

Thermomechanical characterization of 22MnB5 steels with special emphasis on stress relaxation and creep behavior

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Abstract

In hot stamping, the deformed austenitic phase is further quenched in the die to produce a fully martensitic microstructure in the component. During this time, the microstructure is refined with a relaxed state of stress. Therefore, it is important to study the time-dependent effects on the material during forming and quenching. This influences the martensite start temperature and further aids to predict the residual stress in the component accurately.

In this work, thermo-mechanical characterization of 22MnB5 steels is presented with special emphasis on stress relaxation and creep phenomena. In addition to the monotonous experiments at different strain rates, relaxation and creep tests give a deeper insight into the rate dependent behavior of the material. These tests are conducted for two distinct microstructures (austenite and martensite) using Gleeble 3150. In austenitic phase, relaxation test is conducted at 900°C for different strain levels which lead to a significant drop in the stress level of the material due to recovery. Therefore, creep tests are conducted to determine the real elastic limit in the material at high temperatures. As a result, a significant increase in the creep strain is realized once the stress in the material increases 20 MPa. Martensitic microstructure shows stress relaxation behavior at 200°C (below the martensite finish (M_f) temperature) and even at ambient temperature after a total deformation of 3%. These experimental results are the basis for the construction of a viscoplastic constitutive model and the model is able to represent the observed phenomena.

Keywords

Hot stamping, Experimental Characterization, Stress Relaxation and Creep behavior, Hardening model

1 Introduction and motivation

Recent emission regulations have forced the automotive industry to reduce the vehicle weight for improved fuel efficiency without compromising the passenger safety. In order to meet these demands, a significant effort has been directed towards the development of advanced high-

strength steel (AHSS) grades. One method to introduce new grades in the market is to change the chemical composition of the steel, even though efficient, an alternative approach is to tailor the microstructure through heat treatment. The latter is much more feasible than the former which involves scrap recycling, production cost and environmental sustainability, etc.

For example, in hot stamping process; the microstructure of the component is tailored according to the automotive structural requirements. This is an efficient technology for producing high-strength crash resistant components for automobiles with less amount of spring back. For a nice overview of the process, the readers are suggested to refer the work [1]. There has been a significant increase over the production of hot stamped components with 3 million parts per annum in 1987 to 125 million in 2010 and future estimates of around 350 million in 2015 [2].

In this process, the as-received material with ferrite and pearlite microstructure of the steel is fully austenized in a furnace and transferred to the press. In the original process, the component is formed and quenched in the press to obtain fully martensitic microstructures. The advantage of producing tailored microstructures has been investigated recently. Microstructures with phases like bainite, ferrite, and martensite are produced to increase the energy absorption capacity of the components like B-pillar [3]. This can be realized through partial In-die heating [4], post tempering and partial heating in a furnace, etc. Since process variation is out of scope in the present work; interested readers are redirected to the works of [5, 6]. The goal of optimized processes is the production of tailored properties in order to fulfill the design restrictions. A simple parameter to control the state of the microstructure is through the evaluation of hardness.

Extensive material characterizations at high temperatures and real process knowledge are necessary to form the automotive structural components. Naderi et al. [7] and Shi et al. [8] conducted experiments using Bähr dilatometer and Gleeble to study the influence of parameters like austenization soaking time, initial deformation temperature, strain, strain rates, and cooling rate on hot stamping process.

Thermo-mechanical properties of 22MnB5 steels are determined mostly by compression of cylindrical bar specimens [9, 10 & 11]. This is due to the advantage in obtaining the flow stress information at high deformation levels. However, in this approach, bulk material made of the same composition is tested, whereas, the real forming process involves the deformation of rolled sheet. Also, compression tests require inverse estimates to identify the flow stress parameters due to bulging of the specimen. The choice of coulomb friction coefficient used to simulate the compression tests further influences the results. Merklein and Lechler [12] performed isothermal tension tests at different temperatures and strain rates to characterize the flow stress behavior of austenite. The uniform deformation levels attained are limited, and require a large number of specimens similar to compression tests. To minimize the number of specimens, Åkerström et al. [9] performed compression tests under continuous cooling with different start temperatures. Attention must be paid to avoid the formation of ferrite during isothermal deformation [13] tests for particular combination of temperature and strain rates. Isothermally compressed boron

alloyed steels reveal that production of fully martensitic microstructures is not feasible if the deformation temperature is less than 800 °C [14]. Li et al. [15] characterized the flow stress properties of all possible microstructure phases (austenite, ferrite, bainite and martensite) in 22MnB5 steel with the influence of both temperature and strain rate.

In this work, we present the thermo-mechanical characterization of pbs-ultraform material supplied by voestalpine Stahl GmbH. Isothermal tension tests at three different temperatures and strain rates are conducted for austenite phase using Gleeble 3150. Phenomena like deformation induced ferrite transformation, DIFT [16] are observed. It is realized that the control of machine regulation's parameters is necessary to maintain constant strain rate at high strain rates.

As already discussed, thermomechanical tests on hot stamping steels are already covered very well in the literature. The authors were unable to find studies on relaxation and creep as well as the recovery behavior for hot sheet metal forming. During deformation of the austenite phase, a competition between hardening and recovery takes place until the die is completely closed. Before the martensitic start temperature is reached as a result of the quenching process, stress relaxation phenomena lead to a drop in the stress state of the austenite phase. For temperatures below martensite start, the phase transformation from austenite to martensite is influenced by the stress state (e.g. due to volume change). However, the temperature is still high enough that stress relaxation phenomena occur in the martensitic microstructure. Monotonous tension tests at elevated temperatures will give no information about this behavior; therefore, we perform stress relaxation tests on both austenitic and martensitic phases. As a result, the stress level in austenitic phase drops significantly during constant strain hold duration of 10 seconds. Creep tests show the elastic range in austenite at high temperatures. Plastically deformed martensite at ambient temperature also displays the stress relaxation behavior.

In material models for representing hot stamping processes, the time dependent behavior due to relaxation and creep effect are not taken into account. Typically, pure mechanical theories have been extended with incorporation of temperature and an internal variable for modelling the evolution of the volume fraction of martensite [17]. The hypothesis of the experimental investigations and the model approach is that the modelling of hot stamping processes may be improved with incorporation of the most important rate-dependent effects: i.e. strain-rate dependence as well as relaxation and creep.

The article is organized as follows: In section 2, the experimental methods are described. Section 3 introduces the flow stress material model. Section 4 describes the experimental and numerical results and the article is closed with important conclusions at the end.

