

High strain-rate spallation and fracture of tungsten by laser-induced stress waves

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ABSTRACT

This paper presents a study on the high strain-rate spallation and fracture behavior of bulk tungsten and tungsten/tungsten interfaces using laser-induced stress waves. The behavior of bulk polycrystalline tungsten was studied using both Gaussian and shock waves. In-plane cracking and spallations were observed at sufficiently high amplitude stress waves. The in-plane cracks propagated along relatively straight paths and extended to distances two to three times of the loading diameter. Microstructural study of the fracture mechanism show that bulk tungsten spallation was mainly due to grain boundary decohesion resulted from the growth, interconnection and coalescence of micro-voids on grain boundaries. The spallation behavior of tungsten/tungsten interface produced by a tungsten film magnetron sputtered on a tungsten substrate was studied using shock waves. The spallation occurred right at the interfacial region. The spall strength of bulk tungsten was estimated to be between 2.7 ± 0.2 and 3.1 ± 0.3 GPa, and the interfacial strength of tungsten/tungsten interface was estimated to be 875 ± 161 MPa.

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1. Introduction

Tungsten (W) has long been recognized by its high melting temperature, high density and high strength that attracted growing interest in defense applications such as kinetic energy penetrators. These long rod penetrators are used to defeat armored target with emitting rod velocity as high as 2 km/s. Rolled homogeneous armor with thicknesses exceeding 1 m can be pierced [1]. During the penetration process, both the penetrator and the armor materials experience large plastic strains especially surrounding the penetration cavity. The armor material in the path of the penetrator is pushed aside, while the penetrator is inverted and eroded at the moving penetrator–target interface until the penetrator is brought to rest or is completely consumed. Peak stresses with strain rates up to 10^6 s^{-1} [1] occur at initial contact, but quickly diminish as relief waves return from the free surface.

Good performance of kinetic energy penetrators requires special properties, including high density, optimum combinations of hardness, stiffness, strength and also good ballistic performances. Tungsten is a good candidate with its high density and high strength. However, like many other bcc metals, the pure poly-

crystalline tungsten is intrinsically brittle, while single crystal tungsten behaves ductile when loaded in certain crystallographic orientations [2,3]. Due to the difficulty and high expense of producing single crystalline tungsten, tungsten heavy alloys (WHAs) (bcc structured spheroid hard tungsten crystals embedded in an fcc structured Ni–Fe–W matrix) are more promising, which maintain the high strength, high density and high melting temperature, while having improved ductility behavior. In order to obtain good ballistic performances, most of the improvement of WHAs focuses on the formation of adiabatic shear bands to prompt self-sharpening during penetration [4–6]. However it has been argued that the ballistic performance (as well as the penetration performance) is not directly related to the static mechanical properties, and may not be predicted by the quasi-static behavior of materials [7]. Therefore, mechanical evaluation under appropriate strain rates is required.

Numerous works on the characterization of WHAs and polycrystalline tungsten under various strain rates were reported. Strain rates of 10^{-3} to 10^3 s^{-1} were obtained by standard MTS testing systems [8–10]. Split Hopkinson pressure bar (SHPB) and plate impact experiments were usually used to perform experiments at higher strain rates of 10^3 to 10^5 s^{-1} [11–16]. Failure micro-mechanisms were investigated under these strain-rates range [14], which confirms the suggestion that the bond strength of tungsten–matrix interfaces determines the fracture modes in WHAs to some extent [17]. Numerical simulations using advanced cohesive laws were also carried out to investigate the spall strength and frac-

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ture mechanisms of WHAs [18–21]. However the real strength at grain boundaries especially between tungsten and matrix, tungsten and tungsten are difficult to resolve. Also, the high pressure and high strain rate (10^6 s $^{-1}$) experienced in the penetration event is difficult to achieve with traditional methods. It has been shown that the behavior of pure tungsten is very important when considering the overall behavior of WHAs at high strain rates [9,22].

Motivated by the above needs and challenges, a laser-induced stress wave technique [23–30] is used in this work to investigate the high strain-rate spallation and fracture behavior of tungsten. This method is capable of achieving a strain rate up to 10^6 to 10^7 s $^{-1}$, that well satisfies the conditions required in the aforementioned penetration processes. The present study focuses on the high strain-rate behavior (spallation, fracture and strength) of bulk tungsten substrate and tungsten/tungsten interface. Based on the understanding obtained, future work will focus on tungsten heavy alloys.

2. Experimental setup

The experimental setup used in this work is similar to the one reported by Wang et al. [28] and a schematic is shown in Fig. 1. A 5 ns Nd:YAG laser pulse (1064 nm in wavelength) with varying energy from 0 to 300 mJ is fired toward the energy absorbing Al film sandwiched between the back surface of the tungsten substrate and the rigid waterglass confining layer. Upon absorbing the laser pulse energy, the exfoliation of the absorbing Al film generates a compressive stress wave traveling in the substrate toward the tungsten film. Upon reflection from the free film surface, the compressive stress wave becomes a tensile wave which leads to the eventual spallation of the film at a sufficiently high stress amplitude. In the mean time, the transient out-of-plane displacement of the sample front surface is measured using a Michelson interferometer. A 5 W argon ion laser (514 nm in wavelength) is used as the probing laser. The interference signals are detected by a superfast photodetector and subsequently digitized by an ultra-fast 5 GHz oscilloscope at a sampling rate of 20 G sample/s. During the testing, the Nd:YAG laser pulse energy is varied for different test spots (resulting in varying stress wave amplitude) while the loading area on the sample surface was fixed to be 1–2 mm in diameter to maintain the plane wave characteristics.

The stress wave profiles can be obtained from the measured film displacement using wave mechanics. Based on the calculation principles described in a previous publication [28], when the loading

area is large compared to the total thickness of the sample, the substrate stress can be obtained from a simple one-dimensional wave equation:

$$\sigma_{\text{sub}} = -\frac{1}{2}(\rho c_L)_{\text{sub}} \frac{\partial u}{\partial t} \quad (1)$$

where u is the out-of-plane displacement experimentally measured from the free surface of the test film using the optical interferometer; ρ and c_L are the density and longitudinal wave speed of the substrate. The stress loaded at the film/substrate interface is resulted from the superposition between the incoming compressive stress wave and the reflected tensile stress wave from the free surface, and can be calculated as:

$$\sigma_{\text{int}} = \sigma_{\text{sub}} \left(t - \frac{h}{c_L} \right) - \sigma_{\text{sub}} \left(t + \frac{h}{c_L} \right) \quad (2)$$

where h and c_L are the thickness and longitudinal wave speed of the film. For the tungsten sample tested in this work, $\rho = 19.3 \times 10^3$ kg/m ^{3} , and $c_L = 5218$ m/s (calculated using Poisson's ratio $\nu = 0.28$ and Young's modulus $E = 409$ GPa from ultrasonic measurements reported in the literature [31]).

