}} + x_{\text{Cr}} \cdot w_{\text{Cr}} \quad (20b)$$

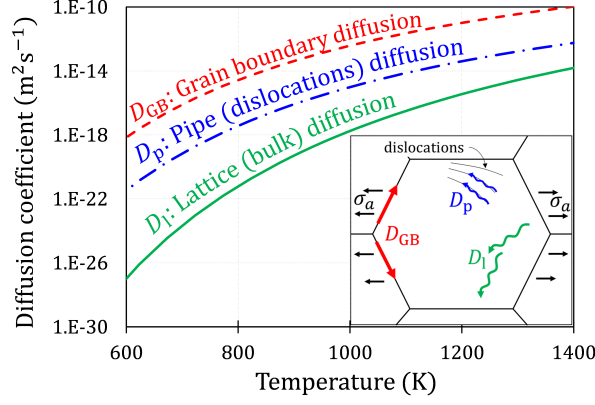


Fig. 4: Temperature dependency of D_{GB} , D_p and D_l used in this work, calculated after the methods of [75, 77–79]. The arrows inside the plot illustrate the vacancy diffusion under load (σ_a).

Table 3: Pre-exponential and activation energy parameters calculating D_l and D_{GB} for 800H alloy.

Parameter	Magnitude per γ lattice constituent			weighted average
	FCC-Fe (after [78])	FCC-Ni (after [77])	FCC-Cr (after [75])	
D_l^0 (m^2s^{-1})	3.260×10^{-5}	8.620×10^{-5}	3.240×10^{-4}	1.424×10^{-4}
Q_l (Jmol^{-1})	2.596×10^5	2.559×10^5	2.874×10^5	2.670×10^5
D_{GB}^0 (m^2s^{-1})	1.880×10^{-5}	3.820×10^{-5}	5.800×10^{-5}	3.788×10^{-5}
Q_{GB} (Jmol^{-1})	1.607×10^5	1.564×10^5	1.842×10^5	1.665×10^5

The pre-exponential and activation energy parameters to calculate D_p are extracted from [76]: $D_p^0 = 4.50 \times 10^{-5} \text{ m}^2\text{s}^{-1}$ and $Q_p = 1.85 \times 10^5 \text{ Jmol}^{-1}$). The temperature dependency of the three diffusion coefficients is depicted in Fig. 4.

4.2.2 Material elastoplastic constants

Elastic constants:

The shear modulus G and Poisson ratio ν of the material were introduced into the MF model as temperature-dependent polynomial functions $f(T)$ of the form shown in Equation (21). The equation and its fitting parameters (A, B, C and D; given in Table 4) are part of the quasi-static hot tensile material characterization conducted for Incoloy 800H by [80].

$$f(T) = A \cdot [1 + B \cdot (T - T_0) + C \cdot (T - T_0)^2 + D \cdot (T - T_0)^3] \quad (21)$$

Table 4: Fitting parameters for Equation (21) describing the temperature dependent shear modulus (G) and Poisson ratio (ν) for Incoloy 800H. Material characterization and parameter identification were performed by [80]. T is in $^{\circ}\text{C}$, and $T_0=20^{\circ}\text{C}$.

Property	A	B ($^{\circ}\text{C}^{-1}$)	C ($^{\circ}\text{C}^{-2}$)	D ($^{\circ}\text{C}^{-1}$)
G	76.74 ± 0.16 (GPa)	-2.50×10^{-4}	-4.32×10^{-7}	3.40×10^{-10}
ν	0.297 ± 0.0022 (-)	-6.05×10^{-4}	2.66×10^{-6}	-2.11×10^{-9}

Table 5: Optimized parameters for the polynomial function Equation (22) describing the temperature-dependency of the 0.2% offset σ_y for Incoloy 800H. Experimental characterization after [81, 82]

α_0 (MPa)	α_1 (MPa $^{\circ}\text{C}^{-1}$)	α_2 (MPa $^{\circ}\text{C}^{-2}$)	α_3 (MPa $^{\circ}\text{C}^{-3}$)	α_4 (MPa $^{\circ}\text{C}^{-4}$)
2.43×10^2	-3.14×10^{-1}	3.88×10^{-4}	5.69×10^{-12}	-2.60×10^{-10}

Yield stress:

Considering the temperature-dependent 0.2% offset yield stress characterization of Incoloy 800H at various temperatures conducted independently by [81, 82], σ_y is defined as a 4th order temperature-dependent polynomial function. It is presented in Equation (22), where T is the temperature in $^{\circ}\text{C}$, $T_0=20^{\circ}\text{C}$, and α_i are fitting parameters. The latter are identified via simple [root-mean squared percentile error \(RMSPE\)](#) minimization and are given in Table 5.

$$\sigma_y(T) = \sum_{i=0}^4 \alpha_i \cdot (T - T_0)^i \quad (22)$$

4.2.3 Physical and fitting parameters

The good balance between the accuracy and predictability of [MF](#) models is attained by combining both physical properties and reliable phenomenological formulations. To this end, correct identification of both physical and fitting parameters is paramount. The physical parameters reported in literature are given in Table 6, whereas the fitting parameters are listed in Table 7. To ensure physical consistency, the latter were identified via inverse modeling following a hierarchical optimization strategy (see section 4.4).

4.3 Complementary thermokinetic simulations

To consider the effect of precipitate evolution on the creep deformation response, this [MF](#) model requires the time and temperature-dependent precipitate history to be introduced as input data. That is, the mean radius r_i and number density N_i of precipitates and their corresponding nucleation sites. This information is outsourced

Table 6: List of physical parameters with literature sources.

Parameter	Description	Magnitude	Source
η_{Cr}	SSH: Cr-FCC shear modulus misfit	1.7541 –	[55]
α_{FCC}	SSH: FCC geometrical constant	16 –	[57]
δ_{Cr}	SSH: Cr-FCC size misfit	0.0298 –	[55]
w/b	SSH: mean atom travel distance ratio	1.70 –	[55]
d_{AGS}	Average grain size (diameter)	150 μm	[48]
a_g	Unit cell size (FCC)	0.356 nm	[83]
b	Burgers' vector (FCC Fe-Ni-Cr)	0.253 nm	[84]
α	Constant σ_T , optimized for 800H alloy	0.6 –	[50]
m_T	Taylor factor, optimized for FCC	3.06 –	[84]
β	Factor for dislocation production at GBs	0.0375 –	[31]
δ_{anh}	Dislocation annihilation coefficient	5 –	[31]
η_v	Coefficient for transfer of defects into jogs	1.00×10^{-6} –	[31]
Ω_{FCC}	Transfer parameter of dislocations into jogs	$1.21 \times 10^{-29} \text{ m}^3$	[31]
Θ_{DW}	Mean misorientation angle of DWs	5°	[74]

from ex-situ thermokinetic simulations conducted with **Thermo-Calc** PRISMA module, using the TCFE13 and MOBFE8 databases. These calculations, intended to recover the kinetics of precipitates in 800H during creep, correspond to isothermal simulations with no stress dependency (i.e., equivalent to aging). Such an approximation is admissible given the low stress levels addressed in this research.

4.3.1 Thermokinetic simulation methodology

The initial number density distribution of the precipitates in Incoloy 800H is largely determined by the mandatory **SA+WQ**. In view of the influence of these precipitates (in particular, **Cr₂₃C₆** nucleating at dislocations [58]) on the **SSH** behavior of this material, accurate determination of the initial precipitate distribution is paramount. To this end, a thermokinetic simulation that replicates the **SA+WQ** process shown in Fig. 5a was first performed. The resulting **number density distribution (NDD)** of the primary precipitates obtained after this simulation are shown in Fig. 5b.

Given the high stability and rapid precipitation of **Ti_x(C,N)** and to ensure numerical convergence in TC-PRISMA, only the initial **NDD** of **Cr₂₃C₆** is considered for the subsequent creep/aging thermokinetic simulations conducted in TC-PRISMA. A total of 5 isothermal simulations were conducted at: 700°C, 800°C, 900°C, 1000°C, and 1050°C. The evolution of the mean radius and number density of **Cr₂₃C₆** precipitates for an isothermal thermokinetic simulation at 1000°C for a total of $1 \times 10^5 \text{ h}$ (≈ 10 years) is shown in Fig. 6. The predicted coarsening of intergranular **Cr₂₃C₆** precipitates (i.e., at grain boundaries, grain edges and grain corners) is consistent with aging, whereas the dissolution of those nucleating at dislocations is the source of the **SSH** phenomenon addressed in this work. The mean radius and number density of precipitates during **MF** creep simulations are continuously calculated internally by means of time and temperature dependent linear interpolations.

