We report a class of multifunctional metal matrix composite (MMC) materials that combine structural and functional properties, potentially providing significantly improved protection against space environmental hazards, without the punishment of increasing weight and size or poor scalability. Formed by a scalable friction stir additive manufacturing (FSAM) process, these MMCs are incorporated with a high level of uniformly distributed ceramic or metallic particles at a fraction of greater than 30%. The microstructures of the metal matrices between these added particles are significantly refined by the FSAM process as well as by the presence of large amounts of the particles, e.g., interparticle space of down to less than 1 μm in aluminum MMCs. Consequently, a combination of this high concentration of ceramic and metallic particles and the refinement of the MMC matrix by the FSAM process results in not only enhancing mechanical properties, e.g., hardness and resistance to wear but also embedding functionalities of these incorporated particles in the MMCs. These embedded functional properties can be controlled to provide effective shielding of particle radiation, improved tolerance to high temperature, increased friction force at contact surfaces, etc., which are critical to mitigate the hazards of the space environment.

Keywords additive manufacturing, friction stir, metal matrix composite, multifunction, space

1. Introduction

All of space exploration, from flying satellites at low-Earth-orbits (LEOs), interplanetary space-flighting, to living and traveling on Moon or Mars, faces the same challenges in protecting humans and hardware against a number of space environmental hazards and threats, including ionizing radiation, charged particles, solar ultraviolet (UV) radiation, extreme temperatures, heat and cold cycling, impact from micromete- roids and orbital debris (MMOD), and degradation of materials by arcing, corrosion, and contamination induced by the presence of ions and free electrons (Ref 1-4). The existing protection relies on optimizing body structures, such as increasing wall thickness, cladding ceramics externally, and lining polymers internally (Ref 4-6). Any future enhancement of protection by this approach, however, is severely restricted by limitations on weight and size, and the lack of scalability of the materials and their fabrication toward high technology and manufacturing readiness levels. Obviously, lightweight and compact protection solutions, based on new materials with significant improvements in the functionalities for shielding from radiation, tolerance to thermal cycles, resistance to impact, temperature extremes, corrosion by atomic oxygen attacks, etc., have to be explored and developed.

Aluminum wrought alloys of the 6xxx and 7xxx series are widely used in constructing spacecraft owing to their light-weight, excellent mechanical properties and their ability to withstand the attacks by atomic oxygen, but only provide a benchmark shielding protection against the space ionizing particle radiation (Ref 6-8). In contrast, boron-containing ceramic particulates, such as B₄C (Ref 9, 10) and BN (Ref 11-13) embedded in polyethylene polymer matrix composites (PMCs) as reinforcements, show superior performance to shield neutron and proton radiation after hydrogenation compared with these Al alloys, which possess high hardness that is essential for improving impact resistance. However, a lack of mechanical performance and the resistance to atomic oxygen corrosion of these PMCs, has prevented them from replacing Al alloys in the external body structures of spacecraft. In addition, the deployment of these PMCs as internal linings is limited due to the additional weight and bulky size of these PMCs. On the other hand, MMCs consisting of a metal alloy matrix and ceramic particles distributed in the matrix have demonstrated the capability to combine advantages of the Al alloys and the functional ceramic particles. For example, aluminum (Al) alloy-based MMCs reinforced with carbide and graphite particles have been shown to have improved both mechanical and functional properties, such as high hardness, resistance to heat, low coefficient of thermal expansion (CTE), and low wear losses (Ref 14, 15). These improvements are directly correlated with the high melting point, high hardness, and other intrinsic functional properties of these ceramic particles. These MMCs,
however, have yet been deployed to replace Al alloys as thin-wall panels, enclosures, and frames in constructing spacecraft, due to the limitation of their fabrication methods to form required microstructures and thin-wall products.

