High temperature and dynamic testing of AHSS for an analytical description of the adiabatic cutting process

S Winter¹, F Schmitz², T Clausmeyer², A E Tekkaya² and M F-X Wagner¹

¹Institute of Materials Science and Engineering, Chemnitz University of Technology, 09125 Chemnitz, Germany
²Institute of Forming Technology and Lightweight Construction, Technical University of Dortmund, 44227 Dortmund, Germany

E-Mail: sven.winter@mb.tu-chemnitz.de, fabian.schmitz@iul.tu-dortmund.de

Abstract. In the automotive industry, advanced high strength steels (AHSS) are widely used as sheet part components to reduce weight, even though this leads to several challenges. The demand for high-quality shear cutting surfaces that do not require reworking can be fulfilled by adiabatic shear cutting: High strain rates and local temperatures lead to the formation of adiabatic shear bands (ASB). While this process is well suited to produce AHSS parts with excellent cutting surface quality, a fundamental understanding of the process is still missing today. In this study, compression tests in a Split-Hopkinson Pressure Bar with an initial strain rate of 1000 s⁻¹ were performed in a temperature range between 200 °C and 1000 °C. The experimental results show that high strength steels with nearly the same mechanical properties at RT may possess a considerably different behavior at higher temperatures. The resulting microstructures after testing at different temperatures were analyzed by optical microscopy. The thermo-mechanical material behavior was then considered in an analytical model. To predict the local temperature increase that occurs during the adiabatic blanking process, experimentally determined flow curves were used. Furthermore, the influence of temperature evolution with respect to phase transformation is discussed. This study contributes to a more complete understanding of the relevant microstructural and thermo-mechanical mechanisms leading to the evolution of ASB during cutting of AHSS.

1. Introduction

The quality of the cutting edge of sheet metal parts manufactured by adiabatic cutting processes is excellent [1]. Moreover, adiabatic cutting processes have a great potential for the processing of advanced high strength steels (AHSS) [2]. A fundamental understanding of the process and the availability of efficient simulation strategies are necessary requirements to establish this technology for the industrial cutting of sheet materials. The demand for high-quality shear cutting surfaces that do not require reworking can be fulfilled by adiabatic shear cutting: High strain rates and local temperatures lead to the formation of adiabatic shear bands (ASB) [3 - 5]. This process is well suited to produce AHSS parts with excellent cutting surface quality, but both a fundamental understanding of the process and efficient simulation strategies are still missing today.

The key goal of this study is an analytical approach for the prediction of the relevant technological and microstructural processes. First, material characterization of two high strength steels is performed with a special focus on the temperature dependent and strain-rate dependent behavior. Based on this information an analytical model is then used to calculate the local temperature increase in shear bands...
for such high deformation rates. These results help to validate subsequent simulations for the adiabatic cutting process and increase the process understanding of adiabatic cutting itself. Finally, this provides the basic knowledge for manufacturing sheet metal parts with high quality functional faces and to estimate the quality of the cut surface. Furthermore, the experimental and numerical results help to determine process parameters for the adiabatic cutting of AHSS.

2. Experimental

2.1 Materials and experimental setup

For this study two high strength steels, obtained as sheet metals from different suppliers, were used. The first material was the manganese boron steel 20MnB5 provided by BILSTEIN GmbH & Co. KG (Hagen, Germany) in 4.7 mm thick sheets in a hardened condition. The second one was the carbon-spring steel C75S provided by Risse + Wilke Kaltband GmbH & Co. KG (Iserlohn, Germany) in 4.5 mm thick sheets in a hardened and tempered condition. The chemical composition was determined by spectroscopic analysis; the results are shown in Table 1. Both steels meet the required specifications of the given values.

Table 1. Chemical composition of the two used high strength steels, determined by spectroscopic analysis.

|          | C    | Si   | Mn   | Cr   | (ppm) B |
|----------|------|------|------|------|---------|
| 20MnB5   | 0.20 | 0.18 | 1.20 | 0.21 | 19.5    |
| C75S     | 0.72 | 0.29 | 0.74 | 0.32 | 2.3     |

Compression specimens with a diameter of 4 mm and a height of 6 mm were extracted out of the sheet metals. These specimens were used for all experiments discussed below. In order to estimate the anisotropy of mechanical behavior, the samples were taken out both in rolling (RD) and transversal directions (TD), figure 1. The initial microstructures (prior to mechanical testing) of both steels were analyzed by optical microscopy.