The tabulated *.csv files containing our thermokinetic simulation results for 800H alloy can be made available upon request.

Table 7: List of fitting parameters of the model and their calibration target in hierarchical order. While parameter magnitudes cannot be openly disclosed for confidentiality reasons, the physical definition and calibration target for each parameter are provided to facilitate model reproducibility.

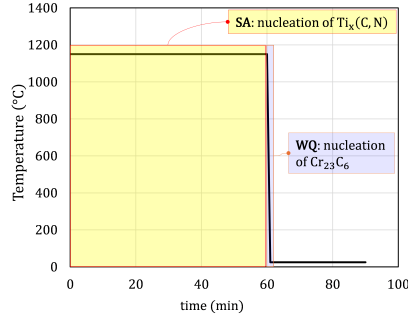
Calibration target	Parameter	Description	Unit
1. Steady-state creep rate ($\dot{\epsilon}_{ss}$)	a_1	Pre-exponential factor for v_g	ms^{-2}
	Q_g	Thermal activation energy for v_g	J
	A_V	Parameters for V_r in σ_{ss} :	—
	B_V	$V_r(\sigma_a) = A_V [1 + B_V \exp(C_V \cdot \sigma_a)] \cdot b^3$	—
	C_V		Pa
2. Solid-solution hardening (SSH)	C_1	f_{ss} – Equation (8)	—
	m		—
	φ_1	ΔE_b – Equation (11)	—
	φ_2		J
	b_1		J
	b_2	Parameters for diffusion of Cr Equation (12) & Equation (13)	—
	b_3		K
	b_4		—
	b_5		—
3. Tertiary creep stage (damage)	m_w	Stored plastic energy:	—
	w_{crit}	$w_s(\sigma_0) = \sigma_a \epsilon^{1/m_w}$	—
	A_1	Cavitation damage parameters:	—
	A_2	$\dot{D}_{cav} = A_1 (1 + A_2 \exp(T/A_3)) \cdot \dot{\epsilon}$	—
	A_3		—

4.4 Parameter uniqueness and identification methodology

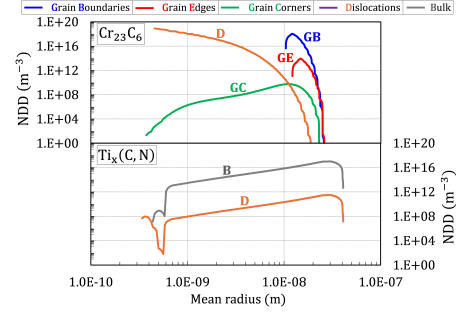
Given the dimensionality of the constitutive framework, an identification strategy was adopted to ensure physical consistency. First, a base model entitled to the case study alloy is defined by constraining the precipitate kinetics information inherited from ex-situ thermokinetic simulations (see section 4.3), physical parameters (see Table 6) and calibration of initial state variables (see section 4.1). Then, the inverse modeling of the creep deformation response addressing the individual contributions of $\dot{\epsilon}_{ss}$, SSH and damage is conducted. This is:

1. Identification of $\dot{\epsilon}_{ss}$ by optimization of parameters for v_g (a_1 , Q_g), and V_r (A_V , B_V , C_V).
2. Identification of SSH by optimization of parameters for f_{ss} (C_1 , m), ΔE_b (φ_1 , φ_2), and Cr-diffusion (b_i).
3. Identification of damage parameters for damage initiation (w_{crit} , m_w) and cavitation (A_i).

This hierarchical approach ensures that each parameter subset remains physically constrained to its respective deformation mechanism.

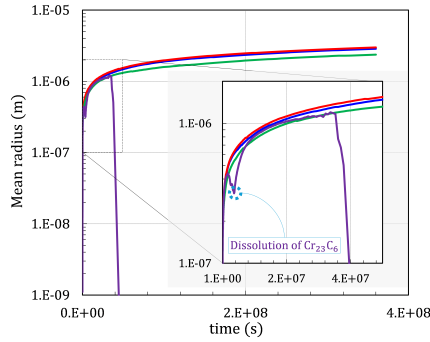


(a) Simulated SA (60 min at 1150°C) followed by WQ.

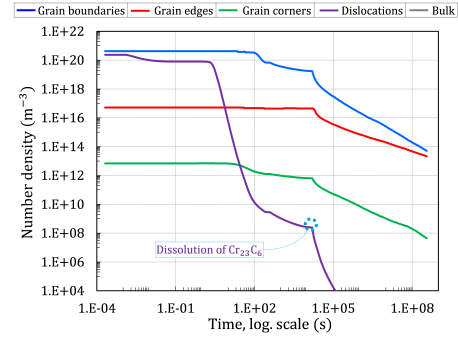


(b) NDD of Cr_{23}C_6 and $\text{Ti}_x(\text{C}, \text{N})$ nucleating during the SA+WQ, calculated with TC-PRISMA.

Fig. 5: Thermokinetic simulation of SA+WQ (Solution-Annealing followed by Water-Quenching) process conducted with Thermo-Calc PRISMA module for Incoloy 800H.



(a)



(b)

Fig. 6: Mean radius (a) and number density (b) history of M_{23}C_6 precipitates per nucleation site calculated after isothermal thermokinetic simulation conducted with Thermo-Calc PRISMA module at 1000°C for 1×10^5 h.

5 Results and validation

5.1 Creep deformation response:

A visual validation of the creep deformation predictions of the developed MF model is given in Fig. 7, where three experimental creep curves at equal temperatures and different stress levels are compared to their numerical counterparts. The exact scale information cannot be provided for confidentiality reasons.

The apparent prediction inaccuracies observed at the very beginning of the creep strain rate curves shown in their normalized form in Fig. 7b stem from the simplified stress-assisted diffusion model. The functions chosen to describe the time, stress and

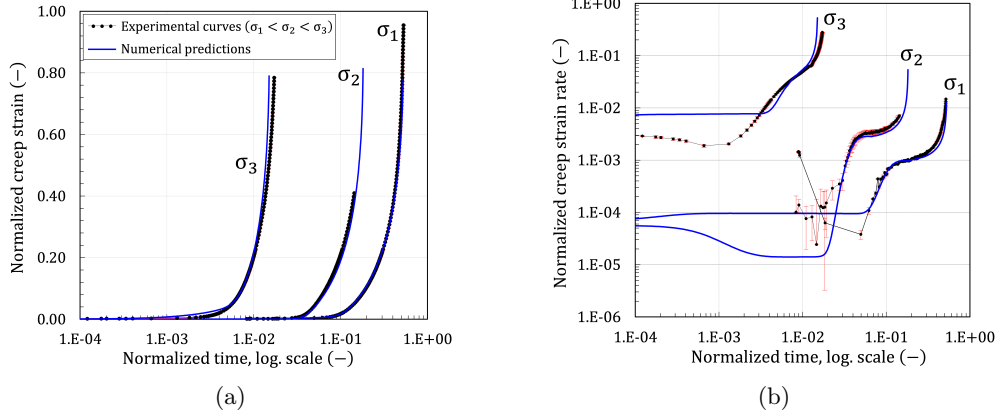


Fig. 7: Superposition of experimental and numerical creep strain (a) and creep strain rate (b) curves for Incoloy 800H for a single temperature and three different stress levels ($\sigma_1 < \sigma_2 < \sigma_3$). All the creep curves, with the exception of those at σ_2 , were conducted until rupture.

temperature dependent SSH (see section 2.1) are suited to recover the creep strain rate hardening effect within an order of magnitude of accuracy. Despite this approximation, the low level of creep deformation attained during the initial SSH creep strain rate regime leads to this error becoming negligible in the log term. Consequently, the SSH-induced creep hardening predicted in the creep strain vs. time curves in Fig. 7a is close to the experimental results.

