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Du, T., Gao, C., Wei, L., Ye, H., Zhang, G., & Du, L. (2026). Morphology-dependent indentation response of pressure-assisted sintered Cu for power electronic packaging. *Materials and Design*, 271, Article 117159. <https://doi.org/10.1016/j.matdes.2026.117159>

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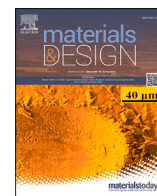
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Morphology-dependent indentation response of pressure-assisted sintered Cu for power electronic packaging

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HIGHLIGHTS

- Particle morphology governs the indentation response of sintered Cu.
- Bimodal Cu shows the highest elastic modulus, while monomodal Cu shows highest yield strength.
- Flake-shaped Cu is weakened by poor through-thickness connectivity.
- Spherical and Berkovich tips activate distinct deformation mechanisms.
- SEM and MD reveal morphology-specific pore collapse and plasticity.

ARTICLE INFO

Keywords:

Sintering
Particle morphology
Nanoindentation
Molecular dynamics
Power electronics packaging

ABSTRACT

Sintered copper (Cu) particles are widely used in catalysis, biomedical devices, energy systems and electronic packaging, yet the role of particle morphology in governing their mechanical response remains insufficiently resolved. Here, we investigate porous sintered Cu produced from monomodal spherical, bimodal spherical and flake-shaped particles by combining nanoindentation, scanning electron microscopy (SEM) and molecular dynamics (MD) simulations. Both spherical and Berkovich indenters were used to interrogate morphology-dependent deformation mechanisms. The bimodal spherical network exhibited the highest elastic modulus, whereas the monomodal spherical network showed the highest indentation yield strength. In contrast, the flake-derived network displayed substantially reduced elastic modulus and strength due to weak through-thickness connectivity. Correlative SEM images and MD simulations reveal morphology-dependent deformation mechanisms: distributed neck plasticity in the monomodal network, hierarchical load sharing between coarse and fine particles in the bimodal network, and anisotropic strain localization along interlamellar interfaces in the flake-derived network. Indenter geometry further modulates the response, with spherical indentation promoting distributed compaction and Berkovich indentation promoting stress concentration at weak interfaces. These findings highlight particle morphology as a critical factor governing the topology-dependent mechanical performance of porous sintered Cu.

1. Introduction

Porous materials are widely used as functional and structural architectures because of their low density, highly accessible surface area and interconnected pore networks [1–3], enabling applications ranging from filtration [4,5], catalysis [6,7], energy storage [8,9] to biomedical implants [10,11] and power electronic packaging [12–14]. Unlike in dense materials, their mechanical behaviour is intrinsically local and heterogeneous, with plastic deformation often initiated by pore collapse, contact yielding and localized densification [15]. These responses are

highly sensitive to processing parameters [16–18], material composition [19,20] and particle morphology [21–23]. Although the mechanical behaviour of porous materials has been extensively investigated, most existing studies have focused on the roles of processing parameters and material composition [24–26]. However, how particle morphology governs the local mechanical response and plastic deformation behaviour of sintered porous structures remains insufficiently resolved.

Nanoindentation provides an effective approach for probing mechanical responses at the micro- to nanoscale [27,28]. Its high spatial

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<https://doi.org/10.1016/j.matdes.2026.117159>

Received 26 July 2026; Received in revised form 14 September 2026; Accepted 28 September 2026

Available online 30 September 2026

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resolution and precisely controlled load-displacement acquisition enable quantitative evaluation of elastic modulus, hardness and time-dependent deformation [29]. It has therefore been widely used to correlate the local mechanical properties of porous materials with processing conditions, strain-rate sensitivity and creep behavior [30,31]. For instance, Zhang et al. [32] showed that sintering temperature dominates densification and porosity reduction in spark-plasma-sintered porous materials, whereas applied pressure promotes microstructural homogenization and phase-dependent nanomechanical evolution. For highly heterogeneous porous sintered networks, however, local mechanical testing must also provide sufficient statistical representativeness. Micropillar compression can reveal deformation mechanisms within a selected microstructural volume, but its response is strongly affected by pillar geometry and the specific pore-neck configuration sampled by each pillar [33,34]. This makes it challenging to define a representative volume element or obtain morphology-averaged mechanical properties from a limited number of pillars. Nanoindentation is therefore used in this study to evaluate the local mechanical response of sintered porous Cu structures with different particle morphologies and to examine the associated plastic deformation mechanisms.

Plastic deformation is inherently structure-sensitive, as strain accommodation and localization are governed by characteristic microstructural length scales, crystallographic constraints and defect populations [35,36]. In dense Cu, plastic deformation is primarily accommodated by crystallographic slip and dislocation-mediated plasticity, with grain size, crystallographic orientation and boundary constraints governing strain localization and work-hardening behaviour [37,38]. For example, Wang et al. [39] demonstrated that nanoindentation of Cu single crystals produces orientation-dependent pile-up patterns governed by crystallographic slip along the intersections between primary slip planes and the indented surface. In porous materials, this deformation response is further complicated by the disrupted continuity of the load-bearing solid, which promotes additional mechanisms such as strain concentration, ligament or neck bending, pore collapse and localized densification [33,40,41]. Thus, resolving how these pore-mediated mechanisms evolve under local contact loading is essential for linking particle-neck architecture to the mechanical reliability of porous sintered Cu. However, experimental characterization remains limited in its ability to resolve the transient and spatially heterogeneous processes that govern plastic deformation, which has motivated the development of computational modelling approaches, including discrete dislocation dynamics [42–44], crystal plasticity modelling [45–47], phase-field methods [48,49] and molecular dynamics (MD) simulations [50–53]. Among these, phase-field and crystal plasticity models are effective for capturing mesoscale microstructural evolution, damage accumulation and crystallographic strain partitioning [54,55]. However, their continuum formulations generally cannot directly resolve dislocation nucleation and migration [56]. In contrast, discrete dislocation dynamics can explicitly describe dislocation-mediated plasticity, but its reliance on predefined dislocation mechanisms and idealized geometries limits its applicability to sintered Cu networks where atomistic nucleation, pore collapse, neck evolution and irregular particle morphology are strongly coupled [57]. By comparison, MD simulation directly resolves atomic diffusion, neck formation, defect nucleation, stacking-fault evolution and pore collapse in particle-based metallic networks [58,59]. Despite its inherent spatial and temporal scale limitations, MD is particularly well suited to identifying relative trends in morphology-dependent deformation, while the associated atomistic defect processes can be directly tracked [60]. Therefore, MD simulations were employed here to elucidate how particle morphology regulates pore evolution, dislocation activity and localized plasticity in sintered Cu networks.

