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Ductile Failure in Upsetting of a Rapid-Solidification-Processed Aluminium Alloy

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Cold upset-tests have been performed on a Rapid Solidification Processed (RSP) aluminium-alloy, produced by the 'melt-spun ribbons'-process out of 70% car-scrap and 30% primary scrap. The ribbons are hot extruded, resulting in 29 mm diameter bar. Its properties regarding plastic flow and fracture are strongly anisotropic. Upset-tests are performed on this alloy to verify a concept for the prediction of ductile failure. The experimental results also demonstrate the mechanical integrity of this material.

Hill's flow-criterion for anisotropic materials is used in FEM-models of the experiments. Results of various experiments concerning failure at the equator of the free surface are compared with fracture strains in tension and tension tests in circumferential direction. Furthermore a surface-instability analysis based on an energy principle is applied. It is shown that a graph of effective strain versus triaxiality \( \varepsilon_f / \sigma_{\text{eff}} \), in addition to an instability-criterion, is a useful tool for the prediction of ductile failure in forging processes.

Therefore the amount of surface roughening in upsetting is negligible. This simplifies the visual detection of cracks at the equatorial free surface. This in contrast with upsetting of aluminium-alloys having normal grain size, where surface roughening complicates crack-detection severely.

Table 2: Tensile test results (properties of a rectangular hollow extrudate)

<table>
<thead>
<tr>
<th>Tensile property</th>
<th>T6: SHT 450 °C 25 min.</th>
<th>YS [MPa]</th>
<th>UTS [MPa]</th>
<th>Elongation [%]</th>
</tr>
</thead>
<tbody>
<tr>
<td>CWQ, 160 °C 6 hrs.</td>
<td>258</td>
<td>342</td>
<td>13.3</td>
<td></td>
</tr>
</tbody>
</table>

For the mathematical treatment of ductility, models and empirical criteria have been developed. Well known are the criteria of Cockcroft and Latham [4] and Rice and Tracy [12, 3]. Here respectively \( \sigma = \sigma (\sigma_1, \sigma_2, \sigma_3) \) and \( \sigma = \sigma (\sigma_1, \sigma_2, \sigma_3) \) should be invariant at failure. Especially the Cockcroft and Latham criterion is often applied in upsetting for the prediction of ductile failure. As is shown by Bolt [3], using torsion-, tensile- and prenotched tensile-test, all of these criteria perform moderately as compared to the ductile failure curve concept [5]. Here fracture is assumed to occur if during a mechanical working process a part of the material crosses the ductile failure curve. This is a curve in the \( (\varepsilon_f, \sigma) \) versus \( \sigma \) plane. The stress triaxiality \( \varepsilon_{\text{eff}} \) is the non-dimensional representative of the state of stress, and is defined as the ratio of the mean stress \( \sigma = (\sigma_1 + \sigma_2 + \sigma_3) / 3 \) and the effective flow stress \( \sigma_{\text{eff}} \). For practical use a failure curve can be obtained by means of some basic material test such as uniaxial and plain strain tensile-tests and a torsion-test.

A failure-curve needs to be accompanied by a local surface instability analysis in processes concerning failure at an expanding free surface. Instable flow at a free surface will result in strain concentration and, as a result, fracture will occur at a higher strain level than can be observed macroscopically. For analyses of local surface instability in the upset-process, the major and minor surface-strains are respectively negative and positive, the Hill instability criterion [6], stating \( \frac{1}{2} \sigma : \varepsilon = (\sigma_1 + \sigma_2 + \sigma_3) / 2 \sigma_{\text{eff}} \) for stable flow, is frequently used [2]. This criterion is based on the assumption of six-fold symmetry of the yield surface in the deviatoric stress-plane. When using a flow criterion incorporating anisotropy, the Hill criterion loses its validity. As an alternative an energy-based instability-criterion, as proposed by Kals and Veenstra [7], is applied.

Table 1: Chemical composition

<table>
<thead>
<tr>
<th>Sl</th>
<th>Fe</th>
<th>Cu</th>
<th>Mn</th>
<th>Mg</th>
<th>Zn</th>
<th>Rest</th>
<th>Al</th>
</tr>
</thead>
<tbody>
<tr>
<td>wt. %</td>
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<td>0.65</td>
<td>1.15</td>
<td>0.14</td>
<td>0.33</td>
<td>0.56</td>
<td>0.40</td>
</tr>
</tbody>
</table>

1.1 Ductility

Ductility, being the ability of materials to undergo plastic deformation without fracture, is an important property in industrial forming processes. Practical quantification of ductility is requested for the selection of materials for forming processes or vice versa.

