The shock and spall response of three industrially important hexagonal close-packed metals: magnesium, titanium and zirconium

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Magnesium, titanium and zirconium and their alloys are extensively used in industrial and military applications where they would be subjected to extreme environments of high stress and strain-rate loading. Their hexagonal close-packed (HCP) crystal lattice structures present interesting challenges for optimizing their mechanical response under such loading conditions. In this paper, we review how these materials respond to shock loading via plate-impact experiments. We also discuss the relationship between a heterogeneous and anisotropic microstructure, typical of HCP materials, and the directional dependency of the elastic limit and, in some cases, the strength prior to failure.

1. Introduction

Hexagonal close-packed (HCP) materials have seen increasing use in structural applications. This has mainly been owing to their high specific strength relative to traditional structural materials, but also owing to more specialized properties, such as good resistance to corrosion and a low absorption cross section for thermal neutrons.
This increased usage has led to a surge in research and development activities in the area of HCP materials for structural applications. Research has primarily been focused on three main elements and their alloys: magnesium, titanium and zirconium owing to their importance in the automotive [1–3], aerospace [4–8], defence [9–14] and nuclear industries [15].

As a result, the previous deficit in understanding of the mechanical behaviour of HCP materials relative to face-centred cubic (FCC) and body-centred cubic (BCC) materials has been drastically improved over the past 10 years [16]. Despite this, significant questions still remain—in particular with regards to how these materials behave under high strain-rate loading. In particular, there is a need to improve our understanding of the role of microstructure and crystallographic texture during shock loading conditions to further exploit this unique class of materials.

Shock loading of materials (where the material is subjected to a discontinuity of stress in a very small time scale) invariably occurs through plate impact, explosive loading or by pulsed laser. A good review of the various techniques available is provided by Field et al. [17]. Spall or dynamic fracture involves the generation of a plane of tensile damage within a dynamically loaded material; its effects can range from incipient spall where failure has just begun to break out to the creation of a detached spall plane. Measured spall strengths depend heavily upon loading conditions [18,19]. There are a number of techniques that can be used to monitor the evolution of spall inside materials. These include approaches to monitor the free surface of loaded systems, visual interferometer system for any reflector (VISAR), heterodyne velocimetry as well as using embedded manganin gauges [20]. However, because such approaches monitor spall remote from the site of its evolution, the signal from the spall plane will inevitably evolve before detection, making interpretation of spall complex [18].

It should be noted that the contributing factors for materials spalling have been studied for decades [21–27] and the phenomenon of spall was documented by Hopkinson [28]. Nevertheless, unlike FCC and BCC metals where the shock response has been reviewed [29,30], no such review exists for HCP metals. In this work, we focus primarily on three industrially important HCP metals to review how they respond to shock and spall—with the addition of some recent results and analysis.

2. Overview of hexagonal close-packed materials

(a) Crystal structure

The distinguishing properties of magnesium, titanium and zirconium are a result of the unique crystallography of the HCP lattice structure. To highlight the hexagonal low symmetry of the HCP lattice, it is often represented by a hexagonal prism, as shown by the dashed lines in figure 1. However, the hexagonal prism is not a true unit cell and consists of three primitive unit cells, as shown by the bold lines in figure 1. The primitive hexagonal cell is described by the axes $a_1 = a_2 \neq c$ and by the corresponding angles $\alpha = \beta = 90^\circ$ and $\gamma = 120^\circ$.

Assuming that atoms in the unit cell can be represented by hard spheres, the HCP crystal structure has a close-packed stacking sequence of ABAB, with an ideal $c/a$ ratio of 1.633. No pure metal has the ideal $c/a$ ratio. The lattice parameters and $c/a$ ratios of magnesium, titanium and zirconium are given in table 1.

A comprehensive review of the crystallography of the HCP lattice is given by Partridge [31].

(b) Plastic deformation

(i) Crystallographic slip

As with most metals, magnesium, titanium and zirconium plastically deform principally by crystallographic slip. The most commonly active slip systems observed in these materials are...
Figure 1. Primitive hexagonal unit cell (bold lines) and hexagonal prism (dashed lines).

Figure 2. Slip systems in HCP magnesium, titanium and zirconium.