With the sample configuration described above, due to the characteristics of the Nd:YAG laser (Gaussian distribution in time and uniform pancake distribution in space), the compressive stress wave generated is also of Gaussian distribution in time and uniform in space. Stress wave generated in this way is referred as “Gaussian stress wave” in this work.

In most practical applications, in order to obtain a sufficiently high stress to delaminate the thin film interface, a shock stress wave with much sharper unloading ramp is desired. Previous studies by Wang et al. [29] have proved that the Gaussian stress wave generated at the absorbing layer can evolve into a shock after traveling certain distance in fused silica due to the unique negative nonlinear elasticity of this material in the low stress region. The shock wave is very beneficial for the interfacial adhesion testing of thin films. In order to study the dynamic response of bulk tungsten and tungsten/tungsten interface under shock wave loading, an alternative sample configuration as shown in the inset of Fig. 1 is also adopted. The stress wave generated using this sample configuration is referred as “shock wave” in this work.

3. Sample preparation

Commercially available polycrystalline tungsten from Eagle Alloys Corp. (Talbott, TN) was used as the substrate in this work.

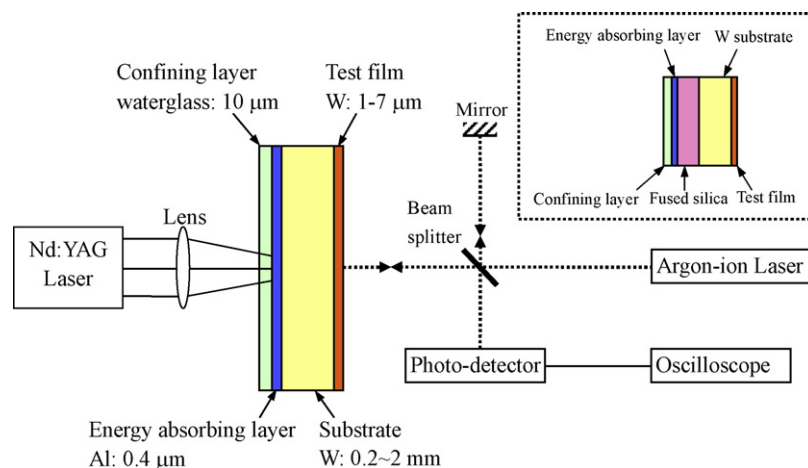


Fig. 1. Schematic of the experimental setup and sample configuration. The inset shows an alternative sample configuration for inducing shock wave loading using fused silica as pulse-shaping material.

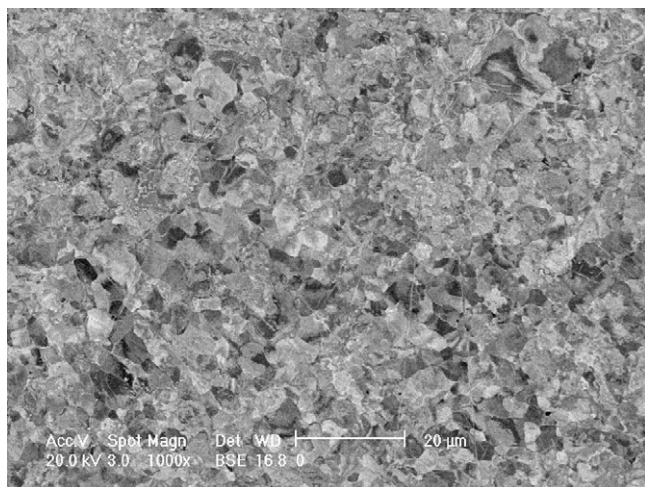


Fig. 2. SEM image showing the W substrate texture.

The as-received material has a purity of 99.9%, and was in the form of tungsten plates of 1 in. in diameter and varying thickness cut from a rod sample, with very rough surfaces on both sides. Before performing experiments, both sides of the tungsten plate were polished using diamond platen of 180, 325, 600 and 1200 mesh sizes, respectively. They were then polished with 9 and 3 μm diamond paste, and finally with a 0.05 μm alumina slurry as the finishing step. Fig. 2 shows a scanning electron microscopy (SEM) image of a polished tungsten surface. A grain size of about 5 μm can be clearly identified from the image.

Following the sample configuration shown in Fig. 1, after finishing polishing on both sides of the tungsten plate, a 0.4 μm thick Al film was deposited on one side (back side) of the substrate as energy absorbing layer using magnetron sputtering (ORION-5-UHV, AJA International). A layer of waterglass around 10 μm thick was spin-coated on top of the energy absorbing Al film as confining layer. The front surface of the tungsten substrate was either deposited with a layer of tungsten film by magnetron sputtering for studying the tungsten/tungsten interface property or left free for studying the bulk tungsten behavior. As described earlier, samples prepared in this way will undergo a Gaussian stress wave loading due to the characteristics of the Nd:YAG laser pulse.

To fully understand the dynamic behavior of tungsten, the shock wave response of tungsten was also studied using the sample configuration as shown in the inset of Fig. 1. To prepare these samples, after the completion of the polishing processes, the tungsten substrate with desired thickness was attached to a 500 μm thick fused silica plate using a thin layer of super glue. A 0.4 μm thick Al layer was then deposited onto the back surface of the fused silica plate as energy absorbing layer followed by a waterglass confining layer. Due to the special negative nonlinear elastic properties of fused silica, the originally generated Gaussian shape compressive stress wave will first be transformed into a shock wave in fused silica and then continue to propagate into the tungsten substrate [29].

4. Results and discussion

4.1. Bulk tungsten spallation and fracture under Gaussian loading

The high strain-rate behavior of bare tungsten substrate was investigated first using the sample configuration shown in Fig. 1 without depositing the tungsten film. In this way a Gaussian stress wave is generated and only the substrate response is studied. The substrate stress profiles measured from tungsten substrates of four

different thicknesses, 0.4, 0.8, 1.5, and 1.8 mm, are shown in Fig. 3. These samples were prepared and tested under the same conditions in order to have consistent and comparable results. Specifically, the thickness of the waterglass confining layer and the aluminum energy absorbing layer were kept the same, and the loading diameter was kept the same for all the samples. The pulse energy from the Nd:YAG laser was set to maximum.

