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On the behaviour of the magnesium alloy, AZ61 to one-dimensional shock loading

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Abstract

The behaviour of the magnesium alloy AZ61 during one-dimensional shock loading has been investigated in terms of the Hugoniot (shock-induced equation of state) and the shear strength variation in terms of variation with impact stress and pulse duration. The Hugoniot has been shown to be near identical with the similar AZ31, as would be expected with related dilute alloy systems. The shear strength has been observed to increase with applied shock stress, in common with many other materials. The shear strength has also been observed to increase with time behind the shock front. It has been suggested that this be due to twinning early in the deformation process, followed by dislocation based at later stages.

Keywords: Magnesium alloys; Impact behaviour; Shock loading; Dislocation mobility; Twinning

1. Introduction

The requirement for high-quality materials and microstructural data for input into constitutive models for high strain rate and high-velocity impact events is becoming increasingly important. For example, in the military for armour design and defeat, from aerospace to adequately design for foreign object damage and bird strike, and satellite protection from orbital debris. Given that such events involve a projectile of arbitrary material, mass, shape, impact velocity and angle onto a target of complex geometry (for example an aerofoil in a jet turbine engine), the impact conditions will be extremely complex, with all states of stress and strain (compression, tension and shear) being present under the impact site. As such, meaningful analysis of this event is rendered extremely complex, usually with only qualitative judgements possible. Consequently, it is usual to generate materials data under much simpler loading geometries and use the results as input data into computer models that can then be compared to more realistic events. Under quasi-static strain rates, materials can be loaded in uniaxial stress in either compression or tension, or via plane strain for fracture toughness. As the strain rate increases (10^2–10^4 s^{-1}), devices such as the split Hopkinson pressure bar (SHPB) are employed, where the sample is again loaded under conditions of uniaxial stress. As the strain rate increases still further, inertial confinement makes uniaxial stress increasingly difficult and eventually impossible to maintain under these conditions, and hence a different approach is required. The use of a planar shock wave will load a material under conditions of uniaxial strain, whereby all deformation is accommodated along the loading axis (x), thus

$$\varepsilon_x \neq \varepsilon_y = \varepsilon_z = 0 \quad \text{and } \sigma_x \neq \sigma_y = \sigma_z \neq 0 \tag{1}$$

where ε and σ are the imposed strain and stress respectively, and the subscripts y and z are the axes orthogonal to the loading axis x. Such loading can be imposed explosively (with a plane wave lens) or via the impact of a flat plate of known properties, again driven explosively or via

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a gun (using either powder or compressed gas). This latter technique requires that both the flyer and target plates are accurately manufactured (flat and parallel to ±5 μm is a common standard), and that the impact itself is planar to within 1 milli-radian. The target can be instrumented using a variety of techniques such that materials properties data can be extracted. A more complete discussion of shock loading is beyond the scope of this paper. However, the interested reader is directed to the review article of Davison and Graham [1] and the textbook by Meyers [2].

Of particular interest is the variation of materials shear strength (τ) with shock conditions, as this has been shown to be an excellent indication of resistance to ballistic attack, shear localization and fragmentation [3]. This is derived from the standard relation

$$\sigma_x = P + \frac{4}{3}\tau \tag{2}$$

where P is the hydrostatic stress. In principle, τ is simple to determine as the pressure can be subtracted from the stress. Unfortunately, P is generally determined from ambient bulk modulus data, and thus a degree of uncertainty occurs. However, the hydrostatic pressure can be considered to be the average of all three orthogonal components of stress, thus

$$P = \frac{\sigma_x + \sigma_y + \sigma_z}{3} \tag{3}$$

and hence, in combination with Eq. (2), an expression for shear strength can be written in terms of stress only

$$2\tau = \sigma_x - \sigma_y \tag{4}$$

In practical terms, both components of stress can be measured using manganin stress gauges [4–6]. Other methods to determine shear strength include the monitoring of rear surface movement during inclined impact (or pressure shear) experiments [7] and comparison of load–unload/load–reload experiments [8,9]. A major advantage of the stress gauge method is that it places the gauge within the material flow during shock loading, and thus it is possible not only to determine the variation of shear strength with shock amplitude, but also with pulse duration as well. However, it should also be acknowledged that this by its very nature is an invasive process. As such, it would be desirable to compare and contrast all these techniques on a single material (preferably from a single batch); unfortunately, to the authors knowledge, this has never been carried out. However, various grades of tungsten heavy alloy (WHA) and pure tungsten have been examined, and in previous works [10,11] it has been possible to compare the lateral stress gauge approach to pressure shear [7] and calculation of shear strength using Eq. (2) [12]. Results show a close degree of agreement of shear strength varying with stress. In addition, as has already been mentioned above, the gauge is placed in the material flow, and hence variations of shear strength with time can be observed. In high stacking fault energy (SFE-γ) face-centred cubic (fcc) metals such as nickel [13] we have observed that the shear strength increases behind the shock front, whilst, in contrast, body-centred cubic (bcc) metals such as tungsten and its alloys [10,11] and tantalum [14] a decrease in shear strength is observed. In the case of the former, it was suggested that a rapid build up in dislocation density caused shear strength to rise. This correlates with earlier shock recovery work on nickel by Murr and Kuhlmann-Wilsdorf [15], who showed that the resultant dislocation cell structure did not reach its final state until 0.5–1 μs after the arrival of the main shock. In the case of bcc metals, it was suggested that the high Peierl's (lattice friction) stress resisted the generation of new dislocations, hence deformation was more dependent upon the motion of dislocations already present within the microstructure [14].

The situation with hexagonal-close packed (hcp) metals is not as well understood, and is further complicated by the fact that the microstructural response is influenced to a significant degree by the *c*/*a* ratio of the unit cell [16]. Common engineering materials such as the titanium alloy Ti–6Al–4V have been studied under high strain rate and shock conditions [17–20] as it has been considered as potential armour material, as well as its use in the aerospace industry. The group IVa (Ti, Zr and Hf) have also been studied by Cerreta and her colleagues [21,22] on a microstructural basis. In a recent paper, lateral stress measurements in the titanium alloy Ti–6Al–4V [23] showed that there was a degree of hardening behind the shock front that was influenced by the orientation of the loading axis to the original axis of the as-received bar stock. It was suggested that the inherent anisotropy in that material influenced the relative amounts of dislocation slip and twinning, which in turn would manifest themselves in the shock-induced mechanical response.

Unlike many metals, the available data on the shock response of magnesium and its alloys is comparatively sparse. Early equation of state work on both pure magnesium and its alloy AZ31 has been summarized by Marsh [24], whilst Fuller and Price [25] investigated release paths for the alloy ZW3, results suggesting that the material showed a strong elastic–plastic response. Schmidt et al. [26] examined the spall response of a number of alloys, including AZ31, as a function of temperature. They reported that the spall strength was ca. 1500 MPa at room temperature, dropping rapidly with temperature to a near constant 300 MPa and finally to 200 MPa at incipient melt. Measurements of the spall strength of nominally pure magnesium [27,28] showed that it dropped precipitously as the melting point was approached. Interestingly, they also observed (in common with aluminium) that the Hugoniot elastic limit (HEL – the yield strength under one-dimensional strain) increased under the same conditions. It was suggested that an increase in point defects (vacancies, atoms occupying interstial positions, etc.) due to thermal activation could hinder the progress of dislocations through the microstructure and hence yield a positive dependence of HEL with temperature. They also suggested