

Strength gradient in impact-induced metallic bonding

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Qi Tang¹, David Veyssset^{2,6}, Hamid Assadi³, Yuji Ichikawa⁴ & Mostafa Hassani^{1,5}✉

Solid-state bonding can form when metallic microparticles impact metallic substrates at supersonic velocities. While the conditions necessary for impact-induced metallic bonding are relatively well understood, the properties emerging at the bonded interfaces remain elusive. Here, we use in situ microparticle impact experiments followed by site-specific micromechanical measurements to study the interfacial strength across bonded interfaces. We reveal a gradient of bond strength starting with a weak bonding near the impact center, followed by a rapid twofold rise to a peak strength significantly higher than the yield strength of the bulk material, and eventually, a plateau covering a large portion of the interface towards the periphery. We show that the form of the native oxide at the bonded interface—whether layers, particles, or debris—dictates the level of bond strength. We formulate a predictive framework for impact-induced bond strength based on the evolution of the contact pressure and surface exposure.

Although supersonic impacts are typically associated with destruction^{1,2}, when they occur at small scales, they can be leveraged to build metallic materials and structural components^{3–6}. Under the extreme conditions of deformation and pressure induced by impact^{7,8}, fresh metallic surfaces are generated and metal atoms can be forced into close proximity needed to form instantaneous solid-state metallic bonding⁹. This phenomenon has been observed in space when dust particles collide with space structures at high velocities¹⁰. Additionally, it has been utilized to build up structural coatings and bulk materials in impact welding^{11–16} and cold spray additive manufacturing^{3–6}.

Considerable effort has been dedicated to understanding the mechanisms behind impact-induced bonding, rendering the measurement and prediction of the critical bonding velocity a scientific focal point for decades^{17–20}. The recent advancement in the controlled launching of individual powder particles coupled with ultra-high-speed imaging^{21–26} has enabled resolving the moment of bonding with micrometer-scale and nanosecond-level spatiotemporal resolutions²⁷. The critical bonding velocity can now be precisely measured for a wide

range of materials^{28–31}, particle sizes³², and temperatures³³. The progress in the processing science of impact bonding now calls for a coordinated mechanistic understanding of the properties of impact-induced bonded interfaces, if it is to be used for the repair and manufacturing of safety-critical components.

Here, we offer a paradigm shift from “when does impact lead to bonding?” in metals to “how strong is the bond?” at the micrometer scale. Understanding the micromechanics of individual impact-induced bonded interfaces is a critical starting point for understanding the mechanical behavior of materials manufactured by assembling many such interfaces. The bottom-up approach presented here is different from the measurements of mechanical properties at the deposit level^{6,34–36}, which involves hundreds of thousands of deformed particles, each with unique and distinct^{37,38} kinetic^{39–41} and thermal histories^{42–44}. Consequently, breaking beyond the average-based correlations between impact parameters (e.g., particle size, temperature, velocity, and impact geometry) and bulk properties to uncover the fundamental origins of property development proves

¹Sibley School of Mechanical and Aerospace Engineering, Cornell University, Ithaca, NY 14853, USA. ²Hansen Experimental Physics Laboratory, Stanford University, Stanford, CA 94305, USA. ³Brunel Centre for Advanced Solidification Technology, Brunel University London, Uxbridge UB8 3PH, UK. ⁴Fracture and Reliability Research Institute, Tohoku University, Sendai 980-8579, Japan. ⁵Department of Materials Science and Engineering, Cornell University, Ithaca, NY 14853, USA. ⁶Present address: Wellman Center for Photomedicine, Harvard Medical School, Massachusetts General Hospital, Boston, MA 02114, USA.

✉ e-mail: hassani@cornell.edu

highly challenging. In contrast, we isolate individual interfaces produced by the impact bonding of individual particles with precisely measured or controlled kinetic, thermal, and geometrical parameters to study interfacial strength. We release the assumption of uniform properties that is implicit in the deposit level measurements and reveal that a significant variation in bond strength exists within one single interface.

We resolve the spatial variation of bond strength using *in situ* micromechanical measurements and with micrometer-level precision. These measurements are conducted at various interfacial locations on individual metallic microparticles bonded to metallic surfaces. The bonding occurs through controlled and reproducible launching of single microparticles to supersonic speeds and their subsequent impact on a metallic surface, a process we record in real time. We report a significant gradient of bond strength across impact-induced bonded interface and discuss it in light of the nanoscale interfacial features. We also discuss the role of the spatial and temporal evolution of the contact pressure and the interfacial strain in governing the impact-induced strength and its gradient.

Results

In situ impact and bond strength measurements

We used laser-induced microparticle impact test (LIPIT) (Fig. 1a) to produce bonded Al particles onto Al substrates with precisely measured particle sizes and velocities. Figure 1b, c shows exemplar images resolving the rebound and bonding of two Al particles impacting an Al substrate at 752 and 1066 m/s, respectively. Having conducted the particle impact experiments in a site-specific fashion, we identified the bonded particles (Fig. 1d) and cross-sectioned them to reveal the interface (Fig. 1e). A distinctive boundary, characteristic of solid-state bonding, is clearly visible between the particle and substrate. We used micro-machining to fabricate microtensile test specimens with the particle-substrate interface located in the gauge section (Fig. 1f). Microtensile tests were conducted using a micromechanical stage

inside a scanning electron microscope (SEM). The peak on the load-displacement curve (Supplementary Fig. 1a) corresponds to the moment of fracture at the interface, which is further confirmed through subsequent SEM observations and elemental mapping (Fig. 1f and Supplementary Fig. 7). The peak load was used to measure the interfacial bond strength. To exclude the effect of the interfacial angle from the measurements, we decomposed the fracture load into the normal and tangential components and then used both normal and shear stresses in the Tsai-Hill failure criterion⁴⁵ to determine a site-specific bond strength (see Supplementary Information Section 1 for details).

Gradient of bond strength

Figure 1g shows impact-induced bond strength as a function of the location along the particle-substrate interface. The lowest bond strength was measured at the location closest to the impact center. Moving away from the center, we measure a rapid twofold increase in bond strength which is then followed by a plateau closer to the edge of the particle. These measurements provide direct evidence confirming the gradient of bond strength along microparticle impact-induced bonded interfaces. Previous microstructural characterizations indicate that micro- and meso-scale microstructural activities intensify progressively from the impact center to the edge^{46–49}. Specifically, higher levels of dislocation density⁴⁶, grain refinement^{50,51}, and dynamic recrystallization⁵² have been reported in regions closer to the edge compared to those near the center. The measured gradient of the local bond strength is therefore in line with the general expectation that regions nearer to the edge should exhibit relatively stronger bonding. However, considering the trend implied by microstructural activities⁴⁸, it is now interesting to note the emergence of a sharp rise and thereafter a substantial plateau covering more than 60% of the interfacial arc length.