Our MF model is identified and optimized for high-temperature and low-stress creep loadings. In the context of creep modeling, the applicability of the model is limited to a stress-temperature range whereupon a ubiquitous combination of creep mechanisms is involved. Venturing on stress levels falling outside this regime entails facing different creep mechanisms where new equations should be required. Let us illustrate this point by comparing the numerical predictions for two experimental creep curves at equal temperatures and different stress levels in Fig. 8. Both the creep strain and the creep strain rate benchmarks indicate that the model provides reasonable predictions for low(σ_4) stress levels. Conversely, the numerical curves obtained for the high stress case ($\sigma_5 \approx 2\sigma_4$) fail to reach the experimental response, dangerously underestimating the deformation response of the material.

During high stress creep deformation (see σ_5 in Fig. 8b), the $\dot{\epsilon}_{ss}$ is found to be an order of magnitude above than that predicted by the MF model. This finding indicates that creep is driven by a deformation mechanism that differs from that considered in Equation (1). In addition, no SSH effects are observed.

The creep deformation error calculated for those tests where the numerical predictions were consistent with the experimental results (e.g., σ_1 to σ_4 in Fig. 7 and Fig. 8) shows that for large creep strains $\epsilon \geq 5\%$, the model error is largely within a reasonable range: between 15% of underestimation and 5% of overestimation. Considering

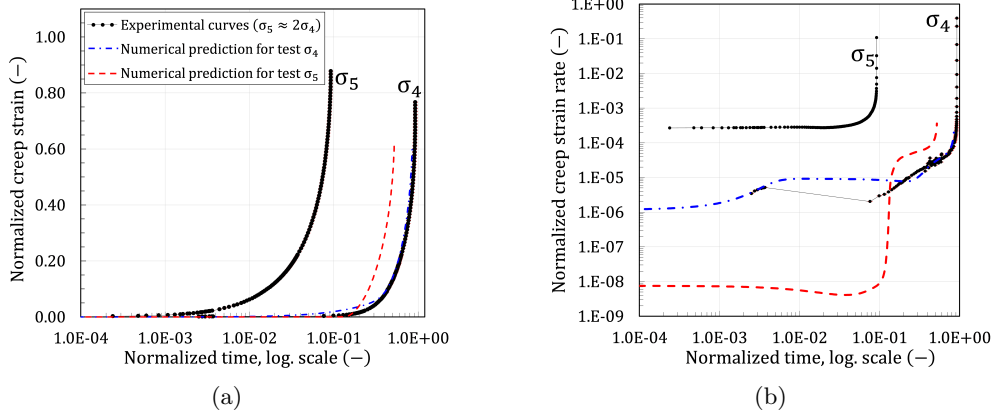


Fig. 8: Superposition of experimental and numerical creep strain (a) and creep strain rate (b) curves for Incoloy 800H for a single temperature and two different stress levels where $\sigma_5 \approx 2 \times \sigma_4$. All the creep tests presented here were conducted until rupture.

that experimental creep curve scattering is often found within a $\pm 10\%$ error, in addition to the fact that the model appears to exhibit a largely conservative error margin (i.e., underestimating the t_r), predictions for high-temperature low-stress loadings are validated. MF creep simulations can therefore be used for creep data extrapolation and creep rupture prediction.

5.2 Dislocation density evolution:

Given the inherent difficulties associated with accurately measuring the ρ_{GND} and the inability to effectively identify the respective contributions from ρ_m , ρ_s and ρ_b , validation between the numerical and experimental time-dependent evolution of these dislocation density types is possible only from a semi-quantitative standpoint [27, 44]. To this end, the ρ_{GND} values measured for 800H under AR condition and during creep (see Fig. 3) are compared to the dislocation density types predicted for the same creep testing conditions in Fig. 9. Let us compare the ρ_{GND} measurements to the predicted ρ_m , given the physical relationship between these two quantities [85]. The results show that the ρ_m history is within 1 order of magnitude of the experimental ρ_{GND} . In the context of MF modeling, such a deviation is considered to be in good agreement with the experimental results.

5.3 Damage evolution:

Creep damage (D_c) combines the effects of particle coarsening D_{ppt} and cavitation D_{cav} contributions. It is implemented following a semi-coupled approach, affecting the creep deformation response through the Orowan equation in Equation (1). In addition, an energy-based damage initiation Equation (17) is implemented. The fitting parameters (3 parameters for D_{cav} and the strain exponent m_w , see Table 7) are optimized for low-stress high-temperature creep tests following an inverse RMSPE error minimization

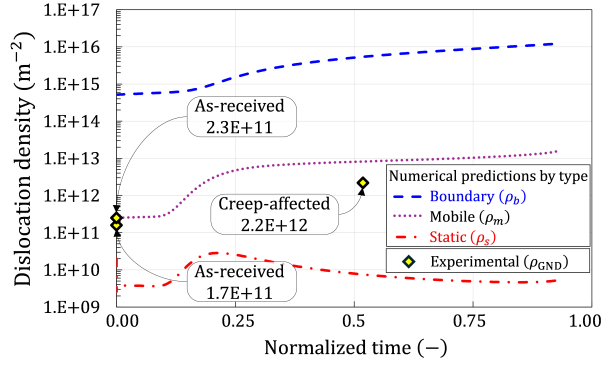
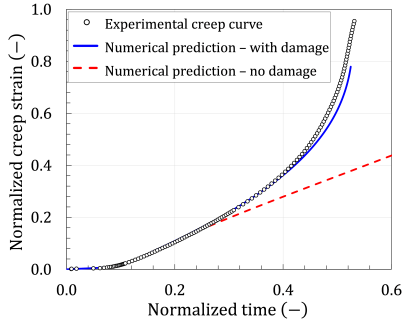
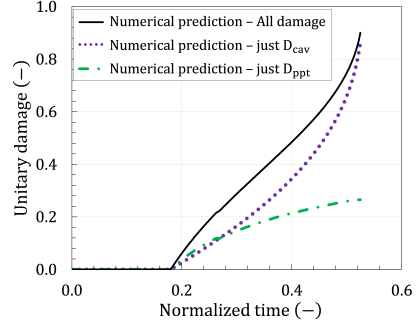


Fig. 9: Numerical prediction of dislocation density evolution by type (ρ_m , ρ_s and ρ_b) and experimental measurements (ρ_{GND}) for the same creep test (see Fig. 3).



(a) Creep strain benchmark for a single creep curve compared to numerical predictions with and without damage.



(b) Time-dependent D_c evolution for the test presented in (a) and contributions from D_{cav} and D_{ppt} .

Fig. 10: Effect of creep damage on the creep deformation response prediction and detailed assessment of the individual contributions from D_{cav} and D_{ppt} .

procedure. In addition, a critical damage level of $D_c^{crit} = 0.9$ is imposed to account for the rupture of the material. The simulation results presented in Fig. 7 and Fig. 8 show that the fracture prediction with these parameters is in good agreement with the experimental results within the stress-temperature applicability range of this model. The effects of individual damage contributions for curve σ_1 from Fig. 7 are illustrated in Fig. 10.

6 Discussion

6.1 About the stress-sensitivity of V_r

The $\dot{\epsilon}_{ss}$ dependency on stress and temperature is the core component of any viscoplastic model. The improved stress dependency of the apparent activation volume (V_r)

developed in this work was the result of several inverse modeling attempts. The optimal magnitude of V_r , identified independently on 4 experimental creep curves at equal temperatures and multiple stress levels, exhibited an inversely proportional Arrhenius-type dependency on σ_a . Similar V_r dependency on the stress has also been reported for 9Cr-1Mo martensitic steel [86], low-carbon steels [87], Ni₃Al polycrystals [88] and 316LN stainless steel [86], substantiating the interest in this generic formulation. Nevertheless, dependency on the strain rate [89], temperature [90], and grain size [91] has also been reported, indicating that the real mathematical form of V_r is strongly material dependent.

A similar Arrhenius-type V_r dependency to σ_a has been recently implemented for the MF modeling of the creep behavior of 30CrMoNiV5-11 martensitic steel by F. Chen et al.[92], further consolidating the applicability and generality of this approach.

Successful $\dot{\epsilon}_{ss}$ prediction entitled to the improved stress-sensitivity of V_r was only achieved for low stresses. The loss of $\dot{\epsilon}_{ss}$ predictability observed for high stresses (see σ_5 in Fig. 8) is attributed to a creep mechanism transition, either toward a diffusion-driven Coble creep or toward grain boundary sliding (GBS).

