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On deformation and damage micromechanisms in strong work hardening 2198 T3 aluminium alloy

Ante Buljac [a, b], François Hild [a], Lukas Helfen [c, d], Thilo F. Morgeneyer [b, *]

[a] Laboratoire de Mécanique et Technologie (LMT), ENS Paris-Saclay, CNRS, Université Paris-Saclay, 61 avenue du Président Wilson, 94235 Cachan, France
[b] MINES ParisTech, PSL Research University, Centre des Matériaux, CNRS UMR 7633, BP 87, 91003 Evry, France
[c] Institute for Photon Science and Synchrotron Radiation, Karlsruhe Institute of Technology (KIT), 76131 Karlsruhe, Germany
[d] European Synchrotron Radiation Facility (ESRF), 38043 Grenoble, France

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The deformation and damage micromechanisms ahead of a notch in a large flat specimen have been assessed using *in situ* synchrotron laminography combined with digital volume correlation for the strain evaluation in the material bulk. Despite the enhanced work hardening of the naturally aged Al-Cu-Li alloy several slanted strained bands were found from very early loading onward in the slanted fracture region. The final slanted crack followed one of the strained bands without significant ductile damage growth. In the high stress triaxiality region, close to the notch that underwent substantial local necking, two damage micromechanisms were observed, namely, i) limited void nucleation and growth from intermetallic particles, and ii) slanted shear cracks, even starting from the specimen surface and also located in single grains as shown by *post mortem* EBSD analyses. In the intermediate region a new deformation mechanism of flip-flopping strain bands and resulting flip-flopping cracks have been revealed using "projection DIC" measurements.

1. Introduction

Tearing resistance of thin sheet material is a key design parameter for transport applications [59], as it controls weight and associated fuel and energy consumption related to thin structures. The ultimate aim of the material science and mechanics community is to optimize and design microstructures and to be able to predict their associated macroscopic failure resistance. To date, this aim is very far from being achieved. Even extrapolating from a fracture test with a given geometry and associated stress state to other stress states remains very difficult [45].

Classical ductile damage growth models [16,41] are not tailored to predict failure for low stress triaxialities (i.e., <1). More recent studies have been undertaken using phenomenological approaches defining the strain to failure ε_f as a function of stress triaxiality η and Lode parameter μ [2,13,40,43,58]. Experimentally, tension-torsion test results obtained by pure experimental [17] and hybrid experimental-numerical [45] procedures reveal the monotonic dependence of the strain to failure on the stress triaxiality and the characteristic asymmetric dependence on the Lode parameter. There are various mechanisms that are accounted for and some which are not understood (e.g., shear ductile failure, slanted cracks and nucleation/coalescence in the context of secondary void populations).

A phenomenon that highlights the complexity of the physics of ductile fracture and its challenges is the flat to slanted crack transition where the crack starts from the notch normal to the tensile direction and then flips to 45° when it propagates [1,4,6,26,31,36]. This phenomenon is very hard to predict and poorly understood [3]. The investigation of the origin of such behaviour has engineering relevance since it has been shown that mixed-mode I/III fracture leads to reduced toughness compared to pure mode I fracture [25]. Matching both stress-strain curves and crack propagation paths is difficult to achieve numerically.

Consequently, CT-like specimens were chosen to carry out *in situ* tests [34,35,38] to gain new insights into these mechanisms. The combined use of laminography [18–20] and global Digital Volume Correlation (DVC) [51] has enabled *in situ* displacement fields to be measured at the microscale. Laminography is a nondestructive imaging technique, similar to tomography. However, it allows regions of interest in flat specimens to be analyzed and a wider range

* Corresponding author.
E-mail address: thilo.morgeneyer@ensmp.fr (T.F. Morgeneyer).

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of stress states to be examined compared with tomography [27,30,32,37,42]. Thanks to DVC, bulk displacement fields can be measured and strain fields calculated [35] based on the recordings of 3D images.

For a methodical assessment of the effect of microstructure and work-hardening on the strain field and flat to slanted fracture, it was decided to proceed in a systematic manner by selecting a specific family of alloys using different heat treatments, i.e., change work hardening without changing the initial grain, particle and void structures. In particular, 2139 (Al-Cu) and 2198 (Al-Cu-Li) aluminium alloys have been studied under naturally and artificially aged conditions, thereby changing the nano-precipitate structure and the work hardening behavior [34,35,38,53]. Representing the latest generation of aeronautical alloys, understanding the underlying failure mechanisms is of great interest.

Extensive analyses on CT-like specimens made of recrystallized AA2198 T8 [9,34] have shown that in a zone close to the notch root (i.e., $\approx 800\,\mu$m [34]), and even in its immediate vicinity [9], strained bands appeared very early in the loading history. In the slanted crack region, plane strain conditions in the crack propagation direction predicted in numerical simulations [5] have been confirmed by kinematic measurements [9]. Slanted strained bands appeared well before damage, i.e., plastic flow is very heterogeneous before any significant signs of void nucleation and growth is detected at micrometer resolutions. The origin and observed behaviour (i.e., activation and deactivation) of strained bands at the microscale still remain unclear. The T8 heat treatment induces relatively high yield stress followed by little work hardening.