Currently, a majority of the lightweight alloy-based MMCs like Al and Mg matrix composites are produced by liquid-state routes, such as stir casting, squeeze casting, and infiltration (Ref 16-19). Stir casting is currently the most popular commercial route for producing aluminum MMCs since its cost is lower than others (Ref 19-21). As a result of the increased viscosity in MMC melts with the addition of ceramic particulates, these fabrication methods are only suitable for manufacturing thick-wall and large-size components to replace traditional cast products, but not for fabricating the thin-wall MMC materials that are needed for constructing spacecraft panels, frames, etc. Beside of only allowing large wall-thickness products, the MMC microstructures contain coarse aluminum phases that are accompanied by the formation of relatively large secondary intermetallic particles in the grain boundaries that often induce crack initiation, especially at elevated temperature. The most challenge in stir casting is the difficulty to distribute the added ceramic particulates uniformly. Therefore, the level of the ceramic addition is capped mostly at 20-25 vol% in a practical stir casting operation, due to low cooling rate and difficult dispersion of micro/nano-reinforcements. In addition, the high temperature liquid-state processing can also develop a brittle interfacial layer between the involved phases, which adversely affects the composite properties (Ref 16). Consequently, the failure strain of the composite is significant reduced, and it becomes almost impossible to reduce gauge of the as-cast MMC ingots or plates to thin-walled MMC products by secondary processing, such as rolling and extrusion.

Different to the liquid-state MMC manufacturing processes, solid-state processes to form MMCs from metal and ceramic powders have also been explored, ranging from powder metallurgy (PM), compact rolling, to cold spray. The most common solid-state process is based on the PM route, which can be used to produce micro- and nanoparticle reinforced MMCs (Ref 22). This method allows for both near-net-shape parts and functionally graded materials by using a progressive increment or reduction of reinforcement volume in specific regions of the part to be produced. However, the drawback of PM is much higher cost of the process than the liquid-state casting processes and has the same difficulty in controlling the reinforcement dispersion when the fraction of reinforcement is high (Ref 23).

Compared with the PM route when manufacturing pure metal materials, solid-state deposition (i.e., the cold spray) has the advantage of low-cost due to its high production efficiency. When producing MMCs, however, rebounding of hard ceramic particles off the metal matrix after impact limits the efficiency for depositing ceramic particles to about 10-20% in Al alloy matrix (Ref 24). Consequently, producing high-fraction MMCs by the cold spray deposition would be accompanied by either huge wasting of ceramic particles or long duration of deposition operation. In addition, the amount of porosity defects in cold-sprayed aluminum MMCs increase significantly, e.g., to 3-5% or greater if the fraction of the SiC in Al matrix is more than 20-30 vol% (Ref 25). Wedged between low efficiency in depositing ceramic particles and large increase of porosity defects, the cold spray deposition is only considered to be sustainable for producing MMCs if the fraction of the ceramic particles in the metal matrix is 10-20% or lower.

Another solid-state process under investigation to form MMCs is friction stir processing that is based on the same principle as friction stir welding, i.e., mixing the materials by severe plastic deformation with the aid of friction-generated heat (Ref 26-28). During this MMC forming process, ceramic powders are pre-packed in either drill-holes (Ref 29) or cut-grooves (Ref 30) on metal alloy substrates. A series of friction stir forming passes are then applied over these grooves or holes to mix the ceramic powders into the metal substrates. While the formed surface of MMCs have a zone of uniformly distributed added ceramic particles, its thickness is only of about a few millimeters and the loading of the ceramic addition in the zone is often less than 10% as a result of the limitation on sizes of the grooves/holes and low packing densities of the loosed powders (Ref 29, 30). These limitations indicate the difficulty to scale up MMC production using the above methods to either produce large-dimensional panel components or optimize functional properties with a high-ceramic loading.

In comparison, a variation on solid-state friction stir processing, i.e., friction stir additive manufacturing (FSAM), has achieved scalability for pure metal alloys (Ref 31-36). During FSAM processing, a large-size component is built by either friction stir joining stacked layers of the materials (Ref 32, 33) or by transferring input metal from a consumable filler through a hollow tool onto the substrate (Ref 35, 36). Until recently, there had been very limited reporting (Ref 37-39) on the use of FSAM processes for fabricating large-dimension MMCs. The results shown in these reports indeed indicate the potentials of the FSAM process to manufacture MMCs with a high-ceramic particle loading of the large area and thickness. Particularly in the case of the stacked FSAM process, its scalability has been demonstrated in two aspects. Firstly, MMCs have been fabricated in an automation operation at a build rate of greater than 1080 cm$^3$/h (Ref 39) that is comparable to FSAM process for aluminum alloys, e.g., 1000 cm$^3$/h (Ref 31, 35), but significantly higher than FSAM build rate of mild steel of 7-300 cm$^3$/h (Ref 40, 41). Secondly, MMCs fabricated by the stacked FSAM process have been readily integrated with other metallic materials, minimizing barriers to integrating MMCs into large structures.

