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On the role of the residual stress state in product manufacturing

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ABSTRACT

This paper concentrates on the effect of the residual stress state during product manufacturing of AISI 420 steel on the final shape of the product. The work includes Finite Element (FE) calculations of the distribution of the residual stresses after metal forming and a heat treatment. The evolution of the residual stress state is predicted employing a creep model which was calibrated through creep experiments ranging between 500 and 700 °C. The creep model is implemented in the FE software together with material specific parameters to simulate a thermal treatment of the product including residual stresses. The complete simulation is validated by a comparison of the predicted shape change with the actual cold-formed and hardened product. An important message of this paper is that, in contrast to what is commonly believed, residual stresses play a crucial role during creep and that a shape change may occur at lower temperatures, i.e. not at higher temperatures and in fact below the transformation temperature in this AISI 420 steel. This surprising conclusion is essential for keeping the shape change during product manufacturing minimal.

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1. Introduction

Modern consumer products need high precision metallic parts so as to meet the quality demands of costumers. However, the manufacturing process of high precision components suffers from a large number of complexities, i.e. the components typically have demanding geometric specifications in three dimensions and other critical-to-quality requirements, such as hardness, surface roughness and density [1–3]. Commonly used consecutive processing steps include metal forming so as to achieve the desired shape and followed by a heat treatment to obtain the required mechanical strength. In general the heat treatment makes the shape of the product less accurate. In this contribution we concentrate on the influence of the residual stress state on the shape change during hardening of a cold formed product. It is emphasized that this paper does not have the intention to measure the precise quantity of residual stresses but rather to demonstrate its influence on the combination of product and shape change.

It should be realized that in practice the complete product manufacturing consists of shaping operations (i.e. forming), heat treatments, finishing operations and assembly. When martensitic chromium steel is used as a base material, the most critical step lies in the middle of the production chain, i.e. the hardening of the steel. Here, residual stresses introduced by forming may cause plastic deformation during

hardening. However, the material is commonly known to produce little shape change during the hardening because in most applications the material is already made stress-free prior to a heat treatment. Obviously, fewer finishing processes are needed when after the hardening the product is already more accurate in shape. A reduction of the finishing processes increases the production efficiency and therefore lowers the fabrication costs. As a consequence there is a strong need to simulate the shape of the components through the various processing steps so that the consequence of any modification of the production process can be predicted in advance. Also novel products can be tested thereby decreasing the design-to-product time.

For the aforementioned hardening, a thermic driven process which should not be confused with work hardening, the typical austenitization temperature for martensitic stainless steel ranges between 925 and 1065 °C for 30 to 90 min [4]. After forming by cold work, which could include stretching and bending, the material has a high residual stress state [5]. In general we may say that there are three mechanisms responsible for the relaxation of residual stresses [6]: (i) At lower temperatures creep mechanisms (diffusional and dislocation creep affected by grain size) allow areas of tensile and compressive stresses to contract or expand; (ii) At high temperatures the yield strength decreases, promoting strain relieve through dislocation glide and dislocation creep mechanisms; and finally (iii) precipitation and ageing effects may lead to volumetric changes that can also relax the residual stress state. We should emphasize that in many circumstances it is a combination of processes and various relaxation mechanisms which may operate at the same time. When diffusional creep (Nabarro–Herring, Coble) occurs

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at low temperature it is still active at high temperature where dislocation creep and dislocation glide maybe become operational depending on the stress state, i.e. the ratio between then effective stress state and the temperature dependent shear modulus.

Several authors have investigated various aspects of the heating cycle: i.e. calculation of internal stresses during martensite forming [7], FE modeling of residual stress [8,9] and stress relaxation [10] during quenching; high temperature deformation [11] and the evolution of distortions during quenching [12]. Surm et al. [13] noticed the importance of the residual stress state during heating, attributing stress relieve to plastic deformation due to the decrease of the yield strength at high temperatures. Relaxation of residual stresses with a creep model has been shown in [14,15], however no predictions for shape change have been done and that is the novel contribution in this paper.

In general, shape change of steel during a hardening treatment is attributed to the transformation induced stress or thermal stress [7,16,17]. But, little attention is paid to the role of residual stresses in the shape change at the lower temperatures of the heating cycle. In this paper we will in particular focus on the contribution of the low temperatures heating region to the overall shape change during hardening and the role of residual stresses.