2 Experimental methods

2.1. Chemical Composition

The material (22MnB5 steel sheet of thickness 1.8 mm) used for the present investigations is supplied by the company voestalpine Stahl GmbH 2015 under the brand name phs-ultraform. The coating on the metal sheet protects from oxidation during austenization. In our investigations, the coating on the material is removed and all our experimental characterizations are carried with initial sheet thickness of 1.65 mm. The chemical composition of this batch has been determined in house and the individual element compositions are in agreement with the published results reported in literature see, Table 1 [11].

Table 1. Chemical composition of phs ultraform (wt %)

C	Si	Mn	P	S	Cr	Al	Ti	B	Ti, V, Nb
0.224	0.227	1.213	0.013	<0.001	0.231	0.047	0.033	0.003	0.036

2.2. As-received material properties

Tensile tests were conducted at ambient temperature according to DIN EN 10002-1 standard on the as received material with (ferrite + pearlite) microstructure in (0^0) rolling, (90^0) transverse and (45^0) directions. The material doesn't show strong anisotropic behavior (Fig. 1) in the as received state. For use in hot stamping applications, the material can be assumed as isotropic [12] in the present work and its flow behavior is approximated using von-Mises yield potential.

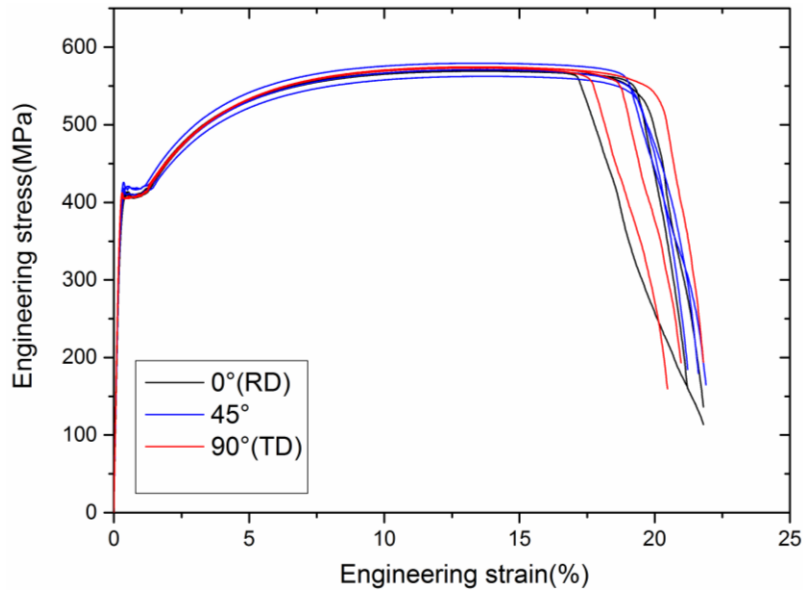


Fig.1: As received uniaxial stress-strain curves of phs-ultraform (22MnB5 steel)

2.3. Thermomechanical tensile tests

In hot stamping, the forming of the component must be completed during the austenitic phase of the material to take the advantage of ductility at this stage; therefore, its plastic behavior at different temperatures and strain rates is determined using Gleeble 3150. In general, the plastic behavior of austenite is accomplished either by tensile [18] or compressive [19, 10] tests. Since the material available for the present investigation is in the form of sheet, which is further used for producing hot stamped components, it is advisable to study the behavior of the rolled sheet using tensile tests instead of compressive tests on cylindrical specimens. Since austenitic phase exists only at temperatures greater than A_{e3} , the cooling must be rapid to avoid the formation of diffusional phase transformations while testing the material at temperatures below A_{e3} . In the present work, sheet specimens (Fig. 2b) are manufactured according to the Gleeble 3150 hand book and tensile tests are performed with the thermomechanical cycle mentioned in Fig. 2a for three different temperatures (650, 750 & 850⁰C) and strain rates (0.01, 0.1 & 1/s). Further stress relaxation and creep studies are carried out using the same specimen (Fig. 2b) shape and heating cycle (Fig 2a).

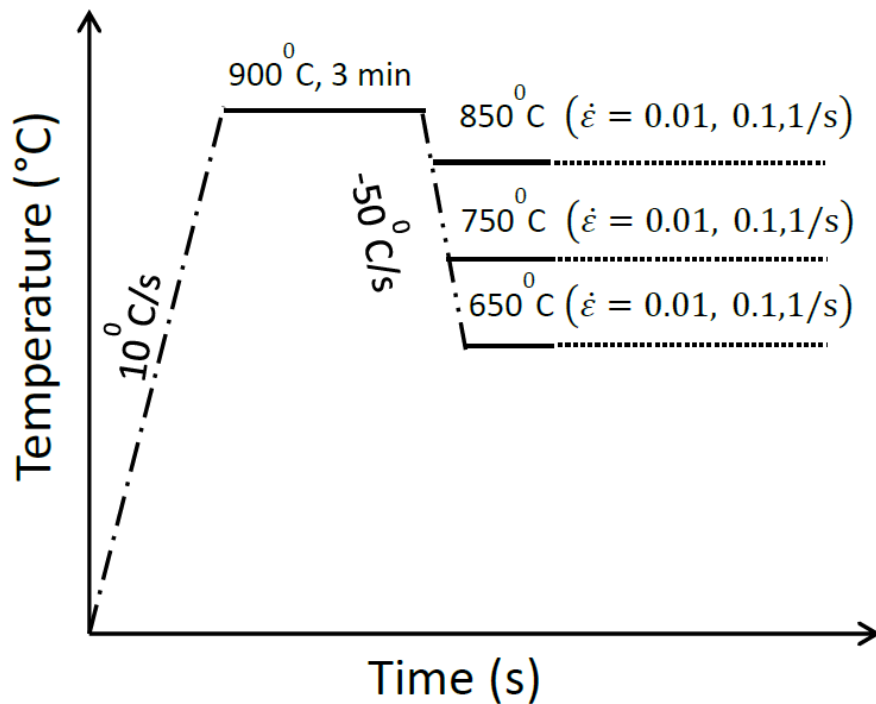


Fig. 2a: Thermo-mechanical cycle for tensile tests at different temperatures and strain rates

