Formability enhancement of EN AW-5182 H18 aluminum alloy sheet metal parts in a flash forming process: testing, calibration and evaluation of fracture models

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Abstract. Currently, it is state of the art to use precipitation hardening 6000-series aluminum alloys to manufacture high-strength aluminum automotive parts by extrusion or in a cold forming process. Alternatively, it is also possible to produce such parts by the use of non-precipitation hardening 5000-series aluminum alloys in a work-hardened condition. Therefore, BENTELER Automobiltechnik GmbH developed a special sheet forming process, henceforth referred to as “flash forming process”. The application of the flash forming process, consisting of a rapid heat treatment and a subsequent cold die stamping, increases the forming capability of the work-hardened 5000-series aluminum sheets and results in high-strength parts with a very good ductility and weldability. In addition, this thermal assisted forming process allows a cost-saving production of such high-strength aluminum parts due to lower material costs of 5000-series aluminum alloys than those of a 6000-series material. Furthermore, the weight-saving effects of “flash formed” parts can be higher compared to extruded or cold formed 6000-series aluminum alloys. The suitability of the process is evaluated by forming a commercial AW-5182 H18 aluminum sheet to a crash-relevant automotive part. However, to accurately simulate the flash forming process itself, a temperature dependent fracture model is necessary. Investigations on a coupon basis also showed that the effect of adiabatic heating due to plastic work cannot be neglected. In cooperation with Paderborn University, a detailed mechanical testing, aided by digital image correlation (DIC) and thermal imaging, is carried out to characterize the yield, hardening and fracture behavior at elevated temperatures. The experimental tests are followed by the calibration of a FLD and an incremental stress state dependent fracture model in LS-DYNA. Finally, the simulation models are validated on a cross die deep drawn cup.

1. Introduction

Owing to a worldwide tightening of emission and crash safety regulations, lightweight and crashworthiness of car body structures is more important than ever before. However, the substitution of conventional steel grades by lighter materials, such as aluminum alloys, is challenging due to inferior formability and cost drawbacks. The limited formability of aluminum alloys can be enhanced through heat treatment or warm sheet forming methods. The solution heat treatment of e.g. 6000-series aluminum alloys increases the critical drawing depth, but needs a subsequent aging for strengthening of the formed part. An isothermal sheet forming process at high temperatures results in a superplastic
material behavior and allows the forming of complex part shapes. However, this process requires a fine grained microstructure and very low strain rates which have a negative impact on the production cycle and costs [1]. As shown in [2] and [3], direct warm forming below the recrystallization temperature also improves the formability of work-hardening aluminum alloys. It could be also proven that temperature gradients within the formed blank allow to enhance the formability even more. Nevertheless, this kind of process requires a sophisticated tooling design with partially heated and cooled tools. Moreover, the warm forming of non-heat treatable aluminum alloys leads to loss in strength which cannot be recovered as in the case of heat treatable alloys. As a consequence of that drawback a heat-assisted multi-stage forming process consisting of cold pre-forming, oven heat treatment and final cold forming was introduced in [4]. The loss in strength during the heat treatment is recovered through work hardening in the final forming stage. However, the strength increases considerably only in areas that have undergone a significant work hardening in the last forming step. In the result, the overall strength to weight ratio for those kind of parts is limited. On that account, to further advance that process, the design of tools with integrated local heat treatment areas was proposed in [5]. A tailored heat treatment layout of the pre-formed part allows to recover the material at targeted points which will undergo a further forming in the final process stage. Yet, the process proposed in [5] as well as the further development shown in [6] needs two forming stages and a special tooling. To enable an affordable and flexible part manufacturing, a new heat assisted forming approach for 5000-series aluminum alloys was introduced in [7]. The so-called “flash forming” process allows to overcome the drawbacks of limited formability, low cycle time and low global strength by a rapid heating and subsequent cold die stamping of AlMg alloys in a work-hardened condition. This novel approach not only reduces the loss in strength but also increases the formability at the same time. To facilitate the part and tooling design, as well as to understand the complexity of this non-isothermal process, an accurate plasticity and fracture model is necessary. Therefore, two numerical approaches are investigated to describe the temperature dependent yield loci and fracture: a FLC based and an incremental stress state-dependent method.