For AA2198 T8, it was shown by simulation of the sample geometry and von Mises plasticity or a Gurson-type model that the predicted strain fields were very different from the measurements, i.e., no slanted strained bands were present in the computations. This result indicated that plasticity was not described accurately by these models in the case studied.

Low work hardening materials are believed to be prone to flow localisation [49]. Following this reasoning, enhanced work hardening materials should be more resistant to flow localisation. However, this was not the case for another aluminium alloy (i.e., AA2139 T3). Numerous slanted strain bands were observed. Compared to AA2198 T8, their activity was substantially more intermittent [38].

The strain partitioning between the regions within and outside the slanted bands were found to be of the order of 2. Even though there was no clear definition, this behaviour may not be considered as classical localisation where all strains concentrate in the localized band. Slanted strained bands, i.e., heterogeneous strains, that cannot be easily predicted by standard plasticity models, were found. 2D plane strain simulations have shown that a soft zone that is 10% weaker than the surrounding material leads to a strain partitioning with a factor of 2 [38]. The question now is what leads to this substantial softening in the material?

Further, damage nucleation and growth set in very late during the loading history in all these studies. Since these two alloys were different, the strain fields for AA2139 with T8 treatment and AA2198 under T3 condition remain unknown.

In this work, results obtained for a CT-like specimen made of 2198 T3 aluminium alloy are presented. The T3 heat-treatment resulted in a microstructure responsible for high work hardening behaviour [12] as hardening precipitates were missing in this heat treated state.

The paper is structured as follows. The material properties of AA2198 T3 are first introduced. The experimental setup and the laminography imaging technique are outlined next. The basic principles of DVC incorporating strain uncertainty assessments are presented. The results are then discussed. The strain fields in the slant region ≈ 1 mm from the notch root are presented first. Damage growth and final fracture are then shown for all regions of interest (ROIs). A strain field analysis is then performed for the regions closer to the notch root. Qualitative 3D observations, but also damage quantification lead to the identification of two different damage micromechanisms. They are further assessed by *post mortem* SEM and EBSD microstructure analyses that are correlated with the phenomena observed from laminography and DVC results.

2. Experimental procedure

2.1. Material

The CT-like specimen used in the present experiment is made of 2198 T3 aluminium alloy in its recrystallized state, solution heat-treated, stretched by 2–4%, and naturally aged to obtain the T351 (i.e., so-called T3) condition. It is produced by *Constellium C-Tec* and represents the latest generation of aluminum-copper-lithium alloys. The composition of the Al-Cu-Li alloy is shown in Table 1. Compared with the first two generations of Al-Li alloys, AA2198 has higher copper and lower lithium contents. This composition results in increased strength and toughness of the material, which is widely used in the transportation (e.g., aeronautical) industry.

Microscopic observations showed that there were no significant differences in the grain structure between the T3 and T8 conditions [11]. The grain size measured by using a mean linear intercept method was equal to 200–300 µm along the rolling direction (L) and also in the transverse direction (T) [11]. This has been determined to be equal to 10–15 µm in the short transverse direction (S) for the recrystallized sheet by EBSD. Hence, typical pancake-shaped grains in L-T sections were found. In the experiment reported here the loading was applied in the T-direction and the L-direction corresponded to that of crack propagation.

Neither T_1 nor Θ' hardening precipitates were found in the T3 condition. Most of the dispersoids were $Al_{20}Cu_2Mn_3$ or Al_3Zr components. The average dispersoid diameter size was 45 nm with a volume fraction of 1.2–1.8%. Transmission Electron Microscopy (TEM) also revealed no presence of precipitate decoration of the observed grain and sub-grain boundaries nor a precipitate free zone for the T3 condition [11].

Intermetallic particles composed of iron and silicon (white) were elongated and aligned along the rolling direction [11]. However, low intermetallic content (≈ 0.3–0.4 vol %) combined with hardly any initial porosity (i.e., < 0.03 vol. %) made the material challenging for DVC analyses as the laminography image contrast was not very pronounced [35], see Fig. 2(b).

The true stress-strain curve and a power-law fit are given in Fig. 1 for a strain rate of $10^{-3}\,s^{-1}$ and loading in the T direction. The gauge section of the sample was 6 mm wide, 2 mm thick and 32 mm long. The fit has been performed using a classical power hardening law

$$\sigma_{true} = K\varepsilon_{true}^{n} \tag{1}$$

with $K = 685$ MPa, and $n = 0.21$. This result shows the enhanced work hardening behaviour of the T3 condition [12] in contrast to the T8 condition that only yielded a strain hardening exponent of $n = 0.08$. The enhanced work hardening of the T3 heat treated material occurred from about 0.04 to 0.135 strain, which covered most of the strain range.

2.2. Laminography

Synchrotron radiation computed laminography is a