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1.2 A Rapid solidification processed aluminium alloy

The RSP aluminium-alloy used for the experiments is produced out of 70% car-scrap and 30% primary scrap by melt-spinning. This results in ribbons of approximately 50-70 \( \mu \)m thick and 3-5 mm wide. Consolidation takes place by chopping, cold precompression and hot extrusion (ca. 423 °C) of the ribbons. Good quality extrudates can be produced with this process without degassing. Table 1 lists the chemical composition of the alloy, and table 2 some results from tensile tests in extrusion direction of the alloy in T6-condition.

Table 1: Chemical composition

The microstructure of the RSP-alloy shows an extremely small grain size.

<p>| | | | | | | | |</p>
<table>
<thead>
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</tr>
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The plastic anisotropy of the rod was determined by using small-sized compression-specimens, electrical discharge machined from the rod in three
orthogonal directions (fig. 1). The specimens are of the Rastegaev-type [9]. These specimens have shallow cavities on the circular faces which contain a solid lubricant (paraffin wax). When properly dimensioned, these specimens do not barrel during testing. Transverse strains are measured at discrete intervals up to an effective strain of ca. 0.35.

Fig. 1: Anisotropy-specimens: position in rod cross-section and geometry

Measured transverse strains are plotted in fig. 2, together with a best fit line. The tests are made in duplicate. As can be seen (fig. 2) the anisotropy values are high and fairly constant over the strain-range measured. The product of the three fitted anisotropy values equals unity within experimental error. This indicates that the Hill yield-criterion for anisotropic materials is an acceptable approximation of the real yield-surface.

2.2 Upsetting

Specimens for the upset-experiments were machined to 20 mm diameter and wet ground with No. 600 emery-paper. Length to diameter ratios of 0.75, 1 and 1.5 were used to obtain different strain paths. The specimens were compressed in small strain increments at slow speed (approx. 1 mm min⁻¹) between concentrically grooved dies. After each strain increment the strains at the equatorial free surface were measured with the naked eye. The small grain size of the alloy allowed detection of cracks of only 0.2 mm length. Each test condition was run in triplicate for averaging.

Fig. 2: Measured transverse strains; directions of ε_x and ε_θ indicated in parentheses; specimen axis in third direction

The stresses in Eq. (1) were calculated out of measured surface strains using Hill’s yield-criterion. For principal directions it states:

\[ H(\sigma_2-\sigma_3)^2 + G(\sigma_2-\sigma_3)^2 + F(\sigma_2-\sigma_3)^2 = 1 \]  

(1)

in which F, G and H are determined by the anisotropy-values and the effective stress*. The stresses in Eq. (1) are principal stresses in usual cylindrical coordinates.

3 FEM-modelling and local surface instability

The stresses on the equatorial free surface of the upset-specimens were

* Formulas relating stresses to surface strains are given in ref. [8]; equal notation is used here.

For the FEM-calculations also three independent shear yield stresses are necessary to describe material-flow. These were estimated to be 1.04, 0.92 and 1.04 * δ_1/3 for τ_m, τ_p and τ_n respectively.
positive levels of stress there is a large direction dependence of ductility.

failure curve can

torsion experiments were determined neglecting anisotropy.

4.1 Prediction of fracture by a ductile failure curve and local surface instability analysis

As mentioned in the introduction, the Hill local fracture limit is not applicable in case of anisotropic materials. Therefore a more generally valid energy-based analysis [7] is used. For an isotropic material this criterion gives instability limits identical to those resulting from the Hill criterion [7]. It states that, for a straight strain increment, instability can occur if the dissipation of work per unit surface increase shows a maximum value during the strain increment. In other words: the stability limit is found when \( \frac{dW}{dA} \) reaches a maximum. The criterion states that local surface instability will not occur until after

\[
\frac{dW}{dA} = \frac{2}{3} A_0 \varepsilon^3 (F + G + H)
\]

reaches a maximum along a straight strain increment (see appendix). Calculated instability points for the upset-experiments are given in section 4.1. It should be noticed that this calculation does not incorporate strain rate effects and stabilising effects of strain gradients perpendicular to the surface. The analysis should therefore be seen as a lower limit for local surface instability.

4 Results and discussion

4.1 Prediction of fracture by a ductile failure curve and local surface instability analysis

From the tests in axial and circumferential direction a linearised ductile failure curve can be constructed. For estimation of the stress-triaxiality in the centre of the neck of the tensile-test specimen, only isotropic FEM-calculations [3] have been used (incorporating anisotropy would require a three-dimensional mesh). For the same reason, also the triaxiality and the effective strain in the torsion experiments were determined neglecting anisotropy. The direction dependence of fracture behaviour is evident from table 4. Especially at high positive levels of stress there is a large direction-dependence of ductility.

