Mechanical behaviour of AZ91D and AZ31B magnesium alloys at wide range of strain rates and temperatures

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**MECHANICAL BEHAVIOUR OF AZ91D AND AZ31B
MAGNESIUM ALLOYS AT WIDE RANGE OF STRAIN
RATES AND TEMPERATURES**

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SCHOOL OF MECHANICAL AND AEROSPACE ENGINEERING

2014

**Mechanical Behaviour of AZ91D and AZ31B Magnesium
Alloys at Wide Range of Strain Rates and Temperatures**

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A thesis submitted to the Nanyang Technological University
in partial fulfilment of the requirements for the degree of
Doctor of Philosophy

2014

STATEMENT OF ORIGINALITY

I hereby certify that the work embodied in this thesis is the result of original research and has not been submitted for a higher degree to any other University or Institution.

29-01-2014

Date

A handwritten signature in black ink, appearing to read 'Iram Raza Ahmad', written in a cursive style.

Iram Raza Ahmad

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Summary

In many industries especially in automobile and aerospace, developments in near future are mainly concerned with the weight reduction to meet the global challenges of fuel economy and reduction of toxic emissions. Magnesium alloys being the lightest of all metals offer great potential to achieve that weight reduction by replacing the most commonly used materials, i.e. steel, aluminium and plastics. Magnesium alloy are being used in variety of non-structural and to some extent in semi-structural applications. However, their use in structural applications such as auto-body structure is very limited. Limited data available for high strain rate deformation of magnesium alloys combined with an insufficient understanding of the underlying deformation mechanisms adds to the reservations for their use as structural materials. For widespread usage of magnesium alloys, their dynamic behaviour must be determined to assess their performance during a crash event. In the present work, an experimental and constitutive study followed by the microstructural analysis has been carried out to investigate the dynamic behaviour of as-cast AZ91D and wrought AZ31B magnesium alloys. These alloys have been tested at strain rates in the range between 10^{-4}s^{-1} and $4 \times 10^3\text{s}^{-1}$ and at temperatures between -30°C and 200°C under compression and under tension between 10^{-4}s^{-1} and 1500s^{-1} strain rates and at 25°C and at 250°C temperatures. Quasi-static tests were performed using universal testing machine and high strain tests were carried out with Hopkinson Bar apparatus. Dry ice was used to achieve below zero temperature while the elevated temperatures were obtained by using coil heater.

Increasing stresses, larger strains and higher energy absorption are observed with increasing strain rate in both alloys under compression as well as in tension. On the other hand, lower stresses and relatively larger strains are observed at elevated temperatures. However, temperature has little effect on the mechanical behaviour of these alloys at dynamic strain rates. The alloys deform by dislocation glide and twinning at room temperature. Number of twins decreases as the strain rate increases. Twins, voids, shear bands, grains refinement and elongation have a key role in alloys response such as strength, ductility, hardenability and energy absorption. AZ31B alloy is strongly anisotropic and shows a considerable tensile-compressive yield asymmetry. The experimental data was fit to the Johnson-Cook model and the results are in reasonable agreement except in the beginning portion of the flow curves, where the fitted curves deviate from the experimental data.

Chapter 1

Introduction

1.1 Background

To reduce the fuel consumption and thus increase the performance of vehicles, lightweight materials are of great interest and magnesium alloys are one of the prime choices. The development of high strength lightweight materials to reduce the weight of the product without compromising the cost, performance and safety is vital to compete today's global market. Technological demanding industries, such as automotive and aerospace sectors, over the past few decades have increased their demand for lightweight and strong materials. Automotive and aerospace manufacturers are looking for ways to reduce the gross weight of the vehicle to economize the fuel consumption as well as to improve their performance. Around 80% of the vehicle weight is due to steel or cast iron used in the body structure [1]. While the use of magnesium alloys is quite low contributing only a fraction of the vehicle total weight.

Magnesium is the lightest metal on the earth with density of 1740 Kg/m³ that is about two-third of aluminium density (2700 Kg/m³), about one-fourth of steel (7900 Kg/m³) and slightly higher than that of reinforced plastics (1100-1400 Kg/m³)[2-5]. Excellent strength to weight ratio, high specific stiffness, high impact resistance and good castability [2-7] make magnesium alloys attractive structural materials. Their specific yield strength (σ_Y/ρ) and tensile strength are comparable with those of aluminium and low carbon steel (Fig. 1-1) [8].

Magnesium alloys are of key importance for the future automobile industry to achieve lightweight and fuel efficient vehicles. They are being considered as an excellent alternative to steel, iron, and even aluminium alloys.

Figure 1-1: Mechanical properties of different sheet materials [8]

1.1.1 Magnesium Alloys: Past, Present and Future Trends

Magnesium due to its lightness has been of interest to the automotive and aircraft industry since the thirties of the last century till to date. The use of magnesium alloys in aircrafts started in 1920's. Efforts were made to establish magnesium in several structural applications in the US. However, Germany was leading particularly in 1930s with using magnesium alloy in several aircraft parts including tail wheels, break shoes, break levers and pistons [9]. In aircrafts, magnesium alloys are used in engine parts that are not exposed to excessive heat, in wheels of about 75% of commercial and military aircrafts, in structural parts such as tail wheel forks, airline hinge brackets, parts of the landing gear and control system etc. [10]. In extremely high speed aircrafts, the outer skin needs to have sufficient strength to resist the skin failure. A thicker skin can sustain excessive loads resulting from high intensity local pressures at diving speed. Lightweight magnesium alloys can be used successfully to overcome this problem without adding to the weight [10]. Major European aerospace industries are

seriously investigating magnesium alloys as a weight reduction alternative to aluminium and to use 10% to 15% of magnesium components in civil aircrafts in 2015-2020 [11]. The potential exists for magnesium alloys in several structural applications in airframe and engine as well as in the interior of aircrafts.

Like aerospace industry, magnesium alloys are being used in lots of automobile applications in passenger vehicles as well as in military vehicles. In mass production vehicles, the important motivation for weight reduction is to reduce the fuel consumption and associated CO₂ and other toxic emissions. The development of new high purity Mg-Al-Zn alloys helped to control the corrosion which enhanced the use of magnesium in automotive applications [9, 12]. The most notable application of magnesium was its use in Volkswagen's produced "Beetle" in [2, 7, 13], with consumption of about 42000 tons per year [4, 5]. Cast magnesium alloys are found in car bodies, gear box housings, steering wheels frames, steering column housings, seat frames, instrument panels and many more applications [1-5, 7-9, 14-20]. In future magnesium die castings may also find applications as structural chassis components and engine cradle [5]. Magnesium wrought alloys are expected to show a huge impact in future.

Various Organizations are making R&D efforts to increase the usage of magnesium alloys in automotive industry. United States automotive materials partnership (USAMP) has set the 2020 strategic goal to replace 630 lbs. of current ferrous and aluminium parts with 340 lbs. of magnesium alloys components thus reducing the average vehicle mass by 290 lbs. [21]. Ortal Die-casting Company a part of TG group, Israel [22], BMW [23], Ford, Toyota, Honda, Meridian Lightweight Technologies Canada [24], Australian Commonwealth Scientific and Industrial Organization (CSIRO) [25], universities in Europe, United States, China,

Korea and Japan with collaboration of manufacturers and R&D organizations [26] are few among the important groups carrying out research and development on magnesium alloys. The magnesium usage is not just in cast housings; the future trend is well ahead, as depicted in Fig. 1-2 [16].

Figure 1-2: Future trend of magnesium alloys in vehicle manufacturing [16]

1.1.2 Materials Characteristics required in Structural Applications

In structural applications where deformation by crash is important; components should have high strength, high energy absorption capability, higher crush-resistance and absence of collapse mechanisms that are accompanied by a catastrophic reduction in load carrying capacity. Usually body, chassis and powertrain are the major systems associated with structural use of materials. Various components need different requirements with respect to their functions; bending stiffness and flexural stiffness are needed for large flat components, compressive and tensile strength for the strength related components and the strain rate dependence, strength, elongation and hardness characteristics are essential requirements for energy absorbing components [1, 4, 5, 15-17, 27]. Chassis components are highly diverse with many of them have structural

functions. Structural parts such as steering-wheel, steering column, cross car beams, seat frames, suspension links, spring and rods and some bumper structures carry specific loads and have to meet certain crash worthiness requirements. High strength, high ductility, high fatigue and high impact strength are essential factors in the area of chassis applications [1, 4, 5, 8, 15-17 and 27]. The engine and transmission encompass various individual complex assemblies. Components perform different functions ranging from structural to kinematics, power transmission, cooling/heating, lubrication, timing etc. characterized with highly stressed conditions caused by high temperatures, rapid motion, friction, high mechanical loads and fluid pressure [2, 3]. Materials used in power train applications should have high strength at elevated temperatures, high elastic modulus, better fatigue strength, higher hardness, and corrosion resistance [1-4, 8, 15, 17, 27, 28].

1.2 Thesis Organization

A brief literature review of the Hopkinson Bar and high strain rate behaviour of AZ91D and AZ31B alloys is laid down in chapter 2. The experimental techniques, data analysis procedures and specimen geometry used for the current study have been laid down in Chapter 3. Experimental results for AZ91D and AZ31B are discussed in Chapter 4 and Chapter 5 respectively, followed by the microstructural analysis in Chapter 6 and constitutive analysis in Chapter 7. Finally, Chapter 8 presents conclusions from the current research as well as provides some recommendations.

Chapter 2

Literature Review

This chapter provides an overview of the high strain rate behaviour of materials in general. Next, a thorough literature review on the mechanical behaviour of AZ91D and AZ31B is presented which includes the strain rate and the temperature effects on their mechanical properties. The motivation of the present research is laid down in the end.

2.1 High Strain Rate Behaviour of Materials

Strain rate has a strong influence on the mechanical behaviour of the material. At high rates of loading, the mechanical response of most materials is quite different from their behaviour under static loading. Such rate dependence of materials behaviour is observed for nearly all materials including metals, ceramics [29], polymers [30] and concrete [31]. In impact or dynamic loading, the sudden application of a force to material gives rise to two important effects, which control the material's response. These effects include the inertia effects and variations in the inelastic properties of the materials (strain rate effect). It is generally observed that the yield strength of materials increases with the strain rate and the material's plastic flow stress depends on the material hardening. Most materials are sensitive to the rate of deformation. In most metals, stress has been shown to be linearly dependent on the logarithm of the strain rate for certain ranges of strain rates. In general three ranges are observed with different stress-strain relationship on stress vs. strain rate (log scale) plot depending upon

the mechanisms governing the plastic flow [32]. For En3Bsteel these regions can be seen in Fig. 2-1 [33].

Figure 2-1: Effective tensile stress as a function of strain rate for En3B Steel [32]

In Fig. 2-1, region-I is governed by long-range obstacles to dislocation motion, in region-II thermally activated dislocation motion is the dominant mechanism to control the deformation, and region-III is primarily controlled by dislocations drag [32]. Plastic deformation can occur by dislocation motion and twinning, twinning might account for up to a few per cent while dislocation processes are principle mechanism for plastic deformation of materials. During plastic deformation, shear stress acting on the slip planes causes these dislocations to move continuously through the lattice. During their movement, they continuously encounter obstacles such as solute atoms, particles, grain boundaries as well as other dislocations that make their motion difficult [32]. These dislocations need energy to overcome these obstacles to continue their motion which is provided either by increasing the applied stress or by temperature variations. Thermal energy only is sufficient to overcome short-range obstacles but long-range obstacles cannot be overcome by thermal energy. Therefore the flow stress

involves both thermal and athermal components [32]. The thermal energy ΔG provided by thermal fluctuations to overcome obstacles as a function of temperature and the strain rate is given by the Eq. (2.1)

$$\Delta G = kT \ln \left(\frac{\dot{\epsilon}_0}{\dot{\epsilon}} \right) \quad (2.1)$$

where k is Boltzmann constant, T is absolute temperature and $\dot{\epsilon}_0$ is a material constant. This equation shows that the short-range energy barrier is reduced as the temperature increases, but is increased by increasing the strain rate. Therefore, increasing temperature tends to soften the material, whereas increasing strain rate strengthens it. Lindholm and Yeakley [34] derived a constitutive model based on thermal activation given in Eq. (2.2)

$$\sigma = \sigma_G + \frac{\Delta G_0}{V} + \frac{kT}{V} \ln \left(\frac{\dot{\epsilon}_0}{\dot{\epsilon}} \right) \quad (2.2)$$

where ΔG_0 is the activation energy at 0°K and V is the activation volume. This equation shows the logarithmic dependence of the flow stress on the strain rate which is in accordance with an important material behaviour, the strain rate sensitivity.

2.2 Deformation of Magnesium Alloys

According to von Mises criterion [35] more than 5 independent slip systems are needed for metals to deform uniformly and without failure at grain boundaries. An independent slip system is one for which slip displacements on it cannot be duplicated by a combination of displacements on other slip systems. Magnesium alloys have a hexagonal closed pack (HCP) structure. Therefore, only a limited number of slip systems are available to accommodate plastic deformation. At

room temperature, magnesium alloys have only four independent slip systems, and the remaining deformation is accommodated by twinning. However, at elevated temperatures additional slip systems become active, providing sufficient independent systems to fulfil the von Mises criterion resulting in improved ductility at elevated temperatures.

2.2.1 Role of Slip in Deformation

In magnesium alloys, slip occurs on the basal, prism and pyramidal planes as shown in Fig. 2-2 [35]. The deformation mode in any slip system is controlled by the local stress upon the slip system. When an external force F , is applied to a crystal, a shear stress can be resolved onto the given plane in a given direction. Slip occurs when the resolved shear stress on the slip plane reaches the critical value for that particular system, the critical resolved shear stress (CRSS).

Figure 2-2: Unit Cell and Major planes and directions[35]

Table 2-1 shows all available slip systems in HCP metals [36]. However, their activation depends on the given set of applied conditions i.e. temperature, strain rate etc.

Table 2-1: Slip system in HCP metals [36]

Slip system	Burger vector type	Slip directions	Slip plane	No. of slip systems	
				Total	Independent
1	a	$\langle 11\bar{2}0 \rangle$	Basal (0001)	3	2
2	a	$\langle 11\bar{2}0 \rangle$	Prismatic-I $\{10\bar{1}0\}$	3	2
3	a	$\langle 11\bar{2}0 \rangle$	First-order Pyramidal-I $\{10\bar{1}1\}$	6	4
4	c+a	$\langle 11\bar{2}3 \rangle$	2nd-order Pyramidal-II $\{11\bar{2}2\}$	6	4
5	c	$\langle 0001 \rangle$	Prismatic-I $\{10\bar{1}0\}$	3	2
6	c	$\langle 0001 \rangle$	Prismatic-II $\{11\bar{2}0\}$	3	2

Due to hexagonal closed pack structure of magnesium alloys, basal slip and twinning are only mechanisms for deformation at room temperature. The critical resolved shear stress for non-basal slip system is much greater than basal slip system at room temperature, so magnesium can be deformed easily only within their basal planes at room temperature. From table 2-1, only two independent slip systems exist along slip direction $\langle 11\bar{2}0 \rangle$. The pyramidal $\langle a \rangle$ slip produces identical shape change as produced by combined basal slip and prismatic $\langle a \rangle$ slip, the resulting number of independent slip systems from all three deformation modes are still 4 [35] and thus cannot accommodate deformation along $\langle c \rangle$ direction to fulfil criterion stated by von Mises. Therefore, some other non-basal slip systems having a component in c-direction (Table 2-1) should be activated [37] or the deformation should occur by twinning.

At elevated temperature at 200°C and above, the critical resolved shear stress for non-basal system decreases rapidly, thus secondary slip systems $\{10\bar{1}0\} \langle 11\bar{2}0 \rangle$ prismatic and $\{10\bar{1}1\} \langle 11\bar{2}0 \rangle$ pyramidal (Fig. 2-3 and Table 2-1) become active. The activation of pyramidal slip system is relatively difficult at room temperature due to larger burger vectors (c+a) and small pyramidal plane distance, but increasing the temperature will ease the activation process of pyramidal slip

system [38]. These additional slip planes explain the increasing ductility at elevated temperature.

Figure 2-3: Pyramidal slip and prismatic slip [35]

2.2.2 Deformation by Twining

In addition to slip, twining is another mechanism of deformation at room temperature. Twining is a sudden localized shear process that involves a certain volume within a crystal. The lattice in a twinned region is related to untwined crystal, usually a mirror image of the lattice of the matrix in which it is formed [39]. A simple example is shown in Fig. 2-4 where twining has taken place along the plane K_1 and the twining direction η_1 . Twinned region differs considerably in orientation from the grain it is formed. Fig. 2-5 illustrates the basic difference in the lattice translation by slip and by twining.

Figure 2-4: A twining transformation [39]

Figure 2-5: A schematic of difference between slip (a) and twinning (b)[39]

Twin initiation or growth occurs when the macroscopic stress across the twin plane resolved in the direction of the shear reaches a critical value. Usually onset of twinning occurs when the critical resolved shear stress (CRSS) of a twin system exceeds that of any other slip system. The dominant twinning mode observed in magnesium is in $\{10\bar{1}2\} \langle 10\ 1\ 1 \rangle$ [40-42] plane as shown in Fig. 2-6. However, there are several additional active planes as well, i.e. $\{10\bar{1}1\}$, $\{30\bar{3}4\}$, $\{10\bar{1}3\}$ and $\{10\bar{1}4\}$.

Figure 2-6: A twinning plane

Under impact loading, most of the metals exhibit twinning. It has been observed that when the strain rates are relatively lower dynamic strain rates (496/s-964/s), the microstructure is predominated by the high density of twinning, while

the volume fraction of twins is decreased by increasing the strain rate up to 2120/s [43]. Two important twinning systems (Fig. 2-7) "tension/extension twinning" $\{10\bar{1}2\} \langle 11\bar{2}0 \rangle$ and "contraction twinning" $\{10\bar{1}1\} + \{10\bar{1}2\} \langle 11\bar{2}0 \rangle$ could accommodate the plastic deformation in magnesium materials [44]. "Tension/extension twinning" in magnesium can only be activated by a tensile stress parallel to the c-axis (perpendicular to the basal plane) when the c/a ratio is less than 1.73 [45]. Likewise, the contraction twinning can only be activated by a compression stress parallel to the c-axis when the c/a ratio is greater than 1.73 [43, 45].

Figure 2-7: Extension twins and contraction twins [44]

Twinning becomes a dominant mode of plastic deformation as the temperature is decreased [45]. Due to the limited slip system at room temperature, twinning is proven to be the dominant mode responsible for plastic deformation in magnesium alloys [46]. At lower temperature, the mobility of lattice dislocations is low which hinders the slip activity. Since slip is a thermally activated activity, therefore at lower temperatures accommodating slip is low, leading to enhanced twin propagation.

Deformation twins, especially twin intersection play an important role in the formation of adiabatic shear bands (ASBs) [47]. Shear bands are commonly observed after severe plastic deformation of AZ31 alloy, they can be thought of places where the plastic deformation is strongly localized. An accompanying effect is extensive grain size refinement inside the shear band, which in this case leads to an average grain size diameter of 1 μm [48]. There is a general agreement that small grains inhibit twinning [38] and grain size had a significant effect on the tendency of twinning in Magnesium [49]. It is common for the twinning stress to increase with decreasing grain size more rapidly than the stress required to activate slip. As a result, the formation of twins becomes more difficult with decrease in grain size. Since twins form as thin lamellae and cluster to form bundles, the interaction of twins with the surrounding matrix plays an important role in the strain hardening behaviour. The formation of twins within existing grains introduces new boundaries which act as barriers to dislocation movement thus contributing to the hardening behaviour of the material. The dislocations in the matrix form pile-ups near a twinning boundary and leads to a stress concentration near the boundary. Therefore, it is believed that twin boundaries and twin-slip interaction lead to high strain hardening in the metal.

2.3 Effect of alloying elements in deformation

Alloying generally increases the strength and decreases the ductility of the host metal. The amount of hardening depends on the relative size and modulus of alloying element and the chemical interactions between the two species. The hardness increases because the dislocation motion is impeded causing a

corresponding decrease in the ductility. Also, alloying affects slip more than twinning [44].

Alloying with aluminium, however, actually increases the ductility as well as the hardness [50]. It has been suggested that alloying with Al activates prismatic slip by changing the axial ratio; it actually increases the c/a ratio which seems to lessen the possibility of slip on $\{1010\}$. The prismatic plane stacking fault energy is higher than the stacking fault energy of the basal plane and alloying with Al reduces the stacking fault energy sufficiently, but there is no experimental evidence for this. The most likely explanation is that alloying reduces the stress on the prismatic plane, making possible for slip to occur on prismatic plane adding to the ductility of the alloy. Zinc is added mainly to improve the corrosion resistance and the strength of magnesium. It helps to overcome the harmful corrosive effects of iron and nickel impurities. However, the amount of zinc more than 1 wt% can lead to hot shortness [50]. Presence of manganese in small amount causes the formation of aluminium-manganese intermetallic compounds Al_8Mn_5 or $\text{Al}_{11}\text{Mn}_4$.

2.3.1 Intermetallic phase in Magnesium Alloys

Intermetallic phases can be found in almost every magnesium alloy and play a vital role in optimizing the microstructure and mechanical properties. These intermetallic compounds can precipitate during casting at high temperatures or form by solid state transformations at low temperatures. Several morphologies of intermetallic compounds are found i.e. blocky, rosette-like, Chinese script, lamellar and needle-like [51-52].

In AZ91D and AZ31B alloys, an intermetallic compound eutectic β -phase ($\text{Mg}_{17}\text{Al}_{12}$) is formed which has Young modulus of about 80GPa whereas for magnesium it is only 45 GPa. The microstructure of as-cast AZ91D is a typical dendritic structure with primary magnesium and eutectic bulk β -phase. The β - $\text{Mg}_{17}\text{Al}_{12}$ phases segregate mostly at the grain boundaries in the form of lamellar and bulk β -phase. The β -phase ($\text{Mg}_{17}\text{Al}_{12}$) precipitates at grain boundaries during deformation in the form of fine precipitates which are known to increase the strength by suppressing basal slip [53]. These precipitated β -phases which are perpendicular to the basal plan [53] (main slip system of α -magnesium) offer more resistance to the dislocations movement than the bulk β -phase adding the strength of the alloy [54].

In Mg-Al alloys, the poor thermal stability of $\text{Mg}_{17}\text{Al}_{12}$ phase (with a eutectic temperature of 437 °C) and its discontinuous precipitation can result in a substantial grain boundary sliding at elevated temperatures. The intermetallic such as $\text{Mg}_{17}\text{Al}_{12}$ located at the grain boundaries decompose at elevated temperatures cause an increase in the content of aluminium in these local regions. As a consequence, the mechanical properties are deteriorated in these regions [54]. Micro-cracks nucleate at the interface between β - $\text{Mg}_{17}\text{Al}_{12}$ and the magnesium-matrix and the reason reported for this is the bcc structure of β - $\text{Mg}_{17}\text{Al}_{12}$, which is not compatible with the hexagonal structure of the α -matrix [54]. This leads to a limited ductility of the alloy. The presence of precipitates also impedes the process of twinning in magnesium alloys [56].

2.4 Mechanical Behaviour of AZ91 and AZ31

The knowledge of quasi-static mechanical properties of magnesium alloys is well known but that of high strain rate behaviour is limited. In the past few years, most of the research related to magnesium alloys was focused to understand their microstructure evolution and their mechanical properties in the quasi-static and low strain rate regime. A limited research so far has been carried out on the high strain rate behaviour of magnesium alloys. Temperature dependence of their dynamic behaviour is even scarcely studied.

2.4.1 Strain Rate Effects

Significant work has been performed in characterization of magnesium alloys at quasi-static and low strain rate regime. However, so far little research has been carried out under dynamic strain rates. In general, the mechanical properties of magnesium alloys are strongly strain rate dependent. An increase in the flow stress followed by a decline with increasing strain rate in the range between 10^2 and 10^3s^{-1} is reported for AZ91 alloy by Sha et al. [54]. According to their study, the work hardening of α -matrix is responsible for increase in the flow stress, while a decrease in the flow stress is caused by the fracture of $\beta\text{-Mg}_{17}\text{Al}_{12}$. Aune et al. [57] studied the tensile behaviour of AZ91D, AM60B and AM50A at a strain rate in the range between 15 s^{-1} to 130 s^{-1} . The results showed that flow stress increases with increasing strain rate and the elongation at fracture is not affected by the change in the strain rate from 15 s^{-1} to 130 s^{-1} [57]. The strain rate also affects the energy absorption capabilities and the work hardening rate of magnesium alloys. Increase in flow stress with strain rate (10^3s^{-1}) for AZ91 was also confirmed by Ishikawa et al. (Fig. 2-8) [58].