Table 1. Lattice parameters for magnesium, titanium [5] and zirconium.

<table>
<thead>
<tr>
<th>material</th>
<th>$a$ (nm)</th>
<th>$c$ (nm)</th>
<th>$c/a$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Mg</td>
<td>0.321</td>
<td>0.521</td>
<td>1.624</td>
</tr>
<tr>
<td>Ti</td>
<td>0.295</td>
<td>0.468</td>
<td>1.586</td>
</tr>
<tr>
<td>Zr</td>
<td>0.323</td>
<td>0.515</td>
<td>1.593</td>
</tr>
</tbody>
</table>

Table 2. Slip systems in magnesium, titanium and zirconium.

<table>
<thead>
<tr>
<th>slip plane</th>
<th>slip direction</th>
<th>independent modes</th>
</tr>
</thead>
<tbody>
<tr>
<td>basal</td>
<td>$(0002)$</td>
<td>$(a)$</td>
</tr>
<tr>
<td>prismatic</td>
<td>$(10\bar{1}0)$</td>
<td>$(a)$</td>
</tr>
<tr>
<td>pyramidal</td>
<td>$(10\bar{1}1)$</td>
<td>$(a)$</td>
</tr>
<tr>
<td>pyramidal</td>
<td>$(1\bar{1}2\bar{2})$</td>
<td>$(c+a)$</td>
</tr>
</tbody>
</table>

listed in table 2, with a diagrammatic representation of the slip systems relative to the HCP primitive unit cell and the hexagonal prism given in figure 2.

Broadly, the HCP lattice structure accommodates the easy glide of dislocations along the basal, prismatic and pyramidal planes in the $(1\bar{1}2\bar{0})$ directions, commonly referred to as $(a)$ type slip systems, as shown in figure 2. However, it has been shown that to accommodate compatible plastic strains in a polycrystalline metal, a total of five independent slip systems are required [32,33]. Owing to the three easily activated $(a)$ type slip systems having the same slip direction, they reduce to only four independent slip systems. Thus, to meet the criteria for compatible plastic
strains, a further non-\((a)\) type slip system must be activated. Typically, this is accommodated by a so-called \((c+a)\) type slip system, as shown in figure 2. However, it has been argued that incompatible plastic strains in polycrystalline HCP metals can also be accommodated by the activation of deformation twinning. Moreover, for some combinations of the loading axis and crystallographic direction, twinning is the preferred mode for plastic deformation.

Crystallographic slip is driven by the shear stresses acting on a given slip plane. The shear stress required to activate a given slip system is referred to as the critically resolved shear stress (CRSS). The CRSS required to activate the different slip systems in magnesium, titanium and zirconium varies significantly depending on the temperature and composition of the material. However, generally, the CRSS of the \((c+a)\) type slip system is higher than that of the \((a)\) type slip systems, leading to anisotropic plastic deformation behaviour at the single-crystal level. It is well established that increasing the strain rate increases the CRSS by decreasing the time for a given dislocation to overcome a barrier to its motion. At extremely high strain rates, the glide of dislocations is further inhibited by mechanisms such as phonon drag [34,35].

(ii) Deformation twinning

Owing to the limited number of slip systems available in the HCP lattice, deformation twinning plays an important role in plastic deformation, and subsequently the ductility, of HCP materials [36,37]. Deformation twinning results in a reorientation of the crystal lattice around a given plane and in a given direction. This reorientation accommodates plastic strain and, more importantly, typically reorientates the crystal lattice for easier crystallographic slip.

The most commonly active deformation twinning systems observed in magnesium, titanium and zirconium are listed in table 3, with a diagrammatic representation of example twinning systems relative to the HCP primitive unit cell and the hexagonal prism given in figure 3.

Deformation twinning is particularly important at high strain rates, where it has been observed that twinning becomes more prevalent [38].

Table 3. Twinning systems in magnesium, titanium and zirconium.

