Dynamic tensile deformation and damage of B₄C-reinforced Al composites: time-resolved imaging with synchrotron x-rays

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Abstract

Dynamic tensile experiments are conducted on 15% and 30% in weight percentage B₄C/Al composites with a split Hopkinson tension bar, along with high-speed synchrotron x-ray digital image correlation (XDIC) to map strain fields at μm and μs scales. As manifested by bulk-scale stress–strain curves, a higher particle content leads to a higher yield strength but lower ductility. S-train field mapping by XDIC demonstrates that tension deformation and tensile fracture, as opposed to shear and shear failure, dominate deformation and failure of the composites. The fractographs of recovered samples show consistent features. The particle-matrix interfaces are nucleation sites for strain localizations, and their propagation and coalescence are impeded by the Al matrix. The reduced spacing between strain localization sites with increasing particle content, facilitates their coalescence and leads to decreased ductility. Designing a particle-reinforced, metallic-matrix composite with balanced strength and ductility should consider optimizing the inter-particle distance as a key parameter.

Keywords: Metal-matrix composites (MMCs); Fracture; Fractography; Radiography.

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1. Introduction

Metal matrix composites (MMCs) have been of considerable interest as advanced structural materials for their excellent mechanical properties [1–3]. Boron carbide (B_4C), a commonly-used reinforcement phase [4], has a low density, but a high hardness only slightly lower than those of diamond and boron nitride. It offers distinct advantages for various applications such as neutron absorption, impact and wear resistance [5, 6]. B_4C particle-reinforced MMCs possesses high hardness, high modulus, low density and excellent toughness simultaneously [7, 8]. High performance B_4C/Al composites have been used in various fields, such as aerospace and military engineering [7–9]. In such applications, deformation and failure of B_4C/Al composites subjected to dynamic (10^2 – 10^3 s^{-1}) loading, as well as underlying mechanisms, should be considered. However, limited high strain-rate tensile data are available for B_4C/Al [4, 6, 10]. Obtaining spatially and temporally resolved deformation and damage dynamics in B_4C/Al composite has been an experimental challenge.

For dynamic tensile loading, split Hopkinson tension bar (SHTB) has been widely used for various materials including B_4C/Al composites [7, 11, 12]. Strain gages are effective for obtaining bulk, rather than meso-scale, responses of materials. Effects of strain rate, particle content, average particle size, metallic matrix, etc., on macroscopic strength and failure properties of B_4C/Al composites have been systematically discussed [5–7, 9, 10, 13]. However, meso- and micro-scale dynamic deformation and fracture mechanisms of B_4C/Al composites with different particle contents have not been fully understood.

In situ characterization is critical for a fundamental understanding of deformation and damage of materials [14]. Conventional techniques such as high speed optical imaging are normally applicable to surface measurements and may suffer from poor signal-to-noise ratios during dynamic loading, while synchrotron x-rays are advantageous for high-speed x-ray imaging and diffraction owing to their high intensity and high penetration capability, high coherence, and high pulse repetition rate. Synchrotron x-ray techniques are particularly useful in elucidating microstructural evolution of advanced/novel materials [15–18]. For example, synchrotron x-ray microdiffraction has been widely used to study mechanical properties of multilayer structures (e.g., Cu/Nb and Al/Nb) [17, 19–21] and thin films (e.g., Si and GeTe) [22–24] under quasi-static loading. The elastic-plastic transition and residual stress of samples at lattice scales can be mapped through diffraction scanning [14]. However, it is difficult for microdiffraction to map *in situ* strain fields of materials subjected to dynamic loading at microsecond scales. Recently, ultrafast x-ray imaging and diffraction have been established at the Advanced Photon Source (APS) beamline 32-ID-B under gas gun and Hopkinson/Kolsky bar loading, for dynamic events with durations ranging from 100 ns to 100 μs , and strain rates from 10^2 s^{-1} to 10^6 s^{-1} [25–28].

