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Modelling plastic behaviour and fracture occurrence in HCP metal sheets deformed in a wide range of temperatures and stress states

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ABSTRACT

Titanium and magnesium alloys both have hexagonal closest packed (HCP) structure with limited slip systems. This structure lead to two features: low formability at room temperature and the anisotropy influence on the mechanical characteristics and fracture behaviour. Therefor, the plastic behaviour and fracture occurrence need to be investigated at elevated temperature and different specimen orientations.

The approach comprises a joint use of experimental and numerical techniques. In terms of experiments, on the one hand, the plastic behaviour was studied and analyzed on the basis of a large number of tensile tests carried out on smooth specimens in a wide range of temperature and strain rate; the fracture behaviour was studied and analyzed by carrying out tensile tests on different specimen geometries as well as Nakajima tests, in order to reproduce various stress states, identified by the related stress triaxiality and deviatoric stress parameters; on the other hand, the alloy microstructure was analyzed on the tested specimens to explain the mechanical characteristic and anisotropy behaviour, including Optical Microscopy (OM), Scanning Electron Microscopy (SEM), Electron Back Scatter Diffraction (EBSD) techniques and micro-hardness testing. In terms of numerical modelling, on the basis of the large number of experimental results, numerical simulations were implemented in the Ls-DynaTM code environment using a dynamic-implicit analysis with a coupled thermo-mechanical solution procedure to reproduce the sheet anisotropic plastic behaviour and fracture characteristics. The yield criterion and damage model were studied and modified to predict the experimental results.

Specifically, for Ti6Al4V titanium alloy, the heating of the alloy below the β -transus temperature is recognized to enhance its formability, reducing the flow stress and increasing the ductility. However, to investigate the effect of the sheet anisotropy on the material flow behaviour and plastic instability at varying temperature and strain rate, uniaxial tensile tests were carried out in a wide range of testing temperatures (from room temperature to 800°C) and strain rates (0.01, 0.1, 1 s⁻¹) to assess the anisotropy effects. Strain hardening, strain rate sensitivity, and Lankford coefficients were evaluated as a function of the testing parameters and specimen orientation. Furthermore, a numerical model of the uniaxial tensile tests was developed and calibrated making use of the Barlat-Lian-1989 yield criterion and a hardening rule, which was adapted to take into account the anisotropic behaviour at different temperatures. It was proved that the developed model was capable of predicting the strain localization in the specimen gauge length due to plastic instability as well as its thickness distribution at varying temperature and strain rate. To investigate the influence of the sheet anisotropy on the material failure, tensile tests on smooth and notched specimens, shear tests and Nakajima-type tests carried out at varying temperature and rolling direction. The fracture strain is measured and the effect of the specimen orientation and stress state is identified. The coupling of the Barlat-Lian 1989 anisotropic yield criterion and GISSMO damage model was introduced to predict the fracture occurrence and

capture its anisotropic characteristics. The two models were calibrated on the basis of an extensive experimental campaign. In order to validate the proposed approach, tests not used in the calibration phase were used, and the comparison between experimental and numerical results was carried out in terms of fracture characteristics at varying temperature and rolling direction. It was proved that the proposed modelling was able to satisfactorily reproduce the different fracture characteristics arising as a consequence of the sheet anisotropy and testing temperature.

Regarding AZ31B magnesium alloy, the same tests carried out on smooth specimen at the range of temperatures from room temperature to 300°C at 0.1 s⁻¹ to evaluate the anisotropy influence on the mechanical and microstructural characteristics. The Lankford coefficients, ultimate tensile strength, diffuse necking strain and fracture strain values were evaluated as a function of the testing temperature and specimen orientation. Furthermore, microstructural features were analyzed as well as micro-hardness was measured for each testing condition to assess the post-deformation characteristics. To describe the fracture behaviour of AZ31 magnesium alloy sheets as a function of the temperature, stress state, and rolling direction. The same fracture tests were carried out at room temperature, 100°C and 300°C using different specimen geometries cut at different orientations with respect to the sheet rolling direction. The fracture strain values were measured making use of the Digital Image Correlation (DIC) approach and the fracture surfaces qualitatively characterized by means of stereoscopy and SEM. Afterwards, the AZ31 fracture locus as a function of the stress state, temperature and rolling direction was constructed. To analytically predict the fracture locus, a temperature-dependent parameter was added to the Modified Mohr Coulomb (MMC) fracture criterion; the newly proposed criterion was calibrated on the basis of the experimental data obtained from the tests previously mentioned, proving to be suitable to predict the AZ31 fracture locus for tension-dominated regions at varying temperature.

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Chapter 1

1 Introduction

1.1 Background and Research objectives

Hexagonal Close-Packed (HCP) metals are used in a wide range of industrial fields, including aerospace, automotive, nuclear, chemical engineering and bioengineering. These metals and alloys are widely used thanks to their attractive mechanical properties and elevated corrosion resistance. As example, titanium alloys are used for corrosion-resistant pipes, fan blades in jet engines and biomedical implants, whereas magnesium alloys are used in lightweight structure applications, especially in the transport industry to reduce energy consumption and greenhouse gas emissions, zirconium alloys in fuel cladding in nuclear fission reactors, zinc alloys as coatings for steels, and beryllium alloys in scientific instrumentation and nuclear fusion reactors. Some other HCP metals are used less extensively, like cadmium, rhenium, hafnium and cobalt. The HCP structure imparts significant anisotropy on the alloy mechanical performances at the macroscale and microscale, which often leads to more difficult processing routes and requires careful exploitation in high-performance applications.

This research is focused on two widespread HCP metals, namely the titanium alloy Ti6Al4V and the magnesium alloy AZ31B. Their HCP structure may induce significant anisotropy of the mechanical characteristics, which can be of great concern in sheet metal forming, since the as-delivered sheet texture may enhance the material anisotropic characteristics. Ti6Al4V alloy is a biphasic alloy, with the dominant alpha phase at room temperature characterized by an HCP structure. The limited number of slip planes of the HCP structures is known to lead to poor formability, which forces to form Ti6Al4V sheets at elevated temperature. Furthermore, the limited number of slip planes induces a significant anisotropy of the material mechanical characteristics, which may also affect its failure behaviour. AZ31B alloy also shows a poor formability exhibited by its HCP crystalline structure and strong basal texture induced by the traditional forming process chain, consisting in either hot rolling or hot extrusion processes. At room temperature, magnesium alloys own two independent basal slip systems and twinning as deformation modes, namely Compression Twins (CTs) and Extension Twins (ETs), which owe the names to their contraction and extension along the c-axis, respectively. By virtue of the limited number of active slip systems, the ductility of magnesium alloys is therefore restrained.

As previously described, to overcome this limitation, it is necessary to understand the suitable deformation conditions to improve the formability of Ti6Al4V and AZ31B alloy sheets, assuring a correct design of the sheet forming operations: to this aim, a proper knowledge of the effect of different stress states, temperature and

rolling direction on the formability and fracture behaviour is mandatory. This research is an attempt to address these concerns and its major objectives are:

- (i) To investigate the anisotropy effects on the Ti6Al4V and AZ31B sheets flow behaviour and plastic instability in a wide range of thermo-mechanical conditions, typical of industrial sheet forming operations.
- (ii) To analyze and numerically model the Ti6Al4V sheets failure behaviour at varying temperature and specimen rolling direction.
- (iii) To determine experimentally the fracture locus of the AZ31B magnesium alloy at varying temperature, stress state, and rolling direction; whereupon to propose a newly MMC fracture criterion to predict the fracture onset both at room and elevated temperature.

1.2 Outline of the thesis

This thesis is divided into six chapters. The first presents the introduction of the work with a brief description of the industrial request on HCP metals formability and research objectives. The second chapter present a literature review on the anisotropy characteristics of HCP metal, and developed anisotropy yield and fracture criterion, focusing principally on titanium and magnesium alloys. The third chapter present the method used for the research work, including experimental and numerical simulation methods. Afterwards, two main chapters explained the results of magnesium and titanium alloys, respectively on anisotropy influence on plastic and fracture behaviour where the technologies of Hot tensile tests were applied, and based on the experimental results, modelling to predict the anisotropic plastic and fracture behaviour. The last chapter summaries and concludes the result of this research work and brief description of the future works.

Chapter

2 Literature review

2.1 HCP metals characteristics

The Hexagonal Close Packed (HCP) has a coordination number of 12 and contains 6 atoms per unit cell. The HCP structure has three layers of atoms. In both the top and bottom layers, there are six atoms that arrange themselves in the shape of a hexagon and a seventh atom that sits in the middle of the hexagon. The middle layer has three atoms nestled in the triangular grooves of the top and bottom plane. Slip in HCP metals is much more limited than in body-centered cubic (BCC) and face-centered cubic (FCC) crystal structures. In general, HCP metals are characterized by high anisotropy of mechanical properties with regard to crystallographic directions. Usually, HCP crystal structures allow slip on the densely packed basal planes. The activation of other slip planes depends on various parameters, such as the c/a ratio. Since there are only three independent slip systems on the basal planes, for arbitrary plastic deformation additional slip or twin systems need to be activated. This typically requires much higher resolved shear stress and results in the brittle behavior of HCP polycrystals. HCP metals like cobalt and zinc are not as ductile as FCC metals. Some HCP metals are very brittle (such as magnesium) and some are very ductile (such as titanium). Many other features depend upon the crystal structure of metals, such as density, deformation processes, alloying behavior, and much more. Thus, it is important to understand metal structures. The hexagonal structure is shown in Figure 2.1. It provided the c/a ratio is 1.633. A $c/a = 1.633$ is required for perfect packing of spheres. Magnesium is nearest to the perfect number with 1.624 c/a ratio, Titanium less-than ideal c/a ratio of 1.587.

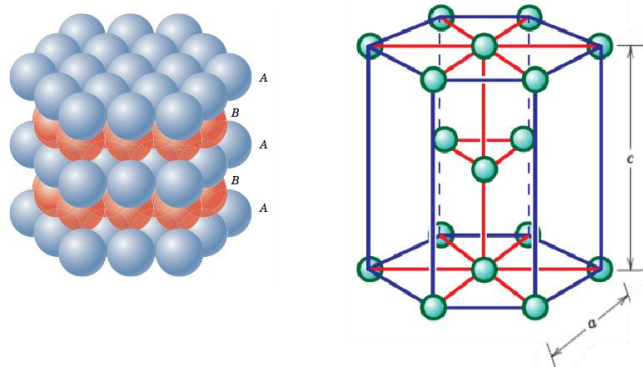


Figure 2.1 The HCP crystal structure [1].

2.1.1 The microstructural deformation mechanism of HCP metal.

In order to understand the microstructure deformation mechanism of HCP metal. A basic knowledge developed on the magnesium alloys as an example, and titanium alloys have similar HCP crystal structure. In magnesium, the basal slip $(0001)\langle 11\bar{2}0 \rangle$ is the most easily activated, because the Critical Resolved Shear Stress of basal slip is much smaller than that of other slip systems. Therefore, it is relatively easy to activate it. Deformation twinning due to c-axis tension $\{10\bar{1}2\}$ twinning is another major deformation mechanism in magnesium [2][3]. Plane strain compression in the direction perpendicular to the c-axis of a single crystal showed that slip systems were activated after almost the entire grain was twinned [4]. In terms of non-basal slip, the principal activation of prismatic slip $\{10\bar{1}0\}\langle 11\bar{2}0 \rangle$ was observed derived from tensile tests perpendicular to the c-axis [5]. In literature [6] reported that pyramidal-1 $\{10\bar{1}0\}\langle 11\bar{2}0 \rangle$ is an active system in a specified direction at room temperature. Whereas, in literature [4] showed that neither prismatic slip nor pyramidal slip accommodate significant amount of strain at room temperature when plane-strain compression along c-axis. The slip system with $\langle 11\bar{2}0 \rangle$ slip direction and $\{1012\}$ twinning, as mentioned above, undergoes no deformation upon compressive strain along the c-axis; therefore, arbitrary deformation cannot be realized [7]. The activation of pyramidal-2 slip $\{11\bar{2}2\}\langle 11\bar{2}3 \rangle$ was also reported in literature [8]. Figure 2.2 (a) shows the slip plane normal vectors and slip direction vectors for these slip systems, and the twinning plane normal vectors and shear direction vectors for twinning. Figure 2.2 (b) reports the schematic diagram of slip and twinning systems..

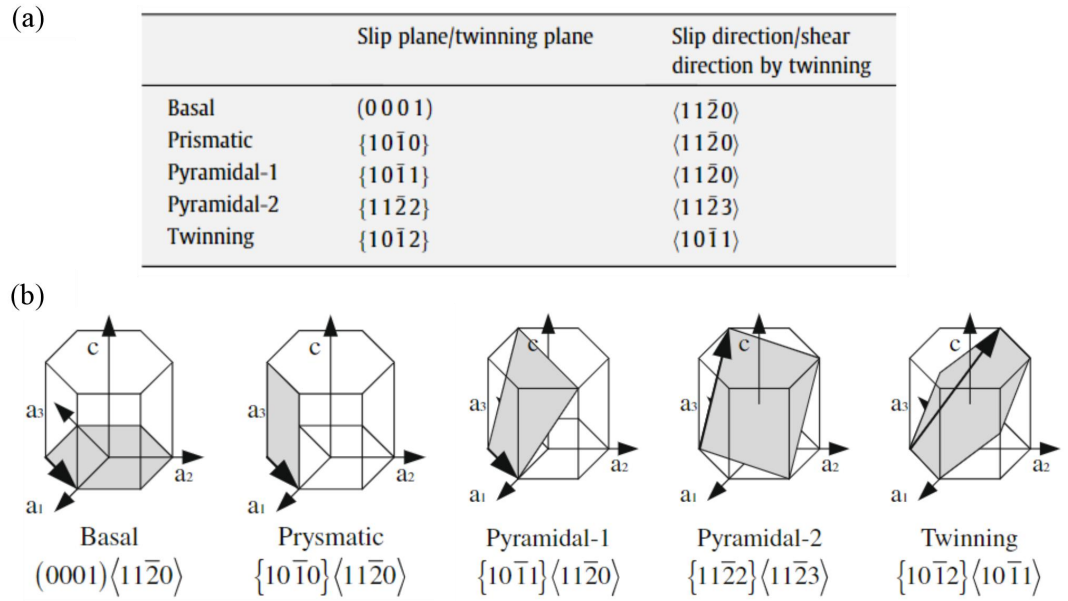


Figure 2.2 (a) Definitions of slip/twinning systems, and (b) Schematic diagram of slip and twinning systems [9].

The texture evolution of magnesium alloys have been investigated as an example. In literature [10], the texture evolution of AZ31B magnesium alloy studied by tracking one single grain orientation, as depicted in Figure 2.3 (a) (b), it was shown that the basal slip that splits the basal texture and prismatic slip causes the two maxim in (10-10) polar figure. Although basal and prismatic slips are the two main deformation modes in RD tension and compression, they actually cause c-axis and $\langle 1010 \rangle$ direction to be rotated to different positions, depending on the loading condition. The normal direction of slip plane is rotated towards the loading condition in compression, while it is rotated towards the direction perpendicular to the loading condition in tension [9]. Figure 2.3 (c) reports that, c-axis can be rotated from the initial position to the center of (0002) PF by basal slip in rolling direction tension. In literature [11], a crystal plasticity finite element model developed to understand the texture evolution of TA15 titanium alloy sheet during hot tension deformation at 750°C.

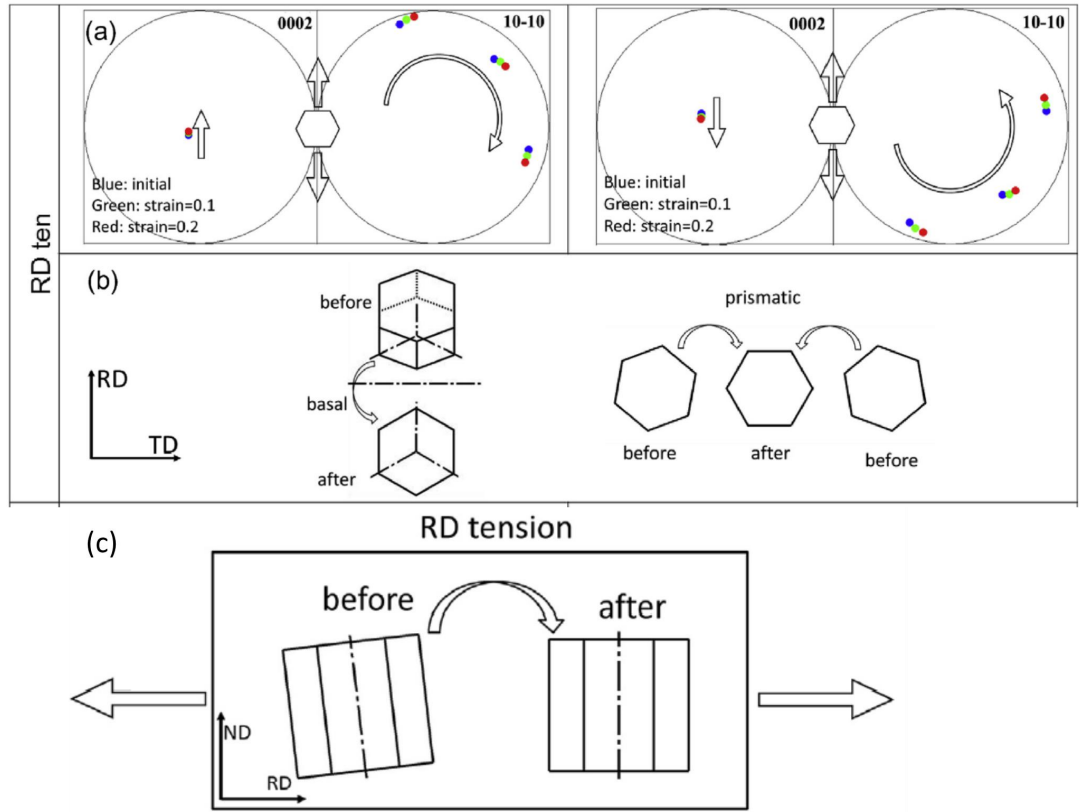


Figure 2.3 Trace of one typical grain rotation and a schematic of grain rotation in RD tension at 200°C: (a) (b) grain rotation in RD tension and the corresponding schematic and (c) a schematic of grain rotation in RD tension [10].

2.1.2 The anisotropic characteristic

Hexagonal Close-Packed (HCP) structure may induce significant anisotropy of the mechanical characteristics and fracture behaviour, which can be of great concern in sheet metal forming, since the as-delivered sheet texture may enhance the material anisotropic characteristics. Therefore, the material anisotropy is generally quantified through two main parameters, namely the normal average anisotropy, $\bar{R}$ (Eq. (2.1)), and the planar anisotropy, ΔR (Eq. (2.2)). The $\bar{R}$ determines thinning behaviour of sheet metals during stretching. The high value of $\bar{R}$ illustrate the thickness resistance is high, much less deformation on the thickness direction than on the width direction. The ΔR exhibits different behaviour in various planar directions. The high value of ΔR illustrate the ears will develop during deep drawing processes.

$$\bar{R} = \frac{r_0 + 2r_{45} + r_{90}}{4} \quad (2.1)$$

$$\Delta R = \frac{r_0 - 2r_{45} + r_{90}}{2} \quad (2.2)$$

The Lankford coefficient or anisotropy coefficient r_i that is the ratio between the specimen strain in the width (ε_w) and the one in the thickness (ε_t). They are

defined in terms of the true strains that the specimen undergoes in tension. (Eq. (2.3)).

$$r_i = \frac{\varepsilon_w}{\varepsilon_t} \quad (2.3)$$

The specimen orientation, namely 0° , 45° and 90° with respect to rolling direction was shown in Figure 2.4 (a) and the strains in Figure 2.4 (b) are used in determining the normal and planar anisotropy of the sheet metal.

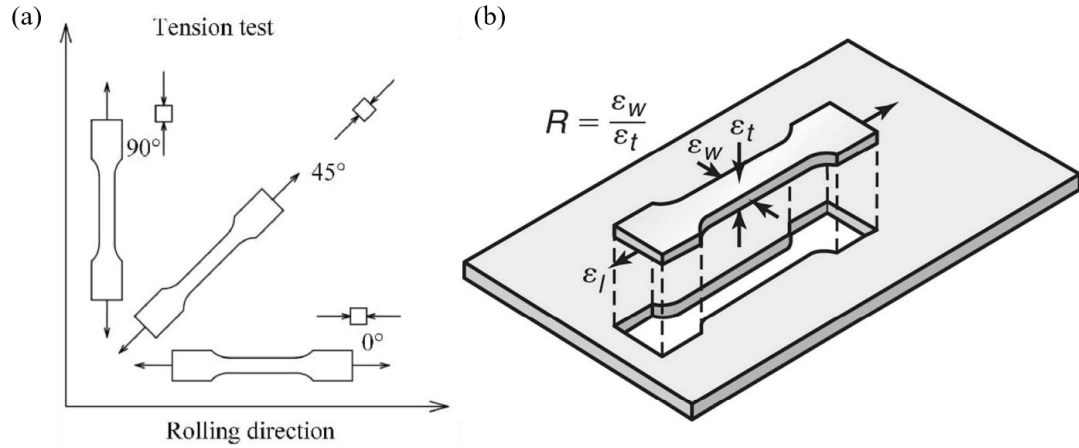


Figure 2.4 (a) Tensile-test specimen cut from the rolled sheet at 0° , 45° and 90° direction with respect to rolling direction [12] and (b) Strains on a tensile-test specimen removed from a piece of sheet metal in direction i [13].

2.2 Titanium alloys

Titanium is typically combined with iron, aluminium, molybdenum and vanadium to form alloys. These alloys have good properties such as, high strength, light in weight, highly resistant to corrosion and can withstand extremely high temperatures. Ti6Al4V alloy is the most commonly used titanium alloys.

Ti6Al4V also called a grade 5 titanium alloy because it is made of 6% aluminium (Al) and 4% vanadium (V), along with about 0.25% iron and 0.2% oxygen, with the remainder being titanium (Ti). It is also called an alpha-beta alloy (because it consistent aluminium, which is an alpha stabilizer and vanadium, which is a beta stabilizer) and able withstand heat treatment. Although its excellent properties, the high cost of titanium limits the use of titanium alloys to military applications, spacecraft, aircraft, and expensive sports cars, electronic goods, and sports equipment, as shown in Figure 2.5.



Figure 2.5 The application of titanium alloys in aircraft and sport fields [14].

The Ti6Al4V alloy is highly resistant to corrosion, mainly due to a continuous and consistent layer of oxide that forms spontaneously when exposed to oxygen. This property helps it withstand the sea water and makes it an ideal choice for underwater and marine applications like submarines, and offshore oil and gas operations equipment, as well as applications in the chemical industry. High tensile strength and light weight/low density: The military and the aviation and aerospace industries have long benefited from these properties of Ti6Al4V. It is used in the structural components of aircraft, hydraulic systems, engine components, helicopter rotor blades, rockets and spacecraft. Its high strength-to-weight ratio is particularly useful in race cars and is used in several components including gear boxes. The military uses it in the manufacture of missiles and artillery. The high strength of this alloy also means that it can be used for dental implants titanium screws are fused to the jawbone to replace a deformed or missing tooth, which looks natural and lasts a lifetime [1], [16][17]. Ti6Al4V can withstand extremely high temperatures and can be easily welded and fabricated, which makes it particularly attractive for the automotive industry, for various engine components, gear boxes etc. Excellent biocompatibility is the amazing properties of Ti6Al4V, making it a wonder alloy for the medical domain [15], namely, it does not react with body fluids, surrounding tissues or bones and helps with osseointegration. Porous structures made from the alloy allow vascularisation and adherence of the tissue to the implant, thereby speeding up recovery in patients who have undergone hip replacement and knee replacement surgeries. The alloy is also used in prosthetics, bone plates and rods (Figure 2.6). Ti6Al4V is non-magnetic, which means that a person with this implant in the body, can safely undergo X-ray, MRI and others scans. Ti6Al4V is used in medical equipment and surgical devices like needles, tweezers, forceps, clamps, scissors, drills etc.

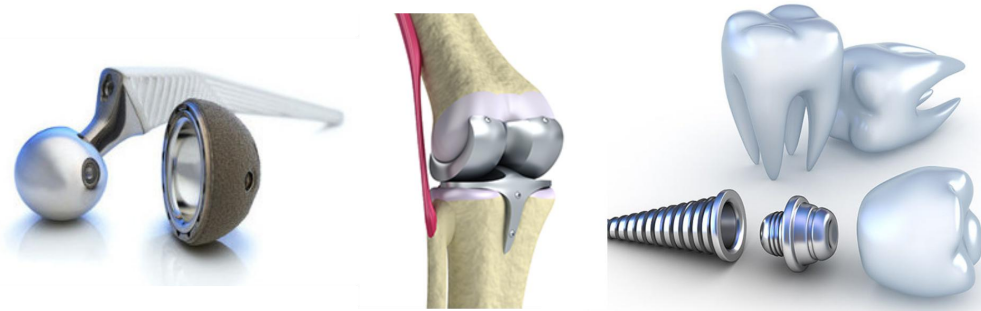


Figure 2.6 The application of titanium alloys in medical domain [15].

This alloy truly has many good properties, but also has a few drawbacks, just like any other materials: Ti6Al4V has poor shear strength and surface wear properties, because of which it tends to slip or seize when in contact with sliding parts. In order to overcome this, oxidizing and nitriding treatments are employed. Though generally resistant to corrosion, the Ti6Al4V alloy is not suitable for use with strong acids like hydrochloric or sulphuric acids, as they cause a breakdown of the protective oxide layer on the alloy, upon contact. The Ti6Al4V has proved to be an amazing material that finds applications in a wide variety of industries - the amount of its use in our daily lives is tremendous, and will only continue to grow with advancement in technology and discovery of new uses for its properties.

2.2.1 The crystal characteristic of Ti6Al4V

Generally, pure titanium can crystallize in two crystal structures: α titanium and β titanium. When it crystallizes at low temperatures (room temperature), the hexagonal close-packed (HCP) structure of alpha titanium is formed. While it crystallizes at high temperatures, the body-centered cubic (BCC) structure of beta titanium is formed. The complete transformation from one to another crystal structure is called allotropic transformation. The respective transformation temperature is called the transus temperature. The HCP α titanium is stable at room/low temperatures, while the BCC β titanium is stable at high temperatures. The β transus temperature for pure titanium is $882 \pm 2^\circ\text{C}$. Ti6Al4V is a biphasic ($\alpha + \beta$) alloy, with the dominant alpha (α) phase at room temperature characterized by a Hexagonal Close-Packed (HCP) structure, as shown in Figure 2.7 On the one hand, the alpha phase, provides for high yield stress but limited ductility, whereas the toughness of this phase depends on its grain size. On the other hand, the BCC phase allows a larger number of active slip planes, resulting in higher ductility but lower yield stress. The crystal structure of hcp α and bcc β phase, the crystal structure was depicted in Figure 2.8. The limited number of slip planes of HCP structures is known to lead to poor formability, which forces to form Ti6Al4V sheets at elevated temperature [18].

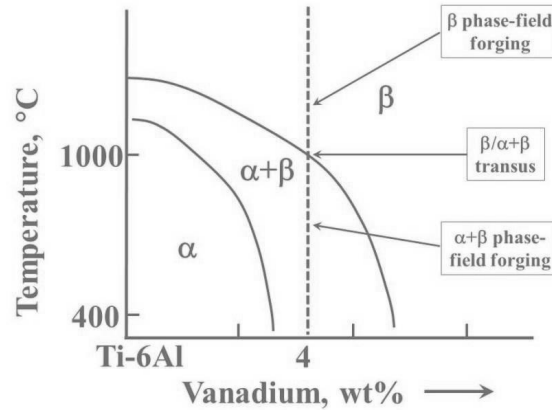


Figure 2.7 A phase diagram showing what happens to phase stability as vanadium is added to Ti6Al4V alloys [19].

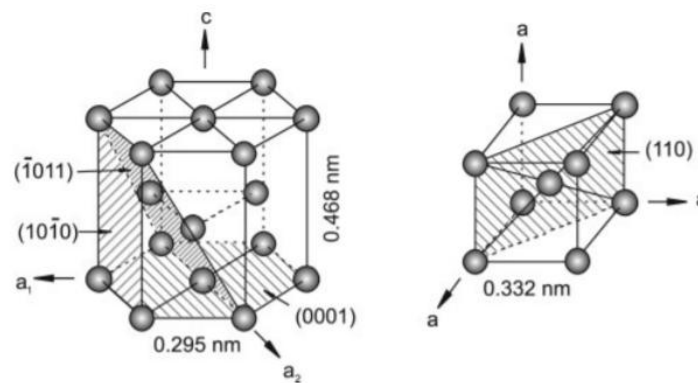


Figure 2.8 The crystal structure of α titanium of HCP and β titanium of BCC [20].

2.2.2 The mechanical characteristic of titanium alloys

Titanium alloys components are more difficult to be fabricated than other traditional metallic materials because of their poor formability at room temperature, high flow stress at elevated temperatures [21]. The microstructure and mechanical properties of two-phase $\alpha+\beta$ titanium alloys after different heat treatment have been studied in literature [22]. The literature concluded that the cooling rate from the β -phase range and ageing conditions had an effect on the microstructure parameters, volume fraction and chemical composition of the β -phase has a significant effect on the mechanical properties of Ti6Al2Mo2Cr alloy. In literature [23], the anisotropy of mechanical properties of commercially-pure titanium have been investigated, the results shown that the yield stress is clearly higher for TD-specimen compare with other direction (Figure 2.9). At the same time, the hardening rate is considerably stronger all over the deformation curve for specimen orientated at rolling than for specimen at transverse direction. The literature review presents a few papers devoted to the experimental evaluation of the anisotropic characteristics of titanium alloys sheets, most of which refer to room temperature testing as in literature [24] the anisotropy of mechanical properties of a near α -titanium alloy Ti60 was evaluated at 600°C, that was shown an obvious anisotropy, namely the deformation in transverse direction was much harder than in rolling direction, no matter the loading stress high or low. The strain rate effect on the Ti6Al4V sheets using tensile

test were shown in literature [25], the true stress decreasing as decreasing strain rate, as reported in Figure 2.10.

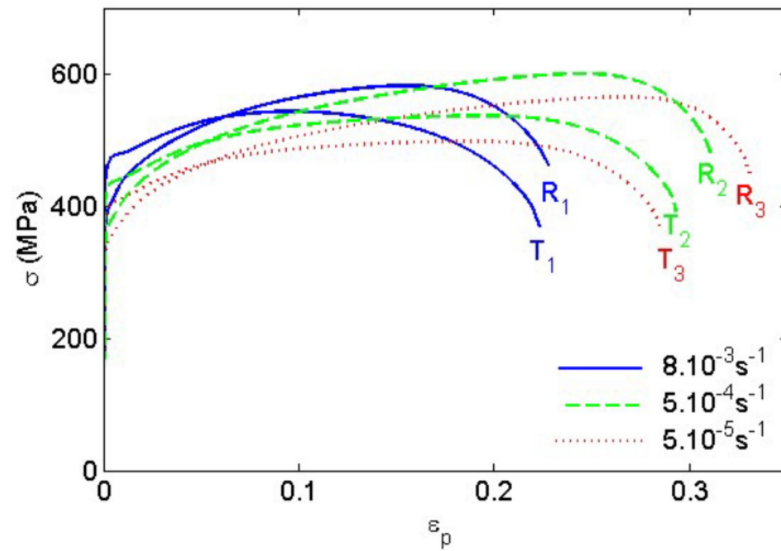


Figure 2.9 True stress versus true strain tensile curves of RD-specimens (referred to as R_{1-3}) and TD-specimens (referred to as T_{1-3}) for three different initial strain rates: $8 \times 10^{-3} \text{ s}^{-1}$, $5 \times 10^{-4} \text{ s}^{-1}$ and $5 \times 10^{-5} \text{ s}^{-1}$ [23].

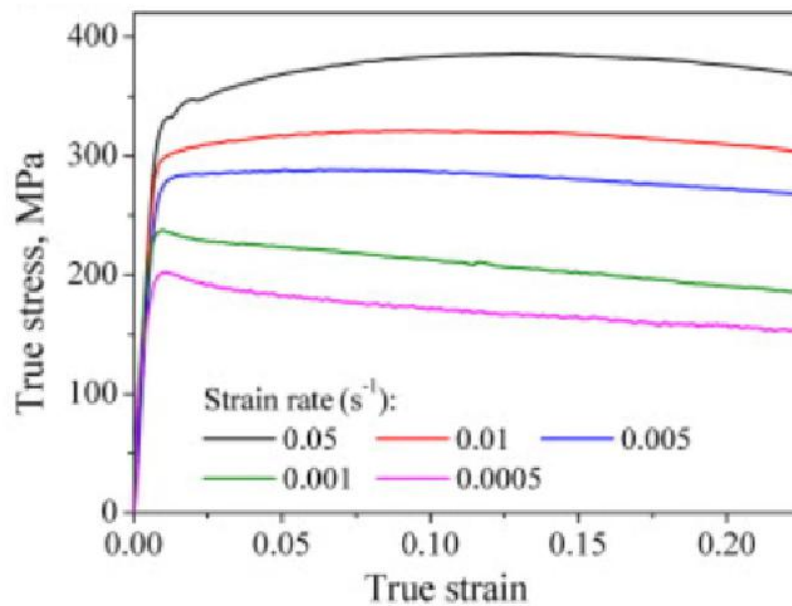


Figure 2.10 True stress–strain curves 600°C for different strain rates [25].

2.3 Magnesium alloys

Magnesium alloys are mixtures of magnesium with other metals like aluminum, zinc, manganese, silicon, copper, rare earths and zirconium. Magnesium is the lightest structural metal currently available in the world. Its approximately 34% lighter by volume than aluminum and 50% lighter than titanium [26]. Therefore, magnesium alloys sheets are ideal candidates for lightweight structure applications, in the field of automotive (Figure 2.12) [27] [28], especially in the transport

industry (Figure 2.11) to reduce energy consumption and greenhouse gas emissions [29], thanks to their low density, high specific strength and stiffness as well as good shock absorption capacity. Moreover, thanks to their high corrosion resistance and biodegradability, they can be also used for biomedical applications, such as for temporary implants [30]. However, despite significant weight saving potential, magnesium alloys remain lowly utilized. The advanced high strength steels were more popular used [31] due to an excellent combination of low cost and high strength; Aluminum alloys with the good properties of low density and relative compatibility with existing manufacturing infrastructure, are experiencing great growth in closure and body applications; Carbon fiber reinforced polymer composites are finding application in lower volume vehicles such as the BMW i3 and Corvette Stingray as development efforts deliver lower cost and new manufacturing technologies. However, magnesium alloys make up less than 0.5% of the weight of an average vehicle [31] due to challenges associated with manufacturing and processing, assembly, in-service performance, and cost [32].



Railway rolling stocks

Vehicle as the rear seat support

Shells in phone and notebook

Figure 2.11 The example of magnesium application [29].