<table>
<thead>
<tr>
<th>Type</th>
<th>(K_1) plane</th>
<th>(\eta_1) direction</th>
<th>(K_2) plane</th>
<th>(\eta_2) direction</th>
</tr>
</thead>
<tbody>
<tr>
<td>Tension I</td>
<td>{10\bar{1}2}</td>
<td>{10\bar{1}1}</td>
<td>{10\bar{1}2}</td>
<td>{10\bar{1}1}</td>
</tr>
<tr>
<td>Tension II</td>
<td>{1\bar{1}2\bar{1}1}</td>
<td>{\bar{1}1\bar{2}6}</td>
<td>{0002}</td>
<td>{1\bar{1}20}</td>
</tr>
<tr>
<td>Compression I</td>
<td>{1\bar{1}2\bar{2}}</td>
<td>{\bar{1}1\bar{2}3}</td>
<td>{11\bar{2}4}</td>
<td>{22\bar{4}3}</td>
</tr>
<tr>
<td>Compression II</td>
<td>{10\bar{1}0}</td>
<td>{10\bar{1}2}</td>
<td>{10\bar{1}3}</td>
<td>{30\bar{3}2}</td>
</tr>
</tbody>
</table>

Figure 3. Twinning systems in Zr, Ti and Mg. (a) \{10\bar{1}2\} \(\bar{10}\bar{1}1\) tensile twin and (b) \{1\bar{1}2\\bar{2}\} \{\bar{1}1\bar{2}3\} compressive twin.
3. Magnesium

Compared with other metals such as copper and aluminium, where there is a wealth of data on their shock behaviour, magnesium and its alloys have received relatively little attention. This is despite their extensive application in the defence and aerospace industries [3,8].

(a) Shock response of magnesium

Early work on the elastic–plastic response of a magnesium alloy was carried out by Fuller & Price [39]. They studied the shock behaviour of several metals including ZW3—a high strength magnesium extrusion alloy (Zn 3.0, Zr 0.6—wt%). They repeatedly observed a relaxation in stress behind the shock front prior to the arrival of the release wave that they attributed to slip only occurring on the basal plane. They reported a Hugoniot elastic limit (HEL) of 0.24 GPa for this alloy.

Figure 4 shows the shock traces of a Mg alloy (Elektron 675—an alloy based on the magnesium–yttrium–gadolinium ternary system) that has been subjected to shock loading to a stress of approximately 2.9 GPa. The experiments were conducted by launching 10 mm thick Al plates at 500 ± 30 m s⁻¹ that impacted the Mg specimens. Here, the shock stress is recorded by means of a calibrated manganin gauge (type LM-SS-025CH-048), electrically insulated by 25 μm of Mylar, and sandwiched between the specimen and a plate of poly(methyl methacrylate) (PMMA).

On the left-hand side, the elastic precursor is seen to diminish as the thickness of the specimen is increased, whereas on the right-hand side of the trace, it is seen that the magnitude of the elastic precursor is maintained for 12 mm. Notably, a ‘hump’ in the trace is visible as the elastic wave is reflected back off the plastic shock wave [40]. Precursor decay has also been observed in annealed bar-stock magnesium alloy such as magnesium MA2–1 [41]; indicative signs of precursor decay have also been seen in the alloy AZ61 [42].

The principal difference between these two sets of results is the processing of each material. The four traces in the left-hand side were for an alloy that has not undergone heat treatment and is in the F condition (as fabricated) by hot extrusion. Whereas the alloy on the right has been...
artificially aged (T5) leading to the precipitation of alloying elements. These precipitates act to pin dislocations. Similar comparisons in behaviour have also been seen when comparing relatively pure forms of aluminium to that of alloys where the alloys such as precipitation hardened AA 6061-T6 show little-to-no precursor decay [43,44], whereas relatively pure forms of aluminium show moderate decay [44,45]. Further, Winey et al. [46] showed that thin (less than 600 μm) 1050 Al samples exhibited attenuation of the elastic precursor, whereas the AA 6061-T6 targets did not. As discussed by Gray III [47], this Mg alloy compares in a similar manner to the previously detailed work and it appears that a coherent conclusion can be garnered on its behaviour. That is, the ageing and consequential precipitation hardening that increases the strength of the alloy also hinders precursor decay. That is, hindering dislocation source activation and/or evolution reduces the propensity for the elastic precursor to attenuate.