Figure 2-8: Stress- strain curves for AZ91 at quasi-static and high strain rates [58]

AZ91 deforms by dislocation glide and twinning at high strain rates and microstructural changes affect the mechanical properties of the alloy. A fewer deformation “twins” in the specimens deformed at 10^3s^{-1} at room temperature were observed as compared to those deformed at 10^{-3}s^{-1} [58]. Liao et al. [59] studied the tensile properties of AZ91D and observed a slight variation in the yield strength, UTS and the failure strain as the strain rate is increased from 300s^{-1} and 1400s^{-1} . A slight increase in the yield and maximum strengths were reported for AZ91 at 1000s^{-1} as compared to quasi-static rates by Dudamell et al. [60]. The strain to failure under compression is not much affected by the strain rate increase. Tensile twinning proposed to be the operative mechanism in both compression and tension. However it is higher in compression than in tension [60]. Li *et al.* [61] and Mukai *et al.* [62] studied the effect of grain size on the dynamic behaviour of extruded magnesium alloys ZK60 and WE43, respectively. Li *et al.* [61] did compressive tests, while Mukai *et al.* [62] performed tensile experiments and both observed an improvement in the ductility at high strain rate as the average grain size is reduced. Li *et al.* [61] also showed that the

compressive ductility improves as the strain rate increases. This result is also confirmed for extruded AZ80 [63].

Strain rate and temperature strongly affects the behaviour of wrought magnesium alloy. The dynamic stresses are higher than the static ones (Fig. 2-9) but the rate sensitivity decreases with increasing strain rate [64]. The adiabatic heating at high strain rate is considered to be the responsible for softening of the alloys. However, the extent of rise in the specimen temperature at 10^3s^{-1} strain rate is not mentioned. Increase in the stress and decrease in the ductility of AZ31 at higher strain rates was observed by Ishikawa and Watanabe [65] and Tan et al. [66]. Significant strain hardening of AZ31 alloy was observed at a strain rate of 10^3s^{-1} [65]. Tan et al. [66] examined AZ31 in tension and compression at different strain rates between 10^{-3}s^{-1} and 10^3s^{-1} . Yang et al. [67] showed that the alloy AZ31B exhibits a clear anisotropy under compression at a of 1200s^{-1} .

Figure 2-9: Stress-strain curves for AZ31B-F [64] and AZ31 [66]

Slip and twinning are dominant mechanism in the deformation of AZ31 alloy. The dislocation creep was suggested the dominant deformation mechanism under compression for AZ31 at low strain rates while dislocation glide and twinning were observed at high strain rates [65-67]. Dislocation glide is the dominant

deformation mechanism in the low strain rate range while dislocation glide and twinning are important at high strain rates [65-66]. Yang et al. [67] suggested that twinning is the dominant deformation mechanism for sample tested for 90° oriented samples and basal and non-basal slip for samples tested 0° and 45° oriented samples.

Tucker et al. [68] performed compressive experiments on AZ31B-H24 plate in normal, rolling and transverse directions at high strain rates. They showed that in addition to strong anisotropy, the strain rate sensitivity of the alloy is also anisotropic as shown in Fig. 2-10. Strong rate dependence of compressive yield strength and ductility were observed in normal direction, while no strain rate dependence of yield strength in rolling and transverse direction is seen (Fig. 2-10). They also reported that as the strain rate increases, the strain to failure under compression increases in normal direction, whereas it slightly decreases in rolling and transverse directions [68].

Figure 2-10: Compressive stress-strain behaviour of AZ31B-H24 showing anisotropy and strain rate dependence [68]

Similar to Tucker [68], Wan et al. [69] also showed that stress-strain response of AZ31 alloy is highly anisotropic and strain rate sensitive. The flow stress

increases with increasing strain rate in normal direction but decreases in transverse and extrusion direction. Normal direction impact shows double twins and tensile twins while tension twinning and dislocation slip are exhibited in transverse direction impact (Fig. 2-11).

Figure 2-11: EBSD orientation micrograph showing twins [69]

The deformation mechanism of AZ31 alloy was investigated in the dynamic strain rate range by Suyuan [70]. The results showed that the dynamic compressive strength of the alloy is much higher than the quasi-static compressive strength and the strain rate has significant effect on the strain rate hardening of the alloy. The deformation is controlled by thermal activation of dislocations at quasi-static strain rate while at high strain rate, dislocation slip is responsible mechanism for deformation [70]

Ulacia *et al.* [71] investigated the dynamic behaviour of AZ31B-O sheets and its microstructural evolution. They performed both tension and compression tests at 10^{-3} s^{-1} and 10^3 s^{-1} strain rates and at different temperatures. The alloy showed anisotropy and tension-compression asymmetry at both quasi-static and higher rates [71]. Ulacia *et al.* [72] studied AZ31B magnesium alloy under tensile loading in rolling and transverse direction and observed that the flow stress and the yield

stress both increase with increasing strain rate. It was also observed that the alloy is anisotropic and the effect of direction is more significant at lower strain rates than at high strain rates. More deformation twins are reported at high strain rates than quasi-static strain rates [72]. No rate sensitivity of AZ31B-F was reported in the range of strain rate between $2.5 \times 10^{-4} \text{s}^{-1}$ and 650s^{-1} under compression while under tension, the stress increased as the strain rate was increased from 10^{-3}s^{-1} to 1375s^{-1} [73]. Energy absorption of the magnesium alloys increases with increasing strain rate. However, it is affected by orientation of loading in wrought alloys. The energy absorption in high strain rate compression of AZ31B increased with strain rate in normal loading direction, while in rolling and transverse direction no significant effect on the energy absorption was noticed with increasing strain rate [68]. A little work has been done regarding the constitutive study where the experimental data is fitted to some constitutive model in order to use in numerical simulation. Johnson-Cook model [100] was employed to study AZ91D under tensile loading by Aune [57] and to AZ31B under tensile loading by Ulicia [74]. The materials parameters for Johnson-Cook models found in above studies are listed in the Table 2-2 below. Ulicia studied AZ31B at 0.001, 100 and 1000s^{-1} strain rates in rolling and transverse direction and good agreement was seen between experimental and constitutive results. The results indicate that the rate sensitivity is nearly same in both loading direction while strain hardening is higher in case of rolling direction.

Table 2-2: Johnson-Cook parameters for AZ91D and AZ31B alloys

Paramete	σ_0 [MPa]	B [MPa]	n	C	
AZ91D[57]	129 (3)	616 (16)	0.5975 (0.0002)	.01-.03	
AZ31B [74]	180 (152-208)	345 (310-279)	0.554 (0.329-0.613)	0.1-0.013	Rolling
	221 (176-265)	206 (183-228)	0.370 (0.144-0.596)	0.1-0.013	Transverse

2.4.2 Temperature Effects

In general flow stress decreases and ductility of materials increases with increase in temperature and this behaviour is commonly attributed to the activation of additional slip systems. Deformation mechanism and fracture mode are affected by temperature. Isikawa and Watanabe [58] studied AZ91 at elevated temperatures. They showed that the flow stress significantly decreased at elevated temperature (573K) as compared to the room temperature and the effect of temperature is relatively small at high strain rate 10^3s^{-1} . Dislocation glide and twining are dominant deformation mechanism even at elevated temperatures [58]. Yield and maximum strength of AZ91 decrease with increase in temperature both at quasi-static and higher strain rate (10^3s^{-1}) [60] but the change is less pronounced at dynamic strain rates. Twining is suggested to be the dominant deformation mechanism in compression even at 400°C [60]. AZ31 magnesium alloy was studied by Ishikawa and Watanabe [65] under compression and by Tan et al. [66] under both compression and tension at different temperatures in the range between 296K and 723K. The dislocation glide and twining were observed at high strain rates even at elevated temperature [68-69]. Twins are rarely found in specimens deformed at 573K [68] and this is also confirmed by Tan et al. [66]. Magnesium alloy AZ31 was studied by Kun et al. [75] at elevated temperatures in the range between 473K and 723K at lower strain rates (0.01 to 10s^{-1}). They showed that the flow stress of the alloy decreases as the deformation temperature increases [75]. Ulacia *et al.* [71] observed a stronger influence of temperature on the stress-strain behaviour of AZ31B-O alloy at higher strain rates as compared to quasi-static rates. The alloy AZ31B-O showed a clear anisotropy and a strong tension-compression asymmetry even at elevated

temperature (up to 400°C) at high strain rates. However, at quasi-static rates tensile and compressive behaviour are similar for temperatures higher than 200°C [71]. They described this behaviour based on the activation of non-basal slip systems rather than twinning at elevated temperatures and at quasi-static rates, whereas twinning remains dominant at higher temperatures at dynamic strain rates [71]. In AZ31 magnesium alloy, the yield and flow stress both decrease noticeably with increasing temperature under compressive loading while decrease gently under tensile loading [72-73]. At quasi-static strain rate, twins are present at room temperature but not observed at 250°C. At high strain rates, increased twinning was observed over the entire range of temperature as can be seen in Fig. 2-12 [72]. The alloy showed a clear anisotropy even at elevated temperatures [72] and tensile-compressive asymmetry [73].

Figure 2-12: Microstructure of the fractured zones of samples tested at different temperatures and strain rate conditions [72]

Ulicia [72] used Johnson-Cook model to study the constitutive response of AZ31B in rolling and transverse directions under tensile loading at 1000s^{-1} strain rate and at 20°C, 150°C and 300°C temperatures. At room temperature, a difference is observed in the model predictions and experimental results and the reason suggested is that the Johnson-Cook model calculates only linear increase of flow

stress with increasing strain rate. Parameters needed by Johnson-Cook model are given in the Table 2-3.

Table 2-3: Johnson-Cook parameters for AZ31B [72]

Parameters	σ_0 [MPa]	B [MPa]	N	C	M
Rolling	224 (215-233)	380 (316-444)	0.628-0.894	0.011-0.014	1.554 (1.504-1.603)
Transverse	232 (214-249)	232 (206-258)	0.359-0.643	0.011-0.014	1.439 (1.4-1,477)

2.5 Objectives

The present research was primarily aimed to get sufficient experimental data in order to extend the structural use of AZ91D and AZ31B magnesium alloys in automotive and aerospace industries. The prime objective of the present study was to investigate the mechanical properties of AZ91D and AZ31B magnesium alloys in the wide range of strain rate and temperature. Understanding the operative mechanisms involved in the deformation of these two alloys and the association of microstructural features such as slip and twinning with the mechanical behaviour of these alloys were among the objectives of the present research.

2.6 Motivation

Lack of information and understanding about the dynamic properties of magnesium alloys add to the uncertainty and reservation about wider use of magnesium alloys. The present research work was aimed to investigate the mechanical behaviour of magnesium alloys AZ91D and AZ31B under wide range of strain rate and at below zero and elevated temperatures. The two alloys have been chosen as they are already being widely used in automotive and aerospace

industries; AZ91D is being used for various die-casting components and have potential to be used in chassis components, airframe etc. and AZ31B is suitable for the vehicle structure and body, fuselage of a plane and in many defence applications. The prime focus of this study was laid to investigate the rate and temperature dependence of the mechanical properties such as strength, hardening behaviour, rate sensitivity and energy absorption capability of the two alloys. It was intended to get sufficient experimental data which could lead to useful engineering models to describe the mechanical behaviour of these alloys and to foresee the response of vehicle components under crash events. In addition, the microscopic study will help to enhance the understanding of the microstructural phenomenon occurring during deformation and their correlation with the changes in mechanical properties of these alloys.

2.7 Missing Links in Literature

- Magnesium alloys AZ91D and AZ31B have not been investigated below zero temperature in compression and tension.
- Compressive mechanical properties of AZ91D have not been investigated at a strain rate higher than 1500s^{-1} at room temperature and at elevated temperature.
- Tensile properties of AZ91D at strain rate $> 10^3\text{s}^{-1}$ are not found in the literature.
- The mechanical behaviour of AZ31B under compression is little investigated at strain rates larger than 10^3s^{-1} . Few studies found at 2000s^{-1} and 4000s^{-1} did not explain the hardening behaviour and rate sensitivity of the alloy at all.

- Compressive properties of AZ31B at strain rate $> 10^3\text{s}^{-1}$ and at elevated temperatures are not found.
- A limited literature found on tensile properties of AZ31B at room temperature and elevated temperature did not explain adequately the deformation behaviour of the alloy.
- The strain rate and temperature dependence of the alloys mechanical properties has been less investigated and its contribution in characterizing the deformation behaviour of magnesium alloys has not been fully understood.
- Microstructure analysis of AZ91D and AZ31B alloys at higher strain rates between 1000s^{-1} and 4000s^{-1} and at temperature ranging between -30°C and 250°C are not found in the literature.
- Limited study is found for the constitutive analysis of these two alloys under tensile loading. No study is found under compressive loadings.

2.8 Scope of Study

In the current study, AZ91D and AZ31B alloys were tested at several strain rates and temperatures under compression and tension. Uniaxial compression tests at quasi-static strain rate of 10^{-4}s^{-1} and at dynamic strain rate ranging between 500s^{-1} and 3500s^{-1} were performed. Tensile tests were performed at 10^{-4}s^{-1} and in the range of strain rate between 300s^{-1} and 1500s^{-1} on both alloys. To see the temperature effects, compression tests were performed at -30°C , 25°C and 200°C temperatures at strain rates of 10^{-4}s^{-1} , 1000s^{-1} and 2200 to 2500s^{-1} . Tensile tests were performed only at 250°C at strain rates of 10^{-4}s^{-1} and about 1500s^{-1} . Quasi-static tests were carried out using uniaxial compression and tension machine, while Split Hopkinson Pressure Bar (SHPB) and Split Hopkinson Tensile Bar

(SHTB) were used to perform high rate tests. Below zero temperature (-30°C) tests were carried using dry ice and elevated temperature tests were by using coil heater. Compression specimens with length to diameter ratio of 0.5 to 0.6 and bar to specimen diameter ratio D_b/D_s between 0.6 and 0.8 were found optimal [82] and used in the present study. For tensile specimen l/d ratio of 2.5 and higher is found appropriate for large strain tensile testing. A Matlab filter is developed to remove the noise from the data that appears in the form of oscillations in the resultant curves.

Chapter 3

Methodology

The strain rate dependence of AZ91D and AZ31B magnesium alloys was studied by performing quasi-static and high rate uniaxial compression and tensile experiments. Temperature dependence of the two alloys was also investigated under compression loading at elevated temperature.

3.1 Quasi-static Experimental Setup

Quasi-static compression and tensile tests at -30°C , 25°C and 200°C were performed using INSTRON 5569 machine with maximum load 50 KN and maximum displacement 50 mm.

Figure 3-1: INSTRON machine for quasi-static strain rate experiments

Below zero temperature of -30°C was obtained using dry ice around the chuck of the machine and the temperature was maintained with the help of thermocouple and digital meter with an accuracy of $\pm 2^{\circ}\text{C}$. High temperature compression tests at 200°C were performed using the built-in facility of the above machine. A close

chamber was heated with built in heaters and the temperature in the chamber was set and controlled by the temperature control unit with an accuracy of nearly $\pm 1^{\circ}\text{C}$. The load cell used on this apparatus has a capacity of 50 kN. Specimen displacement was measured using video extensometer manufactured by INSTRON. Specimens were mounted using v-grips to align the specimen with loading axis to minimize the likelihood of any bending during loading. The required strain rate was set as operating parameter instead of cross head speed.

3.2 The Split Hopkinson Bar

The Split Hopkinson Bar is quite popular and is a widely used technique to analyse the materials behaviour under dynamic loadings. In 1914, Bertram Hopkinson [76] developed a simple technique to measure the pressure generated during impact which lead scientists to develop current modern versions of Hopkinson bar. In the 1949, Kolsky [77] modified the Hopkinson's original design and the apparatus called, "Split Hopkinson pressure bar" became one of the widely accepted methods for testing materials at dynamic strain rates.

The split Hopkinson Pressure Bar (SHPB) also known as Kolsky,s apparatus [77] was used to investigate the high strain rate compressive properties of magnesium alloys. The apparatus consists of a striker bar, an input bar, an output bar and an absorber bar at the end. The cylindrical specimen was placed between the input and the output bar and a gas gun was used to drive the striking bar. Two diametrically opposing strain gauges were mounted at the centre on each of incident and transmitting bar to monitor the strain on these bars. A schematic diagram is shown in Fig. 3-2.

Figure 3-2: A schematic diagram of the Split Hopkinson Pressure Bar (SHPB) setup

Measurement of the strain pulses in two elastic bars on either side of the specimen was used to quantify the mechanical behaviour of the specimen material. In the compressive split Hopkinson bar used in the current study, a gas gun was used to drive the striker bar onto the incident bar. The working of the compressive Hopkinson bar can be explained as that upon impact a rectangular elastic pulse is generated that travels along the incident bar until it reaches the specimen. The time dependent incident strain, $\epsilon_I(t)$, is measured at a strain gauge located on the incident bar. Upon reaching the specimen the pulse partially reflects back towards the impact end. This reflected portion, represented by $\epsilon_R(t)$, is also recorded by the same strain gauge on the incident bar. The remaining portion of the incident wave that transmits through the specimen into the transmitter bar is recorded by the strain gauge mounted on the transmitter bar as transmitted pulse $\epsilon_T(t)$. It is shown ahead that theoretically reflected and transmitted waves are proportional to the specimen's strain rate and stress respectively. Specimen strain can be determined by integrating the strain rate.

Split Hopkinson Tensile Bar (SHTB) was used to investigate the tensile properties at high strain rates. The apparatus consists of one input and one output bar, a striker tube, an anvil bar, and an absorber bar at the end. Four strain gauges, two on input and two on output bar, are mounted 500 mm away from the specimen and 180° apart on the same bar to record the strain when impacted. A dog bone shaped specimen with threaded flanges is screwed between the input bar and the output bar. A gas gun drives the striking tube. A schematic diagram of a tensile Hopkinson bar is shown in Fig. 3-3

Figure 3-3: A schematic diagram of the Split Hopkinson Tensile Bar (SHTB) setup

The principle of Hopkinson Bar in tension is similar to that in compression except the way to generate the tensile pulse, specimen geometry and specimen fixing to bars. Upon impact of the striker tube with the anvil bar which is attached to the input bar, a compressive wave is generated in the anvil bar and a tensile wave is generated in the input bar. The compressive wave propagates through the anvil bar and is absorbed in the absorber bar. The tensile wave, also known as the incident wave, will propagate along the incident bar towards the specimen. It will be measured by the strain gauges on the incident bar as $\epsilon_I(t)$. As in Split Hopkinson pressure bar, the reflected wave, $\epsilon_R(t)$, is captured by the strain gauges on the incident bar and the transmitted wave, $\epsilon_T(t)$, is measured by the strain gauges on the transmitter bar. The time-dependent incident strain $\epsilon_I(t)$,

reflected strain $\varepsilon_R(t)$, and transmitted strain $\varepsilon_T(t)$ are recorded using a digital Oscilloscope. A typical Oscilloscope trace obtained from Hopkinson bar is shown in Fig. 3-4.

Figure 3-4: A typical Oscilloscope trace obtained from Split Hopkinson Bar

When the specimen deforms uniformly, the three strain pulses have the relation [78] as shown in Eq. (3.1)

$$\varepsilon_I(t) + \varepsilon_R(t) = \varepsilon_T(t) \quad (3.1)$$

The dynamic stress $\sigma(t)$, strain $\varepsilon(t)$ and strain rate $\dot{\varepsilon}(t)$ are given by one-dimensional stress-wave theory as in Eq. (3.2) [78].

$$\left. \begin{aligned} \sigma(t) &= E \frac{A_o}{A_s} \varepsilon_T(t) \\ \varepsilon_s(t) &= -\frac{2C_o}{l_s} \int_0^t \varepsilon_R(t) dt \\ \dot{\varepsilon} &= \frac{d\varepsilon_s}{dt} = -\frac{2C_o}{l_s} \varepsilon_R \end{aligned} \right\} \quad (3.2)$$

Where E is the Young's modulus of the pressure bars, A_o is the cross section area of the bars, A_s is the initial cross-sectional area of the specimen and l_s is the initial

length of the specimen. C_o is the sound velocity in the bars and is given by the following Eq. (3.3).

$$C_o = \sqrt{E/\rho} \quad (3.3)$$

In the above equation, E is Young's modulus and ρ is the density of the bar material. The detailed evaluation of Hopkinson bar formulae can be found easily in the literature [76- 79].

3.3 Problems associated with Hopkinson Bars

There are some issues associated with Hopkinson bars needed to be tackled carefully for reliable results. The general theory that governed the Hopkinson testing technique for decades has been well established now. Therefore in recent years, areas like the effects of the specimen geometry, wave dispersion in the bars, friction and specimen inertia effects, pulse shaping techniques, characterizing new materials etc. had been the focus of researchers. These problems need to be tackled carefully for accurate testing and reliable results.

3.3.1 Specimen Geometry Effect

Several researchers found specimen geometry to be very influential and specimen gauge length, specimen diameter and specimen length to diameter ratio (l/d) are widely investigated. Yokoyama and Mayama [80] found that the optimum specimen geometry is a solid cylinder with l/d ratio between 0.50 and 0.75 when diameter ratio d/D between specimen and the bar is 0.5. Gunsekera, Havranek and Littlejohn [81] in investigating the size effect found that an increase in the specimen diameter tends to produce more barrelling and thus giving an overestimation of deformation stresses. High strain rates are obtained

for smaller diameter specimens but at the cost of larger inaccuracies (oscillations) in stress-strain curves [82].

In testing with tensile Hopkinson bar, mostly the dog-bone shaped specimens are used with widely divergent dimensions. In longer specimens, most of the specimen length is in a state of uniaxial tension and the edge effects are negligible. In short gauge length specimen, most or whole of the specimen length is affected by the boundary conditions at the edges which introduce an error in experimentally measured stresses. However, no effect of l/d ratio is observed for l/d greater than 1.60 [83]. Verleysen et al. [84] suggested that “the specimen deformation is more accurate if the length-to-width ratio of the specimen is larger than 1.25 and radius of transition zone is sufficiently small”.

3.3.2 Dispersion of Waves

Hopkinson bar equations are based on the assumption that one-dimensional wave propagation occurs within the bars. However due to non-uniformity at the impact interfaces, the longitudinal waves in bars suffer dispersion that affect the strain pulses, as a result the measured stress-strain behaviour is affected. Various techniques are employed to minimize the dispersion effects from the strain pulses. Efforts have been made by several authors to overcome this problem. Smaller diameter bars and an increase in the pulse rise time [85], elimination of high-frequency components by using low pass filter [84] and are few techniques to minimize the dispersion effects. Introduction of a deformable metallic disc (pulse shaper) sandwiched between the striker and the input bar [85, 87-88] helps to attenuate the high-frequency oscillations resulting from the impact thus producing a smooth incident pulse.

Friction at specimen-bar interfaces in compression testing can be minimized by optimum specimen size: specimens with l/d ratio of 0.5 [89] while Bertholf and Karnes [90] found the reduced friction effects for specimen l/d ratio between 0.5 and 1 and proper lubrication [91]. Friction effects can be minimized by careful selection of specimen geometry and proper lubrication at the specimen/bar interfaces.

3.4 Experiments at -30°C, 25°C and 200°C Temperature

High strain rate tests below zero temperature were performed using dry ice around the bar-specimen assembly as shown in Fig. 3-5. A thermocouple was used to monitor the temperature of the specimen, as soon as the specimen reaches the required temperature the thermocouple is detached and tests are performed immediately to avoid any change in temperature. With this arrangement it was possible to achieve the accuracy of $\pm 2^\circ\text{C}$.

Figure 3-5: Arrangement for below zero temperature high strain rate experiments

Figure 3-6: Arrangement for elevated temperature high strain rate experiments

Elevated temperature tests were performed using a coil heater around the specimen as shown in Fig. 3-6. Using a coil heater of length a little longer (10 mm) than the specimen length has an advantage to avoid the excessive heating of input and output bars. The effect of heating on the strain gauges was also taken into account and no effect was found on the strain gauges at all. At a distance of about 100 mm from the specimen-bar end the temperature of the bars was found to be at room temperature throughout the test as shown in Fig. 3-7. It ensures that the strain gauges mounted at 500 mm from specimen-bar end on each bar were not affected (Equations and properties used in these calculations are given in Appendix D3).

Figure 3-7: Temperature variation along the bars

3.5 Experimental Plan

Compression tests for AZ31B alloy were performed in normal (N), rolling (R), transverse (T) direction as well as at an angle (45°) to the rolling direction. Tensile tests for AZ31B alloy were performed in rolling (R) and transverse direction (T). A complete test matrix can be seen in Table 3-1, Table 3-2 and Table 3-3 for compression at room temperature, tensile at room temperature and

compression at various temperatures respectively. Elevated temperature tests were also performed for AZ91D and AZ31B as given in Table 3-4. For repeatability of the results, four to five tests were conducted for each condition.

Table 3-1: Matrix of experiments performed under compression at room temperature

Material	Average tested strain rate [s^{-1}]							
	10^{-4}	500	1000	1500	2000	2500	3000	3500
AZ91D	X	x	x	x	x	x	x	
AZ31B (45°)	X	x	x	x		x		x
AZ31B (N)	X	x	x	x		x		x
AZ31B (R)	X	x	x	x		x		x
AZ31B (T)	X	x	x	x		x		x

(x) shows that the tests were performed under this strain rate or temperature

Table 3-2 Matrix of experiments performed under tension at room temperature

Material	Average tested strain rate [s^{-1}]						
	10^{-4}	350	500	750	1000	1400	1500
AZ91D	x	x		x			X
AZ31B (R)	x		x		x	x	
AZ31B (T)	x		x		x	x	

Table 3-3 Matrix of experiments performed under compression at various temperatures

Material	Average tested strain rate [s^{-1}]							
	10^{-4}		1000		2250		2500	
	-30°C	200-250°C	-30°C	200-250°C	-30°C	200-250°C	-30°C	200°C
AZ91D	X	x	x	x			x	X
AZ31B (45°)	X	x	x	x	x	x		
AZ31B (N)	X	x	x	x	x	x		
AZ31B (R)	X	x	x	x	x	x		
AZ31B (T)	X	x	x	x	x	x		

Table 3-4: Tensile tests at elevated temperature

Material	Average tested strain rate [s^{-1}]	
	10^{-4}	1500
	250°C	250°C
AZ91D	X	x
AZ31B	X	x
AZ31B	X	x

3.6 Materials and Specimen Geometry

3.6.1 Materials Compositions

Two types of magnesium alloys, as cast AZ91D and rolled AZ31B were characterized as a part of the current study. The densities of about 7 to 8 specimens of each alloy were measured using displacement method (Archimedes Principal) and the following densities were found with lower and upper bounds.

$$\text{AZ91D } 1.804_{+0.009}^{-0.010}$$

$$\text{AZ31B } 1.749_{+0.052}^{-0.055}$$

The chemical compositions of the two alloys are given in Table 3-5.