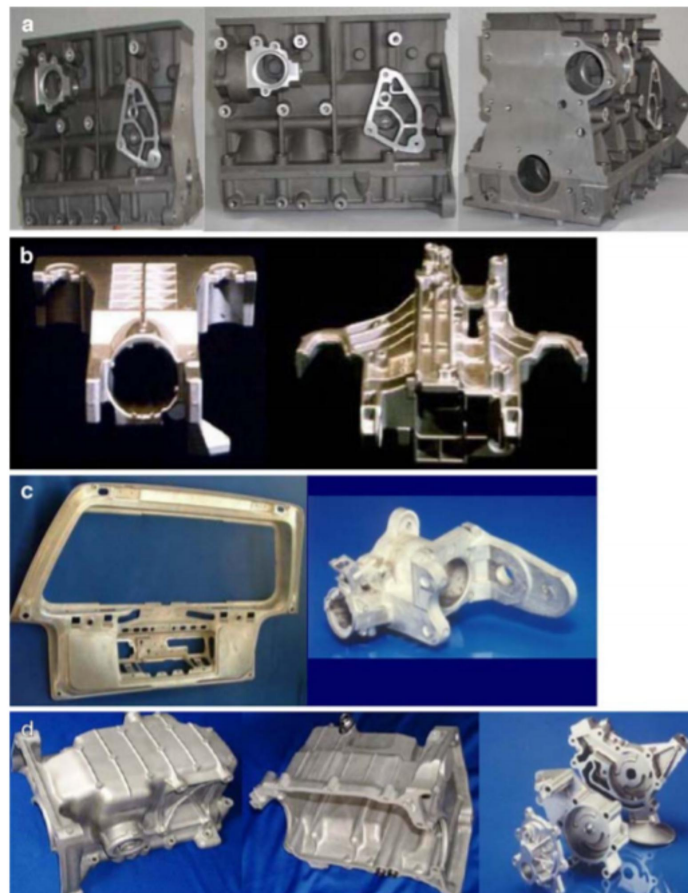


Figure 2.12 Examples of automotive components made of Mg alloys (a: Engine block, b: Steering column module, c: Door frame / Key lock housing, d: Oil pan) [28].

2.3.1 The crystal characteristic of AZ31B alloy

Magnesium sheets with a hexagonal close-packed atomic structure develop a strong (0001) basal texture during rolling [33], with the c-axis parallel to the thickness direction, which makes the deformation along the thickness direction restricted, therefore, promotes early failure. This trait is particularly detrimental when the sheets are deformed at room temperature. Deformation modes of magnesium alloys include basal slip, prismatic slip, pyramidal slip and twinning. The flow stress of a polycrystal is determined by the Critical Resolved Shear Stress (CRSS) of the deformation mechanisms. It is generally assumed that, at low temperatures, $CRSS_{\text{basal}} < CRSS_{\text{twinning}} < CRSS_{\text{prismatic}} < CRSS_{\text{pyramidal}}$ [34][35], as shown in Figure 2.13. Thus, basal slip and twinning are the preferred deformation mechanisms in randomly oriented polycrystals. On the contrary, at higher temperatures, the CRSS of non-basal systems become active leading to an overall enhancement of the magnesium sheets formability [36]. At room temperature, magnesium alloys own two independent basal slip systems and twinning as deformation modes [37], namely Compression Twins (CTs) and Extension Twins (ETs), as shown in Figure 2.14, which owe the names to their contraction and extension along the c-axis, respectively. By virtue of the limited number of active slip systems, the ductility of magnesium alloys is therefore restrained.

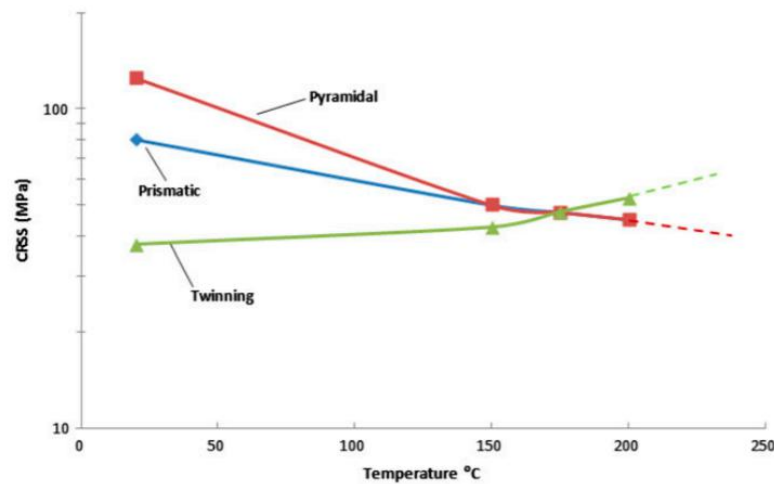


Figure 2.13 CRSS of the deformation systems with respect to the temperature [34].

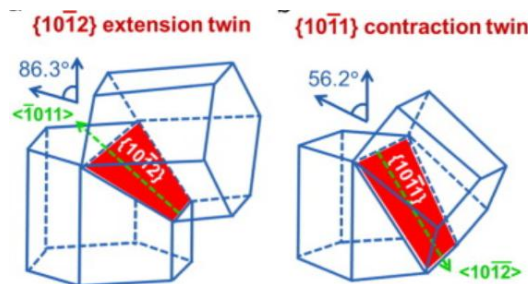


Figure 2.14 Schematic of $\{10\bar{1}2\}$ extension and $\{10\bar{1}1\}$ contraction twin modes in magnesium alloys [38].

2.3.2 The mechanical characteristic of magnesium alloys

At room temperature, magnesium alloys own two independent basal slip systems and twinning as deformation modes [37], which are responsible of their poor formability at room temperature. To achieve sufficient ductility, forming of magnesium alloy sheets is usually carried out at elevated temperature. Therefore, it is mandatory a properly knowledge of the formability and fracture behaviour of magnesium alloys at varying temperature, besides at varying stress state and rolling direction. The mechanical behaviour of AZ31 magnesium alloy sheets was investigated at different temperatures and strain rates in [44] where the temperature and strain rate were found to influence the flow stress behaviour of twin-roll cast, rolled and heat-treated AZ31 alloys tested under uniaxial tension. In [39] the mechanical properties of AZ31 alloy sheets were characterized using a numerical inverse analysis procedure in a range of temperature between 100°C and 300°C based on uniaxial tensile tests, Polycrystal numerical simulations were used in [40] to attempt to find a relationship between anisotropy and texture evolution at elevated temperature. In [41], it was reported twinning formation as a consequence of a tensile load application at the end of uniform elongation and the damage of the AZ31 alloy was related to ductile dimple formation and presence of second-phase particles at room temperature. In [42], it was reported that the tensile-induced twinning played an important role in increasing the toughness of notched specimens of the rolled AZ31 alloy.

In [43], both uniaxial tensile tests and Nakajima tests in the temperature range varying from 200°C to 300°C and at different strain rates were carried out. The temperature increase led to the flow stress reduction and improvement of material formability at necking, whereas a strain rate increase, besides the increase in the flow stress, led to the material formability decrease. In [44], the mechanical behaviour of Twin-Roll Cast (TRC), Rolled (R) and Heat-Treated (HT) AZ31 magnesium alloy sheets was investigated at different temperatures and strain rates under uniaxial tension. Besides confirming the effect of the temperature and strain rate on formability reported in [43], the deformation mechanisms that underlain to the mechanical behaviour of the magnesium alloys were reported and discussed. At high temperature, the occurrence of Dynamic Recovery (DRV) and Grain Boundary Sliding (GBS) led to the reduction of the CRSS requested for the easy movement of dislocations, leading, in turn, to the decrease of the flow stress. On the contrary, at high strain rate, the flow stress increased: it was shown that the strain rate affected the CRSS of the non-basal slip systems, and, at the same moment, induced a higher amount of twins at the grain boundaries. The latter contributed to increase the stress required for achieving plastic deformation, since twins interrupted the free path of dislocations, reducing their accumulation along grain boundaries. In [45][46] it was found that the change in the orientation of twinned grains at moderate temperatures

might cause hardening or softening on the basis of the initial sheet texture. At higher temperature, the fracture strain is higher indicating more ductile behaviour, since grains are refined after recrystallization as well as the CRSS values for the different slip modes including twinning are closer. Sheet anisotropic characteristics are reported in [47] till 200°C, showing larger Lankford coefficients at high strain rate. The temperature influence on the Lankford coefficients of AZ31 magnesium alloy sheets at high strain rate is reported in [48]. However, much less literature records are available about the anisotropy effect on Lankford coefficients, diffuse necking strain and fracture strain at elevated temperature. In [49], the mechanical behaviour of AZ31B magnesium alloy sheets was evaluated on specimens oriented at 0°, 45°, and 90° with respect to the Rolling Direction (RD). It was shown that the specimen orientation led to significant differences in both the yield stress and anisotropy coefficient.

2.4 Anisotropic yield model for metals

An accurate prediction of flow behavior of metals considering the combined effects of strain, strain rate and temperature is essential for understanding flow response of metals. Many anisotropic yield functions such as Hill (1948) [52], Hill (1979) [53], Barlat and Lian (1989) [54], Barlat et al. (2003) [55], Plunkett et al. (2008) [56] have been developed so far. Ti6Al4V elasto-plastic behaviour taking into account the material anisotropy was mainly investigated at room temperature, as in [50] where different anisotropic yield criteria were calibrated through a series of tests performed on specimens cut at different rolling directions. On the contrary, few literature records pertain to modelling of such behaviour at temperatures higher than the room temperature: an example is given in [51] where different anisotropic yield criteria were calibrated to simulate the deep drawing process conducted at 400°C on Ti6Al4V sheets. From the literature review, due to the lack of suitable macroscopic constitutive models, the classical anisotropic yield function like Hill 1948 yield criterion, accompanied with the conventional isotropic hardening approach, is still applied to represent the behaviour of Ti6Al4V in sheet forming simulations [57]. Recently, some other yield criterion developed for HCP structure metal. In literature [58] reported an anisotropic-asymmetric yield criterion, namely, CPB06 which take into account the anisotropy and the strength differential effect related with twinning for pressure insensitive materials. As described in the literature that CPB06 is sufficient suitable in predicting the yield surfaces of CPTi and magnesium [58]. Khan et al. in literature [59] also used the CPB06 criterion to predict the yield locus of Ti6Al4V while resulted a low level of accuracy. Plunkett et al. in literature [60] applied the orthotropic asymmetry yield criteria CPB06, CPB06ex2 and CPB06ex3 in combination with a hardening model which accounts for directional hardening due to texture evolution. With regard to describing the in-plane anisotropy of the sheet, the accuracy of CPB06 is lower compared to the other two yield functions CPB06ex2 and CPB06ex3. Additionally, these yield functions led to significant differences with measured yield surfaces for Ti64 in biaxial and plane stress

conditions. A comparative study has been carried out in literature [61] to assess the accurate of Khan Huang Liang (KHL), Johnson Cook (JC), Fields Backofen (FB) and Mechanical Threshold Stress (MTS) models for prediction of the flow stress behavior for the temperature range of 50°C-400°C at strain rates of 10^{-5} , 10^{-4} , 10^{-3} and 10^{-2} s^{-1} . The result reported that MTS model is more suitable for prediction of flow stress of Ti6Al4V alloy at elevated temperatures due to the physical based model are more preferred than phenomenological model. The literature review presents many papers devoted to the experimental evaluation or modelling of the anisotropic characteristics of titanium alloys sheets, most of which refer to room temperature condition. The investigation at elevated temperature has not been fully discussed.

2.5 Anisotropic fracture model for metals

2.5.1 Fracture-based forming limit criteria

In sheet metal forming, the traditional strain based forming limit diagram (FLD) pioneered by Keeler and Backofen [62] has been widely applied to evaluate localized necking and thickness reduction. It is expressed in the space of the major and minor strain by assuming proportional loading states under the plane stress condition. In sheet metal forming processes, the conventional FLD with the safety margin can sufficiently guarantee successful products without a defect such as a localized thinning of the sheet leading to a tear in sheet metals. The fracture forming limit diagram (FFLD) as an alternative to the necking-based FLD is derived by equating the fracture strain obtained from the fracture criterion with the equivalent plastic strain corresponding to the applied yield function. In literature [63], the author represented the fracture loci and FFLD of AHSS sheets during sheet metal forming processes, based on a theoretical derivation from an anisotropic ductile fracture criterion. To evaluate the anisotropic formability, the FFLC predicted from different directions of the maximum principal stress as shown in Figure 2.15. From the comparison of the fracture-based forming limits suggested, it can be recommended that the formability should be decided by the lowest forming limits over a wide range of loading conditions because the direction of fracture initiation cannot be predicted in complicated forming processes.

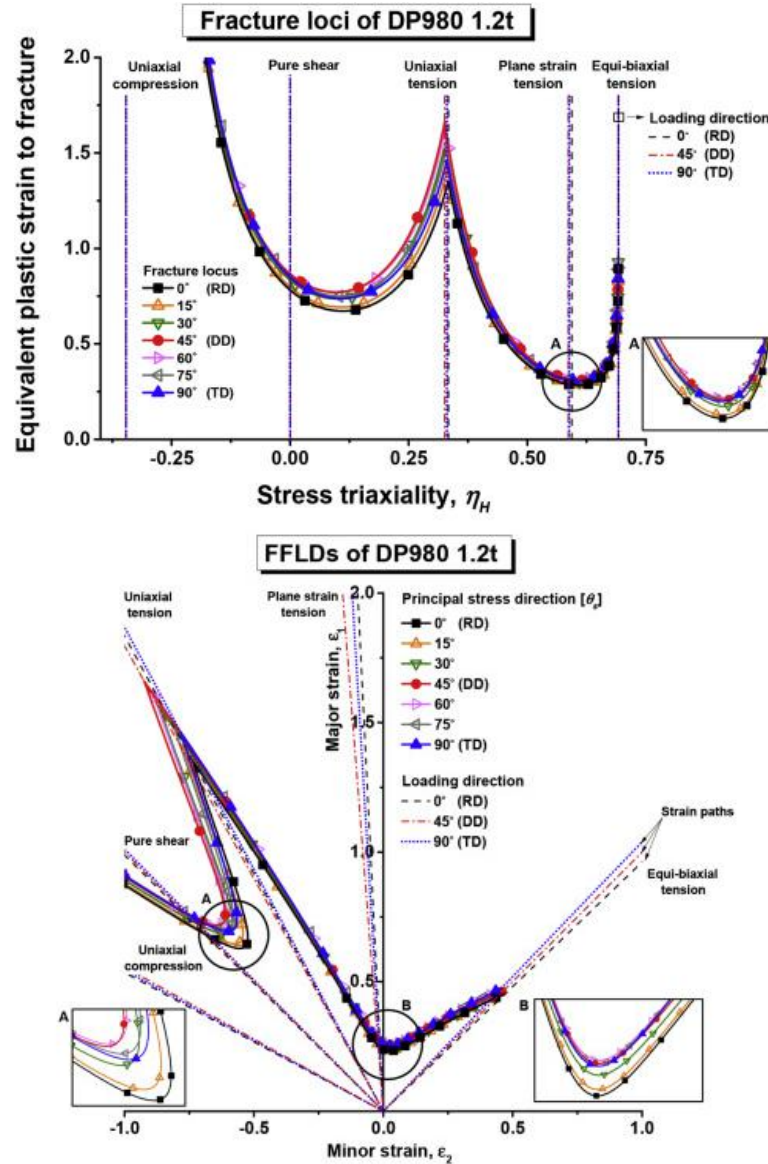


Figure 2.15 Fracture loci and its corresponding FFLDs at different directions of the maximum principal stress: (a) Fracture loci plotted according to the maximum principal stress direction; (b) FFLDs plotted according to the maximum principal stress direction. [63].

2.5.2 Stress state parameters

However, the conventional strain-based FLD is not appropriate for prediction of the fail with little amount of necking different from that with the conventional steel sheets, besides the well-known limitations, their determination at elevated temperature requires dedicated experimental set-up and great experimental effort, which make the method difficult to be effectively applied. A more feasible alternative is to use damage models taking into account the effective state of stress on the sheet during the forming operation.

The stress state of isotropic materials can be expressed with stress triaxiality (η) and Lode angle parameter ($\bar{\theta}$). The stress triaxiality is defined according to Eq (2.4) as the ratio between the hydrostatic pressure and Von Mises equivalent stress. The former has the expression in Eq (2.5), whereas the latter is calculated according to

Eq (2.6), both in the principal stress system. The deviatoric stress invariant J_3 is expressed in terms of the so-called Lode parameter according to Eq (2.7). In the following, the Lode parameter will be considered for the influence of the deviatoric stress. The range of the Lode angle parameter is $-1 \leq \bar{\theta} \leq 1$. The relationship between the stress triaxiality and Lode angle parameter is defined in Eq (2.8), and the relation between the stress state parameter is expressed as shown in Figure 2.16. The stress triaxiality has the values of uniaxial tension ($\eta=1/3$, $\bar{\theta}=1$), plane strain ($\eta=1/\sqrt{3}$, $\bar{\theta}=0$), and equi-biaxial tension ($\eta=2/3$, $\bar{\theta}=-1$) in tension area ($1/3 \leq \eta \leq 2/3$).

$$\eta = \frac{-p}{\bar{\sigma}} = \frac{\sigma_m}{\bar{\sigma}} \quad (2.4)$$

$$p = -\sigma_m = -\frac{1}{3}(\sigma_1 + \sigma_2 + \sigma_3) \quad (2.5)$$

$$\bar{\sigma} = \sqrt{3J_2} = \frac{1}{\sqrt{2}} \sqrt{(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2} \quad (2.6)$$

$$\bar{\theta} = 1 - \frac{2}{\pi} \arccos \left[\frac{3\sqrt{3}}{2} \frac{J_3}{(J_2)^{3/2}} \right] = 1 - \frac{2}{\pi} \arccos \left[\frac{27}{2} \frac{J_3}{\bar{\sigma}^3} \right] \quad (2.7)$$

$$\bar{\theta} = 1 - \frac{2}{\pi} \arccos \left[\frac{27}{2} \eta \left(\eta^2 - \frac{1}{3} \right) \right] \quad (2.8)$$

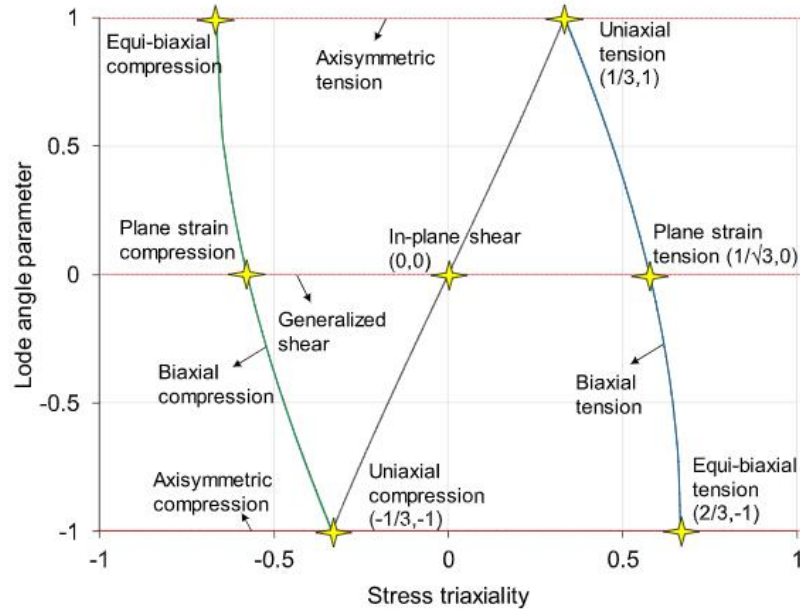


Figure 2.16 Stress states on the stress triaxiality and Lode angle parameter space [64].

Besides, the effect of the stress triaxiality on the flow stress and fracture strain of the AZ31 magnesium alloy at room temperature was studied in [65][66]. Different specimen geometries were used to carry out tensile tests on round specimens at

room temperature in the literature to obtain a wide variety of stress state triaxialities. It was reported an evidence plastic anisotropy influence on stress triaxiality, which was due to the activation of twins. At medium values of stress triaxiality, the fracture process was mainly because of void growth and coalescence instead of twin-size micro-cracks at higher triaxiality values. Therefore, under medium levels stress triaxiality, ductility is the highest and decreases as the stress triaxiality increases. In addition, anisotropy plays an important role in strain regulation during deformation.

2.5.3 Modelling of material anisotropic fracture criterion

Sheet metal typically exhibits varying degrees of anisotropy due to the formation of texture during rolling process. The grain structure and distribution in the normal direction of the sheet may also be different from those within the sheet plane. Due to the anisotropy characteristics, the fracture behaviour may also effected. Lou and Yoon in 2017 [67] stated that there are two mathematical methods to extend the isotropic ductile fracture criterion to an anisotropic one. One of the method is to couple of anisotropy with the equivalent plastic strain increment. In literature [68], the plastic anisotropy on ductile fracture implemented by using the Hill 1948 anisotropic equivalent plastic strain instead of Mises equivalent in the fracture criterion. Also Lou and Yoon [67] introduced the anisotropy into the equivalent plastic strain increment by a linear transformation tensor to develop the anisotropy ductile fracture criterion. Another method is to take into account the anisotropy on the weighting function. Park et al in literature [63] introduced the anisotropic stress triaxiality to the isotropic Lou-Huh fracture criterion based on the Hill1948 criterion, obtained a anisotropic Lou-Huh fracture criterion. Then, Park et al [69] taken into account the equi-biaxial stress state in the anisotropic fracture criterion. Lou and Yoon in 2018 [70] proposed a ductile fracture criterion based on the stress invariant through the coupling the first stress invariant with Druker equation and to extent it to an anisotropic form. A generalized anisotropic ductile fracture criterion developed in literature [71], to describe the anisotropic fracture behaviors of a DP980 sheet over a variety of stress states and it is able to predict the anisotropic fracture behaviors of this DP980 sheet with high accuracy. But it still limited at room temperature. Thus, the effect of anisotropic plasticity is involved in the weighting function. However, less developed anisotropy fracture criterion also consider temperature distribution at the same time. Furthermore, the stress triaxiality and Lode parameter are play a significant role in fracture criterion development. Rice and Tracey [72] firstly introduced a stress triaxiality function to describe the nucleation and growth of voids leading to fracture; this function was proved to affect the fracture onset [73]. Wilkins et al. [74] were instead the first to study the effects of both the stress triaxiality and deviatoric stress parameter on the fracture occurrence at room temperature, proposing in [75] a new model to predict fracture with both these stress parameters implemented. In [76] a Johnson-Cook based criterion incorporating the effect of stress triaxiality and deviatoric stress

parameter was introduced to predict the Ti6Al4V sheets ductility at elevated temperature. Few paper investigate that the damage models taking into account both the stress state and anisotropy characteristics. In literature [77], the fracture loci and surfaces of anisotropic AA7050-T7451 plate in the different stress spaces were investigated to analyze the effect of plasticity anisotropy. As depicted in Figure 2.17(a), and in Figure 2.17(b) the anisotropic relation between stress triaxiality and Lode parameter developed on the base of modified elliptical fracture criterion incorporating Hill48's criterion.

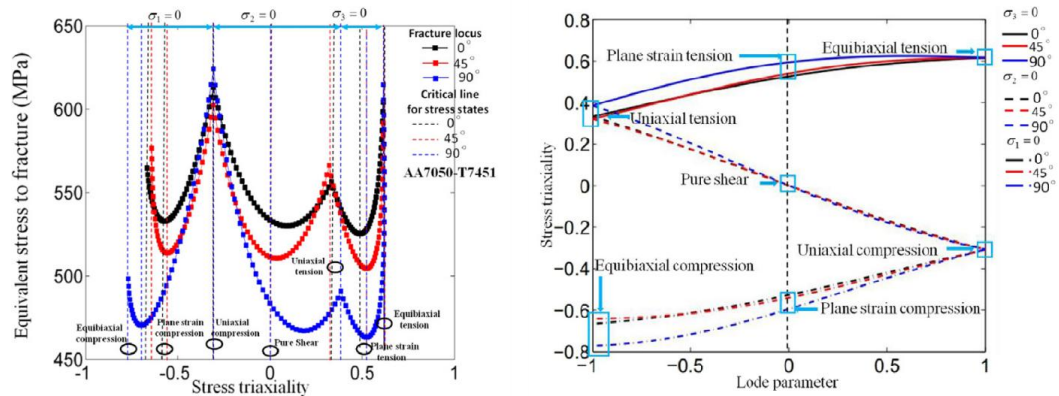


Figure 2.17 (a) The fracture loci of anisotropic AA7050-T7451 and (b) the anisotropic relation between stress triaxiality and Lode parameter with anisotropic parameters of AA7050-T7451 [77].

Modelling of fracture occurrence can be fulfilled using either fracture criteria or physically-based models: the former are quite easy to be calibrated and preferred to be applied in industrial environment thanks to their inherent simplicity, while the latter describe much more closely the fracture phenomenology, despite they need more complex calibration procedures that usually prevent their effective implementation. Within the numerous models available in literature to model damage evolution and fracture occurrence, the phenomenological ones appear the most suited thanks to their inherent easiness in implementation into commercial Finite Element (FE)-based codes devoted to simulation of sheet forming processes and readiness of calibration [78]. Scientific literature reports a few studies about the failure behaviour of titanium alloy sheets deformed at room temperature [79], taking into account the anisotropic behaviour on the basis of the plasticity model, whereas the possible anisotropic failure behaviour at elevated temperature hasn't been reported yet. In this thesis, the phenomenological damage model called GISSMO (Generalized Incremental Stress State dependent Model) is applied to predict the onset of failure of titanium alloys, taking into account the influence of the material anisotropy. The GISSMO model takes into account the dependence of the stress state through the stress triaxiality, whereas the Lode angle dependence is not considered, as shell elements are usually utilized when simulating sheet forming operations. The GISSMO model has been extensively used to predict failure of metals at room temperature: in [80], dual-phase steel sheets were formed and later on subjected to crash loading conditions and their failure behaviour analysed; in

[81], the model was applied to martensitic steel sheets under various stress state, whereas in [82] the comparison between the predictions of the GISSMO model and those of a fracture criterion based on fracture forming limits proved that the former was more accurate; in [83], it was proved that the GISSMO model reduced significantly the simulation time compared to other fracture criteria, being still capable of satisfactory prediction of the material failure. In case of magnesium alloys sheets, modelling of damage evolution and fracture occurrence, just a few examples of application of fracture criteria can be found. In [84], the Modified Mohr Coulomb (MMC) fracture criterion was considered, which provided accurate failure prediction at room temperature as a function of complex stress states, including the dependency on the Lode angle parameter. The MMC fracture criterion was also found suitable to predict the shear-induced fracture in sheet metal forming as in [85], whereas in [86] it was modified by adding a new plastic flow parameter to consider the sheet anisotropic behaviour. Only in [87], an effort to this regard was made: as a result, the MMC model was modified to account for forming temperatures of 180°C and 230°C using a coupled numerical-experimental investigation.

Chapter 3

3 Materials and methods

3.1 Materials

Ti6Al4V titanium alloy and AZ31B magnesium alloy sheets with a thickness of 1 mm which provided in annealed condition were selected as testing material, from which the specimens used in the experimental campaign were laser cut along three directions (namely 0° , 45° and 90° with respect to the rolling direction). The as-received Ti6Al4V alloy characterized by an equiaxed $\alpha + \beta$ microstructure (Figure 3.1 (a)) and the as-received microstructure of the AZ31 sheets (Figure 3.1 (b)) shows a dual grain size distribution, which is as the result of dynamic recrystallization during hot rolling.

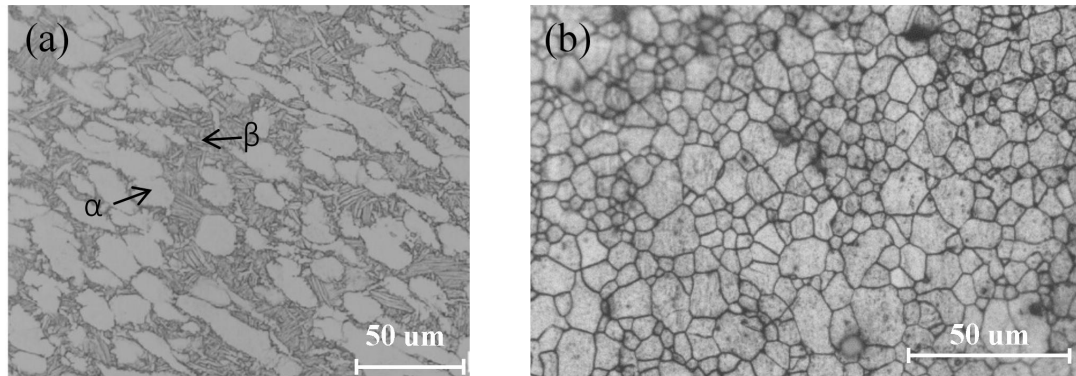


Figure 3.1 Microstructure of the as-received (a) Ti6Al4V and (b) AZ31B sheets.

3.2 Experimental apparatus and procedures

3.2.1 Uniaxial tensile and shear tests

Uniaxial tensile tests were carried out according to the ASTM E8/E8M-16a standard [88] using different specimen geometries. The tensile tests were conducted on a universal 5 t MTS™ testing machine equipped with a resistance heating system to heat up the specimen to the testing temperature. After 30 s of soaking time at the testing temperature to homogenise the thermal field, the specimens were strained until fracture and left to cool in calm air to room temperature. The Aramis™ system from GOM was used to capture the strain field evolution of the specimen during deformation till fracture. In order to record the strain evolution, a thin layer of white and black paint speckle patterns was applied on one side of the specimen. The Aramis™ recorded data were also used to calculate the Lankford coefficients, namely the plastic anisotropy coefficients, on the basis of the width and thickness strain values. The shear tests were carried out by using the same

experimental apparatus and procedure. The experimental set-up used for tensile testing is shown in Figure 3.2

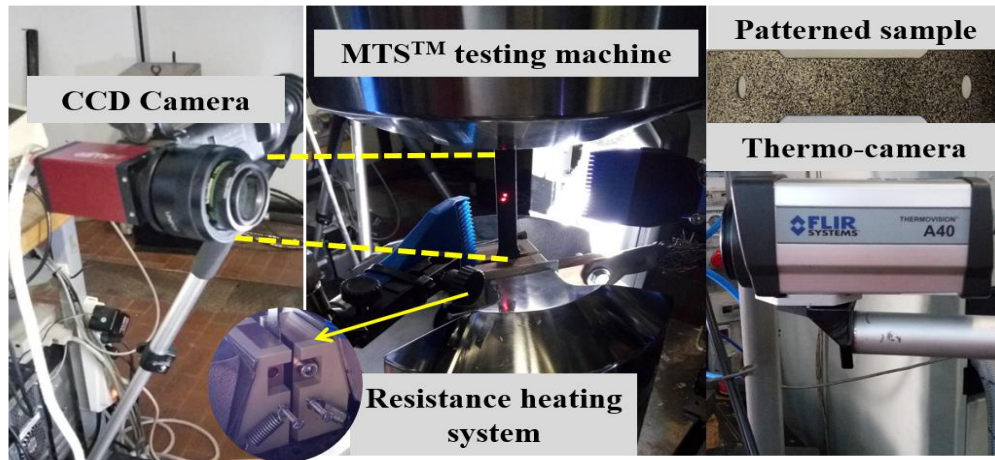


Figure 3.2 MTS™ testing machine equipped with the resistance heating system and Aramis™ system.

3.2.2 Nakajima tests

Nakajima tests were carried out according to the procedure reported in the ISO 12004 standard [89]. The Nakajima testing set-up is shown in Figure 3.3: it consists of a hemispherical punch (100 mm dia.) and a blank holder, mounted on a 200 kN hydraulic testing device. A 0.1 mm thick graphite foil is used as lubricant between the sheet metal specimen and the punch to reduce friction as much as possible. Cartridge heaters are inserted in the hemispherical punch to keep the temperature of the specimen during the experiments constant and equal to the testing temperature. An induction heating coil is used to heat up the specimen to the target temperature, which is continuously monitored through three K-type thermocouples spot-welded at the centre, intermediate and outer diameter of the specimen surface deformed by the punch and the temperature difference between them was kept under 10°C during the test. Two stereoscopic CCD cameras were used to monitor the specimen deformation during the test and the Aramis™ system allowed determining the major and minor strain values. The strain rate controlled in the test using numerical simulation method in which observe the specimen surface strain rate under different punch speeds.

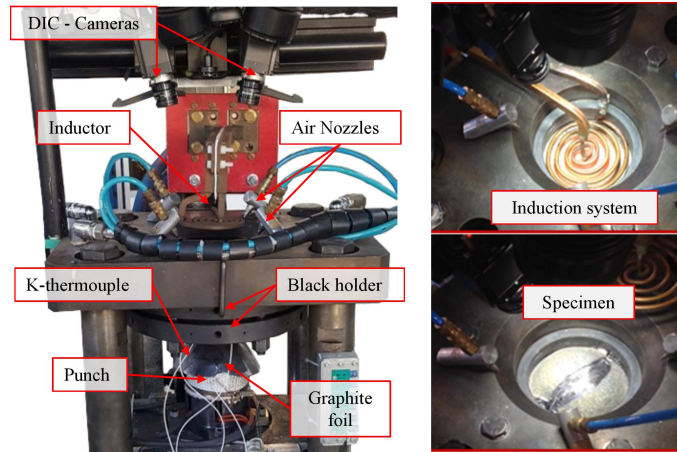


Figure 3.3 Nakajima testing equipment with the induction heating system.

3.2.3 The experimental plan for titanium alloys

Uniaxial tensile tests were carried out on smooth specimen in a wide range of testing temperatures (from room temperature to 800°C) and strain rates (0.01, 0.1, 1 s⁻¹) to assess the anisotropy effects. These tests allowed evaluating the flow behaviour, anisotropic and fracture characteristics at varying temperature, strain rate and rolling direction. The fracture strain values were recorded from the Aramis™ data.

Tensile tests on smooth and notched specimens, shear tests and Nakajima-type tests were carried out to cover a wide range of stress triaxiality, between simple shear and equi-biaxial tension, to both calibrate and validate the models used to describe the elasto-plastic and fracture behaviour of the Ti6Al4V sheets deformed at room and elevated temperature. The notch radii were chosen equal to 2 mm, 5 mm and 10 mm, whereas the shear specimens were machined according to the ASTM B831 standard [90], operating with a single shear zone, and providing the possibility to install the specimen in a universal testing machine. By the arrangement of the slots, the area between the fillets is sheared when the ends of the specimen are loaded with a tensile force (see geometries in Figure 3.4).

These tests were carried out at room temperature and 600°C, being the latter the temperature at which a remarkable change of material behaviour was found (see results in section 4.1) and, therefore, chosen for forming Ti6Al4V sheets at elevated temperature. The strain rate was set equal to 0.1 s⁻¹. All the tests were carried out at varying rolling direction in order to assess the anisotropy influence on the strain at fracture at varying stress state. The strain at fracture of the notched specimens was determined on the basis of the Aramis™ data. The whole experimental plan, comprising tensile tests on smooth and notched specimens, shear tests and Nakajima tests, was reported in Table 3.1.

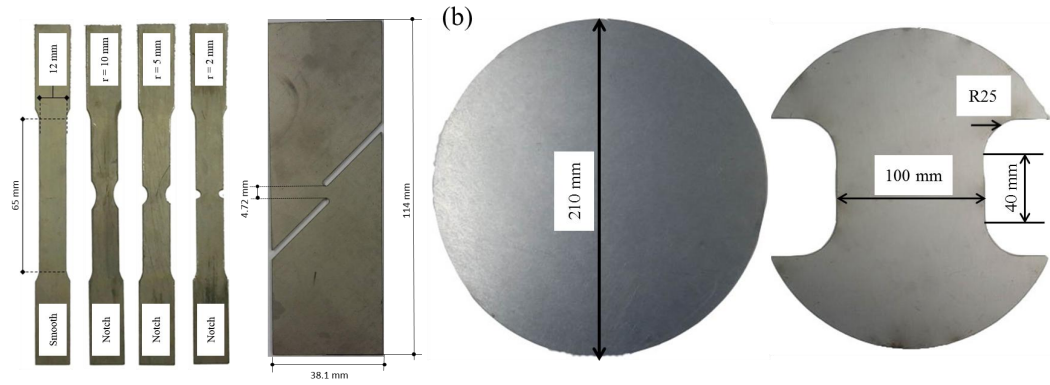


Figure 3.4 Geometries of the specimens.