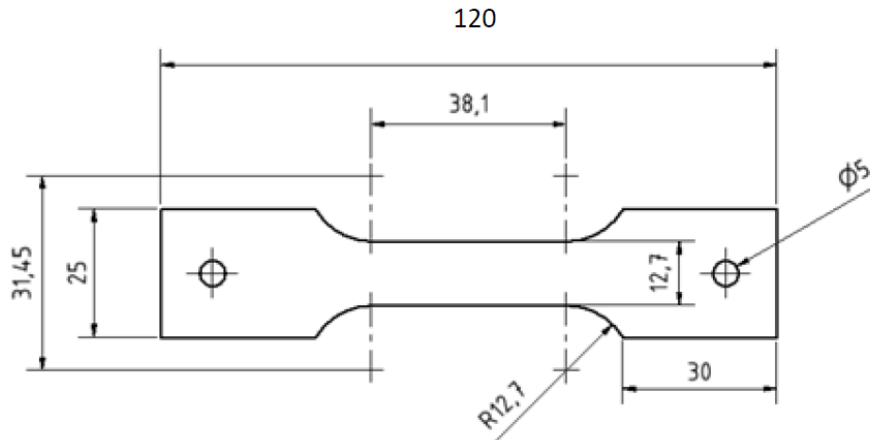


Fig. 2b: Dimensions (in mm) of the specimen used for thermomechanical tension tests

In Gleeble 3150, the material is heated through electric resistance. The choice of an appropriate specimen size is very important due to the existence of a temperature profile in the longitudinal direction of the specimen. In this work, the specimen size recommended by DSI systems for Gleeble 3150 (Fig. 2b) is used and a constant temperature span of 10 mm exists in the center of the specimen. The feedback signal necessary for the closed loop control is obtained through a Cr-Ni thermocouple welded to the specimen surface at the center. The strain in the specimen is measured using a high temperature displacement transducer with a gauge length of 10 mm. Temperature, strain and strain rate are controlled through a computer attached to Gleeble. The default PI controller machine parameters set in the machine are only suitable for attaining low cross head speeds, i.e. a strain rate till 0.1/s can be achieved without large fluctuations during the start of the test. Therefore, for testing the material at strain rates greater than 0.1 /s, the default machine control parameters are modified to maintain the strain rates in the test as desired by the program. But a proper fine tuning of the parameters to control the speed of the jaws as programmed demands a higher number of specimens for testing. Instead, an initial correction on these control parameters is attempted and the observed strain rate is directly used in the simulations to determine the material model parameters which will be discussed in the next section.

3 Viscoplastic flow stress model

To model the flow stress behavior of austenitic microstructure in hot stamping process, several functions exist in literature; most of them are cited in the review article (cf. Table 3, [1]) on hot stamping. The models mentioned in this review article use equations with state variables like plastic strain, strain rate and temperature etc. The major problem with such function like representation of flow stress is their inability to consider the strain path history of the material. In hot stamping process, one can find a process chain with different sub steps like heating, forming and quenching. The microstructure evolves with time after each individual sub step; therefore it is very important to model the flow stress behavior of the material with evolution equations that consider the material microstructure history more accurately. In the present work, Perzyna [20]

type viscoplastic equations are chosen to model the flow stress behavior of austenite phase. The material parameters in this model are considered as linear function of temperature for simplicity.

The material is assumed to be isotropic and yields according to von-Mises criterion with yield function f

$$f = \sigma_{eq} - \sigma_y, \quad \sigma_{eq} = \sqrt{\frac{3}{2} s_{ij} s_{ij}}. \quad \text{Eq. (1)}$$

where σ_{eq} is the von-Mises equivalent stress, which is determined using Eq (1) from the deviatoric stress components s_{ij} .

The yield radius of the material with an initial yield stress σ^0 as a function of temperature (T) in $^{\circ}\text{C}$ is given by

$$\sigma_y = \sigma^0 + R, \quad \sigma^0 = s_1 - s_2 T. \quad \text{Eq. (2)}$$

where R represents the hardening in the material which is followed by

$$\dot{R} = (\bar{\gamma} - R)b\dot{s}_p - \alpha R, \quad \bar{\gamma} = \gamma_1 - \gamma_2 T, \quad b = b_1 - b_2 T. \quad \text{Eq. (3)}$$

$(\bar{\gamma}, b, \alpha)$ are parameters in the hardening Eq. (3), the parameter α is assumed to be temperature independent and is responsible to model the phenomena of static recovery in the material after forming and before the start of martensitic phase transformation. A more physics based approach to model the recovery phenomena using dislocation density theories is due to Estrin [21]. If we neglect the static recovery part in Eq. (3) i.e. ($\alpha = 0$), then the classic exponential hardening law [22] which accounts for dislocation storage and dynamic recovery mechanisms can be retained.

Perzyna type [20] viscoplastic multiplier is used to represent the time dependent effects in the material,

$$\dot{s}_p = \frac{1}{\eta} \left\langle \frac{f}{r} \right\rangle^m = \left\langle \frac{f}{\bar{\eta}} \right\rangle^m \quad \text{Eq. (4)}$$

$$r = 1 \text{ MPa}, \quad \bar{\eta} = \eta^{1/m}, \quad \bar{\eta} = \eta_1 - \eta_2 T.$$

η is related to the dynamic viscosity term with time units and m is the rate sensitivity parameter. In the parameter identification procedure, η tends to become very large, therefore the parameter is taken inside the brackets and a new one $\bar{\eta}$ is introduced. Due to this, η turns out to be a 7th order polynomial in temperature and this makes it tough to represent the units for parameters ($\eta_1 = 256.59$ & $\eta_2 = 0.1308$). For this reason, these parameters are not listed in Table 2. Furthermore, the bracket $\langle x \rangle = (x + |x|)/2$ distinguishes between elastic and plastic states.

A total of 10 parameters in the above described model are identified through the monotonous stress strain curves generated from three different temperatures (650, 750 & 850 $^{\circ}\text{C}$)

and strain rates (0.01, 0.1 & 1 /s) along with the relaxation behavior of austenite at 900⁰ C. Particularly, the stress-strain data at 650⁰C and at 0.01/s strain rate is excluded from the parameter identification due to the fact that austenitic microstructure cannot be retained after a strain level of 0.04. Due to the incorporation of recovery term in the hardening model (Eq. 3), it is not possible to obtain an explicit function like expression to represent hardening with plastic strain, strain rate and temperature. Instead, a full elasto viscoplastic constitutive material model Eq. (1-4) calculation for one material point is evaluated for each time step; such calculation using a finite element program like ABAQUS consumes more time due to the long relaxation durations. The time for each simulation will have more impact when combined with optimization runs. This process is completely avoided and a standalone 3D elasto viscoplastic program to simulate uniaxial strain paths up to uniform elongation is developed in FORTRAN and combined with the optimization tool box SMAT [23] in MATLAB to identify the parameters in the model. This approach is quite faster as there is no need to perform Global iterations like finite element programs. The obtained parameters are listed in Table 2.