This study investigates how particle morphology governs the mechanical response and plastic deformation of porous Cu sintered from monomodal spherical, bimodal spherical and flake-shaped particles. By integrating spherical and Berkovich nanoindentation,

post-indentation scanning electron microscopy (SEM) characterization and MD simulations, the work quantifies local mechanical properties and elucidates morphology-dependent deformation mechanisms from the microstructural to atomistic scales, providing mechanistic guidance for the design of sintered Cu interconnects in power electronic packaging.

2. Experimental and simulation methods

2.1. Sample preparation

The initial morphologies of the three Cu particle systems are shown in Fig. 1(a)–(c), corresponding to monomodal spherical, bimodal spherical and flake-shaped particles, with particle shape and size distribution jointly defining the distinct starting particle architectures. The monomodal system was composed of spherical Cu particles with an average particle diameter of approximately 150 nm. In the bimodal system, small spherical particles of approximately 150 nm were mixed with larger spherical particles of approximately 1.5 μm , forming a highly asymmetric size distribution that promotes the filling of interstitial spaces by the smaller particles [61]. The flake-shaped particles consisted of disk-like Cu particles with lateral dimensions of approximately 2.5–7.5 μm , a thickness of about 150 nm, and corresponding aspect ratios of approximately 17:1–50:1.

The fabrication procedure for the sintered Cu samples is illustrated in Fig. 1(d)–(f). The Cu paste was first screen-printed onto a Cu substrate, followed by preconditioning at 120 °C for 5 min in a nitrogen atmosphere to remove residual solvent. A temporary aluminium (Al) capping film was then placed on the printed paste. Pressure-assisted sintering was carried out at 250 °C under a uniaxial pressure of 30 MPa for 5 min in a reducing 5% H_2/N_2 atmosphere. After cooling to room temperature, the Al capping layer was removed, leaving the exposed sintered Cu layer. Cross-sectional SEM images of the pressure-assisted sintered Cu layers fabricated from the monomodal, bimodal and flake-shaped particles are presented in Fig. 1(g)–(i), respectively, showing morphology-dependent differences in pore distribution, particle packing and sintered-neck formation, with corresponding porosities of approximately 7%, 10% and 20%.

2.2. Nanoindentation tests

Nanoindentation tests were conducted at room temperature using a Nanoindenter G200 system (Agilent Technologies, Chandler, AZ, USA) equipped with a continuous stiffness measurement (CSM) module. Both Berkovich and spherical diamond indenters were employed for each sintered Cu sample to evaluate the mechanical response under different contact geometries. During loading, a harmonic displacement oscillation with an amplitude of 2 nm and a frequency of 45 Hz was superimposed on the displacement signal, allowing continuous determination of contact stiffness from the amplitude and phase shift of the load response. The machine stiffness and indenter area functions were calibrated using the Oliver-Pharr method [62]. A customized insulated Berkovich diamond tip with a Macor shaft (Synton, Switzerland) and a centerline-to-face angle of 65.3° was used to measure indentation modulus and hardness under a sharp-contact field. In parallel, a spherical diamond indenter with a nominal diameter of approximately 10 μm was used to determine the corresponding indentation modulus and hardness and to construct representative indentation stress-strain responses under a broader contact field. For each particle morphology and indenter geometry, three independent indentations were performed at spatially separated locations on the same specimen. For all measurements, the indenter was driven into the sample surface at a target strain rate of 0.05 s^{-1} to a maximum penetration depth of 2500 nm, held at the peak depth for 5 s, and then unloaded. The selected penetration depth provides a micrometer scale probing volume encompassing multiple particle-neck or inter-flake features, enabling a consistent comparison of the local effective mechanical response among the three morphologies.

