

Constitutive modeling of hot forming of austenitic stainless steel 316LN by accounting for recrystallization in the dislocation evolution

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Abstract. Hot compression test data taken from Zhang [1] of metastable austenitic stainless steel AISI 316LN over a range of strain rates and temperatures shows typical dynamic recovery and recrystallization behavior. It is proposed to model this behavior by incorporating not only the hardening and recovery into the Bergström dislocation evolution equation, but also the recrystallization. It is shown that the initial mechanical response before recrystallization can be accurately represented by assuming that the mean free path evolves as the microstructure evolves from homogeneously spaced dislocations to cell-pattern. Results show that this novel continuum mechanical model can predict the observed behavior, showing a good match to the experimental data and capturing the transition from recrystallization to (almost) no recrystallization.

1. Introduction

Metastable austenitic stainless steels are widely used in industry due to their excellent corrosion resistance and formability. During hot forming, the stresses introduced when deforming are considerably lower and thus the resulting residual stresses and springback. In order to design optimal and safe hot-formed components and products a good understanding and constitutive description of the material behavior is vital and an important feature of the behavior during hot forming is dynamic recovery and dynamic recrystallization.

Several authors have presented continuum mechanical models that predict the flow stress evolution of metallic materials experiencing dynamic recrystallization during hot forming, however most of these models fall in one of two classes. The first one employs the Arrhenius equation to link strain rate, deformation temperature and flow stress, the second one treats dynamic recrystallization similar to static recrystallization and is based on Avrami kinetics.

The Arrhenius-based models start out with assuming a phenomenological relation between the flow stress and Zener-Holloman parameter Z and are supplemented with 4th to 7th-order polynomial relations between strain and model parameters to be able to describe stress-strain relations. The recrystallization models based on Avrami kinetics describe recrystallization as a fractional process lowering the stress towards a steady state. In this model, recrystallization evolves from zero to one, suggesting that recrystallization is finished. However, it is commonly recognized that



multiple peak versus single peak recrystallization is due to successive waves of recrystallization, where in single peak the waves are in rapid succession and in multiple peak, recrystallization is (almost) finished before the next wave starts [2].

In this research a new continuum mechanical model is proposed to account for single-peak recrystallization in a more physically based manner. It is based on the assumption that at the saturation stress an equilibrium exists between hardening on one hand and recovery and recrystallization on the other. Experimental curves of hot compression of AISI 316LN are taken from Zhang et al. [1] and used to calibrate the proposed model.

2. Modeling of Hot Forming

The constitutive equations for the new model are based on the decomposition of the flow stress into a work hardening stress σ_w , which accounts for dislocation-dislocation interactions, and a strain and strain rate independent stress σ_0 . The strain and strain rate independent stress is assumed to relate to stresses in the atomic lattice of which the temperature dependence is assumed to be similar to the temperature dependence of the shear modulus [3].

$$\sigma_f(\dot{\varepsilon}, T) = \sigma_0(T) + \sigma_w(\dot{\varepsilon}, T) = \mu(T) \left(\sigma_0 + \alpha \mu_{ref} b M \sqrt{\rho(\dot{\varepsilon}, T)} \right) \quad (1)$$

Where $\mu(T)$ represent the temperature dependence of the shear modulus with respect to the reference value μ_{ref} and is taken linear in this work:

$$\mu(T) = \frac{\mu_{ref} - C_5 T}{\mu_{ref}} \quad (2)$$

The work hardening part will be described in two subsections. In the first subsection the Taylor equation and the corresponding dislocation density evolution are described. Furthermore the relation between the model parameters and strain rate and temperature is explained. In the second subsection the evolution of mean free path with strain is elaborated.

2.1. Taylor and Bergström equation

The work hardening of the model is based on the evolution of the statistical average mobile dislocation density ρ (henceforth dislocation density) and is described by the Taylor equation:

$$\sigma_w = \alpha \mu b M \sqrt{\rho} \quad (3)$$

Where α , μ , b and M are a material constant, temperature dependent shear modulus, the length of the Burgers vector and the Taylor orientation factor. To complete the relation between stress and strain a supplementary equation is needed to describe the evolution of the dislocation density with strain. A novelty of the current model is the incorporation of dynamic recrystallization directly into the Bergström dislocation evolution. After all, the effect of recrystallization is to nucleate and grow new dislocation-free grains into the dislocated substructure, thereby removing dislocations from the structure [4,5] and oppose work hardening. The modified Bergström equation reads:

$$\frac{d\rho}{d\varepsilon} = h\sqrt{\rho} - f\rho - R_{rate}(\rho - \rho_0) \quad (4)$$