Table 2. Material model parameters

s_1	s_2	γ_1	γ_2	b_1	b_2	α	m
(MPa)	(MPa/°C)	(MPa)	(MPa/°C)	(-)	(-)	(1/s)	(-)
94.155	0.10457	481.0497	0.477275	19.6924	0.014176	0.041317	7.05977

4 Results and Discussion

Most of the material models available in literature (Table 3, [1]) to represent the thermomechanical behavior of hot stamping steels are entirely based on monotonous tension or compression tests. The influence of time is only evaluated by conducting tests at different strain rates. In hot stamping, it is believed that the time elapsed after forming the component and before the temperature in the component reaches the martensitic start (M_s) plays a prominent role. During this time period, it is very important to understand the stress state in the material, which further influences the final properties and also helps to assess the residual stress accurately. Therefore, apart from monotonous tension tests, stress relaxation studies are performed in the current work, which will be discussed in this section in detail.

4.1 Monotonous loading behavior:

In this section, we discuss the important issues associated during the monotonous tension tests of austenitic phase at different temperatures and strain rates. The tension tests are performed according to the heat treatment cycle mentioned in section 2. Conducting isothermal tensile tests is the most common way to characterize the material behavior. Fig. 3 shows the tensile behavior of austenitic microstructure for three different temperatures at a strain rate of 0.01/s.

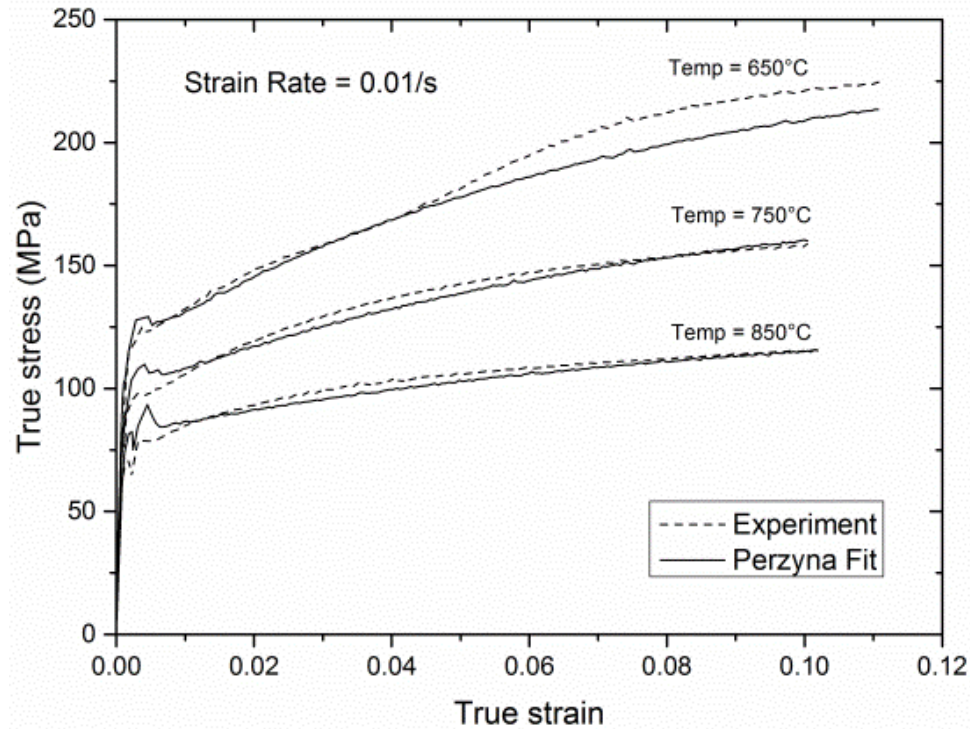


Fig. 3: Stress strain curve of austenitic phase at different temperatures; formation of ferrite at 650⁰C after a total strain of 0.04.

At temperatures below A_{e3} , the tests must be performed with a minimum strain rate to avoid the formation of unwanted phases through diffusional phase transformations. This can be observed with a sudden change in the slope (temp = 650°C, strain rate = 0.01/s) of the stress strain curve (dotted lines in Fig. 3) after reaching a strain of 4%. Therefore, this particular measurement is not included for calibration in the material parameter identification. Even though the flow stress data at 650°C is not used in the material model calibration, the prediction from the material model (continuous lines- Perzyna fit) represents the expected austenite behavior at these conditions.

The experimental results at strain rates 0.1/s and 1/s are shown in Figs. 4 & 5 respectively. It must be noted that the maximum strain rate used in our investigation is 1/s; the default machine controller parameters are modified to maintain the strain rate constant to some extent during the test. Data generated from such tests should not be used to calibrate the parameters in the material model using constant strain rate boundary conditions in the simulation. Such an inverse identification gives fictitious material parameters which results in loss of accuracy when applied to forming simulations.

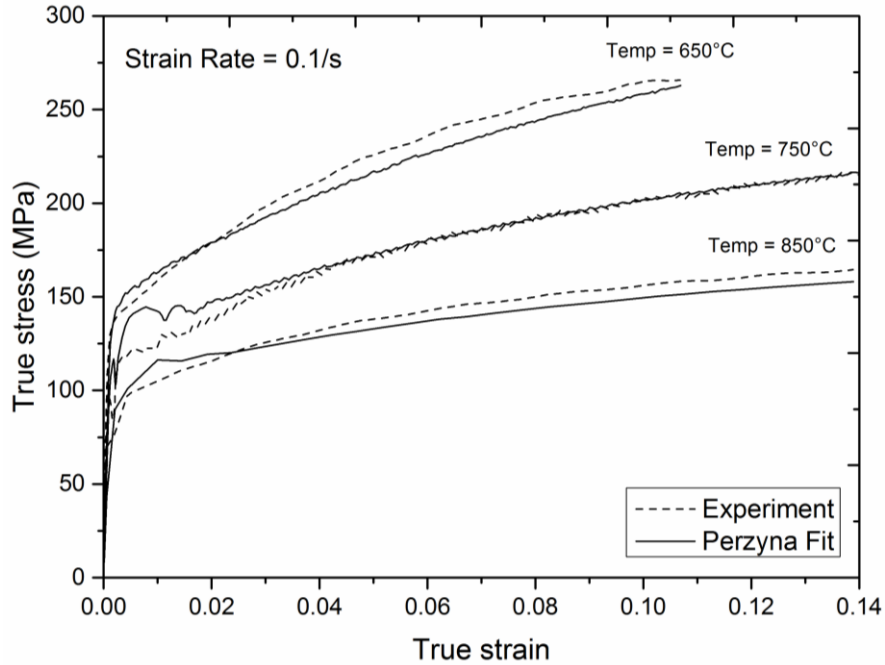


Fig. 4: Comparison of flow behavior of austenitic phase from the hardening model described in Section 3 with experiments carried at a strain rate of 0.1/s