- The Coble creep theory is supported after the work of L.A. Beardsley et al.[18] in 2019. On the basis of the semi-physical BMD (Bird-Mukherjee-Dorn; see [93]) $\dot{\epsilon}_{ss}$ modeling formalism, the authors use a genetic algorithm to identify the combined diffusion and dislocation creep mechanisms involved in the high-temperature creep behavior of Incoloy 800H. The results provide phenomenological evidence of two dislocation-based (low and high-temperature) creep mechanisms and a single diffusion (Coble, where $\dot{\epsilon}_{ss} \propto \sigma^{-1}$) creep mechanism.

A complementary Coble creep term affecting the strain rate of the material in Equation (1a) could be implemented to improve our MF model predictability. Such an approach has already been implemented by F. Riedlsperger et al.[44] in 2023, and, more recently, by F. Chen et al.[92] in 2026, with outstanding results

- GBS is a plastic deformation mechanism in which grains move relative to each other. This shearing grain movement has been observed to promote cavitation damage during creep in materials with large grain structures (such as FCC in Cu- or Ni-alloys [33]). Its phenomenological interpretation follows the relationship $\dot{\epsilon}_{ss} \propto \sigma^{-2}$.

6.2 About the SSH approach

From a macromechanical perspective, SSH is a strengthening mechanism commonly associated with yield and tensile stress increments with negligible changes in ductility. This is consequence of the incomplete precipitation of alloying elements, which is purposely achieved by a combination of material composition and heat treatment optimization. For instance, the SSH effect attributed to variations in the matrix concentrations of elements such as Si, Cr, Mo, C, or N in stainless steels was determined by [94, 95] via regression.

For 800H alloy, while most of its alloying elements (such as Mg, Mn, Si or C) remain in solid solution state, a peak creep minimum attributed to a SSH mechanism

is observed at high temperatures. This creep strengthening is consequence of the incremental matrix concentration of Cr and C atoms following the rapid dissolution of the intragranular submicron Cr_{23}C_6 . Qualitative observation of the polarized light reflection response from these particles in AR material (see Fig. 2a) is corroborated via TEM by [58, 96]. Early attempts to link the first creep strain hardening peak observed in 800-type alloys to the precipitate strengthening effect of carbides achieved limited success. For instance, B. Reichert et al.[97] used a Kocks-Mecking-Estrin (KME)-based dislocation density model in which precipitate kinetics were included by means of an Ostwald-ripening approach with poor experimental validation. A similar but strongly phenomenological methodology was implemented by E. El-Magd et al.[98] for alloy 800HT with better results. Experimental evidence of a SSH process driving the plastic deformation of 800H alloy was obtained after quasi-static hot tensile test experiments conducted by Y. Cao et al.[21] in 2018, where a clear dynamic strain aging (DSA) regime characteristic of a SSH process is observed. In consideration of these latter findings, the solid-solution creep strain hardening peak exhibited by 800H alloy during low-stress high-temperature creep regimes is included in our MF creep model as a σ_i contribution. The combination of physical and phenomenological terms in the σ_{ss} formula (see section 2.1) allows for straightforward parameter optimization, whereas the Cr physical misfit parameters for Ni-Cr alloys given in [55] allow the finalization of the model set-up with no additional material characterization steps. However, accuracy limitations can be partially attributed to the following simplifications:

- The formula describing σ_{ss} (see Equation (8)) is a function of the Cr concentration in matrix ($\mathcal{X}_{\text{Cr}}$). The solid-solution contribution of the C atoms liberated during the Cr_{23}C_6 dissolution process is neglected because of their small size and number compared to those of Cr (see Supplementary material). Additional alloying elements that could contribute to the strengthening of 800H alloy in the form of solid-solution (e.g., P, Cu, Si, etc.) are disregarded in these calculations. Researchers aiming at more deterministic approaches could explore multiple SSH elements, following, for instance, the works of M.A. Moretti et al.[55] where multiple solid solution contributions are considered.
- Cr concentration function is inspired by diffusion in solids theory. The time-dependent Arrhenius function describing the decrease in $\mathcal{X}_{\text{Cr}}$ (see Equation (12)) is based on the theoretical works of P. Shewmon [59]. However the fitting parameters b_i ($i = 1, \dots, 5$) are purely phenomenological.

6.3 Final remarks

The results within low-stress and high-temperature creep loading regimes (i.e., σ_1 to σ_4 across Fig. 7 and Fig. 8) show that the strain rate hardening prediction is in good agreement with the experimental curves. The simplifications and concessions acknowledged here above are confined within a reasonably physical framework. The MF model developed in this work represents a balanced mesoscale approach where prediction accuracy, parameter uncertainty and numerical complexity are considered. This balance follows the modeling advice found in [99].

Exploitation of the physics-based results obtained from the current MF model can, for instance, complement experimental information gathered from conventional standard creep tests, decreasing the time and resources devoted to intensive creep experimental campaigns. Additionally, accurate predictions of creep scalars such as $\dot{\epsilon}_{ss}$, t_r or time-to-1% creep deformation can be used to feed surrogate models or construct physically-based stress-temperature material selection diagrams, exploiting machine learning approaches.

7 Conclusions

In this work, a dislocation forest based mean-field (MF) model adapted to predict the non-classical creep deformation response observed in 800H austenitic stainless steel was formulated and validated. The key conclusions are as follows:

- The integration of thermokinetic simulation results into the MF formulation establishes a direct, physically consistent link between the thermal history of the alloy and its evolving precipitate radius and number density. This eliminates the need for empirical precipitation parameters and enables a mechanistic description of solid solution hardening (SSH), precipitate strengthening, and coarsening-induced damage.
- The improved stress-sensitivity of V_r and the analytical diffusion-based SSH formulation capture the physical effects involved in the low-stress and high-temperature creep deformation response of 800H. Recent MF results reported in [5] showcased an Arrhenius-like V_r dependency on σ_a is a good indication of the generality of our approach. Nevertheless, more research is needed to corroborate the applicability of both the V_r and σ_{ss} formulations developed in this work.
- The creep prediction deviation observed for high stress is attributed to a transition towards an intergranular type of deformation. This finding is physically consistent with mechanisms such as Coble creep, where the strain rate is controlled by grain-boundary diffusion kinetics, or grain boundary sliding, where deformation is controlled by the shearing movement of grains relative to each other. The current model, focused on dislocation glide and climb mechanisms, systematically underpredicted strain for high stress. Identifying this transition provides a rigorous mechanistic basis for the observed discrepancy and clearly delineates the validity window of the dislocation-based approach. Additional high-stress creep tests and microstructural characterization instances are required to verify the deformation mechanism and extend the model to high stresses. The successful extension of a similar mean-field creep model has been recently reported for a 30CrMoNiV5–11 steel alloy by [92].
- The cavitation and precipitate coarsening damage formulations implemented in the MF model allow simulated creep curves to exhibit a strain softening consistent with the loss of load carrying capacity observed during experimental tests conducted until rupture. Additionally, the damage threshold enables accurate prediction of the final tertiary creep stage and fracture during standard creep tests.

Acknowledgements

All ULiège authors acknowledge the financial support of the European project AID4Greenest HORIZON-CL4-2022-RESILIENCE-01-19. Carlos Rojas-Ulloa acknowledges the FNRS FRIA grants B1 4000-8987 and B2 4002-1907. As Research Director F.R.S. - FNRS, Anne Marie Habraken acknowledges the F.R.S. - FNRS (Fonds de la Recherche Scientifique, Belgium, Walloon Region). Víctor Tuninetti and all ULiège authors acknowledge the international research cooperation agreement WBI/AGCID 2023-2025 RI-02 (DIE23-001).

Additionally, the authors acknowledge the helpful support of Dr. Bernhard Sonderegger, Dr. Florian Riedlsperger and PhD. candidate Laura Witzmann from the Institute for Engineering Materials at the Johannes Kepler University (Linz, Austria), and Dr. Surya Deo Yadav from the Department of Metallurgical Engineering at the Indian Institute of Technology (BHU) (Varanasi, India) for sharing their time and expertise on the capabilities of creep-oriented mean-field models. We also acknowledge Mg.Sc.Eng. Amedeo di Giovanni and Mg.Sc.Eng. Alexandre Venramini for their technical support during our experimental campaign.