The role of contact pressure and surface exposure

Two mechanisms must work in tandem to achieve metallurgical bonding upon impact (Supplementary Fig. 4). First, shear-induced

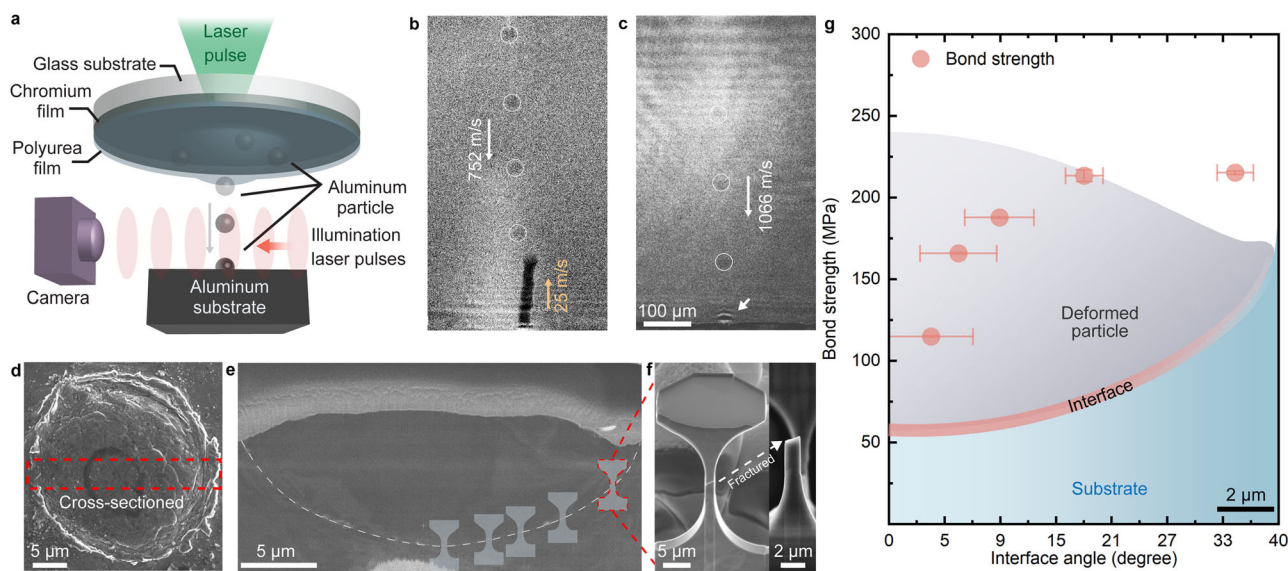


Fig. 1 | *In situ* single particle impact test and site-specific measurements of bond strength. **a** Schematic of the laser-induced microparticle impact testing (LIPIT) setup. Multi-exposure images showing **b** a 22- μ m Al microparticle impacting and an Al substrate at 752 m/s and rebounding and **c** a 22- μ m Al microparticle impacting an Al substrate at 1066 m/s and bonding to it. **d** Top view of an Al particle bonded onto an Al substrate at 1100 m/s. **e** Cross-sectional view of the Al particle shown in (**d**) showing the bonded interface and locations of microtensile specimens on the interface. **f** A microtensile specimen micromachined from bonded particles, and then fractured at the interface during microtensile testing. **g** Experimental

measurements of bond strength at the micron scale reveal a gradient of bond strength across impact-induced bonded interfaces for 23 ± 2 μ m Al particles impacting and bonding onto an Al substrate at 1100 ± 41 m/s. The error bars in *x* coordinate on the experimental data points are the width of the microtensile specimens at the gauge section. The error bars in *y* coordinate on the experimental data points are calculated from the resolutions of microtensile load recording and SEM characterization of specimen geometry. The geometry of the deformed particle and substrate was superimposed on the plot as colored regions to enhance clarity. Source data are provided as a Source Data file.

lateral expansion of contacting surfaces fractures the native oxide layer and exposes the underlying fresh metallic surfaces. Subsequently, the impact-induced pressure brings these clean surfaces into atomic-scale proximity, enabling metallurgical bonding. Multiple mechanisms cooperate to facilitate the lateral expansion of the contacting surfaces, including pressure-gradient-induced jetting^{33,54} and adiabatic shear instabilities¹⁷. Here, we introduce surface exposure, i.e., the ratio of the exposed surface area to the total surface area, as a quantitative measure of the exposed fresh metallic surface that can be used to predict the local bond strength. We also define effective pressure as the difference between the impact-induced pressure and the critical pressure required to initiate the extrusion of the exposed metal through the gaps in the fractured native oxide layer. The critical pressure depends on the instantaneous flow stress and the local surface exposure (Eq. S20). From this perspective, the instantaneous strength of an impact-induced bond (σ_B) can be shown to be proportional to the product of the local effective pressure (P_{eff}) acting on contacting surfaces and the local effective surface exposure (Y_{eff} , see Eq. 1 and Eqs. S15–18). From a physical point of view, the interfacial strength cannot exceed the instantaneous flow stress of the deformed base metal at the moment of bonding. As such, we impose the instantaneous flow stress of the base metal (σ_{fb}) as an upper bound to the local bond strength. Lastly, we introduce the ratio of the instantaneous flow stress during debonding (σ_{fd}) to that at the moment of bonding (σ_{fb}) to account for the significantly different deformation rates and temperatures during the impact and microtensile testing.

$$\sigma_B = \frac{\sigma_{\text{fd}}}{\sigma_{\text{fb}}} \times P_{\text{eff}} \times Y_{\text{eff}} \quad (1)$$

We have recorded contact pressure, surface exposure, and instantaneous flow stress for every interfacial element during the entire course of impact-induced deformation with finite element simulations (see Supplementary Information Section 2 for details). For each interfacial element, we monitored the temporal evolution of σ_B and considered the peak value to be the local interfacial strength (see Supplementary Information Section 3 for details).