Table 3-5: Chemical composition of tested magnesium alloys

Alloy	Al %	Mn %	Zn %	Si %	Cu %	Ni %	Fe %	Mg %
AZ91D	8.5-9.5	0.17-0.3	0.35 -1	0.05	0.025	0.002	0.004	balance
AZ31B	2.5-3	0.3-1	0.7-1.3	-	-	-		balance

3.6.2 Quasi-static Test Specimen Geometry

In the present study the specimen geometries for static compression tests were manufactured according to the ASTM standards E9-89a [92]. The dimensions of the cylindrical specimen for quasi-static compression tests for AZ91D and AZ31B are shown in Table 3-6. The dog bone specimens for tensile static test were manufactured according to the ASTM standard B 557M-07 [93] with gauge length of 30.0 mm and diameter of 6.0 mm as shown in Fig. 3-8

Table 3-6: Specimen dimensions for quasi-static compression test [92]

Alloy	Height L(mm)	Diameter D (mm)	Length to diameter ratio
AZ91D	25.00 ± 1.0	12.50 ± 0.20	2.0
AZ31B	20.00 ± 1.0	10.00 ± 0.20	2.0

Nominal Diameter	Dimensions, mm			
	Standard Specimen		Small-Size Specimens Proportional to Standard	
	12.5	9	6	4
G—gage length	62.50 ± 0.10	45.00 ± 0.09	30.00 ± 0.06	20.00 ± 0.04
D—Diameter (Note 1)	12.50 ± 0.25	9.00 ± 0.10	6.00 ± 0.10	4.00 ± 0.05
R—Radius of fillet, min	9	8	6	4
A—Length of reduced section, min (Note 2)	75	54	36	24

Figure 3-8: Standard tensile specimen B557M-07 [93]

3.6.3 High Strain Rate Test Specimen Geometry

Optimum specimen geometries [82] for compression and tensile tests were chosen to minimize the size effects on the behaviour of material. The diameter and height of the cylindrical specimens used for high strain rate compression tests are 8.0 ± 0.10 and 4.0 ± 0.05 respectively for both AZ91D and AZ31B alloys. All compression specimens were machined using Wire Electrical Discharge Machining (EDM) followed by the fine grinding of the specimen ends to make them flat and parallel for perfect contact with the bar ends. The specimen geometry for high rate tensile tests was chosen with a minimum length to diameter ratio of 2.5 and detailed dimensions are shown in Fig. 3-9. The tensile specimen geometry used was modified (a small portion of 1.5 mm was kept without threads starting from the edge of transition zone) in order to get a better tight fit of specimen into the bars, a loose fit will introduce some noise in the data.

Figure 3-9: High strain rate tensile specimen used in present study

In case of cast alloy AZ91D to see the effect of homogeneity of the material in the ingot; few tests were carried out for specimens manufactured from side G1 and middle G2 and from the centre along the ingot length G3. Also specimens were manufactured from the top T, middle M and the bottom of all three positions as shown in the Fig. 3.10.

Figure 3-10: Orientation of specimens from AZ91D ingot

Specimen from AZ31B rolled plate (L=12 inches, W=12inches, T=0.5 inches) for compression and tensile tests were prepared along normal, rolling and transverse direction and also at 45° angle to rolling direction as shown in Fig. 3-11. Initially rectangular bars were cut from the plate and then specimens were prepared by EDM wire cutting. Tensile specimens were machined to the final shape as shown in Fig. 3-11.

Figure 3-11: Schematic of AZ31B specimens manufactured from rolled plate

3.6.4 Specimen Preparation for Microstructural Analysis

For microstructural analysis, tested specimens were cut, ground, polished and etched for analysing with optical microscope and scanning electron microscope (SEM). Each specimen was cut into two halves. The surface that is parallel to the impact direction as shown in Fig. 3-12 was ground using abrasive paper of 800, 1200 and 2400 grit, polished using 1 micron diamond paste and etched.

Figure 3-12: Cutting and polishing scheme of compressive specimen for SEM

3.7 Processing of Experimental Data

3.7.1 Data Acquisition

The testing output from Hopkinson bars is in the form of voltage histories from strain gauges on incident and transmitter bars which by simple formulation are converted into the stress, strain and strain rate of the specimen. The signals from strain gauges are measured and amplified by dynamic strain meters. The signals from dynamic strain meters are input to digital oscilloscope where these are plotted in the form of voltage and time. The incident strain pulse and reflected strain pulse are recorded and plotted on same channel (channel 1) and the transmitted strain pulse is recorded on channel 2. The output from the oscilloscope are opened in Excel in the form of two column data, the first column presenting incident and reflected voltage signals recorded on channel 1 and the

second column presents the data recorded on channel 2. These voltage signals are converted into incident strain, reflected strain and transmitted strain pulses which are finally transformed into specimen strain rate, strain and stress using the Hopkinson bar equations Eq. 3.25. A step by step procedure for calculating the specimen stress, strain and strain rate from the oscilloscope output is explained in Appendix E1.

3.7.2 Experimental Strain Rate Measurements

Strain rate in high strain rate experiments is needed to be averaged as a constant strain rate is not achieved. In high rate tests, uniform strain rate is achieved after a certain time known as rise time. In present study, the rise time is about $25\mu\text{s}$ and $30\mu\text{s}$ for 2000s^{-1} and 3000s^{-1} strain rate respectively as can be seen in Fig. 3-13. After the rise time, the strain rate is uniform although a constant value of strain rate is not achieved. Therefore, the strain rate is calculated by taking the average of the plateau region (A-B) of the strain rate-time curve as shown in Fig. 3-13. Strain rates shown throughout this research are calculated in the similar way and taken as average strain rate of the plateau region of the curve.

Figure 3-13: Average strain rate calculations during high strain rate testing

3.7.3 Noise Filtering from High Strain Rate Experimental Data

High strain rate data is usually accompanied by some noise which should be separated from the data for accurate analysis. Noise in experimental data can be introduced from electronics used to capture the data. A Matlab code is written to filter the noise appearing in the data. This software given in Appendix E2 is based on Butterworth filter; it reads the input data from excel.xls file filter it and writes the output filtered data in the same file. Fig. 3-14 shows the original data along with the filtered data to show the noise removal from the data.

Figure 3-14: Experimental data: original and filtered data

3.8 Verification of Force equilibrium in the specimen

In Hopkinson Bars, a specimen should be under uniform stress during deformation, that is, the force on the two ends of the specimen given by equation Eq. (3.4) and Eq. (3.5) should be equal.

$$F_1(t) = EA_o[\varepsilon_I(t) + \varepsilon_R(t)] \quad (3.4)$$

$$F_2(t) = EA_o \varepsilon_T(t) \quad (3.5)$$

F_1 and F_2 (Eq. 3.4 and Eq. 3.5) are the forces at specimen-incident bar interface and specimen-transmitter bar interface respectively. Fig. 3-15 shows the force as

a function of strain at specimen-incident bar interface (F_1) and specimen-transmitter bar interface and (F_2). The overlapping of the two forces ensures that the specimen is in dynamic equilibrium. The force equilibrium was checked in each experiment and the difference of forces (F_1 and F_2) was found to be about than 5%. However, it is observed that the force equilibrium at high strain rates is established a little late; after 2 to 3 % strain.

Figure 3-15: Force equilibrium on the two faces of the specimen/bar interfaces

3.9 Pulse Shaping

Pulse shaper is placed between the striker and incident bar. This enables the rise time of incident pulse to increase and achieve the dynamic stress equilibrium. From SHPB experiment without pulse shaper, the incident pulse is flat topped, with short rise time, affecting the response of the reflected pulse. As the strain rate is dependent on the shape of the reflected pulse, the strain rate will be varying appreciably during the test. The initial high strain rate is undesirable as the strain rate history effects could hide any true strain rate dependence. A pulse

shaper placed between the striker and the incident bar deforms plastically and spreads the pulse in the incident bar upon impact. An increase in the rise time enables the specimen to achieve dynamic stress equilibrium and to deform at nearly constant strain rates in the elastic and early yield response regions. Pulse shaper of different materials such as copper [94-95], polymer [96] and steel [97] in thickness ranging between 0.1 mm to 2 mm. Wu et al. [98] utilized a paper as a pulse shaper to reduce high-frequency wave dispersion. In the current study, aluminium and copper discs of 0.5, 1, 1.5 and 2.5 mm thickness and 12 mm diameter were used as pulse shaper. Various thicknesses were used as the thickness of the pulse shaper affects the rise time and the magnitude of the incident pulse [99]. Pulse shaper of 0.5 mm thickness made a very slight difference. The pulse shaper of 2.5 mm thickness increased the rise time but at the expense of incident pulse magnitude thus lowering the impact significantly. However, results of 1 and 1.5 mm thick pulse shaper for both aluminium and copper were close and optimum in respect of increase in rise time and magnitude of the incident pulse. As well as 4 layers and 8 layers of A4 (80 gm.) paper were used as pulse shaper for both AZ91D and AZ31B alloys. Impact velocity in all these tests was 17 ± 0.25 m/s. The results for incident strain pulses for AZ91D and AZ31B are shown in Fig. 3-16. The 4 layers A4 paper pulse shaper was not effective at higher impact velocity ($V \geq 15$ m/s). The 8 layers paper pulse shaper seems to be useful and have effectively increased the rise time of the incident pulse resulting in a uniform strain rate. For entire study of this research, 8 layers of A4 paper were used as pulse shaper for all strain rates. They were cheaper and easy to prepare and effective as well.

Figure 3-16: Incident strain pulses from oscilloscope for different pulse shapers ($V \sim 17 \text{ m/s}$)

3.10 Constitutive Analysis

In order that the experimental results can be utilized in simulations to optimize the processing, design, response of components in crash etc. the experimental data must be evaluated. The most common and easy approach is to use accepted constitutive laws. The Johnson Cook model [100] is commonly used because of simplicity which is essential to save computational resources in large-scale simulations such as car crash. It takes into account the effect of strain, strain rate and temperature on the constitutive behaviour of the material and can be used over a wide range of temperature and strain rate. The mathematical formulation of Johnson and Cook model is represented by the following equation,

$$\sigma = (A + B\varepsilon^n)(1 + C \ln(\dot{\varepsilon}/\dot{\varepsilon}_0))[1 + (T^*)^m] \quad (3.4)$$

Where

σ : True or effective stress

n : Work hardening exponent

ε : Effective plastic strain

$\dot{\varepsilon}$: Effective plastic strain rate and

$\dot{\varepsilon}_0$: Reference strain rate usually taken as 1 s^{-1} .

A , B and C are constants which describes the work hardening behaviour where A is yield stress, B is increase in strength due to work hardening and C is the strain rate sensitivity of the material

T^* : Homologous temperature = $(T-T_0)/(T_m-T_0)$

T_0 is the room temperature, T_m is melting temperature, m is thermal softening exponent. In Eq. (3.4), A , B and n are the three coefficients describing the quasi-static behaviour of the material. The terms within the first set of parentheses in the model equation is a strain hardening term that describes a power law relationship between true stress and effective plastic strain. The second term in parentheses is strain rate hardening term that introduces a logarithmic dependence of material behaviour on strain rate. The third term explains the effect of testing temperature. Johnson and Cook model explained in the equation comprises of four experimentally determined parameters (A , B , C and n).

A large number of studies have shown that numerical simulation employing Johnson-Cook model can produce results of sufficient accuracy for many engineering purposes. Johnson and Cook model assigns a power law hardening behaviour to the material and scales that behaviour up and down depending upon the strain rate [101]. Johnson and Cook model has some limitations as well. The strain and strain rate hardening phenomena are uncoupled that makes it to work at moderate strain rate level. The strain rate hardening increases as the strain rate increases therefore Johnson-Cook model cannot be fit with same power law at all strain rates [101]. Johnson and Cook model is independent of

pressure therefore the predicted yield stress is the same in compression and tension. These should be taken into account when analysing the results.

In the present study, the constitutive parameters A , B , n , C and m required by the model (Eq. (3.4)) are predicted by fitting Johnson-Cook equation (Eq. (3.4)) to the data obtained by experimental trials using a built-in curve fitting MATLAB tool based on nonlinear regression analysis. At high strain rates two temperature were chosen for the fitting: 25°C and 200°C in compression and 25°C and 250°C in tension to account for the adiabatic heat rise of the alloys due to plastic work performed. To determine the Johnson-Cook material constants, the third term explaining the temperature effects is dropped for the room temperature tests and Eq. (3.4) takes the form as Eq. (3.5).

$$\sigma = (A + B\varepsilon^n)(1 + C \ln(\dot{\varepsilon}/\dot{\varepsilon}_0)) \quad (3.5)$$

A , B and n are obtained from tests conducted at low strain rates ($\dot{\varepsilon} \approx 1 \text{ s}^{-1}$), for which the second term is equal to 1 and the Eq. (3.5) is reduced to the following equation.

$$\sigma = (A + B\varepsilon^n) \quad (3.6)$$

The predicted values of A , B and n from the fitted data are then placed in Eq. (3.5) and the only unknown in the equation the strain rate sensitivity C is determined by fitting Eq. (3.5) to the stress-strain data from tests at high strain rates (Hopkinson bar). Similarly the value of m is found out by using the experimental data at elevated temperatures.

3.11 Microstructural Analysis

Microstructure analysis was performed on some selected specimens for both compressive and tensile loadings at quasi-static and high strain rates. Three different types of etchants were used,

1. Etchant-1 picric solution (4.2 g picric acid +10 ml ascetic acid + 70 ml ethanol and 10 ml water)
2. Etchant-2 (4.5 g oxalic acid + 100 ml water)
3. Etchant-3 (1g oxalic acid + 1ml nitric acid + 1ml ascetic acid + 150ml water)

The etchant-1 and etchant 2 were used for AZ91D and etchant-1 and etchant 3 for AZ31B alloy. The etchant time was calculated and the best time was found to be the 5-6 second for AZ91D and 10-15 seconds for AZ13B. The specimen of AZ31B was also gold sprayed for better conductivity to get clear image. Secondary electron method was used to observe the microstructure and the EDX was also performed to estimate the composition of the alloy in different phase.

Chapter 4

Results and Discussion (AZ91D)

In this chapter, the experimental results of as cast AZ91D alloy under compression and tension are presented. The homogeneity of the alloy throughout the ingot was first investigated, followed by the study of strain rate and temperature effects on the mechanical behaviour of AZ91D. Experimental results and microstructural analysis are presented and discussed.

4.1 Material homogeneity throughout the ingot

In cast alloys, ingots or blocks available commercially are usually half a meter long and 4 to 5 inches in height and width. These blocks or ingots are considered to have some non-homogeneity in composition, grain size and micro-porosity (Reasons for this non-homogeneity is not in the scope of the present study) which affects the mechanical properties of the alloy. In the current study, specimens were prepared from different parts of a block as shown in Fig. 3-12 to check the extent of non-homogeneity and its effects on the properties of the alloy. Tests were performed to observe non-homogeneity in the cast AZ91D ingot according to the scheme shown in Fig. 3-12. Stress-strain curves for specimens from three different column positions of the alloy block G_1 , G_2 and G_3 are shown in Fig. 4-1 (a) and from top, middle and bottom of the same column G_2 are shown in Fig. 4-1 (b).

Figure 4-1: Stress-strain relationship (a) different location in ingot (b) top, middle and bottom of same location

Form these figures; it was observed that the specimens prepared from several locations of the ingot did not show significant variation in stress-strain behaviour. The results from different locations are within 3% of each other in the stress level for group G_2 and G_3 and less than 2% for specimens taken from the top as compared to the specimen produced from middle and the bottom of same column. The fracture strain showed little variation for specimens made from three locations. Hence it was assumed that the alloy ingot is sufficiently homogeneous ensuring that the specimens prepared from various locations of the ingot will have similar properties under same testing conditions. Nevertheless, all specimens used in the present study were manufactured from the middle portion of columns parallel to G_1 direction to eliminate the effects of non-homogeneity.

4.2 Compressive Behaviour

Tests were carried out at several strain rates between $10^{-4}s^{-1}$ and $3000s^{-1}$ and temperatures between $-30^{\circ}C$ and $200^{\circ}C$. Results are presented and discussed in the following sections.

4.2.1 Strain Rate and Temperature Effects

Experimental strain rates obtained in compressive loading of AZ91D are plotted against time in Fig. 4-2. Strain rates in the beginning increase sharply, plateau for a long range, then rapidly decreases to zero. The gradual decrease observed in the plateau region of the strain rate vs. time curve is usually due to the hardening of the alloy under compression. As explained earlier, the values shown in the figure are average strain rate values of the flat portion of the curve.

Figure 4-2: Average compressive strain rate measured during experiments

The strain rate effect on the compressive behaviour of AZ91D at room temperature is shown in Fig. 4-3. The stress level increases with increasing strain rate. For a fixed strain of 0.1, an increase of about 53% and 25% in stress is noticed as the strain rate is increased from 10^{-4} s^{-1} to 1500 s^{-1} and from 1500 s^{-1} to 3000 s^{-1} respectively. Over the whole range of dynamic strain rate considered, yield strength (YS, stress at 0.2% offset), peak compressive strength (PCS, the maximum stress as shown in Fig 4-3 in the σ - ϵ curve beyond which the stress starts decreasing either due to the softening overcomes the hardening or some cracks begins to develop leading to failure, PCS for 3000 s^{-1} is about 450MPa) and the

strain to failure all show a monotonic increasing behaviour as seen in Fig. 4-3. It is also noted that at a strain rate of 1500s^{-1} and above, the stress-strain curves become flat after a certain strain, indicating a mitigating effect of hardening and softening of the alloy during deformation.

Figure 4-3: Stress-strain relationship for AZ91D at room temperature

The compressive yield strength (CYS) and the peak compressive strength are plotted as a function of strain rate in Fig. 4.4. Both YS and PCS increase with strain rate. The yield strength shows a significant increase ($\sim 30\%$) when the strain rate changes from 10^{-4}s^{-1} to 1000s^{-1} while it undergoes lesser increase ($\sim 10\%$) as the strain rate increases further from 1000s^{-1} to 3000s^{-1} . On the other hand, PCS increases by 40% as the strain rate increases from 10^{-4}s^{-1} to 3000s^{-1} .

Figure 4-4: Yield stress and peak compressive stress as a function of strain rate

Fig. 4-5 shows experimental stress-strain curves under compression at few temperatures ranging between -30°C (243K) and 200°C (473K) at quasi-static rate. Specimens of 8mm diameter and 4mm length were heated using coil heater as explained in chapter 3. In each experiment, test was performed immediately after the temperature of the specimen reached the desired value to avoid any decrease in the temperature. From experimental results shown in Fig.4-5, it is observed that the stress level decreases as the temperature increases. In the plastic deformation region, the stress for any fix strain decreases by a little (i.e. decreases by 4% at 12.5% strain) as the temperature changes from -30°C to 25°C whereby it decreases significantly (i.e. decreases by 45% at 12.5% strain) for further increase in temperature up to 200°C . Thermal softening strongly affects the deformation at 200°C and responsible for considerable decrease in the stress. However, the plastic strain is sufficiently increased though the specimen is not broken at 200°C .

Figure 4-5: Temperature effect on the mechanical behaviour of AZ91D at 10^{-4} s^{-1}

Fig. 4-6 and Fig. 4-7 show experimental stress-strain curves at various temperatures ranging between -30°C (243K), 25°C (298K) and 200°C (473K) at 1000 s^{-1} and 2500 s^{-1} strain rates to analyse the combined effect of temperature and strain rate. At 1000 s^{-1} , for increase in temperature from -30°C to 200°C the decrease in stress level is approximately 27% and 28% at 3% and 9% strain respectively. While for 2500 s^{-1} , this decrease is about 32% and 30% at 3% and 9% strain. The stress variation with strain tells about the different hardening rates of the alloy at different temperatures. In general, the yield strength and peak compressive strength (PCS) decrease with increasing temperature at all strain rates. For increase in temperature from -30°C to 200°C , the yield strength undergoes a decrease of about 28-30%, at 10^{-4} s^{-1} , 1000 s^{-1} and 2500 s^{-1} . However, the peak compressive strength is decreased by 45%, 15% and 5% at 10^{-4} s^{-1} , 1000 s^{-1} and 2500 s^{-1} respectively indicating a combined effect of strain rate and temperature on the peak compressive strength.

Figure 4-6: Temperature effect on the mechanical behaviour of AZ91D at 1000s^{-1}

Figure 4-7: Temperature effect on the mechanical behaviour of AZ91D at 2500s^{-1}

4.2.2 Work Hardening

Usually metals strain harden and get strengthen upon loading. That is their load bearing capacity increases during deformation. The measure of the ability of metals to get strengthen is known as the hardening exponent n and is calculated by Eq. (4.1)

$$n = d\log(\sigma_t)/d\log(\epsilon_t), \quad (4.1)$$

where σ_t is true stress and ε_t is true strain. Graphically it is the slope of the line fitted to the true stress-true strain data on log-log scale as shown in Fig. 4-8. The Fig. 4-8 illustrates the work hardening behaviour of the alloy at various strain rates. A linear relationship exists between $\log(\text{stress})$ and $\log(\text{strain})$ within a certain region. The hardening exponent n increases slightly with strain rate with an average (averaged over strain rate between 1000s^{-1} and 3000s^{-1}) value of 0.52 ± 0.02 . It is also observed that the hardening exponent n decreases with strain, it reduces to 0.3 approximately after 12 % strain and becomes 0.05 after 17% strain.

Figure 4-8: Effect of strain rate on the hardening behaviour of AZ91D

Fig. 4-9 shows the effect of the temperature on the hardening behaviour of the alloy. The hardening exponent n decreases from 0.53 to 0.46 as the temperature increases from -30°C to 200°C . Similar to what was observed for strain rate effect, there are three regions regarding the alloy's hardening exponent which decreases with strain; it reduces to 0.3 approximately after 12 % strain and becomes nearly zero after 15% strain. At 200°C temperature, the slope of the $\log(\text{stress})$ - $\log(\text{strain})$ results is constant. However, it increases non-linearly at higher values of

strain. The reason for this can be explained on the basis of hardening and softening of the alloy; at 200°C the hardening and softening seems to balance each other after an initial strain of approximately 2 to 3%.

Figure 4-9: Effect of temperature on the hardening behaviour of AZ91D at 2500s⁻¹

The hardening exponent n is plotted against strain rate and temperature as shown in Fig. 4-10. To quantify the hardening behaviour of the alloy and to see how it changes with strain rate and temperature, the hardening exponent n is plotted against strain rate and homologous temperature T^* which is defined by the Equation (4.2), where T_0 (-273K) is the absolute zero and T_m (~620°C [893K]) is the melting temperature in Kelvin.

$$T^* = \frac{T - T_0}{T_m - T_0} \quad (4.2)$$

The effect of strain rate on the hardening exponent n is shown in Fig. 4-10(a) and it increases as the strain rate increases. The effect is more pronounced when strain rate changes from 10⁻⁴s⁻¹ to 10³s⁻¹ as compared to the change in strain rate from 10³s⁻¹ to 3x10³s⁻¹. Fig. 4-10(b) describes the temperature dependence of n , it decreases as the temperature increases and a larger reduction is observed at

200°C and 10^{-4}s^{-1} . Figures 4-10(a) and 4-10(b) are clear indicative of the rate and the temperature dependence of the hardening exponent of AZ91D although the effect is less significant at higher strain rates (10^3 s^{-1} to $3 \times 10^3 \text{ s}^{-1}$) and lower temperatures (-30°C and 25°C).

Figure 4-10: Hardening exponent as a function of (a) strain rate and (b) temperature

4.2.3 Strain Rate Sensitivity

Fig. 4-11 shows the stress as a function of strain rate (log scale) to quantify the strain rate sensitivity of the alloy. The stress is plotted at various strain levels between 5% and 20%. The yield stress (0.2% offset) is also included. Two distinct regions corresponding to different rate sensitivities are observed with a clear change in the rate sensitivity before 500s^{-1} . Moreover, on either side of this threshold, the stress exhibits a log-linear relationship between the stress level and the strain rate. The slope of the log-linear relationship known as the strain rate sensitivity exponent m (given by Eq. (4.3)) is larger at larger strains.

$$m = \left(\frac{\partial \sigma}{\partial \log \dot{\epsilon}} \right)_{\epsilon, T} \quad (4.3)$$

In the above equation, T is the test temperature and ϵ is a certain strain.