Owing to its relatively low melting point (649°C), a modest increase in testing temperature has also been shown to affect the shock response of Mg, primarily the elastic precursor [48]. For instance, large values of the elastic precursor wave were observed at temperatures approaching the melting point. This was also seen in Al [48]. This was attributed to the Frenkel disorder [49], i.e. spontaneous nucleation of point defects at the elevated temperatures from their normal lattice positions to interstitial positions. Theoretically, the end result is increased resistance to activation and motion of dislocations. However, it is not clear that this process would occur quickly enough to result in the observed strengthening during shock loading. Similar behaviour has been seen in heated single crystals of aluminium where the cause of the strength increase was attributed to phonon drag on dislocations [50]. It has also been seen in stainless steels [51]. In this case, it was thought to be owing to the formation of relatively hard intermetallics at the grain boundaries.

(b) Spall response of magnesium

Examination of the spall behaviour of various magnesium alloys has been of interest to a number of researchers [52,53]—particularly at elevated temperatures [48,54,55]. Schmidt et al. [56] measured the spall strength of AZ31B-H24 and reported a value of 1.5 GPa for the onset of incipient spall at room temperature. They reported a rapid drop in spall strength as the internal energy was increased and approached 0.2 GPa at incipient melt. Similarly, the spall strength of Mg95 (99.95 wt% Mg) has been studied by Kanel et al. [48] at various initial temperatures. In this work, they showed that the spall strength dropped off as the temperature approached the melting point of the metal. Similar behaviour has been observed in magnesium alloys [41,57]. Ordinarily, this might appear unsurprising. However, it is reported that when various metals have been shocked to a high stress, and where the temperature increase approaches melt for the material (owing to shock heating), the spall strength remains unaffected [48].

Limited studies have been carried out on the microstructural effects on the shock and spall behaviour of magnesium alloys. One of the more promising types of magnesium alloys for structural applications are the rare-earth alloys. However, processing of rare-earth Mg alloys generally results in a heterogeneous and anisotropic microstructure. For instance, figure 5c shows the microstructure of the Mg alloy Elektron 675-T5 where striations of relatively small grains (indicated by the arrows) are observed. These striations can affect both the spall strength and the HEL [58]. In a recent study, Mg alloy samples possessing the microstructure depicted in figure 5c were shocked either along the extrusion direction or perpendicular to it. Figure 5a shows the effect of the microstructural features of the alloy on its shock response; both the HEL and the spall strengths are lower in samples that were shocked perpendicular to the extrusion direction. Notably, their reload of the spall signal from the dip also appears sharper. This is indicative of a faster rate of damage accumulation, presumably owing to a separation between the harder smaller grains and softer larger grains. Similar to these findings, Rinehart in his seminal 1964 paper [22] discussed the spalling of materials and structures such as laminates, highly textured or rolled plates. In these cases, the heterogeneity acts as planes of weakness that can be separated during a tensile pulse.
Figure 5. (a) Free surface velocity response of two Mg-alloy (T5 temper) targets shocked along the extrusion direction (T5-F3) and perpendicular to it (T5-F1), (b) location of where plate-impact samples originated and (c) orientation map of the virgin microstructure indicating the columns of small grains. (Online version in colour.)

4. Titanium

Titanium alloys have been extensively used in aerospace applications owing to their high specific strength and good resistance to corrosion. Despite these favourable properties, titanium alloys have found limited use as monolithic armour materials, with high production costs and a propensity to fail owing to adiabatic shear banding/spall being limiting factors to their adoption. However, new low-cost variants of traditional titanium alloys and advances in low-cost production methods have resulted in renewed interest in titanium alloys for armour applications.

(a) Shock response of titanium

The equation of state for the most widely used titanium alloy, Ti–6Al–4V, has been well established by numerous studies using longitudinal stress gauges, with the HEL shown to vary between 2.1 and 2.8 GPa [59–61]. It is well known that oxygen content significantly increases the yield strength of titanium alloys owing to the formation of coherent $\alpha_2$ particles [5,62]. Razorenov et al. [63] have studied the shock behaviour of three Ti–6Al–4V alloys of different oxygen contents and measured a marked increase in the HEL with increasing oxygen content; the spall strength appeared largely unaffected by increasing the oxygen content from 0.105% up to 0.24%. Moreover, Ren et al. [64] tested an extra-low interstitial Ti–6Al–4V and found that the HEL was 1.7 GPa, significantly lower than the previous results for Ti–6Al–4V. Again, it was concluded that this was due to the low oxygen content.