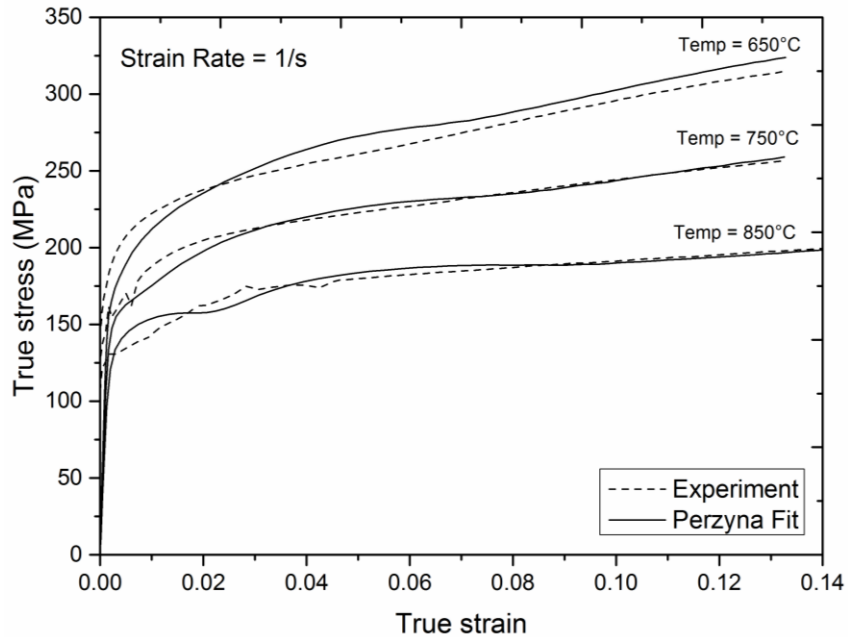


Fig. 5: Comparison of flow behavior of austenitic phase using the hardening model described in Section 3 with experiments carried at a strain rate of 1/s

In the present work, the material model results represented as continuous lines are shown in Figs. 3-5. A viscoplastic material model (Section 3) is calibrated to the experimental data [Monotonous (4.1) and Relaxation (Section 4.3) behavior] and the identified material parameters are reported in Table 2. While simulating the monotonous stress strain paths, the experimental

strain vs. time was given as input to the model; this is to capture the observed strain rate in the simulation during parameter identification. The oscillations in the material model simulations are due to the use of actual observed gauge length displacement of the specimen in simulations.

4.2 Deformation induced ferrite transformations:

Deformation in austenite produces new nucleation sites for diffusional phase transformations and helps to the formation of ferrite. This mechanism is called deformation induced ferrite transformation [16] or shortly DIFT, which is observed in our analysis while testing the material at 650°C with a strain rate of $0.1/\text{s}$. In our experiments, DIFT is observed with the sudden increase in the recorded thermocouple temperature due to phase transformation after certain duration of time. In Fig. 6, the heating cycle of tension test at 650°C with a strain rate of $0.1/\text{s}$ is shown. The blue line (TC1, $^{\circ}\text{C}$) indicates the temperature of the thermocouple attached to the specimen in Gleeble tests for monotonous uniaxial loading and the black line (PTemp) shows the programmed temperature profile, the deformation in the specimen starts at 277^{th} sec approximately, after 3 s duration the specimen reaches the targeted true strain (i.e. 0.2), it is clear that there is a sudden increase in thermocouple temperature during the time elapsed in the absence of the deformation. This can be further confirmed by observing the sudden drop in the machine power angle (Red line) to zero. The increase in thermocouple temperature is not observed while testing the material under similar conditions at 750°C and 850°C . At higher strain rates ($1/\text{s}$), the time elapsed during the test is too short to activate such diffusional transformations in the microstructure. Min et al. [13] also reported the formation of ferrite due to the reduction in the nucleation time due to DIFT while testing 22MnB5 material at the same temperature and strain rate.

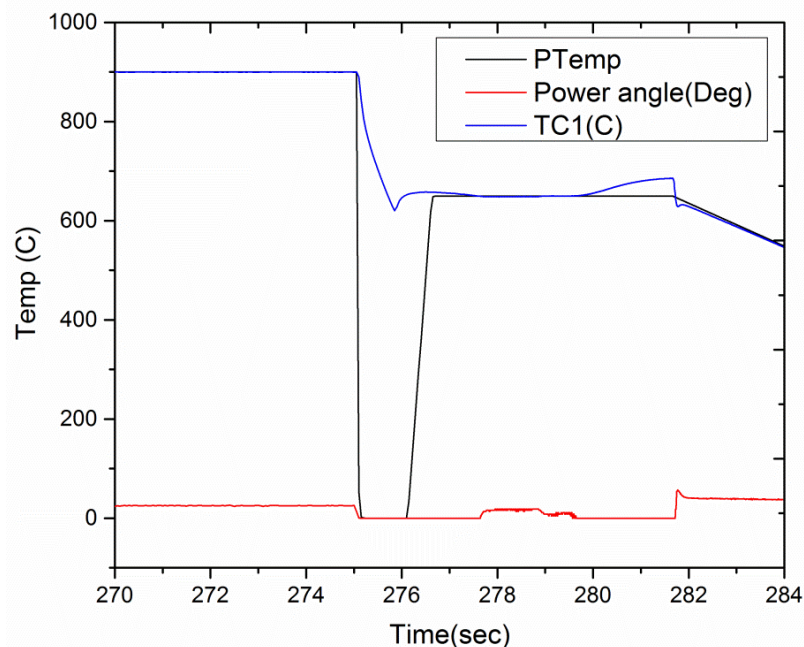


Fig. 6: Sudden increase in temperature (Blue line, TC1) shows the formation of deformation induced ferrite transformation at 650°C

4.3. Stress relaxation studies:

Since austenitic phase can exist only at temperatures greater than A_{e3} , stress relaxation test is conducted in the material at 900°C and at a nominal strain rate of $0.01/\text{s}$; the dimensions of the specimen (Fig. 2b) are similar to the monotonous tests. The as received microstructure (ferrite + pearlite) of the material is heated to 900°C using Gleeble 3150 in 90 s to obtain a uniform austenite grain size with an isothermal hold for three minutes. The specimen is then subjected to various deformation levels (0.5%, 1%, 3%, 5%, & 8 %) with a three minute constant strain hold period after each strain level. Below figures show that variation of stress in austenitic phase with strain (Fig. 7a) and time (Fig. 7b) as independent variables. During the first relaxation cycle that lasts from 0 to 180 s in Fig 7b, an unexpected increase in the material hardening is observed after the first drop in stress. The authors were not able to find an explanation for this observation, the tests are further reproduced and the similar increase in stress is observed once again. From Fig. 7b, it can be realized that the stress level drops to 15 MPa for all deformation levels within duration of 10 s. This is due to the activation of recovery processes in the material and this 10 s duration is also the typical waiting time for the component inside the press during hot stamping. The influence of dynamic recrystallization can be neglected as the working temperature in hot stamping process is around 300°C lower than typical forging process [24].

