Predicting plasticity and fracture of severe pre-strained
EN AW-5182 by Yld2000 yield locus and Hosford-Coulomb
fracture model in sheet forming applications

A A Camberg¹, T Tröster¹, F Bohner² and J Tölle²

¹ Automotive Lightweight Design, Paderborn University, Pohlweg 47-49, 33098 Paderborn, Germany
² BENTELER Automobiltechnik GmbH, Research and Development, An der Talle 27-31, 33102 Paderborn, Germany

E-mail: alan.camberg@uni-paderborn.de

Abstract. The objective of this contribution is to evaluate the capabilities of the Yld2000 yield locus and the Hosford-Coulomb fracture model under isothermal conditions for an EN AW-5182 H18 material in sheet metal forming applications. The calibration of the Yld2000 model is based on uniaxial tension tests in 0°, 45° and 90° in respect to rolling direction and additional layer compression tests. Subsequently, the calibrated plasticity model is evaluated under shear stress and plane strain dominant conditions. Stress state dependent fracture strains are obtained from various tensile tests as well as from FLD tests. The implemented failure prediction is based on a generalized incremental stress-state dependent fracture model (GISSMO) combined with a three-parameter Hosford-Coulomb fracture and instability curve. Finally, the plasticity and fracture models are validated at a cross die cup. It is shown that the models are capable to provide an accurate prediction of the onset of fracture and could be used for a non-isothermal extension in future work.

1. Introduction
Heat treatable 6000 series aluminum alloys are particularly suitable for crash-relevant and lightweight automotive parts, since they are usually formed in a ductile condition and later age hardened during paint bake cycle [1]. Besides precipitation hardening, severe plastic deformation (SPD) represents an alternative processing route to increase the strength of aluminum sheet materials [2]. Thus, highly work-hardened non-heat treatable 5000-series aluminum alloys could provide an alternative and cost-efficient way to produce high-strength and lightweight structural components. However, pre-strain not only leads to an increased yield strength but also results in a reduced formability. Hitherto, the reduction in formability associated with the high pre-strain makes it virtually impossible to manufacture sound components from strain hardened aluminum sheets.

The flash forming process (FFP) developed by BENTELER is a heat-assisted approach for increasing the formability of severe work-hardened sheets without simultaneous loss of strength due to recovery effects. The process begins with a multi-stage sheet rolling, followed by rapid blank heating with subsequent forming and die quenching [3]. However, to accurately simulate this non-isothermal forming process, a temperature dependent plasticity and fracture model is required. The aim of this contribution is to evaluate the capabilities of the Yld2000 yield function [4], the Hosford-Coulomb fracture initiation model [5] and the GISSMO framework [6] at room temperature as a basis for a
temperature dependent extension of these models for the purpose of FFP and other hot stamping processes.

![Figure 1. Scheme of the flash forming process [3].](image)

2. Experimental procedures

2.1. Material
The studied material is a cold rolled EN AW-5182 H18 (AlMg4.5Mn) aluminum alloy with a nominal sheet thickness of 2.00 mm. The chemical composition of the investigated material is shown in table 1.

|   | Al  | Mg  | Mn  | Si  | Cu  | Fe  | Cr  | Zn  | Ti  |
|---|-----|-----|-----|-----|-----|-----|-----|-----|-----|
| Bal.| 4.35| 0.47| 0.05| 0.06| 0.21| 0.04| 0.02| 0.03|     |

The microstructural characterizations of the investigated material are presented in figure 2. Bands of strongly oriented grains, which result from roll-induced deformation, are already visible by optical microscopy. Further investigations on the microstructure are carried out by EBSD maps, where a misorientation cut off of 5° is selected and high angle boundaries are assumed to have misorientations greater than 15°. The EBSD data show a strongly heterogeneous structure with a majority of coarse and severe elongated bands with a small fraction of refined grain layers. The amount of subgrain boundaries increases with decreasing band thickness, suggesting that adjacent refined areas are formed by grain subdivision. This observation is consistent with that reported in previous research on accumulative roll bonding of aluminum sheets [7].