It has to be noted however, since the substrate stress is calculated from the free surface displacement, it only represents the substrate stress close to the free surface, not directly related to the stresses deep inside the substrate before reaching the free surface. It can be noticed from Fig. 3 that the Gaussian stress wave profiles are slightly deformed, with a 5 ns loading ramp followed by a stretched 10 ns unloading ramp. The stress amplitude is much lower when the substrate gets thicker (indicating a stress attenuation with increasing travel distance). Since all the control parameters were the same, these curves can be treated as from different cross-sections from one thick sample (>1.8 mm). It can be inferred that while the Gaussian distribution of the stress wave was well maintained at different travel distances, there is a clear amplitude attenuation with respect to the travel distance. There are two possible mechanisms responsible for the amplitude attenuation, specifically the geometrical attenuation and the tungsten intrinsic wave absorption. However, according to a previous calculation [28], for a 1–2 mm loading diameter, little geometrical attenuation is expected with a traveling distance less than 2 mm. Therefore, the intrinsic wave absorption of tungsten should be the major cause of the stress attenuation.

The maximum compressive stress in Fig. 3 was measured to be 4.2 ± 0.2 GPa from the 0.4 mm thick tungsten substrate. Using this value, the maximum strain can be calculated to be about 0.01. Considering a nominal duration of 5 ns, the maximum strain rate would be $\sim 10^6 \text{ s}^{-1}$. At such high amplitude stress level, substrate spallations were observed as slightly visible surface bumps. The maximum height of the bumps was measured using a DEKTA³ (Veeco) surface profiler to be less than 500 nm. It indicates that 4.2 ± 0.2 GPa could be the threshold substrate stress amplitude that can cause tungsten spallation. While amplitude lower than 4.2 ± 0.2 GPa will not initiate the spallation, a higher amplitude will lead to more obvious and severe tungsten substrate spallation and fracture. This can be achieved by either reducing the tungsten substrate thickness or increasing the Nd:YAG laser energy fluence. However, since the maximum laser energy is fixed as mentioned before, decreasing the substrate thickness would be more practical

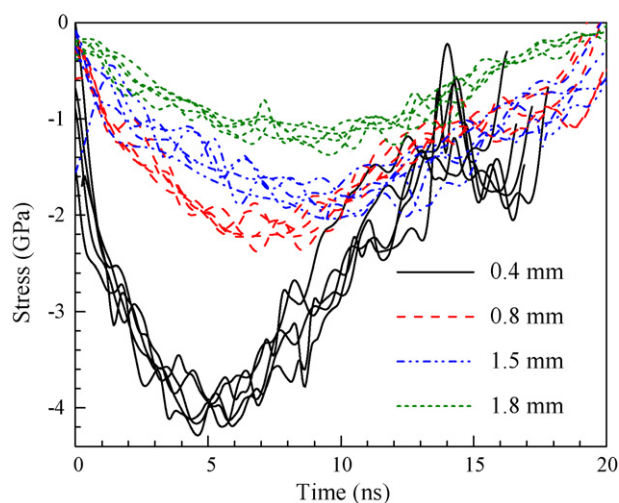


Fig. 3. Substrate stress profiles in W substrates with different thicknesses: 0.4, 0.8, 1.5 and 1.8 mm due to Gaussian wave loading.

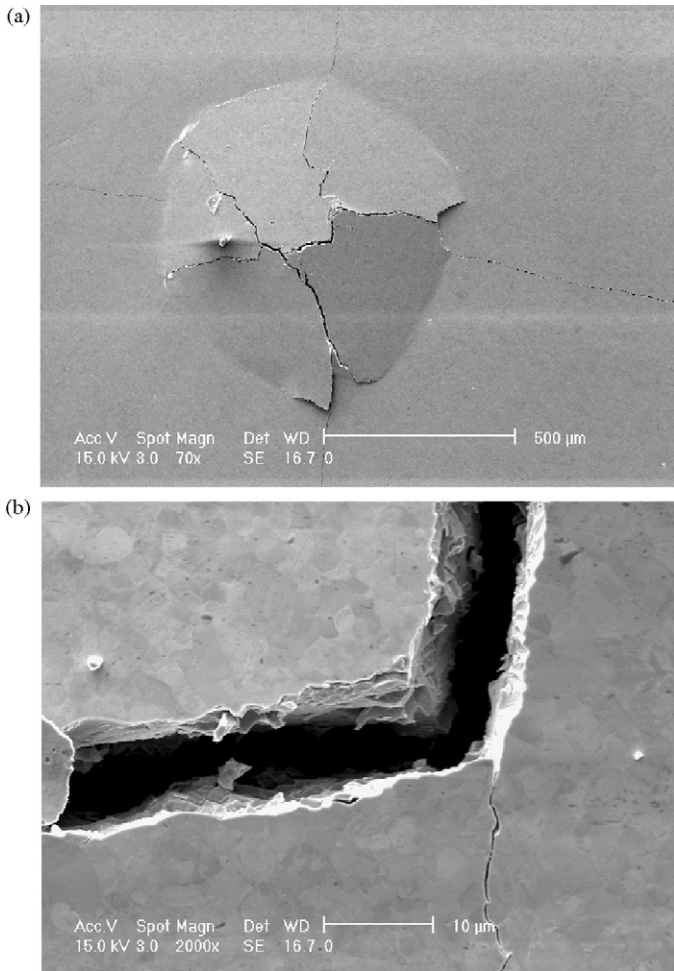


Fig. 4. (a) Spallation and crack observed on a 0.2 mm thick W substrate and (b) zoomed-in view of the in-plane crack.

in order to lessen the stress wave attenuation. When the tungsten substrate thickness was reduced to 0.2 mm, clear substrate spallation and fracture was observed as shown in Fig. 4(a). The spallation is in the form of a surface bump with diameter comparable to the loading diameter (~ 1 mm). All the spallations happen in the similar manner that the loaded area of the substrate was spalled and lifted from its original height with the actual spallation plane buried underneath the bumped surface. In-plane crackings can be seen both on the top of the bump and on the surrounding substrate surface indicated by the crack lines. A zoomed-in view of the in-plane crack as shown in Fig. 4(b) indicates that the in-plane crack did not follow the grain boundaries but rather break through the grains along its path.

4.2. Bulk tungsten spallation and fracture under shock loading

Shock wave response of bare tungsten substrate was studied using the sample configuration shown in the inset of Fig. 1 without the tungsten film. A 500 μm thick fused silica plate was used to evolve the originally generated Gaussian stress wave at the energy absorbing layer into a shock wave. The theoretical shock formation distance can be calculated according to a previous study [29]. For example, for a 1 GPa amplitude Gaussian stress wave, shock wave is estimated to form after traveling about 236 μm in fused silica, and for stress waves with a higher amplitude, a shorter travel distance will be required. Based on the stress amplitude measured in the

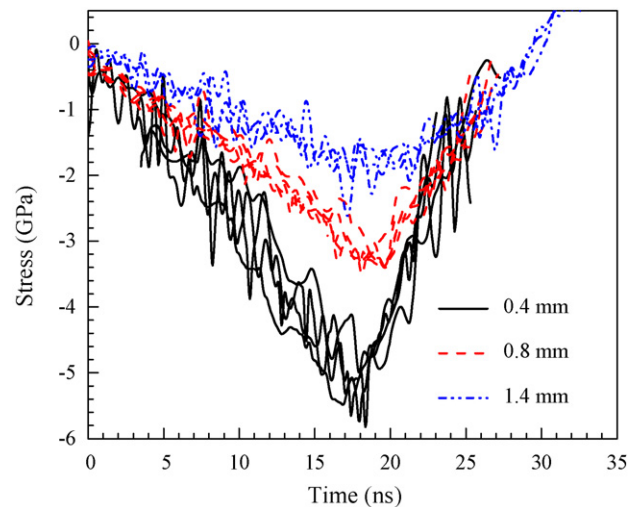


Fig. 5. Substrate stress profiles obtained on W substrate of different thicknesses: 0.4, 0.8 and 1.4 mm due to shock wave loading.

previous section, a 500 μm thick fused silica is expected to be thick enough to generate shock wave while minimizing the stress wave attenuation.