Table 3.1 Experimental plan.

Specimen	Temperature (°C)							Strain rate (s ⁻¹)	Rolling direction (-)
2 mm notched	25			600				0.1	0°, 45°, 90°
5 mm notched	25			600				0.1	0°, 45°, 90°
10 mm notched	25			600				0.1	0°, 45°, 90°
shear	25			600				0.1	0°, 45°, 90°
Nakajima round	25			600				0.1	-
Nakajima notched	25			600				0.1	0°, 45°, 90°
smooth	25	200	400	600	700	800		0.01, 0.1, 1	0°, 45°, 90°

3.2.4 The experimental plan for magnesium alloys

Uniaxial tensile tests were carried out on smooth specimen at a strain rate of 0.1 s⁻¹ on a MTSTM universal testing machine equipped with a resistance heating system. The testing temperature ranged from 25°C to 300°C. For each testing temperature, the tests were carried out using specimens at varying rolling direction.

The tests were considered for positive stress triaxiality (0-0.6), namely uniaxial tensile, shear and equi-biaxial tensile tests. In order to have a more complete description of the material behaviour between the uniaxial tensile and shear stress states, the shear-tension compound type geometry employed in literature [91] was also selected. Hereafter, this specimen type is referred to as “special”. The specimen named “Nakajima round” is the one dedicated to equi-biaxial tensile tests. Along with different stress triaxiality states, tests were performed at different temperatures, namely 25°C, 100°C and 300°C. It is worth underlining that the testing temperatures of 100°C and 300°C were selected in order to appreciate the presence of twinning and recrystallization, respectively [92]. The strain rate was set equal to 0.1 s⁻¹ and kept fixed for all the tests. Each test was repeated at least twice to confirm the reliability of the results. The specimen dimensions for magnesium alloy are shown in Figure 3.5. The whole experimental plan is shown in Table 3.2.

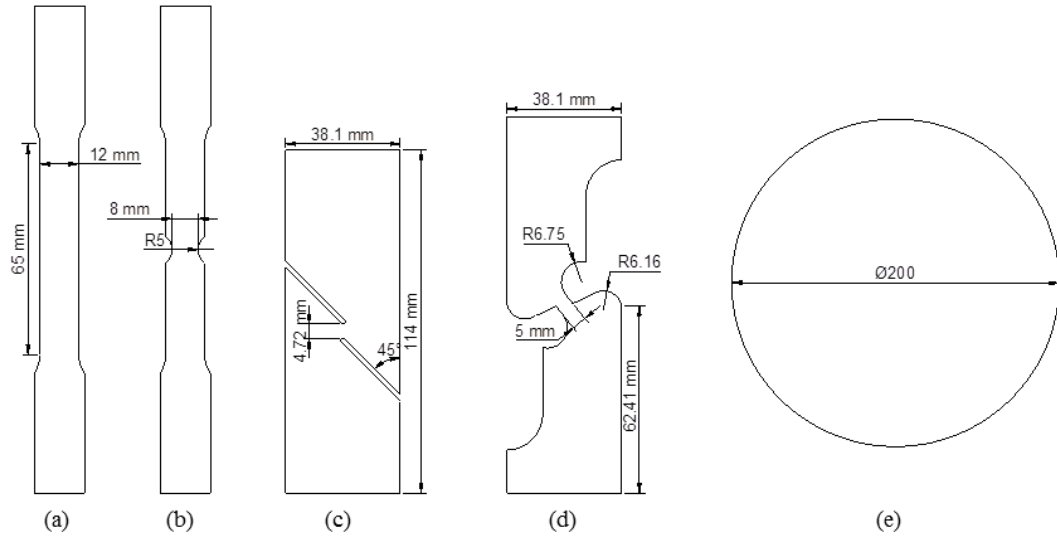


Figure 3.5 Geometries of the specimens: (a) smooth, (b) notched 5 mm, (c) shear b831, (d) special, and (e) Nakajima round.

Table 3.2 Experimental plan.

Specimen geometry	Temperature (°C)	Strain rate (s ⁻¹)	Rolling direction (deg)
Smooth	25, 100, 200, 250, 300		
Notched 5 mm			
Shear b831		0.1	0, 45, 90
Special	25, 100, 300		
Nakajima round			-

3.2.5 Measurement method of anisotropy coefficient

Normal anisotropy coefficient r , which is the ratio between the specimen width and thickness strains. In this study, the evolution of both the width and thickness strains along the cross section in the specimen deformed zone was obtained from the Aramis™ system. The r value was calculated along the longitudinal direction, and a steady state range between 0.15 and 0.2 of true strain was selected for its evaluation.

3.2.6 Measurement method of fracture strain

The fracture strain of smooth, notched, shear specimen measured making use of two different methods depend on the testing temperature and specific condition: (1) Manual measurement. One of the common methods, often referred to as final fracture strain, consists in measuring the specimen cross-section area after fracture and calculating the shrink percentage compared to the initial cross-section area. This method used in section 4.1.3 for measuring the fracture strain of fractured titanium smooth specimen tested at elevated temperature and varying strain rate; in section 5.1.2 for measuring the fracture strain of magnesium smooth specimen. The main disadvantage of this method lies in the fact that fracture surface boundaries are usually difficult to be clearly identified in case of ductile fracture at high temperature because of their sharp slope. Furthermore, for identifying the fracture

strain of the shear specimen in section 4.2.2 the following procedure was identified on the basis of literature [93]: first, three lines were drawn on the deformation zone of the specimen along the horizontal direction, then, after the specimen fracture, a lateral image of the fracture area was acquired using a table microscope, aligning the vertical axis of the specimen with the horizontal direction of the image later on processed in CAD; finally, the angles between the lines that extended from the highlight endpoint and the vertical line were measured and the average angle calculated (Figure 3.6 (b)). (2) The Digital Image Correlation (DIC) approach which developed by ARAMIS™ system was used to directly measure the fracture strain of the smooth, notched, special and shear specimens. The fracture strain was identified as the maximum major strain along the loading direction as soon as the first visible crack appeared on the specimen surface (Figure 3.6 (a)). This method used in section 4.4.5 for calibrating damage model with strain of fracture of smooth, notched and shear specimen and in section 5.2.5 for calibrating the parameters of damage model with smooth, notched, shear, special specimen. In addition, all fracture strain of Nakajima test were measured using DIC method. The details of measurement method can be found in the corresponding section 5.2.1.

$$\gamma = \arctan\left(\frac{b}{a}\right) \quad (3.1)$$

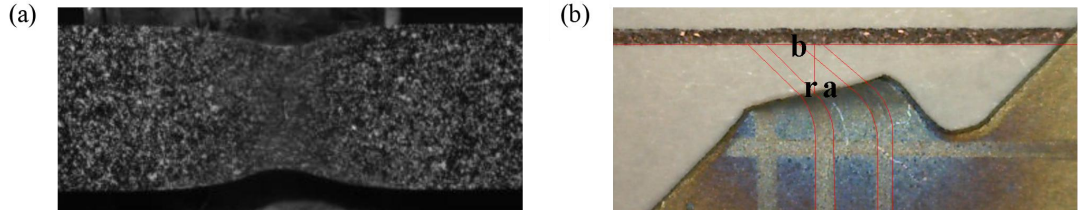


Figure 3.6 (a) Images from the Aramis™ system and (b) fracture strain estimated from table microscope images.

3.2.7 Microstructure analysis

The fractured specimens of titanium alloys were analyzed by a Zeiss™ Stemi C-2000 stereo microscope to acquire images of the specimen side surfaces. Whereupon, the specimen fracture surfaces were inspected using a FEI™ QUANTA 450 Scanning Electron Microscope (SEM) with the Secondary Electron (SE) probe. Images at different magnifications, namely 2000x and 5000x, were acquired across different zones of the fracture surfaces. Moreover, the Electron Backscatter Diffraction (EBSD) was used to analysis the texture of titanium alloys.

The fractured specimen of magnesium alloy where close to the fracture surfaces were cut, polished and then etched (using a solution made of 30 ml acetic acid, 15 ml water, 6 ml picric acid and 100 ml ethanol) for 5 s to analyze the microstructure by a Leica DMRE™ optical microscopy equipped with a high definition digital camera. On the same specimens, micro-hardness measurements were carried out using a Leitz Durimet™ micro-hardness tester on the cross section perpendicular to the loading direction at regular distance intervals, using a 15 g load for 30 s, four

values were recorded for measurement and then the average value calculated. The same SEM machine was used to obtain the fracture surface images.

3.2.8 Measuring the grain dimension

The grain size measured using the line intercept method implemented in MatlabTM. This program allows you to analyze the size of grains in a micrograph with the intersection method arranged in a user-friendly GUI. After loading the micrograph the program puts lines on the surface. Click on the intercept of the lines with the grain boundaries. After calculating, Obtained the length of the grains and frequency of appearance and the histogram including information of Average Grain Size (AGS) and Standard Deviation (SD).

3.3 Numerical modelling

Finite element method (FEM) has been extensively used in metal forming process, as an effective tool to simulating the response of materials and obtain the optimum process parameters. Reproduce the deformation process properly in metal forming is important to reduce manufacturing cost and understand their influence on the deformation behavior. Constitutive models describe the relationship of the dynamic material properties with process parameters and used as input to the finite element (FE) code for simulating the material's response under the specified loading conditions. Flow stress of materials is mostly influenced by hardening and softening behavior which is majorly dependent upon temperature and strain rate. The correctness of simulation results are completely influenced by the accuracy of prediction of the constitutive equation. Ideally, a constitutive equation should contain a reasonable number of material constants, which can be estimated using limited number of experimental data, and at the same time, should be able to predict the flow behavior with acceptable accuracy and reliability over a wide range of temperature and strain rate. In this thesis, on the basis of the large number of experimental results, numerical simulations were implemented in the Ls-DynaTM code environment, using a dynamic-implicit analysis and a coupled thermo-mechanical solution procedure.

Numerical models of the tensile, shear tests and Nakajima tests were developed in order to get the correct values of the stress triaxiality and deviatoric stress parameter for each specimen geometry. The specimens were discretized using shell elements of 1 mm of thickness with 5 through-thickness integration points. After a preliminary sensitivity analysis, the element size was set less than 0.5 mm in the deformation zone to avoid any influence on damage prediction. The dies simulated for the Nakajima set-up were considered rigid. In case of tensile and shear tests models, the specimen testing temperature was simulated implementing a localized heat generation source to the specimen central zone on the basis of recordings of an infrared thermo-camera, whereas, in case of Nakajima tests models, the initial thermal field of the specimen was simulated with the same dimensions of the

inductor heads used for heating up. Graphite foils were used as lubricant, using static and dynamic friction coefficients between the punch and specimen equal to 0.1 and 0.05 at room temperature, respectively, while at 600°C the same values were set equal to 0.2 and 0.1 [98][99]. The actual thermal field of the specimen as a consequence of the electrical heating was implemented (see Figure 3.7) on the basis of the thermo-camera recordings. The center point of the specimen deformation zone was considered to get the values of the stress triaxiality and Lode parameter.

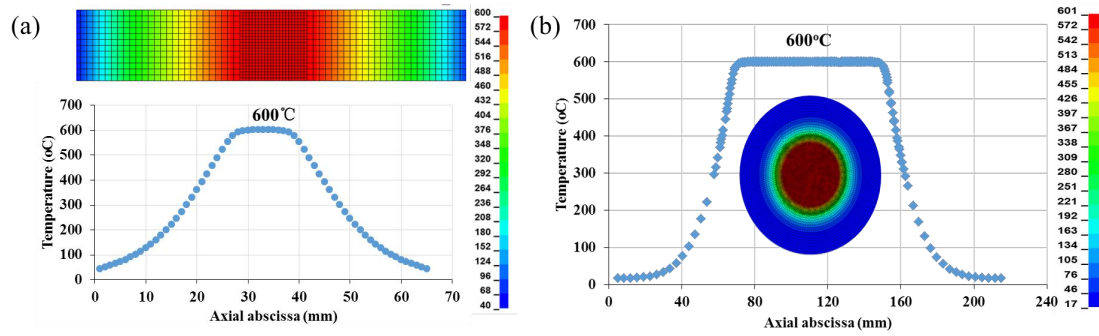


Figure 3.7 Temperature distribution at 600°C implemented in case of tensile test on (a) smooth specimens and (b) Nakajima test.

Chapter 4

4 Results and discussion for titanium alloys

In this Chapter, results and discussion of the mechanical characteristic and numerical modelling involving titanium alloys are presented.

Details on the experimental campaign, materials and reasons that have pushed the study can be found in chapter 3.

4.1 Anisotropy influence on mechanical characteristic of Ti6Al4V sheets deformed in a wide range of temperatures and strain rates

This section present the anisotropy effects on the Ti6Al4V flow behaviour, plastic instability and fracture behaviour in a wide range of thermo-mechanical conditions, typical of industrial sheet forming operations. The uniaxial tensile tests were carried out in a wide range of testing temperatures (from room temperature to 800°C) and strain rates (0.01, 0.1, 1 s⁻¹) to assess the anisotropy effects. Strain hardening, strain rate sensitivity, Lankford coefficients and fracture strain were evaluated as a function of the testing parameters and specimen orientation.

4.1.1 Flow behaviour and plastic anisotropy

Ti6Al4V exhibited a non-identical response in terms of flow behaviour and strain at fracture for different specimen orientations.

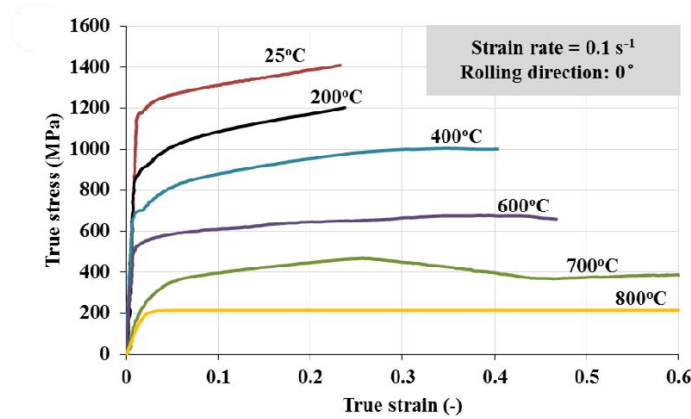


Figure 4.1 Flow behaviour at varying temperature at rolling direction and strain rate of 0.1 s⁻¹.

Figure 4.1 shows the high flow stress sensitivity to temperature at strain rate of 0.1 s⁻¹ for specimens cut at 0° with respect to the rolling direction. The Ultimate Tensile Strength (UTS) that is the ratio between maximum force and the initial cross-section, according to

$$\sigma_{UTS} = \frac{F_{\max}}{A_0} \quad (4.1)$$

The $F_{\max}$ the maximum force, A_0 the initial cross-section area of the undeformed specimen. Figure 4.2 (a) shows the influence of the temperature and specimen orientation on the UTS values for a strain rate equal to 0.01 s^{-1} : as expected, UTS decreased at increasing temperature regardless the specimen orientation. On the other hand, specimens oriented at 45° with respect to the rolling direction showed the lowest strength regardless the testing temperature, whereas specimens oriented at 0° and 90° were characterized by comparable UTS values. The same influence of the specimen orientation regardless the applied strain rate as shown in Figure 4.2(c) (e).

The influence of the temperature and specimen orientation on the strain hardening exponent is shown in Figure 4.2 (b) for the strain rate of 0.01 s^{-1} . As expected, the strain hardening decreases at elevated temperatures, reaching values close to zero at the highest one. Again, a strong influence of the specimen orientation was evident. The same influence of the specimen orientation regardless the applied strain rate as shown in Figure 4.2(d) (f).

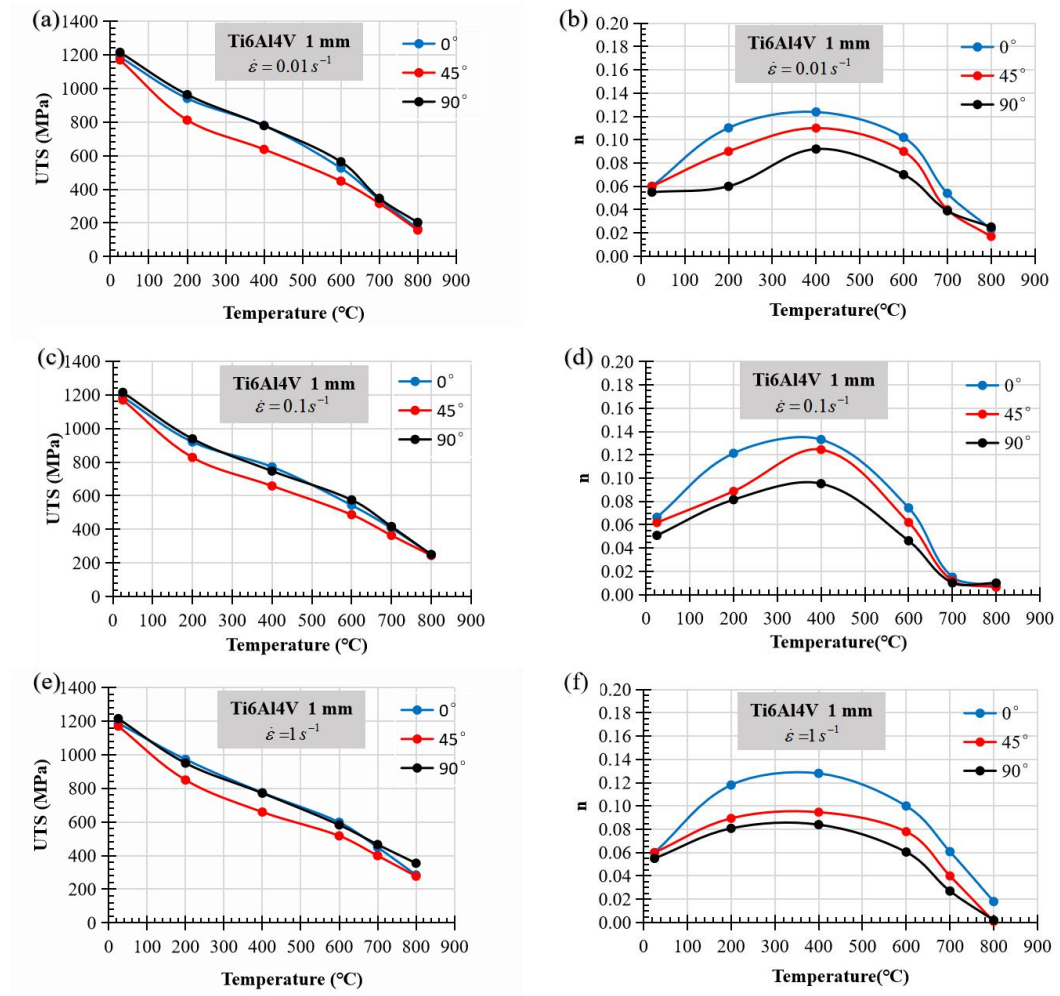


Figure 4.2 UTS and strain hardening exponent as a function of the testing temperature and specimen orientation at strain rate 0.01, 0.1, 1 s^{-1}

The anisotropy coefficient r is shown in Figure 4.3 (a) as a function of the testing temperature and specimen orientation for a strain rate equal to 0.1 s^{-1} . The effect of temperature is noticeable: a decrease of the r values is evident for the specimens oriented at 0° and 90° at increasing temperature. The 0° oriented specimens have always values lower than 1, whereas the 90° oriented specimens tend to be isotropic in the temperature range between room temperature and 400°C . On the other hand, the trend of the 45° oriented specimens is pretty different, with an initial increase of the plastic anisotropy in the low temperature range and subsequent decrease at higher temperatures. Again, the same tendency regardless the applied strain rate as shown in Figure 4.3 (c) (d).

To account for this still anisotropic behaviour at elevated temperature, EBSD analysis was carried out for the specimens oriented at 0° and strained at room temperature, 600°C , and 800°C . After standard metallographic preparation, the specimens were finished with a vibratory polishing using an acid solution containing $0.02 \mu\text{m}$ colloidal silica. Figure 4.3 (b) shows the texture plots for the RD-TD plane at room temperature, 600°C and 800°C for the specimen oriented at 0° , strain rate of 0.1 s^{-1} . It is evident a strong $(0001)_\alpha$ accumulation along the axis in the TD direction and weak $(0001)_\alpha$ accumulation along the axis in the RD direction, indicating strong T-texture and weak basal-texture (B texture) components, more pronounced at room temperature, but still present at elevated temperature. The aforementioned behaviour was observed for all the other tested strain rates.

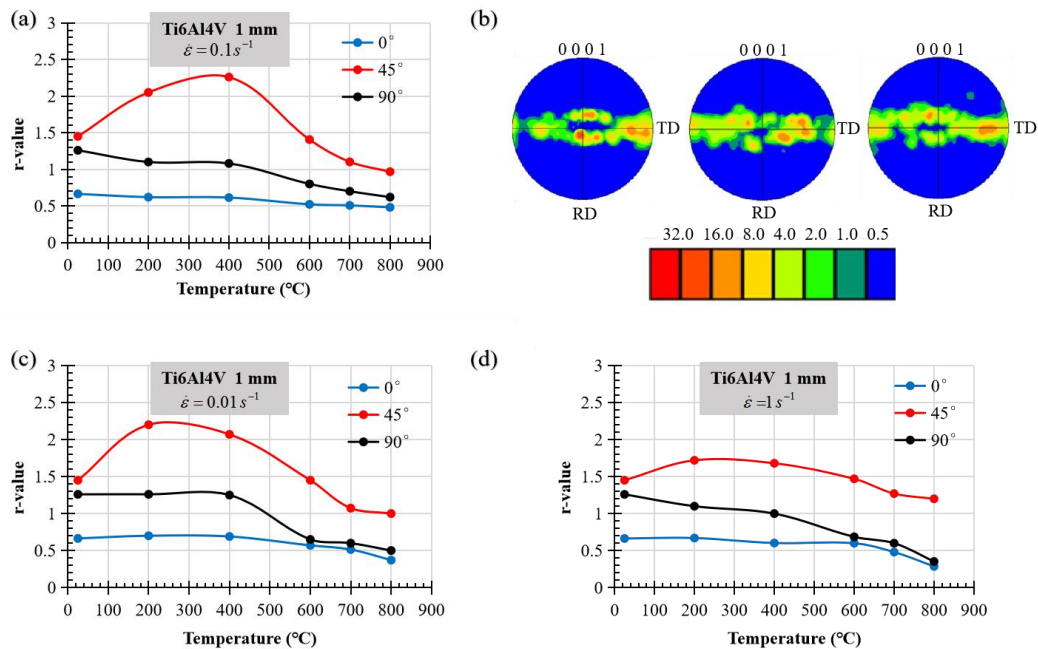


Figure 4.3 Plastic normal anisotropy as a function of the testing temperature and specimen orientation ($\pm 2\%$ uncertainty) at strain rate (a) 0.1 s^{-1} (c) 0.01 s^{-1} (d) 1 s^{-1} , (b) pole figures at room temperature, 600°C and 800°C for the specimen oriented at 0° .

4.1.2 Plastic instability

The strain rate sensitivity exponent m , indicative of the occurrence of either localized or diffuse necking, was calculated using an exponential law to model the

flow stress as a function of the strain rate. Figure 4.4 (a) shows the m values as a function of the temperature and specimen orientation at a strain rate equal to 0.1 s^{-1} . The material always exhibited positive m values for all the testing temperatures. Nevertheless, in the range of temperatures between 0° and 400°C , the values were close to zero regardless of the specimen orientation, meaning that the plastic instability manifested itself as localized necking. Starting from 600°C , the strain rate sensitivity became higher, with an increase at increasing temperature. Furthermore, it can be noted that the m values for the specimens oriented at 45° were always higher than the ones of the specimens oriented at 0° and 90° . The drastic increase of the strain rate sensitivity in the range of elevated temperatures was indicative of the occurrence of plastic instability in form of diffuse necking, as witnessed in Figure 4.4 (b) by the AramisTM images of the specimens strained till fracture. Specimens oriented at 45° showed a larger diffuse necking in the temperature range between 600° and 800°C , consistent with their higher ductility at increasing temperature, compared to the specimens oriented at 0° and 90° . It is worth to note that at 600°C , the 90° specimen showed the lowest m value, and from the AramisTM images the necking was still localized. Same results were obtained for the other tested strain rates.

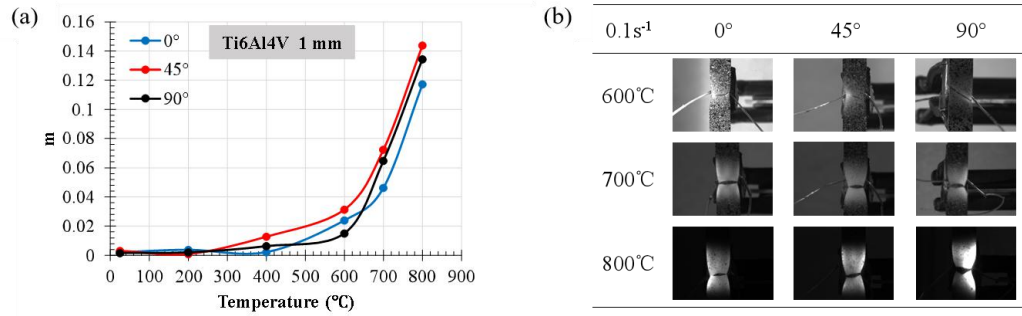


Figure 4.4 (a) Strain rate sensitivity as a function of the testing temperature and specimen orientation; (b) AramisTM images of the specimens strained at elevated temperatures and strain rate of 0.1 s^{-1} .

4.1.3 Fracture strain

The true strain at fracture at varying temperature and strain rate was calculated by measuring the fracture surface according to:

$$\varepsilon_f = \ln \left(\frac{A_0}{A_f} \right) \quad (4.1)$$

where A_0 is the specimen initial cross-section area and A_f the fractured one. Following the procedure reported in [94], the specimen area at fracture was measured through Scanning Electron Microscopy. The image of fractured cross-section shown in Figure 4.5 (a) as an example. Figure 4.5 (b-d) shows the strain at fracture as a function of the testing temperature and specimen rolling direction at 0.01 , 0.1 and 1 s^{-1} , respectively: in general, a significant increase of ductility is seen starting from 600°C , identifying this temperature as the one at which the benefit of elevated temperature forming in terms of enhanced formability can be exploited.

Furthermore, there is a remarkable sensitivity to the rolling direction, as specimens cut at 45° with respect to the rolling direction always show the highest ductility, regardless of the testing temperature and strain rate. This behaviour can be ascribed to the increase of beta phase volume fraction at increasing temperature, which becomes significant at temperature above 600°C [76]. In fact, there is a gradual change from the Hexagonal Closed Packed (HCP) crystal structure, typical of the alpha phase and characterized by a limited number of slip planes as well as high anisotropy, to the Body Centred Cubic (BCC) structure, typical of the beta phase and characterized by a much higher number of slip planes and reduced anisotropy. This change in the crystal structure is responsible of the higher ductility. It is worth noticing that the highest values of Lankford coefficients pertain to specimens cut at 45° with respect to the rolling direction (Figure 4.3), which show to provide the highest ductility, regardless of the testing temperature. Similar results were obtained for all strain rates.

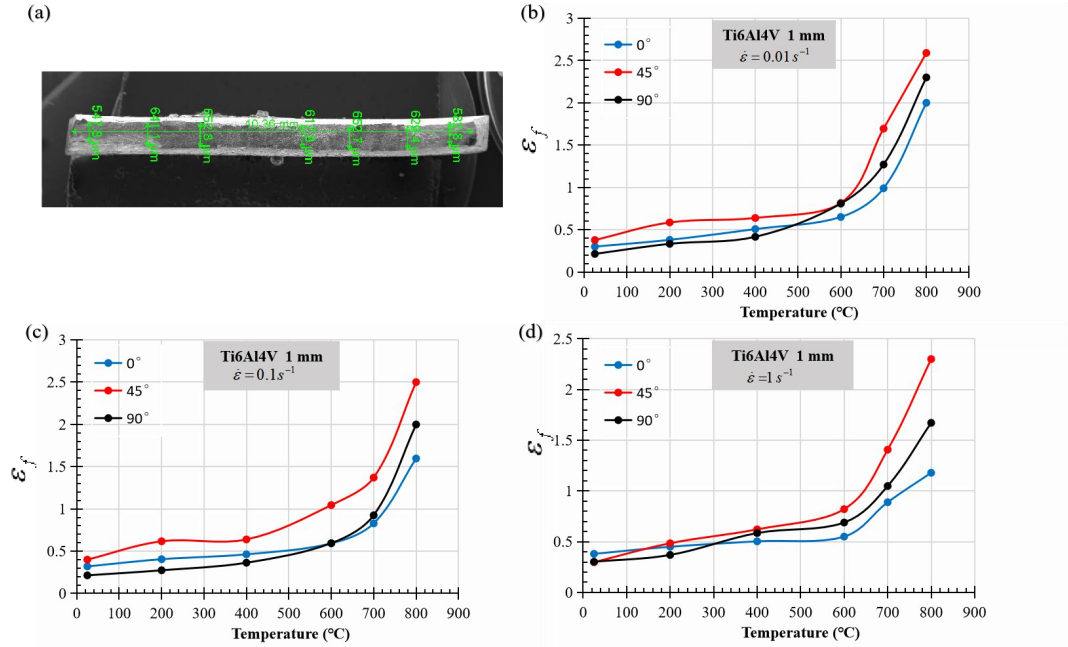


Figure 4.5 (a) SEM image for cross-section of fractured specimen and fracture strain at varying temperature and specimen rolling direction at strain rate of (b) 0.01 s^{-1} (c) 0.1 s^{-1} (d) 1 s^{-1}

On the basis of these results, the reference testing temperature was assumed to be 600°C , since it is the temperature at which the material ductility starts increasing, while the reference strain rate was set equal to 0.1 s^{-1} . At the reference conditions, the strain at fracture sensitivity to the specimen geometry will be discussed in next section, which corresponds to the different stress states.

4.2 Anisotropy influence on the failure of Ti6Al4V sheets under varying stress state

This section presents the anisotropy influence on the failure characteristics of Ti6Al4V titanium alloy sheets. The fracture characteristics are discussed making use of tensile tests on smooth, notched and shear specimens in order to have various stress

states. The reference condition, namely temperature of room and 600°C, and strain rate of 0.1 s^{-1} were used. Finally, the fracture strain is measured and the effect of the specimen orientation and stress state is identified. To determine the actual stress state for each specimen geometry, a numerical model was set up and calibrated using elasto-plastic data from uni-axial tensile tests on smooth specimens in section 4.4.5.

4.2.1 Plastic behaviour under varying stress state

The influence of specimen orientation was first evaluated at diffuse necking to identify the so-called critical strain, namely the value of strain at Ultimate Tensile Stress (UTS) recorded from the engineering stress-strain curve. It is worth to note that, regardless the testing temperature and specimen orientation, the strain at localized necking (i.e. the value of strain at which the true stress-true strain curve drops) is very close to that at diffuse necking. Therefore, here the general definition of instability strain is given. Figure 4.6 shows that the instability strain vs. stress triaxiality as a function of the specimen orientation at room temperature and 600°C: the shown data are relevant just to smooth and notched specimens, as shear specimens do not present necking before fracture. Both at room temperature and 600°C, the instability strain increases significantly for the smooth specimens, whereas the values for the notched values are pretty similar. Furthermore, the influence of the specimen orientation is evident just for the smooth specimens, but with a different sensitivity for the two testing temperature: at room temperature the specimen oriented at 0° behaves the best, whereas at 600°C, the specimen oriented at 45° has the highest instability strain.

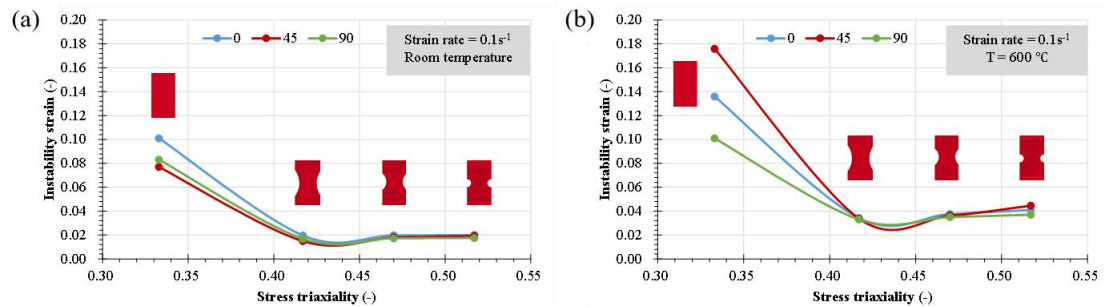


Figure 4.6 Instability strain as a function of stress triaxiality and specimen orientation at (a) room temperature, and (b) 600°C.

4.2.2 Fracture strain under varying stress state

The values of the fracture strain for the smooth and notched specimens were identified on the basis of the images acquired by the AramisTM system: the image at which the fracture strain was taken was the one where the first crack in the deformation zone was evident, before the final rupture of the specimen. Figure 4.7 shows the fracture strain as a function of the stress triaxiality and specimen orientation at room temperature and 600°C: as expected, the fracture strain

increases at decreasing stress triaxiality. As regards the anisotropy effect, the smooth and notched specimens oriented at 45° always present higher fracture strain values regardless the testing temperature, even if the sensitivity at room temperature is lower. On the contrary, the opposite trend characterizes the shear specimens, since the 90° oriented ones show the highest fracture strain and the 45° oriented the lowest, both at room temperature and 600°C . Figure 4.8 shows the fracture strain as a function of the Lode parameter and specimen orientation at room temperature and 600°C . The same trend seen as a function of the stress triaxiality is evident for both the temperatures.

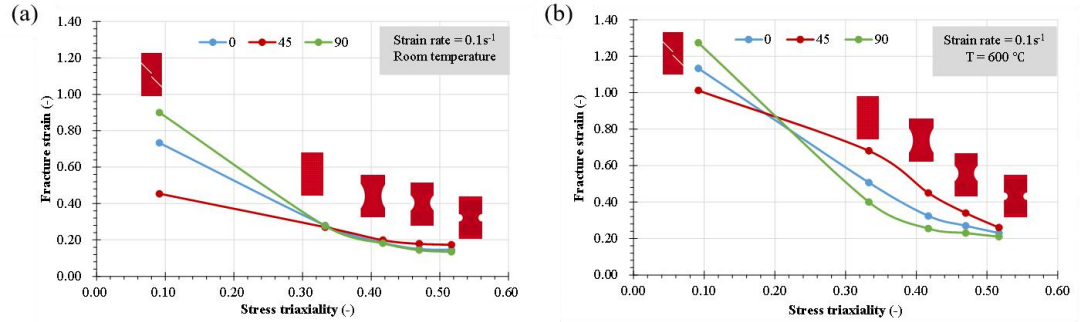


Figure 4.7 Fracture strain as a function of stress triaxiality and specimen orientation at (a) room temperature, and (b) 600°C .