2. Flash forming process

The flash forming process (FFP) can apply for all non-precipitation hardening AlMg(Mn) alloys [7] and will be discussed considering the example of an EN AW-5182 aluminum alloy. In a departure from the “hot blank – cold die (HB-CD)” process proposed in [8], the FFP does not start with an aluminum blank in an O temper but in a strain hardened condition, in this case a H18 temper. It is vital that the pre-strained material exhibits a higher dislocation density in the crystal lattice, what results in an increased yield strength and reduced formability of the initial material. However, to manufacture complex and high-strength aluminum parts in a single cold forming step without the necessity of recovery annealing, nor a second forming step, a new approach is necessary.

Experiments have shown that a rapid heat treatment of a strain hardened 5000-series aluminum alloy up to temperatures which are typical for recovery annealing increases significantly the strain to fracture and reduces the yield stress. At the same time the material strength of the final part is similar to the as-delivered Hxx temper (see section 3). As reported in [5], an induction heating of pre-strained AA5182 specimens within 3 s up to 300 °C combined with a subsequent ambient air cooling leads to a recovery of approx. 80 %. Hence, to suppress a complete recovery of the microstructure and an associated loss in strength the material shall be cooled down in a faster manner. Yet, there is only a small process window within the heat treatment allows to induce a greater formability and does not cause a significant reduction of the accumulated dislocations in the strain hardened material. As shown in Figure 1, the FFP starts with a strain hardened AlMg sheet. In the first process step, the blank is rapidly heated up to a temperature between 270 and 350 °C [7]. The necessary heating rates can be achieved by induction heating [5] or in a direct way by heated contact plates [9]. To ensure a coincident advantage of an increased formability and a slight loss in strength of the final part, the forming step should be carried out in a cold tool directly after blank heating. As pointed out in [8], heating only the blank implies lower energy needs compared to conventional warm/hot forming. Moreover, the process of flash forming is at least energy-equivalent with common cold forming processes of AlMg(Mn) alloy sheets, since the recovery annealing after rolling is no longer required.
A further advantage of that kind of process is the similarity to a press hardening process of boron steels, so current stamping infrastructure can be used and a cost-effective high-volume production of automotive parts can be obtained [8].

![Scheme of the flash forming process.](image)

**Figure 1.** Scheme of the flash forming process.

3. Experimental procedures

3.1. Tensile tests

To enable a material characterization for the FFP, a special experimental setup was developed. In order to facilitate a process of heating and subsequent tensile testing, a pair of isolated chucks is used to heat and clamp the specimen directly in the testing machine. The chucks, and so the specimen, are connected to a DC power supply. Once the electrical current flow through the specimen is turned on, the testing procedure is carried out automatically. The temperature of the specimen is monitored by a pyrometer pointing on the middle of the gauge area. As soon as the specimen reaches the target testing temperature, the heating is turned off and the tensile test starts. In addition to the temperature measurement, a 3D DIC measurement is carried out by a GOM Aramis 4M system. The recorded images are synchronized with test data from the testing machine and the pyrometer, force and temperature, respectively. The FFP approach is evaluated by testing a commercial non-heat treatable EN AW-5182 H18 (AlMg4.5Mn) aluminum sheet with a nominal thickness of 2.0 mm. The sheets were machined along the rolling direction into tensile specimens according to the geometry presented in Figure 2. The tensile tests were carried out at 25 °C (RT), 175 °C, 250 °C, 275 °C and 300 °C, since, as reported in [5], even a slight amount of pre-straining leads to a coarse grain formation at temperatures above 350 °C. To correspond to the target FFP forming speed, all tests are conducted at a nominal strain rate of 0.20 s\(^{-1}\). Furthermore, additional tests at RT with a strain rate of 0.0014 s\(^{-1}\) are carried out to determine the strain rate sensitivity of the tested material.