In this study, we aim to explore the potential of aluminum alloy MMCs (AMMCs) fabricated by the scalable stacked FSAM process with a high fraction of ceramic powders, to create a composite with properties which mitigate the hazards in the space environment, such as shielding ionizing particle radiation and resistance to impact. AMMCs reinforced by ceramic powders were firstly characterized using optical microscopy and scanning electron microscopy (SEM) with energy-dispersive x-ray spectroscopy (EDS) for microstructure evaluation. Then hardness measurements and wear tests as well as proton radiation tests and modeling relevant to the low-Earth-orbit (LEO) radiation environments were carried out to correlate the microstructures to the functionalities that are desirable for application in space.

2. Experiments

As shown in Fig. 1, the ceramic materials used in the stacked FSAM process for reinforcing the AMMCs are magnesium diboride (MgB$_2$) and silicon carbide (SiC) particles. The MgB$_2$ particle feedstock comprises a large amount of
Fig. 1  SEM images and particle size distributions (PSD) of the as-received (a, b) MgB$_2$ agglomerates, (c, d) SiC powders, (e, f) Al powders, and (g, h) A357 alloy powders
agglomeration with irregular particle shapes. The particle size
distribution (PSD) of the MgB$_2$ agglomerates is broad, from
about 1 to 150 $\mu$m with 50% of the agglomeration ($d_{50}$) less
than 14 $\mu$m (Fig. 1b). In comparison, the SiC particles have the
angular/elongated shapes that are less irregular than the MgB$_2$
agglomerates and also have a narrower PSD ranging from 10
$\mu$m to 30 $\mu$m with a $d_{50}$ of 13 $\mu$m (Fig. 1c and d). Metal
materials for the AMMC matrix include pure Al powders with a
PSD of 45-150 $\mu$m (Fig. 1e and f), A357 (Al-7%Si-Mg)
powders with a PSD of 20-60 $\mu$m (Fig. 1g and h), and the A357
alloy ingots. Metal substrates that the AMMCs were built on in
the stacked FSAM process are Al metal and 6XXX Al alloy
sheets of 5.0 mm thickness. Handling of the metal powder
materials was carried out under insert argon atmosphere.

Using the A357/SiC AMMC as an example, before the start of
the FSAM operation, A357/SiC AMMC sheets of 4~6 mm
thickness are prepared by either liquid-state solidification (e.g.,
casting) or solid-state powder metallurgy (PM) processes (Fig. 2a).
This A357/SiC AMMC sheet is then positioned on top of the
6XXX alloy substrate. During the FSAM operation as shown in
Fig. 2 (c), a friction stir tool, rotating at 500 to 1,000 revolution per
minute (rpm) and titled ~3 degrees vertically, plunges into the top
A357/SiC AMMC sheet at one end and penetrates about 1 mm into
the surface of the bottom 6XXX alloy substrate. The rotating stir
tool then travels on the surface A357/SiC AMMC sheet at speed of
100-200 mm per minute toward the other end of the sheet,
completing one friction stir pass. These FSAM procedures are
repeated to leave a number of parallel and overlapping tracks over
the entire surface area of the A357/SiC AMMC sheet.

Typical surface tracks and cross-sectional macro-features of
the AMMC samples produced by the stacked FSAM are shown
in Fig. 3 (a) and (b) indicating that no macro-size defects are
present. Figure 3 (c) shows an optical micrograph showing
microstructures of the AMMC layer and the joint between the
top AMMC layer and the bottom A357 alloy substrate. There
are no defects observed such as cracking, delamination, or
porosities either along the joint or between the FSAM tracks.

For comparison of the FSAM microstructures with other
AMMCs produced by different fabrication methods, AMMC
samples were also sectioned from stir-cast A357/SiC ingots and
cold-sprayed A357/SiC plates. Cold spray is a solid-state
process that involves exposing a substrate to a high velocity
(300 - 1200 m/s) jet of small particles. Powders are injected into
a heated nozzle where they are accelerated by a supersonic gas
stream at a temperature well below their melting point. The powders are carried by pressured inert gasses, travel at supersonic speeds, and impact onto a surface to form semifinished or finished products. The A357/SiC plate was formed by co-spraying A357 and SiC powders that are injected by different powder feeders at a deposition rate of ~2.69 g/min or less for producing ~20 vol% SiC composites. This deposition rate is lower than for cold spraying pure metal alloys by up to 50%, highlighting the difficulty to incorporate high fraction of ceramic particles into a metal matrix by cold spray.