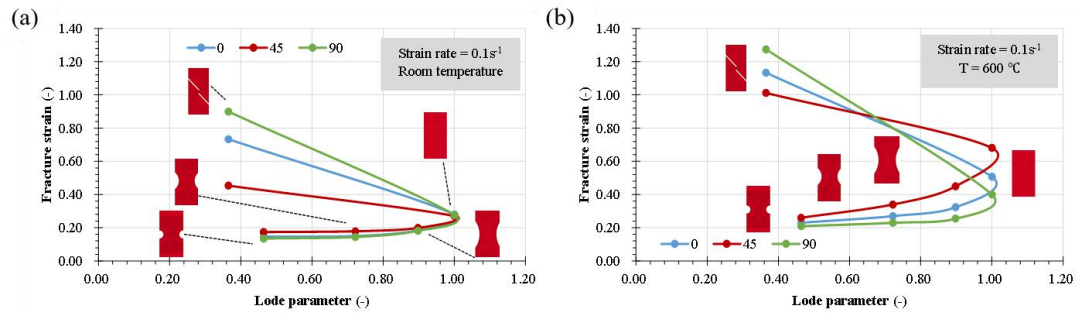


Figure 4.8 Fracture strain as a function of Lode parameter and specimen orientation at (a) room temperature, and (b) 600°C

4.3 Microstructure analysis

4.3.1 Fracture surface characterization

The SEM analyze have been carried out on the fractured smooth specimen at temperature from room temperature to 800°C , and the images from different view side were reported in Figure 4.9. The slant fracture surface observed view from top to bottom when the testing temperature between 25°C to 400°C . While as the temperature rise to 800°C , the slant fracture disappeared and flat fracture surface observed, that was due to at relatively high temperature the higher diffuse necking occurrence before specimen fracture. The same feature can be found on the thickness direction that a quick thickness decrease after 600°C . Meanwhile, the number of dimple enhances with rise in temperature from 25°C to 400°C . when the temperature increase to 600°C , the dimple size of Ti6Al4V titanium alloy are deeper than at lower temperature. From the microstructure of fracture surface, we

can see that the higher ductility as increasing temperature due to nucleation, growth, and coalescence of microvoids.

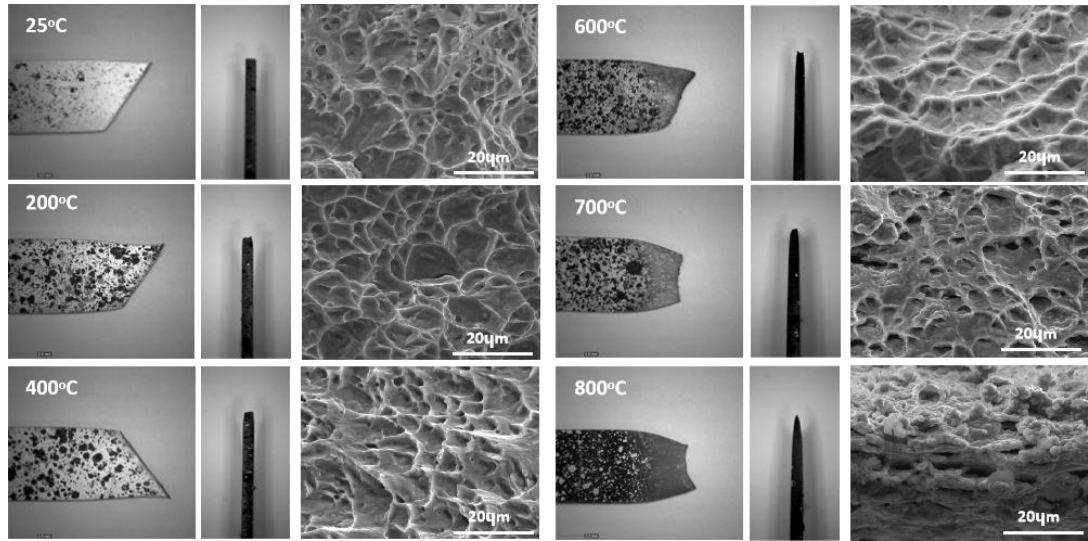


Figure 4.9 SEM images of lateral surface and fracture surface of the smooth specimen after a tensile test at temperature from 25°C to 800°C, strain rate 0.1 s^{-1}

In terms of different specimen geometry. The SEM analysis were showed that the smooth and notched specimens have fracture surfaces characterized by deeper dimples and a higher density of secondary voids, proving that the higher the stress triaxiality the lower ductile the fracture and therefore the lower the fracture strain. This is observed regardless the testing temperature and specimen orientation. On the other hand, as expected, at 600°C, the dimples size becomes bigger, with an increased density of secondary voids, regardless the stress state and specimen orientation. It is evident from Figure 4.10 that the dimples in the specimen with the 45° orientation are bigger and deeper compared to the ones in the specimens with the 0° and 90° orientations, which is consistent with the related recorded values of the fracture strain. Again, this is evident regardless the testing temperature and stress state. The fracture morphology of the shear specimens is drastically different, with features typical of the shear stress state. Figure 4.11 shows that the specimen with the 45° orientation has an evident less density of voids, in accordance with the its lower fracture strain compared to those of the specimens orientated at 0° and 90°.

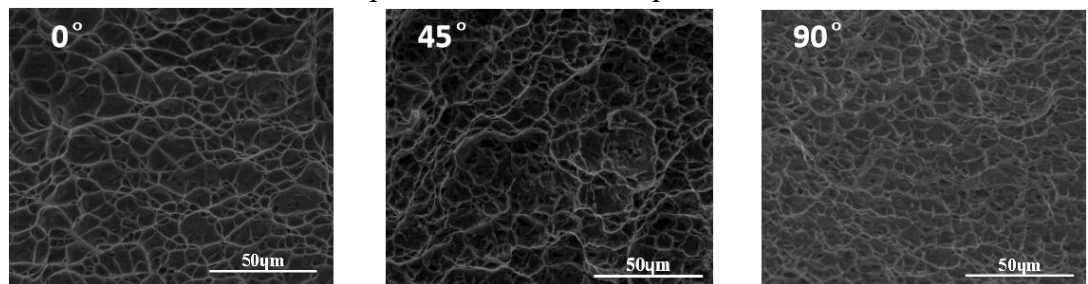


Figure 4.10 SEM images of the fracture surface of 5 mm notched specimens oriented at 0°, 45°, 90° at 600°C.

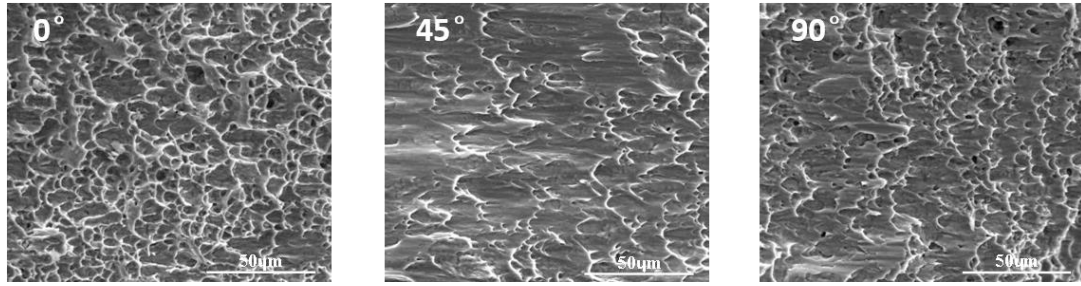


Figure 4.11 SEM images of the fracture surface of shear specimens oriented at 0°, 45°, 90° at 600°C.

4.3.2 EBSD analysis

Electron back-scattered diffraction (EBSD) is the most widely used method to study the grain structure and the texture evolution. The EBSD technique was used to study the texture evolution of titanium alloys based on the fractured smooth specimen at room temperature and at 0°, 45° and 90° respect to the rolling direction, respectively. Figure 4.12 shows the experimental textures recorded on basal (0001), prismatic (1010) and pyramidal (1120) crystallographic planes of the Ti6Al4V smooth specimen cut in the 0°, 45°, 90° direction respect to the rolling direction. The texture intensities are represented by the colour scale bar corresponding to the representative pole figure. The (0001) and (1010) PFs at tensile-fractured specimen along RD from EBSD analysis. We can see that the projection points of (0001) orientations gather into a narrow band locating in the center and perpendicular to the RD. This result implies that basal slip which contributes to the out-plane rotation of basal plan is one of the dominate plastic mechanism in the tensile deformation. After that, the basal plane already parallels to the RD. The evolution of (1010) PF gives the information of lattice rotation within the basal plan. the projections points of (1010) gather to the north and south poles and two narrow bands around the two poles. This result implies that except the out-plane rotation, the hcp crystals also rotate within the basal plan and stabilize when two (opposite with each other) of the six [1010] orientations parallel to the loading direction. This is a typical rolling texture. The (0001) and (1010) PFs at tensile-fractured specimen along 45 rolling direction from EBSD analysis. It present similar texture distribution with the PF on rolling direction. But it show more obvious preferred orientation on (0001) PF (see the maximum value in the colour scale bars). The T texture exhibit 45 tilt to loading direction demonstrate that the slip system cannot activated due to deformed was not enough. This is a possible reason that a weak ductile on 45 rolling direction specimen. The (0001) and (1010) PFs at tensile-fractured specimen along 90 rolling direction from EBSD analysis. A ring texture forms in the (1010) pole figure in which the intensity of the prismatic pole densities are distributed 360° non-uniformly at the periphery of the pole figure.

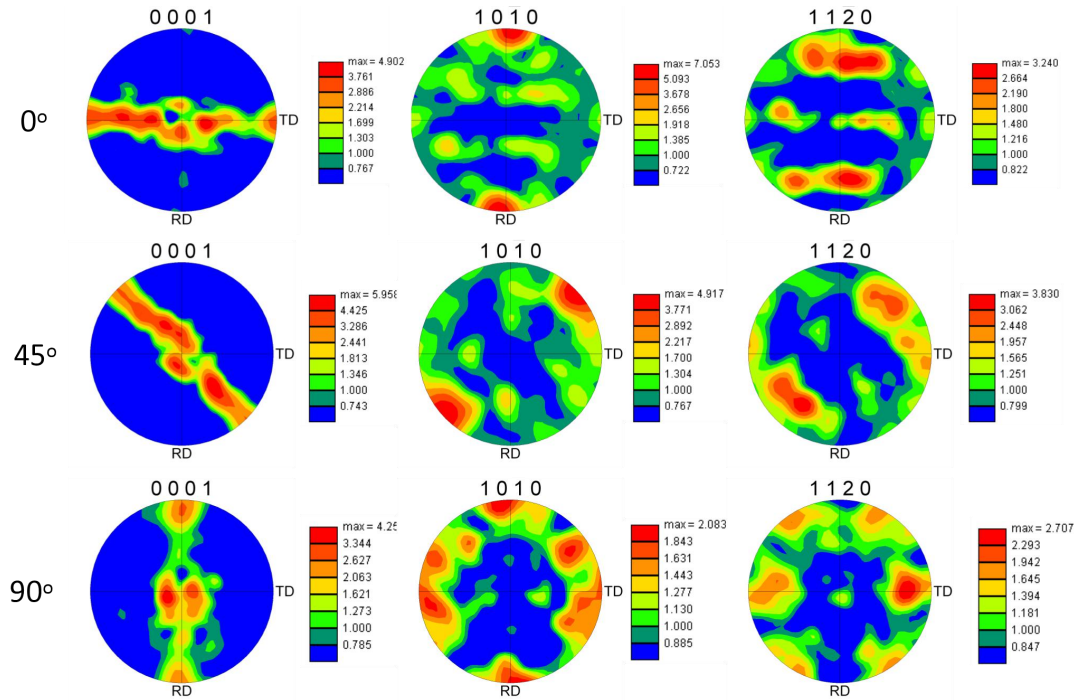


Figure 4.12 The pole figures of the Ti6Al4V specimen along 0°, 45°, 90° direction respect to the rolling direction via EBSD.

4.4 Modelling of the anisotropic plastic behavior and fracture occurrence in Ti6Al4V alloy sheets

This section describes the research work related to the modelling of the material behaviour, whose importance has increased with the interest on the use of FE models. The coupling of the Barlat-Lian 1989 anisotropic yield criterion and GISSMO damage model was introduced to predict the fracture occurrence and capture its anisotropic characteristics. The two models were calibrated on the basis of an extensive experimental campaign. In order to validate the proposed approach, tests not used in the calibration phase were used, and the comparison between experimental and numerical results was carried out in terms of fracture characteristics at varying temperature and rolling direction.

4.4.1 Modelling of the anisotropic plastic behaviour

The Ti6Al4V behavior was modelled with the following characteristics:

- Temperature-dependent values of the Young's, Poisson's moduli, heat capacity and thermal conductivity were implemented from literature [95], whereas the density was kept constant (see Table 4.1).
- Temperature and strain rate-dependent anisotropic yield locus on the basis of the Barlat-Lian-1989 yield criterion, implemented by using the Lankford coefficients at varying temperature and strain rate;

- Crystallographic constant m of the Barlat-Lian-1989 yield criterion set equal to 12, which is stated in literature [50] to be the most accurate value for HCP crystal lattice at room temperature. Being all the tests carried out below β -transus temperature in a range of temperatures where there is still little transformation, the HCP crystal lattice can be assumed to be the dominating structure for all the testing conditions;
- The elasto-plastic behaviour of Ti6Al4V sheets was calibrated on the basis of tensile tests carried out on smooth specimens at varying temperature and strain rate in section 4.1. Temperature and strain rate-dependent hardening rule, implemented by using the experimental flow curves derived from specimens oriented at 0° ;
- Orientation-dependent strain rate sensitivity, implemented changing the hardening rule as a function of the specimen orientation; this feature is essential to capture the behavior highlighted in Figure 4.4.
- Temperature and anisotropy-dependent GISSMO damage model, introduced to predict the fracture occurrence and capture its anisotropic characteristics.

Table 4.1 Ti6Al4V mechanical and thermo-physical properties implemented at room temperature and 600°C.

	Density (g/cm ³)	Young's modulus (GPa)	Poisson's ratio (-)	Heat capacity (J/kgK)	Thermal conductivity (W/mK)
T = 25°C	4.54	105	0.23	580	6.65
T = 600°C		74.2	0.24	725	14.7

The whole mentioned procedure of modelling the plastic and failure behaviour of Ti6Al4V was summarized as a flow chart, as shown in Figure 4.13.

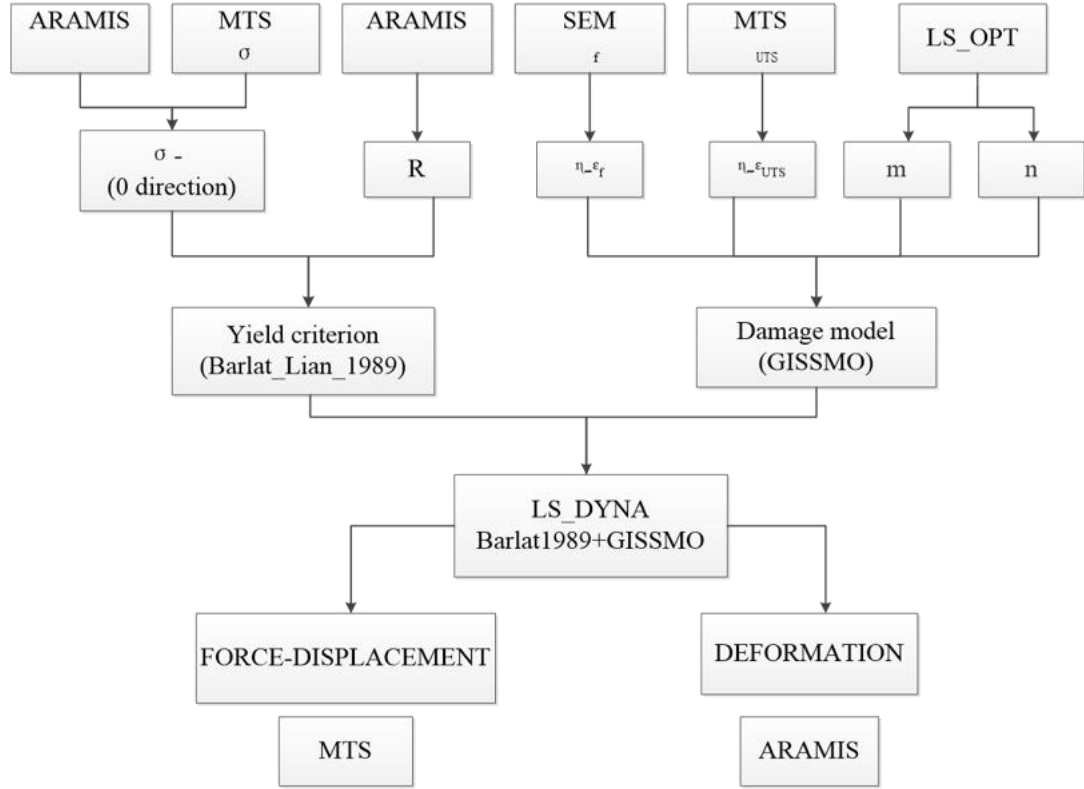


Figure 4.13 The flow chart of modelling the anisotropic plastic and fracture behavior of Ti6Al4V titanium alloy.

4.4.2 Barlat-Lian1989 yield and hardening rule

The anisotropic yield locus at room temperature and 600°C was modelled using the Barlat-Lian 1989 yield criterion [96], developed for materials exhibiting planar anisotropy, as is the case of Ti6Al4V, based on the following formulation:

$$f = a|k_1 + k_2|^M + a|k_1 - k_2|^M + c|2k_2|^M = 2\sigma_e^M \quad (4.2)$$

where k_1 and k_2 are invariants of the stress tensor, M is an integer exponent, which is related to the crystallographic structure of the material, and σ_e is the effective stress.

The constants a , c and h are calculated on the basis of the following equations:

$$k_1 = \frac{\sigma_{11} + h\sigma_{22}}{2}, \quad k_2 = \left[\left(\frac{\sigma_{11} - h\sigma_{22}}{2} \right) + p^2 \sigma_{12}^2 \right]^{1/2} \quad (4.3)$$

$$a = 2 - c = 2 - 2 \sqrt{\frac{r_0}{1+r_0} \cdot \frac{r_{90}}{1+r_{90}}} \quad (4.4)$$

$$h = \sqrt{\frac{r_0}{1+r_0} \cdot \frac{1+r_{90}}{r_{90}}} \quad (4.5)$$

where r_0 , r_{45} , r_{90} are the three Lankford coefficients.

Compared to other yield criteria usually implemented for the numerical simulation of sheet forming processes, the calibration of this yield criterion requires a low number of material constants, namely the three Lankford coefficients r_0 , r_{45} , r_{90} and the constant M that depends on the alloy crystallographic structure. The former

were implemented on the basis of the tensile tests on smooth specimens reported in section 4.1, whereas the latter was chosen equal to 12 as reported in literature [50] for HCP metals at room temperature. It is worth underlining that the same value of M was used at 600°C since the HCP crystal lattice is still the dominating structure at this temperature, as the phase transformation into a BCC structure is almost negligible at 600°C [76].

The associated flow rule was implemented by using the experimental flow curves derived from specimens oriented at 0° at varying temperature and strain rate to account for strain hardening, temperature and strain rate sensitivity.

4.4.3 Prediction of diffuse necking and localization necking of tensile test

The experimental and numerical results were compared in order to assess the capability of the model to predict the plastic straining regime of the specimen. In particular, the major strain and thickness distribution along the specimen axis were assessed, making use of the AramisTM data, consisting of local strain measurements during the whole test duration. The baseline strain rate of 0.1 s⁻¹ was used for the comparison and two temperatures were chosen for the validation, namely 200°C and 700°C, representative of the occurrence of two distinct plastic instability phenomena, localized necking for the former and diffuse necking for the latter. Furthermore, the choice of these two temperatures allowed evaluating the model predictability in cold conditions, where the material strain hardening was predominant, as well as in hot conditions, where the strain rate sensitivity played the major role. In all the conditions the benchmark was carried out after the plastic instability onset, at approximately the same amount of equivalent strain, namely 0.2. Figure 4.14 illustrates the comparison at 200°C. For the 0° and 45° directions, the model was capable of predicting both the major strain field and thickness distribution along the specimen axis. In particular, the thickness distribution prediction was accurate both in the necking zone and outside it, proving that the temperature softening and the strain hardening parameters of the model were properly set. For the 90° direction, the major strain distribution shape was less accurate, although the maximum value was correct as well as the thickness distribution.

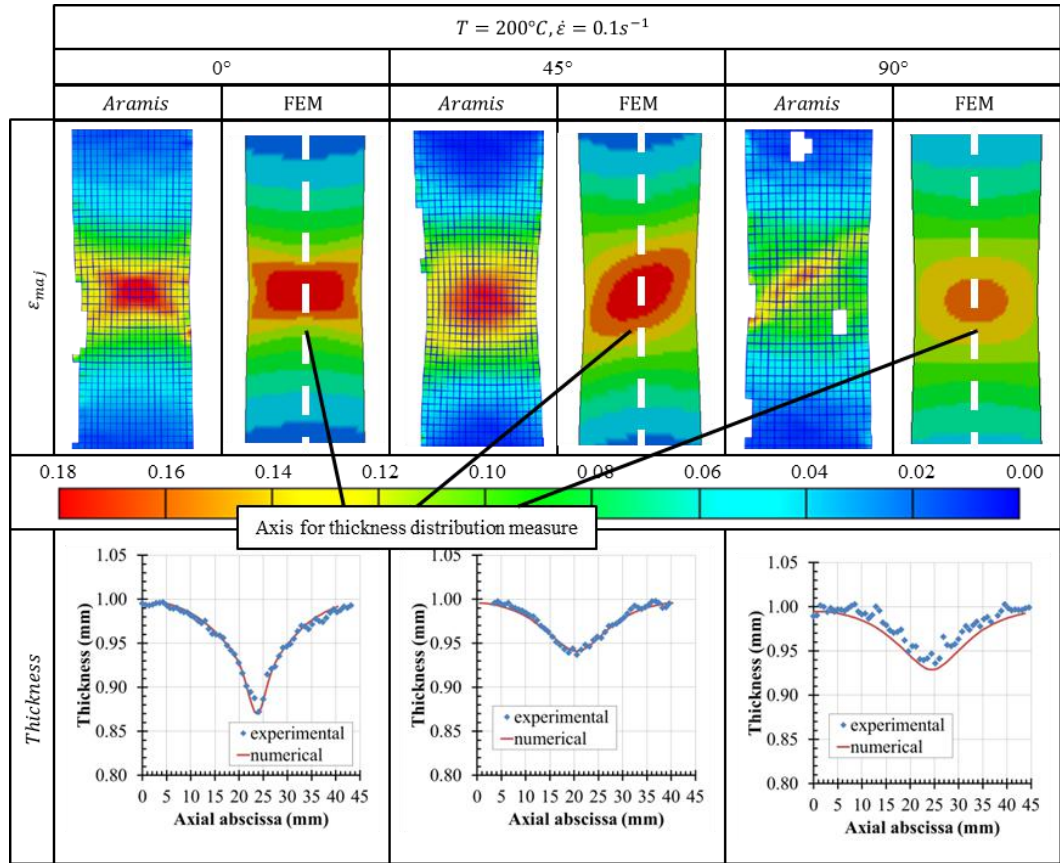


Figure 4.14 Experimental and numerical benchmark at 200°C , strain rate 0.1 s^{-1} .

Figure 4.15 shows the same results for 700°C . The prediction accuracy was also high, particularly the strain distribution along the axis, which showed the typical characteristics of the hot deformation regimes, namely an almost negligible uniform deformation (blue zones at the specimen extremities in both experimental and numerical fringe maps) and a larger post-necking strain diffusion. Figure 4.19 and Figure 4.16 shows at 650°C which were not tested before in order to validate reliability of simulation. The prediction of thickness distribution still show a high accuracy with experiment results.

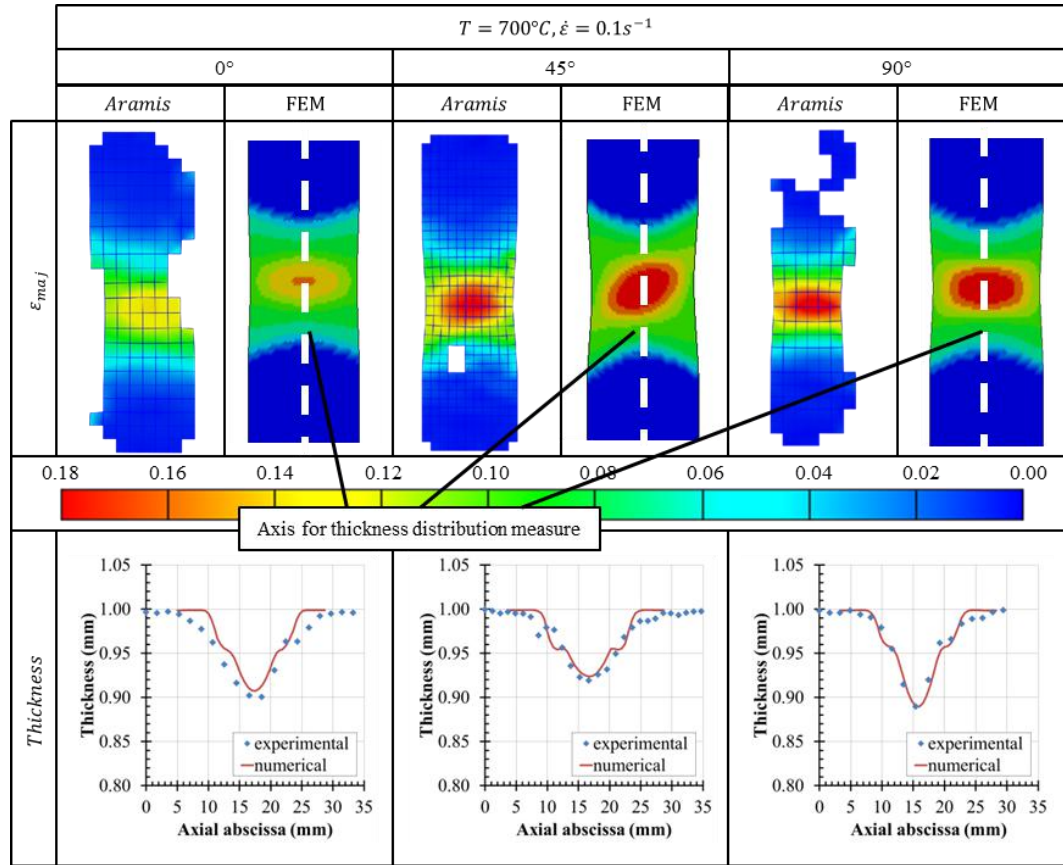


Figure 4.15 Experimental and numerical benchmark at 700°C , strain rate 0.1 s^{-1} .

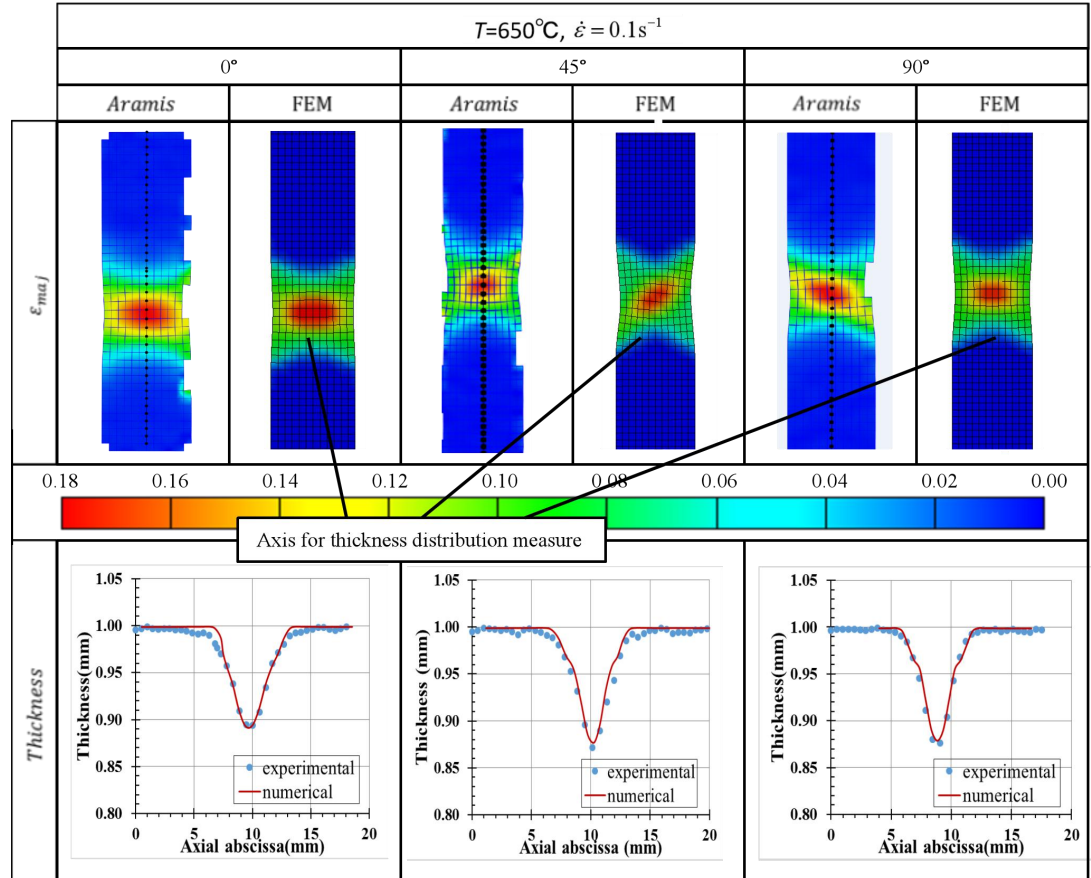


Figure 4.16 Experimental and numerical benchmark at 650°C , strain rate 0.1 s^{-1} .

4.4.4 GISSMO model

The Generalized Incremental Stress State dependent damage MOdel (GISSMO model hereafter), developed by Neukamm et al [97], predicts damage evolution and fracture occurrence in ductile metals using a phenomenological approach. The model combines an incremental damage accumulation formulation, based on a failure curve function of the actual stress state, with an incremental instability accumulation formulation, based on a critical strain curve, again function of the actual stress state. The accumulated damage D and the so-called instability measure F are computed according to Equations (4.6) and (4.7), respectively.

$$\Delta D = \frac{n}{\varepsilon_f(\eta)} D^{\left(1-\frac{1}{n}\right)} \Delta \varepsilon_v \quad (4.6)$$

$$\Delta F = \frac{n}{\varepsilon_{inst}(\eta)} D^{\left(1-\frac{1}{n}\right)} \Delta \varepsilon_v \quad (4.7)$$

Here, $\Delta \varepsilon_v$ is the actual equivalent plastic strain increment, n is a damage exponent that accounts for a non-linear relationship between strain and damage, which is reasonable since damage is expected to accelerate when approaching failure, $\varepsilon_f(\eta)$ and $\varepsilon_{inst}(\eta)$ are the fracture strain and instability strain as function of the stress triaxiality, respectively. Failure occurs when the damage D reaches unity. On the other hand, when the instability measure is equal to 1, which means strain localization occurs, with the accumulated damage that reaches a critical value D_{crit} , the accumulated damage affects the stress σ according to Equation (4.8).

$$\sigma = \tilde{\sigma} \left[1 - \left(\frac{D - D_{crit}}{1 - D_{crit}} \right)^m \right] \quad (4.8)$$

where m is an exponent for damage related stress fadeout.

4.4.5 Numerical modelling of the tests

Using the finite element method described in section 3.3. The numerical results of stress triaxiality and lode parameter can be determined. Figure 4.17. Figure 4.18 proves that there is neither temperature nor anisotropy effect on the stress triaxiality and Lode parameter before necking occurrence (when the curves start decreasing). This means that the specimen geometry plays the major role in determining the stress state. Therefore, the average values of the stress triaxiality and Lode parameter were taken before necking, excluding the first portion of the curves. The evolution of stress triaxiality and lode parameter as a function of the plastic strain is shown in Figure 4.19. The same values were considered regardless the testing temperature and specimen orientation.

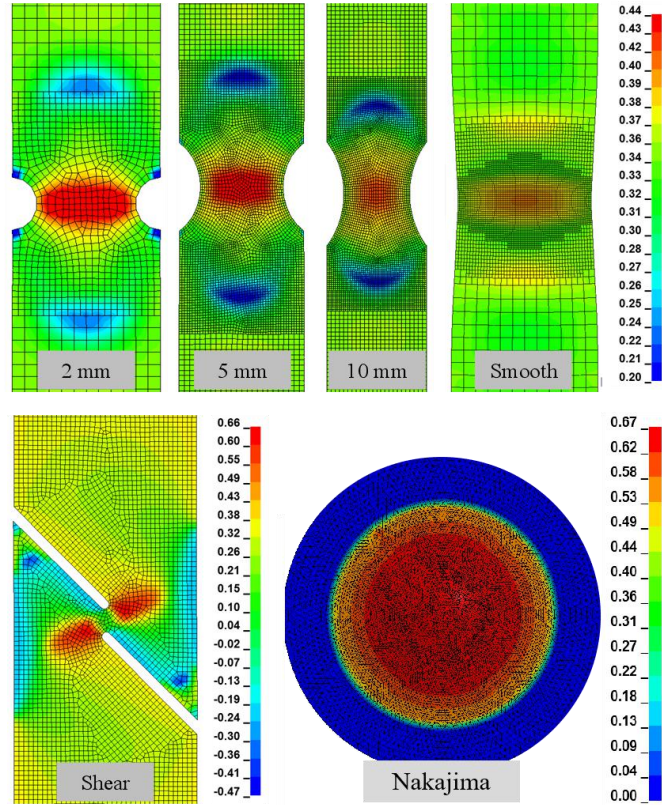


Figure 4.17 Stress triaxiality field for all the specimen geometries.

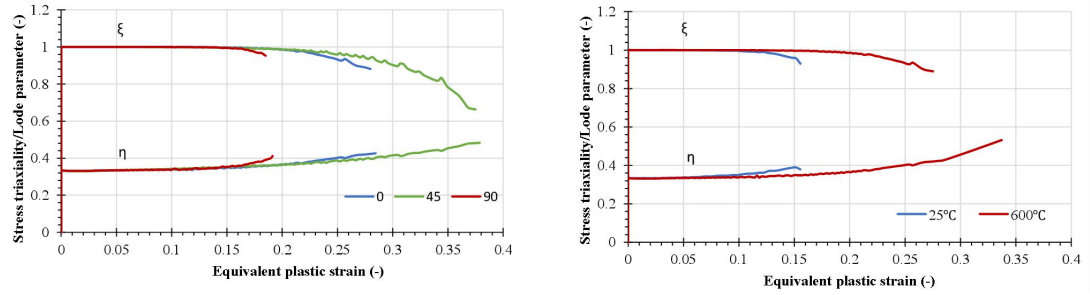


Figure 4.18 (a) Stress triaxiality and Lode parameter of smooth specimen as a function of (a) specimen orientation and (b) testing temperature.

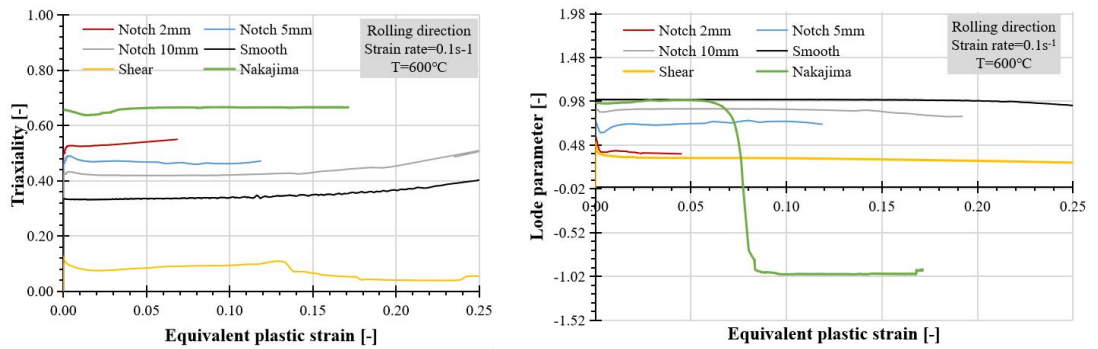


Figure 4.19 (a) Stress triaxiality and (b) Lode parameter as a function of the equivalent plastic strain at 600°C.