Local and mesoscopic deformation dynamics can be characterized with two-dimensional (2D) strain field mapping, using optical digital image correlation [29–31] or x-ray digital image correlation (XDIC) [27, 28, 32]. XDIC is essential-

Figure 1: Schematic diagram of the SHTB loading and high-speed x-ray imaging system (a). 1: striker; 2: incident bar; 3: sample; 4: transmission bar; 5: strain gages; 6: collet; 7: x-ray beam; 8: scintillator; 9: high-speed camera. SEM micrographs of B_4C/Al composites with different particle contents: 15 wt% (b); and 30 wt% (c).

ly a combination of high-speed x-ray phase contrast imaging (XPCI) and DIC. It is advantageous for the penetration capabilities of x-rays [22], and particularly useful to image low- Z materials including B_4C . The reinforcing particles naturally serve as speckles for XDIC. Dynamic XDIC provides spatially and temporally resolved information as a valuable complement to post-mortem recovery analysis with scanning electron microscope or SEM [7, 9]. High-speed synchrotron XDIC has been adopted to explore the dynamic deformation of Al, Mg alloys, and granular materials [27, 28, 32]. However, *in situ* measurements on dynamic strain fields in B_4C/Al composites with high-speed XDIC are extremely rare. Previous dynamic tensile experiments on B_4C/Al composite usually yield bulk-scale stress–strain relationships, and micro- or meso-scale deformation dynamics is largely untouched [28, 32].

In the present study, high strain-rate tensile experiments are conducted on B_4C/Al composites using SHTB and high-speed XDIC. Stress–strain curves indicate that the yield strength increases, while the fracture strain decreases, with increasing particle contents. Strain field mapping is carried out between consecutive images, and the means and standard deviations of the strain fields are calculated. The mean tensile strain appears much higher than the mean shear strain, implying that tensile fracture dominates the failure of B_4C/Al composites under dynamic tensile loading. SEM fractographs of the recovered samples indicate that tensile fracture is particle-matrix debonding. The particle-matrix interfaces are nucleation sites for strain localizations, and their propagation and coalescence are impeded by the Al matrix. The reduced spacing between strain localization sites with increasing particle content, facilitates their coalescence and leads to decreased ductility.

2. Materials and methods

The matrix of the B₄C-reinforced composites is 99.9% pure Al. Two different particle contents, 15% and 30% in weight percentage, are explored in our experiments. The reinforcing B₄C particles with a mean particle size of $\sim 50 \mu\text{m}$ are mixed uniformly in molten Al to obtain 15 wt.% (Fig. 1(b)) and 30 wt.% (Fig. 1(c)) B₄C/Al composites, referred to as 15- and 30-B₄C/Al, respectively.

We implement a mini SHTB device along with a high-speed x-ray imaging system at the beamline 32-ID of the Advanced Photon Source. Relevant experimental details were presented elsewhere [26–28, 32]. As shown in Fig. 1(a), a B₄C/Al sample (3) is held between two collets (6). The dimensions of the sample (dog-bone shaped) perpendicular to the x-ray (7) direction are 2.5 mm (gauge length) $\times$ 1.37 mm, and the thickness along the view direction is 0.48 mm. The striker (1), incident bar (2) and transmission bar (4) of SHTB are all made of high-strength steel, and the bar diameter is 6 mm. After the gas gun is fired, the striker impacts the incident bar and generates an elastic wave propagating through the incident bar (along the x -direction, Fig. 1). When the incident wave arrives at the interface between the incident bar and the sample, it is partially reflected owing to impedance mismatch, while the rest is transmitted into the transmission bar. The incident, reflected and transmitted waves are recorded by strain gages (5). The incident wave signal is used to trigger the high-speed camera (9). Upon loading, the x-rays transmitted through the composite sample form images on the scintillator (8) which are captured by the high-speed camera (9) as image sequences. The incident and transmitted waves are used to derive the stress (σ_s), strain (ε_s) and strain rate ($\dot{\varepsilon}_s$) histories of the sample with the following equations [33]

$$\sigma_s = E_t \varepsilon_t \frac{A_t}{A_s}, \quad (1)$$

$$\dot{\varepsilon}_s = \frac{2C_0}{L_s} (\varepsilon_t - \varepsilon_i), \quad (2)$$

$$\varepsilon_s = \frac{2C_0}{L_s} \int_0^t (\varepsilon_t - \varepsilon_i) d\tau, \quad (3)$$

where E is the Young's modulus, A is the cross-sectional area, and C_0 is the elastic velocity of the bar. L_s is the gage length of the sample, and t is time. Subscripts i, t and s denote the incident bar, transmission bar and sample, respectively.