In this equation $h\sqrt{\rho}$ describes the hardening by storage (immobilization) of mobile dislocations and is linked to the mean free path of dislocations, f governs the dynamic recovery and is dependent on temperature and strain rate due to its thermally activated nature [6].

$$f(\dot{\varepsilon}, T) = f_0 + C_1 Z^{m_1} \quad (5)$$

Now it must be noted that R_{rate} does not describe the state of recrystallization, it describes the current recrystallization *rate*. As discussed in the introduction, at the steady state, equilibrium exists between dislocation generation and annihilation by recovery *and* recrystallization. This means that at the steady state, recrystallization is ongoing and not finished. In this research, a simple mathematical formulation is used to describe the expected recrystallization rate explicitly. It captures both the sigmoidal trend and the constant rate of recrystallization which is expected at the steady state stress.

$$R_{rate} = \frac{A}{1 + \exp\{n(\varepsilon_c - \varepsilon)\}} \quad (6)$$

Here A determines the saturation recrystallization rate (and thus influences the steady state stress), n determines the rate at which the saturation recrystallization rate is reached and ε_c is linked to the critical strain for the onset of recrystallization by shifting the sigmoid over the strain-space. The dependence of ε_c on strain rate and temperature is mirrored from the Bergström approach for recovery, Eq. 5. For A the constant is omitted to allow the recrystallization rate to approach zero at low temperatures and high strain rates, n is taken as a constant.

2.2. Evolution of mean free path on hardening behavior

In a recent paper by Angella [7], focusing on the hardening and recovery behavior of AISI 316L, it was found by means of TEM analysis that annealed AISI 316L starts out with homogeneously spaced dislocations which, upon straining, arrange themselves into a cell pattern. One of the fundamental hypotheses of the family of Bergström and similar Kocks-Mecking hardening models is the principle of similitude, which states that the dislocation structure is self similar, i.e. it scales with the inverse of the applied stress [8]. Obviously, the evolution from homogeneously spaced dislocations to cell-pattern violates the principle of similitude.

To account for this behavior, Angella assumes that the mean free path of dislocations is a state variable (next to the dislocation density) and evolves from an initial value to a saturation value. In the model presented in this paper (for the similar material AISI 316LN) the same strategy is used and a simple evolution equation is presented for the parameter controlling the hardening.

$$\frac{dh}{d\varepsilon} = \frac{K}{\sqrt{\rho}}(h_s - h) \quad (7)$$

In the results it will be shown that the behavior of AISI 316LN is not well represented by the Bergström model alone and that Eq. 7 offers a significant improvement in capturing the hardening behavior. To complete this model a relation is needed to define the rate at which h evolves from the initial value to the saturation value depending on strain rate and temperature. This is achieved through a similar dependence of K on the Zener-Holloman parameter as in Eq. 5.

3. Results

A least squares fitting routine was used to obtain the parameter values shown in Table 1. The model, experimental results and the merit of the mean free path evolution are shown in Figure 1.

Table 1: Model parameters obtained by least squares optimization (α , b , and M were taken constant). Units: σ_0 and μ_{ref} in MPa, Q in J/mol, b in mm, ρ_0 in mm⁻² and h_0 and h_s in mm⁻¹, the rest is dimensionless.

Taylor		Bergström		$\frac{dh}{d\varepsilon}$		f		Z	
α	0.5	ρ_0	4.63e5	h_0	1.71e6	f_0	15.06	Q	3.75e5
b	2.8e-7	h	—	h_s	2.67e5	C_1	7.68e4		
M	3	f	—	K	—	m_1	-0.25		
σ_0	42	R_{rate}	—						
K		A		n		ε_c		$\mu(T)$	
K_0	4.40e5	C_3	2.1e8	n	15.40	ε_{c0}	2e-3	μ_{ref}	8.67e4
C_2	1.5e11	m_3	-0.56			C_4	0.16	C_5	-32.25
m_2	-0.47					m_4	3.48e-2		

4. Discussion and Conclusion

The results show that the presented constitutive equations are capable of accurately capturing the behavior of AISI 316 LN over a broad range of hot working temperatures and strain rates. This includes the transition from pronounced recrystallization at ‘high temperatures - low strain rates’ to negligible recrystallization at ‘low temperature - high strain rates’. The material constants have logical dependencies on temperature and strain rate through simple power law relations with the Zener-Holloman parameter, similar to the one suggested by Bergström for dynamic recovery. Furthermore, from Figure 1, it is clear that the evolution of mean free path improves the accordance between model and experiment significantly. The hardening saturation value is reached around 5% strain, however the effect on the fit between measurement and model is significant up to large strains.