Fig. 5 shows the substrate stress profiles measured after the stress wave traveling through the tungsten thicknesses of 0.4, 0.8 and 1.4 mm, respectively. All the data collected were from experiments conducted under the same conditions. Similar to what was observed under Gaussian stress wave, the stress wave maintained the same shape at several traveling distances with attenuating amplitude due to the wave absorption in tungsten. The stress profile has a stretched linear loading ramp of about 18 ns, and an unloading ramp of about 10 ns, which is not as sharp (~ 1 ns) as reported in the previous work by Wang et al. [29]. The loss of the sharpness of the shock may be attributed to the high stress amplitude. As fused silica has negative nonlinearity only at low stress region of less than 4 GPa, above which, its nonlinearity becomes positive and it can no longer maintain the sharpness of the shock waves.

On tungsten substrates of 0.8 and 1.4 mm thick, the maximum substrate stresses were measured to be 3.4 ± 0.4 and 2.0 ± 0.3 GPa, respectively, and no visible failure was observed. Visible surface bumps due to substrate spallation were observed on the 0.4 mm thick substrate at stress amplitude of 5.4 ± 0.4 GPa. Profiler scan-

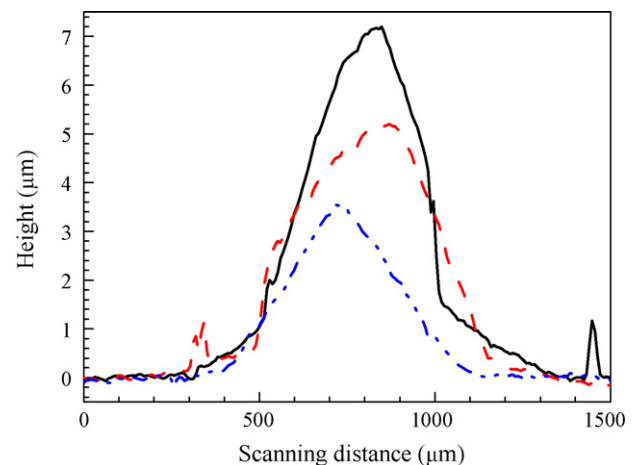


Fig. 6. Profiler scanning traces across several surface bumps on a 0.4 mm thick W substrate due to shock wave loading.

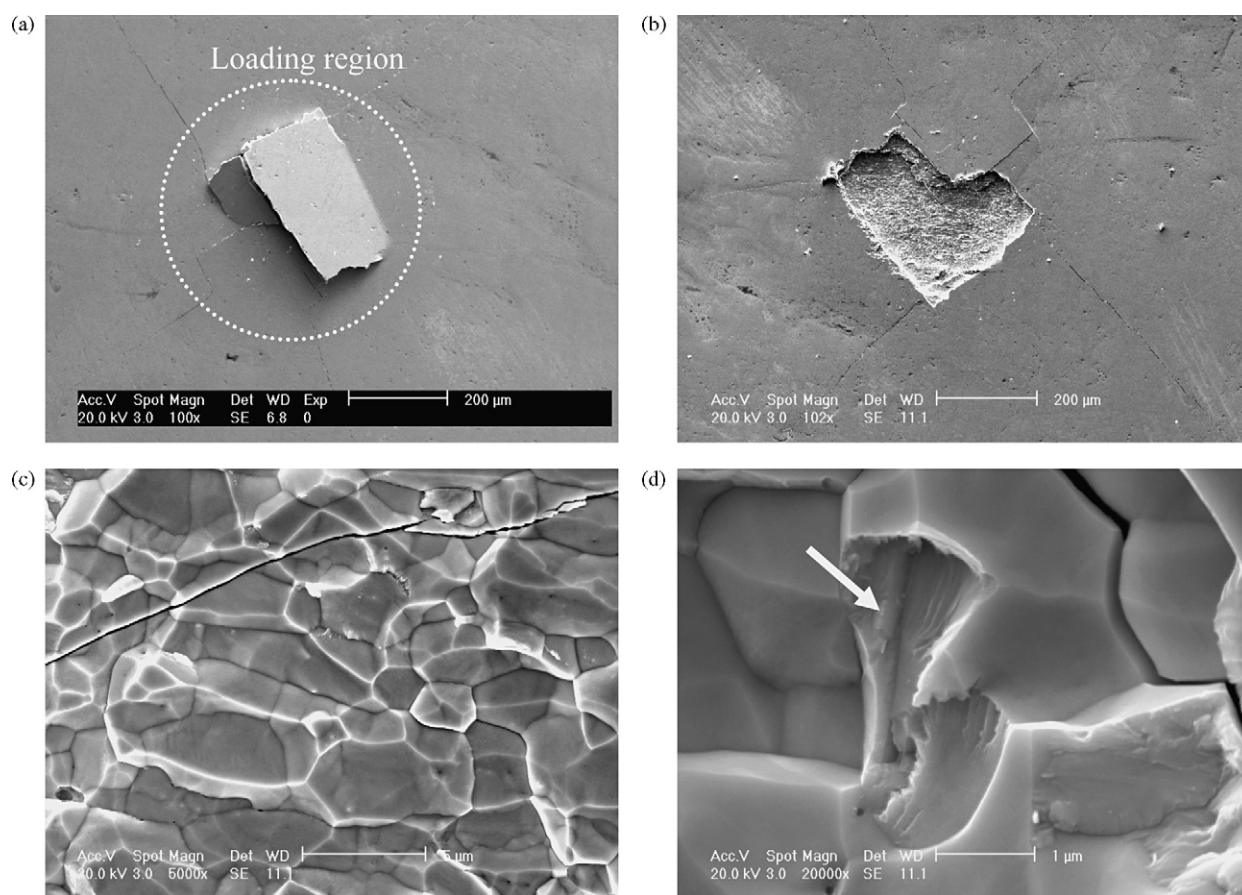


Fig. 7. SEM images showing substrate spallation and fracture on a 0.2 mm thick W substrate attached with 500 μm thick fused silica: (a) surface lifting due to spallation, dashed circle indicates the loaded region; (b) surface layer was removed to expose the underneath spallation and cracking plane; (c) and (d) zoom in view of (b), with (c) showing grain boundary decohesion that caused the spallation and the transgranular cracking lines, and (d) breakage of individual grains due to spallation.

ning traces shown in Fig. 6 confirmed the spallation. The diameter of the bump is about 1 mm which is consistent with the loading area (~1 mm in diameter), and a maximum height was measured to be about 6 μm. This is significantly higher than the spallation-induced bumps on the 0.4 mm thick tungsten plate under the Gaussian stress wave (<500 nm).