3. Results and discussion

The true stress–strain curves of the two composites, under a strain rate of $\sim 2400 \text{ s}^{-1}$, are shown in Fig. 2(a). The two composites show quite different tensile responses. The yield strength of 30-B₄C/Al is 180 MPa and is much higher than that of 15-B₄C/Al, 70 MPa. Moreover, 30-B₄C/Al shows pronounced

Figure 2: True stress–strain curves (a), and snapshots for 15-B₄C/Al (b) and 30-B₄C/Al (c) composites during dynamic tension. Frames f1–f6 correspond to the instants (solid squares for 15-B₄C/Al and solid circles for 30-B₄C/Al) in (a). The frame intervals are 5.6 μs.

strain hardening before its catastrophic failure, while 15-B₄C/Al exhibits approximately a perfectly elastic-plastic response. 30-B₄C/Al fractures at a strain of 0.05, at a maximum tensile stress of 288 MPa. However, 15-B₄C/Al does not fracture even at a strain of 0.11, and displays better ductility than 30-B₄C/Al. The above results agree with previous low strain rate experiments [4, 6, 13]. In addition, Zhang et al. [4] showed that B₄C/Al with low particle contents (5%–15%) exhibits no obvious rate hardening as manifested in macroscopic stress–strain curves, likely because the matrix Al is not rate-sensitive. However, for high particle-content (e.g., 55%) composites, strain rates show significant influence on failure strains or ductility of materials [7], probably indicating a transition in microdeformation mechanisms with varying loading rates.

High-speed image sequences are acquired with x-ray phase contrast imaging

during tension. The camera frame rate is set at $180,000 \text{ s}^{-1}$ and the exposure time is $0.35 \text{ } \mu\text{s}$. The spatial resolution is approximately $4 \text{ } \mu\text{m}$. The central area ($\sim 1.6 \text{ mm} \times 0.8 \text{ mm}$) of the sample is chosen for imaging. As seen from Fig. 1 insets, the microstructure features are of much smaller scales so this area can represent the whole sample. The B_4C particles appear brighter (white dots) than the Al matrix because the electron density of B_4C is lower than that of Al. The snapshots at different instants (frames f1–f6) for 15- and 30- $\text{B}_4\text{C}/\text{Al}$ are shown in Figs. 2(b) and 2(c), respectively. The images for 30- $\text{B}_4\text{C}/\text{Al}$ are brighter than those for 15- $\text{B}_4\text{C}/\text{Al}$ under the same conditions, as expected. Simple visual observations only reveal cracks starting at frame 5 (f5) after the maximum strain for 30- $\text{B}_4\text{C}/\text{Al}$ (indicated by the arrow), and on the stress plateau (marked by the ellipse) for 15- $\text{B}_4\text{C}/\text{Al}$. The crack starting at frame f5 for 30- $\text{B}_4\text{C}/\text{Al}$ grows and propagates quickly through the sample, inducing an obvious stress drop (Fig. 2(a)).

To better understand “microscopic” deformation dynamics and damage mechanisms of $\text{B}_4\text{C}/\text{Al}$ composites under dynamic tensile loading, XDIC is used to map the strain fields in the samples. Details of XDIC have been given in Ref. [28]. The whole image is chosen as the regions of interest for XDIC analyses. The B_4C particles yield speckles on the images formed via transmitted x-rays, and the edge enhancement by phase contrast also benefits image correlation. The particles inherent in samples produce x-ray speckles [32, 34, 35], so additional marker particles as utilized in optical DIC are not needed. XDIC utilizes x-ray speckles in XPCI images to resolve particle displacements by applying image correlation between two adjacent frames. The Lagrangian normal strains (E_{xx} , E_{yy}) and shear strain (E_{xy}) are calculated from displacement (u) gradients as

$$E_{ij} = \frac{1}{2} (u_{i,j} + u_{j,i} + u_{k,i}u_{k,j}), \quad (4)$$

where $i, j = x, y$.