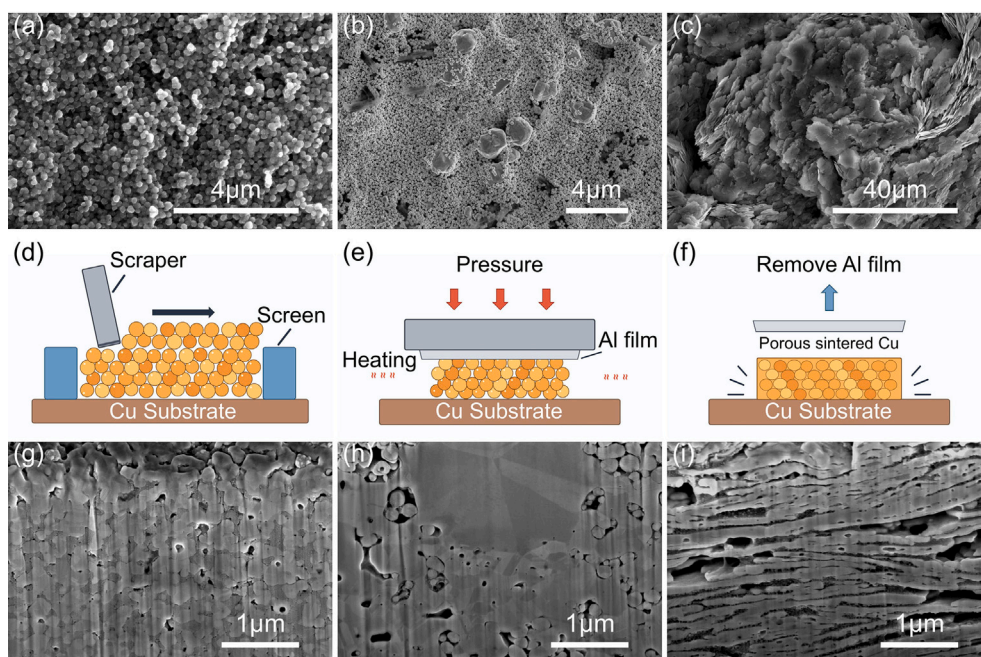


Fig. 1. Representative SEM images of the initial Cu particles with (a) monomodal, (b) bimodal, and (c) flake-shaped particle morphologies; (d–f) schematic illustration of the fabrication process, including stencil printing of the Cu paste onto a Cu substrate, pressure-assisted sintering during heating using a temporary Al film for pressure transfer, and subsequent removal of the Al film; (g–i) cross-sectional SEM images of the pressure-assisted sintered Cu structures fabricated from the corresponding (g) monomodal, (h) bimodal, and (i) flake-shaped Cu particles, respectively.

Table 1
Summary of the construction parameters for the atomistic Cu particle assemblies.

Parameter	Monomodal spherical-particle model	Bimodal spherical-particle model	Flake-shaped particle model
Representative system	Monomodal Cu assembly	Bimodal Cu assembly	Flake-shaped Cu assembly
Simulation domain	$500 \text{ \AA} \times 500 \text{ \AA} \times 500 \text{ \AA}$	$500 \text{ \AA} \times 500 \text{ \AA} \times 500 \text{ \AA}$	$500 \text{ \AA} \times 500 \text{ \AA} \times 600 \text{ \AA}$
Boundary condition	Periodic in x , y and z directions	Periodic in x , y and z directions	Periodic in x and y directions; non-periodic in z direction
Particle morphology	Spherical Cu nanoparticles	Mixture of large and small spherical Cu particles	Disk-like Cu flakes
Particle size	Diameter: $60 \text{ \AA}$ with a standard deviation of $20 \text{ \AA}$	Large spheres: $200 \text{ \AA}$; small spheres: $60 \text{ \AA}$ with a standard deviation of $20 \text{ \AA}$	Diameter: $490 \text{ \AA}$; thickness: $15 \text{ \AA}$
Crystallographic structure	Randomly assigned crystallographic orientations	Randomly assigned crystallographic orientations	Voronoi-defined internal grain structures

2.3. MD simulation

To capture the morphology-dependent evolution of sintered Cu networks, atomistic models representing different initial particle geometries were constructed and subjected to sintering and subsequent mechanical loading. The simulations were implemented in Large-scale Atomic/Molecular Massively Parallel Simulator (LAMMPS) [63,64], while the resulting trajectories were analyzed in Open Visualization Tool (OVITO) [65] to identify local crystal structures, dislocation activity and deformation-induced atomic rearrangements. Cu–Cu interactions were described using the Williams embedded-atom method potential [66], which provides a physically established description of defect formation, structural evolution and mechanical deformation.

Three atomistic particle assemblies were developed to represent the monomodal, bimodal and flake-shaped Cu systems, and the main construction parameters are listed in Table 1. The corresponding initial particle assemblies are presented in Fig. 2, providing a visual comparison of the distinct packing architectures generated for the three Cu powder morphologies. For the two spherical-particle systems, fully periodic boundary conditions were imposed in all three directions to reproduce a bulk packing environment. The flake-based model was instead treated as a stacked architecture, with periodicity applied only in

the in-plane directions. Along the out-of-plane direction, non-periodic boundaries were used and $100 \text{ \AA}$ vacuum regions were placed above and below the stack to allow relaxation normal to the flake layers. The monomodal model was built by randomly placing spherical Cu nanoparticles with normally distributed diameters and randomly assigned crystallographic orientations. The bimodal model followed the same construction principle, but larger spherical particles were first introduced into the simulation cell, followed by the insertion of smaller particles into the remaining interstitial space, producing a more heterogeneous particle-packing configuration. For the flake-shaped system, disk-like crystalline particles were generated with Voronoi-defined internal grain structures and then assembled sequentially in a layer-by-layer manner. To approximate the imperfect alignment expected in practical flake-powder assemblies, neighbouring flakes were initialized with an interlayer spacing of $10 \text{ \AA}$, an in-plane tilt angle of 3.5° , and random lateral offsets within $\pm 100 \text{ \AA}$. After the initial assemblies were generated, overlapping atoms were removed globally to eliminate unphysical interpenetration. The final configurations contained approximately 4.0–5.0 million atoms.