4.3. Spallation and fracture mechanism

To study the spallation and fracture mechanisms in detail, experiments were performed on thinner tungsten substrates of 0.2 mm thick using the shock wave loading and the resulted SEM pictures are shown in Fig. 7. The loading area was focused smaller in these tests in order to obtain a higher energy fluence to induce easily removable surface bumps. It can be seen that in all the damages produced, there exist both tungsten substrate spallation and in-plane cracks. It is natural that when the loaded part of the substrate is spalled off and moves in the out-of-plane direction, it is going to tear the substrate at the loading edge during the separation process. Inevitably, in-plane cracks will be generated at strain rates relative to the spallation. It is clear from the images that these cracks were initiated from the most stress-concentrated crack tips. They propagate for a distance more than two to three times of the loading diameter, just like what is typical for single crystalline materials that cracks propagate along certain crystallographic orientations. It appears that at extremely high strain rates, the in-plane cracks do not distinguish between grain and grain boundaries, and are so energetic that they create a relatively straight path.

To study the spallation mechanism, the spalled layer was uncovered to reveal the buried spallation plane. As shown in Fig. 7(b), after the spalled surface layer was removed, it is evident that the spallation was formed by one layer of the substrate lifting out. The flat remaining plane and the sharp edge confirmed the plane wave characteristics and the relatively uniform thickness of the spalled layer. This thickness was measured by profiler scanning to be about 20 μm. The smooth grain morphologies on the spallation plane shown in Fig. 7(c) indicate the spallation was mainly along grain boundaries. Such an intergranular cleavage mode is opposite to Dummer et al. observation [14] that pure polycrystalline tungsten undergoes a ductile-to-brittle transition at strain rates higher than 10^3 s^{-1} . Though there are also a few places where the breakage of individual grains happened during the spallation as pointed out by the arrow in Fig. 7(d) and river pattern can be seen over the grain surface, transgranular cleavage is evidently a minor effect compared to the intergranular cleavage. It can also be found that the random distribution of voids both in the grains and along the grain boundaries with sizes smaller than 1 μm might contribute to the overall spallation. Fig. 7(c) and (d) shows again the in-plane cracking lines. Just like those on the substrate surface, these cracks are relatively straight lines, and do not follow the grain boundaries but rather propagate through the grains ahead.

To further investigate the spallation micro-mechanisms, cross-sections right through the spallation site was obtained by electric discharge machining (EDM) with a 150 μm diameter wire. The cut cross-section was then polished by 1200 mesh diamond polishing disk to reveal the micro-features of the spallations and cracks. Care

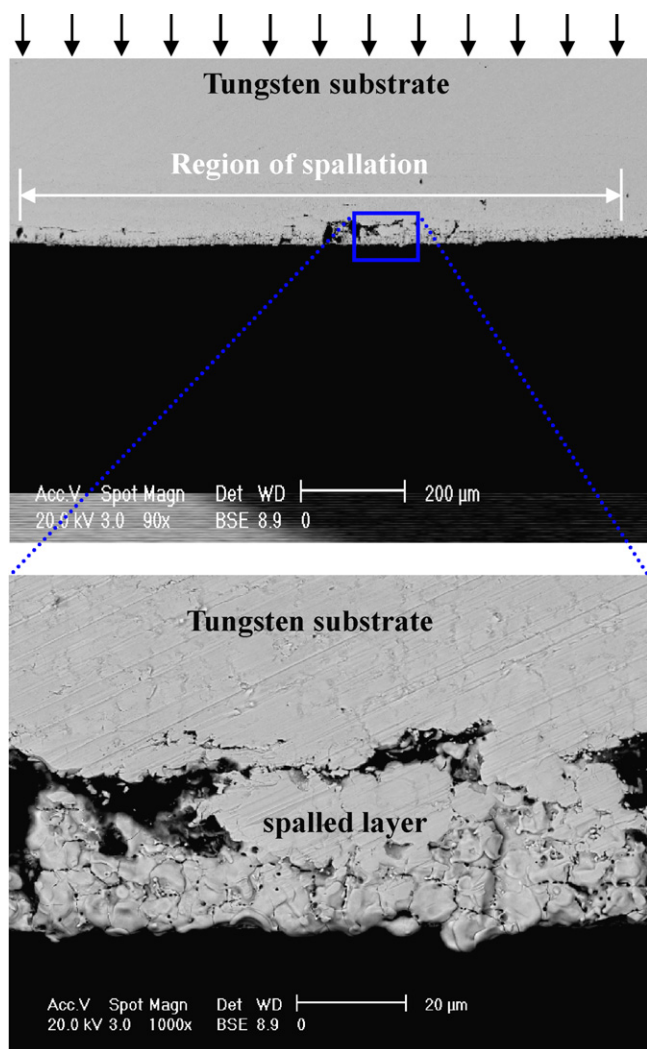


Fig. 8. SEM image of the cross-section of the spallation of a 0.4 mm thick W substrate. The bottom substrate surface in the top image indicates the position of the spallation, and the compressive stress wave comes from the top toward the bottom as indicated by the arrows.