Fig. 7 shows, as already mentioned in section 3.1, the measured and calculated strain paths from the upset-tests. Additionally in fig. 8 the paths to failure, in the effective strain versus triaxiality plane, of a surface element at the equator of the upset-specimens are given. This figure shows also the local surface instability points calculated from the linearised strain paths and the tensile and torsion tests in circumferential direction.

Failure occurred shortly after calculated instability for all three height to diameter ratios. If instability did precede fracture is uncertain, microscopic examination could not confirm strain concentrations in the direction of zero extension near the failure-site. The instability points are close to the linearised ductility-curve, which is determined by the tensile and torsion experiment.

4.2 Alternative fracture criteria

The Cockcroft and Latham [4] criterion is often used with reasonable accuracy for prediction of ductile fracture in upsetting. The Cockcroft and Latham integral at fracture varied from 60 \((h/d=1)\) to 80 \((h/d=1.5)\) for present upset-tests.

The general validity of the criterion however has to be doubted because it is shown that the criterion is extremely insensitive to the shape of the strain path at a free surface [10, 13].

For the Oyane void growth criterion [11] fracture is assumed to occur at a critical density change, that is if \( \int (1 + \sigma_0 \frac{\partial \sigma_0}{\partial A}) dA \) reaches the value of the torsion fracture strain (for which \( \sigma_0 = 0 \)). The fracture points should be located at straight declining line in a graph of \( \int (1 + \sigma_0 \frac{\partial \sigma_0}{\partial A}) dA \) versus \( \varepsilon \). Results for the upset-tests and the circumferential tensile and torsion tests are given in fig. 9. It is clear that results from the torsion experiment do not agree with those from upsetting.

An alternative void-growth model is proposed by Rice and Tracey [12]. Again assuming that fracture occurs at a critical density change, it states that fracture should occur if \( \int \exp(3\alpha_0/\partial A) dA \) equals the value of the torsion fracture strain [3]. A graph of the integral as a function of effective strain is given in fig. 10.
5 Conclusions

The plastic anisotropy of 29 mm diameter rod was measured in three orthogonal directions up to an effective strain of ca. 0.35. The measured anisotropy of the RSP aluminium alloy was large, so that neglecting it would cause severe errors in stress-calculations.

The limit for local surface stability of an anisotropic strain-hardening material was derived using an energy approach. Failure in the upset-experiments occurred shortly after calculated instability. However, microscopic examination could not reveal strain concentrations in the direction of zero extension.

The ductile failure curve concept, with reasonable accuracy, couples the results of failure in the upset-tests and the tension and torsion experiments in circumferential direction.

The Oyane fracture criterion shows good results if the torsion fracture point is neglected. The Rice and Tracey criterion gave unsatisfactory results in predicting fracture in present experiments, especially for failure occurring at low triaxiality levels.

For successful use of FEM in shortening tool and process development-time for industrial forming processes, a simple and accurate failure concept is needed. Present results indicate that the presented failure curve concept can meet this demand.

Acknowledgement

The authors are indebted to A.J. Kneppers, whose dedication to general technical supervision made present work possible. Appreciation is also expressed to the Dutch IOP-metals research programme for financial support and to W.H. Kool (laboratory for materials science, Delft University of Technology) and his research-group for the production of the RSP aluminium-alloy.

References


Appendix

For calculation of the local surface stability limit, $dW/dA$ is calculated from [7]:

$$dA = A d\varepsilon = -A_k \exp(-\varepsilon_k) d\varepsilon$$

and,

$$dW = \sigma d\varepsilon$$

The strain increment in radial direction and the effective strain increment are related by the Lévy-Mises equations for anisotropic materials. In a plain stress state ($\sigma_r=0$) these are [8]:

$$\frac{d\varepsilon}{3(2G+H)} = \frac{-d\varepsilon_{xx}}{2G_{yy}+2F_{xx}} = dA$$

Eq. (2) results from Eq. A1, A2 and A3.

An element at the equator of an upset-specimen first decreases and later increases its surface. At the moment of zero surface increase the quantity $dW/dA$ is infinite. This value will drop rapidly as the surface further increases and a maximum value cannot be found. Therefore the strain path has to be linearised into a sufficiently large number of straight strain increments. These are to be checked for a maximum of $dW/dA$ within the straight strain increment. In doing so, one is calculating the ability of the material to change its strain path locally from the prescribed straight strain path.