Some discrepancies between experiment data and model still exist, e.g. at 900°C, $\dot{\varepsilon} = 10^{-1} s^{-1}$ the experimental curve shows more softening then predicted, furthermore there seems to be pronounced recrystallization at 1200°C, $\dot{\varepsilon} = 10^{-3} s^{-1}$ which is not predicted by the model. Upon close examination it seems that the yield stress at these high temperature-low strain rate conditions is softened *below* the initial yield stress. It is hypothesized that at these extreme conditions concurrent grain coarsening is taking place.

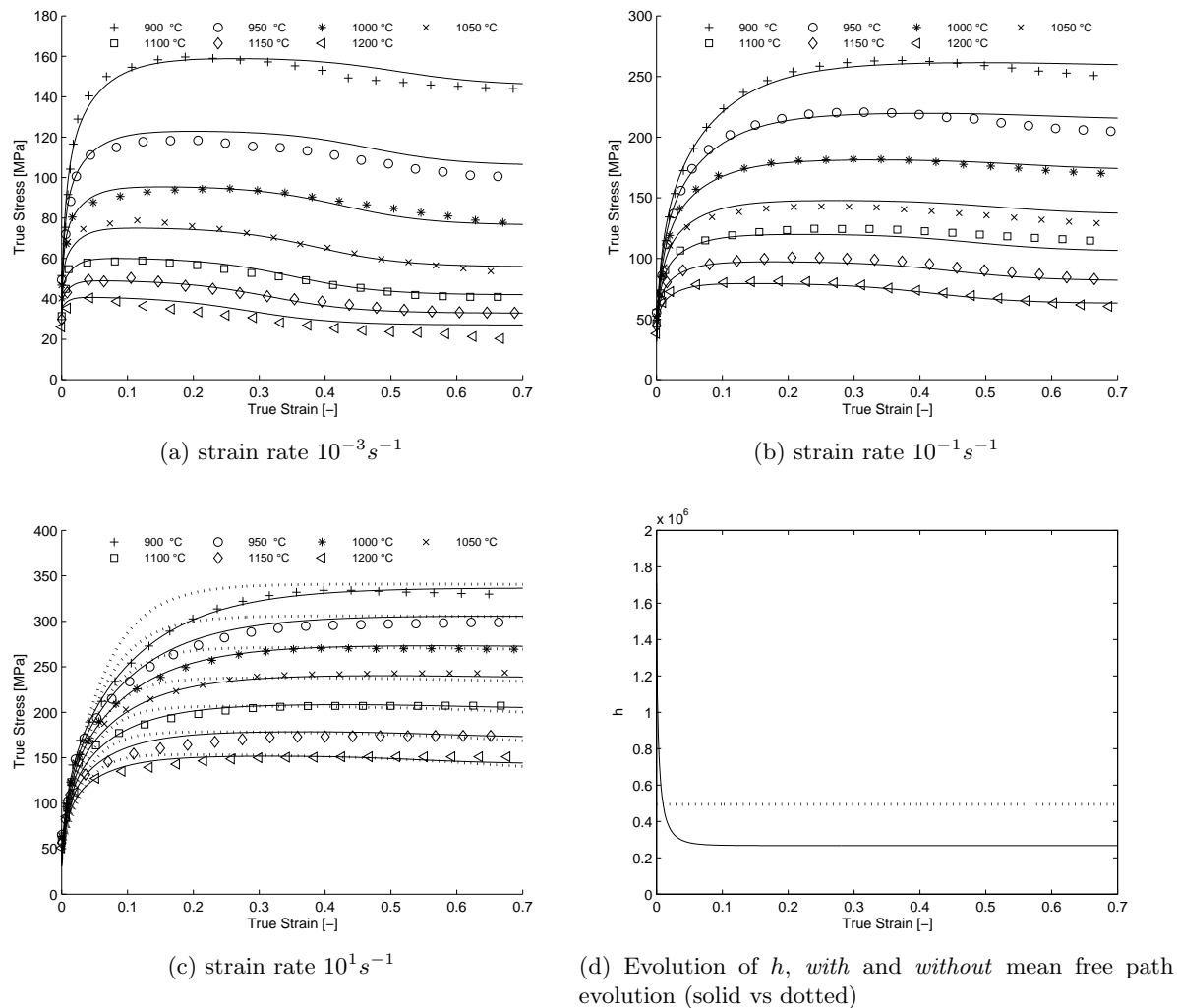


Figure 1: True stress vs true strain curves ($\dot{\epsilon} = 10^{-3} s^{-1} - 10^1 s^{-1}$, $T = 900^{\circ}C - 1200^{\circ}C$), model results shown in solid line and dotted line, respectively *with* and *without* evolving mean free path, subfigure (d) shows the evolution of h *with* and *without* mean free path evolution in the model

References

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