The strain fields (E_{xx} , E_{yy} , E_{xy}) in composites obtained from four successive moments (frames f3–f6) are presented in Fig. 3. The white patches at frame f6 in Fig. 3(b) refer to the areas with poor correlation. $E_{yy}(x, y)$ and $E_{xy}(x, y)$ are much smaller than $E_{xx}(x, y)$ in terms of amplitude. The tensile strain distributions $E_{xx}(x, y)$ exhibit obvious differences between 15- and 30- $\text{B}_4\text{C}/\text{Al}$. For 15- $\text{B}_4\text{C}/\text{Al}$ (Fig. 3(a)), $E_{xx}(x, y)$ increases gradually with increasing stress or bulk strain. Significant strain localizations appear in $E_{xx}(x, y)$ after frame f4 (marked with dashed ellipses), but scatter throughout the sample. A strain localization area is selected when its $E_{xx}(x, y)$ is greater than a threshold strain of $1.5E_{xx}^{\text{av}}$, the areal mean of $E_{xx}(x, y)$. These localizations grow rapidly in amplitude and area, but do not coalesce. At frame f5, a pronounced tension concentration (damage) appears around the microcrack region indicated in Fig. 2(b). For 30- $\text{B}_4\text{C}/\text{Al}$ (Fig. 3(b)), $E_{xx}(x, y)$ is small and homogeneous across the sample in the elastic deformation stage (frames f3). At frame f4 (after the bulk yield point), scattered, low-amplitude, strain localizations (marked by dashed ellipses as well) appear. However, the average distance between these strain-localization areas is much smaller than that in 15- $\text{B}_4\text{C}/\text{Al}$. The tensile

Figure 3: Time evolution of 2D strain fields E_{xx} , E_{yy} and E_{xy} for 15-B₄C/Al (a) and 30-B₄C/Al (b) composites. Also see Fig. 2.

strain localizations coalesce, at frame f5, into a tensile deformation band resulting in the crack marked in Fig. 2(c). Significant localizations also appear in the $E_{xy}(x, y)$ and $E_{yy}(x, y)$ in corresponding regions after failure begins (f5). However, $E_{yy}(x, y)$, induced by Poisson's effect, is so small that no obvious necking is observed in the sample before failure. This is different from previous observations on metals under tension [36, 37]. At frame f6, all tensile strain concentrates around the macro crack, and strains in other regions relax as a result.

Strain fields are calculated with XDIC at all the instants indicated in Fig. 2(a), squares for 15-B₄C/Al and circles for 30-B₄C/Al. Then, the average values (E_{xx}^{av} , E_{xy}^{av} , and E_{yy}^{av}), and standard deviations (E_{xx}^{sd} , E_{xy}^{sd} , and E_{yy}^{sd}) of the strain components are calculated (Fig. 4) from the strain field maps. The normal and s-shear strain both increase with increasing bulk strain. However, the tensile strain is much bigger than the shear strain (approximately by an order of magnitude), indicating that tension and tensile fracture dominate deformation and failure of the B₄C/Al composites. This is consistent with previous observations that B₄C/Al composites usually fail through particle fracture and particle-matrix debonding [4]. In addition, the average shear strain in 30-B₄C/Al becomes much bigger than that in 15-B₄C/Al upon failure. That explains why the main crack in 30-B₄C/Al is inclined (Fig. 2(c)), rather than remains perpendicular to the loading direction as in 15-B₄C/Al (Fig. 2(b)), due to shear deformation.

Standard deviations of strains can characterize the uniformity of strain fields

Figure 4: Average values (E_{xx}^{av} , E_{xy}^{av} , E_{yy}^{av}) and standard deviations (E_{xx}^{sd} , E_{xy}^{sd} and E_{yy}^{sd}) of the strain fields for 15-B₄C/Al and 30-B₄C/Al composites. The symbols refer to frames f2–f6 in Figs. 2 and 3, in the order of increasing time or strain.

[32]. A high standard deviation indicates a more heterogeneous strain distribution. As shown in Fig. 4(b), the standard deviations of the normal and shear strain fields (E_{xx}^{sd} , E_{xy}^{sd} and E_{yy}^{sd}) for 15- and 30-B₄C/Al are approximately consistent at small bulk strains. Small strain localizations occur in both composites. However, the standard deviations of the strain components for 30-B₄C/Al all show a sharp increase right before the catastrophic failure, and becomes much higher than their counterparts for 15-B₄C/Al, which increase gradually instead. The abrupt strain concentration actually comes from a coalescence of small s-strain localizations (Fig. 3(c)), and contributes to the early tensile fracture of 30-B₄C/Al. This is consistent with the macroscopic or bulk stress–strain curve (Fig. 2(a)) of 30-B₄C/Al, which shows low ductility.