In this paper a novel approach will be proposed to predict and validate the shape change during a hardening treatment of a metal formed product in 3-Dimensions. The overall idea is that forming of a metallic product induces inhomogeneous strains resulting in residual stresses. The latter are calculated by using the commercial Finite Element (FE) software package 'Marc'. The residual stress distribution is obtained by making a comparison between the simulated work hardening with the hardness profile of the product.

Key in the simulation of the heat treatment step is the assumption that the residual stresses leads to creep and that the creep strains may predominate the shape change of the product. A creep model is calibrated in single creep experiments to predict creep strains below the austenitization temperature, and implemented in the FE calculations. The accuracy of the model is evaluated by making a comparison with the shape of a product in reality.

2. Experimental procedure

The material under investigation is a cold rolled steel strip of martensitic stainless steel of class AISI 420 with chemical composition as listed in Table 1. The product is fabricated using cold forming by deep drawing. Its shape is characterized by measuring the flatness of the bottom of the cup. This bottom flatness is determined along the circular track indicated in Fig. 1. The height profile of small domains (in the z-direction) on this track was measured with confocal microscopy, completed with a fit through the average of many neighboring domains. The shape change in the product during heating is monitored by measuring the bottom flatness before and after the hardening. (See Table 2.)

The FE calculations of the forming assume planar anisotropy (Hill 1948 [18]) and a dislocation work-hardening description (see [19]). Due to the anisotropy in the material the residual stress state will also become inhomogeneous.

The calculation for the shape change during heating includes: elasticity, thermal expansion, plasticity, and creep. The assumption here is that the various shape changes can be added linearly to the total strain, with the same global stress acting on the various individual elements. Consequently, the underlying constitutive equations for a specific mechanism can be determined individually. Hence the combined total

Fig. 1. Cross section of a deep drawn cup with a radius (red line) on the bottom where the z-values are measured. Top: 3D line indicating the shape change in the z-direction.

strain rate equation is written as the sum of elastic strain rate, thermal strain rate, plastic strain rate and creep strain rate:

$$\dot{\epsilon}_{tot} = \dot{\epsilon}_{elastic} + \dot{\epsilon}_{thermal} + \dot{\epsilon}_{plastic} + \dot{\epsilon}_{creep} \quad (1)$$

and all contributions to the total strain are found by integration of the various strain rate contributions over time in the subsequent order, i.e. first elastic plastic behavior followed by thermal strain, plastic strain and finally the creep strain. The description of the constitutive equations can be found in [20]. For the FE simulation of the deep-drawing (at room temperature), Eq. (1) reduces to: $\dot{\epsilon}_{tot} = \dot{\epsilon}_{elastic} + \dot{\epsilon}_{plastic}$. This initial residual stress state is further evaluated during the heating stage. The creep rate relation used in this work is only depending on temperature and stress, and can be calculated separately. By evaluating the decrease of the residual stress state due to creep strain, it can be observed that the residual stresses stay below the yield point during the heating stage. Therefore the contribution of the plastic strain is not taken into account in this trajectory.

Since creep forms an important phenomenon in our work a brief summary of the equation used in the following is given here. Strain

Table 1
Chemical composition of AISI 420 class of martensitic stainless steel (wt.%) [38].

C	Cr	Si	Mn	P	S	Fe
Min. 0.15	12–14	1	1	0.04	0.030	Bal.

Table 2
Experimentally obtained values for the functional set of variables n , Q_c and A of the creep model (see Eq. (2)) in the temperature range 500–700 °C.

	500 °C	600 °C	700 °C	500–600 °C	600–700 °C
n	2.1	5.4	5.3	$-23 + 0.033 \cdot T(K)$	$-6.6 - 0.001 \cdot T(K)$
Q_c				57 kJ/mol	141 kJ/mol
A				$-123 + 0.16 \cdot T(K)$	$-12 + 0.008 \cdot T(K)$

Table 3

Initial loads used for stress relaxation tests at different temperatures.