![Figure 2. Microstructure of EN AW-5182 H18 at 0° in respect to rolling direction, from left to right: Optical microscope image, EBSD image with high angle boundaries (blue) and low angle boundaries (red), EBSD image with IPF.](image)

2.2. Uniaxial tensile experiments
The stress-strain response of the material is determined by uniaxial tension experiments at a moderate strain rate of 0.20 s⁻¹. The sheets are machined along 0°, 45° and 90° with respect to the rolling direction (RD) to tensile specimens according to the geometry shown in [3]. The tests are performed at 25 °C on a hydraulic driven Zwick/Roell Amsler HB 250 universal tensile testing machine by accessing the full strain field at the specimen’s surface by a GOM Aramis 3D DIC system.

The uniaxial tensile properties for the severely rolled 5000-series aluminum sheet are plotted in figure 3. The engineering stress-strain curves for 0°, 45° and 90° with respect to RD show that the
phenomenon of the Portevin-Le Chatelier (PLC) effect can be already greatly suppressed at the tested strain rate. Unlike expected, the effect of rolling induced anisotropy on the flow stress is relatively low. However, the loading direction shows a crucial impact on the maximum strain to fracture. It can be also seen from figure 3, that the evolution of the anisotropy coefficient with respect to the equivalent plastic strain does not converge to a particular value if the loading axis does not coincide with RD. The common approach of linear regression of the $r$-value does not seem satisfactory to describe the anisotropy for 45° and 90° degree in respect to RD. The mechanical properties of the investigated material are shown in table 2.

![Figure 3. Mechanical properties of EN AW-5182 H18 in respect to the rolling direction.](image)

### Table 2. As-received mechanical properties of EN AW-5182 H18, $t = 2.00$ mm

| ° to RD | E [MPa] | YS [MPa] | UTS [MPa] | $A_0$ [-] | $A_{50}$ [-] | $r$-value [-] (0.01-0.04) |
|---------|---------|----------|-----------|-----------|---------|-------------------------|
| 0       | 68017   | 336.64   | 385.42    | 0.061     | 0.066   | 0.471                   |
| 45      | 66108   | 324.91   | 379.65    | 0.066     | 0.073   | 1.007                   |
| 90      | 67136   | 329.01   | 387.16    | 0.066     | 0.080   | 1.042                   |

2.3. Layer compression experiments
As proposed in [4] and [8], the anisotropy coefficient at equi-biaxial loading $r_b$ and additional stress-strain data beyond UTS are determined from layer compression tests. Here, five discs with a thickness of 2.00 mm and a diameter of 13.77 mm are layered and compressed as a stack. The experiments are carried out at a cross-head speed of 0.125 mm/s and three different loads. To reduce the effect of friction a PTFE film (24 µm) and oil is applied as lubricant between the specimen faces and the machine fixture.

The hardening curve is computed based on results from uniaxial tension experiments and discrete stress-strain points determined by evaluating the height of the stacks after compression. The identified Hockett-Sherby fit of the hardening curve, as well as the experimental data up to an equivalent plastic strain of 0.29, are shown in figure 4. The Hockett-Sherby equation and the corresponding parameters are given in equation (1) and table 3, respectively.

Following the recommendations of Barlat et. al [4], the equi-biaxial $r$-value is obtained by a measurement of the final diagonals of each disc. The slope of the linear fit describing the relationship between the longitudinal $\varepsilon_{xx}$ (0°) and transversal strains $\varepsilon_{yy}$ (90°) gives the $r_b$-value. The value of $r_b$ obtained from layer compression tests is found to be 0.554.

$$k[\bar{\varepsilon}_p] = k_{f,s} - (k_{f,s} - k_{f,0}) * e^{-m\bar{\varepsilon}_p}$$  \hspace{1cm} (1)

### Table 3. Hardening parameters of the investigated EN AW-5182 H18

| $k_{f,0}$ [MPa] | $k_{f,s}$ [MPa] | $m$ | $p$ |
|-----------------|-----------------|-----|-----|
2.4. Forming Limit Diagram
To determine the Forming Limit Diagram (FLD), quasi-static Nakazima experiments are performed with a punch diameter of 100 mm according to DIN EN ISO 12004-2:2008. Five specimen geometries with recesses of different radii (min. strip widths of 20, 45, 60, 80, 100 mm) and a full circle 205 mm diameter specimen are tested and evaluated by the Cross Section Method (CSM) described in the standard. The strain fields at the specimen’s surface are accessed by a GOM Aramis 3D DIC system. To minimize the influence of friction PTFE foils and an oil lubricant are used. The results of the Nakazima experiments are shown in figure 5 for both, the principal strain space as well as the mixed stress-strain space. The lowest FLD value is found for Spec. 4 at $\varepsilon_1 = 0.08$, $\varepsilon_2 = 0.03$ and $\eta = 0.64$, $\bar{\varepsilon} = 0.12$, respectively.