Figure 4-11: Strain rate sensitivity of AZ91D

A comparison of rate sensitivity of AZ91D from current work to that from literature is also presented in Fig. 4-11. It is observed that the present data is well within agreement in the range of strain rate available in the literature. However, at higher strain rates where the data is not available in the literature a comparison is made through employing Johnson-Cook model. A similar behaviour was observed for other materials such as shown in Fig. 2-1 and steel [32] as aluminium [102]. The strain rate at which the dependence of σ on $\log \dot{\epsilon}$ departs from linearity is of significant interest, since a change in the functional dependence of stress on the strain rate indicates that a different deformation mechanism is dominating the plastic flow behaviour [32]. However, more data in between 1s^{-1} and 500s^{-1} is required at lower strain rates such as 50s^{-1} , 100s^{-1} and 150s^{-1} for accurate threshold for rate sensitivity. This has not been done due to unavailability of testing facility for that low strain rates in the university.

The variation of stress with temperature is shown in Fig. 4-12. The flow stress at 5%, 10% and 15% plastic strain is plotted against homologous temperature; it

decreases as the temperature increases indicating the thermal softening of AZ91D in the range of temperature considered in the current study. It is noted that the stress is less affected for temperature change from -30°C to 100°C at quasi static strain rate. However, temperature seems to have considerable effect on the stress level of the alloy at 2500s⁻¹. When materials undergo plastic deformation their temperature rises. At high strain rates, the heat generated due to plastic deformation has very short time to dissipate into the surroundings causing an increase in the temperature of the specimen. It is important to find out the magnitude of this increase in the specimen temperature to know an extent to which it can affect the properties of the material. In the current study, increase in the specimen temperature due to plastic deformation was calculated from the energy absorbed (area under the whole stress-strain curve) by the specimen. The exemplary calculations are shown in appendix E4. The temperature of the specimens deformed at 2500s⁻¹ and 3500s⁻¹ increased to 67°C and 80°C respectively. The temperature of the specimen deformed at 200°C and 2500s⁻¹ increased to 239°C (0.4T_m) high enough to contribute to the softening of the alloy. However, these temperatures were not physically verified in the experiment.

Figure 4-12: Stress as a function of homologous temperature T*

Fig. 4-13 gives the rate dependence of stress at various temperatures ranging from -30°C to 200°C. The strain rate sensitivity exponent is calculated from Fig. 4-13 and plotted in Fig. 4-14. It is noticed that strain rate sensitivity of the alloy decreases with temperature up to 100°C and then increases with further increase in temperature up to 200°C as shown in Fig. 4-14. This is in accordance with the stress-strain behaviour, the alloy exhibited at different temperatures shown in earlier Figures 4-5 to 4-7.

Figure 4-13: Stress as a function of strain rate at various temperatures

Figure 4-14: Temperature dependence of strain rate sensitivity exponent m

4.2.4 Work Hardening Rate

Fig. 4-15 presents the work hardening rate ($d\sigma/d\varepsilon$) as a function of strain at various strain rates and temperatures. The work hardening rate ($d\sigma/d\varepsilon$) is computed by the numerical derivation of experimental true stress-strain curve. For any given strain rate and temperature, the work hardening rate decreases rapidly at smaller strains and gradually at larger strains. The work hardening rate is slightly higher at higher strain rates and the reason for this is believed to be the lower dislocation density at higher strain rates also mentioned for Al6060-T6 [103]. For strain over 0.13, the strain rate dependence of work hardening rate is insignificant and this might be due to the fact that at higher strain rates and larger strains the thermal softening takes over the work hardening of the alloy. However, the ($d\sigma/d\varepsilon$) decreases as the temperature increases from -30°C to 200°C . It is also observed that the work hardening rate has less variation (decrease in slope is relatively smaller) with strain at 200°C indicating that the alloy deforms more uniformly at elevated temperatures.

Figure 4-15: Work hardening rate versus true strain (a) at various strain rates (b) at various temperatures

The work hardening rate is shown as a function of stress at various strain rates and temperatures in Fig. 4-16. For any given strain rate and temperature, the work hardening rate decreases rapidly to a certain stress value after which it decreases gradually before dropping to zero at which the flow stress reaches its peak and keeps steady. The stress dependence of work hardening rate is slightly affected by the strain rate and the temperature variation. However, a slight decrease in the stress corresponding to zero work hardening rate is observed as the temperature increases from -30°C to 200°C . In the work hardening rate curves, mostly work hardening rate ($d\sigma/d\varepsilon$) keeps a level around $d\sigma/d\varepsilon = 0$ after which these curves inflect. This explains that the strain hardening rate and strain softening rate balance each other. Flow stress increasing with strain rate indicates an increase in the hardening ability of the alloy during deformation.

Figure 4-16: Work hardening rate versus true stress (a) at various strain rates (b) at various temperatures

4.2.5 Specimen Strain

The effect of strain rate and temperature on the specimen strain is shown in Fig. 4-17 and Fig. 4-18 respectively. At room temperature, the failure strain decreases

about 30% as the strain rate increases from 10^{-4}s^{-1} to 10^3s^{-1} and it undergoes a substantial increase of 150% for further increase in the strain rate up to 3000s^{-1} . The decrease and increase in the failure strain with increasing strain rate is an indication of the change in the deformation mechanism at different strain rates. The specimens deformed at 200°C have not failed at any strain rate. They deform plastically and some fine cracks are observed.

Figure 4-17: Strain rate dependance of alloy's ductility under compression

It is observed in Fig. 4-18 that the specimen strain is considerably affected as the temperature increases from 25°C to 200°C at 10^{-4}s^{-1} strain rate, while little affected at dynamic strain rates. Increase in the strain to failure is about 33% and 15% at 10^{-4}s^{-1} and 2500s^{-1} respectively as the temperature increases from -30°C to 100°C . However, smaller strain to failure at 10^3s^{-1} at all temperatures indicates the lower ductility of the alloy as compared to 10^{-4}s^{-1} . The temperature effect on the failure strain is less pronounced at dynamic strain rates marked by ellipse in Fig. 4-17 also confirmed in figure 4-18. This indicates that the major deformation mechanism at dynamic strain rates is not changed with change in the temperature.

Figure 4-18: Temperature effect on the specimen strain

The specimens under high strain rate compression fail by ductile shear fracture and a compression cone is formed which is also observed in AA7075 and AZ80 [104]. In AZ91D specimen deformed at 1000s⁻¹ and above, the compression cone can be seen in Fig. 4-19. It is also observed that the shear fracture is approximately at 45° to the loading direction. At higher strain rates i.e. 2500s⁻¹, multiple fractures (shown by arrow in Fig. 4-19) occurred parallel to each other but all at 45° to the directions of force. Specimens tested at 200°C did not fracture and deformed by plastic flow

Figure 4-19: Fractured specimens at high strain rate at 25°C and 200°C

4.2.6 Energy Absorption Capability

Energy absorption at various strain rates is shown in Fig. 4-20. Energy absorption capability during deformation is a measure of its performance against impact or crash situations. Higher energy absorption capability of a material indicates its resistance against impact. The energy absorbed during deformation is calculated

using the definition of area under the stress-strain curve up to failure strain. To evaluate the energy performance of AZ91D alloy, the energy absorbed by the alloy during compression was calculated using Eq. (4.4) and plotted in Fig. 4-20

$$E = \int_0^{\varepsilon_f} (\sigma) d\varepsilon \quad (4.4)$$

where ε_f is the failure strain. The specimens of same size were tested at different strain rates to find out the change in energy with increasing strain rate. In general, the ability of AZ91D to absorb energy is strain rate dependent and it increases with increasing strain rate. However, it is slightly lower at 10^3s^{-1} as compared to 10^{-4}s^{-1} strain rate.

Figure 4-20: Energy absorption capability at various strain rates

The effect of temperature on the energy absorption of the alloy during deformation is plotted against homologous temperature as shown in Fig. 4-21. The temperature effect on the energy absorption is less significant at 10^{-4}s^{-1} . However, at higher strain rate of 2500s^{-1} , the energy absorbed decreases about 7% as the temperature increases from 25°C to 100°C . The energy absorption at 200°C was also calculated by taking the area under the whole stress-strain curve and it is estimated that the energy absorbed by the alloy is substantially affected

by the temperature if the specimen is loaded until fracture. It is worth to note that at all temperatures, the energy absorption capability at 2500s^{-1} is 2.5 to 3.25 times higher than what is observed at quasi-static strain rate.

Figure 4-21: Energy absorption of AZ91D as a function of homologous temperature T^*

4.3 Tensile Behaviour

Tensile tests were performed at 10^{-4}s^{-1} , 350s^{-1} , 750s^{-1} and 1500s^{-1} at room temperature. Some tests were also performed at 250°C for 10^{-4}s^{-1} and about 1500s^{-1} . Results are presented and discussed in the following sections.

4.3.1 Strain Rate and Temperature Effect

The dynamic behaviour of AZ91D alloy under tensile loading at various strain rates is presented in Fig. 4-22 followed by Fig. 4-23 where the average experimental strain rates achieved in the current study are plotted against time.

Figure 4-22: Strain rate effect on the tensile behaviour of AZ91D

Figure 4-23: Strain rate variation with time achieved during tensile testing

In the plastic deformation region, higher stresses are observed at higher strain rates as indicated by Figure 4-22. The yield strength, ultimate tensile strength (UTS) and the energy absorption at 1500 s^{-1} are about 70%, 78% and 112% higher respectively than their respective values at 10^{-4} s^{-1} . The strain to failure at dynamic strain rates is larger than the strain to failure at 10^{-4} s^{-1} . It is about 30% larger at 750 s^{-1} and about 12% larger at 1500 s^{-1} as compared to the strain to failure at 10^{-4} s^{-1} . A decrease (about 14.7%) in failure strain can be noticed as the strain rate is increased from 750 s^{-1} to 1500 s^{-1} . A comparison of mechanical properties is shown in Table 4-1.

Table 4-1: Tensile properties of AZ91D at room temperature (25°C)

Parameters/Strain Rate	10^{-4}s^{-1}	350s^{-1}	750s^{-1}	1500s^{-1}
Yield Strength [MPa]	64	107 ± 2.5	125 ± 3.3	134 ± 4
UTS [MPa]	121	153.3 ± 4	183.7 ± 5.3	216.5 ± 6.5
Strain to failure [%]	3.35	*	4.4 ± 0.1	3.75 ± 0.05
Energy [MJ/m ³]	2.78 ± 0.3	*	5.74 ± 0.15	5.90 ± 0.1

**specimens have not failed*

The stress-strain curves at 10^{-4}s^{-1} and 1500s^{-1} strain rates at 25°C and 250°C temperature are shown in Fig. 4-24. The effect of temperature is not significant at quasi-static strain rate whereby it is considerable at 1500s^{-1} . The yield strength, UTS and energy absorption at 250°C and 1500s^{-1} (Table 4-2) are decreased by 10%, 35% and 29% respectively as compared to what is observed at 25°C at 1500s^{-1} (Table 4-1).

Figure 4-24: Temperature effect on the tensile behaviour of AZ91D

Table 4-2: Tensile properties of AZ91D at 250°C

AZ91D	YS [MPa]	UTS [MPa]	Energy absorbed [MJ/m ³]	Strain to failure [%]
10^{-4}s^{-1}	62	116.5	2.7	3.43
1500s^{-1}	120 ± 5	140 ± 4.5	4.185 ± 0.15	4.65 ± 0.1

4.3.2 Specimen Strain and Energy Absorption

Larger failure strain and higher energy absorption are observed at dynamic strain rates as compared to the quasi-static strain rate. It is observed in Table 4-1 that the failure strain at 1500s^{-1} is about 10% higher than at 10^{-4}s^{-1} strain rate and the energy absorption at 1500s^{-1} is twice as compared to the energy absorption at 10^{-4}s^{-1} . At 10^{-4}s^{-1} strain rate, the failure strain is slightly increased while the energy absorption is little decreased with increase in the temperature from 25°C to 250°C . However, at 1500s^{-1} strain rate, the failure strain is increased by 24% and energy absorption is decreased by 29% respectively as the temperature increases from 25°C to 250°C .

Fractured specimens tested at 1500s^{-1} strain rate and 25°C and 250°C temperature are shown in Fig. 4-25. There is no sizeable necking even in the specimen tested at 250°C . However, it is observed that the fracture in the specimen tested at 250°C is a pure ductile shear fracture at 45° angle to the direction of force.

Figure 4-25: Fractured tensile specimens at 1500s^{-1}

4.3.3 Work Hardening

The work hardening exponent n is plotted against strain rate at room temperature in Fig. 4-26. The exponent n increases with increasing strain rate indicating the increased ability of the alloy to strengthen during deformation at higher impact rates.

Figure 4-26: Hardening exponent as a function of strain rate at room temperature

Fig. 4-27 shows the hardening exponent n as a function of homologous temperature. At $1500s^{-1}$, the alloy exhibited substantial softening at $250^{\circ}C$ whereas at quasi-static strain rate the alloy did not show any notable softening. In Fig. 4-27, it is observed that the hardening exponent n is little affected by temperature change at $10^{-4}s^{-1}$ strain rate while significantly affected at $1500s^{-1}$. The results reveal that at elevated temperatures under dynamic tensile loading, the alloy's hardening ability is drastically reduced.

Figure 4-27: Hardening exponent n versus homologous temperature

4.3.4 Strain Rate Sensitivity

Fig. 4-28 shows strain rate sensitivity of AZ91D alloy at room temperature at different strain levels. A sharp increase in the stress can be observed at $350s^{-1}$

indicating an increase in the rate sensitivity of the alloy in dynamic range of strain rate. The rate sensitivity exponent increases with strain; it is about 3 times higher at 3% strain as compared to 1% strain. The strain rate sensitivity exponent m averaged over strain in the dynamic range of strain rate is 0.0339. A comparison of rate sensitivity of AZ91D from current work to that from literature along with Johnson-Cook model fitted data is also presented in the figure. It is observed that the data fitted by model is close to experimental data. However, the data from Liao et al [59] is little higher in strength level but the rate sensitivity of 0.54 calculated from the data presented in his paper is very close to the present work.

Figure 4-28: Strain rate sensitivity of AZ91D at 25°C

4.3.5 Work Hardening Rate

The work hardening rate as a function of strain and stress is shown in Fig. 4-29 and Fig. 4-30 respectively under tensile loading. In the beginning a trend similar to the compressive loading is seen in tensile loading as well. In general, the work hardening rate increases with increasing strain rate. At 1500s⁻¹, the work hardening rate slightly increases at a strain of 0.02 indicating some additional hardening and then decreases possibly due to softening caused by high strain rate

deformation. In Fig. 4-30, the stress corresponding to zero work hardening rate (work hardening rate curve meets stress axis) increases with increasing strain rate which indicates that the strain to onset of necking increases with increasing strain rate.

Figure 4-29: Work hardening rate versus true strain under tensile loading

Figure 4-30: Work hardening rate versus true stress under tensile loading

Considered proposed a criterion for the onset of necking. According to Considere's criteria, the necking is initiated when the work hardening rate becomes equal to true stress and is given by the Equation (4-6) [105].

$$\frac{d\sigma_T}{d\varepsilon_T} = \sigma_T \quad (4-6)$$

The point of intersection between true stress-true strain curve and the $d(\sigma)/d(\epsilon)$ curve is called the Considere point, the point at which necking begins. This is also the point of maximum load, slope at maximum point equals zero implies that $dP=0$, where P stand for load. Fig. 4-31 shows the Considere points at various strain rates. In tensile loading, it is important to see the effect of strain rate and temperature on the onset of necking. The Considere criteria [105] can be used to estimate these effects on the necking during deformation. In present study, it is useful to estimate the onset of necking as magnesium alloys used in current study did not show significant necking in the range of strain rate between 10^{-4}s^{-1} and 1500s^{-1} .

In Fig. 4-31, Considere point corresponds to the stress where work hardening rate becomes equal to true stress. The strain at that point is Considere strain and it increases with increasing strain rates. However, one can notice that the onset of necking of AZ91D doesn't present a significant sensitivity to strain rate considered in the present study. The alloy exhibits an average strain of about $3\pm0.2\%$ at which the Considere criteria is satisfied.

Figure 4-31: Considere point at various strain rates

4.4 Compressive-Tensile Yield Asymmetry

In cast alloys, the crystals are random in orientation and the effect of orientation is nullified and thus tensile and compressive stress-strain curves are nearly identical. The behaviour of wrought alloys depends on the texture (orientation of their grains) and thus wrought alloys show strong compressive-tensile asymmetry. Usually a material is said to have a compressive-tensile yield asymmetry if it shows different yield stress in compression than that under tension at a same rate of deformation. Magnesium alloys show different values of yield stress under compressive and tensile loading. A comparison of compressive and tensile stress-strain curves at various strain rates (10^{-4}s^{-1} , 750s^{-1} and 1500s^{-1}) is shown in Fig. 4-32. It is observed that the yield stress is sensitive to type of loading. At dynamic strain rates, the tensile yield stresses are slightly higher than their respective values of compressive yield stresses at same strain rates. The difference in the tensile and compressive yield stress is primarily due to different modes of deformation in both types of loading. However, it is noticed that in AZ91D alloy, the tensile-compressive asymmetry is not as significant as in pure magnesium [106], where compressive yield strength was 50% lower than the corresponding tensile yield strength at $2 \times 10^{-3}\text{s}^{-1}$ and $2 \times 10^{-1}\text{s}^{-1}$. Furthermore this asymmetry is also found in maximum stress (peak compressive stresses are higher than Ultimate tensile stresses) and strain to failure as shown in Table 4-3.

Figure 4-32: Compressive-tensile asymmetry in AZ91D alloy

Table 4-3 Comparison between compressive and tensile mechanical properties of AZ91D

Parameters/Strain rate [s^{-1}]		10^{-4}	750	1500
Yield stress [MPa]	Compressive	82.6	100	116
	Tensile	80	125	134
Maximum stress [MPa]	Compressive	223.9	242	358
	Tensile	126	183.7	216.5
Strain to Maximum stress/failure	Compressive	14.4	6.95*	18.3
	Tensile	3.15	3.84	3.58

* Note that specimens have not failed; strain is taken up to maximum stress

4.5 Microstructural Analysis

Optical microscope and scanning electron microscope (SEM) images are used to study the microstructural changes at various strain rate and temperature under compressive and tensile loading. Micrographs for specimens tested under various conditions are shown in the following figures.

4.5.1 Compression

Fig. 4-33(a) shows the micrographs of original AZ91D clearly showing α -Mg matrix and β -Mg₁₇Al₁₂. The β -phase presents a dispersed net-shape distribution surrounding α -matrix to make an apparent dendritic structure. Fig.4-33(b) shows

the magnified structure of the alloy. The β -phase can also be seen in the form of lamellar structure that surrounds the bulk β -phase.

Figure 4-33: (a) α -phase and β -phase are visible (b) lamellar β -phase in dendritic form is present around bulk β -phase ($\text{Mg}_{17}\text{Al}_{12}$)

4.5.1.1 Strain Rate Effects

Micrographs of specimens tested at various strain rates are shown in Fig. 4-34. The microstructure of specimen tested at 10^{-4}s^{-1} is shown in (a) and (b) where few cracks are seen in bulk β -phase. Relatively more cracks in β -phase are seen at 500s^{-1} (c and d) strain rate while severe cracking is observed (e to h) in specimens tested at 2500s^{-1} strain rate. The hard and brittle β -phase is broken into small gains which are distributed along the grain boundaries with orientation perpendicular to the loading direction. At high strain rates lamellar β -phase also shows some deformation and its needle like structure changes into flakes. The formation of local shear bands and a sponge like structure of β -phase is observed at 2500s^{-1} strain rate.

Figure 4-34: At room temperature: (a) and (b) few cracks in bulk β -phase (c) and (d) more number of cracks in bulk β -phase (e) to (h) severe damage in bulk β -phase as well as in lamellar β -phase

4.5.1.2 Temperature Effects

The microstructure of AZ91D tested at 2500s^{-1} strain rate and at -30°C and 200°C temperatures are shown in Fig. 4-35. At -30°C , severe damage in β -phase is observed and the fracture seems to be more brittle in nature thus causing a slight reduction in the ductility of the alloy. Some twins can also be seen at -30°C . At 200°C , only few cracks are seen in the β -phase. However, slight elongation of β -phase along with the formation of pores shows a change in its deformation behaviour towards less brittle thus adding some additional ductility to the alloy. The work hardening and increased resistance of α -Mg to deformation led to the increase of stresses in AZ91D with increasing strain rate.

Figure 4-35: At 2500s^{-1} : (a) to (c) severe damage in bulk β -phase as well as in lamellar β -phase (d) to (f) fewer cracks, voids in bulk β -phase

4.5.2 Tension

4.5.2.1 Strain Rate effects

Microstructure of tensile loaded specimens is shown in Fig. 4-36. Few cracks in bulk β -phase are seen at quasi-static strain rate while more cracks and severe damage is observed at 1500s^{-1} strain rate. Usually more cracks are seen in the vicinity of the fracture and fewer away from the fracture. Both sides of the fracture surface are covered with aluminium rich β -phase ($\text{Mg}_{17}\text{Al}_{12}$) which shows that the cracks propagate along the grain boundaries as shown in Fig. 4-36(d). At high strain rate, the grains size is refined and the grain boundaries are increased thus causing an increase in the strength of the alloy. An increase in the yield strength due to grain refinement was also reported in the literature [107]. In Fig. 4-36, the grain size is about $150\text{-}250\mu\text{m}$ at quasi-static strain rate and 100 to $160\mu\text{m}$ at 1400s^{-1} . Cracking in α -Mg matrix is also observed at dynamic strain rates. In tensile deformation, micro-cracks nucleate at the interface between α matrix and eutectic β -phase. Some pores (shown by arrows) are formed along the grain boundaries at both quasi-static and dynamic strain rates. The formation of micro-pores and cavities along the grain boundaries in inter dendritic areas may

also contribute to the initiation of cracks propagation along the grain boundaries leading to the fracture of the specimen.

The tensile-compressive yield asymmetry is attributed to the difference in the deformation modes under compressive and tensile loadings. More cracks in β -phase, cracks in α -Mg and presence of some twining cause higher strength and larger strains in the compressive loading as compared to tensile loading. However, in as-cast AZ91D the tensile-compressive yield is not significant revealing some similarity of the deformation modes in both compressive and tensile loading in as-cast alloys in initial stages of deformation.

Figure 4-36: Tensile loading at room temperature (a) larger grain size (b) few cracks in bulk β -phase (c) refined grains with larger grain boundaries (d) lots of cracks in β -phase near fracture surface (e) no cracks away from fracture surface (f) cracks in α -matrix as well as in β -phase near fracture surface

4.5.2.2 Temperature Test

Fig. 4-37 presents microstructure of AZ91D at 250°C and 2500s⁻¹. It is observed that grains are elongated and voids are formed within α -matrix. Fewer cracks are seen in β -phase but relatively more cracks are observed near fracture surface. Additionally some eutectic- α is precipitated around bulk β -phase. The lamellar β -phase is partially recrystallized and distributed along the grain boundaries. At a strain rate 1500s⁻¹, the effect of temperature 250°C is less significant due to the mitigating effects of strain hardening and softening due to the elevated temperature also has been observed from stress-strain curves in Fig. 4-22.

Figure 4-37: Tensile loading at 1500s^{-1} : (a) elongated grains and voids (b) some of lamellar β -phase recrystallized around bulk β -phase, cracks within β -phase (c) and (d) severely cracked β -phase near fracture surface

4.5.3 Energy Dispersive X-Ray Analysis (EDX)

Energy Dispersive X-Ray Analysis (EDX) is an x-ray technique used to identify the elemental composition of materials. In this analysis, the microscope scans the specimen and EDX generates the data that consists of spectra showing peaks corresponding to the elements making up the true composition of the sample being analysed. In current study, EDX analysis was performed for some selected specimens and results are shown in Fig. 4-38. The composition of various constituents of the alloy are tabulated along with the micrographs. Several compositions of Mg and Al are present including α -Mg and bulk β -phase with some minor phases rich in Mn and Si in the form of scattered particles. The Si (36-39%) is present with magnesium and aluminium (Mn/Al/Si) and Mn is about 10 to 11% in the form of intermetallic compounds Al_8Mn_5 or $\text{Al}_{11}\text{Mn}_4$. These fine intermetallic particles, in-homogenously dispersed inside grains and along grain boundaries, resist to slip in the primary α and β phase causing relatively high hardening rate in the early stage of plastic deformation. Intermetallic compound present in AZ91D such as eutectic β -phase ($\text{Mg}_{17}\text{Al}_{12}$) and hard and brittle Si-rich and Mn-rich intermetallic compounds as observed in EDX analysis cause an

increase in the resistance to dislocation motion adding to the strength of the alloy.

Figure 4-38: EDX; (a) Various composition of Mg and Al including α -phase and β -phase (b) and (c) some minority phases rich in Mn and Si are present.