Figure 2a shows the variation of the local bond strength across the interface as predicted by the simulations. The predictions are consistent with the experimental measurements and provide a more complete picture of the bond strength gradient across the interface. We identify the central regions (up to a normalized interfacial arc length of -0.16) as the weakest segment, showing no significant variation in bond strength. Following this region, we observe a sudden twofold increase in bond strength, spanning from the normalized interfacial length of -0.16 to -0.32. We observe a plateau-like behavior with a slight decline in bond strength for normalized interfacial arc lengths beyond -0.32 towards the edge.

The characteristics of the bond strength gradient can be elucidated by examining the spatial distribution and temporal evolution of the contact pressure and the surface exposure across the interface (Fig. 2d). A localized surface opening emerges in the very early stage of impact (1 ns). Notably, this opening occurs neither at the center of impact nor at the periphery but rather in between, approximately 3 μm away from the center. At this moment, the contact pressure is relatively high (~5 GPa), resulting in the highest bond strength that persists until the end of the deformation. As deformation progresses, surface exposure increases, and the location of the peak surface exposure shifts toward the edge. While higher surface exposures favor the formation of stronger bonds, this effect is counterbalanced by a continuous decline in the contact pressure. In this case, the latter slightly outweighs the former, resulting in a subtle decline towards the edge, though the bond strength remains consistently high across the outer 60% of the interface. In the central regions, surface exposure remains relatively low for several nanoseconds, during which the contact

pressure significantly drops (see $t = 6$ ns). Consequently, the bond strength remains low in the central segment.

To further validate our predictive framework, we conducted bond strength measurements for two additional sets of particle size and impact velocity: $42 \pm 3 \mu\text{m}$ at 860 ± 20 m/s and $20 \mu\text{m}$ at 862 m/s. These parameters were selected to also enable a systematic variation in particle size and impact velocity when combined with the initial set ($23 \pm 2 \mu\text{m}$ at 1100 ± 41 m/s). Figure 2b, c shows the general characteristics of the bond strength gradient discussed earlier for the initial set. In these two additional cases as well, the central segment exhibits low-strength bonding due to the limited surface exposure. A localized surface opening is observed in the very early stage of deformation (Fig. 2e 2 ns for the $42 \mu\text{m}$ particle and Fig. 2f 1 ns for the $20 \mu\text{m}$ particle), contributing to the formation of the strongest bond in each case. Subsequently, the surface opening expands outward, forming a strongly bonded interface towards the edge. The reasonable agreement between the predictions and the measurements across a range of impact velocities and particle sizes further validates the developed framework to predict impact-induced bond strength.

All these three cases result in a maximum bond strength significantly higher than the yield strength of bulk Al which we have measured to be ~115 MPa using the same microtensile specimen geometry as the one used for interfacial strength measurements (Supplementary Fig. 8). We attribute such significant strengthening at the interface to significant dislocation activities caused by the ultra-high strain rate and severe plastic deformation^{49,55} which is captured in the instantaneous flow stress term in our model. What is more, we note a comparable level of peak bond strength (220–233 MPa) in these cases despite variations in particle size and velocity. At large strains and under adiabatic conditions of impact^{56–58}, the rate of dislocation generation can be counterbalanced by the recovery processes⁵⁹ limiting the attainable dislocation density in the fourth stage of hardening⁶⁰ and giving rise to a “saturation” in strengthening^{61,62}.

The data presented in Fig. 2b, c indicate that maintaining a constant impact velocity while doubling the particle size does not significantly alter the distribution of bond strength along the interface. Specifically, in both cases, the central region of the interface exhibits similar bond strengths lower than that of bulk Al. The transition to the high-strength plateau begins at a normalized interfacial arc length of approximately 0.15 and extends only to around 0.9. Conversely, changing the impact velocity has a more pronounced effect on the interfacial bond strength distribution. Figure 2b, c shows that increasing the impact velocity from 862 m/s to 1100 m/s results in ~25% increase in bond strength within the central weakly bonded region. Additionally, this increase in impact velocity extends the high-strength plateau region beyond the normalized arc length of 0.9, all the way to the edge of the interface. We attribute the former to the higher impact-induced pressure (Supplementary Fig. 9) and the latter to the greater pressure gradient, leading to more severe surface exposure and enhanced upward flow at the particle-substrate periphery.

Distribution of native oxide layer at the interface

We used cross-sectional transmission electron microscopy (TEM) to study the state of intimate contact at the bonded interface. Our analysis reveals three distinct forms of native oxide presence at the bonded interface in Fig. 3: (1) nanometer-thick oxide layers separating the particle and substrate surfaces, (2) oxide particles approximately of size on the order of ~10 nm, and (3) scattered oxide debris of only several nm in size each visible across the interface with the bright contrast. Elemental mapping along a line perpendicular to the nanometer-thick oxide layers (Fig. 3a) reveals a pronounced peak in oxygen content, coupled with a decrease in aluminum content. In contrast, Fig. 3b shows native oxide particles ~16 nm in size at the interfaces, with the presence of oxygen still detectable via elemental mapping, albeit to a lesser degree. In extreme cases, we observe