2.5. Ductile fracture experiments
In addition to the uniaxial tension tests (UT01) from section 2.2 and the full circle specimen (NAK) experiments from section 2.4, two complementary specimen geometries are tested to characterize the fracture response at plane strain and simple shear conditions, respectively. The plane strain experiments are performed with a notched tensile specimen with a notch radius of 5 mm (NT05). The geometry used for simple shear experiments (GS00) is reported elsewhere [6]. All experiments are carried out at a
universal tensile testing machine. Again, the surface strain fields are measured by a GOM Aramis DIC system with a frame rate of ≥ 5 Hz and a strain gauge length of $l_0 \approx 0.6$ mm. Mean surface fracture strains from the last frame before fracture of the investigated geometries are given in table 4.

When comparing the values for plane strain conditions from figure 5 and table 4, it is noticeable that the investigated equivalent strain at the onset of fracture from notched tensile tests is below the equivalent strain at the onset of necking from FLD experiments. The discrepancy in fracture resistance between bending and in-plane membrane loadings is explained by different through-thickness stresses and was extensively studied by Woelkel et al. [9]. Here, this effect is even more pronounced since the investigated material shows a strongly heterogeneous through-thickness microstructure. Therefore, the fracture strain at $\eta \approx 0.58$ is additionally determined from FLD experiments. The mean fracture strain from DIC for Spec. 2 is equal to $\bar{\varepsilon}_f = 0.20$. This is 74% more than the notched tensile specimen NT05. As shown by Gorji [10], this value can be even higher by taking into account thinning at fracture.

3. Plasticity and fracture modeling

3.1. Yld2000-2d yield criterion

The plastic flow behavior is modeled by a non-quadratic Yld2000-2d yield function as it has proven its good accuracy in describing the anisotropy of aluminum alloy sheets. [4]. In a first attempt, all relevant experimental data from section 2 are used to describe the shape of the yield surface. From layer compression tests, the flow stress at equibiaxial tension is identified to be equal with the flow stress at $0^\circ$ with respect to RD. As proposed in [4], an exponent of $a = 8$ is used. Following the recommendations from [11], flow stresses at equal amount of plastic work are used to calculate the Yld2000 parameters.

By evaluating the flow function at different stress states / specimen geometries, an underestimation of the flow stress at plane strain (NT05) was noticed. The best fit of experimental data is achieved by choosing $a = 6$ which is typically recommended for bcc crystal structures. Furthermore, it was necessary to set $n_b = 1.0$ for a reasonable fit in the plane strain dominated yield locus area. However, the flow stress at simple shear is slightly overestimated with these model parameters. The yield surface, as well as the distribution of $r$-values and normalized flow stresses in respect to the rolling direction is given in figure 6. The good agreement with the test data proves the suitability of the Yld2000 model also for severely cold-rolled aluminum sheets. The final parameters of the identified Yld2000-2d model are listed in table 5.

### Table 4. Mean surface fracture strains of the investigated specimen geometries

| Specimen | GS00 | UT01 | NT05 | NAK |
|----------|------|------|------|-----|
| Stress triaxiality $\eta$ | ≈ 0.00 | ≈ 0.33 | ≈ 0.58 | ≈ 0.67 |
| Mean surface fracture strain (DIC) [−] | 0.712 | 0.295 | 0.115 | 0.270 |

### Table 5. Identified Yld2000 yield function parameters

| $\alpha_0$ | $\alpha_1$ | $\alpha_2$ | $\alpha_3$ | $\alpha_4$ | $\alpha_5$ | $\alpha_6$ | $\alpha_7$ | $\alpha_8$ |
|------------|------------|------------|------------|------------|------------|------------|------------|------------|
| 6          | 0.599      | 1.321      | 1.120      | 0.987      | 1.025      | 0.838      | 1.028      | 1.077      |
3.2. Hosford-Coulomb fracture model