After construction of the initial particle models, the sintering treatment was reproduced through a thermal–mechanical sequence. Each model was first subjected to thermal activation by increasing the

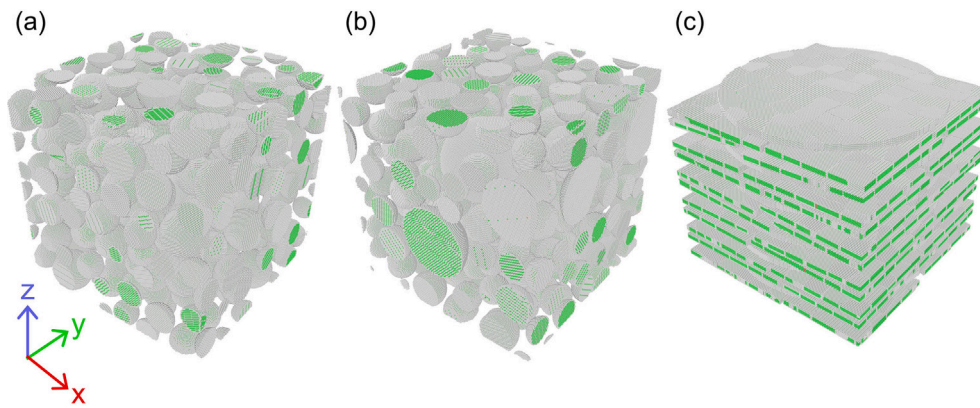


Fig. 2. Initial atomistic configurations of the Cu particle assemblies prior to simulated sintering and mechanical loading: (a) monomodal, (b) bimodal, and (c) flake-shaped particle systems.

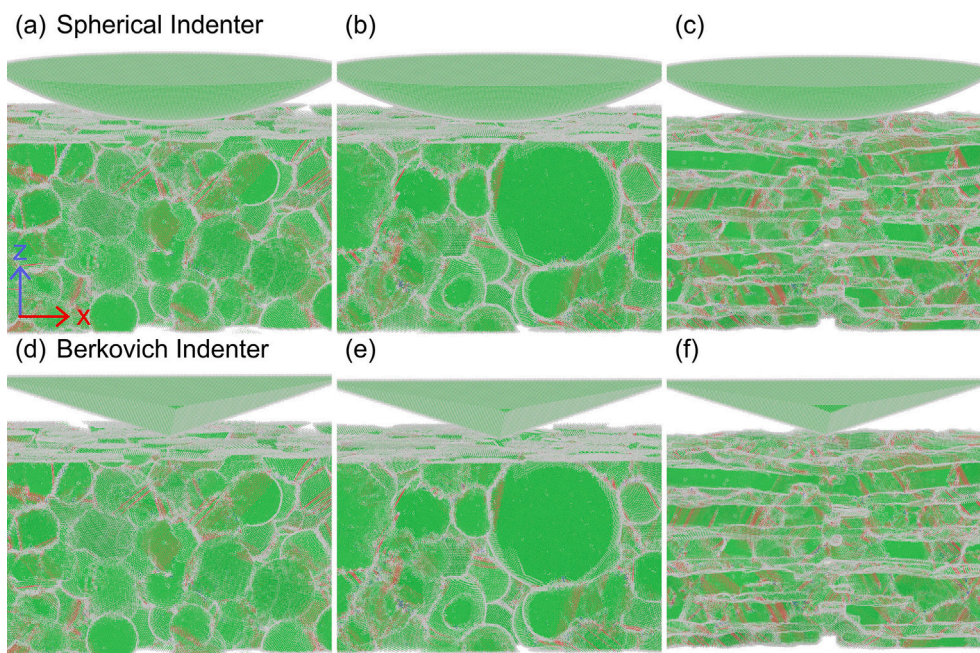


Fig. 3. Representative atomic configurations during spherical and Berkovich indentation of porous sintered Cu structures: (a–c) spherical indentation tests of the monomodal, bimodal and flake-shaped particle systems, respectively; (d–f) Berkovich indentation tests of the corresponding structures.

temperature from 300 K to 523 K within 100 ps under an NVT ensemble controlled by a Nosé-Hoover thermostat, allowing atomic diffusion across particle contacts and the development of sintering necks. Pressure-assisted densification was then introduced by imposing a 50% compressive strain along the z direction at a strain rate of $1 \times 10^9 \text{ s}^{-1}$. To reduce non-equilibrium stresses generated during the rapid compression stage, the compressed structures were subsequently relaxed at 523 K for 100 ps.

Nanoindentation simulations were performed using both spherical and Berkovich indenters to examine the local deformation response of the sintered Cu networks under different contact geometries. Both indenter geometries were treated as perfectly rigid bodies during indentation. The initial simulation configurations prior to indentation are presented in Fig. 3. For the spherical indentation simulations, the indenter geometry was generated from a parent sphere with a radius of 446.43 K. A spherical-cap region with a height of 70 Å along the indentation direction was retained, corresponding to a transverse cap radius of approximately 240 Å. In comparison, the Berkovich indenter was defined

with a height of 70 Å and a geometric parameter of $\tan \beta = 2.171$ to match the experimental centerline-to-face angle of 65.3° . For both indenter geometries, the indenter was initially positioned 2 Å above the sample surface to prevent unphysical atomistic interpenetration. A bottom layer with a thickness of 10 Å was fixed to suppress rigid-body motion, while the remaining sample atoms were integrated using an NVE ensemble and thermostatted at 300 K using a Langevin thermostat. The indenters were displaced along the negative z direction at 1 Å/ps, giving an effective indentation depth of 70 Å. Subsequently, the indenter was retracted to capture the initial unloading response. During indentation and unloading, the total force acting on the indenter was obtained by summing the atomic forces in the z direction, and the corresponding force–depth data were recorded together with atomic configurations and per-atom stress tensors for subsequent analysis of pore collapse, neck-level deformation and local stress redistribution. Given the spatial and temporal limitations of MD, these indentation simulations are intended primarily for comparative analysis of morphology-dependent deformation mechanisms.