Hopkins and Brar inserted manganin gauges into sectioned discs of Ti–6Al–4V to investigate the shear response of the material. They found evidence of shear-hardening with increasing shock stress [61]. Whereas Arrieta & Espinosa [65] investigated the effect of temperature on the HEL and showed that the HEL decreased with increasing temperature; similar results were found by Kruger et al. [66] for the titanium alloy Ti–6–22–22S.

Millett et al. [67] studied the effects of crystallographic texture on the shock response of Ti–6Al–4V by loading material with a strong texture in the longitudinal and radial directions. No variation in the equation of state was found between the two conditions, but it was shown that
the shear strength (again, by using embedded manganin gauges in sectioned targets) was lower in the radial direction which was attributed to the orientation of the HCP unit cell relative to the loading direction.

Titanium and its alloys have been shown to undergo phase transformations during shock loading [53]. At high pressures, the HCP \( \alpha \)-phase of titanium undergoes a martensitic phase transformation to a new HCP structure, known as the \( \omega \) phase, with a \( c/a \) ratio of approximately 0.61 [68]. Owing to a large hysteresis, the \( \omega \) phase is retained after the high-pressure loading is removed [69]. This phase transformation was first observed by Jamieson [70] and has since been extensively studied under various high-pressure conditions [68] and alloy compositions [71,72]. During shock loading of high-purity titanium, Razorenov et al. [73] noted an anomalous shock wave structure at approximately 2–5 GPa, which was attributed to the \( \alpha-\omega \) phase transition. Arrieta & Espinosa [65] investigated the effect of temperature on the \( \alpha-\omega \) phase transition in Ti–6Al–4V and found that increasing temperature suppresses the transition, but that increasing strain rate had the opposite effect. Indications of phase transformations have also been seen in Ti–6Al–4V by Rosenberg et al. [60] and in \( \alpha \)-Ti by Christman et al. [74] at around 10 GPa. More recently, Cerreta et al. [75] studied the role of oxygen content in the \( \alpha-\omega \) phase transition. It was found that the \( \alpha-\omega \) phase transition occurred at 10.4 GPa in the high-purity titanium, whereas the addition of 3700 ppm of oxygen completely suppressed the onset of the \( \alpha-\omega \) phase transition.

(b) Spall response of titanium

The spall strength of Ti–6Al–4V has been studied by Dandekar & Spletzer [76], Church et al. [77] and Tyler et al. [78], all of which report a marked increase in spall strength with increasing pulse duration, indicating significant microstructural and/or dislocation structure evolution during the initial compressive loading stage.

Me-Bar et al. [20] investigated the spall strength of two Ti–6Al–4V materials with a well-developed bimodal microstructure and a lamellar microstructure. Post-experiment microscopy of recovered specimens showed that spall occurred by the nucleation, growth and coalescence of voids parallel to the main spall plane. It was suggested that they had nucleated simultaneously, owing to the similar size of the voids, with nucleation occurring at \( \alpha \) grain boundaries in the bimodal material and at the prior \( \beta \) grain boundaries in the lamellar material. It was suggested that these voids grow and coalesce into facets, connected by intergranular and transgranular shear or cleavage cracks. Microstructural analysis of recovered specimens by Tyler et al. [78] showed that voids seemed to nucleate at the interface between the \( \alpha \) and \( \beta \) phase. Boidin et al. [79] also observed that voids nucleated at the interface between the \( \alpha \) and \( \beta \) phase in Ti–6Al–4V with a lamella-based microstructure. The voids were observed to coalesce owing to highly localized plastic bridging between facets. Based on these observations, Boidin et al. proposed a two-stage fracture model consisting of an initial quasi-brittle failure mode (void nucleation), followed by a ductile failure mode (void growth and coalescence). Similarly, Arrieta & Espinosa [65] investigated the effects of temperature on the spall behaviour of Ti–6Al–4V, where it was shown that the spall strength reduced as the temperature increased. Post-impact microscopy on recovered specimens once again showed that spall occurred via the nucleation of voids at the boundaries between \( \alpha \) grains and propagated via coalescence then bridging.