The micromechanically motivated Hosford-Coulomb model proposed by Mohr and Marcadet [5] had recently been enhanced by Pack and Mohr [12] to ensure applicability to shell elements. In [12], a so-called Domain-of-Shell-to-Solid-Equivalence (DSSE) is introduced to correctly apply for membrane and bending loading with shell elements. In the main, the DSSE is a Hosford-Coulomb model defined for biaxial tension (1/3 ≤ \( \eta \) ≤ 2/3) to capture the onset of instability. Generally, the idea is similar to the GISSMO framework [6], although GISSMO couples the damage parameter to the stress tensor when instability is reached, rather than deleting the integration point as in DSSE. In this contribution the Hosford-Coulomb (H-C) fracture initiation model and its modification as an instability curve (~DSSE) are implemented in the well-established and widely used GISSMO damage indicator framework to investigate its ability for stamping processes. The three-parameter H-C model given by equation (2) promises a time-efficient and direct parameter identification for GISSMO. The H-C model reads:

\[
\varepsilon^f(\eta, \bar{\theta}) = b(1 + c)^{1/a}(0.5((f_1 - f_2)^a + (f_2 - f_3)^a + (f_3 - f_1)^a))^{1/a} + c(2\eta + f_1 + f_3)^{1/n}
\]

(2)

With Lode angle parameter \( \bar{\theta} \) dependent trigonometric functions:

\[
f_1(\bar{\theta}) = \frac{2}{3}\cos[\pi/6(1 - \bar{\theta})], \quad f_2(\bar{\theta}) = \frac{2}{3}\cos[\pi/6(3 + \bar{\theta})], \quad f_3(\bar{\theta}) = -\frac{2}{3}\cos[\pi/6(1 - \bar{\theta})]
\]

(3)

And the unique correlation between \( \bar{\theta} \) and \( \eta \) for plane stress conditions:

\[
\bar{\theta} = 1 - 2\pi \cos^{-1}(-27/2 (\eta^2 - 1/3)) \quad \text{for} \quad -2/3 \leq \eta \leq 2/3
\]

(4)

The parameters of the H-C fracture model, as well as its adaptation as an instability criterion, are identified from experimental DIC data (section 2.4 and 2.5). Since the friction parameter \( c \), as reported in [12], has a weak influence on the so-called “biaxial tension valley”, \( c \) is assumed to be zero for the instability envelope. For low triaxialities (-2/3 ≤ \( \eta \) ≤ 1/3) the instability curve coincides with the fracture curve. The identified H-C parameters for instability and fracture are listed in table 6 below.

| Indicator        | \( a \) | \( b \) | \( c \) | \( n \) |
|------------------|--------|--------|--------|--------|
| Instability Curve| 1.20   | 0.29   | -      | 0.1    |
| Fracture curve   | 1.67   | 0.29   | 0.118  | 0.1    |

Both, the calibrated H-C fracture surface in the 3D space of stress triaxiality, Lode angle parameter and fracture strain, as well as the fracture and instability curve for plane stress in the mixed stress-strain space are shown in figure 7. We observe a weak influence of the stress triaxiality on the fracture strain
for biaxial tension, while the fracture strain increases exponentially for low triaxialities. The adopted DSSE approach for describing the onset of instability seems to provide an acceptable description of the experimental FLD data.

Figure 7. Left: 3D H-C fracture surface, the red line depicts the fracture limit for plane stress conditions, Right: Plane stress fracture and instability curve with experimental data.

4. Validation
In order to validate the predictability of the calibrated yield locus and fracture model, a deep drawing test with a symmetric cross-die cup geometry with a die radius of \( r = 7 \) mm is conducted. The fracture initiation is observed in the experiment at a drawing depth between 13 mm (no fracture) and 14 mm (fracture). The deep drawing simulation of the cross-die cup is performed with the explicit LS-DYNA solver. The blank is discretized with 0.5 mm fully-integrated shell elements with seven through-thickness integration points, while the tool is assumed to be a rigid body. The first onset of fracture in the simulation is predicted at 13.6 mm, what corresponds very well with the experiment. Furthermore, the location and the shape of the crack are the same as in the test. A comparison between the experiment and simulation at a drawing depth of 14 mm is shown in figure 8.

Figure 8. Comparison between experiment and simulation for AA 5182 H18 at RT with Yld2000 + Hosford-Coulomb fracture/instability and GISSMO.