![Figure 1. Schematic illustration: Extraction of the compression specimens out of the sheet metal in rolling and transversal direction. The samples have a diameter of 4 mm and a height of 6 mm.](image)

For the characterization of the quasi-static stress-strain behavior, compression tests were performed at room temperature (RT) in a UPM 1475 Zwick/Roell universal testing machine with an initial strain rate of $10^3$ s$^{-1}$. Due to friction at the faces of the samples, the experiments were stopped at 30% applied (engineering) strain. The strain was measured with strain gauges applied on the specimens and using an additional inductive distance measuring system. Furthermore, measurements of the Rockwell hardness (HRC) were performed in a Mitutoyo HTM ARK600 Rockwell hardness tester. The indentation time was 15 s and the force was 150 kp (1471 N).

The main part of the experimental investigations was the dynamic characterization of the steels with a strain rate up to 1000 s$^{-1}$ in a temperature range from RT to 1000 °C. For these experiments, a Split Hopkinson Pressure Bar (SHPB) was used (setup based on the original design of Kolsky [6], figure 2). The SHPB consists of four bars: striker, incident, transmitter and damping bar. During the experiment, the striker bar is accelerated by compressed air and then hits the incident bar that subsequently deforms the specimen. Force data is determined from the strain gauge data measured at four locations on the incident and transmitter bars. The deformation of the specimen is documented
with a high-speed camera, and an optical digital image correlation system (Aramis by GOM) is used to determine surface strain fields. For the RT experiments, this technique tracks the displacement fields of a speckle pattern on the specimen surface. At higher temperatures, the pattern was applied on the end of the incident bar and the beginning of the transmitter bar in order to allow for an indirect optical measurement of global strain in the specimens.

To perform the compression tests at high strain rates and at temperatures up to 1000 °C, local induction heating was used in the SHPB. For temperature isolation, two ceramic plates were applied between the sample and the bars of the SHPB (figure 3). The experiments at elevated and high temperatures were performed in steps of 200 °C from RT up to 1000 °C. Heating up to the testing temperature was achieved in 5 s, followed by the triggering of the striker bar. The temperature measurement was realized using a pyrometer focused on the specimen. After the dynamic tests, the microstructures of both steels were again analyzed by optical microscopy.

2.2 Analytical description
The aim of the analytical model is to enable the analysis of local process and material effects which cannot be directly determined by experiments. The formation and growth of ASB can be used for technological processes like shear cutting. In order to define a feasible set of process parameters, the material response has to be known. Since both materials considered in this study possess temperature...
dependent flow stresses, the locally generated heat and thus the temperature is a crucial factor for the process. Due to the locality of the adiabatic shear band (~40 µm) and short process times (~1 ms), the temperature in the area of interest cannot be measured for a nonstationary process like shear cutting. Here, the area of interest is also difficult to access with a pyrometer and it is thus challenging to directly measure the temperature. Therefore, the focus of the analytical model will be the prediction of the temperature in this region.

Zener and Hollomon [7] observed that the yield strength and the heat capacity each have an influence on the adiabatic shear behavior. In this work, only the influence of the yield strength is taken into account, which underlines the necessity for experimental information on the flow stress evolution of the materials at high strain rates (~10³ s⁻¹), see section 2.

Finding the link between plastic deformation and temperature evolution is based on the internal dissipation equation. The internal dissipation due to plastic power $\dot{W}$ is coupled with the increase of temperature in the following manner:

$$\dot{T} c_p(T) \rho V = \dot{W} = \beta k_f(T) \dot{\varepsilon}^p V,$$

where $c_p$ and $\rho$ are the specific heat capacity and density, respectively. It is assumed that the deformation only takes place in the volume $V$ (see figure 4, shown in red). Simultaneously, the temperature is assumed to be homogeneous and constant in this area. Since only a part of the plastic work is transformed into heat, the Taylor-Quinney factor $\beta$ is utilized to incorporate this specific effect. Considering a typical amount of energy converted into heat, the parameter is set to $\beta \sim 0.9$ [8].