Author contributions

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- Jérôme Tchoufang Tchuindjang: Data curation, Investigation, Writing - review & editing.
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- Tuan Nguyen-Minh: Formal analysis, Investigation, Writing - review & editing.
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- Anne Mertens: Supervision, Investigation.
- Víctor Tuninetti: Writing - review & editing
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Declaration of interest statement

The authors declare no conflict of interest.

References

- [1] Li, H., Yan, D., Li, G., Tian, C., An, J., Tan, J., Zheng, W., Du, B., Yin, H.: Corrosion mechanisms and differences of Inconel 617 and Incoloy 800H under high-temperature air ingress accident. *Progress in Nuclear Energy* **184**, 105725 (2025) <https://doi.org/10.1016/j.pnucene.2025.105725> . Accessed 2025-06-03

- [2] Guo, X.-f., Nan, N.-n., Liu, B.-x., Fuyang, C.-m., Gong, J.-m., Jia, H.-l., Tan, X.: An improved creep damage mechanics model and its application to evaluate creep strain design codes for hot outlet manifold in steam reformer furnace. *International Journal of Pressure Vessels and Piping* **198**, 104649 (2022) <https://doi.org/10.1016/j.ijpvp.2022.104649> . Accessed 2022-10-28
- [3] Bagui, S., Mandal, M., Sahoo, B.K., Laha, K., Tarafder, S., Mitra, R.: Investigation of non-classical creep behavior of Inconel 617 alloy at 700 C and 800 C through interrupted tests and microstructural characterizations. *Materials Science and Engineering: A* **832**, 142474 (2022) <https://doi.org/10.1016/j.msea.2021.142474> . Accessed 2023-03-06
- [4] Witzmann, L., Riedlsperger, F., Krnsteiner, P., Kulawinski, D., Hald, J., Krenmayr, B., Groiss, H., Sonderegger, B.: Modelling microstructural evolution and creep behaviour of a novel martensitic 10 % Cr steel: a physics-based mean-field approach. *Materials Science and Engineering: A* **948**, 149377 (2025) <https://doi.org/10.1016/j.msea.2025.149377> . Accessed 2025-11-07
- [5] Chen, J., Zhou, Y., Liu, X., Li, K., Wang, A., Fu, H., Zhang, H., Zhu, Z.: Achieving outstanding strength-ductility synergy in Fe-Ni-Cr-Al-Ti high-entropy alloys via dual heterostructures and nanoscale precipitation strengthening. *Materials Science and Engineering: A* **944**, 148929 (2025) <https://doi.org/10.1016/j.msea.2025.148929> . Accessed 2025-11-05
- [6] Li, W., Chen, Y., Gong, X., Wu, J., Hu, W., Liu, Y., Lian, L.: A new creep prediction model in Nibased superalloys via datadriven symbolic regression. *Materials Genome Engineering Advances* **3**(4), 70026 (2025) <https://doi.org/10.1002/mgea.70026> . Accessed 2026-03-06
- [7] Kim, D.-M., Kim, C., Yang, C.-H., Park, J.-U., Jeong, H.-W., Yim, K.-H., Hong, H.-U.: Heat treatment design of Inconel 740H superalloy for microstructure stability and enhanced creep properties. *Journal of Alloys and Compounds* **946**, 169341 (2023) <https://doi.org/10.1016/j.jallcom.2023.169341> . Accessed 2026-03-06
- [8] Bian, Z., Li, M., Liu, H., Zhang, G., Liu, X., He, Y., Li, J.: Comprehensive optimization of tensile and creep properties of Inconel 718 superalloy at room temperature and elevated temperature through grain boundary engineering treatments. *Materials Science and Engineering: A*, 150051 (2026) <https://doi.org/10.1016/j.msea.2026.150051> . Accessed 2026-03-09
- [9] Khan, S., Junaid, M., Khan, F.N.: Nanoindentation and indentation creep analysis of diffusion bonded of inconel 718 and AISI 304L stainless steel joint interface. *Proceedings of the Institution of Mechanical Engineers, Part L: Journal of Materials: Design and Applications* **239**(8), 1620–1632 (2025) <https://doi.org/10.1177/14644207241301338> . Accessed 2025-11-24
- [10] Schirmer, R.W., Selent, M., Abendroth, M., Kiefer, B.: Identification of creep

- parameters from multistep relaxation tests on miniaturised specimen. PAMM **23**(4), 202300195 (2023) <https://doi.org/10.1002/pamm.202300195> . Accessed 2025-11-24
- [11] Cunha Furtado, H., Martins Dos Santos, F.F., Arajo, L.S., Bueno, L.D.O.: Miniaturized-specimen creep testing: an alternative for evaluating the remaining life of Cr-Mo steel components. Academia Materials Science **1**(3) (2024) <https://doi.org/10.20935/AcadMatSci7354> . Accessed 2026-03-06
 - [12] Cano, J.A., Stewart, C.M.: Accelerated Creep Test Qualification of Creep-Resistance Using the WilshireCanoStewart Constitutive Model and Stepped Isostress Method. Journal of Engineering for Gas Turbines and Power **144**(1), 011016 (2022) <https://doi.org/10.1115/1.4052205> . Accessed 2022-05-19
 - [13] International Organization of Standards: ISO-204:2023 (2023)
 - [14] E28 Committee: Test Methods for Conducting Creep, Creep-Rupture, and Stress-Rupture Tests of Metallic Materials. ASTM International (2025). <https://doi.org/10.1520/E0139-24> . <http://www.astm.org/cgi-bin/resolver.cgi?E139-24> Accessed 2025-11-24
 - [15] Maruyama, K., Sawada, K., Koike, J., Sato, H., Yagi, K.: Examination of deformation mechanism maps in 2.25Cr-1Mo steel by creep tests at strain rates of 10⁻¹¹ to 10⁻⁶ s⁻¹. Materials Science and Engineering: A **224**(1-2), 166–172 (1997) [https://doi.org/10.1016/S0921-5093\(96\)10566-9](https://doi.org/10.1016/S0921-5093(96)10566-9) . Accessed 2025-10-27
 - [16] Hatakeyama, T., Sawada, K., Sekido, K., Hara, T., Kimura, K.: Microstructural factors of the complex creep rate change in 18Cr9Ni3CuNbN steel. Materials Science and Engineering: A **831**, 142225 (2022) <https://doi.org/10.1016/j.msea.2021.142225> . Accessed 2023-06-16
 - [17] Swindeman, R.W., Marriott, D.L., Foulds, J.R.: Extended allowable stress values for alloy 800H. ASME standards technology, LLC (2012)
 - [18] Beardsley, A.L., Bishop, C.M., Kral, M.V.: A Deformation Mechanism Map for Incoloy 800H Optimized Using the Genetic Algorithm. Metallurgical and Materials Transactions A **50**(9), 4098–4110 (2019) <https://doi.org/10.1007/s11661-019-05350-6> . Accessed 2022-05-11
 - [19] Hua, Y., Shi, S., Wang, H., Wang, L., Yue, X., Su, C.: Insight into surface microstructure evolution and fracture mechanism of incoloy alloy 800HT under high-temperature conditions. Journal of Materials Research and Technology **35**, 4709–4722 (2025) <https://doi.org/10.1016/j.jmrt.2025.02.138> . Accessed 2026-03-06
 - [20] Abd El-Azim, M.E.: Correlation between tensile and creep data in alloy 800H at 850C. Journal of Nuclear Materials **231**(1-2), 146–150 (1996) [https://doi.org/10.1016/0022-314X\(96\)00011-1](https://doi.org/10.1016/0022-314X(96)00011-1)