![Tensile test specimen geometry in accordance to EN ISO 6892-1.](image)

**Figure 2.** Tensile test specimen geometry in accordance to EN ISO 6892-1.

The true stress-strain results evaluated by a 50 mm digital (DIC) axial extensometer are shown in Figure 3. The flow curves at RT do not show any significant dependence between strain rate and hardening behavior, though a flow stress reduction at a strain rate of 0.20 s\(^{-1}\) can be observed. The negative strain rate sensitivity of the tested material and the serrated flow curve at lower strain rates can be explained by the phenomenon of the Portevin-Le Chatelier (PLC) effect which is a result of an
interaction between solute atoms (Mg) and dislocations [1]. As presented in the test data of Figure 3, the mechanical properties of the tested material show a significant dependence on the testing temperature. While the flow stress decreases with test temperature, the total elongation only increases at temperatures above 175 °C. A similar behavior of a reduced elongation to failure at intermediate temperatures and increased strain rate was also observed in test data presented in [8].

![Figure 3](image)

**Figure 3.** True stress-strain curves of EN AW-5182 H18 at several elevated temperatures and at RT for two different strain rates.

The reason is not unambiguous clarified, but could be also related to the PLC effect. Further, this behavior can be observed within a temperature range where a switch between the dominant deformation mechanisms appears. At ambient temperatures the material deformation is driven by dislocation slipping and changes to diffusion processes and dislocation climbing at elevated temperatures [1]. This leads to an altered strain hardening behavior which affects the ductility of the material and so, as presented in Figure 4, also the fracture mode. The distinct softening after reaching the ultimate tensile strength suggests, that a further dynamic recovery occurs during the deformation and the temperature increase induced by plastic work cannot be neglected. Consequently, the flow curves do not correspond to isothermal conditions and a correction is necessary. This observation could be confirmed by thermal imaging tests conducted in [10], [11] and [12]. Based on the observations done in previous research, the experimental tensile test setup was enhanced by a Flir SC7650 thermal imaging camera and additional tests with a coupled stress-strain-temperature measurement were carried out to access a strain-temperature field within the deformation zone.

![Figure 4](image)

**Figure 4.** Tensile test fracture surfaces for RT, 175 °C, 250 °C, 275 °C and 300 °C. A transition from ductile shear through slant to cup-and-cone / cup-and-cup fracture can be observed.

3.2. Flash forming of an automotive part

The positive elevated temperature response, combined with rapid heating and die quenching promise a suitable forming process for high strength structural EN AW-5182 aluminum alloy parts. In order to validate the flash forming principle presented in section 1 a door beam was stamped from a 3.5 mm thick EN AW-5182 H18 sheet at three different temperatures: RT, < 350 °C and > 350 °C. Beyond the formability it was to determine if the as-delivered material properties could be recovered after the
Stamping process. For that purpose three subsize tensile specimens were cut out from an area of the formed part which undergoes almost no work hardening during the forming process (Figure 5).

**Figure 5.** A successfully flash formed AA5182 door beam (parameter set T2). Specimens were cut from a slightly formed area (marked in green).

While in a stamping trial at RT, several cracks were detected in the formed part, both flash formed components were sound. However, a critical area of deepest drawing at which failure occurs at RT was investigated for necking in both components. The thinning amounts to 15 % and 13 % for < 350 °C and > 350 °C respectively.

**Figure 6.** Door impact beam formed under different conditions: T1 (fracture) and T2 (sound).

Mean results of the tensile tests for different process parameters are given in Table 1. As presumed, there is a slight temperature dependent drop in yield stress (-18.3 %) and ultimate material strength (-9.8 %) accompanied by a simultaneous increase in ductility (-3%) for initial sheet temperatures lower than 350 °C. However, the overall material properties of the parameter set T2 are comparable to those of precipitation hardening aluminum alloys. Heating the blank to more than 350 °C (parameter set T3) leads in fact to an improved formability, but also to a loss in YS (-26 %) and UTS (-15.6 %) which cannot be completely compensated for by work hardening during the forming process.

| Parameter set | Heat treatment prior forming | Necking in critical area | Mean material properties after forming |
|---------------|------------------------------|--------------------------|---------------------------------------|
|               |                              |                          | YS [MPa] | UTS [MPa] | A30 [%] |
| T1            | RT                           | fracture                 | 349      | 398       | 9.0     |
| T2            | < 350 °C                     | 15 %                     | 285      | 359       | 12.0    |
| T3            | > 350 °C                     | 13 %                     | 257      | 336       | 14.0    |