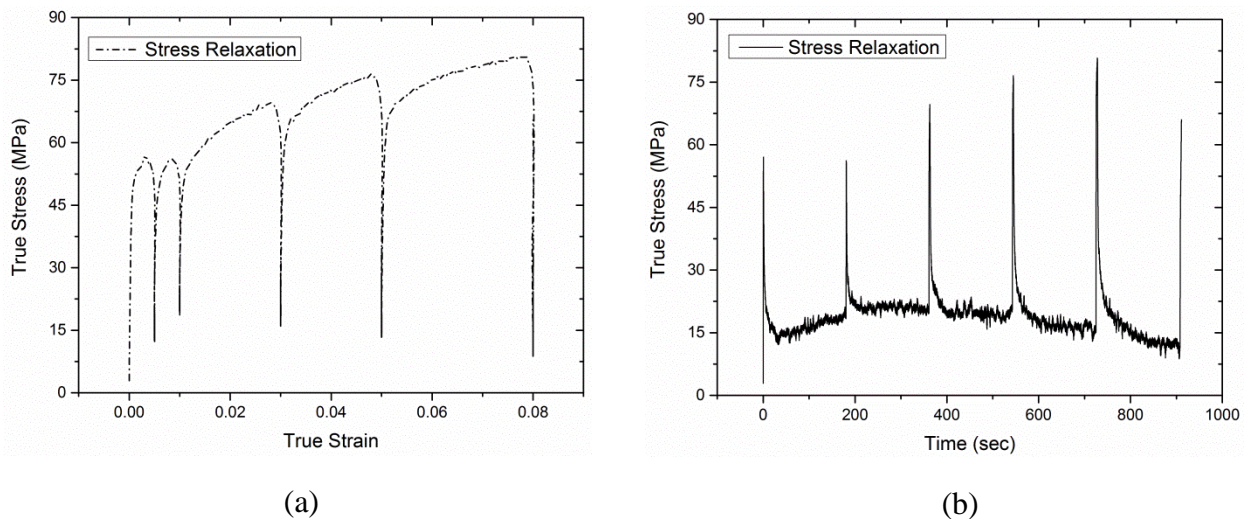


Fig. 7: Stress relaxation behavior of austenitic phase at 900°C with strain (a) and time (b) axis.

Since the stress level in the material drops to a very low level in relaxation experiments (Fig. 7) at elevated temperatures, it is important to know whether there is an elastic range in the material at these temperatures. To realize this, creep tests (see Fig. 8) are conducted with constant stress hold duration of 3 minutes with different stress levels (10, 20, 30, 40 & 50 MPa) at 900°C . There is a significant rise in the strain of the material after 20 MPa, which shows that the material is elasto-viscoplastic after reaching this stress. This information helps to restrict the initial yield radius of the material at elevated temperatures while finding the material model parameters.

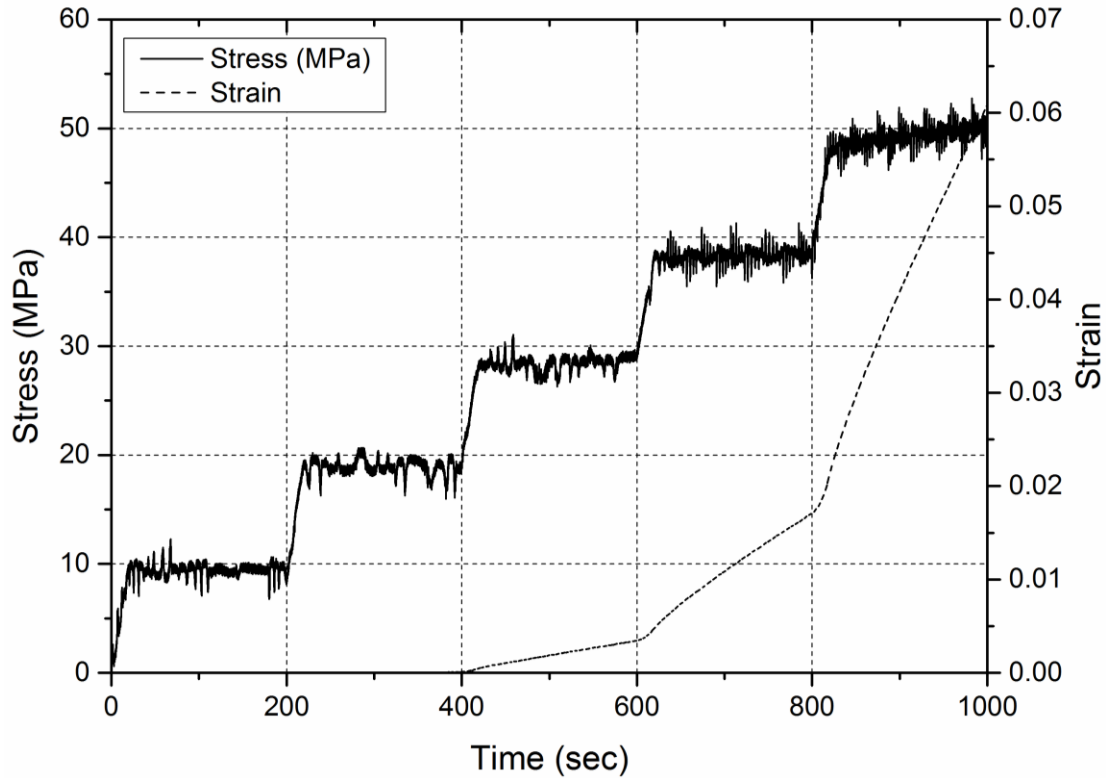


Fig. 8: Elastic range in austenitic phase at elevated temperatures through creep test at 900°C