Fig. 3 Optical images of the A357/SiC AMMC sample produced by the stacked FSAM process. (a) The sample surface with parallel friction-stirred tracks, (b) the cross section of the sample showing the two FSAM layers of AMMCs on the bottom 6XXX alloy substrate plate, and (c) the cross section showing the microstructures of a single top AMMC layer on a 6XXX alloy substrate plate.
3. Results and Discussion

3.1 Fraction and Distribution of Added Particles

In metal matrix composites, particles include externally added ones such as the SiC particles in the A357/SiC MMCs and internally formed ones, e.g., eutectic Si particles and other minor secondary phases that are parts of microstructures of the A357 alloy.

The effects of the stacked friction stir additive manufacturing (FSAM), compared with other fabrication methods, on the fraction and distribution of the added SiC particles are shown in Fig. 4. These optical micrographs were taken on the cross sections of A357/SiC composite samples fabricated by FSAM, commercial stir casting, and cold spray deposition, respectively. As shown in Fig. 4 and Table 1, the fraction of the added SiC particles in A357 alloy matrix of the AMMCs produced by the FSAM process can be more than 30%. The fraction of all the particles including the added SiC particles, the eutectic Si and other secondary particles in the cast A357 alloy (Ref 49) is up to 40%. The distribution of these SiC and secondary particles was highly uniform, and the interparticle spacing between the SiC and secondary particles was measured at about 17 μm (Table 1).

In contrast, in commercial AMMC ingots with the same matrix chemistry and the same SiC particle loadings but fabricated by commercial liquid-state stir casting (Fig. 4b), interparticle spacing is up to 48 μm due to the presence of large primary Al dendrites, and the SiC and Si particles cluster in the interdendritic regions. This high fraction of SiC particles, however, cannot be obtained in the AMMCs manufactured by the cold spray deposition. Figure 4(c) gives the fraction of SiC particles in the cold-sprayed AMMC samples at lower than 20% as a significant proportion of the SiC particles bounced off the deposition surface when they impact on the surface of the deposited AMMCs (i.e., low deposition rates). The distribution of these particles in the cold-sprayed sample is not uniform, being segregated between layers of the Al alloys that can be several hundred micrometers in length. Consequently, the interspacing between these added SiC particles is dependent on the original size of the Al particles and their elongation by impact deformation during the spray deposition. As shown in Fig. 4(c), the interparticle spacing in the cold-sprayed A357/SiC samples can be up to 40–50 μm, which is comparable to the size of the Al alloy powders, i.e., 60 μm diameter, if allowed the deformation by spray impact.

The ability of the stacked FASM process to achieve a high fraction and uniform distribution of ceramic particles is also demonstrated in the AMMCs with pure Al as the matrix and MgB2 as the reinforcing particles. Similar to the A357/SiC AMMCs, the area fraction of the MgB2 particles in the Al/MgB2 AMMCs can also be increased to more than 30% (Table 1) and their distribution in the Al matrix is uniform as shown in Fig. 5. Table 1 shows that the measured fraction of the MgB2 particle is lower than the combined fraction of SiC and secondary particles (40%), while similar to that of the SiC particles (30%). However, the measured interparticle spacing for MgB2 is less than 1 μm in the Al/MgB2 sample, which is significantly less than measured interparticle spacing for all the SiC and other particles (~17 μm) in the A357/SiC composite (Table 1).
Apparently, this significant reduction in the interparticle spacing in the Al/MgB\(_2\) composite is accompanied by a decrease in the size of the added MgB\(_2\) particles from the as-received 14 \(\mu\)m \((d_{50})\), which was primarily due to a high degree of agglomeration, instead of separated single particles as in the case of the SiC feedstock. (Fig. 1a and b). Figure 5 indicates that these MgB\(_2\) clusters are broken up into numerous submicron particles as well as some 1-5 \(\mu\)m particles. The majority of smaller particles also become more faceted, indicating that their measured area fraction can be relevant to their volume fraction. Indicated by EDS images of Fig. 5 that some of these broken up particles could be either Mg-rich or B-rich implying there may be some residual Mg metal in the as-received MgB\(_2\) clusters. Beside the deagglomeration of the as-received MgB\(_2\) clusters, the in situ reduction of interparticle spacing between added particles by the FSAM process is also facilitated by its solid-state mixing action. As shown in Fig. 5, these small particles are uniformly distributed in the matrix by FSAM, leading to a significant reduction in the interparticle spacing to less than 1 \(\mu\)m between the MgB\(_2\) particles in the AMMC. In comparison, the SiC size remains almost the same at 13 \(\mu\)m (Fig. 1c, d, and 4a) and the interparticle spacing is moderately refined to 17 \(\mu\)m after FSAM processing.