5. Conclusions
With the aim to increase the yield strength, the investigated 5000-series aluminum alloy undergoes several cold-rolling passes. As shown by EBSD data, the roll-induced plastic deformation results in a strongly heterogenous and layered microstructure. The mechanical properties of the sheet material are examined by tensile tests, layer compression tests, Nakazima experiments and typical in-plane loading test for characterizing the materials ductility at different stress states. It is found that the fracture resistance of the material is strongly dependent on the loading conditions. Out-of-plane loading leads to
a fracture strain which is 74% higher than at pure membrane loading, which is explained by the inhomogeneous through-thickness microstructure and different stresses in thickness direction for both loadings. The materials flow curve is found to be described well by the Hockett-Sherby hardening law. The associated-plasticity Yld2000 yield function allows a good fit of experimental data, even when the $n_p$ value is assumed to be equal 1. Furthermore, the best adaptation to the experimental data is achieved with an exponent $a = 6$ instead of 8, which is usually proposed in the literature for FCC materials. For failure prediction the Hosford-Coulomb model is implemented within the GISSMO fracture indicator framework as a criterion for fracture and instability. The H-C model parameters are identified by FLD experiments and an additional in-plane shear test. The critical drawing depth of a cross-die cup, as well as the location and shape of the crack can be predicted very well by the calibrated models. Nevertheless, it should be pointed out, that the model will overestimate the fracture strain at pure membrane conditions, since it is calibrated with data from FLC tests. However, considering the bending-dominated application of this model, it provides reliable results. For that reason, an enhancement of the current models for non-isothermal forming conditions will be the subject of future work.

Acknowledgments
The partial financial support through the ILH project TP7 from BENTELER is gratefully acknowledged. Thanks are also due to Florian Hengsbach (LWK, Paderborn University) for carrying out the EBSD measurements.

References
[1] Hirsch J 1997 Aluminium Alloys for Automotive Application Mat. Sci. Forum. 242 33-50
[2] Azushima A, Kopp R, Korhonen A, Yang D Y, Micari F, Lahoti G D, Groche P, Yanagimoto J, Tsuji N, Rosochowski A and Yanagida A 2008 Severe plastic deformation (SPD) process for metals CIRP Annals – Manufacturing Technology 57 716-35
[3] Camberg A A, Bohner F, Tölle J, Schneidt A, Meiners S and Tröster T 2018 Formability enhancement of EN AW-5182 H18 aluminum alloy sheet metal parts in a flash forming process: testing, calibration and evaluation of fracture models IOP Conf. Series: Mater. Sci. Eng. 418 012018 doi:10.1088/1757-899X/418/1/012018
[4] Barlat F, Brem J C, Yoon J W, Chung K, Dick R E, Lege D J, Pourboghart F, Choi S H and Chu E 2003 Plane stress yield function for aluminum alloy sheets – part 1: theory Int. J. Plast. 19 1297-1319
[5] Mohr D and Marcadet S J 2015 Micromechanically-motivated phenomenological Hosford-Coulomb model for predicting ductile fracture initiation at low stress triaxials Int. J. Solids Struct. 67 40–55
[6] 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
[7] Heason C P and Prangnell P B 2002 Grain Refinement and Texture Evolution During the Deformation of Al to Ultra-high Strains by Accumulative Roll Bonding (ARB) Mat. Sci. Forum 396-402 429-34
[8] Tian H, Kang D and Lin J 2004 Determining Sheet Metal Hardening Curve by Laminated Specimen Key Eng. Mat. 274-276 793-98
[9] Woelkel PB, Londono JG, Knoerr LO, Dykeman J and Malcolm S 2018 Fundamental Differences between Fracture Behavior of Thin Sheets under Plane Strain Bending and Tension IOP Conf. Series: Mater. Sci. Eng. 418 012078 doi:10.1088/1757-899X/418/1/012078
[10] Gorji M B 2015 Instability and Fracture Models to Optimize the Metal Forming and Bending Crack Behavior of Al-Alloy Composites Dissertation, ETH Zürich
[11] Abedrabbo N, Pourboghrat F and Carsley J 2006 Forming of aluminum alloys at elevated temperatures - 2: Numerical modeling and experimental verification Int. J. Plast. 22 342-73
[12] Pack K and Mohr D 2017 Combined necking & fracture model to predict ductile failure with shell finite elements Eng. Frac. Mech. 182 32-51