To further simplify the model, it is assumed that plastic deformation solely occurs in the shear zone whereas this zone itself exhibits a uniform deformation that can be described as an amplitude $\Delta$, of simple shear deformation. This amplitude of deformation is normalized with respect to the shear width $l$, such that the shear deformation is given by $\frac{\Delta}{l}$. Hence, the deformation gradient $F = \frac{\partial x}{\partial \mathbf{X}}$ takes the form

$$F = \begin{pmatrix} 1 & \gamma & 0 \\ 0 & 1 & 0 \\ 0 & 0 & 1 \end{pmatrix}.$$

Based on $F$, and the right Cauchy-Green tensor $C = FF^T$, the principal stretches $\lambda_i$ and equivalent plastic strain $\dot{\varepsilon}^p = \frac{1}{\sqrt{3}} \sum_{i=1}^{3} \ln(\lambda_i^2)$ under the assumption of von Mises plasticity can be calculated.

It should be noted that for such high deformations, particularly this formulation of equivalent strain represents a feasible measure [9]. The width of the adiabatic shear band (~40 µm) was determined by experiments analyzing adiabatic blanking of an axisymmetric coin with diameter of 30 mm. The used materials and thickness corresponds to those given in section 2. Figure 4 shows a schematic sketch of
plastic deformation. It is assumed that the crack is initialized after a stroke of 25% of the material thickness \( t \), which corresponds to \( \varepsilon^{p} = 3.8543 = \varepsilon^{P}_{\text{crack}} \).

The temperature at \( \varepsilon^{P}_{\text{crack}} \) can be calculated using the differential equation (1), which was solved by means of the explicit Euler forward integration scheme. Room temperature \( T_0 = 30°C \) and a virgin material with no plastic deformation \( \varepsilon^{P} = 0 \) was considered as initial condition, while the plastic strain rate was assumed to be constant with \( \dot{\varepsilon}^{P} = 10^3 \text{ s}^{-1} \). For all calculations, the temperature dependent tabulated heat capacity of a commercial software, JMatPro, was used. For a first verification, the heat capacity at room temperature was measured. Experimental and numerical data were in good agreement, thus justifying the use of the JMatPro data (\( c_p \sim 460 \text{ [J/(kg K)]} \)). Since the explicit integration needs a sufficiently small time step size, the equation was evaluated in incremental steps of \( \Delta t = 10^{-6} \text{ s} \). Parameter studies of the step size proved no further beneficial effect for shorter step sizes.

3. Results and discussion

3.1 Quasi-static characterization

First we consider the initial microstructure of the two steels. Figure 5a shows the expected fully martensitic microstructure of the manganese-boron steel after hardening. Large martensite plates with a length of 50 \( \mu \text{m} \) can be observed. In contrast, the microstructure of the carbon-spring steel consists of both bainite and martensite, which is indicated by many smaller plates in figure 5b.

![Figure 5](image.png)

**Figure 5.** Initial microstructures of the two high strength steels, investigated by optical microscopy. a) Fully martensitic microstructure of 20MnB5. b) Bainitic and martensitic microstructure of C75S.

The Rockwell hardness of the 20MnB5 material was determined as 43 ± 0.5 HRC, in comparison to 44 ± 0.5 HRC for C75S. Both steels possess nearly the same hardness at RT in the as-delivered conditions. This trend is also evident in the quasi-static compression tests. The results for both materials, in rolling direction and in transversal direction, are plotted in figure 6. It is obvious that the manganese-boron steel has nearly the same yield strength (YS) as the carbon-spring steel. The initial yield strengths for both materials are approximately 1200 MPa and the Young’s moduli are 212–213 GPa, respectively. Strain hardening in the manganese-boron steel up to engineering strains of 2 % is significantly higher, and then similar to that of the carbon-spring steel. Therefore, beyond strains of 2 %, the strength is about 200 MPa higher than the strength of the carbon-spring steel, which is the only difference in the stress-strain behavior between the two steels. We also observe that there is no significant anisotropy between the rolling and transversal directions: Both RD and TD curves are nearly equal for each material. In summary, these results show that both high strength steels exhibit a very similar stress-strain behavior at room temperature under quasi-static compressive loading.
3.2 Dynamic characterization

The temperature dependent flow stress - true strain curves of the manganese-boron steel, measured at a strain rate of 1000 s\(^{-1}\), are shown in figure 7. For comparison, the quasi-static stress-strain behavior at RT is also plotted in the diagram. A very high stress peak at the beginning of the curves in all dynamic tests can be observed. This peak does not result from the material behavior, but it is a measurement artifact related to the experimental setup: The incident bar, the ceramic plates for the temperature isolation and the small specimens have different diameters and elastic stiffnesses. This can lead to complex wave reflection phenomena [10], such that elastic waves are superimposed on the “true” measuring signal and thus form these pseudo stress peaks in all curves. While it is therefore not possible to accurately determine the critical stress for the onset of plastic flow (initial yield stress), yield strengths can be compared reasonably well in the range of larger strains (i.e., beyond the initial stress peaks).