Mechanical properties of particle-reinforced composites strongly depend on inter-particle spacing, which is regarded as an intrinsic length scale [7, 38, 39]. The inter-particle distance decreases with increasing particle fraction. The average distance between particles, λ , is calculated as [7, 38]

$$\lambda = \frac{1 - \phi_v}{\rho_l}, \quad (5)$$

where ϕ_v is the volume fraction of particles (converted from weight fraction in our case), and ρ_l is the line number density of particles intercepting a line of

measurement. The average inter-particle distances in terms of λ are $\sim 83 \mu\text{m}$ and $31 \mu\text{m}$ for the 15- and 30- $\text{B}_4\text{C}/\text{Al}$ composites, respectively, differing by a factor of 2.7.

The hard ceramic particles have high yield strengths themselves, and can block dislocation propagation initiated from the metallic matrix. Therefore, the effective yield strength of a metallic-matrix composite is enhanced relative to the matrix, and increases with increasing particle volume fraction [7]. The mode change in the elastic-plastic transition from the perfect elastic-plastic mode to strain hardening with increasing particle content, appears to attest to the effects of dislocation pileup at the particle-matrix interfaces.

Given the initial heterogeneous microstructures, strain localizations occur near the particle-matrix interfaces in both 15- $\text{B}_4\text{C}/\text{Al}$ and 30- $\text{B}_4\text{C}/\text{Al}$. However, strain localization sites are more “isolated,” or the spacing between them is larger, in 15- $\text{B}_4\text{C}/\text{Al}$ than in 30- $\text{B}_4\text{C}/\text{Al}$ (Fig. 3), because of the larger inter-particle spacing λ for the former composite. As a result, the coalescence of strain localizations becomes more difficult for 15- $\text{B}_4\text{C}/\text{Al}$ since the Al matrix between them serve as a “barrier” to their propagation by diffusing the strain via dislocations in the matrix. This mechanism is also applicable to low strain rates [4, 6]. However, strain localizations in 30- $\text{B}_4\text{C}/\text{Al}$ are more abundant, more closely spaced, and more prone to coalesce into a deformation band and then a macrocrack, given the smaller inter-particle distances, although the strain localizations are less pronounced (in terms of amplitude) compared to 15- $\text{B}_4\text{C}/\text{Al}$ (Fig. 3). The reduced spacing between strain localization sites with increasing particle content facilitates their coalescence, leading to decreased ductility. Therefore, the average inter-particle distance play a crucial role in the strength and ductility of particle-reinforced composites. Optimizing λ via ϕ_v offers one way to design a particle-reinforced, metallic-matrix composite with balanced strength and ductility.

SEM fractographs of the recovered $\text{B}_4\text{C}/\text{Al}$ composites are presented in Fig. 5. Only a small portion of the scanned areas are shown for clarity. B_4C particles are observed on both fracture surfaces of the composites (Figs. 5(a) and (b)) expected for MMCs [4, 6]. However, for 15- $\text{B}_4\text{C}/\text{Al}$, voids and dimples form around B_4C particles during fracture (Fig. 5(a), indicated by dashed circles), indicating ductile void growth and coalescence. On the contrary, few dimples are observed for 30- $\text{B}_4\text{C}/\text{Al}$ but micro cracks form in the Al matrix (Fig. 5(b), arrows), suggesting brittle fracture. Fig. 5(c) shows the details of a dimple in Fig. 5(a) indicated by the solid rectangle. Shear and fibrous zones are observed surrounding the particles, indicating plastic deformation at the particle-matrix interface which consequently contributes to interface debonding and cracking (arrow in Fig. 5(c)). The plastic deformation of the Al matrix leads to better ductility of 15- $\text{B}_4\text{C}/\text{Al}$ (Fig. 2(a)). Fig. 5(d) shows details of the particle in Fig. 5(b) marked by the rectangle for 30- $\text{B}_4\text{C}/\text{Al}$. The stir-casting technique produces compatible particle-matrix interface (dashed rectangles). The interfaces remain “intact” since the Al matrix is subjected to little plastic deformation, while the particle shows fracture features due to strain localization (Fig. 3). This may explain why 30- $\text{B}_4\text{C}/\text{Al}$ exhibits considerable strain hardening but

Figure 5: SEM micrographs of fracture surfaces of the recovered samples: 15-B₄C/Al (a) and (c); 30-B₄C/Al (b) and (d).

low ductility before failure (Fig. 2(a)).