In general, the relationship between stress triaxiality and lode parameter are following the equation (4.9). The comparison between stress triaxiality and lode

parameter simulated by FEM (blue point) and the one calculated by formulation (orange line) was presented in Figure 4.20

$$\xi = -\frac{27}{2}\eta\left(\eta^2 - \frac{1}{3}\right) \quad (4.9)$$

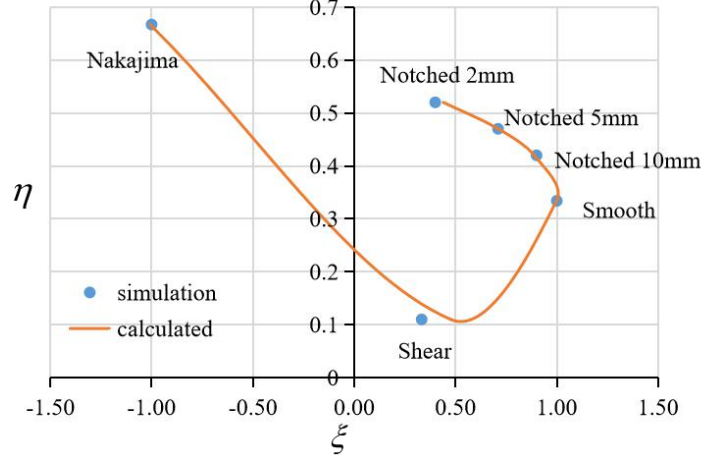


Figure 4.20 The relationship between stress triaxiality and lode parameter according to numerical simulation and calculated results.

For the calibration of the damage model reported in section 4.4.4, the values of the instability strain were also evaluated, namely the values of strain at Ultimate Tensile Stress (UTS) recorded from the experimental engineering stress-strain curves.

Table 4.2 and Table 4.3 reports the values of strain at fracture and instability strain at varying stress triaxiality and specimen rolling direction at room temperature and 600°C, respectively. It is worth underlining that instability strain values are not reported in case of shear specimens, as the simple shear fracture occurs without necking.

Table 4.2 Fracture and instability strain values at room temperature at varying stress triaxiality and specimen rolling direction.

T=25°C		0°		45°		90°	
$\dot{\epsilon}=0.1s^{-1}$	η	ϵ_f	ϵ_{inst}	ϵ_f	ϵ_{inst}	ϵ_f	ϵ_{inst}
shear	0.11	0.733	-	0.454	-	0.9	-
smooth	0.34	0.51	0.101	0.48	0.07	0.55	0.085
10 mm notched	0.42	0.43	0.018	0.45	0.016	0.42	0.017
5 mm notched	0.47	0.35	0.019	0.36	0.018	0.36	0.017
2 mm notched	0.52	0.34	0.019	0.35	0.019	0.35	0.016
Nakajima round	0.66	0.38	0.05	0.38	0.05	0.38	0.050

Table 4.3 Fracture and instability strain values at 600°C at varying stress triaxiality and specimen rolling direction.

T=600°C $\dot{\epsilon}=0.1s^{-1}$	η	0°		45°		90°	
		ϵ_f	ϵ_{inst}	ϵ_f	ϵ_{inst}	ϵ_f	ϵ_{inst}
shear	0.11	1.134	-	1.012	-	1.274	-
smooth	0.34	0.77	0.125	0.85	0.176	0.68	0.090
10 mm notched	0.42	0.68	0.062	0.72	0.085	0.65	0.052
5 mm notched	0.47	0.64	0.070	0.66	0.097	0.62	0.052
2 mm notched	0.52	0.6	0.072	0.64	0.099	0.59	0.053
Nakajima round	0.66	0.66	0.130	0.66	0.130	0.66	0.130

4.4.6 Calibration of the GISSMO model

For calibrating the GISSMO model, the curves of fracture strain ϵ_f and instability strain ϵ_{inst} vs. stress triaxiality as well as the fading exponent m and the damage exponent n are needed. It deserves to mention that the fracture strain presented here for calibrating the GISSMO model was measured by Aramis™ system, not through the SEM method on the fractured surface. Figure 4.21 reports the former curves as a function of the specimen rolling direction at room temperature and 600°C. A significant influence of the rolling direction is appreciable, with the 45° direction showing always a higher ductility in case of smooth, notched and Nakajima round specimens whereas a ductility increase is shown by the 90° direction in case of shear testing, regardless of the testing temperature.

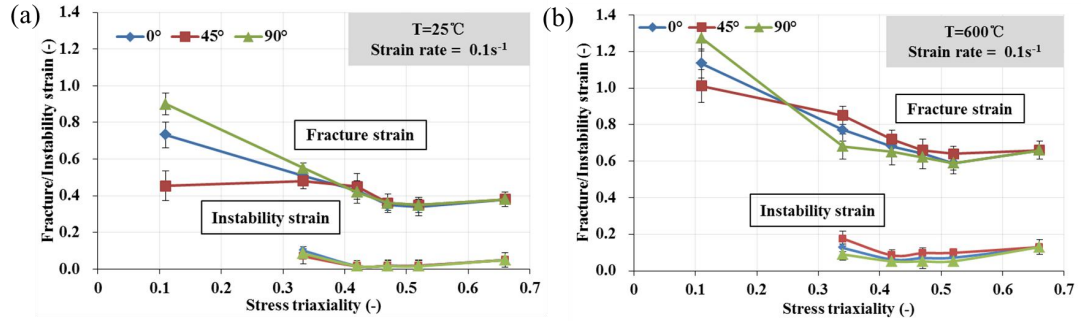


Figure 4.21 Fracture and instability strain curves as a function of the stress triaxiality and specimen rolling direction at (a) room temperature, and (b) 600°C.

The two exponents m and n were identified using the LS-OPT™ optimization software on the basis of the Sequential Response Surface Method (SRS) [100][101], whose procedure consists in varying the two exponents until the mismatch between the experimental and computed force vs. displacement curves is small enough. The experimental force vs. displacement curves chosen for the optimization procedure were the ones from the tensile tests carried out on smooth specimens at different rolling directions at 600°C, which were compared with the ones calculated by the related numerical models where the curves of Figure 4.21 (b) were implemented. The so-called Curve Mapping algorithm (which computes the area between the curves and the x-axis) was used for optimization, as is especially suitable in this case where the force vs. displacement curves are characterized by a

significant drop just before failure. The flow chart showing the optimization procedure is reported in Figure 4.22. After a few iterations, the damage exponent n was identified equal to 4 and the fading exponent m equal to 1.5. The same values were also suitable for the other testing conditions, therefore any further optimization was neglected.

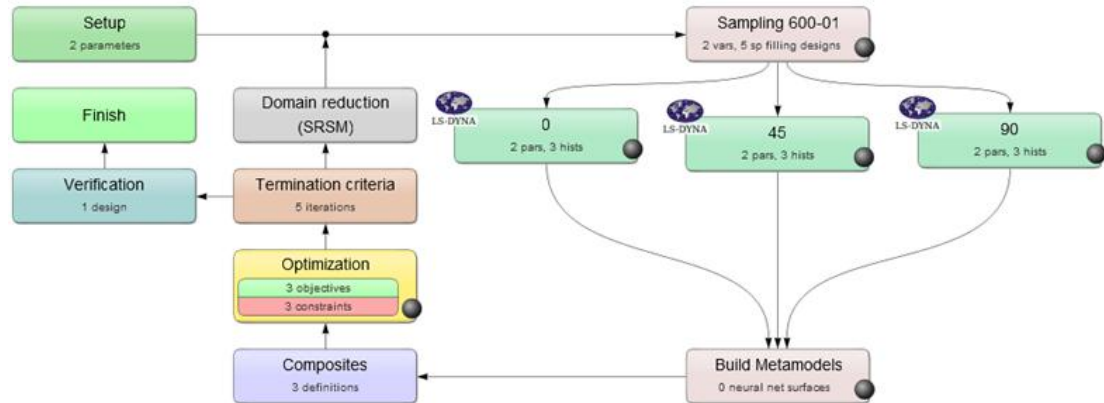


Figure 4.22 Flow chart of the optimization procedure to identify the damage and fading exponents of the GISSMO model.

4.4.7 The comparison between experimental and numerically predicted fracture characteristics

The comparison between the fracture characteristics predicted by the numerical models of the tests calibrated on the basis of the above described procedure and the experimental ones is shown from Figure 4.23 to Figure 4.32, which are relevant to tests carried out at room temperature and 600°C on specimens oriented at 0°, 45° and 90° with respect to the rolling direction. In case of tensile and shear tests, the comparison was carried out considering the strain field just before fracture, the shape of the fracture, and the force vs. displacement curve. In case of Nakajima tests conducted on round specimens, the comparison was about the shape of the fracture and the height of the fractured dome.

Figure 4.23, relevant to smooth specimens deformed at room temperature, shows a good agreement between experimental and numerical results, regardless of the specimen rolling direction. In case of 5 mm notched specimens deformed at room temperature, the agreement is still very satisfactory when testing the 0° and 45° rolling direction, with the 90° specimen showing a simulated fracture that is anticipated of approximately 9.5% compared to the experimental one, but with again a good prediction of the fracture shape. Similar results were obtained for the other notched specimens, and therefore not here reported. In case of shear specimens deformed at room temperature, the model is again capable to catch the fracture shape, but the fracture occurrence is overestimated for the 90° rolling direction of about 20%, even if the shape of the curve final portion is well represented. At room temperature, it is worth noting that the highest ductility shown by the 0° smooth specimen and the 45° notched specimen is well predicted, as is the highest ductility in case of 90° shear specimen. As shown in Figure 4.24 and Figure 4.25, the strain field is well predicted for the 90° rolling direction of the smooth

specimen, whereas the simulated fracture is less accurate for the 90° rolling direction of both the notched and shear specimens. The comparison between experimental and numerically predicted fracture characteristics in case of Nakajima round specimens deformed at room temperature is shown in Figure 4.26: both the fracture shape and the height of the fractured dome are well predicted.

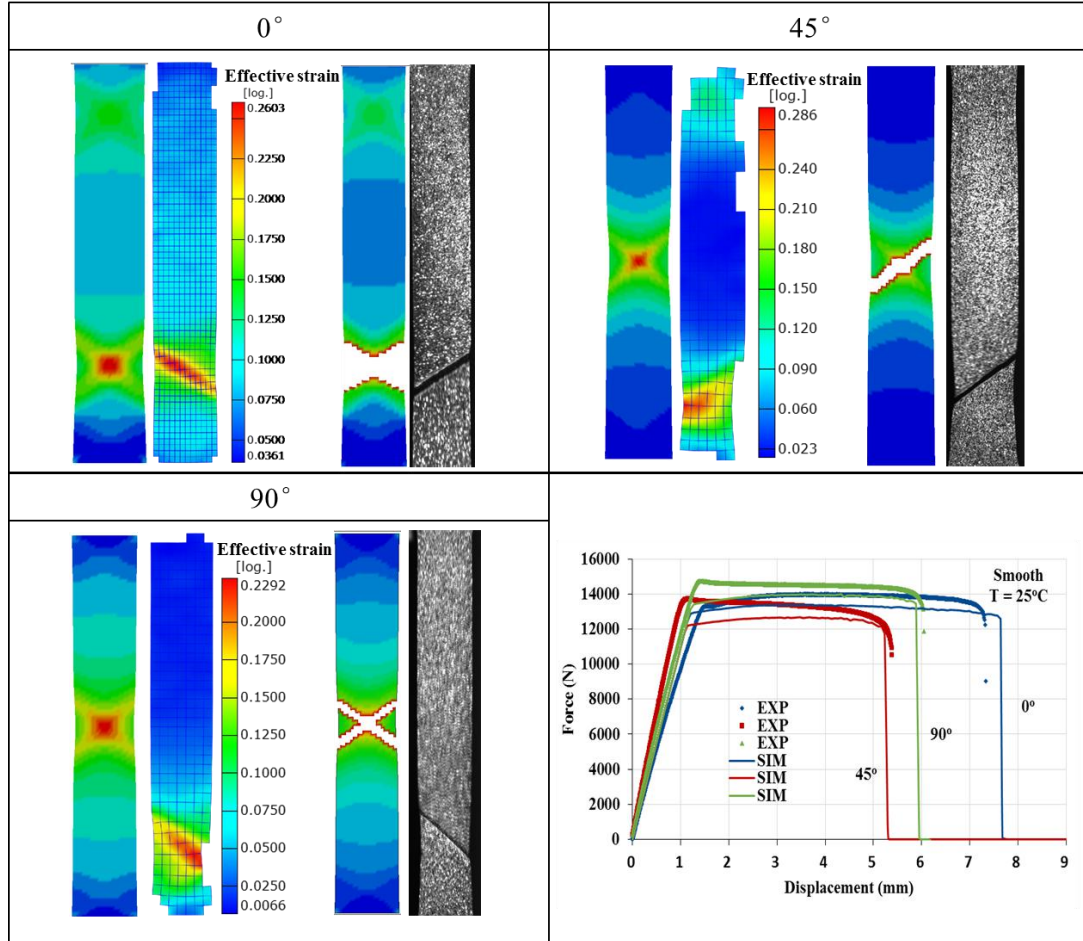


Figure 4.23 Experimental and numerically predicted fracture characteristics of smooth specimens deformed at room temperature at varying rolling direction.

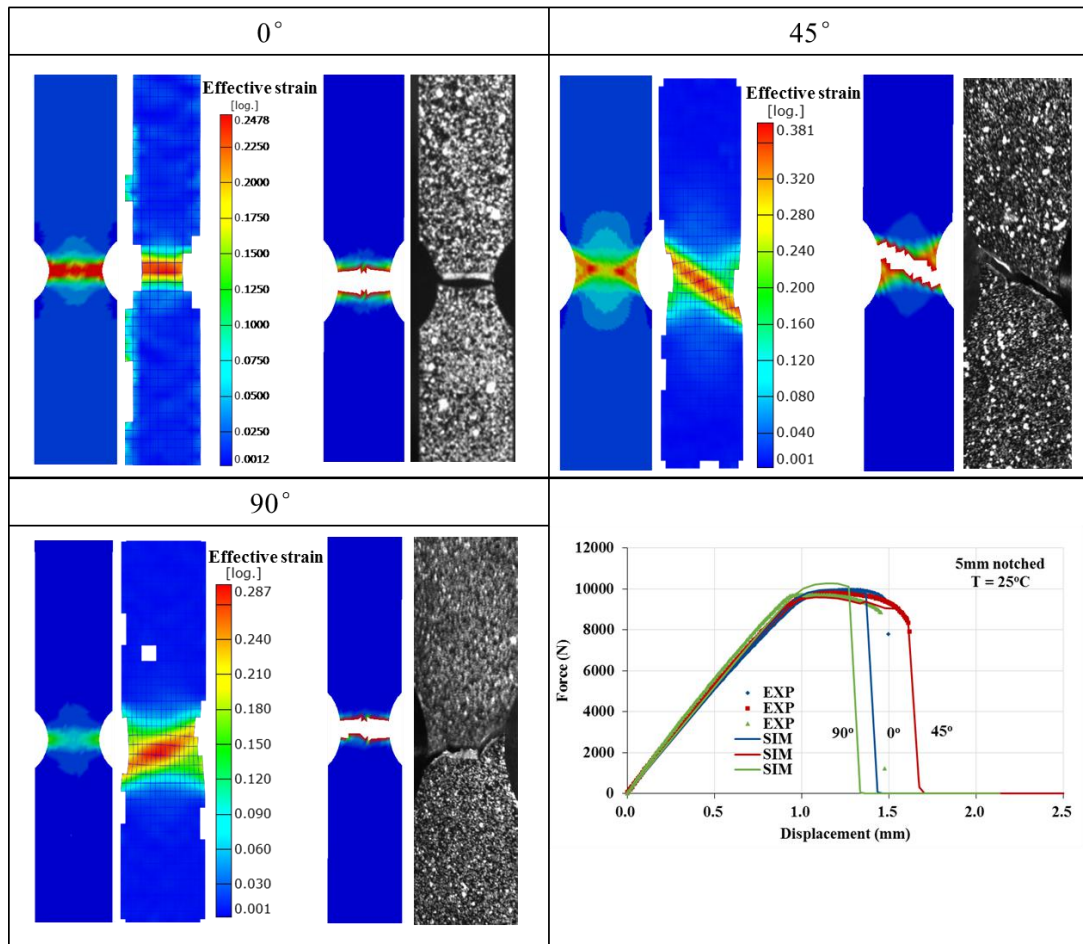


Figure 4.24 Experimental and numerically predicted fracture characteristics of 5 mm notched specimens deformed at room temperature at varying rolling direction.

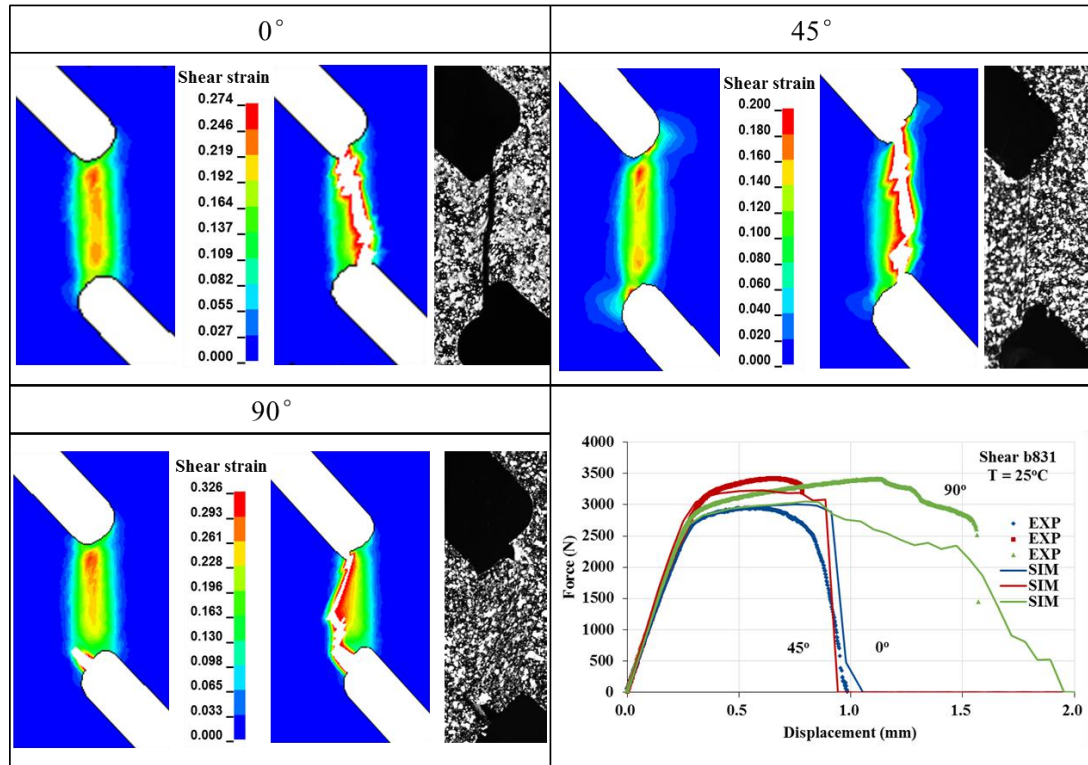


Figure 4.25 Experimental and numerically predicted fracture characteristics of shear specimens deformed at room temperature at varying rolling direction.

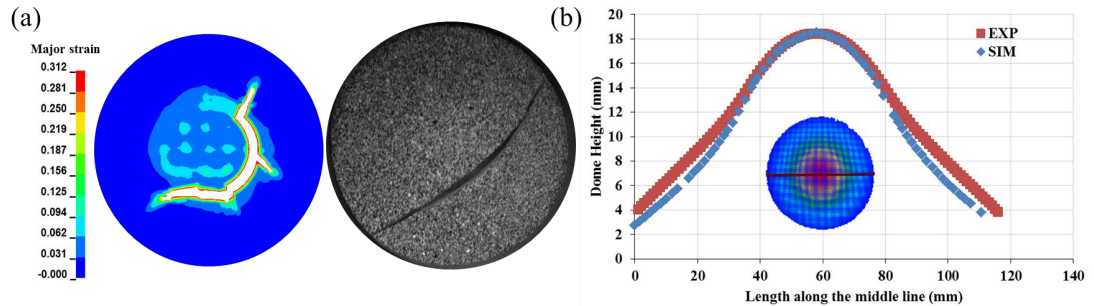
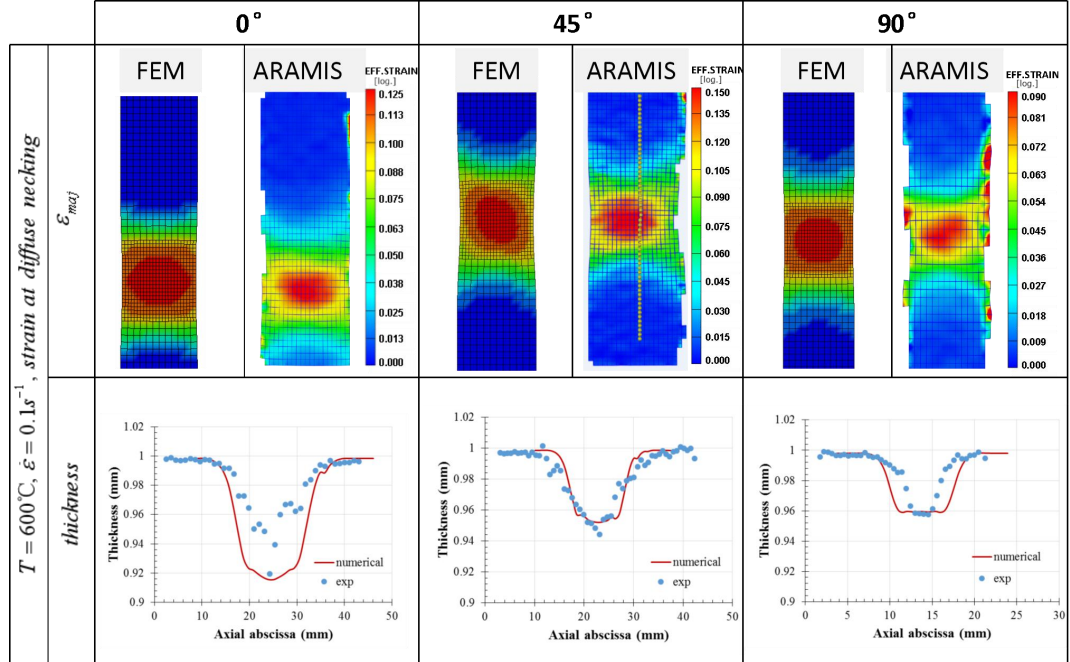


Figure 4.26 Experimental and numerically predicted fracture characteristics of Nakajima round specimens deformed at room temperature.

At 600°C, the fracture characteristics change as a consequence of the material higher ductility. The experimental and numerical results of smooth specimen of strain at diffusing necking and localized necking as well as specimen after fractured are shown in Figure 4.27 and Figure 4.28. The model was capable of predicting both the major strain field and thickness distribution along the specimen axis except the case of 45° rolling direction of localized strain condition, thickness distribution was less accurate, although the major strain distribution shape was correct. For predicting the fracture characteristics of smooth specimen, as shown in Figure 4.28, the strain field is well predicted for the 0° and 45° directions, whereas the experimental strain localization for the 90° agrees less with the numerical outcome. On the contrary, the prediction of the fracture shape is very good for all the three directions: especially at 45° direction even the location of the fracture onset at the middle of the specimen is correctly reproduced by the numerical model, as well as

the force vs. displacement curves. In case of 5 mm notched specimens, as shown in Figure 4.29 and Figure 4.30, the model is able to catch the diffuse necking and localized necking presents in terms of strain field and thickness distribution regardless the specimen orientation. Whereas the fracture behaviour, the model also able to catch the fracture occurrence at the edge of the specimens and the almost flat fracture shape regardless of the specimen rolling direction. At 90° direction, the numerical force vs. displacement curve differs of about 18% from the experimental one in terms of displacement at fracture, whereas the other two directions are well predicted, as is the case at room temperature. Similar results were obtained for the other notched specimens, and therefore not here reported. In case of shear specimens, as shown in Figure 4.31, regardless of the rolling direction, the numerically predicted fracture shape tilts on the left differently from the experimental fractures that tilt on the right. Contrary to room temperature testing, the experimental and numerical force vs. displacement curves relevant to 0° direction differ of about 17% in the displacement at fracture. It is worth underlining that the highest ductility shown by the 45° smooth and notched specimens is always predicted, as is predicted the highest ductility shown by the 90° specimens under shear testing conditions.

Again, the good agreement between experimental and numerical fracture shapes and positions in the Nakajima round specimen is seen in Figure 4.32 when testing at 600°C.



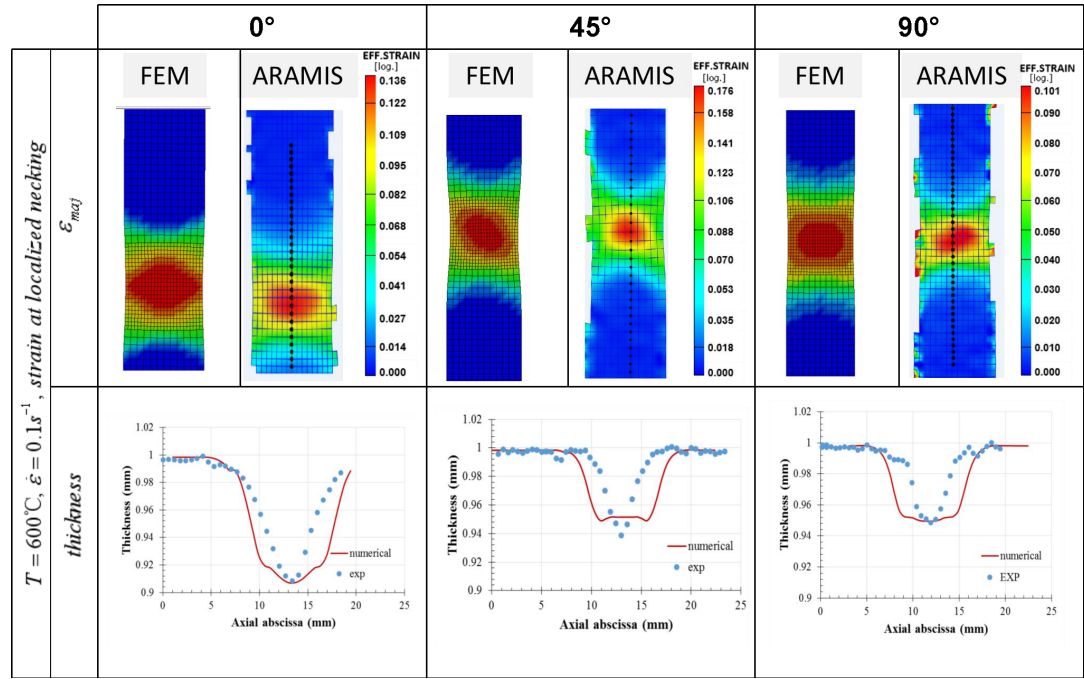


Figure 4.27 Comparison between experimental and numerical results of smooth specimen at diffuse necking and localized necking at 600°C, strain rate 0.1 s⁻¹.

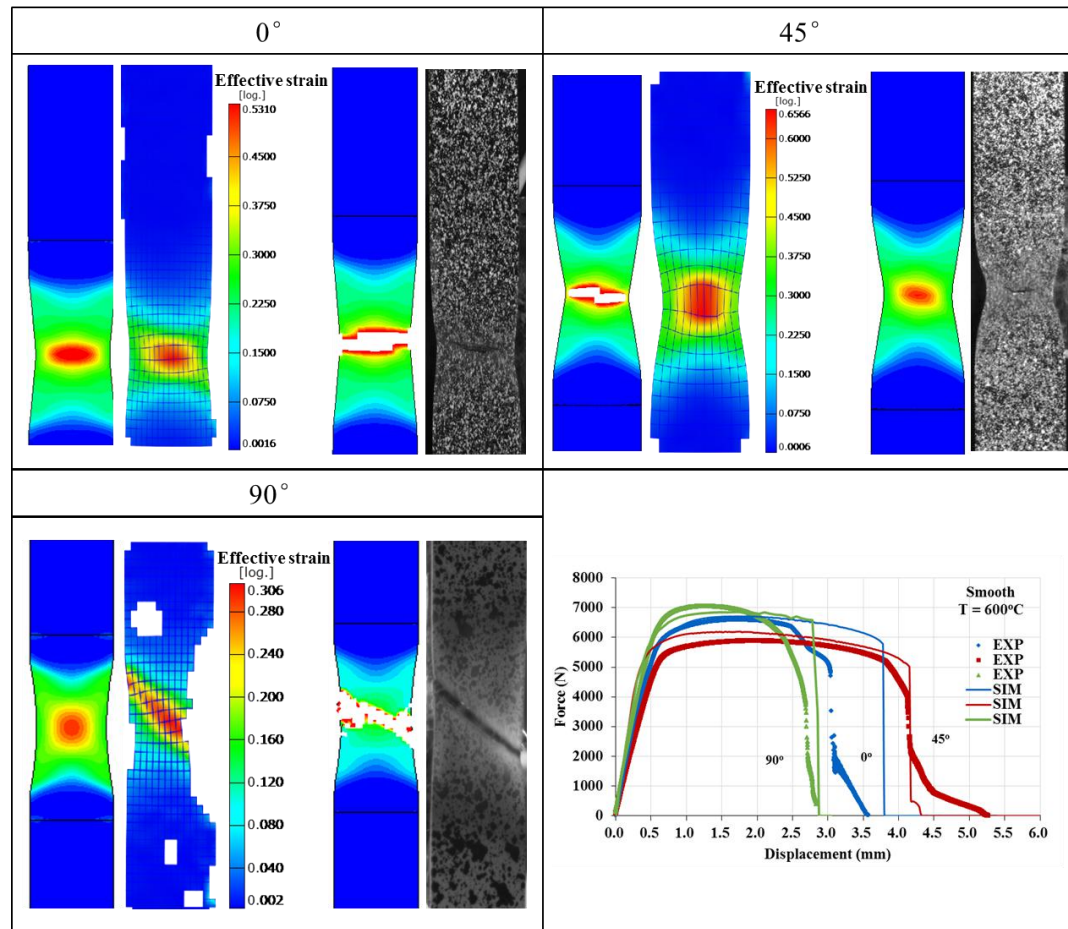


Figure 4.28 Experimental and numerically predicted fracture characteristics of smooth specimens deformed at 600°C at varying rolling direction.

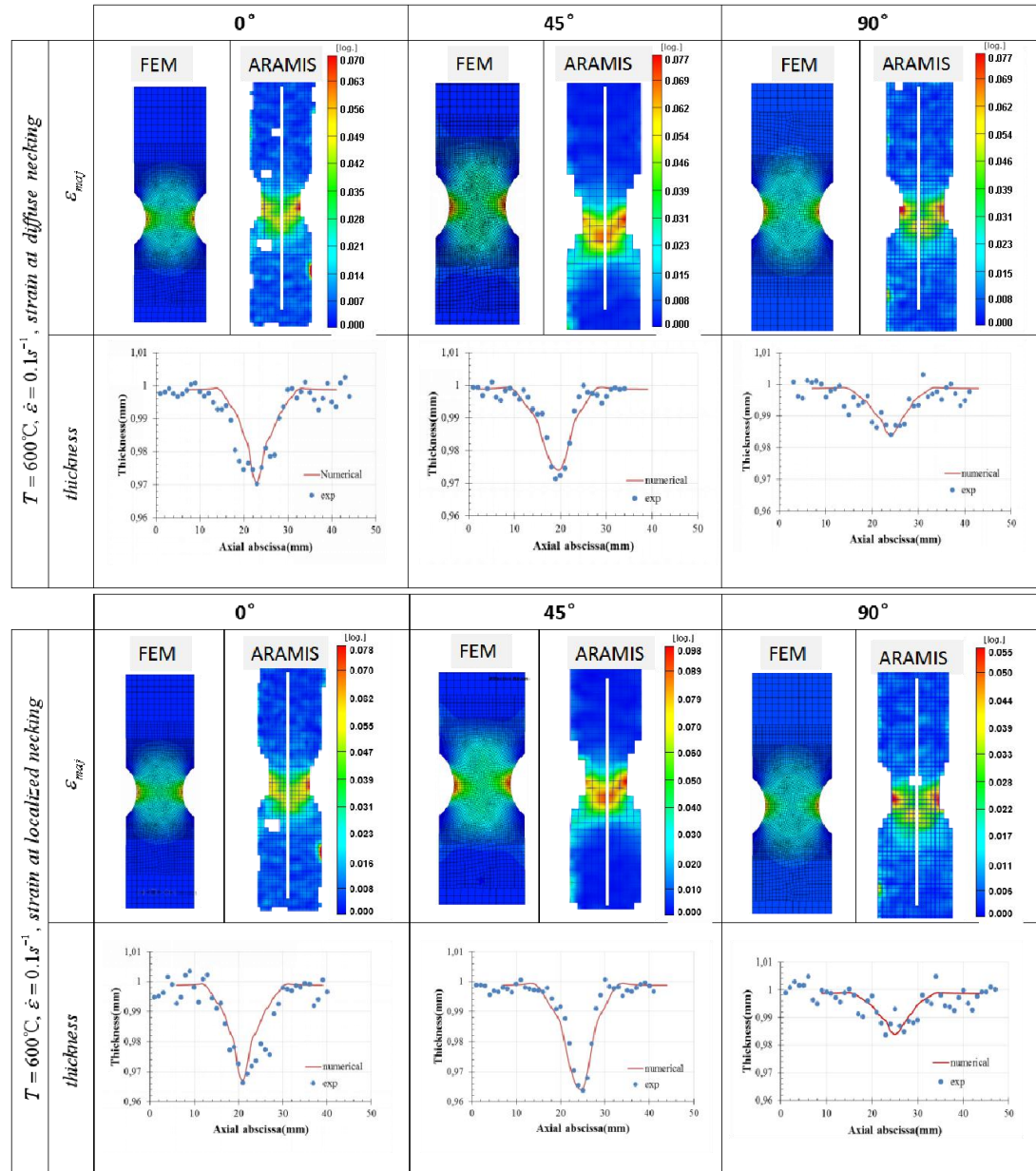


Figure 4.29 Comparison between experimental and numerical results of notched 5mm specimen at diffuse necking and localized necking at 600°C, strain rate 0.1 s⁻¹.

