Acta Materialia 112 (2016) 194–208

Contents lists available at ScienceDirect

Acta Materialia

journal homepage: www.elsevier.com/locate/actamat

Full length article

# Effect of strain rate and temperature on fracture of magnesium alloy AZ31B

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ARTICLE INFO

*Article history:*
Received 26 November 2015
Received in revised form
22 February 2016
Accepted 23 March 2016
Available online 19 April 2016

*Keywords:*
Mg alloys
Ductile fracture
Twin-sized voids
Second phase particles
Void coalescence

ABSTRACT

The fracture behavior of twin roll cast AZ31B magnesium has been studied in tension. Explored ranges of temperature and strain rate were wide enough to encompass noticeable transitions in fracture characteristics at both macroscopic and microscopic scales. Overall properties include strain rate sensitivity, thermal softening, strain to failure, failure mode, plastic anisotropy and work of fracture. The micromechanisms of damage initiation and accumulation were inferred from tests interrupted at incipient macrocrack formation for key test conditions. The competing roles of voids nucleated on aligned inclusion clusters and at presumed contraction double twins are discussed.



## 1. Introduction

Magnesium alloys are high specific strength structural metals of special interest to the aerospace and automotive industries [1]. However, their widespread use hinges on developing fail-safe design methods, improving their ductility and lowering manufacturing cost, e.g. by decreasing forming temperatures or increasing forming rates. At ambient temperatures, Mg alloys are often referred to as quasi-brittle in that a tensile specimen is unable to sustain stable plastic flow beyond the onset of necking [2]. In wrought form, Mg sheets, plates or rods typically exhibit a strong basal texture, which results from the plastic anisotropy inherent to their hexagonal close packed (HCP) structure with a high $c/a$ ratio [3]. This fundamental anisotropy of HCP structure thus imparts wrought Mg products with strong plastic anisotropy. The latter manifests, under simple loading, by the activation of a limited number of deformation systems (slip and twinning) leading to incompatible deformations in polycrystals and, presumably by way of consequence, to limited fracture resistance [2].

At room temperature, the fracture behavior of Mg alloys in tension has two essential characteristics: at the macroscopic level, shear failure occurs with limited post-necking deformation [4]; at the microscale, twin-sized cracks are frequently observed and appear to be associated with $\{10\bar{1}1\}$ $\{10\bar{1}2\}$ contraction double twins [5,6]. The role of brittle second-phase particles has also been discussed [7,8] although the uniaxial fracture behavior seems to be dominated by mechanical instability and twinning-induced cracking [9]. Kondori and Benzerga [9] have recently investigated stress triaxiality effects on damage and fracture in alloy AZ31. They showed that the notch ductility of this material is higher than its uniaxial ductility, in contrast with other metallic materials [10]. They attributed the observed behavior to (i) the change from macroscopic shear failure in uniaxial bars to nominally flat failure in notched bars, and (ii) a change in damage mechanisms from twinning-induced microcracks under uniaxial tension to void formation, notably at second phase particles, followed by void growth to coalescence in notched bars.

As expected, the ductility and formability of Mg alloys improve with increasing temperature [11,12] or decreasing strain rate [3,13] since differences in activation barriers diminish among the various deformation systems. While these trends are common to other metallic alloys (outside the regime of so-called dynamic strain aging) a systematic analysis of damage mechanisms at varying

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http://dx.doi.org/10.1016/j.actamat.2016.03.061
1359-6454/

temperatures and loading rates is lacking in Mg alloys. Deng et al. [14] have examined the changes in fracture morphology due to temperature and strain-rate variations but did not investigate details of the damage accumulation to crack initiation processes. Other work on damage and fracture focused either on a single nominal strain rate condition, mainly between $10^{-4}$ and $10^{-3}$ $s^{-1}$ [4,15], or on superplastic conditions [16].

This paper examines the dependence of macroscopic fracture measures, plastic anisotropy and damage initiation and progression micromechanisms upon temperature and strain rate. The chosen material is from the Al–Zn ternary system (named as AZ alloys). Focus is laid on uniaxial loading of sheet metal and builds on earlier work devoted to microstructure evolution [17]. Most importantly, a series of tension tests were interrupted at incipient macroscopic cracking in order to characterize damage initiation and its connection to twinning. Stress triaxiality effects are not investigated here, although triaxiality effects will be discussed when significant post-necking plastic flow occurs. The parameter space for the fractography analyses that are based on tedious test interruptions and specimen sectioning in various planes is quite large, even with the restriction to uniaxial loading. In order to reduce this parameter space, temperature and strain rate conditions were selected on the basis of non-trivial trends in the work of fracture $W_f$. From the mechanics point of view, $W_f$ is a fracture measure that is weakly dependent on specimen size in the quasi-static regime [18], unlike the strain to failure. As such, it is commonly used as an indicator of crashworthiness or energy absorption capacity [1,12]. The competing damage initiation and progression mechanisms are then described and their sensitivity to temperature and strain rate discussed.