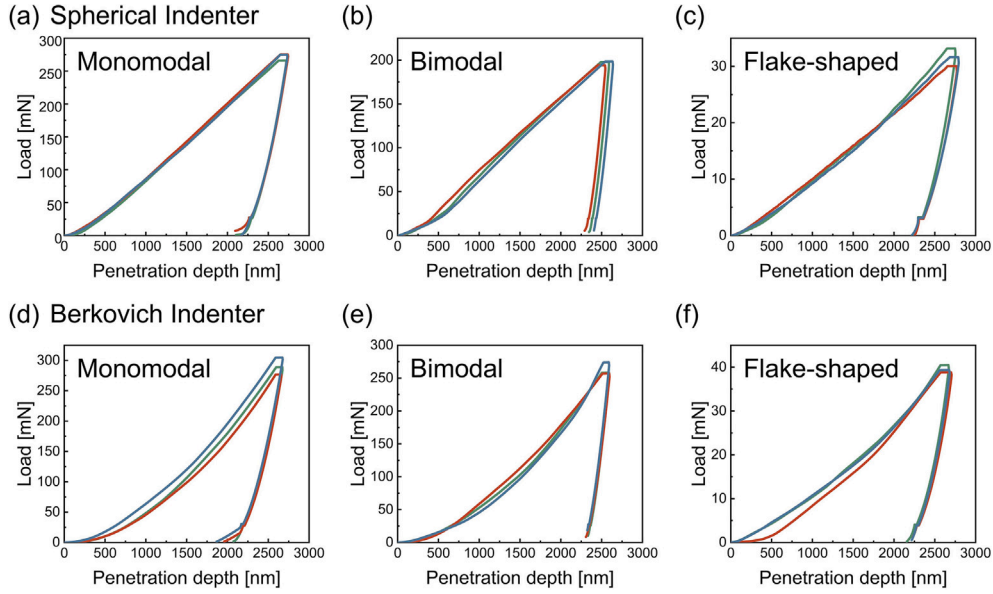


Fig. 4. Load-penetration depth responses of sintered Cu structures with different particle morphologies under nanoindentation: (a–c) spherical indentation curves for the monomodal, bimodal and flake-shaped structures, respectively; (d–f) Berkovich indentation curves for the corresponding structures.

3. Results and discussion

3.1. Load-penetration curves

Fig. 4 shows the representative load–penetration curves of the monomodal, bimodal and flake-shaped structures obtained using spherical and Berkovich indenters. The loading curves for both indenter geometries exhibit no discernible pop-in event, indicating continuous indentation-induced plasticity rather than abrupt displacement-burst-mediated yielding [67,68]. The extracted peak loads reveal a clear morphology-dependent indentation response. Under spherical indentation, the corresponding values are 271.82 ± 5.13 mN, 196.22 ± 2.53 mN and 31.61 ± 0.91 mN, respectively. Under Berkovich indentation, the maximum loads of the monomodal, bimodal and flake-shaped structures are 290.05 ± 13.99 mN, 262.53 ± 9.81 mN and 39.47 ± 0.93 mN, respectively. The Berkovich indenter yields higher peak loads than the spherical indenter for all three particle morphologies. This difference can be attributed to the sharper Berkovich contact geometry, which produces a more localized and constrained deformation field, whereas the spherical indenter generates a smoother contact field and a more gradual elastic-plastic response [69,70]. For the monomodal structure, the Berkovich and spherical indenters produce comparable peak loads, whereas more pronounced indenter-dependent differences are observed for the bimodal and flake-shaped structures. This greater sensitivity can be associated with the sharper Berkovich geometry, which generates a more localized and confined deformation field and preferentially activates structurally weak regions [69,70]. In the bimodal network, pores and heterogeneous necks within incompletely sintered regions intensify local deformation, while the limited through-thickness load-bearing capacity and abundant residual pores of the flake-shaped structure further facilitate localized collapse. The substantial residual penetration depths after unloading indicate that the indentation response is dominated by irreversible deformation with limited elastic recovery. In dense Cu, residual indentation is primarily accommodated by crystallographic slip, pile-up and dislocation-mediated plasticity [39]; in the present porous Cu structures, it additionally involves irreversible collapse of the pore-neck network through neck yielding and bending, local pore closure and densification.

3.2. Indentation-derived mechanical response

Fig. 5 presents the depth-dependent elastic modulus obtained from nanoindentation for the three sintered Cu structures. The modulus

was calculated from the contact stiffness using the Oliver-Pharr framework [62]. For each indenter geometry, the reduced modulus at a given penetration depth was obtained as

$$E_{r,j}(h) = \frac{\sqrt{\pi}}{2} \frac{S_j(h)}{\sqrt{A_{c,j}(h)}}, \quad (1)$$

where j denotes the indenter geometry, $S_j(h)$ is the contact stiffness obtained from the CSM signal, and $A_{c,j}(h)$ is the projected contact area evaluated from the calibrated indenter area function. For an ideal Berkovich indenter, the projected area is

$$A_{c,Berk}(h_c) = 24.5h_c^2, \quad (2)$$

whereas, for an ideal spherical indenter with radius R , the projected contact area is

$$A_{c,sph}(h_c) = \pi a_c^2 = \pi (2Rh_c - h_c^2). \quad (3)$$

In the present analysis, instrument-calibrated area functions were used to account for deviations from ideal indenter geometries. The reduced modulus was then converted to the sample modulus according to

$$\frac{1}{E_{r,j}(h)} = \frac{1 - \nu_s^2}{E_{s,j}(h)} + \frac{1 - \nu_i^2}{E_i}, \quad (4)$$

where an assumed Poisson's ratio of $\nu_s = 0.3$ was used for the sintered Cu structures, which is close to values commonly adopted for sintered Cu [25]. The elastic constants of the diamond indenter were taken as $E_i = 1141$ GPa and $\nu_i = 0.07$. The resulting $E_{s,j}(h)$ is therefore interpreted as an indentation-derived elastic modulus.