4. Conclusions

In summary, we have performed dynamic tension experiments on 15% and 30% in weight percentage B₄C/Al composites using a split Hopkinson tension bar and high-speed synchrotron x-ray imaging and XDIC. The bulk-scale stress-strain curves and "mesoscale" strain field maps show consistent deformation features. As suggested by the stress-strain curves, higher particle content leads to a higher yield strength but lower ductility. Strain field mapping by XDIC demonstrates that the average tensile strain is much bigger than the shear strain, so tensile deformation and fracture dominate deformation and failure of B₄C/Al composites. SEM fractographs of the recovered samples show voids and dimples in 15-B₄C/Al, indicating plasticity and ductile fracture. On the other hand, microcracks and broken particles in 30-B₄C/Al is characteristic of brittle fracture. Therefore, the particle-matrix interfaces are barriers to dislocation propagation and thus boost yield strength. They are also nucleation sites for strain localizations. The propagation and coalescence of strain localizations are impeded by the Al matrix, and benefit from smaller spacing between strain localization sites, eventually from smaller inter-particle spacing. The reduced spacing between strain localization sites with increasing particle content facilitates their coalescence, leading to decreased ductility. Designing a

particle-reinforced, metallic-matrix composite with balanced strength and ductility should consider optimizing the inter-particle distance as a key parameter.

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References

- [1] M. Kouzeli, L. Weber, C. S. Marchi, A. Mortensen, Influence of damage on the tensile behaviour of pure aluminium reinforced with ≥ 40 vol. pct alumina particles, *Acta Mater.* 49 (18) (2001) 3699–3709.
- [2] S. Ozden, R. Ekici, F. Nair, Investigation of impact behaviour of aluminium based SiC particle reinforced metal–matrix composites, *Compos Part A* 38 (2) (2007) 484–494.
- [3] D. Miracle, Metal matrix composites—from science to technological significance, *Compos. Sci. Technol.* 65 (15) (2005) 2526–2540.
- [4] H. Zhang, M. W. Chen, K. T. Ramesh, J. Ye, J. M. Schoenung, E. S. C. Chin, Tensile behavior and dynamic failure of aluminum 6092/B₄C composites, *Mater. Sci. Eng. A* 433 (1) (2006) 70–82.
- [5] K. B. Lee, H. S. Sim, S. Y. Cho, H. Kwon, Tensile properties of 5052 Al Matrix composites reinforced with B₄C particles, *Metall. Mater. Trans. A* 32 (8) (2001) 2142–2147.
- [6] P. Zhang, Y. L. Li, W. X. Wang, Z. P. Gao, B. D. Wang, The design, fabrication and properties of B₄C/Al neutron absorbers, *J. Nucl. Mater.* 437 (1) (2013) 350–358.
- [7] B. Liu, W. M. Huang, L. Huang, H. W. Wang, Size-dependent compression deformation behaviors of high particle content B₄C/Al composites, *Mater. Sci. Eng. A* 534 (2012) 530–535.
- [8] X. G. Chen, M. da Silva, P. Gougeon, L. St-Georges, Microstructure and mechanical properties of friction stir welded AA6063–B₄C metal matrix composites, *Mater. Sci. Eng. A* 518 (1) (2009) 174–184.
- [9] I. Sabirov, O. Kolednik, Local and global measures of the fracture toughness of metal matrix composites, *Mater. Sci. Eng. A* 527 (13) (2010) 3100–3110.