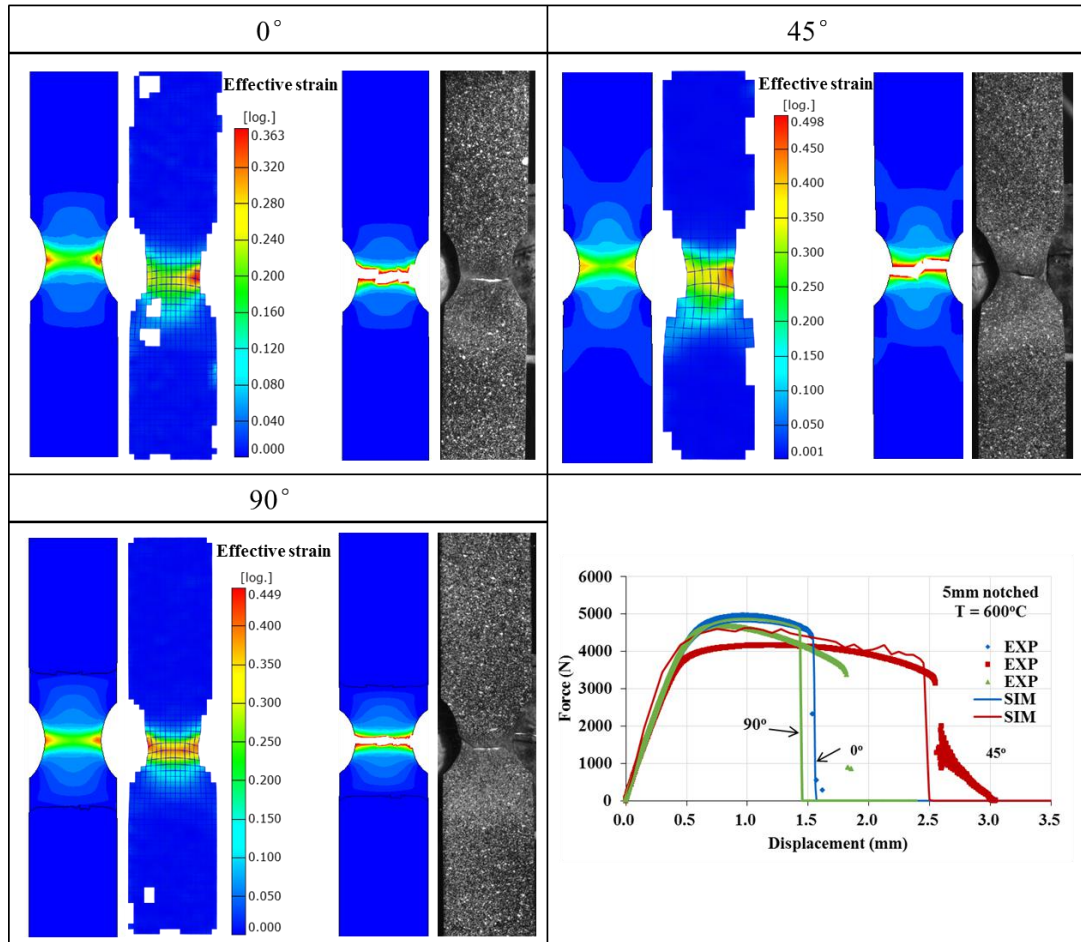


Figure 4.30 Experimental and numerically predicted fracture characteristics of 5 mm notched specimens deformed at 600°C at varying rolling direction.

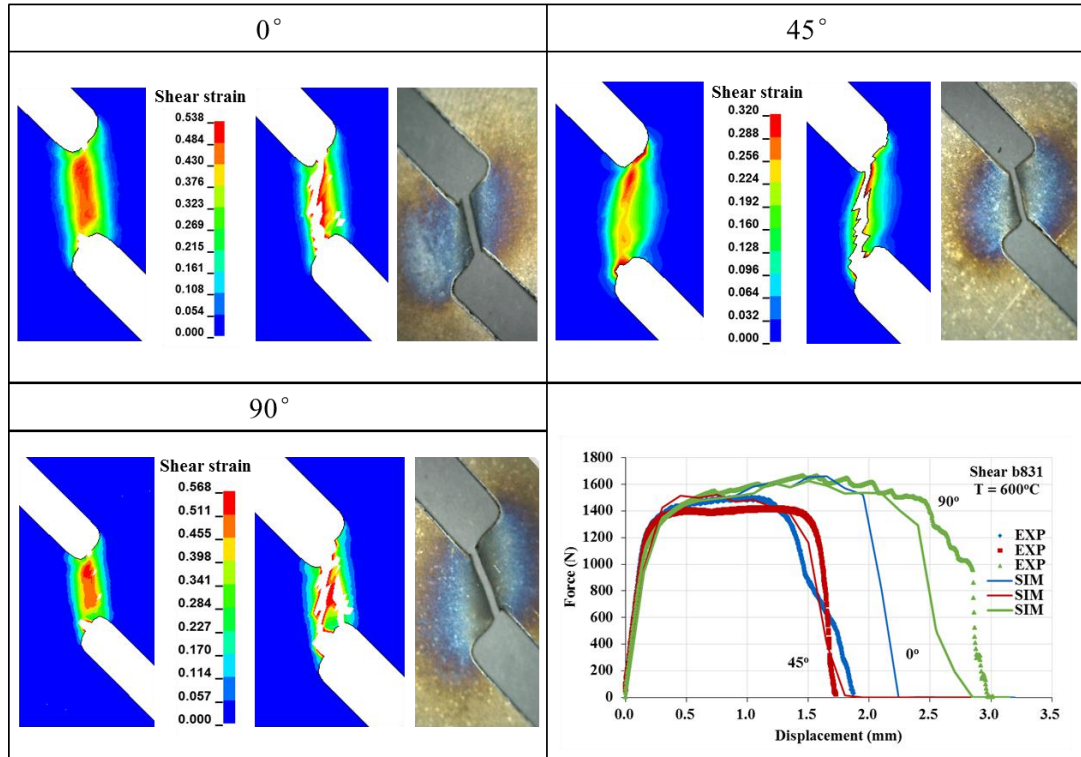


Figure 4.31 Experimental and numerically predicted fracture characteristics of shear specimens deformed at 600°C at varying rolling direction.

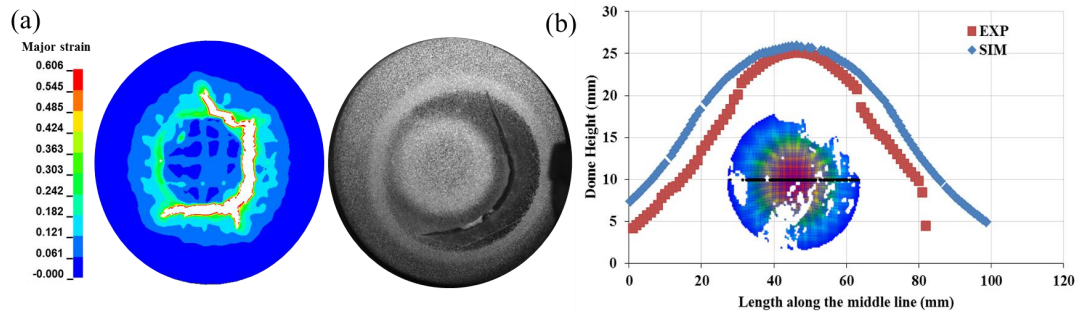


Figure 4.32 Experimental and numerically predicted fracture characteristics of Nakajima round specimens deformed at 600°C.

4.4.8 Modelling validation

The experimental outcomes from testing Nakajima notched specimens, not included in the calibration procedure, were used to validate the modelling approach proposed in section 4.4.4. Figure 4.33 (a) shows the numerical model used for such test, whereas Figure 4.33 (b) shows the comparison between the experimental and numerical dome heights. As expected, as a consequence of the material increased ductility, the dome height at room temperature is lower than the one at 600°C. Both at room temperature and 600°C, the agreement between experimental and numerical results is very satisfactory, with a maximum difference less than 15%. Figure 4.34 shows the comparison between the experimental and numerically predicted fracture morphology and location at different specimen rolling direction, both at room temperature and 600°C. It is worth noting that the GISSMO model is capable of predicting both the location and fracture shape. Even the tilted crack shape characteristic of the 45° oriented specimen is well predicted.

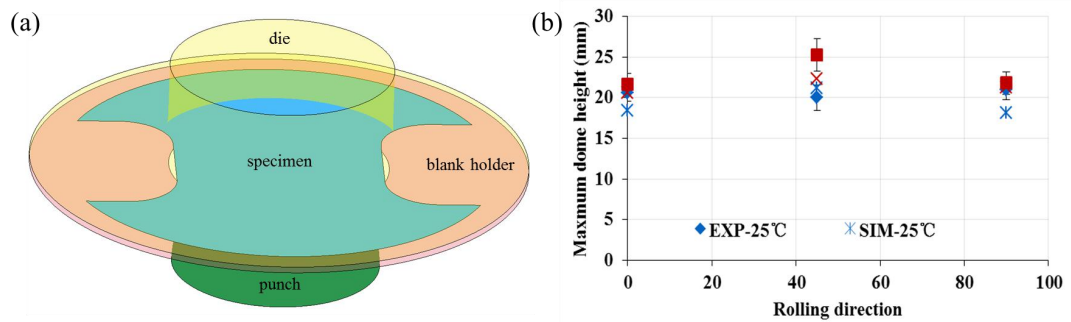


Figure 4.33 (a) FE model of the Nakajima set-up for testing notched specimens. (b) Experimental and numerically predicted dome heights at room temperature and 600°C at varying rolling direction.

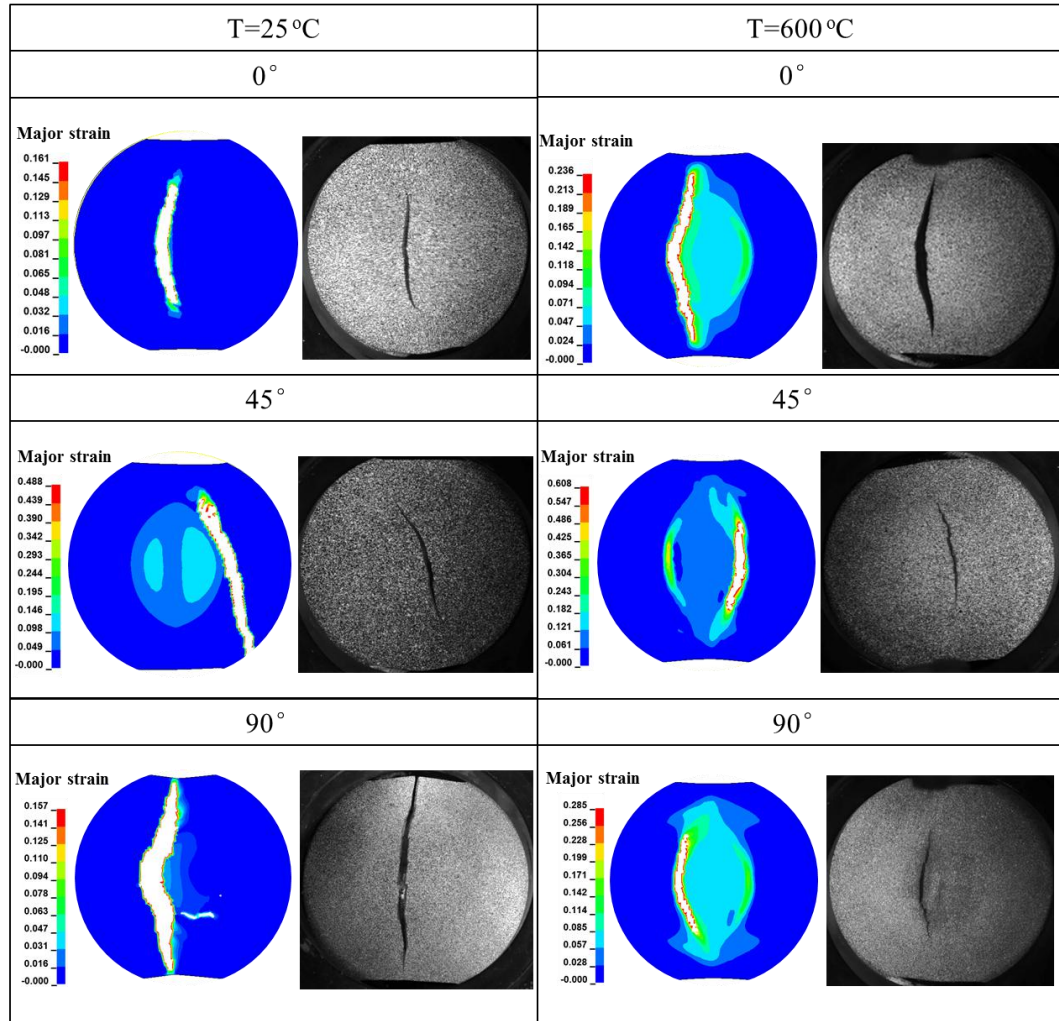


Figure 4.34 Experimental and numerically predicted fracture characteristics of Nakajima notched specimens deformed at room temperature and 600°C at varying rolling direction.

4.4.9 Conclusions

This chapter showed that the anisotropic characteristics of Ti6Al4V sheets influenced the material behaviour in a wide range of temperature from room temperature to 800°C, on the basis of the outcomes from tests using smooth, notched and shear specimens. The obtained results show that smooth and notched specimens oriented at 45° has the highest fracture strain, whereas shear specimens oriented at 90° has the highest ductility, regardless the testing temperature. This means that anisotropic effect have to be taken into account when modelling damage evolution and fracture occurrence at varying stress state.

Then the research focused on the numerical prediction of ductile failure of Ti6Al4V titanium alloy sheets when deformed at room and elevated temperature taking into account the sheet anisotropic behaviour. In order to do that, the Barlat-Lian 1989 yield criterion and the GISSMO model were used, suitably calibrated on the basis of an extensive experimental campaign that included tensile tests on smooth and notched specimens, shear tests and Nakajima-type tests in order to have a wide range of stress triaxiality values. The tests were carried out on specimens machined

at different rolling orientation to account for the material anisotropic behaviour. The numerical model was then validated on the basis of the outcomes of tests not included in the calibration phase.

The following conclusions can be drawn:

- The Ti6Al4V sheets retained their anisotropic behaviour till 700°C, being both the strain at fracture and instability strain significantly influenced by the rolling direction at varying stress triaxiality; in particular, the 45° direction always showed the highest ductility, except in case of shear specimens, both at room temperature and 600°C.
- The Barlat-Lian1989 yield criterion was calibrated on the basis of the Lankford coefficients, and the associated hardening rule on the basis of the material flow stress at varying temperature and strain rate, whereas the GISSMO model by means of the strain at fracture and instability strain values at varying stress triaxiality and rolling direction, at room temperature and 600°C.
- The numerical models of the tensile and shear tests with implementations of the calibrated Barlat-Lian 1989 and GISSMO models proved to give a satisfactory agreement with the experimental outcomes, in terms of strain field, thickness distribution after plastic instability, force vs. displacement curves, fracture location and characteristics; the same agreement was found in case of Nakajima-type tests in terms of dome height and fracture location and characteristics, considering both the geometry used for calibration and the one for validation.
- The models were, in general, able to catch the differences in fracture occurrence due to the different anisotropic characteristics, with just a slight disagreement in case of 90° oriented specimens, especially under shear conditions, which need further investigations.

Chapter 5

5 Results and discussion for magnesium alloys

5.1 Anisotropy influence on plastic and fracture behaviour of AZ31B alloys at room and elevated temperature

The objective of this section is to evaluate the anisotropy influence on the mechanical and microstructural characteristics of AZ31B magnesium alloy sheets deformed at room and elevated temperature. To this aim, uniaxial tensile tests were carried out along the rolling, transverse and diagonal direction in the range of temperatures from room temperature to 300°C at 0.1 s⁻¹. The Lankford coefficients, ultimate tensile strength, diffuse necking strain and fracture strain values were evaluated as a function of the testing temperature and specimen orientation. Furthermore, microstructural features were analyzed as well as micro-hardness was measured for each testing condition to assess the post-deformation characteristics.

5.1.1 Flow behaviour and plastic anisotropy

The Ultimate Tensile Strength (*UTS*), namely the maximum stress value in the engineering stress-strain curve, is shown in Figure 5.1 as a function of temperature and specimen rolling direction for a strain rate equal to 0.1 s⁻¹. As expected, the *UTS* values decrease at increasing temperature regardless of the specimen orientation. The *UTS* values are higher at 90° rolling direction than at 0° and 45° at room temperature and 100°C, whereas the differences between 200°C and 300°C are almost negligible. The same sensitivity was presented in [102][103]. The Lankford coefficients *R*, evaluated as the ratio between the strains in the specimen width and thickness, are reported in Figure 5.1 (b). A point at the center of the necked area on the specimen surface was selected as measuring point, according to the procedure reported in [104]. The Lankford coefficients here reported are representative of the steady state region of the true strain - *R* curve as acquired through the AramisTM system. The *R* values are always higher than 1 regardless of the testing temperature and specimen rolling direction, due to the fact that one of the main deformation mechanisms is the prismatic <a> slip resulting in greater strain in the width than in the thickness direction [105][106]. An increase of the *R* values is shown from room temperature to 100°C, and then a decrease until 250°C, with almost the same values kept at 300°C. It is worth noting that still high *R* values demonstrate a strong participation of prismatic <a> slip on the overall deformation mechanisms. The increase of anisotropy from room temperature to 100°C can be explained by easy slip of <a> on basal and prismatic planes, while, at temperatures above 100°C, the pyramidal <c+a> slip is more and more influential, leading to *R* values reduction. The same trend is reported in [105] in case of coarse grained

sheets. The more anisotropic behavior evident at 90° rolling direction can be a consequence of the transition from basal to non-basal $\langle a \rangle$ slip systems [107].

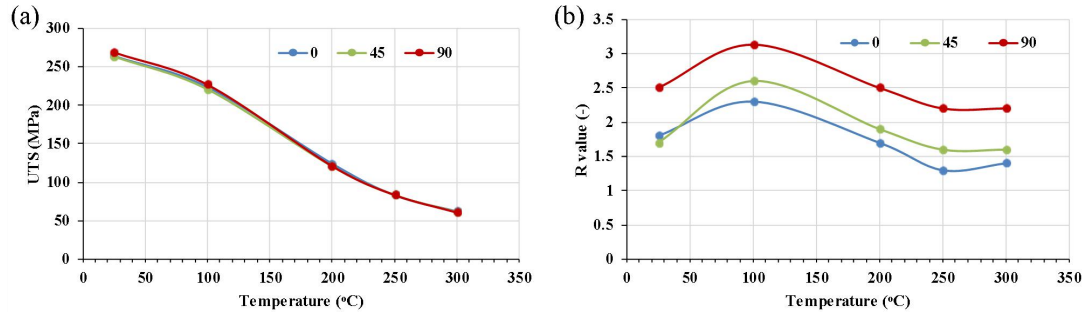


Figure 5.1 (a) UTS , and (b) R values as a function of the testing temperature and specimen rolling direction

Figure 5.2 (a) reports the width strain - true strain and thickness strain - true strain curve at different specimen orientation at room temperature: at 90° rolling direction, the width strain is larger as well as the thickness strain is smaller compared to the rolling and 45° direction. If straining happens along the thickness direction, the contraction along the c-axis due to twinning or pyramidal slip must occur. However, the CRSS is higher on c-axis direction, which makes the width direction experience more strain than the thickness direction. This is also observed in paper [66]. The same was observed at the other testing temperatures, and therefore not here reported. Figure 5.2 (b) shows that the average normal anisotropy is always higher than 1.5, which means much less deformation on the thickness direction than on the width direction, namely the thickness resistance is high to avoid localized thinning. On the contrary, the planar anisotropy is close to 0, which means ears will not develop during deep drawing processes. The same outcomes are given in [108].

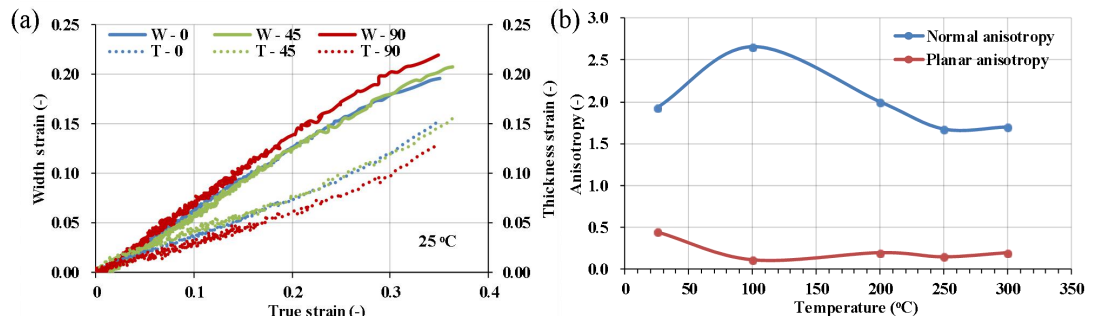


Figure 5.2 (a) Width and thickness strains as a function of specimen rolling direction at room temperature, and (b) average normal and planar anisotropy at varying temperature.

The strain at diffuse necking, also called instability strain, is identified when necking starts developing showing a symmetric decrease of the width of the specimen. The value is the true strain at UTS . Figure 5.3 (a) reports the diffuse necking strain as a function of the testing temperature and specimen rolling direction: specimens cut at 90° with respect to the rolling direction show earlier the onset of diffuse necking. In addition, the width strain is larger at 90° direction, which is also consistent with the diffuse necking appearing earlier at 90° direction (Figure 5.2 (a)). The instability strain increases up to 100°C and then decreases

regardless of the specimen rolling direction. This different behavior is due to the deformation mechanism: at room and 100°C, plastic deformation is mainly controlled by twinning (Figure 5.4) [109]. On the other hand, above 100°C, at increasing temperature, mechanical twinning effect decreases whereas dislocation motion is enhanced as a consequence of the activation of more slip systems as well as recrystallization occurred making the microstructure finer (Figure 5.4): both these phenomena promote the development of diffuse necking.

5.1.2 Fracture strain

The fracture strain at varying testing temperature and specimen rolling direction is shown in Figure 5.3 (b). The strain at fracture increases at increasing temperature, regardless of the specimen rolling direction, nevertheless being always the highest at 45° rolling direction. Below 100°C, even if twinning is active as deformation mechanism, the lowest ductility is due to the twins that act as barriers for the dislocation motion [110]; on the other hand, above 200°C, the activation of non-basal slip systems increases the ductile behavior, while the effect of mechanical twinning is reduced deformation, being the temperature of 200°C the activation temperature of additional slip systems [44].

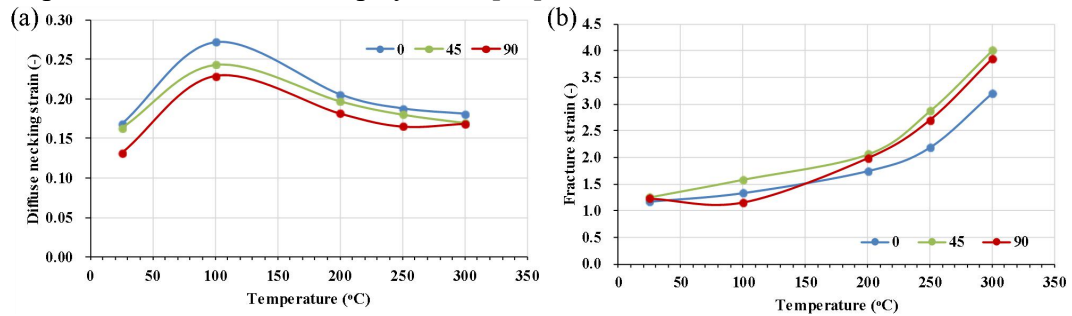


Figure 5.3 (a) Diffuse necking strain, and (b) fracture strain as a function of the testing temperature and specimen rolling direction.

5.1.3 Microstructure and fracture surface characterization

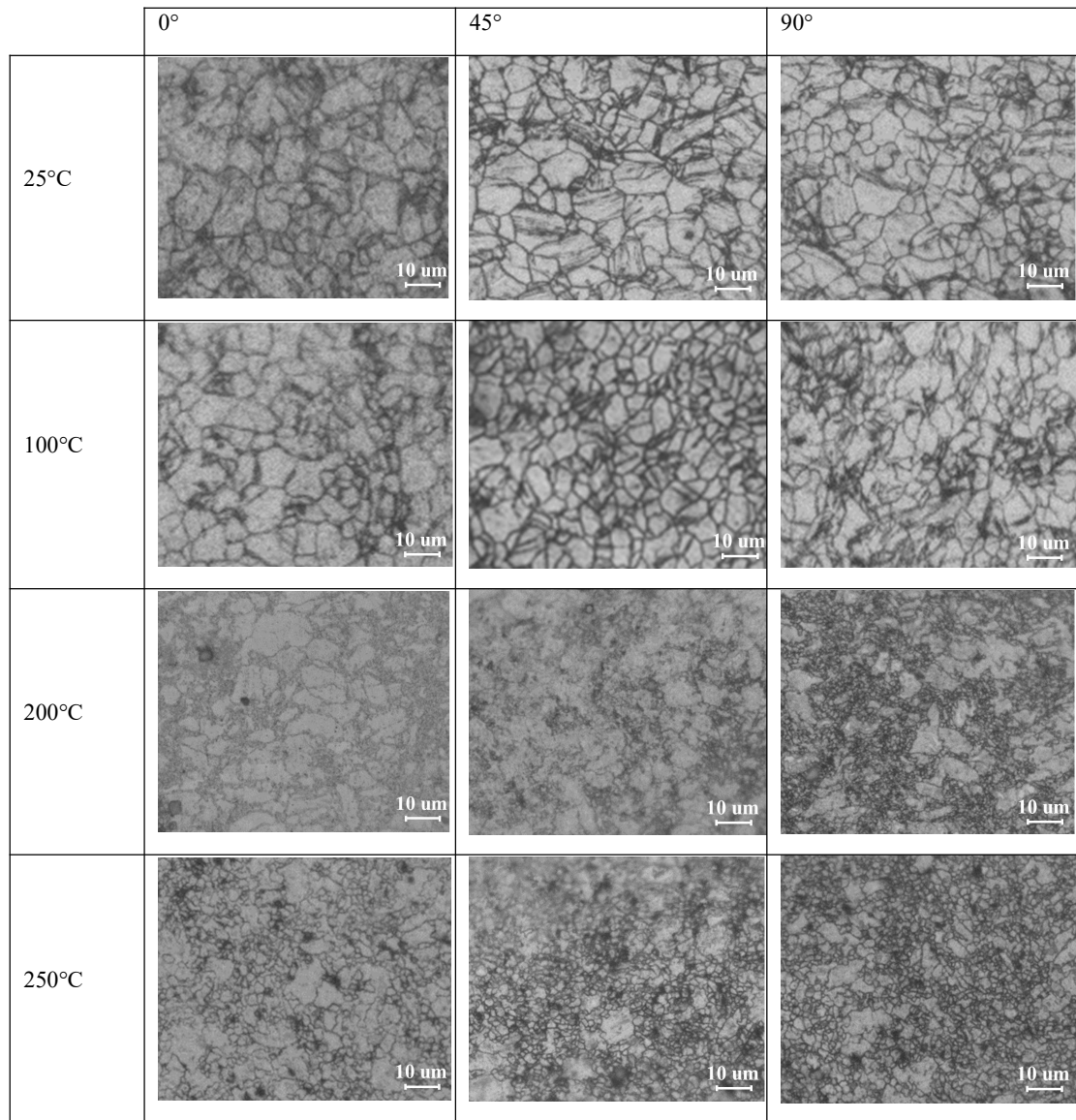
Figure 5.4 reports the microstructure of the deformed specimens at varying testing temperature and specimen rolling orientation. As expected, the grain size decreases at increasing temperature. Both at room temperature and 100°C, a large number of twins is visible in all the rolling directions, indicating that twinning was the dominant deformation mechanism at relatively low temperatures. On the contrary, above 100°C, twins disappear, some small grains start forming along the grain boundaries of coarser grains as a consequence of the onset of dynamic recrystallization, with an increasing grain refinement at 250°C and 300°C. No significant differences in the microstructure were found at different specimen orientation, meaning that there is no anisotropy sensitivity on microstructure.

Figure 5.5 (a-c) reports the grain size distribution in the as-received sheet and nearby the fracture zone at room temperature and 300°C. The grain size measured using the line intercept method. In the as-received sheet, the Average Grain Size

(AGS) is approximately $9\ \mu\text{m}$ with a high Standard Deviation (SD) equal to $4.8\ \mu\text{m}$, meaning a highly inhomogeneous microstructure. The AGS is smaller at room temperature, namely $5.4\ \mu\text{m}$, as a consequence of the specimen straining, and even smaller at 300°C , namely $3.5\ \mu\text{m}$, as a consequence of dynamic recrystallization, which makes the microstructure much more homogeneous.

Figure 5.5 (d) shows the Vickers micro-hardness after deformation of the same specimens, which increases at temperature increase as a consequence of the grain size refinement, which is consistent with Hall-Petch equation, according to

$$\sigma_0 = \sigma_i + k / \sqrt{D} \quad (5.1)$$



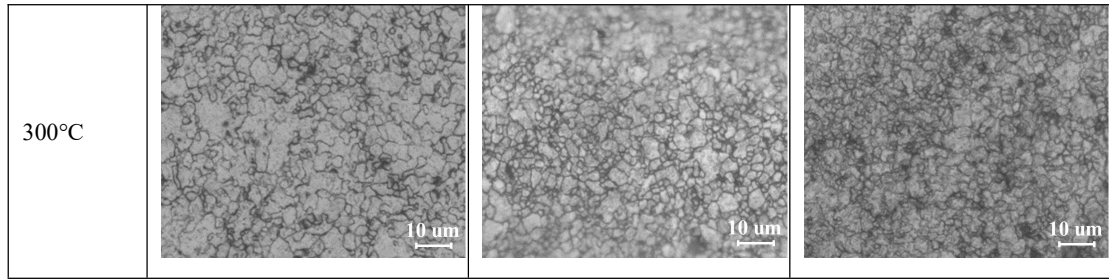


Figure 5.4 Microstructure after deformation as a function of the testing temperature and specimen rolling direction.

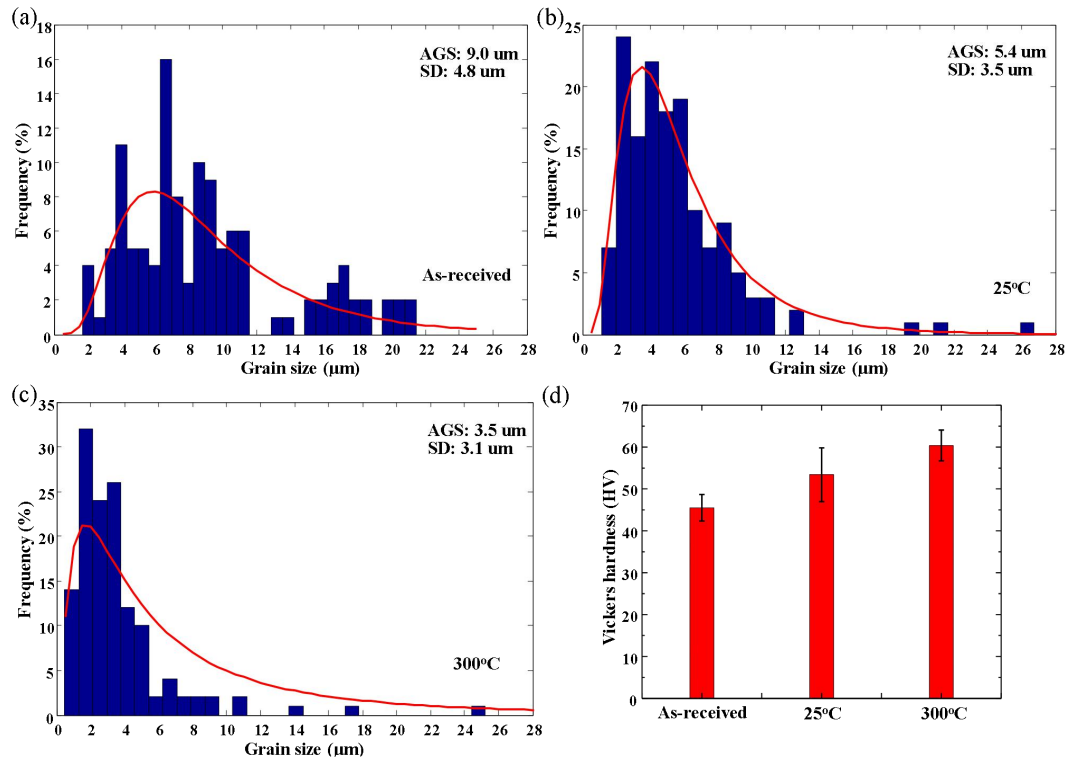


Figure 5.5 Grain size distribution of sheets in: a) as received condition, b) at 25°C, c) at 300°C, and d) Vickers micro-hardness as a function of testing temperature at rolling direction.

5.1.4 Conclusions

This chapter shows the anisotropy influence on the mechanical and microstructural characteristics of AZ31B sheets deformed at room and elevated temperature based on uniaxial tensile tests. The obtained results shows that the diffuse necking strain is higher at rolling direction with the maximum value shown at 100°C, whereas, the highest ductility is shown at 45° rolling orientation, regardless of the testing temperature. This proves the need to consider the anisotropy influence when modelling sheet forming processes carried out on AZ31 sheets even at elevated temperature. On the contrary, the influence of the specimen rolling direction on the microstructure is negligible.

5.2 Anisotropic fracture behaviour of AZ31B alloy as a function of the stress state and temperature

The objective of this chapter is to determine experimentally the fracture locus of the AZ31B magnesium alloy at varying temperature, stress state, and rolling direction and model it by adding a temperature-dependent parameter to the MMC fracture criterion. A comprehensive experimental campaign was carried out, including uniaxial tensile tests on smooth and notched specimens, tension-shear compound tests, shear tests, and Nakajima-type tests at varying temperature and specimen rolling direction.

5.2.1 Procedure for assessing the fracture strain

The onset of fracture during mechanical tests is always challenging to get observed. One of the more common methods to measure fracture strain, often referred to as final fracture strain, consists in measuring the specimen cross-section area after fracture and calculating the shrink percentage compared to the initial cross-section area. The main disadvantage of this method lies in the fact that fracture surface boundaries are usually difficult to be clearly identified in case of ductile fracture at high temperature because of their sharp slope. To overcome this limitation, a hybrid experimental-numerical method was recently applied with the aim of measuring the fracture strain during tests [111]. However, this method needs both an accurate material yield criterion and a reliable numerical analysis. Many yield criteria for isotropic and anisotropic metals are available but their calibration is usually based on uniaxial tensile tests carried out on smooth specimens, therefore neglecting the mechanical response for other loading conditions. This is especially true for the AZ31 magnesium alloy whose mechanical response significantly varies along with the loading condition, as previously demonstrated.

In this chapter, the Digital Image Correlation (DIC) approach was used to directly measure the fracture strain of the smooth, notched, special and shear specimens. The fracture strain was identified as the maximum major strain along the loading direction as soon as the first visible crack appeared on the specimen surface. The method is very appropriate at room temperature and 100°C when the initial crack is easy to be recognized. As example, this method was applied in literature [112] to measure the fracture strain of the AA6082 aluminium alloy deformed using different specimen geometries at room temperature. However, at relatively high temperature, the material showed a higher ductile behaviour developing pronounced necking before fracture. Therefore, it became more and more difficult to determine the initial crack location, since the pattern sprayed on the specimen for the Aramis™ acquisition peeled off due to the large deformation prior to fracture, thus making impossible the Aramis™ computation.

The notched specimens cut along the rolling direction are used as an example to describe the method used to identify the fracture strain values at room temperature, 100°C and 300°C. As shown in Figure 5.6 (a-c), the load-stroke curves and strain-

stroke curves acquired at the same temperature are drawn on the same figure. The data acquisition frequency of MTS™ and Aramis™ helped putting the two curves at the same axial stroke value. The major strains both at the middle and at the edge along the cross-section were considered. From the presented results, it is worth noting that the major strain was larger at the edge than at the middle regardless of the testing temperature, which means the first crack occurred at the edge of the minimum cross section. Therefore, at room temperature and 100°C, the fracture strain was determined at the intersection between the load-stroke curve and the edge strain curve one, evidenced in Figure 5.6 (a-b) with a black star. The obtained fracture strain values are very much consistent with the ones got just using the DIC method. On the contrary, at 300°C the above reported is not doable (see the position of the black star in Figure 5.6 (c)); therefore, another method is here suggested, which makes use of the load first derivative representing the rate of the load change: the minimum of this curve represents the occurrence of localized necking, which was taken as indicative of the first crack occurrence (see Figure 5.6 (d)).