4. Material modeling

4.1. Forming Limit Curve based approach
In a first attempt to predict failure in the flash forming process a FLC-based approach is used. The plastic flow behavior is modeled by a von Mises yield criterion with a stress vs. strain vs. temperature surface to account for thermal effects (*MAT_106 in LS-DYNA) [13]. The flow surface is given by temperature dependent flow curves that are based on a fit of the experimental data from section 3.1 to Gosh’s equation (1) for each temperature.
\[ \sigma = a(b + \varepsilon)^n - c \]  

(1)

Since an isothermal characterization of the forming limits would lead to a recovered crystal lattice within the initially strain-hardened alloy sheet, the temperature dependent Forming Limit Curves were determined in a process analogous to the FFP.

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\[ \sigma_y = k_t(\varepsilon_p, \dot{\varepsilon}_p) \ast \frac{kt(\varepsilon_p, T)}{kt(\varepsilon_p, T_R)} \]  

(2)

\[ \varepsilon_{pf} = f(\eta, \Theta) \ast g(\dot{\varepsilon}_p^p) \ast h(T) \ast i(l_{el}) \]  

(3)

\[ \eta = \frac{-I_1}{\sqrt{3}J_2} \]  

(4)

In contrast to former research, this contribution investigates the suitability of the isotropic Tabulated Johnson-Cook material model for the simulation of non-isothermal forming processes with shell elements. The theory of *MAT_224 assumes that the fracture characteristics for all temperatures are basically the same. The failure prediction is based on one fracture curve determined at RT and an additional scaling function for elevated temperatures, which describes the ratio between the fracture strains at uniaxial tension for different temperatures. As proposed in [16] the basic fracture curve as a function of failure strain and triaxiality is determined by an inverse fit of experimental force-displacement curves for four different specimens in a triaxiality range between 0 and -2/3, compare...
triaxiality definition from equation (4). The experimental results of the uniaxial tensile test specimen (UT01) from Figure 2 are broadened by three additional specimen geometries: a notched tensile specimen (NTR5), a shear 0° tensile specimen (GST00), and an equibiaxial Nakajima test (NAK). Due to a lack of shell elements in the post-necking stress state description and non-linear strain paths of the specimens prior failure, the fracture curve has to be determined by a trial-and-error strategy starting with failure strain values obtained from DIC measurements. The comparison between experimental test data and simulations with the Tabulated Johnson-Cook model are presented in Figure 8.

![Figure 8](image)

**Figure 8.** Comparison between experiment and *MAT_224 simulation for AA 5182 H18 at RT.

The calibration specimens show, except for the shear tension case, good agreement with experimental data. In the case of the shear tension specimen, the failure is initiated by a triaxiality of $\eta = -1/3$ in the outermost elements of the notch which leads to a premature onset of fracture. The sheet forming relevant triaxiality range between $\eta = -1/3$ and $\eta = -2/3$ is validated by forming a cross-die cup. Here, *MAT_224 strongly underestimates (-63%) the critical drawing depth, but the onset of failure is predicted at the same location as the experiment.

$$D = \int \frac{d\varepsilon_p}{\varepsilon_p(\eta)}$$

(5)

This underestimation, as well as the wide variance for the shear specimen, can be traced back on the linear damage accumulation of *MAT_224 that is adopted from the classical Johnson-Cook formulation, see equation (5). The damage accumulation starts once the yield stress is reached and is even more pronounced for shell elements than for solids.