The stacked FSAM process has demonstrated two unique abilities compared with other fabrication methods, in forming A357/SiC and Al/MgB\(_2\) composites: (i) incorporating a high-level addition of ceramic particles into AMMCs free of coarse microstructures, segregation and other defects and (ii) refining interparticle spacing further by either re-distributing secondary particles (A357/SiC) or breaking up the added ceramic particles (Al/MgB\(_2\)). The uniqueness is resulted by the creation of mass flows in the solid state by the FSAM stirring tool. These mass flows include lateral mass flows accompanying the lateral motion of the stirring tool and vertical mass flows generated by the rotating tool that is tiled (Fig. 2c). Effectively transporting and mixing of particles and matrix materials by these flows are the keys for uniformly incorporating a greater fraction of ceramic particles within a metal matrix. Defects such as segregation, porosities, shrinkage, and cracking that have hampered the production of high-ceramic-loading AMMCs by liquid-state stir casting processes are absent in AMMCs made by solid-state FSAM. In addition, the mixing action of mass flows in the multiple directions can distribute added ceramic particles homogenously, unlike some linear or 2D solid-state processes, e.g., extrusion, rolling, or deposition, which can form layers and bands of segregated ceramic particles in the metal matrices. Furthermore, severe deformation as a result of these mass flows in the FSAM processing can break up primary, secondary, or added particle clusters, refining the metal matrix microstructures.

### 3.2 Effects of High Fraction of Ceramic Particles on Hardness and Coefficient of Friction

Figure 6 includes measurements of the micro-Vickers hardness corresponding to \(\sim 30\%\) of SiC and MgB\(_2\). Such high-level loading of ceramic particles leads to an increase of
the hardness to ~180 HV in both AMMCs. Figure 6 shows a series of HV measurements in the cross sections of hybrid sheets, clearly demonstrating the enhancement of hardness of Al matrix by the addition of ceramic particles. The cross sections include a top AMMC layer, a bottom Al metal/alloy substrate, and a transition region between them. For the hybrid sheet with the A357/SiC composite layer on the top in Fig. 6(a), the microhardness is about 110 HV in the 6xxx Al alloy substrate where there are no ceramic particles added. It increases to 140-150 HV in areas of the transition from the Al alloy to the AMMC where the SiC fraction increases. A maximum measured hardness of 170-180 HV occurs in the top region of the AMMC layer where the fraction of SiC particle is about 30%.

Similarly, Fig. 6(b) shows a series of the microhardness measurements across a cross section of the hybrid sheet with the Al/MgB2 composite layer on top of an Al substrate, which also exhibits the same enhancement of Al metal hardness by a high fraction of MgB2 particles. The hardness measured in the Al metal substrate is between 45 and 55 HV. It was then doubled to more than 120 HV in the Al-to-AMMC transition region due to the presence of MgB2 particles. Middle way through the AMMC layer, the measured hardness reaches 175 HV where the fraction of MgB2 is close to 30%. A maximum hardness of more than 210 HV was recorded at the top of the Al/MgB2 composite layer.

In addition to evaluating the effect of a higher fraction of ceramic particles on hardness, the dependence of the coefficient of friction on the ceramic fraction was also investigated, from the perspective of assessing resistance to sliding of a harder object over the AMMC surface which could potentially break up the protective layers on the Al metal in the surface. Figure 7 shows the result of dry sliding tests on the A357/SiC AMMC samples using a WC pin, indicating the resistance force, presented by the coefficient of friction, is higher with the increase of SiC particle loadings in the AMMCs. There is an 80% increase in the measured coefficient of friction when SiC additions increase from 20 to 32%. Concurrently, mass loss of the A357/SiC AMMC sample decreases from about 7 miligram (mg) to < 1 mg with the increase of the SiC loadings from 20 to 32%.