It is clear that austenitic stress level in the real component drops drastically (Fig.7) before phase transformation occurs. The analysis of [8] shows that the phase transformation temperatures are significantly impacted by the variation in prior stress/strain. This shows the importance of the stress relaxation behavior (Fig. 7) in predicting the final properties of the component accurately. The objective in the present work is to deal with the production of hot stamped components with 100% martensitic microstructures. At temperatures below M_f (250°C), the final microstructure of material investigated in the present work is fully martensitic. During stamping, the component is still inside the press with external loads at these temperatures. Before it further cools down to ambient temperatures, stress relaxation can also occur in martensitic phase. In order to observe such phenomena, first fully martensitic specimens are produced in furnace and the temperature dependence of the martensitic flow stress is analyzed under monotonous loading. The martensitic microstructure hardening behavior is observed to be insensitive to temperature (Fig. 9); this is in line with the experimental observations of [15]. The microstructure is not tested for temperatures more than 200°C to avoid the activation of tempering processes.

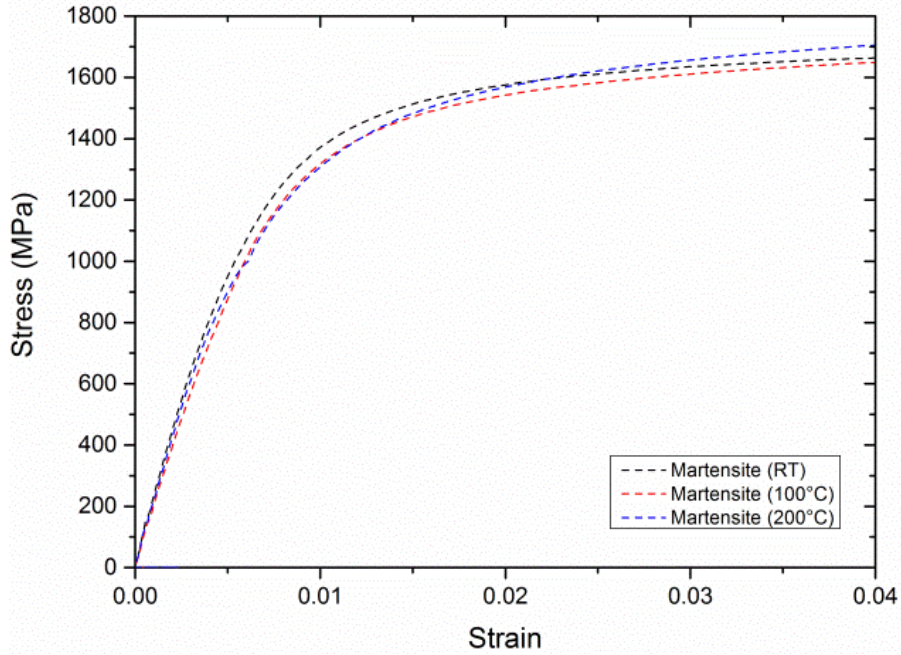


Fig. 9: Temperature dependence on the hardening behavior of martensitic phase

Relaxation tests in martensitic phase are performed at ambient temperature and at 200⁰ C after a deformation of 3% with constant strain hold duration of 10 min (Fig. 10). As a result, martensitic phase shows stress relaxation phenomena even at ambient temperature and the stress drops by 200 Mpa at 200⁰ C after 10 s. The relaxation phenomenon in the martensitic microstructure is observed after a certain amount of plastic deformation in the present work, the origin of such plastic strains in the martensitic microstructure during hot stamping is due to the TRIP effect during phase transformations from austenite to martensite under the influence of external press load.

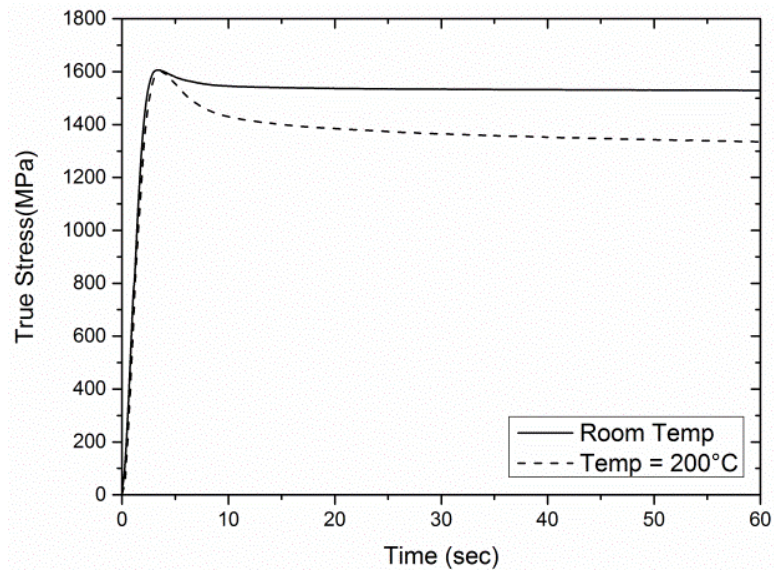


Fig. 10: Stress relaxation behavior of fully martensitic specimens after a pre-strain of 3%

From the above stress relaxation studies in austenitic and martensitic phases we learn that to represent the physical phenomena in hot stamping accurately, the material model must be calibrated with both monotonous and relaxation behaviors. In this framework, as discussed in section 3, an elasto viscoplastic material model that represents the recovery process in the material accurately is developed and calibrated for the present steel. Therefore a recovery parameter α is introduced in Eq. (3) of section 3 to predict the static recovery phenomena during stress relaxation of austenite. If the material parameters in the viscoplastic model are calibrated only using the monotonous data sets (with $\alpha = 0$), the material model underestimates the stress relaxation in the material during the holding time as shown in Fig. 11.