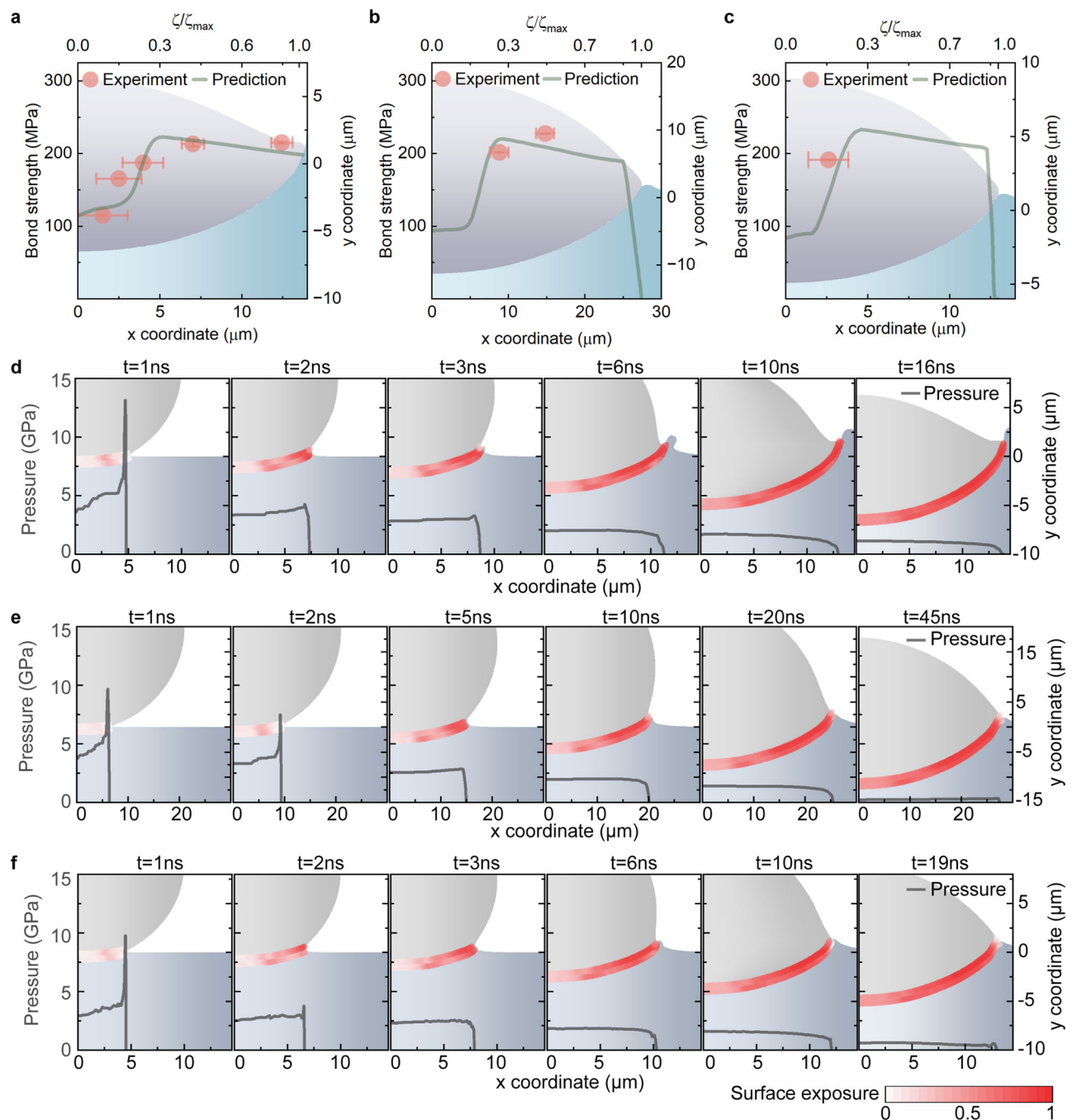


Fig. 2 | Finite element simulations of impact-induced deformation and predictions of bond strength. Comparison between theoretical predictions and experimental measurements of interfacial strength gradient for **a** $23 \pm 2 \mu\text{m}$ Al particles impacting an Al substrate at $1100 \pm 41 \text{ m/s}$, **b** $42 \pm 3 \mu\text{m}$ Al particles impacting an Al substrate at $860 \pm 20 \text{ m/s}$ and **c** $20 \mu\text{m}$ Al particles impacting an Al substrate at 862 m/s . Simulation snapshots showing the temporal evolution and the spatial distribution of the surface exposure and contact pressure at the interface for **d** a $23\text{-}\mu\text{m}$ Al particle impacting Al at 1100 m/s , **e** a $42\text{-}\mu\text{m}$ Al particle impacting Al at 860 m/s and **f** $20 \mu\text{m}$ Al particles impacting an Al substrate at 862 m/s . ξ/ξ_{max} represents the normalized interfacial arc length. Contact pressure is plotted as a

function of the x coordinate. Surface exposure on both the particle and substrate surfaces is represented by a color scale, with white indicating an unexposed surface ($Y=0$) and red indicating a fully exposed surface ($Y=1$). The error bars in x coordinate on the experimental data points are the width of the microtensile specimens at the gauge section. The error bars in y coordinate on the experimental data points are calculated from the resolutions of microtensile load recording and SEM characterization of specimen geometry. The geometry of the deformed particle and substrate were superimposed on the plot as colored regions to enhance clarity. Source data are provided as a Source Data file.

minute native oxide debris (Fig. 3c), only a few nanometers in size, dispersed within the interfacial layer, in which case oxygen signals are absent from elemental mapping. Interestingly, we measured strengths of 94, 201, and 212 MPa for the three locations in Fig. 3a–c, indicating a correlation between increasing bond strength and the transition of the native oxide form at the interface from layers to particles and

eventually debris. Having identified the characteristic forms of native oxide, we conducted a comprehensive examination of the entire interface which we discuss next.

Figure 4a shows TEM observations of the entire bonded interface, from the impact center to the edge. Starting from the impact center, the bonding is extremely localized and discontinuous, separated by

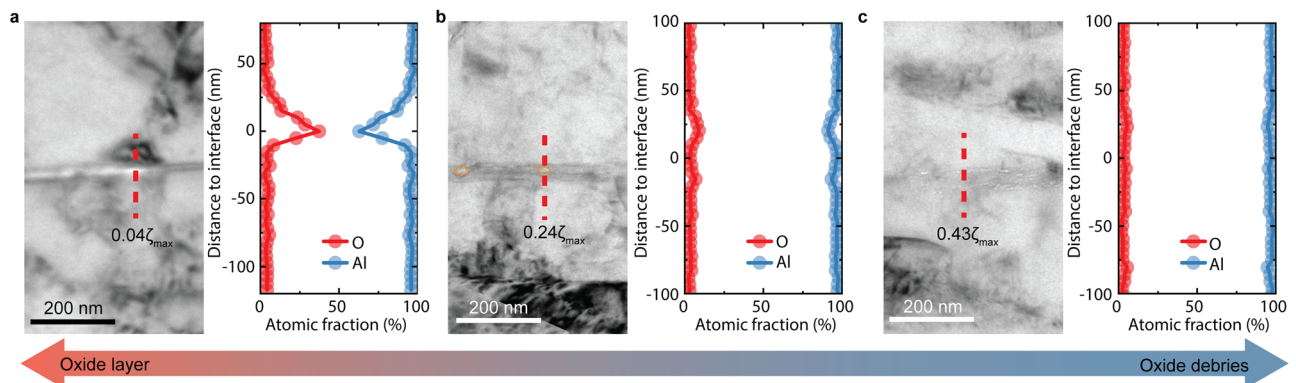


Fig. 3 | Element analysis revealing different forms of native oxide at the bonded interface. **a** Nanometer-thick continuous oxide layer at $0.04 \zeta_{\max}$. **b** Oxide particle with a diameter of approximately 16 nm at $0.24 \zeta_{\max}$. **c** Few-nanometer-sized oxide

debris at $0.43 \zeta_{\max}$. The elemental mappings correspond to the location of the red dashed lines in the TEM images. $\zeta_{\max}$ is the maximum interfacial arc length. Source data are provided as a Source Data file.

