Modelling of Dynamic Strain Aging with a Dislocation-Based Isotropic Hardening Model and Investigation of Orthogonal Loading

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Summary

Based on experimental results, a dislocation material model describing the dynamic strain aging effect at different temperatures is presented. One and two stage loading tests were performed in order to investigate the influence of the loading direction as well as the temperature influence due to the hardening mechanism. Bergström’s theory of work hardening was used as a basis for the model development regarding the thermal isotropic behavior as well as the Chaboche model to describe the kinematic hardening. Both models were implemented in an in-house FE-Code in order to simulate the real processes. The present paper discusses two hardening mechanisms, where the first part deals with the pure isotropic hardening including dynamic strain aging and the second part involves the influence of the loading direction regarding combined (isotropic and kinematic) hardening behavior.

1. Introduction

The aim of this study was the investigation of a high strength low carbon steel concerning dynamic strain aging behavior as well as load direction dependence of the material. Dynamic strain aging is a phenomenon in which the interstitial atoms diffuse around dislocations and inhibit dislocation motion [1]. Modelling of the load direction dependence requires complex material models which involve tensor variables. The widely used kinematic hardening model of the Armstrong-Frederick type as proposed by Chaboche [2] was applied to describe the reverse loading deformation as well as the orthogonal strain-path case. In general, this model is used to fit tension-compression data, or in other words, it is used for proportional loading and especially for small strains as investigated in Chaboche [2]. For experiments with large strains and complex strain paths a more appropriate method has to be used to fit the material parameters.

2. Material Modelling

2.1. Dislocation Based Material Modelling

Work hardening is caused by accumulation of dislocations. Therefore, the flow stress is a function of the total dislocation density. In the well established stress-strain relationship [3] the total dislocation density $\rho$ is a function of the equivalent plastic strain $\varepsilon_{eq}$:

$$\sigma_y = \sigma_0 + cG\sqrt{\rho(\varepsilon_{eq})}$$  \hspace{1cm} (1)

where $\sigma_0$ is the strain independent yield stress, $c$ is a constant which incorporates the Burgers vector and other crystal properties and $G$ is the shear modulus. The major problem on formulating a dislocation theory is to derive the relationship between the dislocation density and the deformation. The dislocation theory of Bergström [4] is based on the average behavior of a large number of dislocations, mobile as well as immobile dislocations. Bergström proposed the following evolution equation of the dislocation density $\rho$:

$$\frac{\partial \rho}{\partial \varepsilon_{eq}} = B - R\rho$$  \hspace{1cm} (2)

where $B$ is the immobilization rate of dislocations and $R$ is a parameter describing the remobilization...
of dislocations. Under some assumptions and especially the strain independence of the immobilization rate \( B \), equation (2) can be integrated analytically, and one obtains a relationship between dislocation density \( \rho \) and the equivalent plastic strain \( \varepsilon_{eq} \),

\[
\rho(\varepsilon_{eq}) = \frac{B}{R} \left( \rho_0 - 1 \right) \exp(-R\varepsilon_{eq}) + 1
\]  

(3)

where \( \rho_0 \) is the normalized initial dislocation density. In order to describe the dynamic strain aging effect the explained model was modified. The following equations were derived from an investigation of the experimental data which is discussed in detail in [6]. The temperature dependence of the shear modulus and of the yield stress was approximated with a quadratic and a linear function, respectively. Whereas the experimental data from [5] were used to fit the model of the shear modulus. It was assumed that the immobilization of dislocations will reach a maximum at a certain temperature. Therefore a Gauss-type function (4) was proposed to describe the behavior of the immobilization rate

\[
B(T) = 1 + \gamma \exp\left( -\left( \frac{T - T_0}{\tau} \right)^2 \right)
\]  

(4)

where \( \gamma \) represents the concentration of the free interstitial atoms, \( T_0 \) defines the temperature at which the immobilization rate achieved the maximum and \( \tau \) is the parameter which defines the width of the function, or in other words, it defines the region at which dynamic strain aging occurs. The temperature dependence of the remobilization parameter was also approximated with an energy-type function as shown in equation (5)

\[
R(T) = R_0 + \alpha \exp\left( -\frac{\beta}{T} \right)
\]  