was taken to protect the separated surface layer during the polishing. SEM images of the polished cross-section are shown in Fig. 8. The slight curvature of the bottom substrate surface in Fig. 8(a) indicates the shape of the bump due to spallation. Fig. 8(b) shows magnified image of the center spallation from Fig. 8(a). From these images, several characteristics of the tungsten spallation can be identified. First, the spallation follows the style that one layer of tungsten was spalled off from the rest of the substrate and changed the surface curvature. It clearly confirms that the bump was formed by the spallation of bulk tungsten. Second, due to the uniform distribution of the plane wave, the thickness of the separated layer was uniform along the spallation plane. The thickness was measured to be around 20–25 μm , which is consistent with that measured by profiler scanning ($\sim 20 \mu\text{m}$) on the substrate surface. However, the most important of all is the observation that the voids grow to bigger size as shown in Fig. 8(b). These voids experience tensile stresses and grow bigger during the spallation, while the voids observed on the spallation plane in Fig. 7 do not due to their specific orientations. Some of the voids were in the grains while most of them were distributed along the grain boundaries. Considering the voids distribution in Fig. 7 to be the original distribution before shock loading, it can be seen that more voids were initiated

during the spallation. Under the reflected tensile wave, the voids grow and interact with each other, forming channels, passes, and coalescence which eventually lead to the overall observed spallation. Conventional void growth mechanisms like vacancy diffusion and shock heating are not able to account for at such high strain rates. Lubarda et al. [32] proposed a dislocation–emission-based mechanism for void growth considering prismatic and shear dislocation loops. Deng et al. [33] studied the role of dislocations in the void growth in nanocrystalline copper at high strain rates by laser shock loading using molecular dynamics simulation. Surprisingly, these works on copper can be used to explain the behavior of tungsten at similar strain rates. It is probably the main reason why tungsten does not act brittle, because the plasticity at micro-scale contributes to and dominates the bulk ductile cleavage. Two tentative conclusions can be drawn: that grain boundaries are more sensitive and amenable to void nucleation and growth, and that copper and tungsten have similar failure mechanisms at high strain rates despite the differences in their crystal structures (bcc/fcc), grain size, density, etc. Of course, more experiments will be helpful in further validating these statements.

4.4. Estimation of tungsten spall strength

The spallation style and its location in bulk tungsten provide evidence that the spallation is due to the tensile stress wave reflected from the free surface. Because the thickness of the spalled layer is uniform and consistent to be about 20–25 μm for different samples, it can be considered as a very thick tungsten film deposited on a substrate surface. In this way, the tensile stress at the spallation plane can be estimated as the bulk tungsten spall strength. The minimum substrate stress amplitude obtained to introduce substrate spallation was $4.2 \pm 0.2 \text{ GPa}$ on a 0.4 mm thick tungsten substrate for very slight failure. So substrate stress corresponding to this amplitude can be used to estimate the upper bound of the tungsten spall strength using Eq. (2). For a spalled layer between 20 and 25 μm , the average upper bound for the tungsten spall strength was calculated to be $3.1 \pm 0.3 \text{ GPa}$ from 11 tests. The lower bound of the threshold amplitude was estimated from another shock wave loaded tungsten substrate of 0.8 mm thick with no substrate failure. The average spall strength of considering the spalled layer to be 20 and 25 μm was $2.7 \pm 0.2 \text{ GPa}$ from six tests. So the spall strength of tungsten is bounded between 2.7 ± 0.2 and $3.1 \pm 0.3 \text{ GPa}$. The average upper and lower bounds of the spall strength and the corresponding substrate stress amplitudes are shown in Fig. 9. Surprisingly, the spall strength of tungsten measured here is very similar to that of nanocrystalline copper ($\sim 3 \text{ GPa}$) measured under similar conditions [34]. And it is significantly higher than the tungsten failure strength measured under quasi-static loading ($\sim 1.2 \text{ GPa}$) by standard material testing system, at strain rates of 10^3 to 10^4 s^{-1} by split Hopkinson bar ($\sim 2 \text{ GPa}$) [14,15,35], and at strain rates of 10^5 s^{-1} ($\sim 0.5 \text{ GPa}$) by plate impact-induced shock loading experiment [36]. However, such a trend that the failure strength increases dramatically with the increasing strain rate is consistent with what was observed in the above mentioned works.

4.5. Interfacial strength of W/W interface under shock wave loading

In addition to the bulk tungsten spallation and fracture study, the tungsten/tungsten interfacial properties were investigated using the shock stress waves. Tungsten films of varying thicknesses were deposited on tungsten substrates by a magnetron sputtering system at 10^{-7} mTorr base vacuum pressure and 150 W DC power. As shown in Fig. 10(a), the deposited films have a grain size of less than 100 nm. The XRD (X-ray diffraction) data in Fig. 10(b) show

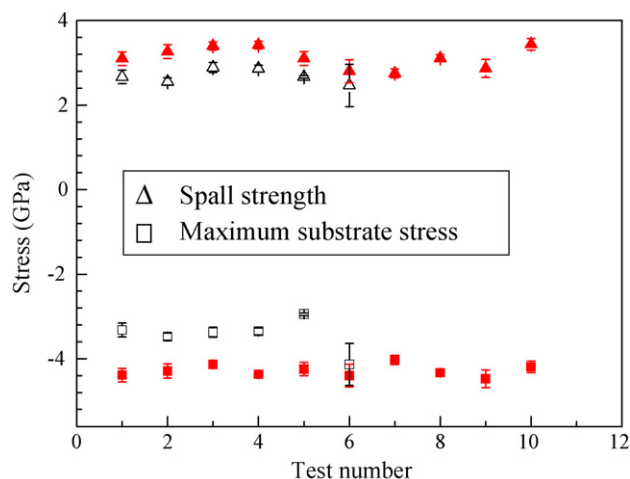


Fig. 9. Estimation of W spall strength of the upper bound and the lower bound. Filled symbols represent upper bound and open symbols represent lower bound. Error bar was estimated from the numerical noise when taking the derivative of the displacement using Eq. (1) to get substrate stress.

three clearly distinguishable peaks at 2θ angles of 40.3° , 58.3° and 73.3° which correspond to the tungsten (α -phase) (110), (200) and (211) planes, respectively. It is worth to mention that the tungsten substrate had exactly the same three peaks in the XRD spectra (not shown), indicating the effect of substrate texture on the film growth.

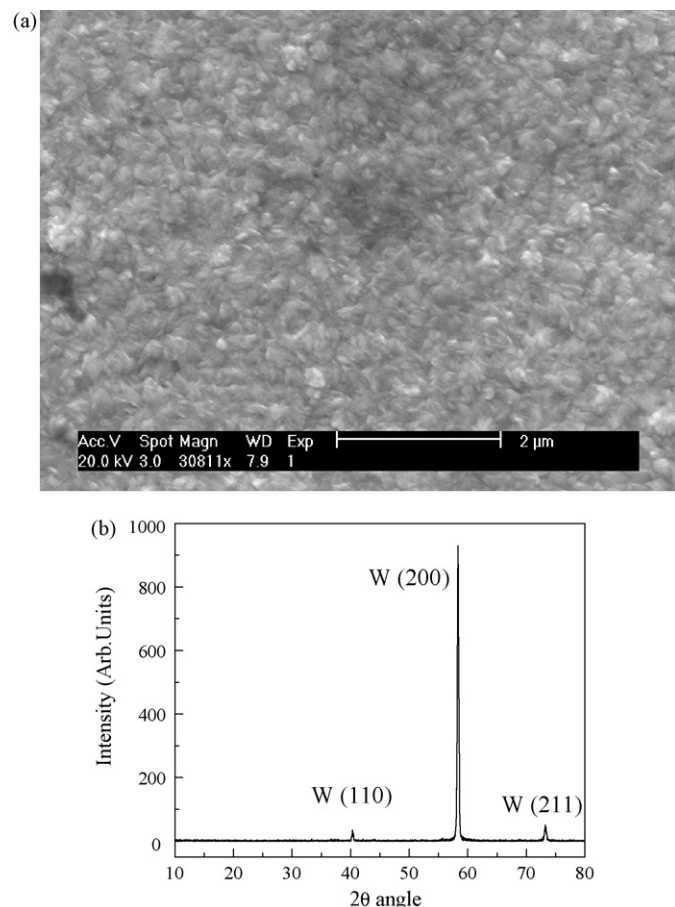


Fig. 10. SEM image (a) and XRD pattern (b) from a $1.1\ \mu\text{m}$ thick W film deposited on a $0.5\ \text{mm}$ thick W substrate.