5. Conclusions

In the present contribution a novel non-isothermal forming approach for work-hardened AlMg(Mn) alloys was introduced. The material behavior was characterized using tensile tests at several temperatures from 25 °C up to 300 °C. The formability as well as the ductile fracture showed a strong dependency on the temperature. A transition from ductile shear and slant fracture at lower temperatures to cup-and-cone fracture at elevated temperatures was observed. Stamping tests of a door beam and subsequent material testing proved the applicability of the process for high-strength crash relevant parts with mechanical characteristics comparable of those of the as-delivered work-hardened blanks. For forming simulation and failure prediction two different approaches were investigated. A
FLC-based method was not able to predict the failure of the non-isothermal process sufficiently. The correlation of the Tabulated Johnson-Cook model with calibration specimens was fairly good, but the critical drawing depth of a cross-die cup was strongly underestimated by the model. That can be explained by the linear damage accumulation adopted from the initial Johnson-Cook model. For that reason an enrichment of the current model by a non-linear damage accumulation will be investigated in future. Furthermore, it is recognized that to achieve the necessary predictability of ductility and fracture of the examined metastable material condition, new constitutive equations for a correlation between heat treatment and crystal lattice recovery have to be developed in future work. More data are in need to better define this relationship what as a consequence demands a deliberate experimental program.

References
[1] Ostermann F 2014 *Anwendungstechnologie Aluminium* (Berlin Heidelberg: Springer Vieweg)
[2] Abe Y and Yoshida M 1994 Warm forming of 5182 aluminum alloy sheets into double square sinks *J. Japan Institute of Light Metals* 44 240-45
[3] Schmoeckel D, Liebler B C and Speck F D 1994 Deep drawing of aluminium in partially heated tools *Prod. Eng.* 12 55-8
[4] Siefert K, Merklein M, Nester W and Grünbaum M 2010 *Enhacement of Forming Limits of Aluminum Alloys Using an Intermediate Heat Treatment* AIP Conf. Proc. 1315 pp 359-64
[5] Siefert K, Sulzberger A and Merklein M 2011 *Investigation on Induction Heat Treatment to Enhance the Formability of Aluminum Alloy AA5182* Proc. 10th Int. Conf. Techn. Plast. (ICTP2011) pp 414-19
[6] Sulzberger A 2013 *Serienmaße Auslegung der Prozesskette zur wärmeunterstützten Umformung von Aluminiumblechwerkstoffen* Dissertation, Friedrich-Alexander-Universität Erlangen-Nürnberg
[7] Dörr J, Garcia R G and Pellmann M 2010 *Method for producing a molded sheet metal part from an as-rolled, non-hardenable aluminum alloy* Filed: 09.02.2010. US, Patent Application Publication, Pub. No.: US 2010/0218860 A1
[8] Zhang N, Abu-Farha F 2015 Characterizing and Modeling the Deformation of AA5182 for Hot Blank – Cold Die (HB-CD) Stamping *Light Metals* ed Hyland M pp 315-20
[9] Hogg M 2006 *Herstellung und Umformung lokal wärmebehandelter Platinen* Dissertation, Universität Stuttgart
[10] Charpentier P L, Stone B C, Ernst S C and Thomas Jr J F 1986 Characterization and modeling of the temperature flow behavior of aluminum alloy 2024 *Metallurgical Transactions A* 17A 2227-37
[11] Hodowany J, Ravichandran G, Rosakis A J and Rosakis P 2000 Partition of plastic work into heat and stored energy in metals *Experimental Mechanics* 40 2 113-23
[12] Martinez J A R 2010 *Advanced constitutive relations for modeling thermo-viscoplastic behavior of metallic alloys subjected to impact loading* Dissertation, University Carlos III of Madrid
[13] LSTC 2015 *LS-DYNA Keyword User’s Manual* (Livemore: Livemore Software Technology Corporation, LSTC)
[14] Buyuk M 2013 *Development of a Tabulated Thermo-Viscoplastic Material Model with Regularized Failure for Dynamic Ductile Failure Prediction of Structures under Impact Loading* Dissertation, George Washington University
[15] Haight S, Du Bois P and Kan C-D S 2016 *A comparison of isotropic (*MAT_224*) and anisotropic (*MAT_264*) material models in high velocity ballistic impact simulations* Proceedings of 14th International LS-DYNA Users Conference pp 1-12
[16] Andrade F X C, Feucht M, Haufe A and Neukamm F 2016 An incremental stress state dependent damage model for ductile failure prediction *Int. J. Fracture* 200 DOI: 10.1007/s10704-016-0081-2