As a first result, we observe that the strain rate sensitivity of 20MnB5 is not very high at RT: the stress strain curve measured during dynamic loading exhibits only a slightly increased strength (by about 80 MPa) level compared to the quasi-static one. This is a relatively small effect given that the strain rate is increased by 6 orders of magnitude. Furthermore strength significantly decreases and ductility increases with increasing temperatures up to 1000 °C. We also note that there is a significant strength drop between the stress strain curves measured at 400 °C and 600 °C, respectively. In this temperature range the strength loss is about 500 MPa and simultaneously the ductility is increased considerably. This observation may well have important consequences for the industrial cutting process because only a small amount of impact energy is needed to reach the thermal softening range between 400 °C and 600 °C in the 20MnB5 sheet metal. This softening then results in shear band initiation in the adiabatic cutting process, and can finally lead to a good cutting result (e.g., in terms of surface quality) of the steel.
Figure 7. Stress-true strain behavior of 20MnB5 from RT up to 1000 °C (measured at a strain rate of 1000 s\(^{-1}\)). For comparison, the quasi-static material behavior at RT is also plotted (dashed line). Flow stresses significantly decrease and ductility increases with increasing temperatures. There is a significant strength drop between 400 °C and 600 °C.

To rationalize the strength drop observed in the temperature range between 400 °C and 600 °C, we analyzed the microstructure of the corresponding specimens after the dynamic experiments, figure 8. The specimen tested at 400 °C is characterized by large martensite plates. While, in comparison to the initial microstructure at RT (see figure 5), the plate size is slightly reduced, the material still exhibits a microstructure that is dominated by martensitic features. After heating to 600 °C, we also observe small martensite needles, but compared to the sample heated to 400 °C the ferrite phase volume fraction (brighter areas in figure 8) has been increased. This indicates that relaxation of the martensite phase by carbon diffusion has taken place. When the carbon content in the martensite is reduced, it transforms back to ferrite, which in turn leads to the lower strength observed in the corresponding mechanical test. Similar microstructural features of so-called high tempered martensite have been reported before, e.g. in [11, 12] for the manganese-boron steel 22MnB5, which has nearly the same chemical composition as the 20MnB5 steel used in the present study. These previous studies documented the temperature dependence of the yield strength from RT up to 450 °C. At RT, the 22MnB5 material had a yield strength of 1500 MPa; at 300 °C, it was decreased to 1400 MPa, and at 450 °C, the yield strength dropped significantly to 950 MPa. The authors of both studies rationalized this material behavior with tempering effects, particularly the transformation from martensite to ferrite; obviously, the new results reported here for 20MnB5 confirm these earlier observations.

The results of the dynamic testing of the carbon-spring steel are shown in figure 9. The grey stress-strain curve again represents the quasi-static behavior at RT. As a first result we observe that the strain rate sensitivity of the C75S material is higher than for the manganese-boron steel: under dynamic loading conditions at RT, the yield strength is higher by about 150 MPa compared to quasi-static testing. Similar to the manganese-boron steel, the flow stresses also significantly decrease (and ductility increases) with increasing temperatures. The carbon-spring steel also exhibits a significant strength drop, but this reduction of flow stresses only occurs between 600 °C and 800 °C. Comparing the two materials, obviously the necessary thermal softening for the initiation of a shear band is only reached at higher temperatures for the C75S materials. This results in a need for higher amounts of energy in the adiabatic cutting process to achieve similar cutting surface qualities in both materials.
Figure 8. Microstructure of the 20MnB5 steel, investigated (after the experiments at increased temperatures in the SHPB) by optical microscopy. a) Predominantly martensitic microstructure after testing at 400 °C. b) Smaller martensite plates are observed after testing at 600 °C, and in comparison to the sample deformed at 400 °C, the amount of ferrite (brighter areas) is increased. This indicates relaxation of the martensite due to carbon diffusion, which results in a lower flow stress of the material at 600 °C.

Figure 9. Flow stress-true strain behavior of C75S from RT up to 1000 °C and at a strain rate of 1000 s\(^{-1}\). Strength significantly decreases and ductility increases with increasing temperatures up to 1000 °C. A strength drop occurs between 600 °C and 800 °C.