Tungsten/tungsten interfacial failure was obtained on a tungsten substrate with a $6\ \mu\text{m}$ thick tungsten film as shown in Fig. 11(a). The tungsten film in the center loading region was spalled by the stress wave, leaving the substrate surface exposed. The center of the remaining surface is smooth, while on the edge of the delamination, layered tungsten film can be seen as shown in Fig. 11(b), which was due to the multiple deposition runs to obtain a very thick tungsten film. The interfacial strength of tungsten/tungsten interface was obtained by gradually increasing the laser fluence from spot to spot to achieve a range of spallation behavior including no failure, slight failure and extensive failure. The stress bounded by the slight failure and no failure was quoted as the interfacial strength. Fig. 12 shows the interfacial strengths and the corresponding substrate stresses measured on a $5.0\ \mu\text{m}$ thick tungsten film deposited on a $0.4\ \text{mm}$ thick tungsten substrate using the shock wave loading. From the total 17 measurements, the interfacial strength was averaged to be 875 ± 161 .

When the substrate stress amplitude was sufficiently high, the substrate spallation occurred together with the interfacial failure. Fig. 13 shows profiler scanning traces across different testing spots with tungsten/tungsten interfacial delamination. The small hill at the bottom line indicates the spallation of the substrate. These hills disappear as the laser energy fluence (proportional to the amplitude of the substrate stress) is decreased, which indicates that

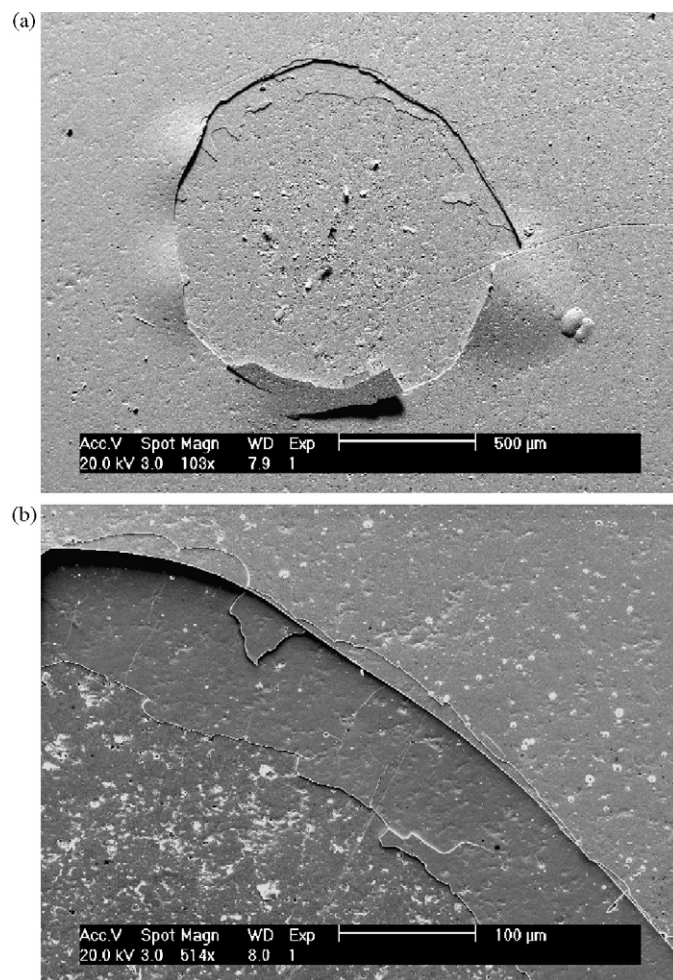


Fig. 11. Interfacial delamination between a $0.4\ \text{mm}$ thick W substrate and a $6\ \mu\text{m}$ thick W film under shock wave loading: (a) full view of the spallation shows the removal of the W film in the loaded region and (b) zoom-in view near the edge of the loading region reveals the layered structure of the film.

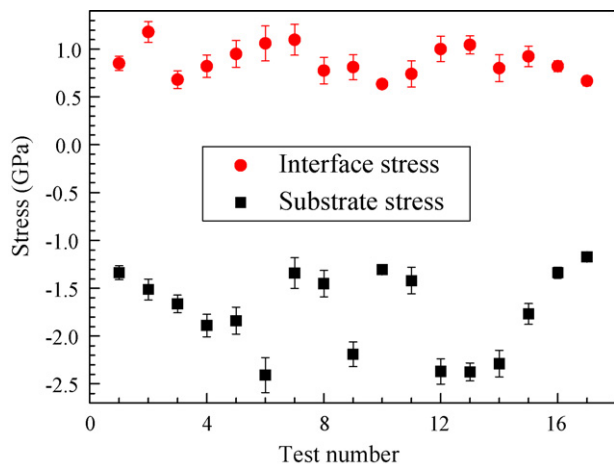


Fig. 12. Interfacial strength of W/W interface measured from a 5 μm thick W film deposited on a 0.4 mm thick W substrate.

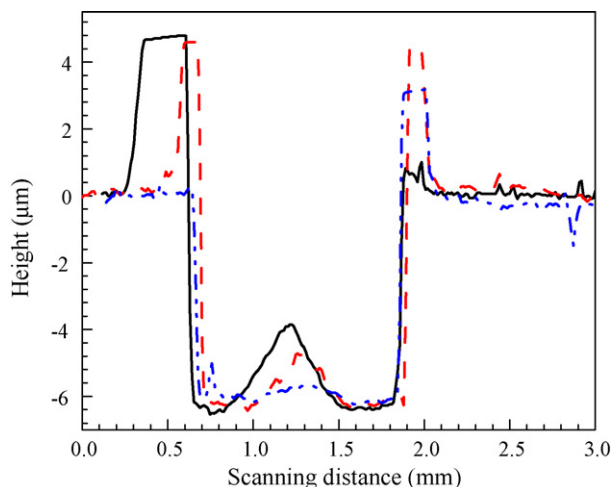


Fig. 13. Profiler scanning traces of different W/W interfacial delaminations.

the interfacial delamination took place before the tungsten substrate spallation. And the interfacial strength between tungsten and tungsten is lower than the bulk tungsten spall strength. Also the tungsten substrate spallation was due to the tensile stress wave reflected from the free surface, which again reinforces our conclusion in the estimation of the tungsten spall strength.