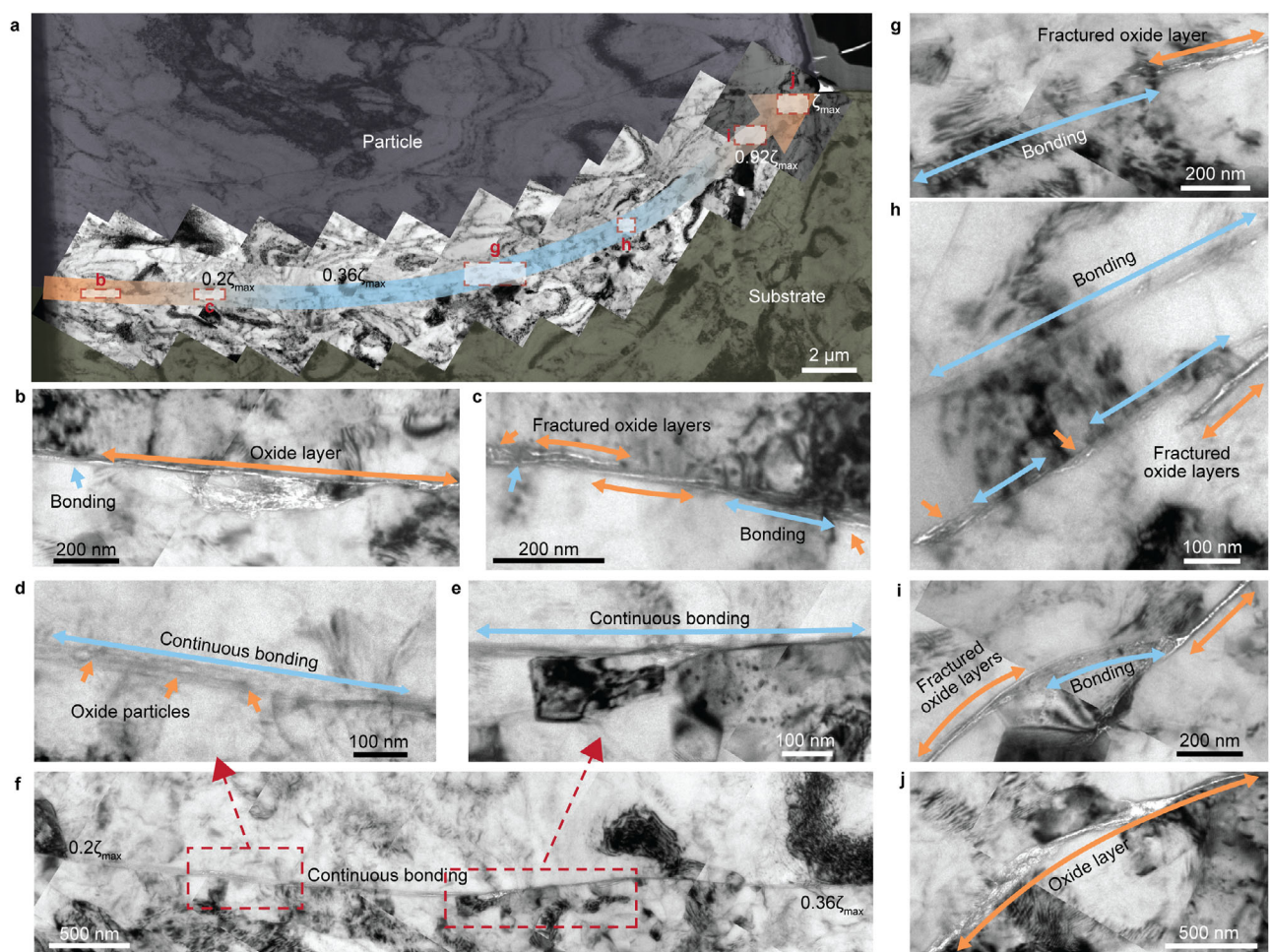


Fig. 4 | High-resolution TEM observations of an impact-induced bonded interface. **a** Cross-sectional SEM image of a $40 \mu\text{m}$ bonded Al particle to Al at 821 m/s along with the TEM images at the interface. High magnification views of the specific locations from the impact center to the edge showing **b** native oxide layers and localized bonding, **c** fractured oxide layers and few-hundred-nanometer-long metallic bonds, **d–f** $5\text{-}\mu\text{m}$ long continuous bonding with isolated oxide particles and

debris, **g** continuous bonding interrupted by fractured oxide layers, **h** extruded metal bonding with both particle and substrate and a transition from **i** the last detectable bonded region to **j** non-bonded region close to the edge. The native oxide layers, particles, and debris are highlighted with orange lines and arrows, and the regions with no discernable native oxide are highlighted with blue lines and arrows.

micrometer-length unfractured native oxide layers (Fig. 4b). As we move away from the center, the increased lateral expansion of the contacting surfaces causes the native oxide to break into shorter pieces, facilitating the formation of more bonded spots (Fig. 4c). Beyond a

normalized interfacial arc length of 0.2, we observe a sharp transition to a $5\text{-}\mu\text{m}$ long continuously bonded interface (Fig. 4f). Higher magnification images of this region reveal only isolated nanometer-sized oxide particles or debris within the bonded interface (Fig. 4d, e). In the

following region, while the lateral surface deformation remains significant, the exposed clean metal surfaces on the particle and substrate sides do not always perfectly align. The incoherent exposure results in interrupted bonded regions interspersed with sub-micrometer-long native oxide layers (Fig. 4g). Overlapped double interfaces in this region (Fig. 4h) provide additional evidence of this incoherency. Our observations suggest that the extruded metal in between double interfaces can form local bonding with both sides, although to a greater extent with the particle surface where greater amounts of clean metal are exposed and fewer native oxide layers and debris are observed (see Supplementary Fig. 10). The oxide fracture and metallurgical bonding persist until near the edge of the interface, extending to a normalized interfacial arc length of 0.92 (Fig. 4i). Beyond this point lies the non-bonded region where the native oxide is barely fractured and clearly acts as a barrier to metallic bonding in Fig. 4j.