McDonald et al. [80] probed the three-dimensional nature of spall in Ti–6Al–4V using X-ray microtomography and showed that the volume fraction of voids/cracks increased with increasing shock stress, but that the spall strength was effectively strain-rate independent.

As noted previously, crystallographic slip in HCP materials at a single-crystal (or grain-by-grain) level is highly anisotropic, owing to \( (c+a) \) type slip having a higher CRSS relative to \( (a) \) type slip. This results in the ease of activation of plastic deformation in a given grain being highly dependent on its orientation relative to the loading conditions. Recently, Wielewski et al. [81] investigated the role of microtexture in the activation of
spall in Ti–6Al–4V by electron backscatter diffraction analysis of recovered specimens from plate-impact experiments. Interestingly, it was found that voids nucleated at the interface between plastically hard (high Taylor factor) and soft (low Taylor factor) \( \alpha \) grains, shown in figure 6a, b. Further, it was shown that the hard/soft grain interaction sites could be found at the corners of a facet that had formed via the nucleation and coalescence of voids, as shown by the circles in figure 6d. This closely mirrors the results of Dunne et al. [82,83] and Rugg et al. [84] under fatigue loading conditions, where it was shown that facets nucleate at boundaries between plastically hard and soft \( \alpha \) grains. This failure mechanism is yet to be probed in other HCP metals. Notably, no twins were evident in the recovered plate-impact specimens, which was contrary to observations of the same material loaded in a Taylor test [38].

5. Zirconium

Owing to its relatively high ductility, good corrosion resistance and low absorption of neutrons, \( \text{Zr} \) has received considerable attention from a variety of industries. Thus, the number of studies to understand the mechanical response of \( \text{Zr} \) subjected to various loading conditions is the most prolific out of the three materials discussed herein.
(a) Shock response of zirconium

One of the primary areas of research in Zr subjected to shock loading is the investigation of the complex series of phase transformations with increasing temperature or pressure/stress. It is known that Zr moves from the HCP $\alpha$ phase through a hexagonal with a three atom basis/hex-3 $\omega$ phase to the BCC $\beta$ phase at elevated pressures [85–87]. Greeff [86] established a high strain-rate phase diagram and Hugoniot equation of state which clearly illustrated the transition through these three phases. In particular, under shock loading conditions, zirconium has been observed to undergo an $\alpha$–$\omega$ phase transformation across a range of pressures from 2.3 to 8.5 GPa. This phase transformation has also been shown to display a hysteretic behaviour. That is to say, high fractions of metastable $\omega$ phase are retained in specimens that were shock-loaded to stresses in excess of the threshold for phase changes and then unloaded to ambient pressure.

Cerreta et al. [87] conducted a series of one-dimensional plate-impact experiments using a single-stage gas-gun to impact Zr samples of 3 mm thickness. Via choice of one high-purity (35 ppm Hf; less than 50 ppm O) and two low-purity (350/14 000 ppm Hf and 390/1200 ppm O, respectively) metals, they were able to investigate the effects of interstitial oxygen on the shock response of Zr. They found that the $\alpha$–$\omega$ phase transition increased from 7.1 to 8.3 GPa as the oxygen content increased from less than 50 to 350 ppm, with this transition entirely inhibited for an oxygen content of 14 000 ppm. This response was attributed to the fact that the oxygen atoms were located at octahedral sites within the HCP lattice rather than the tetragonal sites. Consequently, the presence of these large interstitial oxygen atoms displaced space within the HCP $\alpha$-Zr lattice required for the $\alpha$–$\omega$ phase transformation. Recent work by Rigg et al. [85] used plate impact as well as a Z pinch machine to quasi-isentropically load similar Zr materials as those used by Cerreta et al. The $\alpha$–$\omega$ phase transition was again observed to be heavily dependent on impurity content, with no difference observed with regards to the stress at which this transition occurred with sample thickness. However, it was observed that the 1200 ppm O grade of Zr showed a much more pronounced phase-change signal using the Z pinch machine than where data were gained by the plate-impact experiments.