Under spherical indentation, the elastic modulus rapidly converges to a relatively stable, depth-independent range, whereas Berkovich indentation shows larger transient fluctuations and greater scatter, particularly at larger penetration depths. The indentation-derived elastic moduli obtained using the spherical indenter are 60.9 ± 2.7 GPa, 90.5 ± 3.9 GPa and 7.6 ± 1.0 GPa for the monomodal, bimodal and flake-shaped structures, respectively. The corresponding Berkovich values are 57.0 ± 4.5 GPa, 85.5 ± 10.9 GPa and 8.1 ± 1.5 GPa. Both indenter geometries yield the same morphology-dependent ranking, with the bimodal structure showing the highest elastic modulus, followed by the monomodal and

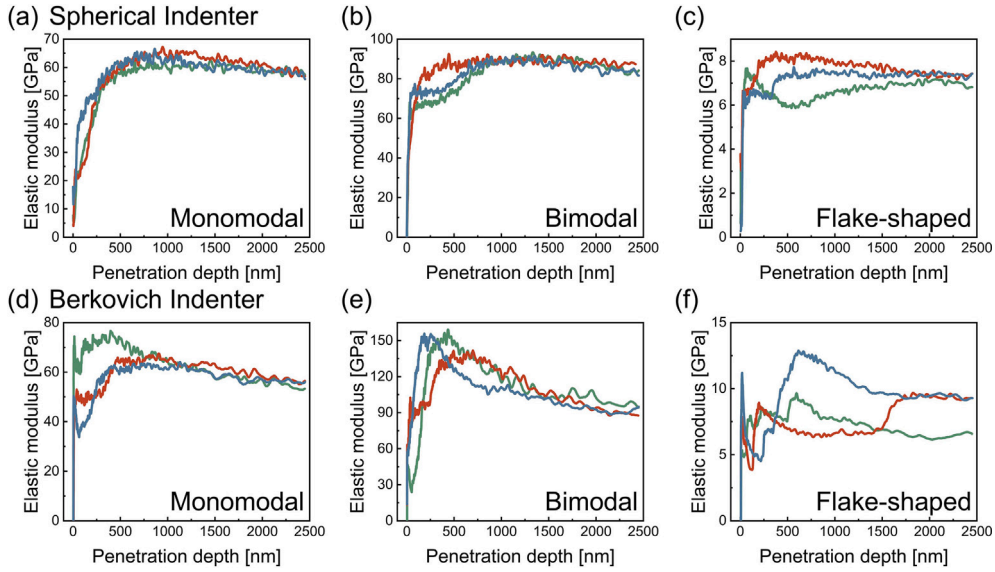


Fig. 5. Indentation-derived elastic modulus of sintered Cu structures: (a–c) spherical indentation results for the monomodal, bimodal and flake-shaped structures, respectively; (d–f) Berkovich indentation results for the corresponding structures.

flake-shaped structures. The larger scatter under Berkovich indentation indicates its greater sensitivity to local microstructural heterogeneity, whereas the lower scatter under spherical indentation suggests a more spatially averaged elastic response [71].

To further evaluate the morphology-dependent plastic response, spherical indentation was used to construct indentation-derived hardness-strain and representative stress-strain curves. The projected contact radius was first obtained from the contact area as

$$a_c(h) = \sqrt{\frac{A_c(h)}{\pi}}, \quad (5)$$

and the representative indentation strain was calculated as

$$\varepsilon_i(h) = \frac{4h_c}{3\pi a_c(h)}. \quad (6)$$

The indentation hardness was then determined from

$$H(h) = \frac{P(h)}{A_c(h)}, \quad (7)$$

where $P(h)$ is the applied load and $A_c(h)$ is the projected contact area determined from the calibrated spherical indenter area function. The representative indentation stress was estimated by converting hardness using the depth-dependent constraint factor:

$$\sigma_r(h) = \frac{H(h)}{C^*(h)}, \quad (8)$$

where $C^*(h)$ is the constraint factor obtained from spherical indentation analysis and is presented in Fig. S1 [72], following an approach that has also been applied to porous metallic systems [73]. The resulting σ_r should be interpreted as an indentation-derived representative stress reflecting the local constrained deformation response beneath the spherical indenter.

Fig. 6 compares the hardness-strain and representative stress-strain responses obtained using the spherical indenter. All three morphologies exhibit a continuous increase in hardness and representative stress at low strain, followed by gradual saturation at larger strain. The monomodal structure shows the highest indentation resistance, with hardness reaching approximately 2.3 – 2.4 GPa and representative stress approaching

850 – 900 MPa. The bimodal structure exhibits an intermediate response, with hardness of about 1.7 – 1.8 GPa and representative stress of 600 – 650 MPa, whereas the flake-shaped structure shows much lower values of only 0.24 – 0.26 GPa and below 100 MPa, respectively. These results demonstrate that the plastic resistance of sintered Cu is strongly governed by particle morphology, reflecting differences in load-bearing connectivity and densification capacity.