- [21] Cao, Y., Zhang, C., Zhang, C., Di, H., Huang, G., Liu, Q.: Influence of dynamic strain aging on the mechanical properties and microstructural evolution for Alloy 800H during hot deformation. *Materials Science and Engineering: A* **724**, 37–44 (2018) <https://doi.org/10.1016/j.msea.2018.03.074> . Accessed 2022-05-10
- [22] Orowan, E.: Problems of plastic gliding. *Proceedings of the Physical Society* **52**(1), 8–22 (1940) <https://doi.org/10.1088/0959-5309/52/1/303> . Accessed 2023-12-11
- [23] Bailey, J.E., Hirsch, P.B.: The dislocation distribution, flow stress, and stored energy in cold-worked polycrystalline silver. *Philosophical Magazine* **5**(53), 485–497 (1960) <https://doi.org/10.1080/14786436008238300> . Accessed 2025-07-15
- [24] Goetz, R.L., Seetharaman, V.: Modeling Dynamic Recrystallization Using Cellular Automata. *Scripta Materialia* **38**(3), 405–413 (1998) [https://doi.org/10.1016/S1359-6462\(97\)00500-9](https://doi.org/10.1016/S1359-6462(97)00500-9) . Accessed 2026-02-26
- [25] Liu, Y.-X., Lin, Y.C., Li, H.-B., Wen, D.-X., Chen, X.-M., Chen, M.-S.: Study of dynamic recrystallization in a Ni-based superalloy by experiments and cellular automaton model. *Materials Science and Engineering: A* **626**, 432–440 (2015) <https://doi.org/10.1016/j.msea.2014.12.092> . Accessed 2026-03-11
- [26] Buzolin, R.H., Lasnik, M., Krumphals, A., Poletti, M.C.: A dislocation-based model for the microstructure evolution and the flow stress of a Ti5553 alloy. *International Journal of Plasticity* **136**, 102862 (2021) <https://doi.org/10.1016/j.ijplas.2020.102862> . Accessed 2026-03-11
- [27] Kumar, N., Ferraz, F.M.B., Buzolin, R.H., Shahryari, E., Poletti, M.C., Yadav, S.D.: A meso-scale model to predict flow stress and microstructure during hot deformation of IN718WP. *International Journal of Plasticity* **187**, 104271 (2025) <https://doi.org/10.1016/j.ijplas.2025.104271> . Accessed 2025-07-06
- [28] Chen, T., Fu, H., Dong, S., Zhou, Y., Gu, Y., Zhou, C., Sakidja, R.: Predicting the high temperature deformation behavior of Haynes282 by a dislocation-density based crystal plasticity model. *Materials Science and Engineering: A* **923**, 147690 (2025) <https://doi.org/10.1016/j.msea.2024.147690> . Accessed 2026-03-11
- [29] Sandström, R.: Modelling creep properties of alloys without adjustable parameters. *Journal of Materials Science* **60**(39), 18201–18224 (2025) <https://doi.org/10.1007/s10853-025-10996-2> . Accessed 2026-03-11
- [30] Mehrotra, P., Kumar, N., George, A., Sahoo, K.C., Ganesan, V., Ahmadi, M.R., Trivedi, S., Yadav, S.D.: An advanced mean field dislocation density reliant physical model to predict the creep deformation of 304HCu austenitic stainless steel. *Materials Today Communications* **32**, 104128 (2022) <https://doi.org/10.1016/j.mtc.2022.104128>

- [31] Ghoniem, N.M., Matthews, J.R., Amodeo, R.J.: A Dislocation Model for Creep in Engineering Materials. *Res Mechanica* **29**(1), 197–219 (1990)
- [32] Anderson, P.M., Hirth, J.P., Lothe, J.: *Theory of Dislocations*, 3rd edn. Cambridge University Press, ??? (2017)
- [33] Sandström, R.: *Basic Modeling and Theory of Creep of Metallic Materials*. Springer Series in Materials Science, vol. 339. Springer, Cham (2024). <https://doi.org/10.1007/978-3-031-49507-6> . <https://link.springer.com/10.1007/978-3-031-49507-6> Accessed 2025-03-05
- [34] Basirat, M., Shrestha, T., Potirniche, G.P., Charit, I., Rink, K.: A study of the creep behavior of modified 9Cr1Mo steel using continuum-damage modeling. *International Journal of Plasticity* **37**, 95–107 (2012) <https://doi.org/10.1016/j.ijplas.2012.04.004> . Accessed 2023-08-22
- [35] Yadav, S.D., Sonderegger, B., Stracey, M., Poletti, C.: Modelling the creep behaviour of tempered martensitic steel based on a hybrid approach. *Materials Science and Engineering: A* **662**, 330–341 (2016) <https://doi.org/10.1016/j.msea.2016.03.071> . Accessed 2023-12-11
- [36] Yadav, S.D., Scherer, T., Prasad Reddy, G.V., Laha, K., Sasikala, G., Albert, S.K., Poletti, C.: Creep modelling of P91 steel employing a microstructural based hybrid concept. *Engineering Fracture Mechanics* **200**, 104–114 (2018) <https://doi.org/10.1016/j.engfracmech.2018.07.027> . Accessed 2026-02-12
- [37] Kozeschnik, E.: Mean-Field Microstructure Kinetics Modeling. In: *Encyclopedia of Materials: Metals and Alloys*, pp. 521–526. Elsevier, ??? (2022). <https://doi.org/10.1016/B978-0-12-819726-4.00055-7> . <https://linkinghub.elsevier.com/retrieve/pii/B9780128197264000557> Accessed 2026-03-12
- [38] Xiao, B., Yadav, S.D., Zhao, L., Tang, Z., Han, Y., Yang, X., Kai, J.-J., Yang, T., Xu, L.: Deep insights on the creep behavior and mechanism of a novel G115 steel: Micromechanical modeling and experimental validation. *International Journal of Plasticity* **147**, 103124 (2021) <https://doi.org/10.1016/j.ijplas.2021.103124> . Accessed 2026-02-12
- [39] Kumar, N., Joseph, A.S., Mehrotra, P., Yadav, S.D.: An improved dislocation density reliant model to address the creep deformation of reduced activation ferritic martensitic steel. *Forces in Mechanics* **9**, 100117 (2022) <https://doi.org/10.1016/j.finmec.2022.100117> . Accessed 2026-02-12
- [40] Kumar, N., Buzolin, R.H., Shahryari, E., Pradeep, K., Poletti, M.C., Yadav, S.D.: A Meso-Scale Model to Describe the Creep Response and Microstructure

- Evolution of Superalloy IN-718. *Transactions of the Indian Institute of Metals* **78**(2), 46 (2025) <https://doi.org/10.1007/s12666-024-03532-4> . Accessed 2026-02-12
- [41] Nilesh Kumar, Ganesan, V., Vasudevan, M., Yadav, S.D.: Microstructure-Based Creep Strain Modeling of Steel 316LN at 140275 MPa/600650C. *Physics of Metals and Metallography* **126**(5), 581–595 (2025) <https://doi.org/10.1134/S0031918X25600010> . Accessed 2026-02-12
 - [42] Riedlsperger, F., Krenmayr, B., Zuderstorfer, G., Fercher, B., Niederl, B., Schmid, J., Sonderegger, B.: Application of an advanced mean-field dislocation creep model to P91 for calculation of creep curves and time-to-rupture diagrams. *Materialia* **12**, 100760 (2020) <https://doi.org/10.1016/j.mtla.2020.100760> . Accessed 2023-08-21
 - [43] Riedlsperger, F., Zuderstorfer, G., Krenmayr, B., Sonderegger, B.: Application of a physically-based dislocation creep model to P92 for constructing TTR diagrams. *Materials at High Temperatures* **39**(2), 161–166 (2022) <https://doi.org/10.1080/09603409.2022.2037326> . Accessed 2025-08-20
 - [44] Riedlsperger, F., Wojcik, T., Buzolin, R., Zuderstorfer, G., Speicher, M., Sommitsch, C., Sonderegger, B.: Microstructural insights into creep of Ni-based alloy 617 at 700 C provided by electron microscopy and modelling. *Materials Characterization* **198**, 112720 (2023) <https://doi.org/10.1016/j.matchar.2023.112720> . Accessed 2024-06-21
 - [45] Zuderstorfer, G., Riedlsperger, F., Sonderegger, B.: CreeSo software for creep simulation of complex alloys. *Materials at High Temperatures* **39**(6), 596–602 (2022) <https://doi.org/10.1080/09603409.2022.2058237>
 - [46] Guttmann, V., Timm, J.: On the Influence of the Thermal Pretreatment on Creep and Microstructure of Alloy 800H / ber den Einflu der Wrmeverbehandlung auf die Kriechgeschwindigkeit und das Gefge der Legierung Alloy 800H. *International Journal of Materials Research* **81**(6), 428–432 (1990) <https://doi.org/10.1515/ijmr-1990-810607> . Accessed 2025-05-07
 - [47] Rojas-Ulloa, C., Morch, H., Tuninetti, V., Duchne, L., Habraken, A.M.: Implementation of a modified Graham-Walles viscosity function within a Chaboche viscoplastic constitutive model. *Computers and Mathematics with Applications* **155**(1), 165–175 (2024) <https://doi.org/10.1016/j.camwa.2023.12.002>
 - [48] Rojas-Ulloa, C., Morch, H., Tuninetti, V., Tchoufang Tchuindjang, J., Pensis, O., Di Giovanni, A., Mertens, A., Duchne, L., Habraken, A.M.: Microstructure evolution of Incoloy 800H in industrial environment and correlation with creep mechanisms from literature. *Materials at High Temperatures* **41**(2), 311–321 (2024) <https://doi.org/10.1080/09603409.2024.2342602> . Accessed 2025-06-01