(5)

in which \( R_0 \) is the temperature independent remobilization parameter, \( \alpha \) and \( \beta \) describe the saturation behavior of the remobilization. Furthermore, the evolution equation of the total dislocation density (2) was adapted in order to predict the hardening behavior of the second deformation by splitting of the total dislocation density into two parts

\[
\frac{\partial \rho}{\partial \varepsilon_{eq}} = B - R(\rho - \rho_L)
\]  

(6)

where \( \rho_L \) is the density of the locked dislocations and the difference \( \rho - \rho_L \) is the so called free dislocation density where dislocations have the ability to be remobilized. From this hypothesis can be deduced that the remobilization parameter \( R \) is no longer applied to the total dislocations but only to those dislocations which did not interact with interstitial atoms during the first deformation.

2.2. Modelling of Kinematic Hardening

In the von Mises theory the plasticity criterion is defined in the following form:

\[
f = J_2(\sigma - X) - Q \leq 0
\]  

(7)

where \( J_2 \) is the second invariant of the deviatoric stress tensor, \( X \) is the back-stress and \( Q \) describes the isotropic hardening of the material. In this work, the evolution equation of the back-stress as proposed by Chaboche [2] was used

\[
dX_i = \frac{2}{3} C \dot{\varepsilon}^p - \eta_i X_i d\varepsilon_{eq}
\]  

(8)
where \( C_i \) and \( \eta_i \) are material constants which describe translation of the yield surface in the stress space. In a phenomenological sense, this equation is associated also to the dislocation theory, where the first term is related to the generation of dislocations, whereas the second term retards the dislocation motion. In order to describe complex material behavior, the back-stress is defined as a sum of nonlinear as well as of linear components

\[
X = \sum_{i=1}^{n} X_i \tag{9}
\]

The isotropic part of the model is a common exponential function with a strain independent yield stress \( Q_0 \), a saturation value \( Q_\infty \) and a saturation rate \( b \).

\[
Q = Q_0 + Q_\infty \left( 1 - \exp(-b \epsilon_{eq}) \right) \tag{10}
\]

3. Experiments and Parameter Identification

The equipment and the specimens which were used to perform the required experiments in order to observe the material behavior and to determine the material parameters are depicted in Fig. 1.

Fig. 1: Dilatometer and the specimens for compression-compression tests (left), Specimen for compression-tension tests (right).

3.1. Dislocation Based Material Modelling

Experiments were performed on a Bähr DIL 805 dilatometer using cylindrical specimens with a length-to-diameter ratio of 3:2, see Fig. 1 (left). In order to reduce friction effects during the deformation, graphite lubrication was used. The temperature of the specimen was controlled with one thermocouple which was spot welded on the specimen.

Fig. 2: Temperature dependence of the material coefficients \( \sigma_0 \): yield stress, \( cB^{1/2} \): product of the constant \( c \) and the immobilization rate \( B \) of dislocations, \( R \): remobilization parameter of dislocations.

The experimental data were fitted with the functions defined in the previous section which are shown also in Fig. 2. The whole data and the fit with the proposed model are depicted in Fig. 3 (left).
3.2. Combined (Kinematic-Isotropic) Hardening Model

Model parameters were determined using an inverse approach of the measured compression-tension (CT) and compression-compression (CC) data. All experiments regarding the kinematic hardening were performed at room temperature. CT experiments were performed with a special geometry of the specimen (Fig. 1, right) in order to achieve large strains and to avoid buckling of the specimen during the compression test. The specimen does not have any cylindrical part in the region of the deformation; it is an axisymmetric concave region with a certain radius. In order to perform experiments with an orthogonal strain path, cubic specimens (Fig. 1, left) were first compressed in the axis direction and after that the specimen was reloaded (compression) perpendicular to the first load direction. The procedure of the parameter identification is schematically illustrated in Fig. 4.