Fig. 7 further summarizes the elastic modulus, indentation yield strength, and Berkovich hardness. Here, the indentation yield strength was determined from the spherical-indentation representative stress-strain curve using the 0.2% offset criterion [74] and is used as a local comparative measure of the onset of plastic deformation [70]. Compared with dense bulk Cu (110–130 GPa), the monomodal and bimodal structures retain a considerable fraction of the intrinsic Cu stiffness, whereas the flake-shaped structure shows a severe loss of elastic load-bearing capability [75–77]. Notably, the plastic resistance does not follow the same order as the elastic modulus. Although the bimodal structure exhibits the highest elastic modulus, the monomodal structure shows the highest indentation yield strength and hardness. This contrast suggests that elastic load transfer is enhanced in the bimodal structure by the coarse-particle framework and fine-particle filling, whereas plastic resistance is more effectively promoted in the monomodal structure by its relatively uniform particle-neck network. The flake-shaped structure represents the lower bound in both stiffness and plastic resistance, reflecting weak through-thickness connectivity, platelet morphology, and high residual porosity.

3.3. SEM characterization

Fig. 8 compares the residual indentation morphologies and deformation mechanisms of the three sintered Cu structures. The top-view SEM images show that all structures undergo irreversible plastic deformation, but the deformation morphology is strongly dependent on the initial particle morphology (Fig. 8a–c). Fig. 8(d) and (g) reveal a compacted zone beneath the indent, characterized by pore closure and localized severe plastic deformation (LSPD) [78,79]. The bimodal structure exhibits a more heterogeneous deformation morphology (Fig. 8e and h). The co-existence of large and small particles creates a hierarchical contact network, in which the fine particles partially fill the interstitial regions and form additional contacts between neighbouring coarse particles. During the initial elastic stage of indentation, this mixed particle network can

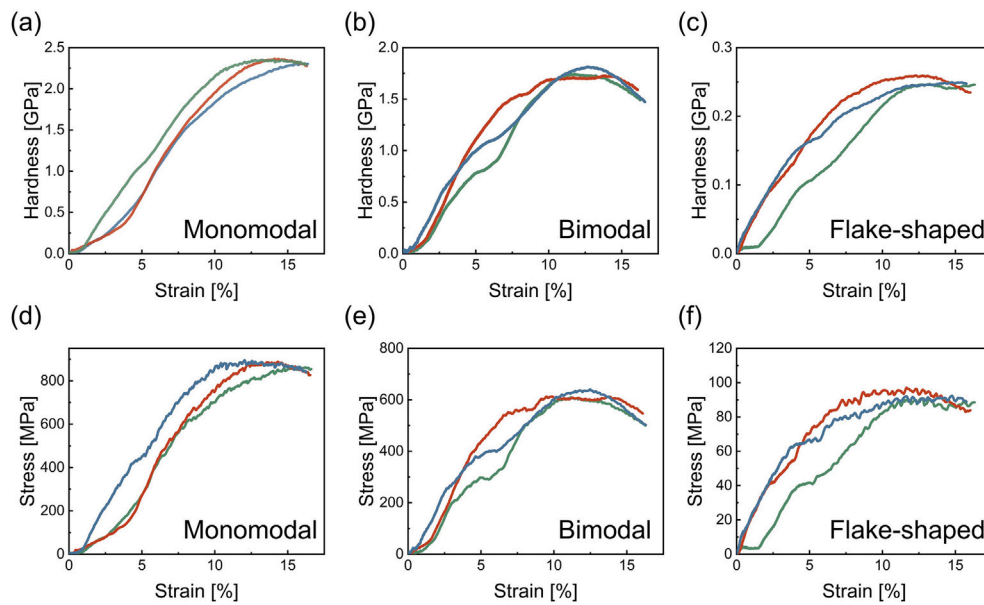


Fig. 6. (a–c) Hardness–strain curves of the monomodal, bimodal, and flake-shaped structures, respectively; (d–f) corresponding stress–strain curves obtained using the spherical indenter.

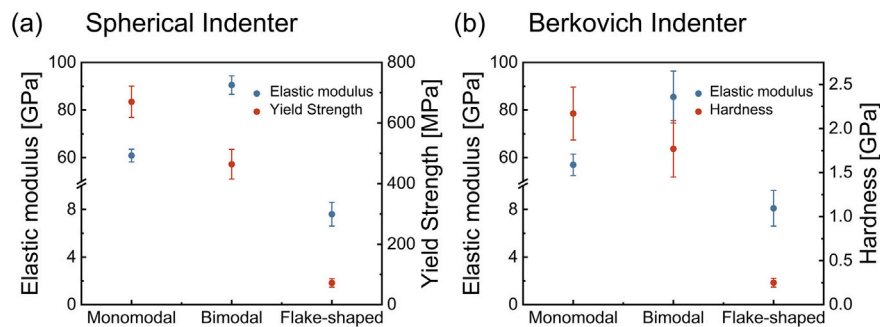


Fig. 7. (a) elastic modulus and indentation yield strength obtained from spherical indentation of the monomodal, bimodal and flake-shaped structures. (b) elastic modulus and indentation hardness obtained from Berkovich indentation of the corresponding structures.

increase the effective contact stiffness by reducing the compliance associated with large open pores and by shortening the load-transfer path through a coarse-particle skeleton, which may contribute to the higher elastic modulus than the monomodal structure [80]. However, the same structural heterogeneity becomes detrimental once plastic deformation is activated. In the bimodal structure, plastic deformation is primarily accommodated by the coarse-particle framework. The larger particle/grain domains provide longer dislocation glide distances and fewer internal barriers to plastic flow than the finer monomodal network, thereby potentially facilitating earlier plastic yielding once the contact stress exceeds the local resistance [81,82]. Meanwhile, the surrounding fine-particle regions and non-uniform necks constrain the deformation of the coarse particles and introduce local deformation incompatibility at coarse/fine-particle interfaces. As a result, plastic strain becomes heterogeneously partitioned within the bimodal network, with pronounced local distortion and discontinuous pore collapse developing around particle contacts and heterogeneous neck regions (Fig. 8e and h). This deformation partitioning may help explain why the bimodal structure exhibits a higher elastic modulus but a lower indentation yield strength and hardness than the monomodal structure.