Compared with the cold-sprayed A357/SiC samples with a SiC fraction of < 20% and Vickers hardness of 80 HV (Table 1), the sevenfold decrease of wear loss and significant increase in the hardness in these A357/SiC AMMCs to a level comparable with some steels can be attributed to the high fraction of the SiC (i.e., > 30 %) homogenously distributed in the metal matrix by FSAM, resulting in a larger reduction in the interparticle spacing. The 10% increase of the fraction of the SiC particles from the cold-sprayed A357/SiC samples appears to be moderate. However, a greater reduction in the interparticle spacing to 17 μm is observed (Table 1), which was facilitated.
by the uniform mixing attainable by the FSAM operation. In comparison, segregation of the SiC particles in the cold-sprayed AMMCs together with a lower fraction of the SiC particles attained (<20%) results in a larger interparticle spacing across the coarse Al grains of 40-200 μm (Table 1).

It is also noted that the hardness of “softer” Al metal and “harder” Al alloy can be equally enhanced to a similar level, or even higher for the Al metal by MgB$_2$. This result indicates that further enhancements to the hardness of the AMMC metal or alloy matrices play a minor role in increasing the overall

Fig. 6  Macrographs of cross sections of hybrid sheets consisting of an AMMC top layer and an Al or Al alloy bottom substrate showing measured microhardness and the measuring locations: (a) A357/SiC on 6XXX Al alloy substrate and (b) Al/MgB$_2$ on Al metal substrate

Fig. 7  Dry sliding test results on two AMMC samples with different SiC loadings
hardness of the AMMCs when the ceramic fraction is >30%.

Another observation is that when the ceramic fraction in the AMMCs is the same, e.g., at 30%, a smaller MgB2 interparticle spacing of less than 1 \textmu m (Table 1), appears to be more effective for increasing the AMMC hardness than larger SiC interparticle spacing of around 17 \textmu m (Table 1). This finding is further supported when considering that the hardness of MgB2 of \sim 1,488 HV (Ref 50) is much lower than that of SiC at \sim 2,563 HV (Ref 46) and that the combined fraction of SiC (~30%) and secondary particles (~10%) at 40% is also higher than the fraction of MgB2 in the matrix. This implies that a reduction in interparticle spacing is more critical in enhancing the hardness of AMMCs compared to a higher fraction of the ceramic particles.

3.3 Mitigating Space Hazards Under the Low-Earth-Orbit Environment: Proton Radiation and Surface Damages

It has been established that incorporating hard and high-melting-point ceramic particles can lead to improved resistances to wear, impact, and heat, as a result of the excellent structural and functional properties of these ceramic materials, e.g., high melting points, extreme high hardness, radiation shielding, etc. (Ref 16, 17), which are relevant for mitigating hazards in the space environment such as launch vibration, extreme temperature cycles, and ionizing particle radiation. In this study, we focus on exploring AMMC materials with a high fraction of ceramic particles for reducing surface damages induced by vibration and proton radiation damage in a LEO environment.

Surface damages during the launches can occur to the folded deployment mechanisms on spacecraft and satellites (Ref 42, 43). These mechanisms can move laterally in oscillatory motions (fretting) between a ball-to-flat contact as a result of extensive vibration during launch. This movement damages and removes the naturally formed protective layer, e.g., oxides on the metal surface, and exposes “fresh” metal underneath. It is discussed in the last Section that an increase of SiC particle loading from 20 to 32% leads to an 80% increase in the coefficient of friction (Fig. 7). A higher coefficient of friction gives a higher friction force for a given normal force at the contact against the motion between a friction pair. This implies that an increase of SiC particle loading from 20 to 32% in the AMMCs increases the resistance against the oscillatory motions (fretting) by 80% in the contacts of the folded deployment mechanisms during launch, potentially minimizing the degree of the motions and thus reducing surface damages.

Proton radiation shielding tests for comparing MgB2 reinforced AMMC with Al and Mg metal utilized the energy range of 10 to 15 MeV that is a part of the proton particle energy spectrum encountered under the LEO environment. The test results in terms of linear energy transfer (LET) and the mass stopping power (MSP) are plotted in Fig. 8. The results show that for a given proton energy the Al/MgB2 sample exhibits a 20-40% improvement of the LET and MSP compared with the Al and Mg metal samples. The SRIM calculation was also carried out over the same energy range and the results are shown in Fig. 9, predicting both pure MgB2 and Al/MgB2 composites have higher LETs than Al and Mg metals. When comparing the two figures, a similar trend was observed between the calculated and the experimental LET/MSP, i.e., the LET of the Al/MgB2 sample is higher than those of Al and Mg metal samples. However, there is a large discrepancy between the measured and calculated improvements of LET by MgB2 in terms of the degree and absolute values of the improvements. The measured LET and MSP of the Al/MgB2 sample are 30-40% and 25-30%, respectively, higher than the Al and Mg metals, while the calculated increases are negligible.