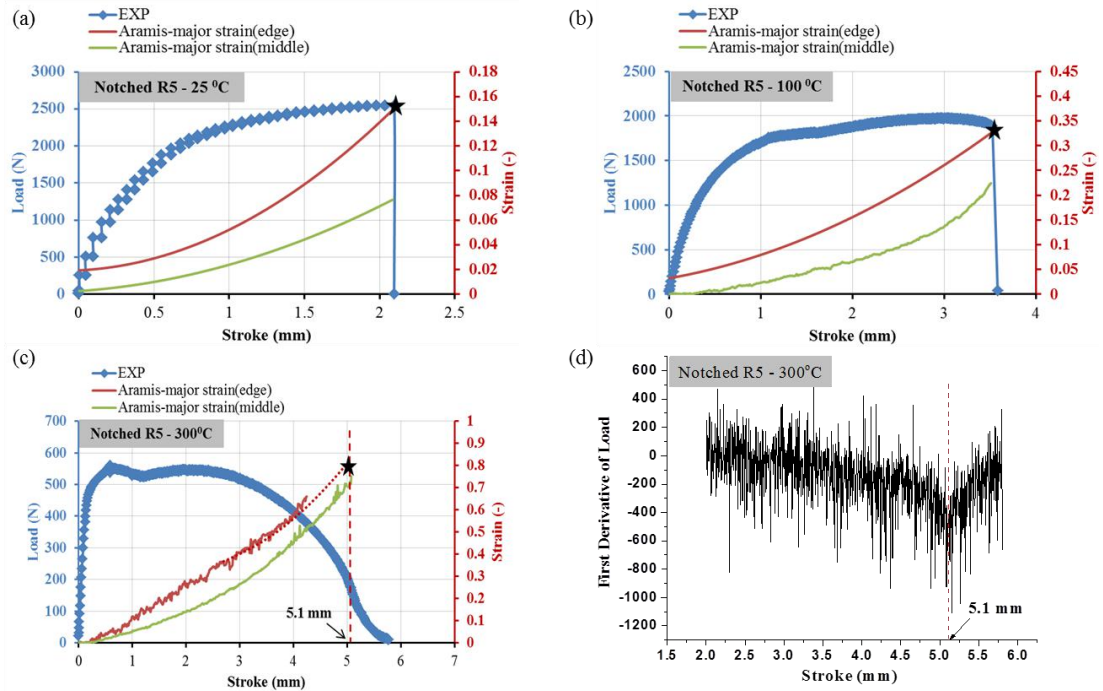


Figure 5.6 Load-stroke curves and strain-stroke curves of notched specimens at: (a) room temperature, (b) 100°C and (c) 300°C; (d) load first derivative curve of the notched specimen at 300°C.

The same procedures were used for all the specimen geometries. In the case of the Nakajima round specimen, the fracture strain was directly recorded from the Aramis™ strain evolution, being the initial crack easy to be identified even at 300°C (see Figure 5.7).

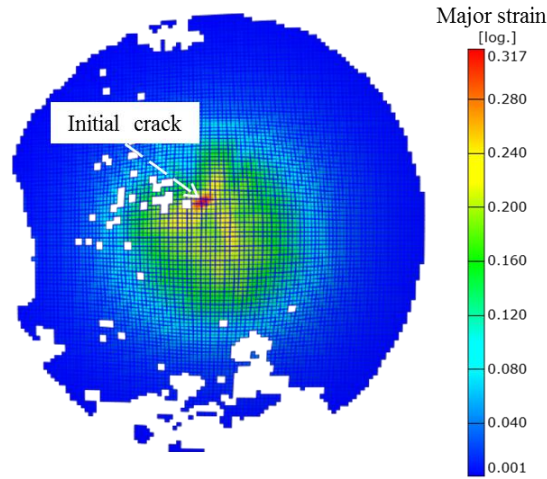


Figure 5.7 Major strain field just before fracture for the Nakajima round specimen at 300°C.

5.2.2 Stress triaxiality values

Since the evolution of the stress triaxiality and Lode angle parameters could not be obtained directly from the experiments, their values were extracted from the numerical simulations of the experiments carried out on the different specimen geometries. The numerical simulations were carried out as implicit analysis using the Ls_Dyna™ code version R 9.1.0 (Ls_Prepost V4.7) with shell elements. The hardening law was calibrated on the basis of the tensile tests conducted on the smooth specimens at varying temperature [113]. The anisotropic yield criterion Barlat-Lian 1989 [96] was adopted and the related anisotropy coefficients at varying temperature were employed to simulate the anisotropy characteristics of the AZ31B sheets. The developed FEM model was validated using experimental data other than the ones used for its calibration.

The actual thermal fields of the specimens were implemented on the basis of the infrared thermo-camera recordings: examples of the ones of the smooth and shear specimens at 300°C are reported in Figure 5.8. Examples of the stress triaxiality fields at room temperature at equivalent plastic strain of 0.1 s^{-1} are shown in Figure 5.9 for all the specimen geometries. The middle point of the minimum cross-section was selected for all the specimen geometries as the reference point to get the stress triaxiality values. Firstly, The distribution of triaxiality and lode parameter along the cross-section of notched, special and shear specimen at room temperature on rolling direction is shown in Figure 5.10. The result of smooth specimen has been neglected here, since triaxiality value and lode parameter are uniform. The strain at 0.15 for notched and shear specimen and the strain at 0.6 for special specimen where before necking and after elastic part are used, the figure reveal that the triaxiality and lode parameter are not average distribution along the intersection. Figure 5.10 (a) present that the maxi triaxiality value of 0.47 for notched specimen appears in the middle of the cross-section, the min value of 0.33 appears at the edge of cross-section. While the lode parameter shown an opposite trend, the min value of 0.7 in the middle and the maxi value of 1 at the edge. A wave curve of triaxiality and lode parameter along the cross-section of both special and shear specimen is

reported in Figure 5.10 (b-c). It is worth to note that triaxiality and lode parameter have a certain and same value at the edge of cross-section regardless specimen geometry, which is 0.3 and 1 for triaxiality and lode parameter, respectively. Therefore, on middle point of the cross-section is selected for all geometry specimens as the reference point. Moreover, the relationship between triaxiality and lode parameter for all geometry specimens in Figure 5.10 are also following the formulation, according to

$$\xi = -\frac{27}{2}\eta\left(\eta^2 - \frac{1}{3}\right) \quad (5.2)$$

In order to know the reason why wave trend shown in Figure 5.10 (b) of special specimen. The evolutions of triaxiality and lode parameter as a function of equivalent strain at three position in cross-section shown in Figure 5.10 (d), the position point 1 (edge), point 2 (between edge and middle), and point 3 (middle) are convergence during the tensile process and separate after which means a stress state varying from shear to uniaxial.

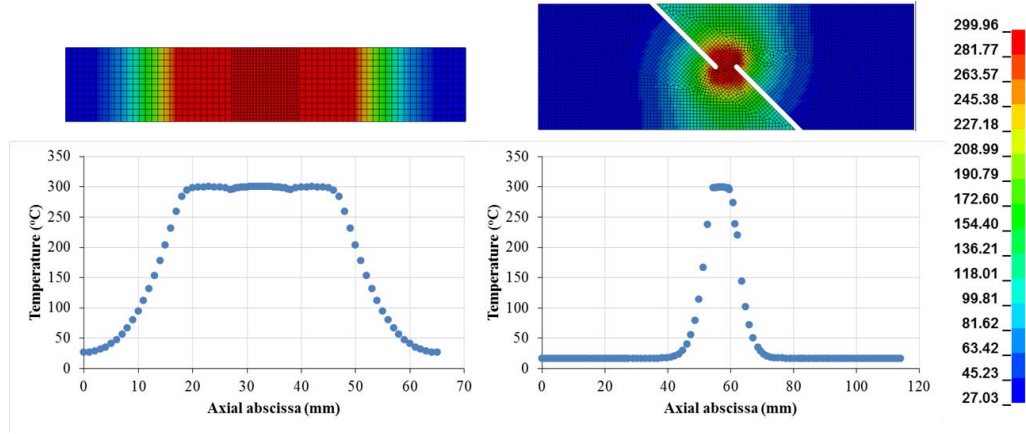


Figure 5.8 Example of temperature distribution at 300°C for the smooth specimen (on the left) and shear specimen (on the right).

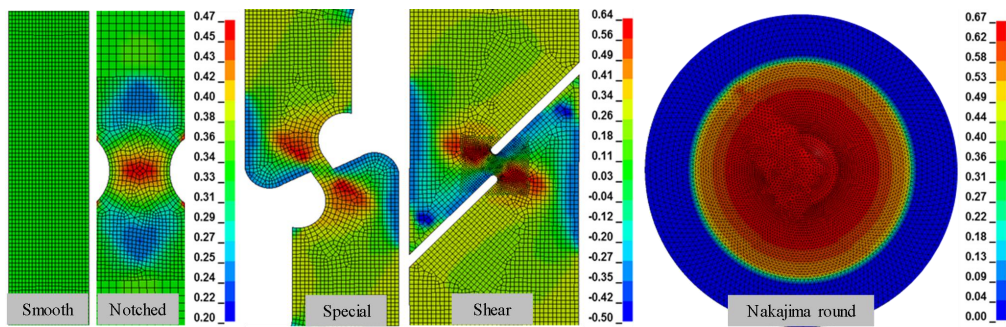


Figure 5.9 Stress triaxiality fields for all the specimen geometries at room temperature.

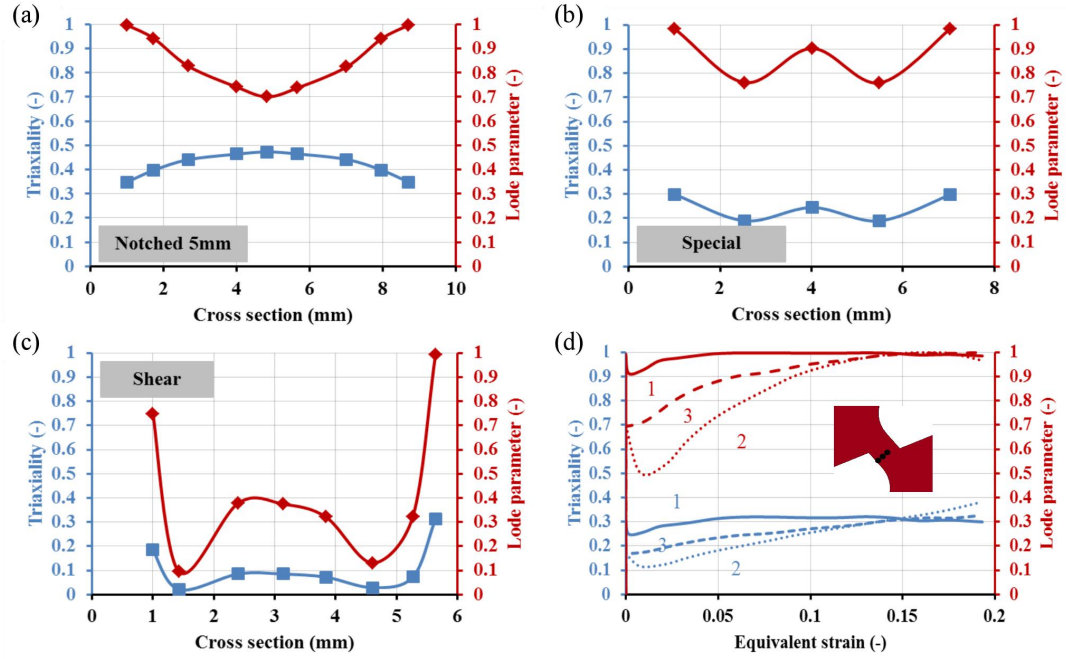


Figure 5.10 The distribution of triaxiality and lode parameter along the cross section of (a) notched, (b) special and (c) shear specimen at room temperature on rolling direction and (d) The evolutions of triaxiality and lode parameter as a function of equivalent strain at three position in cross-section.

Then, the effect of the rolling direction and temperature on the stress triaxiality was investigated. As example, the stress triaxiality evolution as a function of the equivalent plastic strain and temperature for the smooth specimen cut along the rolling direction is presented in Figure 5.11 (a). The stress triaxiality is almost constant at the beginning, regardless of the testing temperature, and then it starts increasing due to the strain localization. The temperature contributes to postpone the appearance of localized necking, since it appears later at increasing temperature. This implies that the stress state is no longer plane during necking with severe through-thickness stress gradients. Figure 5.11 (b) displays the stress triaxiality evolution of the different specimen geometries as a function of the equivalent plastic strain and rolling direction at room temperature, showing that the effect of the rolling direction is negligible. On this basis, the nominal stress triaxiality and Lode angle values for all the specimen geometries were used for the further analysis (see Table 5.1).

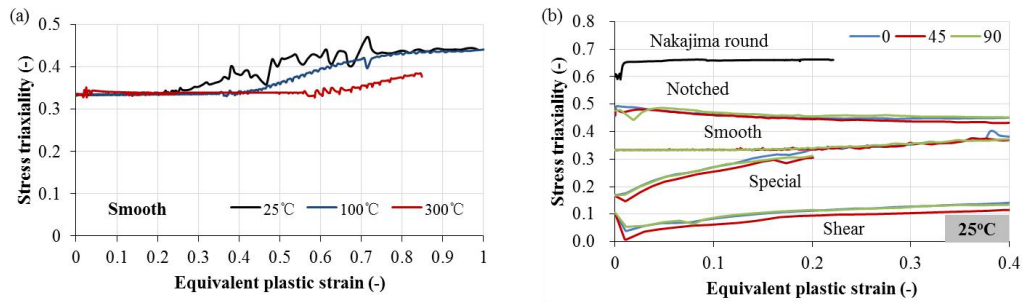


Figure 5.11 (a) Stress triaxiality as a function of the equivalent plastic strain and testing temperature for the smooth specimens cut along the rolling direction, and (b) stress triaxiality of different specimen geometries as a function of the equivalent plastic strain and rolling direction

at room temperature (The middle point of the minimum cross-section was selected for all the specimen geometries as the reference point to get the stress triaxiality values).

Table 5.1 Stress triaxiality and Lode angle values for the different specimen geometries.

	Stress triaxiality	Lode angle
Shear b831	0.1	0.29
Special	0.17	0.49
Smooth	0.33	0.99
Notched 5mm	0.47	0.51
Nakajima round	0.67	-0.73

5.2.3 Triaxiality effect on the mechanical behaviour

The mechanical response of the specimens of different geometry (Nakajima round specimen excluded) as a function of the rolling direction and temperature is shown in Figure 5.12. The triaxial loading affected both the maximum load and maximum stroke at fracture. As expected, the maximum load was shown by the smooth specimen while the minimum load by the shear specimen regardless of the rolling direction and temperature. The loading effect on the maximum stroke at fracture, namely the stroke at which the load drops, was again the one expected: it reduced gradually from smooth, special, notched to shear specimen regardless of the rolling direction and temperature. Interestingly, the maximum load shown by the special specimen was considerably lower than the one of the notched specimen even if their fracture strokes were almost the same. On the other hand, the temperature had an important influence on the maximum load, namely it decreased dramatically at increasing temperature for all the specimen geometries. As regards the stroke at fracture, the temperature had less effect in the case of the smooth specimen compared to the other geometries. In fact, the average stroke at fracture of the smooth specimen was around 12 mm, regardless of the testing temperature. On the contrary, the strokes at fracture of the other specimen geometries had an evident increment at increasing temperature, especially at 300°C. Therefore, it can be stated that the temperature influence on the fracture stroke depends on the specimen geometry. On the contrary, the rolling direction has less influence on the maximum load for all the specimen geometries at room temperature and 100°C, while the results are more scattered at 300°C due to a less stable thermal field. For smooth specimens, at room temperature and 100°C, the maximum load is a bit higher for the specimen cut at 90° with respect to the rolling direction than at 0° and 45°, whereas the differences are almost negligible at 300°C. Overall, the mechanical behaviour of specimens with different geometries is similar at room and 100°C regardless of the rolling direction.

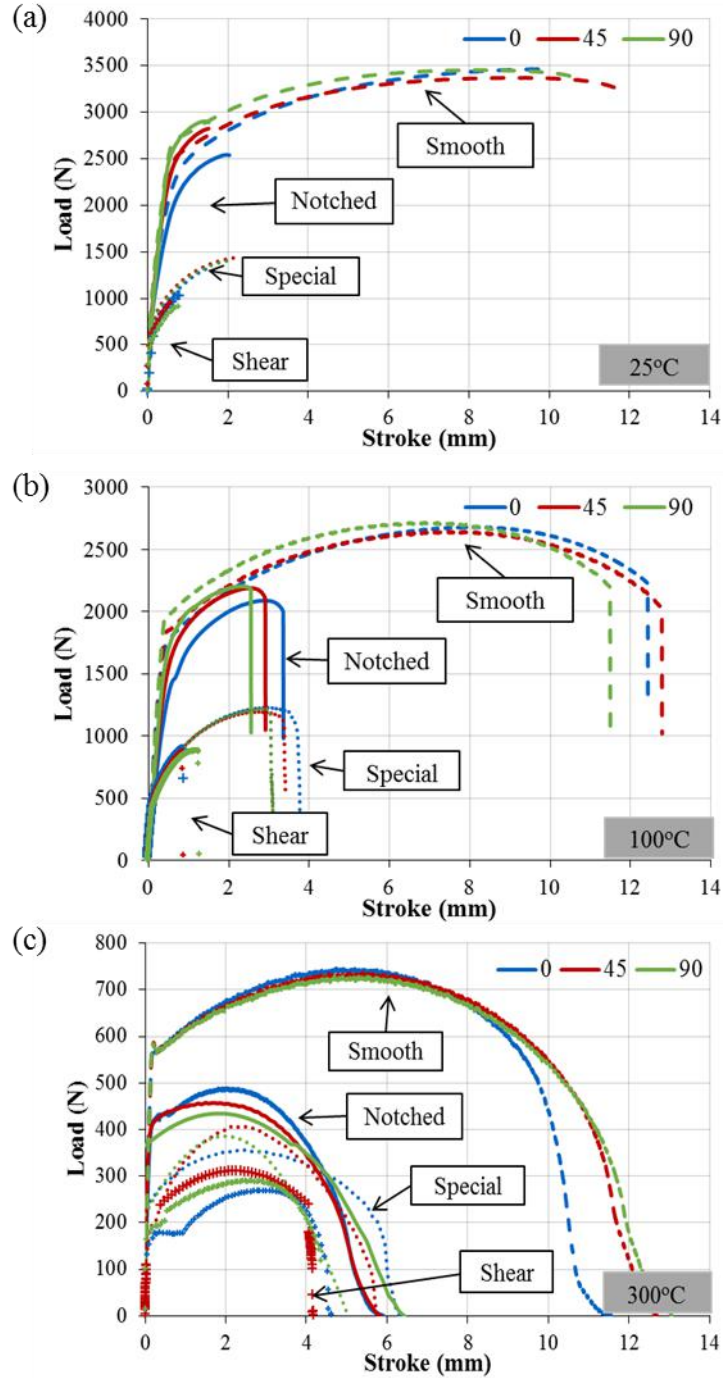


Figure 5.12 Experimental load-stroke curves as a function of the specimen geometry and rolling direction at: (a) room temperature, (b) 100°C and (c) 300°C.

5.2.4 Triaxiality effect on the plastic anisotropy

The anisotropy characteristics of the different specimen geometries were evaluated in terms of the Lankford coefficient, also called the normal anisotropy coefficient r , which is the ratio between the specimen width and thickness strains. In this study, the evolution of both the width and thickness strains along the cross section in the specimen deformed zone was obtained from the Aramis™ system. The r value was calculated along the longitudinal direction, and a steady state range between 0.15

and 0.2 of true strain was selected for its evaluation. As example, the evolution of the r values for the notched and special specimens at 100°C is reported in Figure 5.13: the r values change across the cross section of the notched specimen, reaching the minimum value in the middle (see Figure 5.13 (a)); conversely, the maximum r value for the special specimen appears in the middle of the cross section (see Figure 5.13 (b)). Nevertheless, the central point on the specimen minimum cross section was selected to have the r values for all the specimen geometries.

The evolution of the width strain (W) and thickness strain (T) under different triaxial loadings and rolling direction is depicted in Figure 5.14 (a-c). The width strain for the smooth and special specimens was higher than the thickness strain, regardless of the rolling direction. The opposite trend was shown by the notched specimen as shown in Figure 5.14 (b). As a consequence, the r value was lower than 1 for the notched specimens and higher than 1 for the smooth and special specimens. The same findings were in literature [66] for notched and smooth bars. The results of the shear specimens are not reported, since slight thickness strain variations caused not reliable increase of the r values. Along with the specimen geometry, the rolling direction had remarkable effects on the r values. For both smooth and notched specimens, the maximum width strain appeared at 90° with respect to the rolling direction, whereas appeared at 45° for the special specimen: this means necking started earlier for smooth and notched specimens cut at 90° with respect to the rolling direction whereas it started for 45° special specimens.

Furthermore, also from Figure 5.14 (a-c), the width strain was larger as well as the thickness strain was smaller at the same true strain point for all the specimen geometries; this trend can be explained by the volume constancy in the plastic strain regime: $\varepsilon_{true} = -(\varepsilon_w + \varepsilon_t)$. As a result, the r values as a function of the stress triaxiality and rolling direction at room temperature, 100°C and 300°C are reported in Figure 5.14 (d-f). It can be concluded that the r value was mainly influenced by the stress state as it decreased as the stress triaxiality increased for all the temperature conditions. The high triaxiality weakened the anisotropy characteristic of the AZ31 magnesium alloy. For the notched and smooth specimens, the r value at 90° was higher than at 0° and 45°; for the special specimen, the maximum r value was at 45° regardless of the testing temperature. This is consistent with the width strain trend reported in Figure 5.14 (a-c). Finally, as expected, the temperature increase induced a more isotropic behaviour of the specimens. At 300°C, the r values were closer to 1 compared to the other testing temperatures. It is worth noting that the r values of the special specimens were quite high, indicating that the thickness strain was much lower than the width strain (see Figure 5.14 (c)).

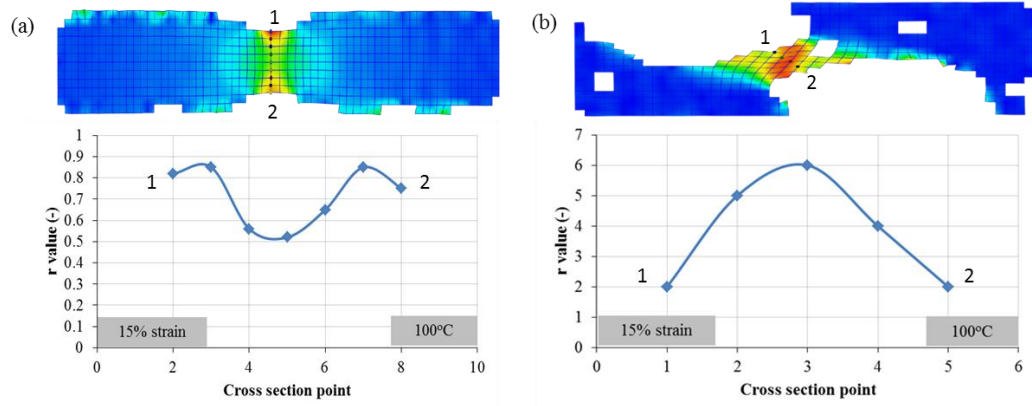


Figure 5.13 Evolution of the r value along the minimum cross section in case of: (a) notched and (b) special specimen at 100°C.

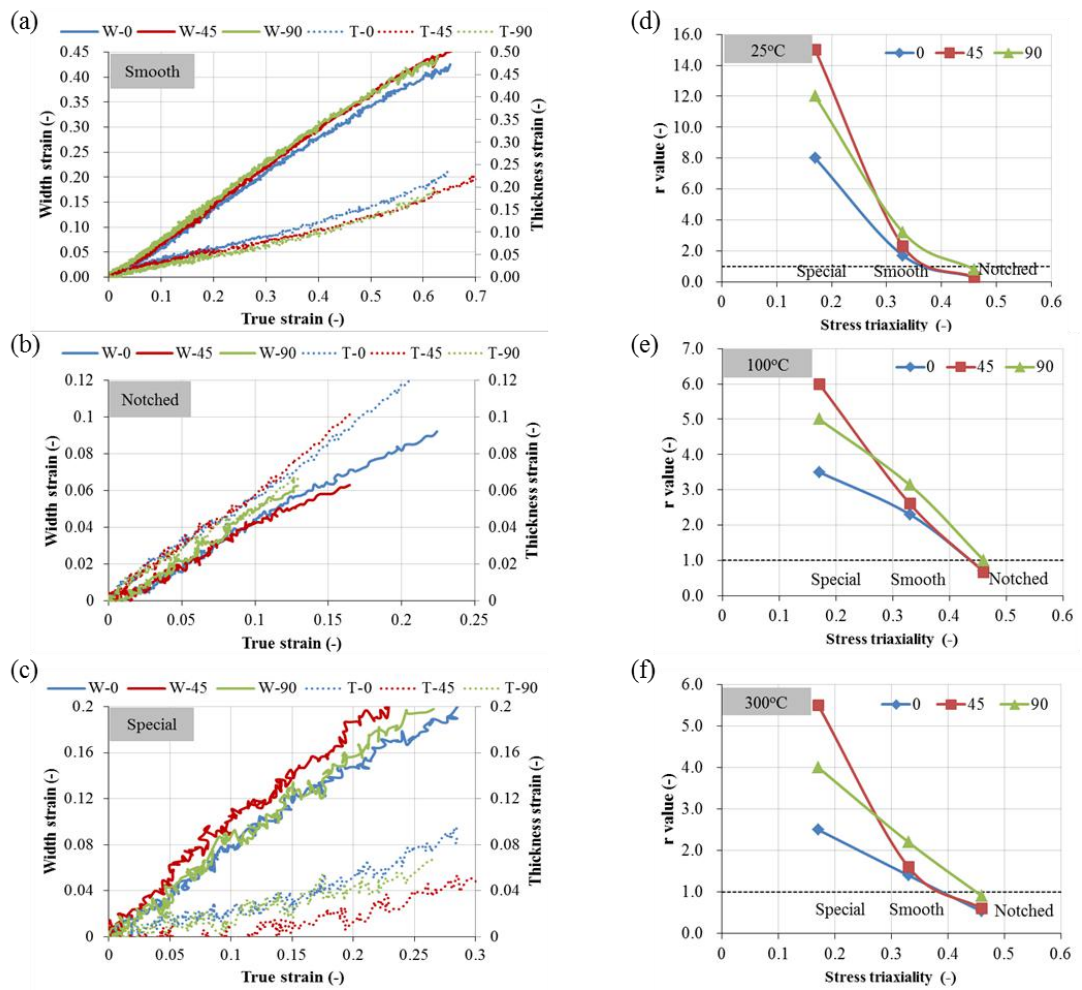


Figure 5.14 Width strain vs. true strain and thickness strain vs. true strain at varying rolling direction for: (a) smooth, (b) notched, and (c) special specimen at 100°C; r values as a function of the specimen geometry and rolling direction at: (d) room temperature, (e) 100°C, and (f) 300°C.

5.2.5 Triaxiality effect on the fracture strain

The AZ31 fracture loci as a function of the stress triaxiality and rolling direction at room temperature, 100°C and 300°C are shown in Figure 5.15. The results indicate that the fracture strain primarily depends on the loading condition and temperature, and secondly on the rolling direction. In terms of loading condition, the smooth specimens showed the highest fracture strain, followed by the special ones, while the round specimens showed the lowest ductility. These findings are true irrespective of the temperature and rolling direction. Fracture strain values are lower in shear loading condition than in uniaxial tensile one, which is somehow unexpected: this may be ascribed to the fact that shear strains measured microscopically near the fracture surface, resolved from grain rotations, can be much higher than the strains reported from DIC measurements that average the strains over a larger area. Nevertheless, similar results are also reported in [114] for the ZEK100-O magnesium alloy. Furthermore, also the Nakajima round specimens showed fracture strain values lower than expected, but again similar results can be found in [115].

In terms of rolling direction, at room temperature and 100°C, the fracture strain under the uniaxial tensile loading condition for smooth specimens cut at 45° with respect to the rolling direction was higher than the ones of the 0° and 90° specimens. While, under shear loading condition, the 90° specimen exhibited the highest ductility. The fracture strain sensitivity to the rolling direction in the case of special and notched specimens was instead negligible. At 300°C, the rolling direction had always a negligible effect on the fracture strain, regardless of the specimen geometry. On the contrary, the testing temperature had the highest effect on the fracture strain, with an increment of 50% from 100°C to 300°C for all the investigated loading conditions.

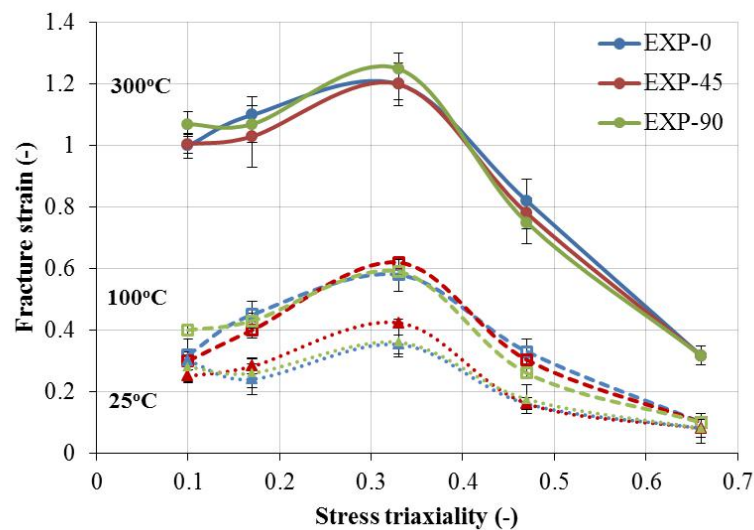


Figure 5.15 AZ31 fracture loci as a function of the stress triaxiality, rolling direction and temperature.

5.2.6 Fracture surfaces

Magnesium sheets are characterized by strong plastic anisotropy inherent to their crystalline structure (hexagonal close packed) with high c/a ratio. Figure 5.16 shows the overall appearance of the specimens after testing, while Figure 5.17 shows stereomicroscope images of representative surface fractures of notched and special specimens as a function of the testing temperature. All the fracture surfaces occur within the ductile mode since they appear slanted, hence, revealing that a considerable amount of deformation took place. From the figure, it can be seen that, in the case of notched specimens, the rolling direction did not affect the failure mode regardless of the testing temperature, whereas, in the case of special specimens, the fracture surfaces appeared very different from each other depending on the rolling direction. However, as can be expected and supported in literature [116], the temperature increase contributed to mitigate this effect: in fact, at 300°C, the surface fractures of the special specimens appeared all oriented to 45° with respect to the load application (see Figure 5.17 (l), (o), (r)).

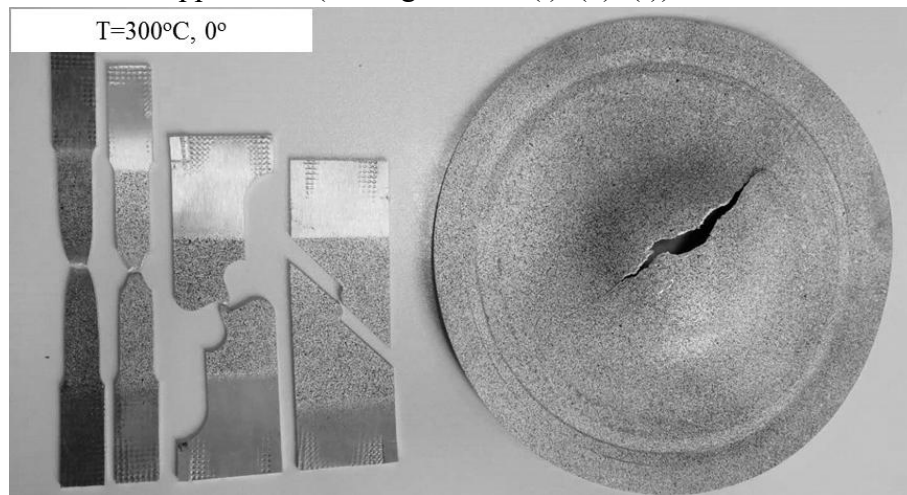


Figure 5.16 Fractured specimens after testing at 300°C.

The rolling process aligns the basal plane so that it is parallel to the sheet plane thanks to the easy slip that occurs on the basal plane. This explains the pronounced plastic anisotropy found at room temperature and the lower ductility of the 90° specimen. On the other hand, it is generally accepted that increasing the temperature generally reduces the anisotropy characteristics, as a consequence of the activation of temperature-dependent deformation mechanisms, such as Grain Boundary Sliding (GBS), and the increase in pyramidal slip activity. In their study, Barnett et al. [54] investigated the mechanical behavior of specimens with different grain sizes at different temperatures. They found that larger grain sizes lessened the contribution that such sliding made to the overall deformation. In [49], the authors found that at elevated temperatures, with a decrease of the Critical Resolved Shear Stress (CRSS) for no-basal slip systems, prismatic slip, and the first-order pyramidal slip system in a $\langle a \rangle$ slip direction as well as the pyramidal slip system in a $\langle c+a \rangle$ slip direction were activated to compatible strains on c -axis.

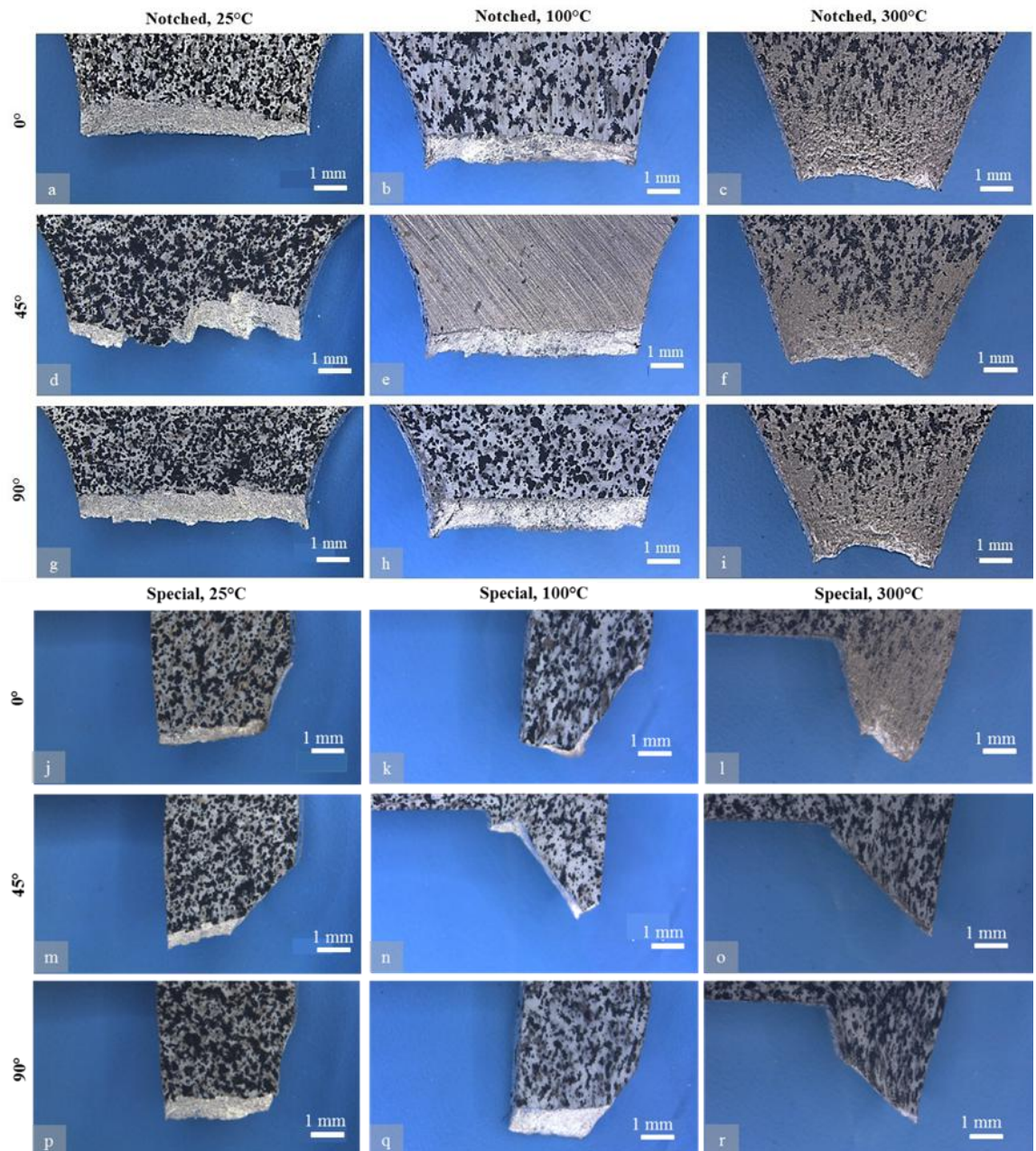


Figure 5.17 Stereomicroscope images of the surface fracture side.