4. Results and Discussion

Experiments show that the effect of dynamic strain aging is very pronounced which is manifested with an increase in dislocation density and therefore with an increase in yield stress, see Fig. 3 (left).
The maximum value of the immobilization rate $B(T)$ is achieved when the velocity of the interstitial atoms is the same as the velocity of the mobile dislocations. If the material does not contain interstitial atoms, the parameter $\gamma$ is equal to zero and the immobilization rate $B$ does not describe dynamic strain aging. Simulation results and experimental data of two stage compression tests are shown in Fig. 3 (right). The temperature behavior of the yield strength is similar to the behavior of the immobilization rate $B(T)$. The higher the diffusion of the interstitial atoms into mobile dislocations during the first plastic deformation, the higher the yield strength of the second deformation at room temperature, see Fig. 3 (right). If the first and the second deformation are conducted at or near room temperature, no dynamic strain aging is observed. Fig. 5 shows some simulation examples of the two stage compression tests including the hardening behavior of the second deformation at room temperature, where the first deformation was carried out at $T = 300{^\circ}C$ and a strain rate of $d\varepsilon/dt = 1s^{-1}$. It is obvious that there is discontinuous stress behavior due to the dynamic strain aging. It is important to emphasize that only the data of the first deformation were fitted and the prediction of the hardening behavior of the second stage is due to the assumption of the dislocation splitting into locked and free dislocations (6). Increasing of the yield stress due to the dynamic strain aging for a large plastic pre-strain leads to a lower ductility and one observes almost a perfectly plastic material behavior (Fig. 5).

The figures below show the material behavior in a compression-tension (Fig. 6) and in a compression-compression (Fig. 7) loading case at room temperature. The decomposition of the back-stress in linear and non-linear parts (9) plays a major role in modelling the reverse loading. In order to approximate the almost linear behavior of the yield stress at larger strains, besides the nonlinear components of the total back-stress a linear component was used. Fig. 6 shows the influence of the plastic pre-straining with respect to the subsequent tension loading. For plastic strains greater than $\varepsilon \approx 0.15$ the evolution of the micro structure plays a significant role and the material shows different behavior in compression and tension, in the case without plastic pre-straining, see Fig. 6 a).

The large plastic pre-strains do not seem to influence the stress response in the second deformation, which is almost identical to the undeformed material. To make an educated statement regarding this
behavior, investigations of the micro-structure should be carried out. A similar behavior of the material is observed also by CC tests, see Fig. 7. In general, experiments with non-proportional loading require a more complex material model. Performing adequate experiments and developing a more complex identification procedure (Fig. 4), one can fit the material parameter of the kinematic as well as of the isotropic part of the model (8), without increasing the number of the model parameter.

Fig. 7: Different plastic pre-straining in compression load case (first deformation $0^\circ$), second deformation $90^\circ$.

5. Conclusions and Outlook

In this work, an enhanced Bergström material model was presented. The simulation results are in good agreement with the experimental data. Especially the temperature dependence of the immobilization rate $B(T)$ is well described. In order to describe the two stage deformation process with respect to dynamic strain aging, the evolution equation of the total dislocation density was modified. The developed model is applicable to a temperature range where dynamic strain aging occurs as well as at room temperature, where at room temperature the presented model coincides with the classical Bergström model. Kinematic hardening behavior of the material at room temperature was described by the classical Chaboche model. Determination of the Chaboche model parameters was performed by an inverse modeling of the experiments CT and CC. The good agreement between the experimental data and the model could be achieved with eight material parameters. The range of applicability could be enhanced when one replaces the isotropic part of the Chaboche model by the proposed enhanced Bergström model in order to simulate dynamic strain aging regarding non-proportional loading. This combination requires also an investigation of the temperature dependence of the kinematic hardening parameters, which could lead to a more difficult optimization process, because of the expected increase in the number of model parameters.

Acknowledgements

The Authors are very grateful to CTI (The Swiss Innovation Promotion Agency) for financial support of this study.

References