	500 °C	600 °C	700 °C
Initial load #1 (MPa)	210	160	100
Initial load #2 (MPa)	160	120	75

rate due to creep in solid solutions is described as a diffusion controlled process [21–24] by:

$$\dot{\epsilon}_{\text{creep}} = A \left(\frac{b}{d} \right)^p \left(\frac{\sigma - \sigma_0}{E(T)} \right)^n e^{-\frac{Q_c}{RT}} \quad (2)$$

where b represents the magnitude of the Burgers vector, d is the grain size diameter p represents the grain size exponent. Q_c represents the activation energy for self-diffusion, in bulk or along grain-boundaries or along dislocation cores, depending on the actual operating mechanism. The external applied stress is represented by σ , the threshold stress is σ_0 and $E(T)$ represents the temperature dependent Young's modulus. Each of these parameters has to be determined experimentally.

As a consequence, creep and tensile tests at elevated temperature were performed with a setup consisting of a Zwick/Roell Z30 tensile bench, equipped with a three-zone induction furnace and a Fiedler Laser Extensometer (LEX). True strain ϵ was derived by measuring the extension of two markers on the tensile bar with the LEX. Markers on the tensile bars were spaced 50 mm and their distance was measured with a precision better than 1 μm (0.002%) at 700 °C.

3. Results and discussion

In this paper we concentrate on the initial stress state after forming. Residual stresses could also arise from thermal gradients and phase transformation, which however are not considered. The reason is that, as we will show, the initial residual stresses due to forming cause the majority of shape change, before a phase transformation. In another publication we will investigate more explicitly the contribution of thermal gradients and phase transformation to improve the shape change calculations.

3.1. Creep model calibration

Tensile tests in the range of RT–900 °C were performed to determine the dependence of the yield stress on temperature, and the results are displayed in Fig. 2. The minimum value at 800 °C marks the onset of austenitization. The E-modulus also has a temperature dependence [25–30] as was derived from the tensile experiments. This is not always very reliable since at elevated temperature creep effects make the elastic regime shorter and often not even linear. For the experiments above room temperature, the markers on the tensile bar were measured with a high precision, prior to the beginning of the tensile test. The start of the tensile tests at 300 °C caused disturbance in the recording of the strain, with the result of a derivation of the elastic modulus which is too high. The values of 500–700 °C have been obtained from tensile tests with various strain rates, i.e. 0.012–0.00012 s^{-1} . Especially at these temperatures creep phenomena influenced the elastic response and made the elastic regime shorter and for the highest temperatures and lowest strain rates, non-linear. Given these results, tensile tests at elevated temperatures are not the optimal experiments to determine the Young's modulus. The Refs. [25–29] supplemented with the work of Masumoto [30] with acoustic waves on similar steels, indicate a mainly linear decrease of Young's modulus up to 700 °C. Therefore we assume a linear fit to the data of Fig. 2 is appropriate.

Creep tests at 500 °C, 600 °C and 700 °C were done at constants load in the elastic regime. The steady state creep rates are displayed in Fig. 3 and show an increase over several orders of magnitude between 500 and 600 and 600–700 °C for a stress around 100 MPa. The stresses

have been chosen to about maximum 75% in the elastic zone for every temperature. Therefore the temperature dependent flow stress determined in Fig. 2 acts as an upper limit. For higher stresses the creep regime changes from a power-law to exponential creep, whereby the steady state creep is a relative short state and tertiary creep is active.

The slope of the double log plot yields the stress exponent n of 2.1, 5.4, and 5.3 at 500, 600 and 700 °C, respectively. The creep stress exponent is an indication of the creep rate controlling mechanism. For $n = 3$, viscous dislocation glide is rate controlling; for $n = 4$ –6 dislocation creep is rate controlling through dislocation climb [31,32]. Likely both creep mechanisms are active simultaneously. Therefore, dislocation glide is the predominant mechanism between 500 and 600 °C and dislocation climb is the predominant phenomenon at 600–700 °C. For the threshold stress in Eq. (2) σ_0 a value of zero was derived.

Strain rates at different temperatures provide information about the activation energy for creep Q_c [23] and two creep regimes were found, i.e. Q_c is 57 ± 30 and 141 ± 2 kJ/mol for 500–600 °C and 600–700 °C respectively. The latter value of Q_c is close to self-diffusion activation energy whereas the former points to carbon diffusional processes.