- [49] Frank, F.C., Read, W.T.: Multiplication Processes for Slow Moving Dislocations. *Physical Review* **79**(4), 722–723 (1950) <https://doi.org/10.1103/PhysRev.79.722> . Accessed 2024-03-26
- [50] Orlov , A.: Relation between the internal stress measured in creep and the stress generated by the dislocation structure in the fcc metals. *Philosophical Magazine* **84**(32), 3419–3426 (2004) <https://doi.org/10.1080/4786430412331284441> . Accessed 2025-04-23
- [51] Labusch, R.: Statistical theories of solid solution hardening. TT F-14,711 N73-11112, NASA, Washington (1972)
- [52] Feltham, P., Kauser, N.: Solid-Solution Hardening. *physica status solidi (a)* **117**(1), 135–140 (1990) <https://doi.org/10.1002/pssa.2211170114> . Accessed 2025-08-07
- [53] Leyson, G.P.M., Curtin, W.A.: Solute strengthening at high temperatures. *Modelling and Simulation in Materials Science and Engineering* **24**(6), 065005 (2016) <https://doi.org/10.1088/0965-0393/24/6/065005> . Accessed 2025-07-06
- [54] Varvenne, C., Leyson, G.P.M., Ghazisaeidi, M., Curtin, W.A.: Solute strengthening in random alloys. *Acta Materialia* **124**, 660–683 (2017) <https://doi.org/10.1016/j.actamat.2016.09.046> . Accessed 2025-07-06
- [55] Moretti, M.A., Lindgren, L.-E., kerström, P.: Physics-Based Flow Stress Model for Alloy 718. *Metallurgical and Materials Transactions A* **54**(5), 1985–1997 (2023) <https://doi.org/10.1007/s11661-022-06819-7> . Accessed 2025-07-06
- [56] Fleischer, R.L.: Substitutional solution hardening. *Acta Metallurgica* **11**(3), 203–209 (1963) [https://doi.org/10.1016/0001-6160\(63\)90213-X](https://doi.org/10.1016/0001-6160(63)90213-X)
- [57] Mishima, Y., Ochiai, S., Hamao, N., Yodogawa, M., Suzuki, T.: Solid Solution Hardening of Nickel —Role of Transition Metal and B-subgroup Solutes—. *Transactions of the Japan Institute of Metals* **27**(9), 656–664 (1986) <https://doi.org/10.2320/matertrans1960.27.656> . Accessed 2025-08-07
- [58] Young, A.M., Kral, M.V., Bishop, C.M.: Carbide formation accompanying internal nitridation of austenitic stainless steel. *Materials Characterization* **184**, 111662 (2022) <https://doi.org/10.1016/j.matchar.2021.111662> . Accessed 2022-12-07
- [59] Shewmon, P.: *Diffusion in Solids*, 2nd edn. The Minerals, Metals & Materials Series, vol. 1. Springer, Department of Metallurgical Engineering, Ohio State University (2016)
- [60] Young, A.M., Kral, M.V., Bishop, C.M.: Effects of aging and nitridation on

- microstructure and mechanical properties of austenitic stainless steel. *Journal of Materials Science* **58**(26), 10716–10735 (2023) <https://doi.org/10.1007/s10853-023-08659-1> . Accessed 2023-11-06
- [61] Gao, M., Xia, S., Huang, C., Hu, X., Liang, S., Zhang, W., Li, Q.: Study of Microstructure and Mechanical Properties of 800H Alloy During Creep. *Materials* **18**(4), 912 (2025) <https://doi.org/10.3390/ma18040912> . Accessed 2025-03-14
- [62] Murch, C., Leen, S.B., ODonoghue, P.E., Barrett, R.A.: A physically-based creep damage model for effects of different precipitate types. *Materials Science and Engineering: A* **682**, 714–722 (2017) <https://doi.org/10.1016/j.msea.2016.11.044> . Accessed 2025-08-19
- [63] Sonderegger, B., Kozeschnik, E.: Particle strengthening in fcc crystals with prolate and oblate precipitates. *Scripta Materialia* **66**(1), 52–55 (2012) <https://doi.org/10.1016/j.scriptamat.2011.10.003> . Accessed 2025-11-01
- [64] Lemaitre, J., Rodrigue, D.: *Engineering Damage Mechanics*, 1st edn. (2005). <https://link.springer.com/book/10.1007/b138882>
- [65] Spindler, M.: ECCC data sheet Steel X8NiCrAlTi 32-21 (1.4959, Alloy 800H). ECCC (1998)
- [66] Tan, L., Rakotojaona, L., Allen, T.R., Nanstad, R.K., Busby, J.T.: Microstructure optimization of austenitic Alloy 800H (Fe-21Cr-32Ni). *Materials Science and Engineering A*, 7 (2011) <https://doi.org/10.1016/j.msea.2010.12.052>
- [67] Schneider, C.A., Rasband, W.S., Eliceiri, K.W.: NIH Image to ImageJ: 25 years of image analysis. *Nature Methods* **9**(7), 671–675 (2012) <https://doi.org/10.1038/nmeth.2089> . Accessed 2025-11-07
- [68] Subcommittee, E.: Standard Test Methods for Determining Average Grain Size. *Book of Standards* **3**(1) (2025) <https://doi.org/10.1520/E0112-25> . Accessed 2025-09-23
- [69] Zheng, X., Zhang, X., Shi, H., Zhang, S., Zhao, L., Gu, J., Meng, W., Wang, J., Song, K., Hu, Y., Wang, H.: Oxidation behavior of Super304H and HR3C steels in high-temperature steam condition. *Journal of Materials Research and Technology* **40**, 2219–2232 (2026) <https://doi.org/10.1016/j.jmrt.2025.12.292> . Accessed 2026-01-13
- [70] Seplveda Hernandez, E., Nguyen-Minh, T., Traka, K., Castro Cerda, F., Kestens, L.A.I.: Orientation dependence of dynamic compared to static recovery in Interstitial Free steel. *Scripta Materialia* **249**, 116151 (2024) <https://doi.org/10.1016/j.scriptamat.2024.116151> . Accessed 2025-11-18
- [71] Orlov, A., adek, J.: Internal stress in homogeneous and dispersion-strengthened