Figure 10 shows optical micrographs taken from samples that were deformed under dynamic loading conditions at 600 °C and 800 °C, respectively. The sample deformed at 600 °C (figure 10a) is characterized by a bainitic and martensitic microstructure, which is similar to the microstructure observed at RT. After dynamic testing at 800 °C, we instead observe a small amount of bainite, and no more martensite (figure 10b). The microstructure mainly consists of austenite (brighter areas) and hard pearlite (darker areas). This microstructure results from the onset of the phase transformation from \(\alpha\)-Fe to \(\gamma\)-Fe followed by air-cooling after the experiment. The \(A_{c3}\) temperature of the carbon-spring steel
is about 730 °C, and due to the carbon amount of 0.72 %, the steel is very close to the eutectic point (0.80 % C). At the testing temperature of 800 °C, a substantial volume of the initial bainitic and martensitic microstructure is transformed to \( \gamma \)-Fe, and this in turn leads to a significantly lower strength.

**Figure 10.** Microstructures of the C75S steel, investigated by optical microscopy after the dynamic experiments in the SHPB at different temperatures. a) Bainitic and martensitic microstructure after testing at 600 °C. b) After testing at 800 °C, the microstructure primarily consists of austenite (brighter areas) and hard pearlite (darker areas), and small amounts of residual bainite.

### 3.3 Evaluation of the analytical model

Given the information of the flow curves determined by experimental analysis, the next step is to incorporate the findings into the analytical approach presented in section 2.2. The analytical model is used to estimate the temperature evolution in the shear zone as described in section 2.2. For an appropriate comparison between the analytically calculated temperatures, the flow curves from the experiments first had to be modified in the sense that the initial pseudo stress peaks were excluded from the data sets. The reasoning to do so is given in section 3.2. Furthermore, based on a similar line of reasoning, the noise occurring in the experimental data was reduced using a simple filter algorithm. Since the solution of the differential equation (1) requires information about a continuous temperature set of flow curves, but the experiments were conducted at only six discrete temperatures, the remaining data was obtained by means of linear interpolation. With the appropriately processed data at hand, the differential equation for the temperature evolution could then be solved. In figure 11a, the temperature evolution is plotted as a function of the equivalent plastic strain for both materials, 20MnB5 (yellow) and C75S (blue). The final temperature between 766 °C (20MnB5) and 822 °C (C75S), which is predicted by the analytical model, is in good agreement with the literature [13]. As can be seen, the temperature increases monotonically with increasing strain. However, for C75S, based on the experimental data, a short plateau-like behavior can be observed at strains around \( \varepsilon^p = 3 \). Due to their similar flow curves, the temperature evolution between both materials is in good agreement, because a similar amount of energy is dissipated during the deformation process. As pointed out in the analysis of the materials’ flow curves, at a specific temperature the flow curves drop significantly. This temperature is different for the two materials (strength drop for 20MnB5 between 400-600 °C and for C75S between 600-800 °C). With this information the increasing difference with increasing equivalent plastic strain (starting at about \( \varepsilon^p = 1.5 \)) between the two temperature curves based on experimental data (yellow and blue) can be explained. The aforementioned plateau can, in contrast, be explained by a phase transformation of the material occurring at around 714 °C. The energy necessary for this type of transformation is considered by increasing the heat capacity at this temperature. The absorbed energy between \( \varepsilon^p = 2.5 \) to 3 causes the temperature to remain constant.
until the material is fully transformed. Since afterwards the heat capacity returns to its actual value, the temperature gradient evolves again according to a decreasing flow stress.

![Figure 11](image-url)

**Figure 11.** Analytical description of the sheared area by the use of experimentally and numerically determined flow curves. a) Monotonic increase of temperature; a plateau-like behavior can be observed when phase transformation takes place. b) Softening behavior due to the evolution of process temperatures.