The large degree of enhancement of the LEO radiation shielding by the addition of MgB2 deviates from the established understanding of interactions between proton ion radiation and shielding materials. Traditionally, hydrogen was considered to be the most effective element for scattering and modifying particle radiation such as proton ions (Ref 51). As the hydrogen nucleus is made of a single proton with a mass matching a proton ion, it is considered that energy can effectively transfer to hydrogen nuclei from proton ions in a nuclear collision, especially for the proton radiation in deep space environments in which energy often exceeds 250 MeV. The boron-containing materials are well known for capturing thermal neutrons, such as shielding thermal neutrons at nuclear reactors for storage of...
spent nuclear fuel (Ref 11-15), but not to play a role like hydrogen-rich materials to scatter and modify the incipient proton particle radiation.

Possible mechanisms for improving shielding effectiveness by MgB$_2$ are sought with the consideration of two factors: (i) the lower energy of the LEO proton radiation and (ii) the complex electronic structures of MgB$_2$. Firstly, when the level of the proton radiation energy is of several hundred MeV, the collision of proton ions with the shielding materials nuclei, not their electrons, would be the key interaction for them to lose energy. Under the LEO environment where low-energy protons can be present, it is suggested that the electronic structures of the shielding materials could also participate in transferring energy from the incipient proton radiation. Secondly, MgB$_2$ has a unique layered structure alternating B and Mg in the layers with a complex electronic bonding structure (Ref 52). The formation of the electronic structures involves transferring electron from the Mg to B layer (ionic interlayer bonding), electron sharing (covalent in-plane bonding), and freeing up electrons (metallic in-plane bonding) within the B layer and reduced Mg–Mg bonding distances. As a result of this bond structure, the electronic structure of MgB$_2$ as a whole is extraordinarily responsive to energy propagation through vibrating crystal lattice (i.e., phonons), from which its superconductivity is originated (Ref 52-54). A similar role that the electronic structures may play in modifying the incipient proton radiation at a lower energy level relevant to the LEO radiation should be investigated further in the future.

4. Conclusion

Multifunctional AMMCs materials with a high fraction of ceramic particulate additions (> 30 %) produced by a stacked friction stir additive manufacturing technology (FSAM) provide a new opportunity to attain advanced protection performance against space environmental hazards and threats for wide-range applications in satellites, spacecraft, surface vehicles, human habitats, etc., while achieving lightweight, compact size, and the scalability of the FSAM fabrication to high technology and manufacturing readiness levels.

Incorporating high-fraction, homogenously distributed and extreme hard ceramic particles in AMMCs such as A357/SiC, leads to a significant decrease in interparticle spacing down to a size comparable to the feedstock particle size and to an increase in the hardness comparable to some steels. These AMMCs demonstrate advanced performance in minimizing wear loss and enhancing resistance to lateral motion at a sliding contact, which are to contribute to overcome surface damages to the folded deployment mechanisms during launches.

Embedding desired functionality such as low-Earth-orbit (LEO) proton radiation shielding into these structural materials can be achieved by forming a hybrid sheet consisting of a Al/MgB$_2$ composite layer and Al substrate. The effectiveness of shielding LEO proton radiation by the Al/MgB$_2$ composite in terms of linear energy transfer and mass stopping power is improved by 25-40% compared with Al metal. The interaction between the MgB$_2$ complex electronic structures and the proton radiation is proposed for future study to understand the mechanism for improving the radiation shielding effectiveness.

While the stacked FSAM process demonstrates its scalability in manufacturing aluminum-based MMCs with a high fraction of added functional particles, it is expected, however, that further development of the process is required for fabricating high-melting-point alloy-based MMCs, e.g., titanium (Ti) alloys. The challenges facing friction stir welding of the Ti alloys, such as alloy contamination and oxidation at high operational temperatures, will be multiplied in incorporating a high fraction of ceramic or intermetallic particulate additions into the Ti alloy matrix MMCs by FSAM.

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