4.6 Discussion

At room temperature, yield stress, peak compressive stress, strain to failure and energy absorption is higher at dynamic strain rates than quasi-static strain rates. Yield strength is little affected by strain rate increase also supported by the fact that the alloy is less rate sensitive at lower strains. This less sensitivity of the alloy in early stage of deformation is due to the reason that the principal mechanisms responsible for deformation are basal slip and tensile twinning and both these phenomena are little sensitive to the strain rate. At elevated temperatures, both the yield stress and the peak compressive stress decrease however this change is little at dynamic strain rates. A possible reason for this is that the effect of temperature (thermal softening) is mitigated by the effect of strain rate (strain hardening). At room temperature, the fracture strain at 500s^{-1} and 10^3s^{-1} strain rates is lower as compared to the quasi-static fracture strain. It is suggested that this is due to the higher strain hardening of the alloy at 500s^{-1} and 10^3s^{-1} as compared to the quasi-static strain rate (Fig. 4-8) resulting due to the contribution of hard and brittle intermetallic β -phase ($\text{Mg}_{17}\text{Al}_{12}$) to the deformation of alloy and the increased resistance of α -matrix to deformation. Higher aluminium content (9%) present in AZ91D causes the formation of inter-dendritic β -phase. It gathers at grain boundaries in the form of fine precipitates that causes an increase in the strength of the alloy by suppressing basal slip. The presence of inter-dendritic β -phase at grain boundaries has also been reported to impede the process of twinning in magnesium alloys [56] which leads to a limited ductility of AZ91D. With further increase in the strain rate beyond 10^3s^{-1} , the stress and ductility of the alloy both increase. At higher strain rates, the thermal

effects due to deformation heating lower the resistance to deformation possibly due to increased number of slip system [108]. Furthermore, α -phase is likely to produce twining under dynamic strain rates which unlike slip causes a lattice rotation to make the slide system move easily. A large amount of rupture in β -phase as seen in Fig. 4-34 promotes the grain boundary deformation thus causing the strain to increase with strain rate. Under tensile loading, the effect of temperature is considerable at dynamic strain rates while insignificant at quasi-static strain rate. This suggests that the deformation mechanism at quasi-static strain rate even at elevated temperatures is same while at dynamic strain rates it changes, some non-basal slip systems become active. At room temperature and lower strain rates, the stress-strain curves under compression are little concave up in the early stage of deformation indicating the tensile twining activity. In tensile loading, the stress-strain curves are concave down at all strain rates and temperatures showing that the tensile twining is higher in compression as compared to tension. At elevated temperatures and dynamic strain rates, stress-strain curves show significant hardening indicating the deformation by twining and slip at dynamic strain rates even at elevated temperatures. Both in compression and tension, the alloy fails with shear fracture at room temperature (Fig. 4-19 and Fig. 4-25) at an angle of 45° to the loading direction with shear bands formation at higher strain rates (Fig. 4-39). At elevated temperatures, the specimens did not fail in compression while in tension the nature of fracture is shear at 45° to the loading direction.

Figure 4-39: Shear bands under compressive loading

In compression, cracks initiate at several locations, grow and join to form a fracture zone while in tension crack path is inter-granular as shown in Fig. 4-40 below. The brittle eutectic β -phase segregation along the grain boundaries seems to cause inter-granular fracture. Micro-cracks nucleate at the interface between β - $\text{Mg}_{17}\text{Al}_{12}$ and the magnesium-matrix and the reason reported for this is the bcc structure of β - $\text{Mg}_{17}\text{Al}_{12}$, which is not compatible with the hexagonal structure of the α -matrix [55]. This also leads to decrease in the ductility of the alloy. The inter-granular crack path is also due to the resistance of discontinuous net-shaped β - $\text{Mg}_{17}\text{Al}_{12}$ against the crack propagation and this was also observed even at lower strain rates by Gupta et al. [109]. In tensile loading, usually fracture occurs at grain boundaries due to increased aluminium contents in the region as a result of decomposition of intermetallic $\text{Mg}_{17}\text{Al}_{12}$ phase.

Figure 4-40: Cracks and fracture surface under dynamic loading

Usually the microstructural changes such as grain refinement results in grain boundary strengthening causes higher yield strength. In the present study, it is observed that an average reduction in the grain size at 1400s^{-1} is about 40% ($250\mu\text{m}$ to $160\mu\text{m}$) as compared to 10^{-4}s^{-1} contributing to higher yield strength. The β -phase and shear bands both seem to have contributed to the strength and improved ductility of the alloy at higher compressive strain rates. At elevated temperature, the lamellar β -phase is partially recrystallized and distributed along the grain boundaries which lower the strength of the alloy.

The work hardening rate increases with increasing strain rate in the dynamic range of strain rates due to higher strain hardening which happens due to creation, multiplication and mobility of dislocations. It slightly decreases with increasing true strain at all strain rates due to severe breakage of intermetallic β -phase which facilitates the dislocation movement due to thermal softening a result of deformation heating. However in the current study, it is observed that the average temperature of the specimen after being tested at room temperature and 3000s^{-1} strain rate is increased up to $75 \pm 5^\circ\text{C}$ ($0.12T_m$). This temperature rise seems insufficient to activate non-basal slip activity (Fig. 2-8 and Fig. 2-9) and thus has little effect on the stress-strain behaviour of the material. The work

hardening rate decreases slightly at 200°C testing temperature. The activity of non-basal slip system becomes active at temperature above 200°C. At higher temperatures, cross slip becomes active[58], and is responsible for softening of the material. As a result, the work hardening rate should decrease and which is observed experimentally in the current study. Increase in maximum flow stress with strain rate (Fig. 4-16a) indicates the increase in the hardening ability of the alloy during deformation. However, the maximum flow stress is considered to decrease with increasing temperature being a characteristic of the basal slip process. In the current study, the maximum flow stress is little changed (Fig. 4-16b) as the effect of temperature is compensated by the strain rate effect.

Chapter 5

Results and Discussion (AZ31B)

In this chapter, the experimental results from quasi-static and high strain rate tests of AZ31B alloy under compressive and tensile loading are presented. Effects of strain rate and temperature on the mechanical properties of AZ31B are investigated. The anisotropy and tensile-compressive asymmetry of AZ31B is also studied. Microstructure analysis is performed and experimental results have been interpreted on the base of microstructural features.

5.1 Compressive Behaviour

The mechanical behaviour of the alloy is investigated at several strain rates between 10^{-4}s^{-1} and 3500s^{-1} . Loading is applied in normal, rolling and transverse directions as well as at 45° angle (45° angle to the rolling direction). Effect of temperature is studied at -30°C and 200°C in compression and at 250°C in tension. Microstructures of specimens tested at quasi-static and about 2250s^{-1} strain rate and at various temperatures (-30°C , 25°C , 200°C and 250°C) under compression and tension are presented and illustrated.

5.1.1 Strain Rate and Temperature Effects

The stress-strain curves under compression in normal, rolling, transverse and at 45° angle orientation are presented in Fig. 5-1 to Fig. 5-4 respectively. It is observed that the dynamic stresses are considerably higher than static stresses in all loading orientations.

Figure 5-1 Stress-strain curves obtained in the normal direction

Figure 5-2: Stress-strain curves obtained in the rolling direction

Figure 5-3: Stress-strain curves obtained in transverse direction

Figure 5-4: Stress-strain curves obtained in 45° loading direction

In the dynamic strain rate range, the stress level increases with increasing strain rate. However, the stress level drops as the strain rate is increased to 3500s^{-1} in normal and transverse loading direction. For the specimens tested in normal direction, the yield stress, elongation at failure and hardening rate increased significantly. For specimen tested in other directions, the yield stress is little affected by the strain rate. Whereas a decline in stress after peak value (peak compressive stress) is because of thermal softening caused due to the rise in specimen temperature as a result of adiabatic deformation at high strain rates and larger strains. The temperature rise, where temperature is the average over the whole specimen, is about 40°C to 55°C with local temperature increase even higher as seen in the form of shear bands and local plastic flow in microstructural analysis that is explained in the later sections. It is also observed that except for the normal loading orientation, stress-strain curves are sigmoidal (concave upward) at all strain rates and exhibit strong hardening behaviour after an initial period of lower strain hardening. The concave down stress-strain curves usually reflect the deformation by slip and concave upward (early stage of deformation in

three directions other than normal) period of stress-strain curve is ascribed to dominant tensile twinning activity.

Table 5-1: Compressive mechanical properties of AZ31B alloy in normal direction

Magnesium alloy AZ31B (Normal direction)						
Strain rate $\dot{\epsilon}$ [s ⁻¹]	10 ⁻⁴ (2) ⁱ	500(4)	1000(4)	1500(4)	2500(4)	3500(4)
Density [Kg/m ³]	1.749 ^{-0.055} _{+0.052}					
Yield strength [MPa]	140	63.5±2.1 ⁱⁱ	97.8±2.5	171.8±3.0	183.8±3.0	123.5±6.45
Peak Compressive Strength (PCS)[MPa]	268.50	178.3±3.4	397±6.1	463.2±3.1	474.3±4.5	419.5±3.5
% strain to failure	12.1	*4.84±0.06	*11.4±0.19	17.8±0.6	17.65±0.5	17.6±0.6
Absorbed Energy (MJ/m ³)	23.9	*5.73±0.16	*29.9±0.95	73.5±1.5	75.2±2.5	69±2

ⁱNumber in brackets () indicates the number of tests conducted in each condition

ⁱⁱ Mean ± standard error, * Specimens did not break at these strain rates therefore strain and energy absorption are calculated up to peak compressive stress

Table 5-2: Compressive mechanical properties of AZ31B alloy in rolling direction

Magnesium alloy AZ31B (Rolling direction)						
Strain rate $\dot{\epsilon}$ [s ⁻¹]	10 ⁻⁴ (2) ⁱ	500(4)	1000(4)	1500(4)	2500(4)	3500(4)
Density [Kg/m ³]	1.749 ^{-0.055} _{+0.052}					
Yield strength [MPa]	70	72.75±1.1	75.5±1.5	79.3±1.75	81.4±1.5	89.5±5.2
PCS [MPa]	311.4	159.3±1.8	344.8±5.8	495.4±7.9	523.5±8.4	576.4±7.4
% strain to failure	17.8	*6.03±0.8	*10.2±0.15	19.5±0.35	24.85±0.5	27.8±0.5
Absorbed Energy (MJ/m ³)	30.5	*6.53±0.4	*18.9±0.45	68.1±2.2	83.5±2.5	110.2±2.0

Table 5-3: Compressive mechanical properties of AZ31B alloy in transverse direction

Magnesium alloy AZ31B (Transverse direction)						
Strain rate $\dot{\epsilon}$ [s ⁻¹]	10 ⁻⁴ (2) ⁱ	500(4)	1000(4)	1500(4)	2500(4)	3500(4)
Density [Kg/m ³]	1.749 ^{-0.055} _{+0.052}					
Yield strength [MPa]	71	72.5±2 ⁱⁱ	75.3±1.5	78.2±2.2	81.25±2.2	81.75±2.0
PCS[MPa]	293.75	167.3±4.9	368.5±7.1	534.4±2.8	570.5±7.8	516.4±3.7
% strain to failure	18.75	*6.1±0.06	*9.38±0.16	19.75±0.5	24.85±0.5	18.9±0.7
Absorbed Energy (MJ/m ³)	33.6	*6.35±0.3	*17.1±0.6	69.2±1.9	110.5±3.5	85.5±5.0

Table 5-4: Compressive mechanical properties of AZ31B alloy in 45° direction

Magnesium alloy AZ31B (45° direction)						
Strain rate $\dot{\epsilon}$ [s ⁻¹]	10 ⁻⁴ (2)	500(4)	1000(4)	1500(4)	2500(4)	3500(4)
Density [Kg/m ³]	1.749 ^{-0.055} _{+0.052}					
Yield strength [MPa]	65.50	71.25±3.5	74.5±2.5	85.8±3.4	89.3±3.24	91.5±2.3
PCS [MPa]	292.850	166.8±3.5	306.3±8.1	541.3±3.2	552±9.1	573.6±5.3
% strain to failure	19.57	*6.85±0.05	*9.2±0.15	20.20±0.2	22.10±0.2	26.65±1.0
Absorbed Energy (MJ/m ³)	34.4	*7.23±0.07	*14.3±0.45	72.8±0.9	85.0±2.0	104.5±4.6

Mean values of strength, strain to failure and energy absorption of AZ31B in different directions with standard deviation are given in Table 5-1 to Table 5-4.

The yield strength, peak compressive strength, strain to failure and energy absorption increase with increasing strain rate. However, a slight decrease in these properties is observed in normal and transverse loading directions for 3500s^{-1} strain. In all four loading directions, yield strength, strain to failure and energy absorption at strain rates of 500s^{-1} and above are considerably higher than what are observed at quasi-static rate. However, it is notable that the specimens tested at 500s^{-1} and 1000s^{-1} are not failed. The smaller stress and strain values at 500s^{-1} and 1000s^{-1} are due to fact that the applied dynamic load has ended before the specimens fail. Therefore, maximum stress and failure strain are not reached at these two strain rates.

The stress-strain curves at -30°C , 25°C and 200°C temperature in four loading directions are shown in Fig. 5-5 to Fig. 5-8.

Figure 5-5: Temperature effect on the mechanical behaviour in normal direction

Figure 5-6: Temperature effect on the mechanical behaviour in rolling direction

Figure 5-7: Temperature effect on the mechanical behaviour in transverse direction

Figure 5-8: Temperature effect on the mechanical behaviour in 45° loading direction

In general, the stress level decreases with increasing temperature. The decrease in stress with temperature increase is strain dependent as well. For 5%, 10% and 15% strain, the stress is reduced by 52%, 48% and 51% respectively in normal loading, 18%, 35% and 48% respectively in rolling direction, 52%, 51% and 57% respectively in transverse direction and 14%, 25% and 35% respectively in angular (45°) direction as temperature increase from -30°C to 200°C. It is noted that stress level at -30°C is slightly higher as compared to 25°C with an exception at 45° angle loading where it is slightly lower. At 200°C temperature, the stress-strain curves are concave down in all loading directions and stress-strain curves reflect the deformation by plastic flow.

Figure 5-9: Temperature effect on the strength of AZ31B

Temperature dependence of yield strength and peak compressive strength is shown in Fig. 5-9. Both yield strength and peak compressive strength decrease as temperature increases. The yield strength is reduced by 20% while peak compressive strength is reduced by 35% to 53% (least in 45° and maximum in transverse loading) in different loading directions as the temperature increases from -30°C to 200°C. However, the yield strength in normal direction is more

sensitive to temperature; it decreases by 60% as the temperature is increased from -30°C to 200°C.

5.1.2 Work Hardening

Hardening behaviour of the alloy in different loading direction is illustrated in Fig. 5-10 to Fig. 5-13 where true stress is plotted against true strain at several strain rates from 500 to 3500/s.

Figure 5-10: Effect of strain rate on the hardening behaviour in normal direction

Figure 5-11: Effect of strain rate on the hardening behaviour in rolling direction

Figure 5-12: Effect of strain rate on the hardening behaviour in transverse direction

Figure 5-13: Effect of strain rate on the hardening behaviour in 45° loading direction

In general, hardening exponent ' n ' increases with increasing strain rate except in normal loading where it shows a mixed trend of increase and decrease. The average hardening exponent (averaged over the whole range of dynamic strain rate considered) is quite low for normal (0.4998) loading as compared to rolling (0.8085), transverse (0.8273) and 45° angle (0.7662) loading indicating lower capability of hardening in the plastic range in the normal direction. One possible reason for this low n may be that while manufacturing of rolled plate the grains

are pre-hardened in the normal direction and have less capacity for further hardening during deformation.

The alloys hardening behaviour at -30°C and 200°C is presented in Fig. 5-14 and Fig. 5-15 respectively. In all loading directions, the hardening exponent decreases as the temperature increases from -30°C to 200°C. The alloy displayed the lowest hardening exponent (0.371-0.590) in normal loading direction at both -30°C and 200°C temperatures, again indicating lower capability of the alloy to strengthen further during the plastic deformation in normal direction loading.

Figure 5-14: Hardening behaviour in different loading orientations at -30°C

Figure 5-15: Hardening behaviour in different loading orientations at 200°C

Figure 5-16: Hardening exponent as a function of homologous temperature

In Fig. 5-16, the hardening exponent ' n ' is plotted against homologous temperature. In each loading direction, the hardening exponent decreases as temperature increases ($n_{-30^{\circ}\text{C}} > n_{25^{\circ}\text{C}} > n_{200^{\circ}\text{C}}$). The decrease in the hardening exponent (softening) seems to have been caused due to increase in temperature during deformation. At high strain rates, the deformation is adiabatic and specimen temperature increases due to conversion of plastic energy into heat. In the current study, rise in temperature is calculated for specimens deformed at same strain rate ($2400 \pm 100 \text{ s}^{-1}$) and at 25°C and 200°C temperature. For temperature calculations, it is assumed that 90% of the total energy from plastic deformation is converted into heat so the fraction of plastic work converted into heat ' β ' is 0.9 [72, 110]. Increase in the specimen temperature is 37°C (normal) to 54°C (transverse) when at room temperature, while it is 29°C (normal) to 34.2°C (rolling) when deformed at 200°C .

5.1.3 Strain Rate Sensitivity

Stress is plotted versus strain rate at various strains in Fig. 5-17 to Fig. 5-20. Two different regions corresponding to two different rate sensitivities are observed, a

steeper slope of fitted lines indicates a change in the strain rate sensitivity at 500s^{-1} . However, stress-strain data at 30s^{-1} , 100s^{-1} and 200s^{-1} is needed to understand a true trend of alloy's rate sensitivity. It is also noticed that the strain rate sensitivity of the alloy is strain dependent too; it increases first and then decreases with strain. The strain rate sensitivity exponent m , averaged over the whole range of strain considered, is found to be approximately 0.0226 ± 0.0036 with little variation in four different loading directions. Over all the average strain rate sensitivity exponent m is higher for normal loading as compared to other loading orientations. At higher strain rates, a linear relationship of stress and strain rate is often observed. The rate sensitivity of the alloy from other authors is also presented in the following figures. With the range of strain rate found in the literature that is less or around 10^3 s^{-1} , the data from current work agrees with the literature data. A very less data is available at strain rates higher than 1000 s^{-1} therefore at higher strain rates comparison is made by fitting Johnson-Cook model. The model fitted data with little variation verifies the experimental data in all loading orientations.

Figure 5-17: Strain rate sensitivity of the alloy in normal direction

Figure 5-18: Strain rate sensitivity of the alloy in rolling direction

Figure 5-19: Strain rate sensitivity of the alloy in transverse direction

Figure 5-20: Strain rate sensitivity of the alloy in 45° loading direction

Figure 5-21: Stress as a function of homologous temperature

Stresses at 5%, 10% and 15% plastic strains have been plotted versus homologous temperature in Fig. 5-21. As can be seen in the Fig. 5-21, the alloy shows thermal softening as the temperature increases from -30°C to 200°C. However, the smaller effect is noted for temperature change from -30°C to 25°C as compared to what is seen for further increase in the temperature up to 200°C. The rate sensitivity exponent ' m ' is plotted against temperature in Fig. 5-22. The rate sensitivity of AZ31B is considerably affected by the temperature. In general, the rate sensitivity exponent m is lower at -30°C than 25°C, indicating that the alloy is more rate sensitive at room temperature than below zero. At 200°C, the rate sensitivity is lower than that at room temperature showing the softening of the alloy at elevated temperature.

Figure 5-22: Temperature dependence of strain sensitivity exponent m

5.1.4 Work Hardening Rate

Fig. 5-23 presents the work hardening rate ($d\sigma/d\varepsilon$) versus strain at various strain rates and temperatures. For a fix strain rate, the work hardening rate for each loading direction decreases rapidly at smaller strains (<2%), after which it gradually increases with increasing strain. This increase in the work hardening rate is the result of strain hardening of the alloy that occurs during deformation. It is also observed that the work hardening rate is slightly higher at higher strain rates except in normal loading direction, where it is lower for $3500s^{-1}$ as compared to $2500s^{-1}$. At larger strains, little dependence of work hardening rate on the strain rate is observed and this happens because at higher strain rates and larger strains thermal softening takes over the work hardening and alloy shows softening behaviour.

Figure 5-23: Work hardening rate as a function of true strain at different strain rates

Figure 5-24: Work hardening rate as a function of true stress at different strain rates

Fig. 5-24 illustrates the work hardening rate ($d\sigma/d\varepsilon$) as a function of stress at different strain rates. It is observed in all loading directions that the work hardening rate decreases rapidly in the beginning and then decreases gradually before dropping to zero. The work hardening rate is slightly higher for larger strain rates. However, in normal and transverse loading it is lower for 3500s^{-1} as compared to 2500s^{-1} and this tally with lower stress level seen in Fig. 5-2 and Fig. 5-4 for normal and transverse loadings at 3500s^{-1} . The stress at zero work hardening rate (maximum flow stress) increases with increasing strain rate. However, it is lower in normal and transverse loading directions at 3500s^{-1} that shows the reduction in strength of the alloy in normal and transverse direction. Increase in maximum flow stress with increasing strain rate indicates the improved hardening ability of the alloy during deformation, a positive attribute of AZ31B alloy for its use as structural material.

Work hardening rate versus true strain at -30°C and 200°C temperatures is shown in Fig. 5-25. The work hardening rate is comparatively higher at -30°C than at 200°C . At -30°C , the work hardening rate slightly increases with strain and then gradually decreases, while at 200°C it shows a gradual decrease with increasing strain after about 1% strain. The work hardening rate varies a little with strain at 200°C indicating that the alloy undergoes softening and deforms more uniformly at elevated temperatures. Work hardening rate as a function of true stress is shown in Fig. 5-26. The work hardening rate is higher for -30°C as compared to 200°C . A significant decrease in the stress at zero work hardening rate is seen as temperature increases from -30°C to 200°C indicating a reduction in strength of AZ31B at 200°C . It also indicates that the ability of the alloy to resist load (strengthening during deformation) reduces at elevated temperature.

Figure 5-25: Work hardening rate as a function of true strain at different temperatures

Figure 5-26: Work hardening rate as a function of true stress at different temperatures

5.1.5 Specimen strain

The effect of strain rate and temperature on the strain of the alloy is shown in Fig. 5-27 and Fig. 5-28 respectively. Specimens at 500s^{-1} and 1000s^{-1} did not fail and the strain mentioned in Fig. 5-27 for these specimens is not the failure strain. The mentioned dots of strain corresponding to 500s^{-1} and 1000s^{-1} are only the strain experienced by the specimens at these strain rates until the loading is finished. Specimens tested at 1500s^{-1} and above all failed and the failure strain was observed to increase with increase in the strain rate from 1500s^{-1} to 3500s^{-1} . However, for normal and transverse loading directions, the failure strain is decreased at 3500s^{-1} strain rate.

Figure 5-27: Effect of strain rate on the compressive failure strain

Figure 5-28: Effect of temperature on the compressive strain at $2250\pm 50\text{s}^{-1}$

In Fig. 5-28, the failure strain increases with increasing temperature from -30°C to 25°C. Increase in the failure strain at higher strain rates as compared to quasi-static strain rates is likely because of thermal softening due to increase in the temperature of specimen during deformation as explained in section 5.1.2. Specimen deformed at 200°C temperature did not fail even at dynamic strain rates and the strain mentioned in Fig. 5-29 is the strain until the loading is finished at a particular strain rate and temperature. The fracture mode of specimens tested at 3500s⁻¹ as can be seen in Fig. 5-29 also reveals that the specimens are plastically deformed due to rise in temperature. Increased strain to failure at impact strain rates is worthwhile property of AZ31B and an essential requirement in crash events.

Figure 5-29: Specimens fractured at room temperature (Upper) and at 200°C (lower)

Like AZ91D, AZ31B specimens also fail by ductile shear fracture at higher strain rates. In AZ31B specimens deformed at 1500s⁻¹ and above, the compression cone

can be seen in Fig. 5-29. It is also observed that the shear fracture is approximately at 45° to the loading direction. At higher strain rates i.e. 3500s^{-1} , multiple fractures occurred at 45° to the direction of force. Different types of fractures are observed in different loading directions indicating the anisotropy in deformation. It is also worth to note that the specimens in all loading directions are more deformed in one direction (x) as compared to other (y) direction signifying the directional nature of wrought AZ31B alloy. Deformed specimens are joined together to the best possible effort and x, y dimensions are measured. The (x/y) ratio in normal specimen is minimum (1.6) and in rolling direction is maximum (1.82). Specimens at 200°C are not fractured in any loading direction even at $2250\pm 100\text{s}^{-1}$ strain rate specifying the increased ductility of the alloy at 200°C . The (x/y) ratio at 200°C is lesser, it varies between 1.2 (45°) and 1.5 (normal) indicating a decrease in the anisotropy of the alloy.

5.1.6 Energy Absorption Capability and Anisotropic Behaviour

The energy absorbed by the alloy during compression was calculated using Eq. (4.3) and plotted in Fig. 5-30 as a function of strain rate and in Fig. 5-31 it is plotted against homologous temperature for different loading directions. At room temperature, the overall energy absorption increases with increasing strain rate; it is about 2.5 to 3 times higher for 2500s^{-1} strain rate as compared to the energy absorption at quasi-static strain rate in all four loading directions. However, as the strain rate is increased to 3500s^{-1} , the energy absorption in normal and transverse loading direction decreases. As the energy absorption is calculated from the area under the stress-strain curve therefore the decrease in the energy absorption is in accordance with the stress-strain results.