The micromechanical measurements of local bond strength, the FE predictions of surface exposure and contact pressure, and the TEM observations all independently and consistently support the presence of a bond strength gradient across a single interface. First, the very localized intimate contact of metallic surfaces observed at the impact center (Fig. 4b) corresponds to the low bond strength measured and predicted in the same region. Furthermore, the region displaying the most intimate and continuous metallic bonding (Fig. 4d–f), aligns with the peak in bond strength observed in both our micromechanical measurements and FE simulations (Fig. 2b). Additionally, the marginal presence of native oxide in this region, predominantly in the form of native oxide debris, suggests that the continuous bonding originated from initially localized clean metal exposure, which then expanded coherently across both particle and substrate contacting surfaces. This is further supported by FE simulations, showing the emergence of a localized coherent high surface exposure region near the impact center early in the deformation process (see 2 ns in Fig. 2e), which then expands laterally to several micrometers long at the end of the deformation (see 45 ns in Fig. 2e). Further moving towards the edge, TEM observations reveal the presence of multiple fractured native oxide layers. These layers reduce the probability of forming continuously bonded areas despite the high local surface exposures predicted by FE at the similar location.

In summary, we used in situ microparticle impact experiments followed by site-specific microtensile testing to study the bond strength across impact-induced bonded interfaces. We revealed a significant gradient of bond strength across impact-induced bonded interfaces. We observed relatively low bond strength in regions near the impact center, followed by a sharp twofold increase to a peak that is significantly higher than the yield strength of the bulk material and, ultimately, a plateau covering over 60% of the interface towards the periphery. The significant strength gradient of $\sim 50 \text{ MPa}\cdot\mu\text{m}^{-1}$ is associated with a localized surface opening in the very early stage of impact, occurring neither at the center nor at the periphery but rather in between. With high-resolution transmission electron microscopy, we showed that the form of the native oxide residue at the bonded interface (i.e., layers, particles, or debris) dictates the level of bond strength. We attribute the strong metallic bonding to two factors: (1) high surface exposure (approaching unity), which facilitates the fracture of the native oxide layer transforming it into fine nanosized particles and debris and exposing clean metallic surfaces, and (2) high local contact pressures (several GPa), bringing particle and substrate surfaces into close atomic-scale proximity. On this basis, we formulated a predictive framework for impact-induced bond strength, which shall prove useful in the performance-oriented design of structure materials and processes relying on supersonic impact-induced bonding.

Methods

Laser-induced microprojectile impact test (LIPIT)

We purchased two batches of aluminum (Al) powder particles with nominal particle sizes of 20 μm and 45 μm from Valimet (Stockton, USA).

A 3.175-mm thick Al plate was purchased from OnlineMetals (Seattle, USA) and was cut into $10 \times 10 \times 3.175 \text{ mm}$ plates. The Al plates were then ground and polished to a nominal 0.04 μm surface finish before being utilized as substrates in the microprojectile impact experiments.

In LIPIT experiments, Al particles were initially dispersed on a launch pad composed of a glass substrate, a 100-nm thick chromium (Cr) layer and 60- μm thick polyurea (PU) film. A laser pulse with 9-ns duration and 532-nm wavelength was focused onto selected single Al particles to accelerate them to velocities ranging from 100 to 1200 m/s. The impact process was recorded using a train of illumination laser pulses (duration of $\sim 500 \text{ ps}$) with predefined intervals (ranging from 12.8 ns to 6.6 μs) and a digital CMOS camera (ORCA-fusion C14440-20 UP). We use the PU film to separate the metal particles from the ablation layer (Cr layer), preventing heat transfer from the laser or ablation process. While the PU film may reduce the ablative force and result in lower particle velocities compared to other launch pad designs^{63,64} that allows direct contact between the metal particles and the ablation layer, it is essential for this study. The film ensures that the particles do not melt or experience a temperature rise that would affect the process of impact-induced bonding.

Cross-sectional transmission electron microscopy

Transmission electron microscopy (TEM) characterization was conducted on the bonded interface produced by Al particles impacting an Al substrate. TEM lamellas were prepared using a focused ion beam/scanning electron microscope (FIB/SEM) microfabrication instrument (Helios 5 UX DualBeam, Thermo Fisher Scientific). Subsequently, bright-field observation was performed using a TEM (JEM2100F, JEOL Ltd.) at an acceleration voltage of 200 kV. The energy dispersive spectroscopy (EDS) analysis (JED2300, JEOL Ltd.) installed in the TEM was used to identify the oxide film at the interface.

In situ microtensile testing

In situ microtensile tests were conducted on Al particles bonded on Al substrates. The bonded Al particles were identified and cross-sectioned using a Helios G4 UX Dual Beam FIB/SEM System. For each bonded Al particle, a 5- μm thick slab revealing the bonded interface was milled and further polished to produce high-quality damage-free surfaces. Microtensile specimens were fabricated from cross-sectional slabs, positioning the bonded interface at the center of the gauge area. For interfaces resulting from particles with diameters of $23 \pm 2 \mu\text{m}$ and impact velocities of $1100 \pm 41 \text{ m/s}$, five microtensile specimens were made at locations spanning from the impact center to the periphery of the particle. For further validation of the finite element simulations, three additional microtensile specimens were produced from different particle sizes and velocities, namely, $42 \pm 3 \mu\text{m}$ at $860 \pm 20 \text{ m/s}$ from the normalized interfacial arc lengths of 0.275 and 0.48, and 20 μm at 862 m/s from the normalized interfacial arc lengths of 0.17.

A micromechanical stage (Alemnis AG, Switzerland) was integrated inside a LEO 1550 SEM (Zeiss, Germany) to conduct in situ microtensile testing and to measure the local bond strength of the Al–Al impact-induced bonded interfaces. The microtensile gripper and the specimen were aligned inside the SEM. The shoulder of the microtensile test specimens was gripped and displaced under a constant displacement rate of 50 nm/s until the specimen fractured. Load-displacement data was continuously recorded, and the interfacial fracture was captured in situ (see Supplementary Movie 1).

Numerical modeling of particle impact

The detailed method for the numerical modeling of particle impact shown in Fig. 2 is described in Supplementary Information Section 2.

Theoretical prediction of bond strength

The detailed calculation procedure of the bond strength prediction shown in Fig. 2 is described in Supplementary Information Section 3.

Data availability

The data that support the findings of this study are available within the manuscript and its Supplementary Information. Source data are provided with this paper.