In a more complex shock loading condition, Martinez et al. [88] and Escobedo et al. [89] studied the effects of high strain rates on a highly textured Zr via a series of impact experiments using a modified Taylor cylinder system. Zr bullet-shaped projectiles were accelerated at velocities ranging from 400 to 600 m s$^{-1}$ and impacted into a steel anvil with a tapered central section, which extruded the incident projectiles, inputting peak stresses of ca 3.5 GPa. Post-mortem examination of recovered specimens showed that deformation occurred by a combination of slip and twinning, with the precise sequence of deformation mechanisms dictated by the initial texture. The end result was a larger ductility exhibited by the material whose texture had the higher number of readily available slip systems.

(b) Spall response of zirconium

In one of the most complete works available, Gray III et al. [90] have recently studied the effect of both crystallographic and microstructural anisotropy on spall in a variety of BCC and HCP materials. In particular, they undertook a study looking at incipient spall in crystalline Zr plate (clock rolled and annealed to produce an equiaxed ca 25 $\mu$m grain size with a strong basal texture along the through-thickness direction of the plate). The Zr chosen was high purity, with only 54 ppm of Hf and 40 ppm O, suggesting that the $\alpha$–$\omega$ phase transition should occur at a stress of around 7 GPa. A 50 mm bore gas-gun was used to impact metallurgically pure 10 mm diameter samples of Zr encased in annular lower purity Zr rings. These samples were then nominally soft captured in loose rags. A pair of shots was carried out at 330 m s$^{-1}$ using Cu flyers (giving a peak stress of 5 GPa, below the expected phase change on this material) on samples cut both in-plane and transverse to the rolling direction of the plate. The resultant free surface velocity traces, shown in figure 7, were broadly similar. The main difference between the traces lay in the initial elastic behaviour, with a more ramped response and higher HEL apparent in the through-thickness
direction. The pullback signal seemed relatively unaffected by loading orientation; further, in both cases, the subsequent reloading as the tensile releases reflected off the back of the spalled region was comparable with a two-stage response attributed elsewhere to elastic–plastic behaviour (figure 7).

Interestingly, as highlighted in figure 7, there are different reloading gradients for the two loading directions—suggestive of different spallation mechanisms. Chen et al. [91] have previously linked a ‘shoulder-like’ feature to a transition between ductile and brittle-like failure; this explanation has been adopted by several other authors [81], whereas others have put this down to the rate of damage accumulation [92,93]. This supposition is backed here by optical micrographs presented by Gray III et al. which showed a slight difference in tensile failure. In both the in-plane and transverse orientations, ductile voids were apparent. However, in the transverse case, these voids had begun to coalesce with some microcracking apparent. This effect was attributed to the textural differences in these two orientations, with the Zr behaving in a ‘softer’ manner orthogonal to the transverse direction resulting in faster void coalescence.

These results tie in with recent work by the authors of this paper. Plate-impact experiments were conducted on commercially pure 19 mm diameter Zr rod sourced from Goodfellow Cambridge Ltd (UK) and of composition (in ppm): 2500 Hf; 1000 O; 200 C; 200 Cr; 200 Fe; 100 N; 10 H. The relatively high oxygen content is necessary to preserve strength [94,95], but as discussed previously will likely have influenced any phase changes while under test. The supplied material had an equiaxed microstructure, with a grain size of ca 25 μm—broadly similar to that of the Zr studied by Gray III et al. [90].

A series of five plate-impact shots were carried out with a 50 mm bore single-stage gas-gun using approximately 50 mm diameter, 2.5 mm thick Al 6082-T6 flyer plates to impact 5.0 mm thick sections of the as-supplied Zr rod. This configuration was chosen such that spall occurred within the bulk of the material; typically, the spall plane formed approximately 1–2 mm from the rear of the target—allowing spall evolution, along with rapid transmission of reloading data. The free surface of the Zr targets was monitored using a single heterodyne velocimetry channel. Resultant free surface velocity traces are presented in figure 8a.