Figure 5-30: Strain rate dependance of energy absorption at room temperature

Figure 5-31: Temperature dependance of energy absorption

As the specimens are not broken at 200°C and energy is calculated from area under the whole stress-strain curve at 2250s⁻¹. Still the comparison shown in Fig. 5-31 can give a good approximation of temperature dependence of energy absorption. The energy absorption capability of the alloy at 200°C is significantly lower as compared to -30°C. A minimum decrease of 10% is observed in normal direction and a maximum decrease of 27% is noticed in transverse direction. It can be estimated that even if the specimen is loaded up to failure the energy absorption will be substantially lower at elevated temperature as compared to lower temperature.

To analyse the anisotropy, stress-strain curves are plotted in Figures 5-32 to 5-34 in all four loading directions at 1000s^{-1} , 2400s^{-1} , 3500s^{-1} strain rates.

Figure 5-32: Anisotropic behaviour of AZ31B at room temperature (1000s^{-1})

Figure 5-33: Anisotropic behaviour of AZ31B at room temperature (2400s^{-1})

Figure 5-34: Anisotropic behaviour of AZ31B at room temperature (3500s⁻¹)

From these figures, it is clear that the alloy is anisotropic and anisotropy is strain and strain rate dependent. At 1000s⁻¹ strain rate, the stress level is highest in normal loading and minimum in 45° angle loading. At strain rates of 2400s⁻¹ and 3500s⁻¹, approximately up to 10% strain the stress level in normal loading direction is higher as compared to other loading directions. While, in other loading directions the alloy did not show any monotonic trend in stress level.

Figure 5-35: Anisotropic behaviour of AZ31B at -30°C

Figure 5-36: Anisotropic behaviour of AZ31B at 200°C

Fig. 5-35 and Fig. 5-36 show anisotropic behaviour at -30°C and 200°C under dynamic compressive loading. It is useful to analyse the anisotropy in properties of the alloy under the combined effect of strain rate and temperature. It will be helpful for evaluating the material behaviour in situations such as crash where both impact rates and increasing temperatures simultaneously occur during deformation. The alloy shows anisotropy in stress and strain at all temperatures. However, the alloy did not show significant anisotropy at lower temperatures as compared to elevated temperature. At 200°C, the alloy showed similar trend in 45°, rolling and normal loading up to 5% strain while after 5% strain variation in stress is significant. The strain is little affected by loading orientations at 200°C. A quantitative analysis of anisotropy in compressive yield strength and peak compressive strength at room temperature is given in Fig. 5-37. The compressive yield strength in the normal direction at 2500s⁻¹ strain rate is nearly twice the yield strength in other three loading directions. The peak compressive strength is less affected by loading orientation. However, the anisotropy in peak compressive strength is relatively higher in the dynamic range of strain rate as compared to quasi-static one.

Figure 5-37: Anisotropy in the compressive strength of AZ31B under various strain rates at room temperature

The alloy also shows anisotropy in energy absorption. It is observed from Fig. 5-30 and Fig. 5-31 that at higher strain rates, the anisotropy in energy absorption is larger as compared to lower strain rates and quasi-static strain rate. However, the anisotropy in the energy absorption decreases as the temperature increases. It is considerably lower at 200°C as compared to -30°C and room temperature.

Figure 5-38: Hardening exponent as a function of strain rate at room temperature

The average hardening exponent and the average rate sensitivity of the alloy in four loading directions are plotted in Fig. 5-38 and Fig. 5-39. Beyond 1000s⁻¹, the hardening exponent n is nearly same (~ 0.9) in transverse, rolling and 45° loading directions while it is quite low (0.4 to 0.6) in normal direction. The lower hardening exponent in normal direction is possibly due to a balance between

hardening (strain rate effect) and softening (due to rise in specimen temperature) at larger strains in the plastic region of deformation. Average strain rate sensitivity versus loading directions is shown in Fig. 5-39. It is highest in normal loading direction, while it is nearly equal in all other orientations of loading. However, for strain rates higher than 2500s^{-1} the alloy showed negative strain rate sensitivity in all direction of loadings. This trend shows that the alloy AZ31B experiences a decrease in stress-level at higher strain rates and it is more pronounced in the case of normal loading.

Figure 5-39: Strain rate sensitivity in different orientations of loading

Work hardening rate in different loading direction is plotted against true strain and true stress in Fig. 5-40 and Fig. 5-41 respectively. At 1000s^{-1} strain rate the work hardening rate is slightly higher in normal loading as compared to other three loading directions. At 2500s^{-1} , the lower work hardening rate after 4% strain in normal direction also confirms the negative rate sensitivity of AZ31B in normal direction. The stress dependence of work hardening rate is not significantly different at 1000s^{-1} . However, it is significantly higher in normal loading in the beginning at 2500s^{-1} as compared to other three loading directions implying the higher hardening ability of the alloy in normal loading at lower

strains. At higher impact rates and larger deformations, the alloy AZ31B shows better resistance to impact in rolling, transverse and angle directions as compared to the normal loading.

Figure 5-40: Work hardening rate versus true strain in different loading directions

Figure 5-41: Work hardening rate versus true stress in different loading directions.

5.2 Tensile Behaviour

In this section, the experimental results for high strain rate tensile testing of AZ31B alloy are presented in rolling and transverse directions at room temperature and 250°C. The strain rate and temperature dependence of the alloy's tensile behaviour is investigated. To ensure the repeatability of the results, 3 to 4 tests were performed for each strain rate and the resultant stress-strain curves are shown in appendix D. Microscopy is performed and experimental results are explained on the basis of microstructure.

5.2.1 Strain Rate and Temperature Effects

Tensile stress-strain curves for tests performed in the range of strain rate between 500s^{-1} and 1400s^{-1} in rolling and transverse direction are shown in Fig. 5-42 and Fig. 5-43 respectively.

Figure 5-42: Effect of strain rate on the tensile behaviour of AZ31B (rolling direction)

Figure 5-43: Effect of strain rate on the tensile behaviour of AZ31B (transverse direction)

In both rolling and transverse directions, higher stresses are observed as the strain rate increases. Almost two times higher stresses are observed at 1400 s^{-1} than those observed at quasi-static strain rates in both loading directions. It should be noted that the specimens tested in the current study did not fail at any strain rate in both rolling and transverse direction. Therefore, the failure strain cannot be analysed and compared. However, it is observed that the total strain (total strain is the strain, the specimen undergoes till the end of the applied load) increases roughly twice as the strain rate increases from 500 s^{-1} to 1000 s^{-1} and thrice as the strain rate and from 500 s^{-1} to 1400 s^{-1} . Table 5-5 and Table 5-6 show the quantitative analysis of mechanical properties of AZ31B in rolling and transverse directions respectively. The yield stress and ultimate tensile stress increase monotonically with increasing strain rate in the range between 500 s^{-1} and 1400 s^{-1} . As the specimens did not fail hence the absorbed energy was calculated up to UTS (area under the curve up to the strain at UTS). Therefore, the energy absorption data presented in the tables carries little significance.

Table 5-5: Tensile mechanical properties of AZ31B alloy in rolling direction

Magnesium alloy AZ31B (Rolling loading)				
Strain rate $\dot{\epsilon}$ [s^{-1}]	$10^{-4}(1)^i$	550(4)	1000(4)	1400(4)
Tensile yield strength [MPa]	120	126.7 ± 5.0^{ii}	139.5 ± 3.5	181.5 ± 4
UTS [MPa]	286.2	308 ± 3	351 ± 3.5	396 ± 8.8
*Absorbed Energy (MJ/m^3)	45	29.4 ± 1.3	35.2 ± 1	42.5 ± 0.95

Table 5-6: Tensile mechanical properties of AZ31B alloy in transverse direction

Magnesium alloy AZ31B (Transverse loading)				
Strain rate $\dot{\epsilon}$ [s^{-1}]	$10^{-4}(1)^i$	550(4)	1000(4)	1400(4)
Tensile yield strength [MPa]	122	130.7 ± 2.8^{ii}	140.7 ± 4.0	185 ± 3.5
UTS [MPa]	265.8	288 ± 6	364 ± 2.5	414 ± 3.8
*Absorbed Energy (MJ/m^3)	28	12.5 ± 0.5	28.3 ± 0.5	55.6 ± 0.5

ⁱ Number in brackets () indicates the number of tests conducted for each experiment

ⁱⁱ Mean $\pm$ standard error, * the energy absorbed by the specimens is calculated up to UTS

Fig. 5-44 and Fig. 5-45 illustrate the effect of temperature on the stress-strain behaviour of the alloy at quasi-static and $1400s^{-1}$ strain rates for rolling and transverse directions respectively. Mechanical properties are quantified and tabulated in Table 5-7.

Figure 5-44: Effect of temperature on the tensile behaviour of AZ31B (rolling direction)

Figure 5-45: Effect of temperature on the tensile behaviour of AZ31B (transverse direction)

Table 5-7: Tensile mechanical properties at 250°C and 1400s⁻¹

AZ31B	YS [MPa]	UTS [MPa]	*Energy absorbed [MJ/m ³]	Total strain
Rolling	115	274.75	22.4	15.0
Transverse	100.5	273	18.35	15.95

* The energy absorbed by the specimens is calculated up to UTS

The alloy is sensitive to temperature at both quasi-static and 1400s⁻¹ strain rate. In comparison to room temperature at 1400s⁻¹ strain rate the yield strength and ultimate tensile strength are substantially affected with a percentage reduction of 37% and 30% for rolling direction and 45% and 36% for transverse direction respectively. However, the total strain (the strain a specimen experienced at a certain strain rate and temperature until the loading is finished) is not much affected at high strain rate; it increases by 10% and 5% at 250°C as compared to the room temperature for rolling and transverse direction respectively.

5.2.2. Specimen Strain and Energy absorption

Strain and energy absorption are quantified and tabulated in Tables 5-5, Table 5-6 and Tables 5-7. At dynamic strain rates, tensile specimens did not break even at

1400s^{-1} therefore failure strain cannot be estimated and the total strain experienced by the specimen at a certain strain rate is considered. At 1400s^{-1} , the total strain in the rolling direction is smaller than the strain to failure at quasi-static strain rate and is slightly larger in transverse direction. The elongation of tested specimens in rolling and transverse direction is about 1.4 ± 0.2 mm (13% to 16% when gauge length is 10mm). No obvious necking is seen in tested specimens in both loading directions as observed in Fig. 5-46. However, the energy absorption at 1400s^{-1} calculated up to UTS is comparable to quasi-static strain rate in rolling direction and almost twice in case of transverse direction. These experimental observations help to conclude that the failure strain (in case of fracture) and the energy absorption are higher at higher strain rates as compared to the quasi-static strain rate.

Figure 5-46: Tensile specimens used in current study (specimen tested at 1400s^{-1})

5.2.3 Work Hardening

The effect of strain rate on the hardening behaviour of the alloy is shown in Fig. 5-47 and Fig. 5-48 in rolling direction and transverse direction respectively. The hardening exponent n increases with strain rate from 0.2439 at 550s^{-1} to 0.3169 at 1400s^{-1} in rolling direction and from 0.3609 at 550s^{-1} to 0.3771 at 1400s^{-1} in transverse direction.

Figure 5-47: Effect of strain rate on the hardening behaviour of AZ31B (rolling direction)

Figure 5-48: Effect of strain rate on the hardening behaviour of AZ31B (transverse direction)

The average hardening exponent (averaged over three strain rates considered in this study) is higher (0.377) in transverse direction as compared to the average hardening exponent (0.2862) in rolling direction. Strain hardening occurs during deformation due to increased dislocation density that will be explained later

5.2.4 Strain Rate Sensitivity

The rate sensitivity of the alloy is given by Fig. 5-49 and Fig. 5-50 in rolling and transverse directions respectively, where stress is plotted against strain rate at various strain levels. A comparison of rate sensitivity of the alloy obtained from current experiments is compared with that from literature and Johnson-Cook

model. The data from current work is observed to be in agreement with the literature data and model fitted data with slight variation that may be due to slight variation in composition of the alloy, the specimen size etc.

Figure 5-49: Strain rate sensitivity of AZ31B (rolling direction)

Figure 5-50: Strain rate sensitivity of AZ31B (transverse direction)

A linear relationship exists between stress and strain rate for both loading directions. However, the slope of the fitted lines is steeper in the dynamic range after 500s^{-1} as compared to what is observed between 10^{-4}s^{-1} and 500s^{-1} . The rate sensitivity exponent m in the dynamic range increases with strain for both rolling and transverse directions with larger m in transverse direction.

5.2.5 Work Hardening Rate

Work hardening rate is plotted against true strain and true stress at different strain rates for rolling and transverse direction in Fig. 5-51 and Fig. 5-52 respectively. The work hardening rate in quasi-static tests has excessive oscillations therefore a 20th order polynomial is fitted as shown in the figures.

Figure 5-51: Work hardening rate a a function of true strain

Figure 5-52: Work hardening rate a a function of true stress

It is noted in Fig. 5-51 that the work hardening rate in tension drops continuously with increasing strain before it becomes negative. In general, work hardening rate is higher at higher strain rates in both loading directions indicating the alloy's ability to become stronger as the deformation proceeds.

Figure 5-53: Stress-strain curves with Considere point [105]

The Considere criteria [105] applied to the rolling and transverse direction and experimental data is presented in Fig. 5-53. In tensile loading, especially at high strain rates it is important to see the effect of strain rate and temperature on the onset of necking. In the current study, AZ31B tensile specimens are not broken but only deformed plastically with small amount of necking. Considere criteria can be used to estimate the onset of necking (Eq. 4-6) and the effect of strain rate on it. It can be observed from Fig. 5-53 that the strain rate has a considerable effect on the onset of necking (Considere point/strain where necking starts) in both rolling and transverse direction. Larger Considere strain at higher strain rates also helps to conclude that the alloy will experience larger failure strains at higher strain rates.

5.2.6 Anisotropic Behaviour in Tension

The stress-strain curves in rolling and transverse directions are plotted in Fig. 5-54 to analyse the anisotropic behaviour of the alloy.

Figure 5-54: Anisotropic behaviour under tensile loading

Figure 5-55: Effect of loading orientation on the yield strength of AZ31B

For a fix strain, slightly higher stresses are observed for transverse direction as compared to the rolling direction in the dynamic range of strain rate. It is also clear from Fig. 5-55 that the yield strength is little affected by the orientation of loading with slightly higher (4 to 5%) values in transverse direction. At quasi-static strain rate, the alloy is more ductile in the rolling direction with approximately 40% higher failure strain in rolling direction.

Figure 5-56: Effect of loading orientation on the hardening exponent n

The average hardening exponent n and average rate sensitivity exponent m are plotted as a function of orientation of loading in Fig. 5-56. Higher values of n and m are noticed for transverse direction as compared to rolling direction. Under tensile loading, AZ31B alloy shows relatively higher ability to resist the deformation in transverse direction as compared to the rolling direction.

5.3 Compressive-Tensile Yield Asymmetry

Fig. 5-57 and Fig. 5-58 show the stress-strain behaviour of AZ31B alloy in rolling and transverse direction both in compression and tension at several strain rates.

Figure 5-57 Stress-strain curves in compressive and tensile loading (rolling direction)

Figure 5-58: Stress-strain curves in compressive and tensile loading(transverse direction)

Usually magnesium alloys exhibit yield asymmetry between tension and compression. In the current study, about twice higher yield strength in tensile loading is observed than those in compressive loading in both rolling and transverse directions. It is seen that the yield strength under tension strongly depends on the strain rate while in compression it is less dependent on the strain rate. It can be observed from Fig. 5-36 and from figures 5-44 and 5-45 that under the same strain rate, the compressive-tensile asymmetry observed at low strain rates and room temperatures is also noticed at high strain rates and elevated temperatures.

Figure 5-59: Tensile-compressive yield asymmetry in the rolling direction

Figure 5-60: Tensile-compressive yield asymmetry (transverse direction)

A quantitative analysis of compressive-tensile yield asymmetry is given in Fig. 5-59 and Fig. 5-60 for rolling and transverse directions. Around 70% to 128% higher stresses are observed in tensile loading in rolling direction and even further higher in case of transverse direction. This asymmetry in the compressive-tensile yield stress is rate sensitive and increases with strain rate, almost 175% higher at $1400s^{-1}$ than that of what is seen at $550s^{-1}$ for both rolling and transverse direction. The hardening and the rate sensitivity exponents are plotted in Fig. 5-61 for compressive and tensile loadings. The hardening exponent n is lower for tensile loading in both rolling and transverse directions. However, the rate sensitivity exponent m is slightly higher under tension.

Figure 5-61: Hardening exponent under compressive and tensile loading

5.4 Microstructural Analysis

5.4.1 Compression

Microstructural analysis was performed using optical microscope and scanning electron microscopy (SEM). Microstructure of AZ31B alloy at various strain rates and temperatures is investigated and correlated with the stress-strain behaviour of the alloy.

5.4.1.1 Strain Rate Effects

Fig. 5-62 to Fig. 5-65 show the microstructure of specimens tested in four loading directions respectively at 10^{-4}s^{-1} , 500s^{-1} and 2500s^{-1} strain rates. The Fig. 5-62 shows the microstructure of the alloy in 45° orientation. Lots of twins can be seen at 10^{-4}s^{-1} strain rate and the twin density decreases at 500s^{-1} strain rate and even fewer twins are observed at a strain rate of 2500s^{-1} . Some micro voids (shown by red arrows in Fig. 5-62) nucleated at grain boundaries can be seen both at quasi-static and higher strain rates. Some randomly dispersed minor phases are also observed. At 2500s^{-1} , partial grains refinement is observed (Fig. 5-62 e, f) with grains of 6 to 8 μm size.

Figure 5-62: Strain rate effects on the microstructure in 45° loading direction; (a, b and c) uniformly distributed voids at grain boundaries (d and f) presence of twins (e) uniformly distributed minor phases

Microstructures for AZ31B tested in normal direction are shown in Fig. 5-63. A large number of twins are seen at quasi-static strain rate and number of twin decreases as the strain rate increases. Some double and intersecting twins can be seen at all strain rates. It is observed that the twins are thickened and this thickening usually takes place when twinning dislocations spread on adjacent planes. Voids are seen at all strain rates and micro-cracks parallel to twins and along the grain boundaries are observed at 2500s⁻¹ strain rates that seems to be the result of coalescence of voids. Some localized adiabatic shear bands are also formed as well as grain refinement is observed in the specimen tested at 2500s⁻¹. Under dynamic loading there is little time available for heat converted from plastic work to dissipate from the material, so the dynamic deformation is

considered to be adiabatic. The refined grains are about 3 to 5 microns which is 3 to 4 times smaller than the grain size (15 to 20 microns) at quasi-static strain rate. The grains within the shear bands are severely deformed and not clearly identifiable. Microstructures at 3500s^{-1} shows increased twin density with lots of double twins formed. Beside thicker twins the large grains are seen at 3500s^{-1} that favours twinning.

Optical microscopic pictures

SEM pictures

Figure 5-63: Strain rate effects on the microstructure in normal direction; (a) lots of twins (b) elongated grains and presence of twins (c) shear band and refined grains (d) thick twins (e) coalescence of voids (f) twins and minor phases present (g) and (h) thicker twins and intersecting twins.

Fig. 5-64 shows the effect of strain rate on the microstructure of AZ31B in rolling direction. Elongated grains and twins are observed at higher strain rates. A large number of twins can be seen at quasi-static strain rate and number of twins decreases as the strain rate increases. Twins almost disappeared at 2500s^{-1} and grains are deformed through conventional plastic flow. The alloy partly melts at high strain rates and deforms plastically due to thermal effects.

Figure 5-64: Strain rate effects on the microstructure in rolling direction; (a, b) twins formation (c) refined grains and voids (d) elongated grains and twins (e and f) deformation of grains by plastic flow

Fig. 5-65 shows the micrographs of specimen tested in transverse direction at quasi-static strain rate, 500s^{-1} and 2500s^{-1} . More twins are seen at quasi-static strain rate as compared to higher strain rates. Elongated grains and twins are seen at higher strain rates. A few numbers of double and intersecting twins are seen at all strain rates. Unlike other orientations, in transverse direction a considerable number of twins are present even at 2500s^{-1} . At 2500s^{-1} , partial refinement of grains occurred with smaller grains of 4 to 8 microns.

Optical microscopic pictures

SEM pictures

Figure 5-65: Strain rate effects on the microstructure in transverse direction; (a, b and c) twins and voids seen (d) elongated grains and twins (e) uniformly distributed minor phases and (f) minor phases broken and dispersed (g) and (h) fewer twins

5.4.1.2 Temperature Effects

Micrographs of specimens tested in four different loading directions at 10^{-4}s^{-1} and $2250\pm 100\text{s}^{-1}$ strain rates and -30°C and 200°C temperatures are shown in Fig. 5-66 to Fig. 5-69. Twinning is a dominant mode of deformation at low temperature. In general, at quasi-static strain rate, lots of twins are observed at -30°C in all orientations of loading and hardly any twins are observed at 200°C temperature except in rolling direction where fewer twins are seen (Fig. 5-68d). Adiabatic shear bands are formed at quasi-static strain rate in 45° angle, normal and transverse loading. Within the shear bands, grains are elongated and severely distorted eventually resulting in the formation of voids. Large number of twins are observed at -30°C and 2250s^{-1} in all orientations of loadings except in 45° where only few twins are observed. At 200°C and 2250s^{-1} , twins are drastically reduced in all loading directions with almost no twins in transverse loading direction. At 200°C , grains are plastically deformed and elongated perpendicular to the compressive loading direction. In magnesium alloys, the dynamic recrystallization starts around 225°C [71] and the current study indicates the presence of dynamic recrystallization phenomena with the formation of equiaxed grains at 200°C and quasi-static strain rate and partial dynamic recrystallization at 200°C and 2250s^{-1} . It is possibly due to rise in temperature, as explained in section 5.2.3, of specimen deforming at high strain rate that has contributed to the recrystallization process.

Figure 5-66: Temperature effects on the microstructure in 45° loading direction; (a) lots of twin (b) shear bands (c) twins (d) shear band and voids formation due to plastic deformation (e) fewer twins and voids (f) elongation of grains (g) fewer twins (h) fewer twins with recrystallized grains

Figure 5-67: Temperature effects on the microstructure in normal direction; (a) twins (b) shear bands (c) twins (d, e) shear bands and elongated grains (f) plastically deformed grains (g) thick twins across boundaries (h) twins

Figure 5-68: Temperature effects on the microstructure in rolling direction; (a)twins (b) grain refinement and increase in grain boundaries (c) twins (d) broken grains (e) multiple twins and twins accross grain boundaries (g) thicker twins and (f and h) twins within grains

Figure 5-69: Temperature effects on the microstructure in transverse direction; (a and e) twins and evenly distributed voids (c) twins (d) plastically deformed grains within shear band (b) adiabatic shear bands (g) fewer twins (h) recrystallized grains

5.4.1.3 Energy Dispersive X-Ray Analysis (EDX)

EDX was also performed for some selected samples for elemental analysis and results are shown in Fig. 5-70. Some minor phases rich in Mn (Mg/Al/Mn, Mg/Mn, Al/Mn) and Si (Mg/Si, Al/Si) are randomly dispersed in grains and within grain boundaries. In these minor phases, the percentage of Mn varies between 26 to 77 % and Si varies from 11 to 48%. Si-rich particles are very scarcely seen and usually found within grain boundaries, therefore these might be due to some foreign inclusion of Si during specimen preparation. However, the exact reason is unclear. These brittle and hard phases seem to contribute positively to the strength of the alloy while having a negative effect on the ductility of the alloy.

Processing option: All elements analyzed
(Normalized)

Spectrum	Mg	Al	Si	Mn
Spectrum 1	3.78	19.35		76.86
Spectrum 2	100.0			
Spectrum 3	50.25		49.75	
Spectrum 4	100.0			
Spectrum 5	100.0

Processing option: All elements analyzed
(Normalized)

Spectrum	Mg	Al	Mn
Spectrum 1	75.74	24.26	
Spectrum 2	39.55	54.50	5.96
Spectrum 3	70.34	29.66	
Spectrum 4	17.72	56.25	26.03
Spectrum 5	86.44	13.56	
Spectrum 6	93.38	6.62	
Spectrum 7	100.0		
Spectrum 8	100.0		

All results in weight%

Figure 5-70: EDS analysis; (a) Mn and Si rich phases (b) and (c) Mn-rich phase (d) Si-rich phase present at grain boundaries

5.4.2 Tension

Microstructure analysis carried out for selected tensile specimens at quasi-static and 1500s^{-1} strain rate. Tests were also performed at 25°C and 250°C temperatures. EDS analysis was also performed for selected specimens.

5.4.2.1 Strain Rate Effects

Fig. 5-71 to Fig. 5-73 show the micrographs of AZ31B alloy in rolling and transverse directions tested under tensile loading at 10^{-4}s^{-1} and 1400s^{-1} respectively. A large number of twins are seen at quasi-static strain rate both in

rolling and transverse loadings. Some double and intersecting twins are observed in transverse directions. It is also observed that the twins have changed the direction across the grain boundaries which may have caused by twin slip phenomena (Fig 5-72 f). In rolling direction; twins are observed to collate to produce cracks. At quasi-static strain rates, more number of twins and cracks has given relatively higher ductility to the alloy in rolling direction. On the other hand, the presence of double twins might have contributed to the lower ductility in transverse direction. At higher strain rate of 1400s^{-1} , still a good number of twins are present in both orientations. However, the alloy in rolling direction exhibits relatively more number of twins as compared to transverse direction. Some micro-voids (*shown by red arrows*) of 1-2 micron size are formed in Mn rich (MgAlMn) phase as shown in Fig. 5-75 c. Some minor phases rich in Mn (MgAlMn) and Si (MgSi) are found to be distributed within the grain boundaries. Mn varies between 20 to 30% and Si is found between 30 to 40%. It is also observed that Mn-rich phase is mostly found in rolling direction and Si-rich phase is mostly formed in transverse direction. However, the reason for this is not clear.