- [10] J. Ye, B. Q. Han, J. M. Schoenung, Mechanical behaviour of an Al-matrix composite reinforced with nanocrystalline Al-coated B₄C particulates, *Philos. Mag. Lett.* 86 (11) (2006) 721–732.
- [11] R. U. Vaidya, S. G. Song, A. K. Zurek, Dynamic mechanical response and thermal expansion of ceramic particle reinforced aluminium 6061 matrix composites, *Philos. Mag. A* 70 (5) (1994) 819–836.
- [12] N. Behm, H. Yang, J. H. Shen, K. Ma, L. J. Kecskes, E. J. Lavernia, J. M. Schoenung, Q. M. Wei, Quasi-static and high-rate mechanical behavior of aluminum-based MMC reinforced with boron carbide of various length scales, *Mater. Sci. Eng. A*.
- [13] I. S. Lee, C. J. Hsu, C. F. Chen, N. J. Ho, P. W. Kao, Particle-reinforced aluminum matrix composites produced from powder mixtures via friction stir processing, *Compos. Sci. Technol.* 71 (5) (2011) 693–698.
- [14] C. C. Aydiner, D. W. Brown, N. A. Mara, J. Almer, A. Misra, In situ x-ray investigation of freestanding nanoscale cu/nb multilayers under tensile load, *Appl. Phys. Lett.* 94 (3) (2009) 031906.
- [15] M. J. Burek, A. S. Budiman, Z. Jahed, N. Tamura, M. Kunz, S. Jin, S. M. J. Han, G. Lee, C. Zamecnik, T. Y. Tsui, Fabrication, microstructure, and mechanical properties of tin nanostructures, *Mater. Sci. Eng. A* 528 (18) (2011) 5822–5832.
- [16] A. S. Budiman, S. M. Han, N. Li, Q. M. Wei, P. Dickerson, N. Tamura, M. Kunz, A. Misra, Plasticity in the nanoscale cu/nb single-crystal multilayers as revealed by synchrotron laue x-ray microdiffraction, *J. Mater. Res.* 27 (03) (2012) 599–611.
- [17] A. S. Budiman, K. R. Narayanan, N. Li, J. Wang, N. Tamura, M. Kunz, A. Misra, Plasticity evolution in nanoscale cu/nb single-crystal multilayers as revealed by synchrotron x-ray microdiffraction, *Mater. Sci. Eng. A* 635 (2015) 6–12.
- [18] T. Tian, R. Morusupalli, H. Shin, H. Y. Son, K. Y. Byun, Y. C. Joo, R. Caramto, L. Smith, Y. L. Shen, M. Kunz, N. Tamura, A. S. Budiman, On the mechanical stresses of cu through-silicon via (tsv) samples fabricated by sk hynix vs. sematech-enabling robust and reliable 3-d interconnect/integrated circuit (ic) technology, *Proc. Eng.* 139 (2016) 101–111.
- [19] A. S. Budiman, N. Li, Q. Wei, J. K. Baldwin, J. Xiong, H. Luo, D. Trugman, Q. X. Jia, N. Tamura, M. Kunz, K. Chen, A. Misra, Growth and structural characterization of epitaxial cu/nb multilayers, *Thin Solid Films* 519 (13) (2011) 4137–4143.
- [20] Y. Kim, A. S. Budiman, J. K. Baldwin, N. A. Mara, A. Misra, S. M. Han, Microcompression study of al-nb nanoscale multilayers, *J. Mater. Res.* 27 (03) (2012) 592–598.