The general form of these traces is very similar to that observed in [90]. The HEL is clearly apparent and appears to be relatively independent of the impact conditions. Applying the Zr material properties (\( \rho_0 = 6.506 \text{ g cm}^{-3} \), \( c_L = 4.77 \text{ mm} \mu \text{s}^{-1} \)) presented by Marsh [96], this HEL is
calculated to be 0.60 ± 0.02 GPa. For comparison, the HEL values extracted from [90] are 0.62 and 1.24 GPa, for the in-plane and through-thickness directions, respectively. The in-plane HEL matches the value found along the extrusion direction of the Zr rod considered here within experimental error. However, the substantial differences in the HEL for the two directions clearly illustrate the importance of microstructure and anisotropy with regards to the high strain-rate response of Zr—as previously discussed with Mg.

Of particular note in figure 8a is the relatively short duration of the velocity plateaus. To discount the influence of release waves from the periphery affecting the shock duration, a calculation of the time of release arrival from the radial periphery of the target was carried out. This calculation used the ‘worst-case scenarios’ of the elastic sound speed and a shortened release path assuming a sample with an 8 mm radius. This suggests that releases from the sample edge could not arrive before an absolute minimum of 1.68 µs after impact (i.e. 0.5 µs in figure 8a); this is significantly beyond the post spall-minimum reloading data in all cases.

The shapes of the pullback spall signals are shown in figure 8b. Overall, a linear decrease in pullback magnitude with increasing impact stress is apparent for the Zr rod tested here. To the authors’ best knowledge, this strain-rate softening behaviour has not been explicitly highlighted for Zr elsewhere; however, it is consistent (although more marked) with the known response of other HCP materials such as Ti64 and a Mg alloy [58]. It is also interesting to note that the free-surface velocity drop measured from figure 7 lies below the new experimental data (approx. 155 m s⁻¹). This is despite the similarities in the measured HEL values.

Optical micrographs highlighting the extent of damage around the resultant spall planes are presented in figure 9; these micrographs are taken from the central section of the spalled targets. These micrographs clearly show the classical evolution of ductile voids, which coalesce with increasing impact stress to form defined spall planes by the 544 m s⁻¹ shot.

The amount of damage in the micrographs presented in figure 9 appears to increase with impact stress up to an impact velocity of 5.3 GPa (figure 9c). Figure 9e shows evidence of void coalescence to form extended damage zones (indicated by the arrows); at 7.6 GPa, however, the spall damage appears more diffuse with less evidence of void coalescence (figure 9f). Furthermore, this change in behaviour occurs at an impact stress between 5.3 GPa and 7.6 GPa which is consistent with the observed α–ω phase change threshold of 7.1–8.3 GPa for Zr. Cerreta et al. [97] have shown that the hardness of shocked Zr increases substantially at stresses above the threshold for phase change. It is well known that high strength materials resist the growth of cavities during ductile spall [25,98], and thus it is possible that the harder ω phase inhibits the void coalescence and this may explain the reduction in damage evolution.
Figure 9. Optical micrographs showing spall damage for shock stress: (a) 2.6; (b) 3.9; (c) 5.3; (d) 7.6 GPa. Scanning electron micrographs of the spall damage at (e) 5.3 and (f) 7.6 GPa. The shock direction in all micrographs is from bottom to top.

6. Conclusion

We have reviewed the shock and spall response of three industrially important HCP metals and their alloys, with emphasis on how microstructure and crystallographic texture affect the dynamic mechanical behaviour. The most salient features of their shock response are as follows. (i) The oxygen content affects the stress threshold for both the elastic–plastic transition and the phase changes. (ii) The commercially available materials exhibit a highly heterogeneous and anisotropic microstructure, which in turn causes a directional dependency of the elastic limit and, in some cases, the spall strength. (iii) For the case of Zr, the shock-induced phase change seems to alter the damage evolution during spall.
All these effects are inherently related to the processing of these materials. This leaves the following questions open: can we tune the microstructure through commercial/industrial processing to optimize their mechanical behaviour under shock loading conditions? Specifically, can we tailor the oxygen content; generate an optimum amount of metastable high-pressure phases; minimize the morphological and texture anisotropy; and ultimately control the number of ‘directionally’ hard and soft grains to increase the spall strength of these materials? If so, an increase in the use of this unique class of materials for industrial applications that involve shock loading should be expected.

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References