References

- Johnson, B. C., Minton, D. A., Melosh, H. J. & Zuber, M. T. Impact jetting as the origin of chondrules. *Nature* **517**, 339–341 (2015).
- Wiggins, S. E., Johnson, B. C., Collins, G. S., Melosh, H. J. & Marchi, S. Widespread impact-generated porosity in early planetary crusts. *Nat. Commun.* **13**, 4817 (2022).
- Assadi, H., Kreye, H., Gärtner, F. & Klassen, T. Cold spraying—a materials perspective. *Acta Mater.* **116**, 382–407 (2016).
- Schmidt, T. et al. From particle acceleration to impact and bonding in cold spraying. *J. Therm. Spray Technol.* **18**, 794 (2009).
- Schmidt, T., Gärtner, F., Assadi, H. & Kreye, H. Development of a generalized parameter window for cold spray deposition. *Acta Mater.* **54**, 729–742 (2006).
- Ichikawa, Y., Tokoro, R., Tanno, M. & Ogawa, K. Elucidation of cold-spray deposition mechanism by auger electron spectroscopic evaluation of bonding interface oxide film. *Acta Mater.* **164**, 39–49 (2019).
- Melosh, H. J. *Impact Cratering: A Geologic Process* (Oxford University Press, 1996).
- Thevamaran, R. et al. Dynamic creation and evolution of gradient nanostructure in single-crystal metallic microcubes. *Science* **354**, 312–316 (2016).
- Lu, Y., Huang, J. Y., Wang, C., Sun, S. & Lou, J. Cold welding of ultrathin gold nanowires. *Nat. Nanotechnol.* **5**, 218–224 (2010).
- Brownlee, D. et al. Comet 81P/Wild 2 under a microscope. *Science* **314**, 1711–1716 (2006).
- Blazynski, T. Z. *Explosive Welding, Forming and Compaction* (Springer, 1983).
- Zhang, Y. et al. Application of high velocity impact welding at varied different length scales. *J. Mater. Process Technol.* **211**, 944–952 (2011).
- Akbarimousavi, A. & Alhassani, S. Numerical and experimental studies of the mechanism of the wavy interface formations in explosive/impact welding. *J. Mech. Phys. Solids* **53**, 2501–2528 (2005).
- Daehn, G. S. & Lippold, J. C. Low-temperature laser spot impact welding driven without contact. US patent PCT/US09/36299 (2009).
- Zhang, J. et al. Increasing joint strength in magnesium-aluminum dissimilar impact welding by surface nanocrystallization of magnesium alloy sheet. *Scr. Mater.* **251**, 116190 (2024).
- Liu, B., Vivek, A. & Daehn, G. S. Joining sheet aluminum AA6061-T4 to cast magnesium AM60B by vaporizing foil actuator welding: input energy, interface, and strength. *J. Manuf. Process* **30**, 75–82 (2017).
- Assadi, H., Gärtner, F., Stoltenhoff, T. & Kreye, H. Bonding mechanism in cold gas spraying. *Acta Mater.* **51**, 4379–4394 (2003).
- Papyrin, A., Kosarev, V., Klinkov, S., Alkimov, A. & Fomin, V. *Cold Spray Technology* (Elsevier, 2007).
- Champagne, V. K. *The Cold Spray Materials Deposition Process: Fundamentals and Applications* (Woodhead, 2007).
- Assadi, H. et al. On parameter selection in cold spraying. *J. Therm. Spray Technol.* **20**, 1161–1176 (2011).
- Veyssset, D. et al. Dynamics of supersonic microparticle impact on elastomers revealed by real-time multi-frame imaging. *Sci. Rep.* **6**, 25577 (2016).
- Lee, J.-H. et al. High strain rate deformation of layered nanocomposites. *Nat. Commun.* **3**, 1164 (2012).
- Lee, J.-H., Loya, P. E., Lou, J. & Thomas, E. L. Dynamic mechanical behavior of multilayer graphene via supersonic projectile penetration. *Science* **346**, 1092–1096 (2014).
- Xie, W. et al. Dynamics and extreme plasticity of metallic microparticles in supersonic collisions. *Sci. Rep.* **7**, 5073 (2017).
- Xue, S. et al. High-velocity projectile impact induced 9R phase in ultrafine-grained aluminium. *Nat. Commun.* **8**, 1653 (2017).
- Veyssset, D. et al. High-velocity micro-projectile impact testing. *Appl. Phys. Rev.* **8**, 011319 (2021).
- Hassani-Gangaraj, M., Veyssset, D., Nelson, K. A. & Schuh, C. A. In-situ observations of single micro-particle impact bonding. *Scr. Mater.* **145**, 9–13 (2018).
- Hassani-Gangaraj, M., Veyssset, D., Nelson, K. A. & Schuh, C. A. Melt-driven erosion in microparticle impact. *Nat. Commun.* **9**, 5077 (2018).
- Hassani-Gangaraj, M., Veyssset, D., Nelson, K. A. & Schuh, C. A. Impact-bonding with aluminum, silver, and gold microparticles: toward understanding the role of native oxide layer. *Appl. Surf. Sci.* **476**, 528–532 (2019).
- Hassani, M., Veyssset, D., Sun, Y., Nelson, K. A. & Schuh, C. A. Microparticle impact-bonding modes for mismatched metals: from co-deformation to splatting and penetration. *Acta Mater.* **199**, 480–494 (2020).
- Hassani-Gangaraj, M., Veyssset, D., Nelson, K. A. & Schuh, C. A. Melting can hinder impact-induced adhesion. *Phys. Rev. Lett.* **119**, 175701 (2017).
- Dowding, I. et al. Particle size effects in metallic microparticle impact-bonding. *Acta Mater.* **194**, 40–48 (2020).
- Chaban, I., Sun, Y., Veyssset, D., Nelson, K. A. & Schuh, C. A. The effect of substrate temperature on the critical velocity in microparticle impact bonding. *Appl. Phys. Lett.* **119**, 011903 (2021).
- Lima, C. R. C. & Guilemany, J. M. Adhesion improvements of thermal barrier coatings with HVOF thermally sprayed bond coats. *Surf. Coat. Technol.* **201**, 4694–4701 (2007).
- Li, W.-Y., Li, C.-J. & Liao, H. Significant influence of particle surface oxidation on deposition efficiency, interface microstructure and adhesive strength of cold-sprayed copper coatings. *Appl. Surf. Sci.* **256**, 4953–4958 (2010).
- Marrocco, T., McCartney, D. G., Shipway, P. H. & Sturgeon, A. J. Production of titanium deposits by cold-gas dynamic spray: numerical modeling and experimental characterization. *J. Therm. Spray Technol.* **15**, 263–272 (2006).
- Kosarev, V. F., Klinkov, S. V., Alkimov, A. P. & Papyrin, A. N. On some aspects of gas dynamics of the cold spray process. *J. Therm. Spray Technol.* **12**, 265–281 (2003).
- Fukanuma, H., Ohno, N., Sun, B. & Huang, R. In-flight particle velocity measurements with DPV-2000 in cold spray. *Surf. Coat. Technol.* **201**, 1935–1941 (2006).
- Goldbaum, D. et al. The effect of deposition conditions on adhesion strength of Ti and Ti6Al4V cold spray splats. *J. Therm. Spray Technol.* **21**, 288–303 (2012).
- Champagne, V. K., Helfrich, D. J., Dinavahi, S. P. G. & Leyman, P. F. Theoretical and experimental particle velocity in cold spray. *J. Therm. Spray Technol.* **20**, 425–431 (2011).
- Gilmore, D. L., Dykhuizen, R. C., Neiser, R. A., Roemer, T. J. & Smith, M. F. Particle velocity and deposition efficiency in the cold spray process. *J. Therm. Spray Technol.* **8**, 576–582 (1999).
- Li, W.-Y. et al. An investigation on temperature distribution within the substrate and nozzle wall in cold spraying by numerical and experimental methods. *J. Therm. Spray Technol.* **21**, 41–48 (2012).
- Schmidt, T., Gaertner, F. & Kreye, H. New developments in cold spray based on higher gas and particle temperatures. *J. Therm. Spray Technol.* **15**, 488–494 (2006).