Figure 5-71: Optical microscopic pictures for specimens tested at 10^{-4}s^{-1} (a) 10x, voids dispersed (b) 50x, voids at grain boundaries (c) 10x and (d) 50x, twins and voids

Figure 5-72: SEM pictures for specimens tested at 10^{-4}s^{-1} (a) grain boundaries and twins within grains (b) dislocation pile up (c) deformed grains (d) twins within grains (e) voids (f) twin slip along grain boundaries

Figure 5-73: SEM pictures for specimens tested at 1400 s⁻¹; (a) to (c) Twins and micro-voids (d) to (f) larger grains are plastically deformed.

5.4.2.2 Temperature Effects

The microstructure at 250°C and 1400s⁻¹ in rolling and transverse directions are shown in Fig. 5-74. Lots of twins are observed at in rolling direction even at 250°C while fewer twins are seen in transverse direction. Grain refinement occurs in transverse direction along with the formation of micro-voids along the grain

boundaries. Shear bands and plastically deformed grains are observed in transverse direction.

Figure 5-74: SEM pictures for specimens tested at 250°C and 1400s⁻¹ (a) to (c) lot of twins (d) refined grains, fewer twins and micro-voids (e) and (f) fewer twins, larger grains are plastically deformed.

5.4.2.3 EDX Analysis

Fig. 5-75 shows the results from EDX analysis for AZ31B alloy. Minor phases such as Mn-rich and Si-rich phase are found in grains and grain boundaries. Mn varies between 20% (Mg/Al/Mn/Zn) to 30% (Mg/Al/Mn) and Si is about 36% to 42% (Mg/Si). As in compression Si seems to be an inclusion introduced from grinding or polishing.

Figure 5-75: EDX analysis for specimens tested in tension; (a) and (b) Mn-rich phase (c) and (d) Si-rich phase can be seen

5.5 Discussion

In general, stress and failure strain increase with increasing strain rate under compressive deformation of AZ31B. However, orientation of loading affects these properties. A decrease is observed in the stress level in normal and transverse loading directions at 3500s^{-1} and it is due to thermal softening caused by the adiabatic heating of the specimen during deformation. Yield stress is highly rate sensitive in normal loading while little sensitive in other loading directions at room temperature. Under tensile loading, the dynamic yield stress is strongly rate sensitive and increase significantly with increasing strain rate. At -30°C , the mechanical behaviour of the alloy is little different from room temperature

indicating that the alloy behaviour is not affected by the temperature in the range below zero. At elevated temperatures, the compressive stresses decreased appreciably indicating higher sensitivity of the alloy to elevated temperatures. Under tension, the decrease in stresses is not as much as that observed in compression.

An interesting thing worth to note is that the concave up stress strain curves in rolling, transverse and 45° loading orientations give the impression that might some shock phenomenon has taken place in these experiments. As the rate-independent theory of plastic wave propagation assumes that the stress-strain curve is concave toward the strain axis (the slope is a decreasing function of strain amplitude) and thus concave down. During high velocity impacts, a shock wave may form and propagate thus producing concave up stress-strain curves. However, shock waves are rarely formed in uniaxial stress state; they are generally found in uniaxial strain state as in plate impact experiments where the impact velocities are 2 km/s or greater [111]. In the present study, the concave up stress-strain curves are characteristic curves of AZ31B alloy in rolling, transverse and 45° loading orientations. Even the stress-strain curves at 10^{-4} s^{-1} , where the speed of loading is very small (a few mm per minute), are concave up further supporting that this phenomena belongs to the alloy's behaviour in particular loading orientations. The concave up behaviour of the stress-strain curves along rolling, transverse and 45° directions also reported in the literature [65, 40, 70]. This particular concave up behaviour of the alloy is due to strong work hardening after an initial low strain hardening region attributed to strong twinning activity. The grain boundary sliding and twin slip at grain boundaries are also possible reasons for this significant variation in the hardening behaviour of

the alloy. The concave down stress curves in normal loading at all strain rates as seen in present study (Fig. 5-1) indicates the deformation is mostly accommodated by slip along the basal and pyramidal planes also observed by Tucker [68] for AZ31, for rolled magnesium by Kelley and Hasford [112] and for extruded magnesium by Barnett [41-42]. The work hardening rate ($d\sigma/d\varepsilon$) has different stages of strain hardening; a rapidly decreasing, gradually increasing followed by the gradual decreasing work hardening rate. The slip is the only mechanism in first stage, in second stage twining is dominant and again twining is severely limited in third stage where the hardening rate decreases. This is also in accordance with the initial low strain hardening portion of the stress-strain curves (slip dominates) followed by the high strain hardening portion of the curve (slip and twining) and low strain hardening region of the curve (slip dominates again) next. By the time the third stage of deformation begins, sufficient twining has occurred and the twinned regions turn into softer orientations that favour slip and eliminate the need of more twining to accommodate deformation.

Magnesium alloys, in general have different mechanisms of deformation at quasi-static and higher strain rates. At room temperature, the alloy deforms by slip on basal planes $\langle 11\bar{2}0 \rangle$ and by twining on the pyramidal planes $\{10\bar{1}2\}$ [72]. At elevated temperatures and high strain rates, the alloy deformed by slip and twining, twins can be seen in figures 5-65 -5-69, 5-73 and 5-74. However, some non-basal slip systems such as pyramidal slip become active responsible for increase in ductility of the alloy. AZ31B work-hardens at elevated temperature about 200°C to 250°C by deformation due to twining. In magnesium alloys, deformation twins are very common and play important role in their mechanical

behaviour. In compression, lots of twins are formed at quasi-static and low strain rates even at elevated temperatures. At room temperature and lower strains, there is significant twinning with low strain rate-insensitivity while increased strain rate sensitivity at larger strains supports the conclusion that the dominant deformation mechanism is slip. In high strain rate deformation of AZ31B, it is observed that the twin density decreases and twins are uniformly distributed in small grains as the strain rate increases. Decreasing twin density is also observed in AZ31 [64-65] and Zk60 [61]. Fewer twins at higher strain rates may be attributed to the localization of the flow. According to Barnett et al. [40] and Jiang et al. [113], twinning should belong to $\{10\text{-}12\}$ which has a strong influence on the work hardening rate. The stronger work hardening as seen in 45° angle, rolling and transverse loading directions in the present study is due to increased number of intersecting twins. These $\{10\text{-}12\}$ extension twins are also connected with the concave up curves with high strain hardening. While increased density of double twins, thickening of twins and presence of large grains which favours twinning (Fig. 5-76) are responsible for the lower compression stresses for specimens tested in normal and transverse directions at 3500s^{-1} . However, double twinning has little effect on the strain to failure in these specimens contrary to what was concluded by Barnett [41] that “Double twinning can decrease the elongation in compression”. At elevated temperatures and even at higher strain rates ($\geq 10^3\text{s}^{-1}$), twins are scarcely found in magnesium alloys.

Figure 5-76: SEM picture of specimen tested in compression at $3500s^{-1}$ (under normal loading direction and at room temperature)

At room temperature, plastic flow by basal slip causes anisotropic deformation behaviour of magnesium. The result is strongly directional mechanical properties under tension and compression. The stress-strain curves in rolling, transverse and 45° , show strong hardening behaviour after an initial period of low strain hardening under compression as compared to normal orientation. More twinning in rolling direction as compared to other orientations of loading defines the twinning as influential participant in the deformation of the alloy in rolling direction. In 45° and normal orientations, less number of twins indicates the dislocation slip as the main deformation mechanism. Under compression at room temperature, the alloy is more ductile in 45° , transverse and rolling directions as compared to the normal direction and this is possibly due to the orientation of basal planes parallel to the rolling direction and increased number of twins. Under tensile loading, this orientation of basal planes promotes relatively easier activation of slip in rolling direction as compared to transverse direction resulting in slightly higher dynamic stresses in transverse direction.

Micro-voids and cracks observed almost at all strain rates play an important role in the deformation of specimens. Twin boundaries and grain boundaries are

potential sites for voids formation as seen in Fig. 5-76, twin boundaries are impassable barriers for dislocation movement and their mutual interactions yield local stress areas which support the early development of micro-voids. Nucleation of voids found to occur predominantly at α/β interface than within the α -grains as seen in the micrographs. This is due to the difference in the deformation rates as a consequence of the difference in the strain accommodating ability of the two phases. The α -matrix is expected to accommodate larger amount of strain as compared to β -phase. Formation of twins within existing grains introduces new boundaries which subdivide the grains and act as barrier to dislocations motion contributing to the hardening of the alloy. Under tensile loading, the failure is mainly due to voids growth and coalescence leading to the final fracture of the specimen and the failure is ductile in nature. Isothermal shear bands are formed at quasi-static strain rate are responsible for the slight reduction in the plastic deformation of the alloy with increasing strain rate. Under dynamic loading adiabatic shear bands are found. These shear bands usually formed by the grains which have preferential twin orientation, so not necessarily to be formed uniformly throughout the tested specimen. Severe deformation, large amount of twins and micro-cracks formed and grow within the shear bands cause the alloy to fracture. At elevated temperatures and dynamic strain rates, more shear bands are formed as the material is in softer state as compared to the room temperature contributing to the smaller increase in the ductility of the alloy at elevated temperatures.

Difference in deformation mechanism in compressive and tensile loadings is responsible for tensile-compressive asymmetry in both alloys. AZ91D shows slight difference in properties while a significant difference is seen in AZ31B

properties in compressive and tensile loadings. Slip in tension and twinning and work hardening in compression are responsible for tensile-compressive yield asymmetry. Higher rate sensitivity of yield strength in tensile loading also contributes to this yield asymmetry. With increasing temperature at low strain rates, the tensile-compressive asymmetry decreases due to the fact that non-basal slip becomes active. However, under dynamic loadings even at elevated temperatures the asymmetry is retained as twinning remains active. Such as seen in the deformation at 2500s^{-1} strain rate and elevated 200°C temperature. Therefore it is concluded that even at 2500s^{-1} and 200°C , the twinning played an important role in tensile-compressive asymmetry.

Increase in the hardening exponent n with increasing strain rate and higher strain rate sensitivity exponent m are due to enhanced dislocation generation and dislocation pile up followed by twinning. At elevated temperatures, non-basal slip system becomes active thus causing the alloy to soften. As a result, the work hardening rate should decrease and this is observed experimentally in the current study. In the present study, increase in the specimen temperature is 29°C (normal) to 34.2°C (rolling) when deformed at 200°C . It means the specimen temperature after deformation goes beyond 225°C at which non-basal slip becomes active in magnesium alloys contributing to the ductility of the alloy. The alloy's hardening and softening behaviour can be explained using dislocation mechanics. Dislocations during their movement continuously come across obstacles such as solute atoms, particles, grain boundaries as well as other dislocations that make their movement difficult [32]. These dislocations require energy to surpass these obstacles to carry on their movement. As explained in section 2.3 that thermal energy only is sufficient to overcome short-range

obstacles but long-range obstacles cannot be overcome by thermal energy and need external energy to overcome. The thermal energy ΔG given by Equation 2.1 indicates that as the temperature increases the energy needed to overcome obstacles is reduced and material softens. While as the strain rate increases, the energy needed to overcome these obstacles is higher and the material hardens.

Chapter 6

Constitutive Evaluation

Numerical modelling is an important tool in the analysis of many industrial processes and impact events, such as vehicle crash. To improve the accuracy of simulations, good models are essential to represent the material's constitutive response. In the present study, Johnson-Cook model [100] as given by Eq. (3.4) is used to evaluate the experimental results. The material parameters A , B , n , C and m required by the model were estimated and the predicted stress-strain response of the two alloys at different strain rates and temperatures was plotted in the next sections. For each testing condition considered in the constitutive fit, a median flow curve was calculated from at least three experimental curves and was used to fit the constitutive model. The flow curve considered was reduced to the plastic range for which, (i) the strain rate was approximately uniform; and, (ii) the stress was considered prior to the peak compressive stress (PCS) in compression and UTS in tension. Some results of the model curve fitting were captured and shown in the appendix F. MATLAB program used to read the experimental data from excel file is given in the appendix. In most of the fits, R-squared value is about 0.9 ± 0.05 .

6.1 AZ91D

6.1.1 Compression

The experimental data was fit to the Johnson-Cook constitutive model at different strain rates at room temperature and results are shown in Fig. 6-1 followed by

the Fig. 6-2 which presents the results at high strain rate and elevated temperatures. The Johnson-Cook fitted curves are identified by dashed lines and experimental curves are represented by the solid lines. As observed in Fig. 6-1 and 6-2, the estimated compressive yield stress increases with increasing strain rate. However, the it is little sensitive to the temperature at a strain rate of 2500s^{-1} as observed in Fig. 6-2. The hardening behaviour also showed little dependence on the temperature at higher strain rate.

Figure 6-1: Experimental and Johnson-Cook model curves at room temperature and different strain rates

Figure 6-2: Experimental and Johnson-Cook model curves at elevated temperature and high strain rate

6.1.2 Tension

The Johnson-Cook model fitted curves under tensile loading are shown in Fig. 6-3 and Fig. 6-4 at different strain rates at room temperature and at about 1500s^{-1} strain rate and 250°C temperature respectively. Increasing yield stress with increasing strain rate is obvious as can be seen in Fig. 6-3. The alloy showed increased hardening behaviour (larger n) at higher strain rates. The estimated yield stress decreases at 250°C even at higher strain rate of 1500s^{-1} .

Figure 6-3: Experimental and Johnson-Cook model curves at room temperature and different strain rates

Figure 6-4: Experimental and Johnson-Cook model curves at elevated temperature and high strain rate

The corresponding material parameters calculated for AZ91D are given in the Table 6-1 at 95% confidence level. For each fit, the lower and upper bound of the 95% confidence intervals are also given. These bounds can help to assess the "uncertainty" ($uncertainty = (estimate - lower\ or\ upper\ bound) / estimate$) that the 95% confidence interval represents on the parameters. This "uncertainty" gives a good assessment of the accuracy of each parameter. It is observed that the behaviour of AZ91D is not significantly different under compressive and tensile loadings. However, the alloy is more temperature sensitive under tensile loading as compared to compressive loading.

Table 6-1: Parameters of the Johnson-Cook model for AZ91D magnesium alloy

Johnson-Cook parameters (95% confidence)						
Parameter	Compression			Tensile		
	Estimate	Lower Bound	Upper Bound	Estimate	Lower Bound	Upper Bound
A [MPa]	108.6	93.45	122.9	104.69	91.15	118.32
B [MPa]	737.8	661.8	813.7	673.03	621.3	724.75
N	0.636	0.5357	0.7362	0.6016	0.60	0.605
$*C$	0.06189	0.04982	0.06731	0.03897	0.02151	0.05452
M	2.551	2.475	2.7676	1.183	1.177	1.188

$*C$ is same as m used in experimental results to represent the strain rate sensitivity exponent.

6.2 AZ31B

6.2.1 Compression

The experimental data for AZ31B in normal, rolling and transverse directions was fit to the Johnson-Cook model at different strain rates and temperatures and results are shown in figures from Fig. 6-5 to Fig. 6-10. The results for the alloy tested in 45° loading direction are not shown as the stress-strain behaviour of the alloy in 45° loading direction is quite similar to its behaviour in rolling and transverse directions. In these figures, it can be seen that the yield stress is

moderately strain rate sensitive in normal loading while little rate sensitive in rolling and transverse loading directions. At elevated temperature, the alloy showed a decrease in the yield stress as well as in the peak compressive stress with increasing strain rate in all loading directions. In rolling and transverse directions, the experimental curves are concave up in the beginning while no such concave upward behaviour is observed in the fitted data. This might be due to the limitation of the model to incorporate the effect of twinning responsible for this concave up shape of stress-strain curves.

Figure 6-5: Experimental and Johnson-Cook model curves at room temperature and different strain rates

Figure 6-6: Experimental and Johnson-Cook model curves at and elevated temperature and high strain rate

Figure 6-7: Experimental and Johnson-Cook model curves at room temperature and different strain rates

Figure 6-8: Experimental and Johnson-Cook model curves at elevated temperature and high strain rate

Figure 6-9: Experimental and Johnson-Cook model curves at room temperature and different strain rates

Figure 6-10: Experimental and Johnson-Cook model curves at curves at elevated temperature and high strain rate

6.2.2 Tension

The Johnson-Cook model fitted curves under tensile loading at different strain rates and room temperature for rolling and transverse loading directions are shown in figures below. A significant increase in the yield stress with increasing strain rate can be observed for both loading directions in Fig. 6-11 and Fig. 6-13. The estimated yield stress in rolling and transverse directions decreases at 250°C even at higher strain rate of 1400 s^{-1} .

Figure 6-11: Experimental and Johnson-Cook model curves at room temperature and different strain rates

Figure 6-12: Experimental and Johnson-Cook model curves at curves at elevated temperature and different strain rate

Figure 6-13: Experimental and Johnson-Cook model curves at room temperature and different strain rates

Figure 6-14: Experimental and Johnson-Cook model curves at curves at elevated temperature and different strain rate

The corresponding material parameters calculated for AZ31B in different loading directions are given in Table 6-2 to Table 6-4 at 95% confidence level. In normal direction, parameters are not shown as the tensile tests can't be performed in normal direction due to the limitation of the plate thickness. The alloy showed positive rate sensitivity in all loading directions and significant hardening under compression as compared to tension. The alloy is more temperature sensitive in tension than compression. Unlike AZ91D, AZ31B showed strong compressive-tensile asymmetry as can be seen from the Tables 6-2 to Table 6-4. The alloy is strongly anisotropic in both under compressive and tensile loadings and the anisotropy exists even at elevated temperature and higher strain rates.

Table 6-2: Parameters of the Johnson-Cook model for AZ31B (Normal direction)

Johnson-Cook parameters (95% confidence)						
Parameter	Compression			Tensile		
	Estimate	Lower Bound	Upper Bound	Estimate	Lower Bound	Upper Bound
A [MPa]	217.343	203.886	230.80	-	-	-
B [MPa]	755.75	709.7	801.9			
<i>n</i>	0.5575	0.550	0.565			
C	0.03435	0.02982	0.03969			
<i>M</i>	1.804	1.795	1.813			

Table 6-3: Parameters of the Johnson-Cook model for AZ31B (Rolling direction)

Johnson-Cook parameters (95% confidence)						
Parameter	Compression			Tensile		
	Estimate	Lower Bound	Upper Bound	Estimate	Lower Bound	Upper Bound
A [MPa]	73.451	71.55	75.353	254.3	209.1	289.9
B [MPa]	2533.3	2000	2800	430.4	430	430.8
<i>n</i>	1.035	1.01	1.063	0.3939	0.3848	0.410
C	0.03597	0.03083	0.03708	0.01611	0.01386	0.0192
<i>M</i>	1.358	1.296	1.42	2.002	1.971	2.033

Table 6-4: Parameters of the Johnson-Cook model for AZ31B (Transverse direction)

Johnson-Cook parameters (95% confidence)						
Parameter	Compression			Tensile		
	Estimate	Lower Bound	Upper Bound	Estimate	Lower Bound	Upper Bound
A [MPa]	83.2335	81.578	84.889	212.086	176.99	241.223
B [MPa]	3047	3000	3161	465.3	421.6	520
n	1.045	1.02	1.07	0.3617	0.351	0.385
C	0.03609	0.03152	0.03726	0.02583	0.0129	0.03499
M	1.035	1.012	1.058	1.585	1.547	1.622

A comparison of estimated Johnson-Cook parameters of Az91D and Az31B alloy tested in present work with that of done by other researchers available in the literature is presented in Table 6-5. It is observed that the parameters obtained in the current work are within the range mentioned by others researchers. Liao et al [59] found quite different parameters from that obtained in the current study as well as from Aune et al. [57]. The reason is not clear for this significant difference in hardening behaviour and strain rate sensitivity of the alloy.

Table 6-5: Comparison of Johnson-Cook model parameters for AZ91D and AZ31B alloys from the current work with that from literature

Johnson-Cook parameters for AZ91D						
		A [MPa]	B [MPa]	n	C	M
Current work		105 (91-118)	673 (621-724)	0.6016 (0.060-0.0603)	0.0390 (0.0215-0.0545)	1.18
Liao et al [59]		121	318	0.2145	0.0221	
Aune et al [57]		129	616	0.5975	0.019	
Johnson-Cook parameters for AZ31B						
Current work	Rolling	254 (209-290)	430 (430-431)	0.394 (0.385-0.41)	0.0161 (0.0138-0.0192)	1.93
	Transverse	212 (177-241)	422 (421-520)	0.3617 (0.351-0.385)	0.025 (0.0129-0.034)	1.58
Ulicia et al [72]	Rolling	224 (215-233)	380 (316-444)	0.761 (0.628-0.894)	0.012 (0.011-0.014)	1.55
	Transverse	232 (214-249)	232 (207-258)	0.501 (0.359-0.643)	0.013 (0.011-0.014)	1.44
Ulicia et al [74]	Rolling	180 (152-208)	344 (310-380)	0.554 (0.329-0.613)	0.012 (0.011-0.013)	-
	Transverse	221 (176-265)	206 (183-228)	0.370 (0.144-0.596)	0.011 (0.010-0.013)	-

Furthermore, the Table 6-6 showed the properties of aluminium alloys and commonly used structural steel. It is interesting to observe that the alloys show better resistance to deformation (the high hardening) than some aluminium alloys and even some steel. This indicates the potential of AZ91D and AZ31B alloys to be used in structural applications in both civil and defence sector and replacing if not fully by the time partially the aluminium and steel.

Table 6-6: Comparison of Johnson-Cook model parameters for AZ91D and AZ31B alloys from the current work with that of aluminium alloys and steel from literature

Johnson-Cook parameters for different alloys					
	A [MPa]	B [MPa]	n	C	M
AZ91D	105	673	0.6016	0.03897	1.18
AZ31B	254	430.4	0.3939	0.0161	1.93
Al6061-T6 Zhu et al [114]	236	430	0.376	0.024	
Al6061 Lesues et al [115]	324	114	0.42	0.002	1.34
Mild steel Vedantam et al [116]	217	233.7	0.6248	0.0756	
Steel 1006 Johnson-Cook [100]	350	275	0.36	0.022	1
Al2024 T351 Johnson-Cook [100]	265	426	0.34	0.015	1.0

6.3 Discussion

Johnson-Cook model was fitted to the experimental data and the results are in reasonable agreement with the experimental data at high strain rates and at room temperature except in the beginning portion of the flow curves, where the experimental data is different from the fitted curves. This might be due to the fact that the strain rate is not stringently constant in the early stage of deformation thus introducing this lack of fit. A lack of fit is also observed at elevated temperature and high strain rates. In general, the Johnson-Cook model does not capture the change in rate sensitivity as the strain rate changes. The model assumes constant log-linear strain rate sensitivity while in fact it changes with

strain rate. It is also observed in the constitutive analysis of the two alloys that for the data fit considered here, the non-linear curve fitting gives a wide range of strain rate sensitivity parameter C for 95% confidence interval as can be observed in Tables 6-1 to Tables 6-4. This uncertainty in the rate sensitivity has also an effect on the accuracy of the other parameters in the constitutive equation. The thermal softening is well captured by the model for both alloys except for AZ31B in rolling and transverse loading directions under compression. Also the model is unable to fit the concave upward behaviour of the stress-strain curves in the early stage of deformation. It might be due to the reason that the Johnson-Cook model is not able to model the effect of twinning responsible for this concave-up stress-strain behaviour. A modified model is needed that can take into account the deformation mechanism effects to predict the accurate response of the alloys under particular strain rate and temperature conditions.

Chapter 7

Conclusions & Recommendations

The conclusions drawn from the work presented in the previous chapters can be divided into two groups. The grouping is based on whether the conclusions are related to as cast AZ91D or wrought AZ31B alloy.

7.1 AZ91D

1. Mechanical properties of AZ91D are strongly strain rate and temperature dependent. However, at dynamic strain rates, the effect of temperature is less significant. At room temperature, as the strain rate increases from 10^{-4}s^{-1} to $3 \times 10^3\text{s}^{-1}$, the compressive yield strength and the failure strain increases by 40% and 150%, respectively. Under compression at a strain rate of $2500 \pm 100\text{s}^{-1}$, the yield stress decreases by 30% as the temperature increases from -30°C to 200°C . The energy absorption at 3000s^{-1} is about four times higher than the energy absorption at 10^{-4}s^{-1} strain rate, while it is little decreased as the temperature increases from -30°C to 200°C . Under tensile loading, the yield stress, UTS, energy absorption and strain to failure at 1500s^{-1} are 75%, 78%, 112% and 12% higher than their respective values at 10^{-4}s^{-1} . For temperature increase from 25°C to 250°C at 1500s^{-1} , yield strength, UTS, energy absorption decrease by 44%, 35% and 29% respectively, while strain to failure increases by 25%. The hardening exponent increases with as the strain rate increases and decreases with increasing temperature in both compression and tension. However, in the temperature range considered in this study, the alloy shows strong hardening behaviour under compression as compared to under

tension. The strain rate sensitive of the alloy is strain dependent; it increases with increasing strain in both compressive and tensile loading.