The microstructure was investigated in detail with electron back scatter diffraction (EBSD). A specimen from the strip of material (as-received) was heated 30 min at 700 °C prior to analysis. More than 1000 grains from each sample where measured, but no significant grain growth was observed compared to the as-received material. Therefore detailing of the grain size exponent p is not needed.

Finally, stress relaxation calculations based on this creep model are done using a FE bending test. The sample is heated to the temperature of interest and bended between two tilted dies until the desired stress

Fig. 2. a: Yield stress ($\sigma_{0.2}$) as a function of temperature, obtained from tensile tests with $\dot{\epsilon} = 0.012 \text{ s}^{-1}$. b: E-modulus derived from tensile test (squares) with a linear fit.

Fig. 3. Secondary creep rate as a function of load at 500 °C (triangles), 600 °C (dots) and 700 °C (squares).

(load) is reached in an element on the outside of the sample (inducing tensile stress). Thereafter the dies are fixed in their final positions and the temperature is kept constant. In every time step the creep strain due to the stress is evaluated resulting in a decrease of the stress. This test is repeated at different loads and at various temperatures (Table 3). Fig. 4 shows the resulting equivalent Von Mises stress together with the tensile tests. Here the specimen is deformed elastically due to a particular load, after which the stress relaxation is recorded by the load cell. The calculated stress relaxation is in good agreement with the tests: i.e. for the initial stage 0–10 s (rapid decrease) and the second stage where the stress level remains quasi-static. It demonstrates that the creep model is suitable to calculate stress relaxation below the austenitization temperature of 800 °C.

3.2. Forming

Metal forming by deep drawing introduces residual stresses, which are not uniform distributed along the cup [33]. In reference [34] it is shown that residual stresses increase due to pre-straining, with a relation that indicates a work-hardening (or strain-hardening) mechanism. In order to validate the FE calculations after deep drawing, we have visualized the residual stress by comparing the hardness of the actual formed cup with the simulation. Therefore the hardness of pre-strained material has been used to translate the calculated stress to hardness. This assumption/method has no influence on the further calculations

of the evolution of the stresses during the heat treatment. Hardness measurements have been chosen, as they are an indirect but simple experiment to perform. The justification of this procedure can be found in [35], where a linear relation between the hardness and residual stress is derived.

The deep-drawing process has been done following the well-known procedure described in [36], resulting in a cup with 20 mm in diameter and 10 mm height. The FE simulation of the forming has been done by implementing the deep-drawing parameters into the Marc software. Material specific properties as yield stress and Young's modulus at room temperature have been stated in the software. The deep-drawing is simulated using planar anisotropy and dislocation work hardening (as mentioned in Section 2). The parameters of the Hill'48 planar anisotropy model are derived from the magnitude of the flow stress in various directions with respect to the rolling directions. The dislocation work hardening description of [19] is calibrated with the data presented in Fig. 5.

It is known that the hardness changes depending on the amount of cold work, i.e. work hardening [19]. Experimentally this is verified with tensile tests on pre-strained material. The stress at the onset of plastic deformation (yield stress) for various strains is recorded. Thereafter the hardness of each strained sample is measured. The hardness with the corresponding flow stress after the tensile test is depicted in Fig. 5. Since the flow stress depends on the amount of strain, an indirect relation between the strain (thus residual stress) and the hardness exists. The calculated residual stresses of a cup after forming can therefore be expressed by a calculated hardness profile. The prediction of the hardness is verified in Fig. 6 by comparing the calculation with a hardness profile of a fabricated cup.

The cup was first cold mounted in an epoxy resin and then dissected twice. The mounting prevented spring-back of the cut section and therefore a relaxation of the residual stress. The surface was then carefully prepared for hardness measurements by consecutive refining steps of grinding and polishing: a common method of sample preparation for hardness measurements and EBSD microscopy (where surface stress perturbations by cutting and polishing should be erased). The hardness is measured by a grid of indentations. The contour is obtained by interpolation of the hardness values for the body, and extrapolation to the edges. This method is demonstrated and described in [37].

Based on symmetry, only half of the product cross-section is shown. The black line in Fig. 6a represents the contour of the simulated cup. The difference can be due to geometrical errors in the settings of the press, which cause more deformation of the fabricated cup compared to the FE simulation. The overall hardness distributions match quite well, indicating that the calculation of the residual stresses is validated.