## 2. Experimental procedure

### 2.1. Material

The material used in this study is a twin roll cast (TRC) AZ31B sheet (2.5–3.5% Al, 0.7–1.3% Zn, 0.2–1.0% Mn) rolled to a 3 mm thickness and supplied by POSCO. The three principal directions of the sheet are denominated: rolling or longitudinal (L), transverse (T) and through thickness or short transverse (S), Fig. 1. Metallographic sections were prepared by grinding with SiC paper and polishing with diamond and colloidal silica suspensions. For etching, an acetic-picral solution, with a 5:2 acetic-to-picric acid ratio was used. The initial microstructure of the LT and LS views are presented in Fig. 2 The average grain size was measured as ~8 μm in LT sections and ~11 μm in LS sections. Note that more twins are observed in LS and TS (not shown) sections simply because the LT section was observed close to the free surface. As shown in a

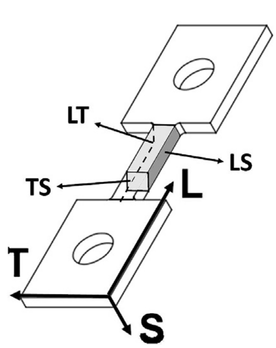


**Fig. 1.** Tensile specimen geometry and reference directions: longitudinal (L), transverse (T) and short transverse (S). Orientations of the three metallographic sections analyzed: LT, LS and TS.

previous study [19], the microstructure is somewhat graded which manifests in through-thickness variations in both the twin density and grain size. The initial orientation of the material was obtained through Electron Backscatter Diffraction analysis EBSD [17]. The as-received sheet exhibits a basal texture with the basal planes parallel to the LT plane as shown in the pole figures of Fig. 2c.

### 2.2. Tensile testing

Rectangular prismatic tensile specimens were machined according to the geometry in Fig. 1 with a gage length $L_0 = 19.05$, width $w_0 = 6.35$ and thickness $t_0 = 3.0$ all in mm. The specimens were loaded along the rolling direction. Tensile tests were conducted at true axial strain rates between $10^{-4}$ and $10^{-1}$ $s^{-1}$ and temperatures between room temperature (RT) and 300 °C using a MTS Insight 30 kN machine equipped with a ThermoCraft environmental chamber. The crosshead speed was adjusted so as to keep a constant axial strain rate. This implies a constant true strain rate prior to necking. In addition to the tests conducted to complete rupture [17], new tests were carried out to better characterize the scatter in the strain to fracture. Also, a series of interrupted tests were conducted at 100 °C for two strain rates $10^{-3}$ and $10^{-1}$ $s^{-1}$, and at 200 °C for a strain rate of $10^{-1}$ $s^{-1}$. The specific choice of these conditions will become clear in context. The samples were kept in the environmental chamber for 30 min before the test, in order to achieve homogeneous heating of the specimen without causing grain growth [17]. In some experiments, a laser extensometer was used to measure elongation and enable test interruption at incipient load drop, especially at high strain rates. The strain to failure $\varepsilon_f$ was calculated as follows:

$$\varepsilon_f = \ln\left(\frac{w_0 t_0}{w_f t_f}\right) \tag{1}$$

where $w_0$ and $w_f$ are the initial and final widths of the specimen, respectively; and $t_0$ and $t_f$ are the initial and final thicknesses, respectively. The lateral strains at fracture were defined as (dropping the index 'f' for convenience):

$$\varepsilon_T = \ln\left(\frac{t_0}{t_f}\right) \tag{2}$$

$$\varepsilon_S = \ln\left(\frac{w_0}{w_f}\right) \tag{3}$$

so that $\varepsilon_f = \varepsilon_T + \varepsilon_S$. Also, the Lankford coefficient, or r-value, was calculated at fracture as $r^L = \varepsilon_T/\varepsilon_S$. The value of $r^L$ was also measured at test interruption for the conditions reported above.

The work of fracture $W_f$ was obtained by integrating the area under the nominal stress–strain curve:

$$W_f = \int_0^{\varepsilon_L} \sigma d\varepsilon \tag{4}$$

where $\varepsilon_L = \delta L_f / L_0$ is the elongation to fracture and σ is the nominal stress.

A series of tests were interrupted in order to elucidate damage mechanisms. While this could be done at various stages of the deformation process, here we focus our attention to incipient macroscopic cracking. Such analyses were performed for some carefully selected loading conditions, as will be explained below. For each condition, at least two realizations of the test were first conducted to complete rupture so as to obtain an estimate of the nominal strain to failure initiation and scatter. The nominal strain