2. In AZ91D alloy, the deformation is accommodated by both α -matrix and β -phase. The inter-dendritic β -phase present along the grain boundaries is responsible for lower strength and limited ductility. The grain boundary strengthening, resistance to deformation of hard and brittle β -phase and shear bands formation contribute to the strength and improved ductility of the alloy at higher strain rates. Fewer cracks are observed in the bulk β -phase at quasi-static strain rates while lots of cracks are seen at dynamic strain rates. Moreover, the thermal effects caused by the high strain rate deformation could make the specimen temperature rise and the resistance to deformation is decreased. Additionally, the α -Mg is prone to twinning at high strain compression that changes the orientation of crystal to make slide system move easily. At dynamic strain rates, the large amount of rupture of β -phase is responsible for promoting the grain boundary deformation which causes the strain to increase with increasing strain rate. While, at 200°C, decrease in the work hardening rate and the maximum flow stress indicate that some non-basal slip activity responsible for softening of the alloy becomes significant. At elevated temperatures, voids generation and coalescence leads to the ductile fracture while at -30°C and 2500s⁻¹, large amount of cracks in β -phase (Mg₁₇ Al₁₂) are seen and fracture seems to be brittle in nature contributing to lower ductility. Under tensile loading, cracks are seen only in the vicinity of the fracture surface. In tension, the brittle eutectic β -phase segregation along the grain boundaries seems to cause inter-granular fracture. Both surfaces of the fracture are covered with aluminium

rich β -phase which shows that the cracks occur basically by inter-crystalline splitting along the boundaries of the original dendrite. Most of the strain accommodated by β -phase and the twinning in α -matrix are possible reasons for larger strain to failure under compression as compared to tension.

7.2 AZ31B

1. Like AZ91D, the mechanical properties of AZ31B are also sensitive to strain rate and temperature. Under compression, the yield stress increases about 35% in normal loading and about 15% in rolling, transverse and 45° loading directions as the strain rate increases from quasi-static to 2500s⁻¹. While it decreases about 55% in normal loading and about 15-20% in other three loading directions for an increase in temperature from -30°C to 200°C at a strain rate of 2250s⁻¹. Under tensile loading, as the strain rate is increased from 10⁻⁴s⁻¹ to 1400s⁻¹, an increase of 50% is observed in the yield strength while UTS is increased by 38% and 55% for rolling and transverse directions, respectively. The alloy is strongly sensitive to the temperature. As, the temperature is increased from room temperature to 250°C, the yield strength and the UTS is reduced by 37% and 30% for rolling direction and 45% and 36% for transverse direction, respectively. In the temperature range considered here, the alloy is more ductile (30% in transverse direction and 55% in rolling direction) at higher strain rates as compared to the quasi-static strain rate. The energy absorption is 2 to 3 times (it varies in different loading directions) higher at 3500s⁻¹ strain rate as was observed at quasi-static strain rate. However, it is slightly decreased as the temperature increases from -30°C to 200°C at a strain rate of 2250s⁻¹.

2. In each loading orientation, the hardening exponent n in general increases with increasing strain rate and decreases as the temperature increases under both compressive and tensile loading. The alloy is anisotropic and the anisotropy is larger at higher strain rates and lower temperature. However, the anisotropy in tension is relatively lesser as compared to compression. Under compressive loading, the alloy showed strong ability of strengthening (high hardening exponent n) during deformation as compared to tension. As well as the alloy is more strain rate sensitive in compression than in tension. AZ31B alloy shows strong tensile-compressive yield asymmetry. Around 70% (10^{-4}s^{-1}) to 130% (1400s^{-1}) higher yield stresses are observed in tensile loading in both rolling and transverse directions. The energy absorption is considerably dependent on the loading orientation.
3. At room temperature, AZ31B deforms by dislocation glide in the basal plane (0001) $\langle 11\bar{2}0 \rangle$ and twining on the pyramidal planes $\{10\bar{1}0\}$. Slip is the dominant deformation mechanism in normal loading direction while large number of twins defines the twining as influential participant in the deformation of the alloy in rolling, transverse and 45° angle loading directions. Lots of twins are formed at quasi-static strain rate while fewer twins are observed at higher strain rates. In rolling, transverse and 45° loading directions, concave upward stress-strain curves at room temperature are due to the increased number of twining activity and due to the grain boundary sliding. Increased numbers of intersecting twins are responsible for strong work hardening as seen in 45° angle, rolling and transverse loading directions and decrease in the stress and strain to failure in normal and transverse loading at 3500s^{-1} seems to be an effect of increased density of double twins

and thickening of twins. Under tensile loading, lots of single and double twins are seen at quasi-static strain rate for both rolling and transverse loading as compared to dynamic strain rate. In tension, deformation is controlled by twining and plastic deformation of grains even at 250°C and 1400s⁻¹. Grains are elongated along the stress direction in both rolling and transverse directions. The alloy deformed in rolling direction has high density of twining as compared to transverse loading even at elevated temperature and this is possibly due to the orientation of grains favouring the twining. Twining has played an important role in tensile-compressive asymmetry even at elevated temperature and higher strain rates.

4. Johnson-Cook model was employed to the experimental data and the results predicted by the model are in reasonable agreement with the experimental data except in the beginning portion of the flow curves, where the Johnson-Cook fitted curves differ from the experimental data. This lack of fit between experimental and model data is because of a constant strain rate is not achieved in the early stage of deformation in dynamic tests.
5. Both alloys showed strong potential to be used in structural applications and have capability in particular AZ31B to replace aluminium alloys and commonly used steel. The significant high dynamic increase factor (DIF) even better than 6061-t6 aluminium and mild steel as shown in Fig. 7-1 and Fig. 7-2 indicate the positive prospects of these two alloys to be used as structural materials.

Figure 7-1: Comparison of DIF of various alloys in compression

Figure 7-2: Comparison of DIF of various alloys in tension

7.3 Recommendations

1. It is planned to do microstructural analysis on nano-scale using TEM (Transmission electron microscopy) for enhanced understanding of the role of twinning and the mechanics behind twinning. Within the twin domains the lattice is sheared and reoriented during twinning. As a result, the macro and micro scale observations provide a little insight into the atomic scale processes responsible for twin nucleation and growth. Knowledge of slip activity, slip bands, dislocations and dislocation pileup, twinning and lattice orientation after deformation needs to study the microstructure on submicron level. Second phase particles, not observable using optical metallography, will be analysed (their size, shape and distribution) using TEM to understand their association with failure initiation sites.
2. A modified model is needed that can take into account the deformation mechanism effects (twinning, slip etc.) to predict the accurate response of the alloys under particular strain rate and temperature conditions. Some explicit dynamic software such as SDYNA should be used to simulate the high strain rate tests

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APPENDICES

APPENDIX A: ENGINEERING STRESS-STRAIN CURVES (AZ91D)

A1: Compressive

Fig A1-1: Stress strain curves at 500 s^{-1}

Fig A1-2: Stress strain curves at 1000 s^{-1}

Fig A1-3: Stress strain curves at 1500 s^{-1}

Fig A1-4: Stress strain curves at 2000 s^{-1}

Fig A1-5: Stress strain curves at 2500 s^{-1}

Fig A1-6: Stress strain curves at 3000 s^{-1}

Fig A1-7: Stress strain curves at 1000 s^{-1} and -30°C

Fig A1-8: Stress strain curves at 2500 s^{-1} and -30°C

Fig A1-9: Stress strain curves at 1000 s^{-1} and 200°C

Fig A1-10: Stress strain curves at 2500 s^{-1} and 200°C

Fig A1-11: Stress as a function of strain rate

Fig A1-12: Stress as a function of strain rate

A2: Tensile

Fig A2-1: Stress-strain curves at 350 s^{-1}

Fig A2-2: Stress-strain curves at 750 s^{-1}

Fig A2-3: Stress-strain curves at 1500 s^{-1}

Fig A2-4: Stress as a function of strain

Tensile, 250°C

Fig A2-5: Stress-strain curves at 250°C

APPENDIX B: ENGINEERING STRESS-STRAIN CURVES (AZ31B)

B1: Compressive, (45° Angle Loading)

Fig B1-1: Stress-strain curves at 500s⁻¹

Fig B1-2: Stress-strain curves at 1000s⁻¹

Fig B1-3: Stress-strain curves at 1500 s^{-1}

Fig B1-4: Stress-strain curves at 2500 s^{-1}

Fig B1-5: Stress-strain curves at 3500 s^{-1}

Fig B1-6: Strain rate versus time

B2: AZ31B (Normal Loading)

Fig B2-1: Stress-strain curves at 500 s^{-1}

Fig B2-2: Stress-strain curves at 1000 s^{-1}

Fig B2-3: Stress-strain curves at 1500 s^{-1}

Fig B2-4: Stress-strain curves at 2500 s^{-1}

Fig B2-5: Stress-strain curves at 3500 s^{-1}

Fig B2-6: Strain rate versus time

B3: AZ31B (Rolling Loading)

Fig B3-1: Stress-strain curves at 500 s^{-1}

Fig B3-2: Stress-strain curves at 1000 s^{-1}

Fig B3-3: Stress-strain curves at 1500 s^{-1}

Fig B3-4: Stress-strain curves at 2500 s^{-1}

Fig B3-5: Stress-strain curves at 3500 s^{-1}

Fig B3-6: Strain rate curves

B4: AZ31B (Transverse Loading)

Fig B4-1: Stress-strain curves at 1000 s^{-1}

Fig B4-2: Stress-strain curves at 1000 s^{-1}

Fig B4-3: Stress-strain curves at 1500 s^{-1}

Fig B4-4: Stress-strain curves at 1000 s^{-1}

Fig B4-5: Stress-strain curves at 3500 s^{-1}

Fig B4-6: Strain rate curves

Appendix C: Compressive, -30°C and 200°C

C1: AZ31B (45° Angle Loading)

Fig C1-1: Stress-strain curves at (-30°C)

Fig C1-2: Stress-strain curves at (-30°C)

Fig C1-3: Stress-strain curves at (200°C)

Fig C1-4: Stress as a function of strain rate

C2: AZ31B (Normal Loading)

Fig C2-1: Stress-strain curves at (-30°C)

Fig C2-2: Stress-strain curves at (-30°C)

Fig C2-3: Stress-strain curves at (200°C)

Fig C2-4: Stress as a function of strain rate

C3: AZ31B (Rolling Direction)

Fig C3-1: Stress-strain curves at (-30°C)

Fig C3-2: Stress-strain curves at (-30°C)

Fig C3-3: Stress-strain curves at (200°C)

Fig C3-4: Stress as a function of strain rate curves

C4: AZ31B (Transverse Direction)

Fig C4-1: Stress-strain curves at (-30°C)

Fig C4-2: Stress-strain curves at (-30°C)

Fig C4-3: Stress-strain curves at (200°C)

Fig C4-4: Stress as a function of strain rate

Appendix D: Tensile, Room Temperature

D1: AZ31B (Rolling Loading)

Fig D1-1: Stress-strain curves at 500 s^{-1}

Fig D1-2: Stress-strain curves at 1000 s^{-1}

Fig D1-3: Stress-strain curves at 1400 s^{-1}

Fig D1-4: Strain rate versus time curves

Elevated Temperature Curves (250°C)

Fig D1-5: Strain rate versus time curves at 250°C

D2: AZ31B (Transverse Loading)

Fig D2-1: Stress-strain curves at 500s^{-1}

Fig D2-2: Stress-strain curves at 1000s^{-1}

Fig D2-3: Stress-strain curves at 1400 s^{-1}

Fig D2-4: Strain rate versus time curves

Elevated Temperature Curves (250°C)

Fig D2-5: Stress-strain curves at 250°C

Appendix E: Data Acquisition

E1: Extracting the results

The output data from oscilloscope that contains the information of the specimen deformation behaviour during testing is in ASCII format and needs some processing to extract incident wave, reflected wave and transmitted wave from the data. First it is important to choose the starting and ending point of the incident, reflected and transmitted pulses obtained from oscilloscope (Fig. 3-4). These points are shown in Tabl3 E1. The table shows the separated incident pulse, reflected pulse and transmitted pulse file from oscilloscope data. All three pulses start from zero and end at zero. However, the transmitted pulse is usually not taken up to zero as after unloading or fracture of specimen data is of no use.

Index	Incident (V)	Index	Reflected (V)	Index	Transmitted (V)
1	0.00E+00	1	0.00E+00	1	0.00E+00
2	-1.67E-01	2	8.33E-02	2	-2.50E-01
3	-3.33E-01	3	8.33E-02	3	-3.33E-01
Start point of IP		Start point of RP		Start point of TP	
.	.	.	.	.	.
138	-1.67E-01	137	3.33E-01	140	-5.00E-01
139	-0.833E-02	138	1.67E-01	141	-4.17E-01
140	0.00E+00	139	0.00E+00	142	-3.33E-01
End point of IP		Endpoint of RP		Endpoint of TP	

In next step, these voltage signals are converted into incident (ϵ_I), reflected (ϵ_R) and transmitted (ϵ_T) strain pulses as shown in Table below. This is done by dividing the incident, reflected and transmitted voltage signals by the Gain factor

(Corresponding the RANGE factor used for amplification of voltage signals) which is in the present study is 500×10^{-6} (this can vary if needed).

	ϵ_I	ϵ_R	ϵ_T	$\int \epsilon_R dt$	Strain	Stress	Strain Rate
1	0.00E+00	0.00E+00	0.00E+00	0.00E+00	0.00E+00	0.00E+00	0.00E+00
2	8.35E-05	4.17E-05	4.15E-05	4.17E-11	1.07E-04	2.12E+01	-1.07E+02
3	1.67E-04	-4.17E-05	8.30E-05	0.00E+00	0.00E+00	4.23E+01	1.07E+02
4	2.50E-04	-4.17E-05	8.30E-05	-4.17E-11	1.07E-04	4.23E+01	1.07E+02
5	3.75E-04	-4.17E-05	1.25E-04	-8.33E-11	2.14E-04	6.38E+01	1.07E+02

To get the Strain, we need to integrate ϵ_R (Column L) to derive at $\int \epsilon_R(t) dt$ (Column N). Hence, Column N, $\int (\epsilon_R) dt$ is achieved by multiplying each row of Column L by 0.0000001 (the time between two data points and is equal to $(1/\text{sample rate}(10^7))$ and adding the previous row value to get accumulative sum a shown in Table below. After which the specimen stress, strain and strain rate are calculated using Eq. 3.26

	ϵ_I	ϵ_R	ϵ_T	$\int \epsilon_R dt$	Example of how to tabulate $\int \epsilon_R dt$ = (0.000001 * K1) = (0.000001 * K2) + (N1) = (0.000001 * K3) + (N2) = (0.000001 * K4) + (N3) = (0.000001 * K5) + (N4)
1	0.00E+00	0.00E+00	0.00E+00	0.00E+00	
2	8.35E-05	4.17E-05	4.15E-05	4.17E-11	
3	1.67E-04	-4.17E-05	8.30E-05	0.00E+00	
4	2.50E-04	-4.17E-05	8.30E-05	-4.17E-11	
5	3.75E-04	-4.17E-05	1.25E-04	-8.33E-11	
6	5.00E-04	-8.35E-05	2.08E-04	-1.67E-10	

E2: Filtering of Raw Data using MATLAB

A low pass filter given below is used to filter the noise in the data. The code read input data from excel file (.xls) in two column format (time and data to be filtered) and after filtering the output is written in desired column.

Low Pass filter

```
[filename,filepath]=native_uigetfile('.xls','open the file',300,400);  
data_source=strcat(filepath,filename);
```

```
%read in data
```

```
F_P=500000;
```

```
F_S=10000000;
```

F_P and F_S are to make sure the filtered graph

```
data_in=xlsread(data_source,'A2:B151');%read in raw data A column is
```

```
time column is data
```

```
sample_rate=10e7;
```

```
stime=1/sample_rate;
```

```
t_in=data_in(:,1);
```

```
inci_reflec_in=data_in(:,2);
```

```
%begin to design filter
```

```
Wp=F_P/(sample_rate/2); %NORMALIZED CUTOFF FREQUENCIES
```

```
Ws=F_S/(sample_rate/2); %NORMALIZED CUTOFF FREQUENCIES
```

```
Rp=3;
```

```
Rs=5;
```

```
[N,Wn]=buttord(Wp,Ws,Rp,Rs);
```

```
[B,A]=butter(N,Wn);
```

A2 represent the first input reading (column A, second row) and B151 is the last input reading (column B, last row) at

G2 represent the first output reading (column G, second row) and G151 is the last output reading (column G, last row). The length of output data (column G) depends on the length of input data (column B).

```
inci_reflec_out=filtfilt(B,A,inci_reflec_in); % inci_reflec_in is
```

```
raw valuef-inci_reflec_out is the value after filter
```

```
write_result1=xlswrite(data_source,inci_reflec_out,'G2:G151');%result
```

```
output to F column
```

Figure (filtered and unfiltered together)

```
aa=1:length(inci_reflec_in);
bb=1:length(inci_reflec_out);
plot(aa,inci_reflec_in,'k:',bb,inci_reflec_out,'k-')
xlabel('Time (uSec)')
ylabel('Stress (MPa)')
```

Figure E2-1: Graph window

E3: Heat Conduction in Input and Output Bars

As the working temperature range is not very high (maximum is 250 C°) and specimen size (thickness less than 4 mm) is very small therefore some assumption have been made for the simplicity of the problem:

1. No convection from the lateral surface of the specimen
2. Specimen is considered to be at constant working temperature for the time it is in contact with the bars
3. The total test time for the specimen for which it is exposed to room temperature including its contact with the bars is very short (10 to 15 seconds)
4. Symmetrical heat flow is considered on both sides from specimen to bars
5. A very thin and uniform layer of air between specimen and bars exists

Properties used in the calculation are

S304 Steel (Bar Material)

Density $\rho = 7900 \text{ Kg/m}^3$, Thermal conductivity $k (300\text{K}) = 14.9 \text{ W/K-m}$, $C_p(300\text{K}) = 477 \text{ J/kg-K}$,

Thermal diffusivity $\alpha = 0.395 \times 10^{-5} \text{ m}^2/\text{sec}$,

Magnesium AZ91D (Specimen Material)

Density $\rho = 1800 \text{ kg/m}^3$, Thermal conductivity $k (300\text{K}) = 72 \text{ W/K-m}$, $C_p(300\text{K}) = 1024 \text{ J/kg-K}$,

Thermal diffusivity $\alpha = 3.96 \times 10^{-5} \text{ m}^2/\text{sec}$,

Magnesium AZ31B (Specimen Material)

Density $\rho = 1774 \text{ kg/m}^3$, Thermal conductivity $k (300\text{K}) = 76 \text{ W/K-m}$, $C_p(300\text{K}) = 1026 \text{ J/kg-K}$,

Thermal diffusivity $\alpha = 4.17 \times 10^{-5} \text{ m}^2/\text{sec}$,

Contact coefficient $h_c (\text{Steel/Al}) = 3200 \text{ N/m}^2\text{-K}$, $\text{erf}(z) = 1 - 1.5577e\{-0.7182(z+0.7856)^2\}$, $z = \frac{x}{2\sqrt{\alpha t}}$

$$q = Q/A = -k \frac{\Delta T}{\Delta x}, \quad \frac{T(t) - T_\infty}{T_0 - T_\infty} = e^{-mt}, \quad m = \left(\frac{h}{\rho c_p} \right) \left(\frac{A}{V} \right), \quad \frac{A}{V} = \frac{2(L+R)}{RL}, \quad \alpha = k/(\rho c_p)$$

A is surface area and V is the volume

E4: Temperature rise in specimen due to deformation

The increase in specimen temperature due to deformation is calculated assuming that

- Deformation process totally adiabatic, no heat loss from the specimen into the environment and bars.
- The 90% of the total energy from plastic deformation is converted into heat that the fraction of plastic work converted into heat β is 0.9

Heat in the specimen is given by the following equation,

$$Q \text{ (joules)} = (\rho V)C\Delta T$$

$$E = \frac{Q}{V} \text{ (J/m}^3\text{)} = \rho C\Delta T$$

$$\Delta T = \beta \left(\frac{1}{\rho C} \right) E$$

$$E = \int \sigma_T d\varepsilon_p$$

In the above equations,

In the above equations, Q is the heat due to temperature rise ΔT , E is the energy absorbed by the specimen during plastic deformation in J/m^3 . ρ (Kg/m^3) is the density of the material, V (m^3) is the volume and C (J/kg-K) is the specific heat of the material.

Constants for AZ91D

Density= $\rho=1800 \text{ kg/m}^3$, specific heat = $C(300\text{K})=1024 \text{ J/kg-K}$,

Constants for AZ31B

Density= $\rho=1774 \text{ kg/m}^3$, specific heat = $C(300\text{K})=1026 \text{ J/kg-K}$,

Example Calculation (AZ91D)

At 25°C and 2500s⁻¹

E = Energy absorbed during deformation at 2500s⁻¹ = 98.2 MJ/m³

$\Delta T = (T_2 - T_w) = 0.9 (98.2/1800 \times 1024) \times 10^6 = 48 \text{ K}$, $T_2 = 340 \text{ K}$ (67°C)

At 25°C and 3000s⁻¹

E = Energy absorbed during deformation at 3000s⁻¹ = 110.5 MJ/m³

$\Delta T = (T_2 - T_w) = 0.9 (110.5/1800 \times 1024) \times 10^6 = 54 \text{ K}$, $T_2 = 346 \text{ K}$ (73°C)

At 200°C and 2500s⁻¹

E = at 200°C and 2500s⁻¹ = 80.5 MJ/m³

$\Delta T = (T_2 - T_w) = 0.9 (80.5/1800 \times 1024) \times 10^6 = 39 \text{ K}$, $T_2 = 508 \text{ K}$ (235°C)

T_w is working temperature= 25°C, T_2 is temperature of the specimen after deformation.

Example Calculation (AZ31B)

Normal direction, at 2500s⁻¹ strain rate and room temperature

E= Energy absorbed during deformation at 2500s⁻¹ = 75.2M J/m³

$\Delta T = (T_2 - T_w) = 0.9 (75.2 / 1774 \times 1026) \times 10^6 = 37.2 \text{ K}$, $T_2 = 335.2 \text{ K}$ (62.2°C)

At 200°C and 2250s⁻¹

E= at 200°C and 2250s⁻¹ = 58.65 MJ/m³

$\Delta T = (T_2 - T_w) = 0.9 (58.65 / 1774 \times 1026) \times 10^6 = 29.0 \text{ K}$, $T_2 = 502 \text{ K}$ (229°C)

T_w is working temperature= 25°C, T_2 is temperature of the specimen after deformation.

Working Temperature	Rise in specimen temperature in different loading directions (ΔT°)			
	Normal	Rolling	Transvers	45°
25°C	37.2	41	54.8	42.1
200°C	29	34.3	32.7	33.2

References

K. D. Hagen. 1999. Heat Transfer with Applications. Prentice Hal, New Jersey, USA.

The Error function $\text{erf}(z) = 1 - 1.5577e \{-0.7182(z+0.7856)^2\}$, Appendix A

For values of k , C_p , α , Appendix C

Contact coefficient, h_c Table 2-2

Appendix F: Constitutive Analysis

Program to read experimental data from Excel files

```
% Calculation for "A, B and n"

[filename,filepath]=uigetfile('.xls','Data File For Calculation Of B,
n and C for J/C material model');
data_source=strcat (filepath,filename);

%read in data
data_in=xlsread(data_source,'A1:B90'); %A1 represents the first data
entry in Column A and B90 is the last data entry. These vary with the
starting of the data and the length of the data

strain=data_in(:,1);           %experimental strain
stress=data_in(:,2);          %experimental stress


% Calculation of "C"

[filename,filepath]=uigetfile ('.xls','Data File for Calculation of
C');
data_source=strcat(filepath,filename);

%read in data
data_in=xlsread(data_source,'A1:B810');%A1 represents the first data
entry in Column A and B810 is the last data entry. These vary with
the starting of the data and the length of the data

Strain=data_in(:,1);
Stress=data_in(:,2);
```

F1: AZ91D

Fig. F1-1: Johnson-cook parameters estimation using non-linear curve fitting in MATLAB (AZ91D, Compression)

Fig. F1-2: Johnson-cook parameters estimation using non-linear curve fitting in MATLAB (AZ91D, Tensile)

Fig. F1-3: Johnson-cook parameters estimation using non-linear curve fitting in MATLAB (Elevated Temperature, AZ91D Compression)

F2: AZ31B

Fig. F2-1: Johnson-cook parameters estimation using non-linear curve fitting in MATLAB (AZ31B Normal, Compression)

Fig. F2-2: Johnson-cook parameters estimation using non-linear curve fitting in MATLAB (AZ31B Rolling, Compression)

Fig. F2-3: Johnson-cook parameters estimation using non-linear curve fitting in MATLAB (AZ31B Rolling, Tensile)

Fig. F2-4: Johnson-cook parameters estimation using non-linear curve fitting in MATLAB (AZ31B Transverse, Compression)

Fig. F2-5: Johnson-cook parameters estimation using non-linear curve fitting in MATLAB (AZ31B Transverse, Tensile)

Fig. F2-6: Johnson-cook parameters estimation using non-linear curve fitting in MATLAB (Elevated Temperature, AZ31B Compression); a: Normal direction, b: Rolling direction, c: Transverse direction

Publications by year

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