Fig. 4. Stress relaxation tests with fixed strain, lines 2 (squares) & 4 (diamonds); together with the calculated stress relaxation by creep of a bending test, lines 1 (triangles) & 3 (circles).

Fig. 5. Hardness of the pre-strained samples (circles) as a function of flow stress measured in tension, with a continuous fit.

Fig. 6. Hardness profile measured on the product cross-section (a), and FE simulation (b). The black line in (a) is the contour of the simulated cup (b) and indicates the deviation between the product and the simulation.

3.3. Stress relief during hardening

The creep model is implemented in the FE software and enables calculation of the evolution of the residual stresses and creep strains during a heat treatment. The residual stresses in the cup calculated by FE modeling of the forming process are shown in Fig. 7a. This is the initial state before heat treatment. In the simulation of the heat treatment a temperature profile from 20 to 700 °C in 1000 s is applied. Due to the convergence of the constitutive equations, the calculation can be done in just 11 FE steps. The stress decreases significantly to nearly 0 MPa. Although the yield strength decreases with increasing temperature considerably (as shown in Fig. 2), the residual stresses remain in the elastic zone during the entire heat treatment. The relaxation by creep lowers the residual stress faster than the decrease of the yield strength as a function of temperature.

The resulting creep strains affect the bottom flatness and are therefore responsible for shape change during heating. It has been noticed that the shape change of a full hardened cup is equal to a cup heated to 700 °C. This observation is supported by the calculation in Fig. 7, showing that almost all stresses have been relaxed after heating to 700 °C. The bottom flatness as described in Section 2 has been measured on a fabricated cup after forming and after a full hardening cycle.

The maximum distortion in the bottom flatness after forming is 5.4 and 2.9 μm for the fabricated and simulated cup, respectively. The normalized profiles for bottom flatness after forming and hardening are presented in Fig. 8. For a clear comparison of the relative increase after

Fig. 8. Bottom flatness of the fabricated (solid lines) and simulated cup (dashed), normalized at 90°. Bottom flatness after forming is presented in red and after heating in black.

hardening, all graphs are normalized. The maximum values at 90° after forming, are the reference points for the simulated and fabricated cup. The typical shape, i.e. two peaks and two valleys around the circular track on the cup, is caused by the anisotropic behavior of the cold rolled material. The profile of the calculated cup after forming and hardening shows a very good agreement with the processed manufactured cup. The relative increase of the shape change due to heating is 22% for the fabricated and 15% for the simulated cup. As shown in Fig. 6, the fabricated cup is slightly more deformed during cold forming due to unforeseen geometrical errors in the settings of the press. This resulted in slightly higher stresses in the bottom of the cup. Higher stresses lead to higher creep rates (Fig. 3, Eq. (2)). Indeed, the resulting higher creep strains could have caused the difference in shape change during the hardening of the produced cup, compared to the calculated bottom flatness.

4. Conclusions

A full simulation of the cold forming of a martensitic steel cup and a following heat treatment has been performed. The distribution and magnitude of the residual stresses was successfully predicted in comparison with the micro hardness profile of a fabricated cup.

Fig. 7. Cross-section of simulated cups with contour plots of the equivalent Von Mises stress. a: local high stresses after forming. b: relaxation of the stresses after heating to 700 °C.

The relaxation of the residual stresses can be described with a creep model. Two creep regimes were experimentally identified in the heating trajectory to 700 °C. The relaxation by creep lowers the residual stresses faster than the decrease of the yield stress by temperature. Observations on the fabricated products show that the effect of the completion of the heating cycle (hardening) after heating to 700 °C has only a negligible effect on the shape change. Therefore the shape change after forming is mainly due to residual stress induced creep strain.

The calculated relative shape change during heating to 700 °C was found to be in agreement with the observed shape change on a fully hardened product.

An important message of this paper is that, in contrast to what is commonly believed, residual stresses play a crucial role during creep and that the shape change occurs at lower temperatures, i.e. not at higher temperatures and in fact below the transformation temperature. This surprising conclusion is the key for keeping the shape change during product manufacturing minimal.

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