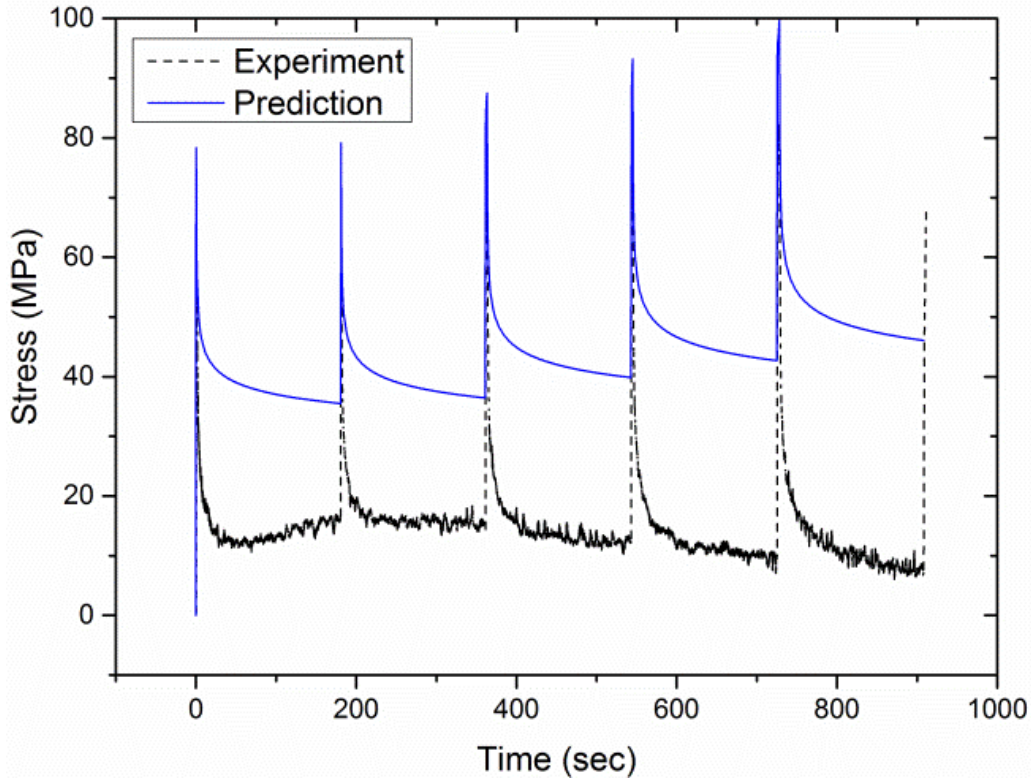


Fig. 11: Comparison between the experiment and simulation stress relaxation behavior in austenite at 900°C without the influence of static recovery parameter ($\alpha = 0$)

The stress relaxation prediction in the material can be effectively improved (see Fig. 12) through the introduction of recovery parameter ($\alpha = 0.0413$) in the hardening model [Eq. (3)]. Since austenitic microstructure cannot be retained at temperatures less than Ae_3 , it is not possible to make relaxation tests at other temperature conditions. This is the reason for choosing a constant static recovery parameter (α) in this work. The material parameters (Table. 2) in the material model are identified using eight different data sets from monotonous loading behavior together with a single data set for relaxation behavior. Even though the relaxation behavior is not close to the observed data, the idea of the present work is to develop a simple approach that can represent all possible important physical mechanisms observed during hot stamping. The stress relaxation simulations with the new hardening approach are in decent agreement (see Fig. 12)

with the experimental findings of the present work, whereas the monotonous loading behaviors (Figs. 3-5) are in good agreement considering the simple linear dependence on temperature.

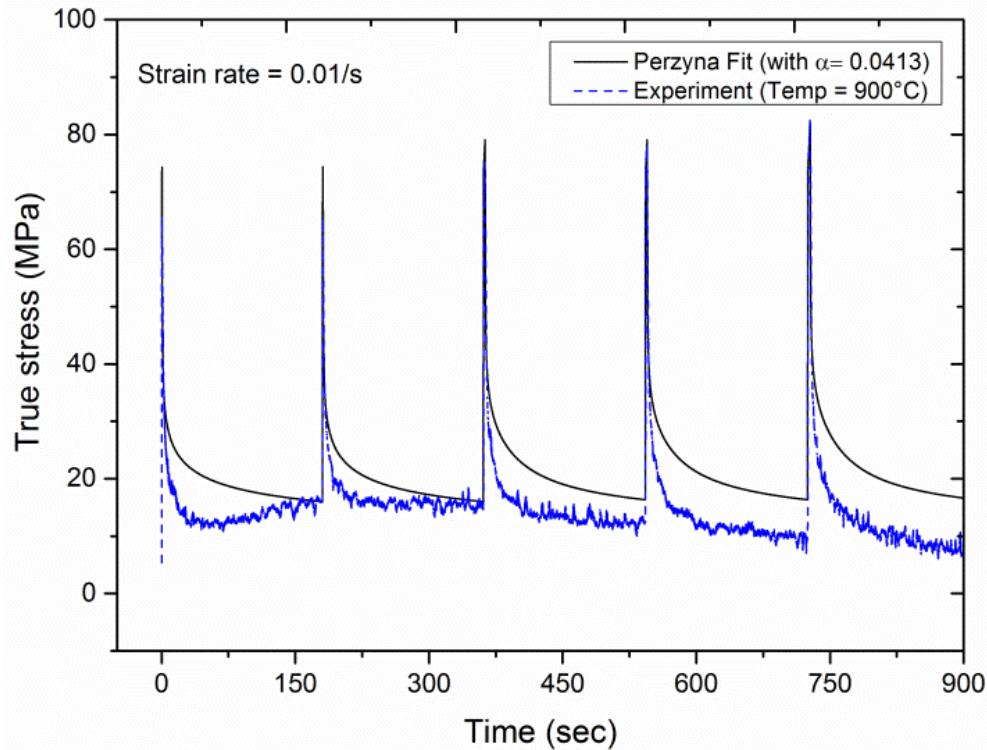


Fig. 12: Influence of the static recovery parameter on the prediction of the stress relaxation behavior in austenite at 900⁰C

Conclusions

In this study, the flow stress behavior of 22MnB5 steel grade is characterized using Gleeble 3150. Thermomechanical tests are performed at different temperatures and strain rates for austenitic and martensitic microstructures through uniaxial tension tests on sheet specimens. The importance of studying time dependent effects like stress relaxation and creep behavior in hot stamping process is demonstrated efficiently through the reported experiments. Representation of static recovery in 22MnB5 steels using Perzyna type viscoplastic equations with a single recovery parameter improves the relaxation predictions without compromising the monotonous behavior of the material at high temperatures and strain rates. The present approach is a phenomenological attempt to represent the physical phenomena (like work hardening, dynamic recovery and static recovery mechanisms) during hot stamping process. The need of using the experimentally observed strain rate in material model simulations for the accurate representation of hardening is discussed. A unique attempt has been made in hot stamping steel grades to identify the material model parameters using both monotonous and stress relaxation behavior of the microstructure phases. Finally, the authors would like to conclude that the present work will enhance the accuracy of FEM simulations and further improves the property predictions in hot stamped components.

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