A different approach, which is more appropriate for application purpose, is to look at the current flow stress as a function of the current, process-related equivalent plastic strain. This is illustrated in figure 11b. The seemingly softening behavior is a result not of such a material behavior itself but rather of the increasing temperature during the process, which has the effect of reducing the flow stresses of the (still hardening) material. It is therefore useful to think of the temperature evolution, as seen in the left figure, superimposed to the developing equivalent plastic strain in the right figure, for each material respectively. In order to assess the general work required for the entire cutting process, the areas below the curves are analyzed. This provides information about the deformation energy and, hence, the process forces if the distance covered by the tools during the process is known. Since the linkage between the left and right figure is given by the flow curves, see for instance figure 9 for C75S, the curvatures can be explained as follows. At the point, where the absolute value of the gradient increases significantly (for instance: C75S, blue curve, $\varepsilon^P = 1.8$) the temperature is known from the left figure (600 °C). Compared with the flow curves, it relates to the experimental data point (temperature) at which the distinctive drop in flow stress due to the phase transformation occurs (flow curves: 600 to 800 °C). However, it should be noted that the linear interpolation between 600 °C ($\varepsilon^P = 1.8$) and 800 °C ($\varepsilon^P = 3.7$) causes the gradient to decrease prematurely, i.e. the temperature band in which the phase transformation and thus the significant flow stress drop occurs, is overestimated. Further investigation is required to determine more accurate flow curves in this particular temperature band of interest (C75S: 600-800 °C).
4. Summary and conclusions
The mechanical behavior of two high strength steels (20MnB5 and C75S) was characterized in dependence of temperature and strain rate. In addition to conventional quasi-static compression tests at room temperature, compression tests in a Split Hopkinson Pressure Bar with an initial strain rate of 1000 s⁻¹ were performed in a temperature range between 200 °C and 1000 °C. The experimental results show that high strength steels with nearly the same mechanical properties at RT may possess a considerably different behavior at higher temperatures, which can lead to different cutting results in the adiabatic cutting process. The manganese-boron steel exhibits a significant strength drop between 400 °C and 600 °C. In contrast, the strength of the C75S steel decreases between 600 °C and 800 °C. This means that the necessary thermal softening for the initiation of a shear band over the whole cross section of a sheet will only start at considerably higher temperatures. This in turn results in a higher need of energy in the adiabatic cutting process for cutting the carbon-spring steel with a comparable cutting surface quality to the 20MnB5 steel. The strength drops observed in both materials can be directly related to microstructural changes in the corresponding temperature ranges. The analytical model presented here shows good agreements with the temperature increase during deformation given in the literature [13]. In order to do obtain a more detailed understanding of the thermo-mechanical behavior of both materials, further microstructural analyses may provide additional information on temperature gradients and grain sizes. Moreover, a more detailed data set (i.e., flow curves) particularly in the temperature range between 600 and 800°C is needed to identify the accurate temperature and amplitude where the strength drop occurs. This will also lead to an enhancement of the results of the analytical model and thus also improve the process simulation, which is necessary for the process design itself.

Acknowledgements
The authors would like to thank the “Arbeitsgemeinschaft industrieller Forschungsvereinigungen (AiF)”, the Federal Ministry for Economic Affairs and Energy (BMWi) and the “Forschungsvereinigung Stahlanwendung e. V. (FOSTA)” for support of the research project “Material science-based development of simulation strategies for the application of adiabatic cutting in sheet metal part manufacturing”, P 1127/28/2015 / IGF 18865 BG.

References
[1] Neugebauer R, Bouzakis K D, Denkena B, Klocke F, Sterzing A, Tekkaya A E and Wertheim R 2011 CIRP Annals.-Manu. Tech. 60 (2) 627
[2] Lazzarotto L and Michon R 2008 Tagungsband Chemnitzer Karosseriekolloquium CBC 5 242
[3] Meyers M A and Wittmann C L 1990 Metall. Trans. 21 3153
[4] Bai Y and Dodd B 1992 Theories and Applications (Oxford: Pergamon Press)
[5] Odeshi A G and Bassim M N 2008 Mater. Sci. Eng. A 488 235
[6] Kolsky H 1949 Proc. Phys. Soc. B 62 676
[7] Zener C and Hollomon J H 1944 J. Appl. Phys. 15 22
[8] Taylor G I and Quinney H 1934 Proc. Royal Soc. London 143 307
[9] Onaka S 2012 Mater. Trans. 53 1547
[10] Meyers M A 1994 Dynamic Behavior of Materials (New York: Wiley)
[11] Knezar K, Manzenreiter T, Faderl J and Radlmayr K M 2007 Tagungsband Erlanger Workshop Warmblechumformung 2 93
[12] Laumann T and Pfestorf T 2007 Tagungsband Erlanger Workshop Warmblechumformung 2 149
[13] Lins J F C 2007 Mater. Sci. Eng. A 457 205