44. Nastic, A., Jodoin, B., Poirier, D. & Legoux, J.-G. Particle temperature effect in cold spray: a study of soft particle deposition on hard substrate. *Surf. Coat. Technol.* **406**, 126735 (2021).
45. Azzi, V. D. & Tsai, S. W. Anisotropic strength of composites. *Exp. Mech.* **5**, 283–288 (1965).
46. Liu, T., Vaudin, M. D., Bunn, J. R., Ungár, T. & Brewer, L. N. Quantifying dislocation density in Al-Cu coatings produced by cold spray deposition. *Acta Mater.* **193**, 115–124 (2020).
47. Liu, T., Leazer, J. D. & Brewer, L. N. Particle deformation and microstructure evolution during cold spray of individual Al-Cu alloy powder particles. *Acta Mater.* **168**, 13–23 (2019).
48. Li, J. et al. Process characteristics and interfacial microstructure in spot impact welding of titanium to stainless steel. *J. Manuf. Process* **50**, 421–429 (2020).
49. Wang, H., Liu, D., Lippold, J. C. & Daehn, G. S. Laser impact welding for joining similar and dissimilar metal combinations with various target configurations. *J. Mater. Process Technol.* **278**, 116498 (2020).
50. Wang, Q., Ma, N., Takahashi, M., Luo, X. & Li, C. Development of a material model for predicting extreme deformation and grain refinement during cold spraying. *Acta Mater.* **199**, 326–339 (2020).
51. King, P. C., Zahiri, S. H. & Jahedi, M. Microstructural refinement within a cold-sprayed copper particle. *Metall. Mater. Trans. A* **40**, 2115–2123 (2009).
52. Rokni, M. R., Widener, C. A. & Champagne, V. R. Microstructural evolution of 6061 aluminum gas-atomized powder and high-pressure cold-sprayed deposition. *J. Therm. Spray Technol.* **23**, 514–524 (2014).
53. Hassani-Gangaraj, M., Veyssset, D., Champagne, V. K., Nelson, K. A. & Schuh, C. A. Adiabatic shear instability is not necessary for adhesion in cold spray. *Acta Mater.* **158**, 430–439 (2018).
54. Hahn, M. et al. Vaporizing foil actuator welding as a competing technology to magnetic pulse welding. *J. Mater. Process Technol.* **230**, 8–20 (2016).
55. Amiri, M., Soto Leytan, K. N., Apelian, D., Mumm, D. R. & Valdevit, L. Controlling splat boundary network evolution towards the development of strong and ductile cold sprayed refractory metals: the role of powder characteristics. *Mater. Sci. Eng. A* **902**, 146559 (2024).
56. Bae, G. et al. Bonding features and associated mechanisms in kinetic sprayed titanium coatings. *Acta Mater.* **57**, 5654–5666 (2009).
57. Chen, Q. et al. High-strain-rate material behavior and adiabatic material instability in impact of micron-scale Al-6061 particles. *J. Therm. Spray Technol.* **27**, 641–653 (2018).
58. Yang, Y., Xinming, Z., Zhenghua, L. & Qingyun, L. Adiabatic shear band on the titanium side in the Ti/mild steel explosive cladding interface. *Acta Mater.* **44**, 561–565 (1996).
59. Huang, F., Tao, N. R. & Lu, K. Effects of strain rate and deformation temperature on microstructures and hardness in plastically deformed pure aluminum. *J. Mater. Sci. Technol.* **27**, 1–7 (2011).
60. Cai, J., Griesbach, C., Ahnen, S. G. & Thevamaran, R. Dynamic hardness evolution in metals from impact induced gradient dislocation density. *Acta Mater.* **249**, 118807 (2023).
61. Flanagan, T. J. et al. Mechanical properties of supersonic-impacted Al6061 powder particles. *Scr. Mater.* **171**, 52–56 (2019).
62. Bedard, B. A. et al. Microstructure and micromechanical response in gas-atomized Al 6061 alloy powder and cold-sprayed splats. *J. Therm. Spray Technol.* **27**, 1563–1578 (2018).
63. Kajihara, M., Nagaami, K., Miyagawa, T., Kondo, T. & Yonezu, A. Development of a velocity measurement method for a micro-particle projectile and high-speed impact testing of metallic materials for grain refinement. *Acta Mater.* **262**, 119467 (2024).
64. Veyssset, D., Sun, Y., Kooi, S. E., Lem, J. & Nelson, K. A. Laser-driven high-velocity microparticle launcher in atmosphere and under vacuum. *Int. J. Impact Eng.* **137**, 103465 (2020).

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Author contributions

Conceptualization and supervision: M.H. Methodology: Q.T., D.V., H.A., Y.I., M.H. Investigation and visualization: Q.T. Analysis: Q.T., D.V., H.A., Y.I., M.H. Funding acquisition: M.H., Y.I. Writing—original draft: Q.T., M.H. Writing—review and editing: Q.T., D.V., H.A., Y.I., M.H.

Competing interests

The authors declare no competing interests.

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Correspondence and requests for materials should be addressed to Mostafa Hassani.

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