- austenitic alloys in high-temperature creep. *Metal Science* **11**(12), 586–591 (1977) <https://doi.org/10.1179/msc.1977.11.12.586> . Accessed 2025-03-21
- [72] Brnger, E., Wang, X., Gottstein, G.: Nucleation mechanisms of dynamic recrystallization in austenitic steel alloy 800H. *Scripta Materialia* **38**(12), 1843–1849 (1998) [https://doi.org/10.1016/S1359-6462\(98\)00124-9](https://doi.org/10.1016/S1359-6462(98)00124-9) . Accessed 2024-11-18
- [73] Wang, X., Brnger, E., Gottstein, G.: The role of twinning during dynamic recrystallization in alloy 800H. *Scripta Materialia* **46**(12), 875–880 (2002) [https://doi.org/10.1016/S1359-6462\(02\)00072-6](https://doi.org/10.1016/S1359-6462(02)00072-6) . Accessed 2024-11-18
- [74] Cao, Y., Di, H., Zhang, J., Yang, Y.: Dynamic behavior and microstructural evolution during moderate to high strain rate hot deformation of a Fe-Ni-Cr alloy (alloy 800H). *Journal of Nuclear Materials* **456**, 133–141 (2015) <https://doi.org/10.1016/j.jnucmat.2014.09.030> . Accessed 2022-05-11
- [75] Paul, A.R., Kaimal, K.N.G., Naik, M.C., Dharwadkar, S.R.: Lattice and grain boundary diffusion of chromium in superalloy Incoloy-800. *Journal of Nuclear Materials* **217**(1-2), 75–81 (1994) [https://doi.org/10.1016/0022-3115\(94\)90306-9](https://doi.org/10.1016/0022-3115(94)90306-9) . Accessed 2024-06-14
- [76] Stechauner, G., Kozeschnik, E.: Self-Diffusion in Grain Boundaries and Dislocation Pipes in Al, Fe, and Ni and Application to AlN Precipitation in Steel. *Journal of Materials Engineering and Performance* **23**(5), 1576–1579 (2014) <https://doi.org/10.1007/s11665-014-0921-z> . Accessed 2024-06-14
- [77] Naik, M.C., Paul, A.R., Kaimal, K.N.G., Venkateswarlu, K.S.: Mass transport of cobalt and nickel in Incoloy-800. *Journal of Materials Science* **25**(3), 1640–1644 (1990) <https://doi.org/10.1007/BF01045364> . Accessed 2024-06-14
- [78] Paul, A.R., Naik, M.C., Kaimal, K.N.G.: Mass transport of iron in Incoloy 800. *Materials Science and Technology* **7**(1), 8–11 (1991) <https://doi.org/10.1179/mst.1991.7.1.8> . Accessed 2024-06-14
- [79] Randle, V.: Twinning-related grain boundary engineering. *Acta Materialia* **52**(14), 4067–4081 (2004) <https://doi.org/10.1016/j.actamat.2004.05.031>
- [80] Hammond, J.P., Ratcliff, L.T., Brinkman, C.R., Moyer, M.W., Nestor, C.W. Jr.: Dynamic and Static Measurements of Elastic Constants with Data on 2 1/4 Cr-1 Mo Steel, Types 304 and 316 Stainless Steels, and Alloy 800H. UC-79B, -h, -k ORNL-5442, Oak Ridge Laboratory, Oak Ridge, Tennessee, U.S. (1979)
- [81] Wright, J.K., Carroll, L.J., Cabet, C., Lillo, T.M., Benz, J.K., Simpson, J.A., Lloyd, W.R., Chapman, J.A., Wright, R.N.: Characterization of elevated temperature properties of heat exchanger and steam generator alloys. *Nuclear Engineering and Design* **251**, 252–260 (2012) <https://doi.org/10.1016/j.nucengdes.2011.10.034> . Accessed 2022-05-09

- [82] Kim, W.-G., Kusuma, I.N.C., Park, J.-Y., Sah, I., Kim, E.-S., Kim, S.-J., Kim, M.-H.: Hot Tensile Behavior for Alloy 800H Base and Weld Metals. In: Transactions of the Korean Nuclear Society Autumn Meeting, Yeosu, Korea, pp. 1–3 (2018). <https://api.semanticscholar.org/CorpusID:202626990>
- [83] Gardiner, B.: High temperature creep performance of alloy 800H. PhD. thesis, University of Canterbury (2014)
- [84] Frost, H.J., Ashby, M.F.: Deformation-mechanism Maps : the Plasticity and Creep of Metals and Ceramics. Oxford [Oxfordshire] ; New York : Pergamon Press, ??? (1982). <http://lib.ugent.be/catalog/rug01:003160335>
- [85] Kubin, L.P., Mortensen, A.: Geometrically necessary dislocations and strain-gradient plasticity: a few critical issues. *Scripta Materialia* **48**(2), 119–125 (2003) [https://doi.org/10.1016/S1359-6462\(02\)00335-4](https://doi.org/10.1016/S1359-6462(02)00335-4) . Accessed 2025-10-07
- [86] Praveen, C., Christopher, J., Ganesan, V., Reddy, G.V.P., Albert, S.K.: Influence of varying nitrogen on creep deformation behaviour of 316LN austenitic stainless steel in the framework of the state-variable approach. *Materials Science and Engineering: A* **803**, 140503 (2021) <https://doi.org/10.1016/j.msea.2020.140503> . Accessed 2025-04-29
- [87] Kumari, M., Ray, K.K.: Effect of the mode of deformation on activation volume of a material. *Materials Science and Engineering: A* **650**, 335–344 (2016) <https://doi.org/10.1016/j.msea.2015.10.060> . Accessed 2025-05-20
- [88] Martin, J.L., Lo Piccolo, B., Kruml, T., Bonneville, J.: Characterization of thermally activated dislocation mechanisms using transient tests. *Materials Science and Engineering: A* **322**(1-2), 118–125 (2002) [https://doi.org/10.1016/S0921-5093\(01\)01124-8](https://doi.org/10.1016/S0921-5093(01)01124-8) . Accessed 2025-03-28
- [89] Blum, W.: Discussion: Activation volumes of plastic deformation of crystals. *Scripta Materialia* **146**, 27–30 (2018) <https://doi.org/10.1016/j.scriptamat.2017.10.029>
- [90] Raj, S.V.: Apparent activation volume for creep of copper and alpha brass at intermediate temperatures. *Journal of Materials Science* **24**(9), 3196–3204 (1989) <https://doi.org/10.1007/BF01139041> . Accessed 2025-07-30
- [91] Li, Z., Voisin, T., McKeown, J.T., Ye, J., Braun, T., Kamath, C., King, W.E., Wang, Y.M.: Tensile properties, strain rate sensitivity, and activation volume of additively manufactured 316L stainless steels. *International Journal of Plasticity* **120**, 395–410 (2019) <https://doi.org/10.1016/j.ijplas.2019.05.009> . Accessed 2025-07-30
- [92] Chen, F., Carlos, R.-U., Tchuindjang, J.T., Dedry, O., Bharadwaj, R., Somani, M., Mertens, A., Habraken, A.M.: Microstructure-informed creep model for

- 30CrMoNiV11-5 steel. *Materials & Design* **266**, 116152 (2026) <https://doi.org/10.1016/j.matdes.2026.116152> . Accessed 2026-05-12
- [93] Mukherjee, A.K., Bird, J.E., Dorn, J.E.: Experimental correlations for high temperature creep. Scientific report UCRL-18526, Berkeley, California, Lawrence Radiation Laboratory (1968)
- [94] Eliasson, J., Sandström, R.: Proof strength values for austenitic stainless steels at elevated temperatures. *Steel Research* **71**(6-7), 249–254 (2000) <https://doi.org/10.1002/srin.200001224> . Accessed 2025-11-22
- [95] Sieurin, H., Zander, J., Sandström, R.: Modelling solid solution hardening in stainless steels. *Materials Science and Engineering: A* **415**(1-2), 66–71 (2006) <https://doi.org/10.1016/j.msea.2005.09.031> . Accessed 2025-06-06
- [96] Cyrska-Filemonowicz, A., Krautwasser, P., Ennis, P.J.: Quantitative TEM and SANS Studies of Aged and of Creep Deformed Alloy 800H. In: *Proc. 13 Int. Congr. on Electron Microscopy*, vol. 1, pp. 33–34. Les Ed. de Physique, Paris (1994)
- [97] Reichert, B.: Extension of a constitutive model describing the inelastic behaviour of a precipitation hardened alloy. *Steel Research* **70**(3), 115–120 (1999) <https://doi.org/10.1002/srin.199905610> . Accessed 2024-02-20
- [98] El-Magd, E., Nicolini, G., Farag, M.: Effect of carbide precipitation on the creep behavior of Alloy 800HT in the Temperature Range 700 to 900 . *Metallurgical and Materials Transactions A* **27**(3), 747–756 (1996) <https://doi.org/10.1007/BF02648962> . Accessed 2023-07-24
- [99] Forest, S., Mesarovic, S.D., Zbib, H.: *Mesoscale Models: From Micro-Physics to Macro-Interpretation*. CISM International Centre for Mechanical Sciences, vol. 587. Springer, ??? (2019)