5. Conclusions

The dynamic behavior of bulk polycrystalline tungsten and tungsten/tungsten interface was investigated at a strain rate around 10^6 s^{-1} using laser-induced stress waves. Spallation and fracture was observed in bulk tungsten at sufficiently high stress levels. The spallation was characterized by a layer of the tungsten substrate lifted off from its original plane to form a surface bump. The fracture was characterized by in-plane cracks extending from the loading edge to two to three times of the loading diameter and traveling along a relative straight path through the grains and grain boundaries. The spallation of tungsten was dominated by grain boundary decohesion combined with minor effect from grain breakages. Cross-sectional SEM revealed that the spallation micro-mechanism can be attributed to micro-voids nucleation, growth

and coalescence at the grain boundaries under the reflected tensile wave. The spall strength of bulk tungsten was estimated to be between 2.7 ± 0.2 and 3.1 ± 0.3 GPa. This strength is much higher than that reported at lower strain rates, but may be considered reasonable due to the rate sensitivity of tungsten. In addition, the tungsten/tungsten interfacial strength was measured to be 875 ± 161 MPa, significantly lower than the spall strength of bulk tungsten.

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References

- [1] L.S. Magness, *Mechanics of Materials* 17 (1994) 147–154.
- [2] A.S. Argon, S.R. Maloof, *Acta Metallurgica* 14 (1966) 1449–1462.
- [3] R.M. Rose, D.P. Ferriss, J. Wulff, *Transactions of the Metallurgical Society of AIME* 224 (1962) 981–990.
- [4] Z.G. Wei, J.L. Yu, S.S. Hu, Y.C. Li, *International Journal of Impact Engineering* 24 (2000) 747–758.
- [5] Q. Wei, K.T. Ramesh, E. Ma, L.J. Kescskes, R.J. Dowding, V.U. Kazykhanov, R.Z. Valiev, *Applied Physics Letters* 86 (2005) 101907.
- [6] R.C. Batra, B.M. Love, *Journal of Thermal Stresses* 28 (2005) 747–782.
- [7] W.D. Cai, Y. Li, R.J. Dowding, F.A. Mohamed, E.J. Lavernia, *Reviews in Particulate Materials* 3 (1995) 71–131.
- [8] R.L. Woodward, N.J. Baldwin, I. Burch, B.J. Baxter, *Metallurgical Transactions A: Physical Metallurgy and Materials Science* 16 (1985) 2031–2037.
- [9] A.K. Zurek, G.T. Gray, *Journal De Physique III* 1 (1991) 631–637.
- [10] W.S. Lee, S.T. Chiou, *Composites Part B: Engineering* 27 (1996) 193–200.
- [11] T. Weerasooriya, P.A. Beaulieu, *Materials Science and Engineering A: Structural Materials Properties Microstructure and Processing* 172 (1993) 71–78.
- [12] G. Subhash, Y.J. Lee, G. Ravichandran, *Acta Metallurgica Et Materialia* 42 (1994) 331–340.
- [13] M. Zhou, R.J. Clifton, *Journal of Applied Mechanics-Transactions of the ASME* 64 (1997) 487–494.
- [14] T. Dummer, J.C. Lasalvia, G. Ravichandran, M.A. Meyers, *Acta Materialia* 46 (1998) 6267–6290.
- [15] W.S. Lee, G.L. Xie, C.F. Lin, *Materials Science and Engineering A: Structural Materials Properties Microstructure and Processing* 257 (1998) 256–267.
- [16] A.M. Lennon, K.T. Ramesh, *Materials Science and Engineering A: Structural Materials Properties Microstructure and Processing* 276 (2000) 9–21.
- [17] R.M. German, J.E. Hanafee, S.L. Digiallonardo, *Metallurgical Transactions A: Physical Metallurgy and Materials Science* 15 (1984) 121–128.
- [18] J.D. Clayton, *International Journal of Solids and Structures* 42 (2005) 4613–4640.
- [19] J.D. Clayton, *Journal of the Mechanics and Physics of Solids* 53 (2005) 261–301.
- [20] J.D. Clayton, *Theoretical and Applied Fracture Mechanics* 45 (2006) 163–185.
- [21] T.J. Vogler, J.D. Clayton, *Journal of the Mechanics and Physics of Solids* 56 (2008) 297–335.
- [22] K.T. Ramesh, R.S. Coates, *Metallurgical Transactions A: Physical Metallurgy and Materials Science* 23 (1992) 2625–2630.
- [23] J.L. Vossen, *ASTM STP640* (1978) 122–133.
- [24] V. Gupta, A.S. Argon, J.A. Cornie, D.M. Parks, *Materials Science and Engineering A: Structural Materials Properties Microstructure and Processing* 126 (1990) 105–117.
- [25] J. Yuan, V. Gupta, *Journal of Applied Physics* 74 (1993) 2388–2396.
- [26] V. Gupta, J. Yuan, *Journal of Applied Physics* 74 (1993) 2397–2404.
- [27] J. Yuan, V. Gupta, A. Pronin, *Journal of Applied Physics* 74 (1993) 2405–2410.
- [28] J.L. Wang, R.L. Weaver, N.R. Sottos, *Experimental Mechanics* 42 (2002) 74–83.
- [29] J.L. Wang, R.L. Weaver, N.R. Sottos, *Journal of Applied Physics* 93 (2003) 9529–9536.
- [30] J. Wang, N.R. Sottos, R.L. Weaver, *Journal of the Mechanics & Physics of Solids* 52 (2004) 999–1022.
- [31] D.I. Bolef, J. De Klerk, *Journal of Applied Physics* 33 (1962) 2311–2314.
- [32] V.A. Lubarda, M.S. Schneider, D.H. Kalantar, B.A. Remington, M.A. Meyers, *Acta Materialia* 52 (2004) 1397–1408.
- [33] X.L. Deng, W.J. Zhu, H.L. He, D.X. Wu, F.Q. Jing, *Acta Physica Sinica* 55 (2006) 4767–4773.
- [34] Y.G. Wang, H.L. He, M. Boustie, T. Sekine, *Journal of Applied Physics* 101 (2007) 103528.
- [35] J.C.F. Milliet, G.T. Gray, N.K. Bourne, *Journal of Applied Physics* 101 (2007) 033520.
- [36] J.R. Asay, L.C. Chhabildas, D.P. Dandekar, *Journal of Applied Physics* 51 (1980) 4774.