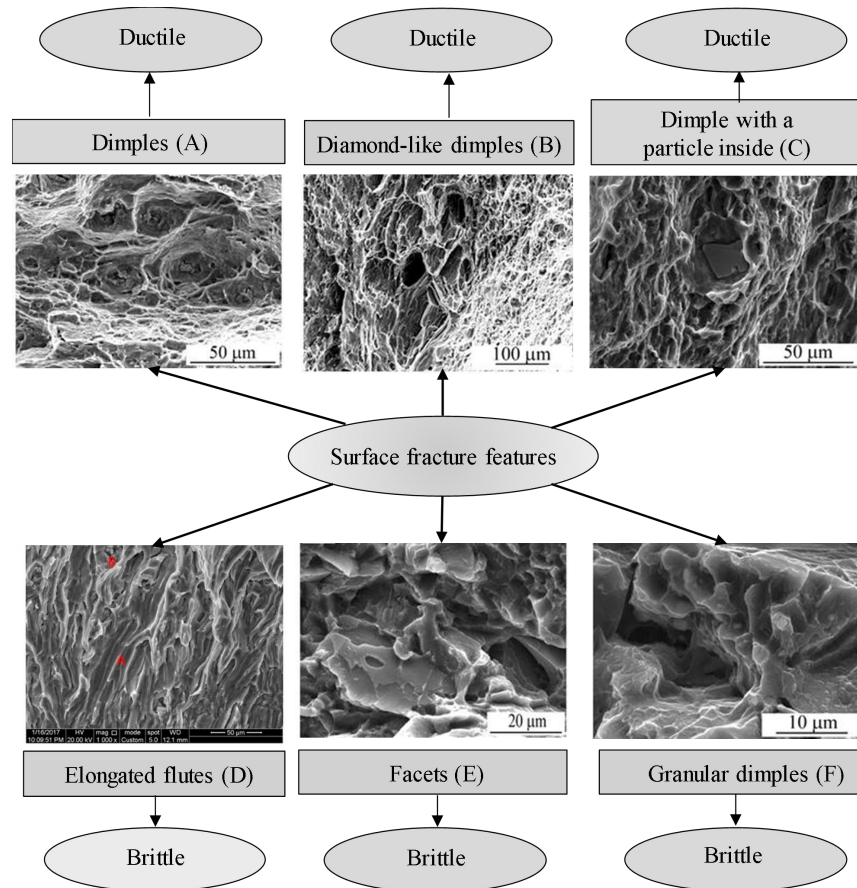


Figure 5.18 Summary of the main surface fracture features based on [66] and [118].

For completeness, surface fractures were also analyzed using SEM. Ductile and brittle type of fracture are distinguishable by their characteristic patterns. In [66], the effect of stress triaxiality on the failure mode of AZ31 magnesium alloy specimens of different geometries was studied by means of tensile tests at room temperature. It was found that both twins and second phase particles played a role in inducing fracture. In [118], the effect of notch acuity on the AZ31 fracture behavior was investigated by means of four points bending tests: it was found that the apparent toughness kept constant and reached the same value as obtained by fatigue pre-cracked specimens when a critical notch tip radius was achieved. Characteristic fracture features were extrapolated from these works, and summarized in Figure 5.18 to be used to recognize distinctive features of the surface fractures obtained in this study. It is worth nothing that different SEM magnifications are essential to catch all of the fracture possible features.

Bumpy pits, generally recognized as dimples, are commonly associated with ductile type of fracture. Both shape and morphology of dimples can provide additional information about the nature of fracture. Shallower dimples are usually associated to a more brittle type of fracture compared to deeper ones. Even characterized by a different shape, also diamond-like dimples are classified as indicator of ductile type of fracture [66]. On the contrary, features like elongated flutes and facets are usually correlated to quasi-brittle fracture. In fact, for most brittle crystalline materials, crack propagation corresponds to the successive and repeated breaking of

atomic bonds along specific crystallographic planes. Since the fracture cracks pass through the grains, this type of fracture is called transgranular (or transcrystalline). Macroscopically, the fracture surface is marked by faceted appearance as a consequence of changes in orientation of the cleavage planes from grain to grain. Usually, flat facets exhibit ‘river marking’ caused by crack moving through the crystal along parallel planes. Granular shaped dimples, which appear as the results of grain pull-out, are usually associated with intergranular crack propagation during brittle fracture. The shapes and sizes of the grains along the fracture path are easy to be observed.

On the basis of the summary shown in Figure 5.18 the same salient features were identified on the fractured specimens of the present study. To this purpose, Figure 5.19 and Figure 5.20 show the surface fracture morphology of the AZ31 specimens along the rolling direction for different specimen geometries and temperatures at different magnifications.

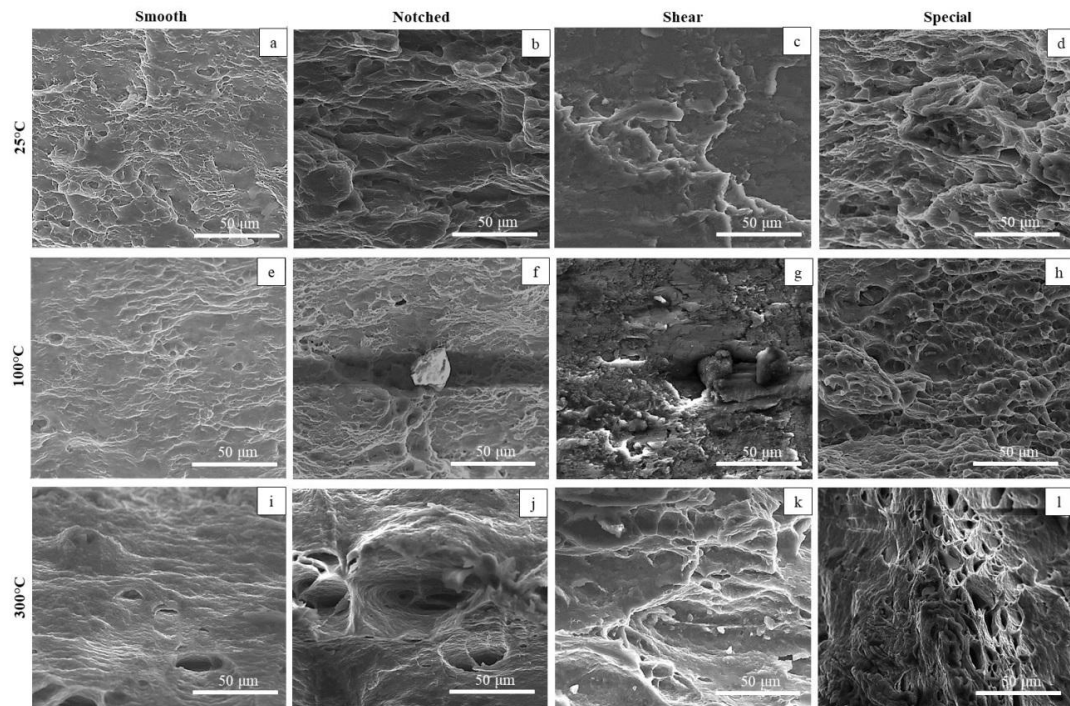


Figure 5.19 SEM images of the fractured surfaces of the specimens at 0° rolling direction (2000X magnification).

Several dimples were observed through the entire fractured surface of smooth specimens, indicating that ductile fracture had occurred (see Figure 5.19 (a)). Under the action of tensile stress, plastic deformation took place and voids were formed at the interfaces. As void evolution kept going, the holes became linked to each other and were consecutively strained until final failure occurred, creating lower size secondary dimples and leading to an assigned distribution of cavities. A completely different morphology characterized the fracture surface of the shear specimen, as reported in Figure 5.19 (c). A series of shallower dimples, stretched along the shear direction, are clearly recognizable. They are usually associated with low stress triaxiality conditions, which obstruct the growth of voids and therefore straitening ductility. Besides shear dimples, signs of intergranular fracture are evident. Figure

Figure 5.19 (b) and (d) report notched and special specimens that show a mixture of the previous type of fractures, thereby indicating that the matrix was subjected to both tensile stress and shear stress. In addition, other features, like elongated flutes and river patterns were also found on those fracture surfaces. In all of the reported figures, second phase particles are located at the center of shallow dimples or widespread inwardly them. This implies that, besides voids, also second phase particles played a role in the damage initiation process. The crack might have nucleated in correspondence of these inclusions, leading to particle/interface decohesion. On the basis of these observations, it can be concluded that the smooth specimen possesses the highest ductility, the shear ones the lowest while notched and special specimens lie in the middle. These findings are in accordance with the previously reported results of the mechanical tests (see Figure 5.12).

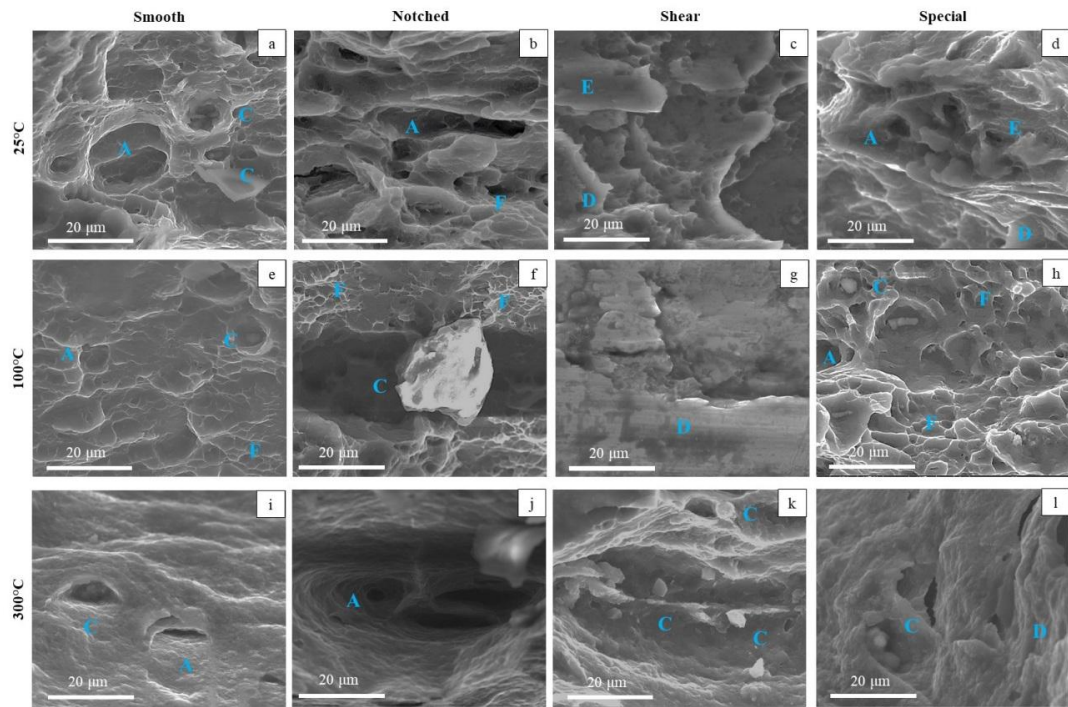


Figure 5.20 SEM images of the fractured surfaces of the specimens at 0° rolling direction (5000X magnification).

5.2.7 Modelling of the fracture locus

5.2.8 Proposed Modified Mohr-Coulomb (MMC) fracture criterion

The Modified Mohr-Coulomb (MMC) fracture criterion was applied to predict the fracture strain of AZ31 magnesium alloy sheets under various loading conditions, temperature and rolling direction. The fracture locus of the original MMC fracture criterion [84] is given by Equation (5.3):

$$\bar{\epsilon}_f = \left\{ \frac{A}{c_2} [1 - c_\eta (\eta - \eta_0)] \left[c_\theta^s + \frac{\sqrt{3}}{2 - \sqrt{3}} (c_\theta^{ax} - c_\theta^s) \left(\sec\left(\frac{\bar{\theta}\pi}{6}\right) - 1 \right) \right] \left[\sqrt{\frac{1 + c_1^2}{3}} \cos\left(\frac{\bar{\theta}\pi}{6}\right) + c_1 \left(\eta + \frac{1}{3} \sin\left(\frac{\bar{\theta}\pi}{6}\right) \right) \right] \right\}^{\frac{1}{n}} \quad (5.3)$$

$$c_{\theta}^{ax} = \begin{cases} 1 & \text{for } \bar{\theta} \geq 0 \\ c_{\theta}^c & \text{for } \bar{\theta} < 0 \end{cases} \quad (5.4)$$

where A is a material constant, n the strain hardening exponent, and c_{η} , η_0 , c_{θ}^s , and c_{θ}^c are parameters to describe both the pressure dependence and Lode angle dependence of the material plasticity. A total of eight material constants (A , n , c_{η} , η_0 , c_{θ}^s , c_{θ}^c , c_1 , c_2) need to be identified. If a von Mises yielding function is used ($c_{\eta} = 0$, $c_{\theta}^s = c_{\theta}^c = 1$), then Equation (5.3) reduced to Equation (5.5):

$$\bar{\varepsilon}_f = \left\{ \frac{A}{c_2} \left[\sqrt{\frac{1+c_1^2}{3}} \cos\left(\frac{\bar{\theta}\pi}{6}\right) + c_1 \left(\eta + \frac{1}{3} \sin\left(\frac{\bar{\theta}\pi}{6}\right) \right) \right] \right\}^{-\frac{1}{n}} \quad (5.5)$$

Here Equation (5.5) is simplified into Equation (5.6) by rearranging the material constants:

$$\bar{\varepsilon}_f = \left\{ c_1 \left[\sqrt{\frac{1+c_2^2}{3}} \cos\left(\frac{\bar{\theta}\pi}{6}\right) + c_2 \left(\eta + \frac{1}{3} \sin\left(\frac{\bar{\theta}\pi}{6}\right) \right) \right] \right\}^{c_3} \quad (5.6)$$

where η is the stress triaxiality, θ the Lode angle, and $\bar{\varepsilon}_f$ the fracture strain, and c_1 , c_2 , c_3 are material constants identified on the basis of the data obtained from the experimental tests. In the case of plane stress condition the stress triaxiality and Lode angle depend on each other, according to Equation (5.7). Hence, Equation (5.6) can be reduced to a single variable, namely the stress triaxiality.

$$\cos\left[\frac{\pi}{2}(1-\bar{\theta})\right] = -\frac{27}{2}\eta\left(\eta^2 - \frac{1}{3}\right) \quad (5.7)$$

It is worth underlining that the MMC fracture criterion was proposed for isotropic metals, and without taking into consideration any temperature effect. Therefore, the aim of this section is to introduce a new term into the MMC criterion to account for the temperature effect and apply it to different rolling orientations. A temperature-dependent parameter was introduced into the MMC fracture criterion: in particular, an exponential function of the temperature was added, which was the most suitable to describe the evolution of the fracture strain with temperature, as can be seen from Figure 5.21 where the experimental fracture strain of smooth specimens at varying rolling direction at elevated temperature was predicted making use of an exponential function of the temperature.

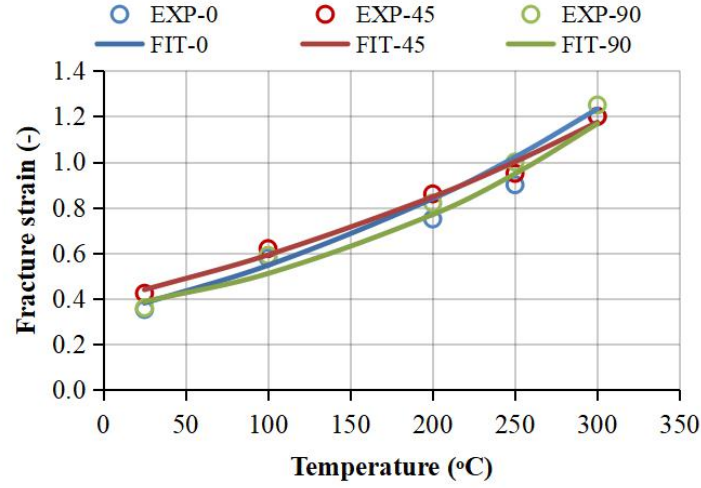


Figure 5.21 Experimental and predicted fracture strain values in case of smooth specimens as a function of the temperature and rolling direction using an exponential function of the temperature.

On this basis, the newly proposed MMC fracture criterion with a temperature-dependent parameter is represented by Equation (5.8):

$$\bar{\varepsilon}_f = \left\{ c_1 e^{mT} \left[\sqrt{\frac{1+c_2^2}{3}} \cos\left(\frac{\bar{\theta}\pi}{6}\right) + c_2 \left(\eta + \frac{1}{3} \sin\left(\frac{\bar{\theta}\pi}{6}\right) \right) \right] \right\}^{c_3} \quad (5.8)$$

Where e^{mT} is a temperature control term in which T (°C) is the testing temperature. The four material constants, namely c_1 , c_2 , c_3 and m , were identified at fixed rolling direction on the basis of the stress triaxiality, temperature and fracture strain values from the tests carried out on different specimen geometries. The material constants are given in Table 5.2 for each rolling direction. The comparison between the experimental and predicted fracture loci at varying temperature is shown in Figure 5.22 (a-c). However, if all the loading conditions are considered for the material constants identification, some discrepancies are evident regardless of the rolling direction. This is due to the fact that the fracture strain after an equi-biaxial state of stress represented by the Nakajima test on round specimens is lower for AZ31B alloy, which betrays the curve trend of the MMC model. To overcome this, the fracture criterion was calibrated again but without considering the Nakajima test results. According to literature [111], it is not unusual that phenomenological fracture models usually overestimate experimental data relative to fracture that occurs during Nakajima tests. The new predictions are shown in Figure 5.22 (d-f), where a much higher reliability of the fracture criterion to model the fracture loci is evident. This is also witnessed by the correlation coefficient R and the unbiased statistic parameter root mean square error RMSE whose values are reported in Figure 5.22 for the two sets of data used for the fracture criterion calibration.

Table 5.2 Material constants of the temperature-dependent MCC fracture criterion at varying rolling direction.

Identification using all the data	Identification without Nakajima data
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Rolling direction	0°	45°	90°	0°	45°	90°
c1	1.962	1.942	1.856	2.382	2.256	2.151
c2	0.181	0.169	0.187	0.120	0.112	0.143
c3	-7.705	-8.468	-10.214	-3.967	-4.721	-5.257
m	-0.001	-0.001	0.000	-0.001	-0.001	-0.001

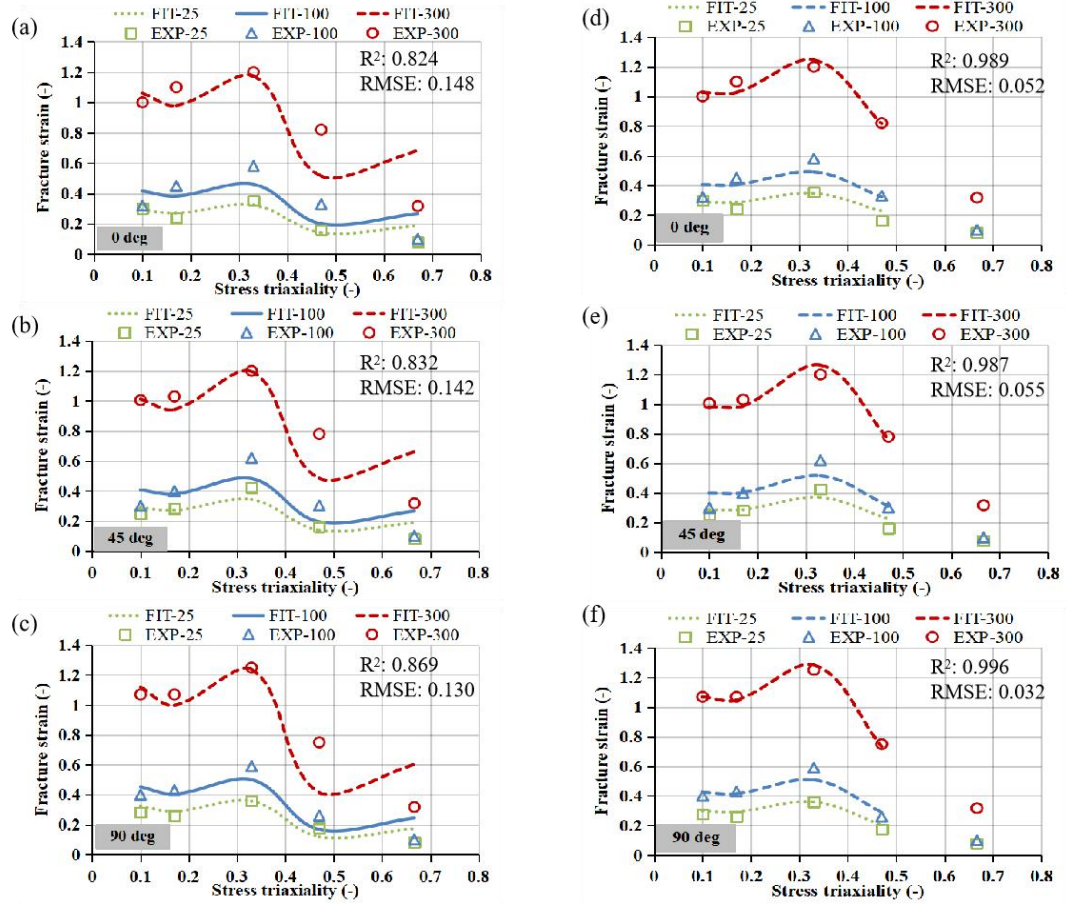


Figure 5.22 Comparison between experimental and predicted fracture loci using the newly proposed temperature-dependent MMC fracture criterion: (a-c) material constants identification using all the experimental data, and (d-f) without using Nakajima data.

5.2.9 Conclusions

Tensile, shear and Nakajima-type experiments were performed using different specimen geometries extracted from AZ31 magnesium alloy sheets to evaluate and model the material fracture behaviour at different temperatures, stress states and rolling directions. The main findings can be summarized as follows:

- The fracture strain values at room temperature and 100°C were measured by using the DIC method, whereas at 300°C, due to the extremely ductile nature of the fracture, a novel method was proposed making use of the load first derivative to identify the onset of fracture.
- The triaxial loading condition had a significant effect on the maximum load and fracture stroke regardless of the temperature and rolling direction. An increase

of temperature contributed to dramatically decrease the maximum load, while, on the contrary, the effect of the rolling direction was shown to be secondary.

- The anisotropy coefficient r decreased as stress triaxiality increased regardless of the temperature. Similarly, when the temperature raised, the anisotropy was weakened.
- The experimentally determined fracture loci indicated that the fracture strain was highly influenced by the loading condition and temperature, while the effect of the specimen rolling direction was secondary.
- Stereoscopy images of the overall fracture surfaces revealed always a ductile fracture mode.
- Magnified images of the fractured surfaces obtained by SEM showed that, besides ductile features, like dimples, also elongated flutes, facets and granular dimples were present, which are indicative of a certain brittle nature of the rupture. Therefore, the degree of ductility was different as a function of the adopted testing parameters.
- The classic MMC fracture criterion was modified by introducing a new term, namely an exponential function, for taking into consideration the effect of the temperature on the fracture strain.
- However, the reliability of the newly MMC fracture criterion is diminished when considering the bi-axial tension stress state.

Chapter

6 Conclusions and future works

In this chapter, the findings and results of the current research are summarized with final remarks and conclusions. The concluding remarks are given in order of appearance in the chapters.

The directions and recommendations of the future work are discussed briefly based on the results and observations presented in this dissertation.

6.1 Ti6Al4V titanium alloy

- The anisotropic characteristics of the Ti6Al4V sheets influenced the flow behaviour and plastic instability in a wide range of temperature from room temperature to 800°C. Ultimate Tensile Strength (UTS) decreased at increasing temperature regardless the specimen orientation. On the other hand, specimens oriented at 45° with respect to the rolling direction showed the lowest strength regardless the testing temperature. The strain hardening decreased at elevated temperatures, reaching values close to zero at 800°C. The strain rate sensitivity became higher, with an increase at increasing temperature, and the strain rate sensitivity - m values for the specimens oriented at 45° were always higher than the ones of the specimens oriented at 0° and 90°. The drastic increase of the strain rate sensitivity in the range of elevated temperatures was indicative of the occurrence of plastic instability in form of diffuse necking. The highest values of Lankford coefficients pertain to specimens cut at 45° with respect to the rolling direction, which show to provide the highest ductility, regardless of the testing temperature, whereas the 0° and 90° specimen oriented tend to be isotropic regardless the temperature. On the other hand, the trend of the 45° oriented specimens is pretty different, with an initial increase of the plastic anisotropy in the low temperature range and subsequent decrease at higher temperatures.
- The influence of anisotropy on the failure characteristics of Ti6Al4V sheets deformed at room temperature and 600°C. Tests using smooth, notched and shear specimens were carried out. The obtained results show that the Ti6Al4V sheets retained their anisotropic behaviour till 700°C, being both the strain at fracture and instability strain significantly influenced by the rolling direction at varying stress triaxiality; in particular, the fracture strain increased at decreasing stress triaxiality, the 45° direction always showed the highest ductility, except in case of shear specimens, both at room temperature and 600°C. whereas shear specimens oriented at 90° had the highest ductility, regardless the testing temperature. This means that the anisotropic effect has to

be taken into account when modelling damage evolution and fracture occurrence at varying stress state.

- The SEM analysis showed that the Ti6Al4V smooth and notched specimens had fracture surfaces characterized by deeper dimples and a higher density of secondary voids, proving that the higher the stress triaxiality the lower ductile the fracture and therefore the lower the fracture strain. This was observed regardless the testing temperature and specimen orientation. On the other hand, as expected, at 600°C, the dimples size became bigger, with an increased density of secondary voids, regardless the stress state and specimen orientation. Furthermore, the dimples in the specimen with the 45° orientation were bigger and deeper compared to the ones in the specimens with the 0° and 90° orientations, which was consistent with the related recorded values of the fracture strain. Again, this was evident regardless the testing temperature and stress state.
- A numerical model of the uniaxial tensile tests was developed and calibrated making use of the Barlat-Lian-1989 yield criterion and a hardening rule, which was adapted to take into account the anisotropic behaviour at different temperatures. It was proved that the developed model was capable of predicting the strain localization in the specimen gauge length due to plastic instability as well as its thickness distribution at varying temperature and strain rate.
- The numerical prediction of ductile failure of Ti6Al4V titanium alloy sheets when deformed at room and elevated temperature took into account the sheet anisotropic behaviour. The Barlat-Lian 1989 yield criterion and the GISSMO model were used, suitably calibrated on the basis of the extensive experimental campaign that included tensile tests on smooth and notched specimens, shear tests and Nakajima-type tests in order to have a wide range of stress triaxiality values. The tests were carried out on specimens machined at different rolling orientation to account for the material anisotropic behaviour. The numerical model was then validated on the basis of the outcomes of tests not included in the calibration phase.
 - The Barlat-Lian1989 yield criterion was calibrated on the basis of the Lankford coefficients, and the associated hardening rule on the basis of the material flow stress at varying temperature and strain rate, whereas the GISSMO model by means of the strain at fracture and instability strain values at varying stress triaxiality and rolling direction, at room temperature and 600°C.
 - The numerical models of the tensile and shear tests with implementations of the calibrated Barlat-Lian 1989 and GISSMO models proved to give a satisfactory agreement with the experimental outcomes, in terms of force vs. displacement curves, strain field, fracture location and characteristics; the same agreement was found in case of Nakajima-type tests in terms of dome

height and fracture location and characteristics, considering both the geometry used for calibration and the one for validation.

- The models were, in general, able to catch the differences in fracture occurrence due to the different anisotropic characteristics, with just a slight disagreement in case of 90° oriented specimens, especially under shear conditions, which need further investigations.

6.2 AZ31B magnesium alloy

- The anisotropy influence on the mechanical and microstructural characteristics of AZ31B sheets deformed at room and elevated temperature was evaluated on the basis of uniaxial tensile tests. The obtained results showed that the UTS values decreased at increasing temperature regardless of the specimen orientation. The UTS values were higher at 90° rolling direction than at 0° and 45° at room temperature and 100°C, whereas the differences between 200°C and 300°C was almost negligible. The Lankford coefficients-R values were always higher than 1 regardless of the testing temperature and specimen rolling direction, due to the fact that one of the main deformation mechanisms is the prismatic $\langle a \rangle$ slip resulting in greater strain in the width than in the thickness direction. The diffuse necking strain was higher at rolling direction with the maximum value shown at 100°C, whereas, the highest ductility was shown at 45° rolling orientation, regardless of the testing temperature. This proves the need to consider the anisotropy influence when modelling sheet forming processes carried out on AZ31B sheets even at elevated temperature.
- The influence of anisotropy on the failure characteristics of AZ31B sheets deformed at elevated temperature was evaluated using different specimen geometries cut at different orientations with respect to the sheet rolling direction. The triaxial loading condition had a significant effect on the maximum load and fracture stroke regardless of the temperature and rolling direction. An increase of temperature contributed to dramatically decrease the maximum load, while, on the contrary, the effect of the rolling direction was shown to be secondary. The anisotropy coefficient r decreased as stress triaxiality increased regardless of the temperature. Similarly, when the temperature raised, the anisotropy was weakened. The AZ31B fracture locus as a function of the stress state, temperature and rolling direction was constructed. The experimentally determined fracture loci indicated that the fracture strain was highly influenced by the loading condition and temperature, while the effect of the specimen rolling direction was secondary.
- The SEM analysis of AZ31B showed no significant differences in the microstructure at different specimen orientation, meaning that there was no anisotropy sensitivity on microstructure. Stereoscopy images of the overall fracture surfaces revealed always a ductile fracture mode. Magnified images of the fractured surfaces obtained by SEM showed that, besides ductile features,

like dimples, also elongated flutes, facets and granular dimples were present, which are indicative of a certain brittle nature of the rupture. Therefore, the degree of ductility was different as a function of the adopted testing parameters.

- Tensile, shear and Nakajima-type experiments were performed using different specimen geometries extracted from AZ31B magnesium alloy sheets to evaluate and model the material fracture behaviour at different temperatures, stress states and rolling directions.
 - The fracture strain values at room temperature and 100°C were measured by using the DIC method, whereas at 300°C, due to the extremely ductile nature of the fracture, a novel method was proposed making use of the load first derivative to identify the onset of fracture.
 - The classical MMC fracture criterion was modified by introducing a new term, namely an exponential function, for taking into consideration the effect of the temperature on the fracture strain. The newly proposed criterion was calibrated on the basis of the experimental data obtained from the tests previously mentioned, proving to be suitable to predict the AZ31B fracture locus for tension-dominated regions at varying temperature.

6.3 The summary for both titanium and magnesium alloys

The Ti6Al4V and AZ31B alloys are both HCP structures, There are many similarities and many differences, from macro mechanical characteristic and micro structure features.

- The UTS values decrease at increasing temperature regardless of the specimen orientation for both Ti6Al4V and AZ31B alloys. The UTS values around 1200MPa for Ti6Al4V which is much higher than AZ31B, 260Mpa.
- Specimens cut at 45° with respect to the rolling direction always show the highest ductility for both Ti6Al4V and AZ31B. However, the AZ31B shown a better ductile than Ti6Al4V regardless the specimen orientation and temperature.
- For Ti6Al4V, the highest values of Lankford coefficients pertain to specimens cut at 45° with respect to the rolling direction, which show to provide the highest ductility, regardless of the testing temperature. However, for AZ31B, the highest R value shown on 90° rolling direction. From the normal anisotropy, the anisotropic behaviour of Ti6Al4V pertains to temperature up to 600°C, which gradually reduces to disappear at 800°C. This behaviour can be ascribed to the increase of beta phase volume fraction at increasing temperature, which becomes significant at temperature above 600°C. Whereas AZ31B still keep a high normal anisotropy at 300°C. Which means much less deformation on the thickness direction than on the width direction, namely the thickness resistance is high to avoid localized thinning. On the contrary, the planar anisotropy is

close to 0, which means ears will not develop during deep drawing processes. The anisotropy effect on mechanical behaviour of Ti6Al4V is stronger than AZ31B.

- For Ti6Al4V, a significant increase of ductility is seen starting from 600°C, identifying this temperature as the one at which the benefit of elevated temperature forming in terms of enhanced formability can be exploited.
- The coupling of the Barlat-Lian 1989 anisotropic yield criterion and GISSMO damage model was introduced to predict the fracture occurrence and capture its anisotropic characteristics. The classic MMC fracture criterion was modified by introducing a new term, namely an exponential function, for taking into consideration the effect of the temperature on the fracture strain.

6.4 Future research directions

One of the future research work focus on the cryogenic tests which are being carried out on different specimen geometries of AZ31B alloy to investigate the anisotropy mechanical behaviour and fracture feature at temperatures below zero (-100°C and -50°C).

To do that, the uniaxial tensile tests will be carried out till fracture inside an environmental chamber, with the use of liquid nitrogen to cool down the specimen in case of low-temperature testing. The strain at uniform elongation and strain at fracture will be measured to evaluate the material ductility. Microstructural and surface fracture analyses will be carried out to explain the material behaviour at different temperatures. The experimental force and displacement curves of smooth specimen for different testing temperature from -100 to 300°C were shown in Figure 6.1

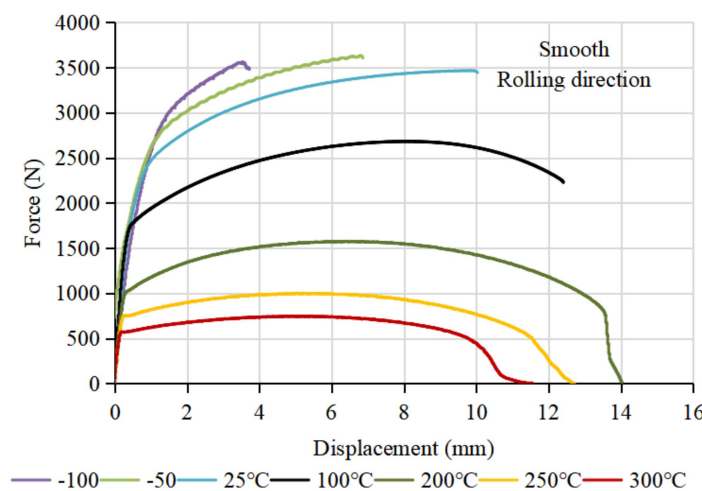


Figure 6.1 Experimental Force and displacement curves for different testing temperatures at strain rate $0.1s^{-1}$.

Another future research work will focus on the fracture model - temperature dependent MMC model proposed in this dissertation, the numerical simulation of the model is still need to be reach.

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