- [21] R. Shivakumar, S. K. Tippabhotla, V. A. Handara, G. Illya, A. A. O. Tay, F. Novoa, R. H. Dauskardt, A. S. Budiman, Fracture mechanics and testing of interface adhesion strength in multilayered structures—application in advanced solar pv materials and technology, *Proc. Eng.* 139 (2016) 47–55.
- [22] A. S. Budiman, G. Illya, V. Handara, W. A. Caldwell, C. Bonelli, M. Kunz, N. Tamura, D. Verstraeten, Enabling thin silicon technologies for next generation c-si solar pv renewable energy systems using synchrotron x-ray microdiffraction as stress and crack mechanism probe, *Sol. Energy Mater. Sol. Cells* 130 (2014) 303–308.
- [23] K. N. Rengarajan, I. Radchenko, G. Illya, V. Handara, M. Kunz, N. Tamura, A. S. Budiman, Low stress encapsulants? influence of encapsulation materials on stress and fracture of thin silicon solar cells as revealed by synchrotron x-ray submicron diffraction, *Proc. Eng.* 139 (2016) 76–86.
- [24] C. Y. Khoo, H. Liu, W. A. Sasangka, R. I. Made, N. Tamura, M. Kunz, A. S. Budiman, C. L. Gan, C. V. Thompson, Impact of deposition conditions on the crystallization kinetics of amorphous gete films, *J Mater Sci* 51 (4) (2016) 1864–1872.
- [25] S. N. Luo, B. J. Jensen, D. E. Hooks, K. Fezzaa, K. J. Ramos, J. D. Yeager, K. Kwiatkowski, T. Shimada, Gas gun shock experiments with single-pulse x-ray phase contrast imaging and diffraction at the advanced photon source, *Rev. Sci. Instrum.* 83 (2012) 073903.
- [26] M. Hudspeth, B. Claus, S. Dubelman, J. Black, A. Mondal, N. Parab, C. Funnell, F. Hai, M. L. Qi, K. Fezzaa, S. N. Luo, W. Chen, High speed synchrotron X-ray phase contrast imaging of dynamic material response to split Hopkinson bar loading, *Rev. Sci. Instrum.* 84 (2) (2013) 025102.
- [27] D. Fan, L. Lu, B. Li, M. L. Qi, J. C. E, F. Zhao, T. Sun, K. Fezzaa, W. Chen, S. N. Luo, Transient x-ray diffraction with simultaneous imaging under high strain-rate loading, *Rev. Sci. Instrum.* 85 (11) (2014) 113902.
- [28] L. Lu, D. Fan, B. X. Bie, X. X. Ran, M. L. Qi, N. Parab, J. Z. Sun, H. J. Liao, M. C. Hudspeth, B. Claus, K. Fezzaa, T. Sun, W. Chen, X. L. Gong, S. N. Luo, Note: Dynamic strain field mapping with synchrotron X-ray digital image correlation, *Rev. Sci. Instrum* 85 (7) (2014) 076101.
- [29] S. A. Hall, D. M. Wood, E. Ibraim, G. Viggiani, Localised deformation patterning in 2D granular materials revealed by digital image correlation, *Granul. Matter* 12 (1) (2010) 1–14.
- [30] V. Tiwari, M. A. Sutton, S. R. McNeill, Assessment of high speed imaging systems for 2D and 3D deformation measurements: methodology development and validation, *Exp. Mech.* 47 (4) (2007) 561–579.

- [31] J. Y. Huang, Y. Li, Q. C. Liu, X. M. Zhou, L. W. Liu, C. L. Liu, M. H. Zhu, S. N. Luo, Origin of compression-induced failure in brittle solids under shock loading, *Phys. Rev. B* 92 (14) (2015) 144101.
- [32] J. Y. Huang, L. Lu, D. Fan, T. Sun, K. Fezzaa, S. L. Xu, M. H. Zhu, S. N. Luo, Heterogeneity in deformation of granular ceramics under dynamic loading, *Scripta Mater.* 111C (2016) 114–118.
- [33] H. Kolsky, *Stress waves in solids*, Vol. 1098, Dover, 1963.
- [34] S. A. McDonald, L. C. R. Schneider, A. C. F. Cocks, P. J. Withers, Particle movement during the deep penetration of a granular material studied by X-ray microtomography, *Scripta Mater.* 54 (2) (2006) 191–196.
- [35] C. M. Kong, J. J. Lannutti, Localized Densification during the Compaction of Alumina Granules: The stage I–II transition, *J. Am. Ceram. Soc.* 83 (4) (2000) 685–690.
- [36] A. Rusinek, R. Zaera, J. R. Klepaczko, R. Cheriguene, Analysis of inertia and scale effects on dynamic neck formation during tension of sheet steel, *Acta Mater.* 53 (20) (2005) 5387–5400.
- [37] S. Osovski, D. Rittel, J. A. Rodríguez-Martínez, R. Zaera, Dynamic tensile necking: influence of specimen geometry and boundary conditions, *Mech. Mater.* 62 (2013) 1–13.
- [38] M. Kouzeli, A. Mortensen, Size dependent strengthening in particle reinforced aluminium, *Acta Mater.* 50 (1) (2002) 39–51.
- [39] Y. W. Yan, L. Geng, A. B. Li, Experimental and numerical studies of the effect of particle size on the deformation behavior of the metal matrix composites, *Mater. Sci. Eng. A* 448 (1) (2007) 315–325.