In contrast, the flake-shaped structure exhibits a distinct deformation mode that appears to be associated with its incomplete interparticle contact during sintering. Owing to the relatively high stiffness of Cu, the

high-aspect-ratio platelets are difficult to fully conform to neighbouring flakes during consolidation [83,84]. Interparticle bending and geometric mismatch therefore limit intimate face-to-face contact and neck development, leaving elongated pores and weak through-thickness connectivity in the sintered network. Under indentation, these structural features may interrupt vertical load transfer and promote deformation through platelet bending, interlayer sliding, local buckling, and collapse of elongated pore regions (Fig. 8f and i), rather than by more homogeneous pore compaction within an isotropic particle-neck network.

Fig. 9 shows cross-sectional SEM images of the monomodal and bimodal sintered Cu structures after Berkovich nanoindentation. In contrast to spherical indentation, which imposes a relatively broad contact field and promotes distributed subsurface compaction (Fig. 8), the sharp Berkovich geometry concentrates deformation beneath the indenter and generates stronger local strain gradients (Fig. 9) [85]. In the monomodal structure, the deformed region remains comparatively continuous, with pore compression and particle-neck distortion distributed over a relatively homogeneous network (Fig. 9a). This indicates that the sharp-contact deformation can still be accommodated by collective deformation of interconnected particles and necks. By contrast, the bimodal structure exhibits a more heterogeneous deformation field after Berkovich indentation. The cross-section shows pronounced deformation incompatibility between coarse particles and surrounding

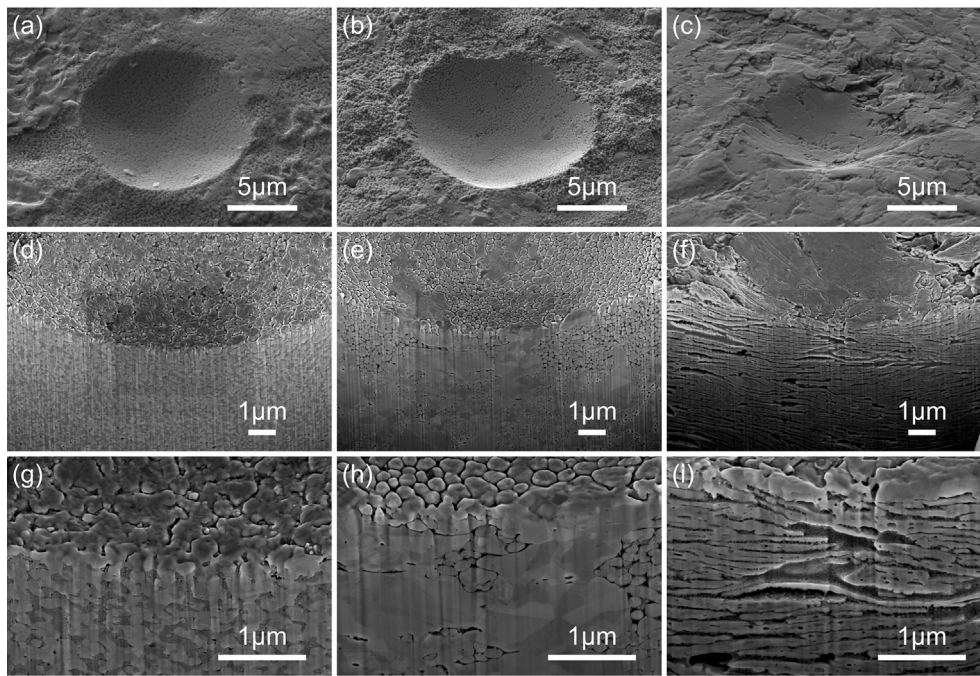


Fig. 8. (a–c) Top-view SEM images of the spherical indentation imprints in the monomodal, bimodal, and flake-shaped structures, respectively; (d–f) corresponding cross-sectional SEM images showing the deformation zone beneath the indents; (g–i) high-magnification cross-sectional images of the regions beneath the indentation sites, revealing morphology-dependent local deformation and pore/neck rearrangement.

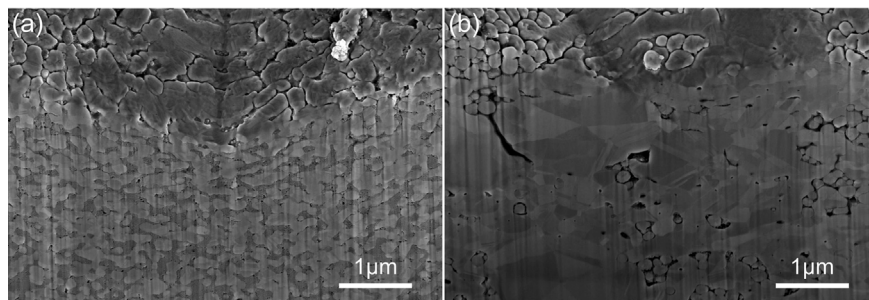


Fig. 9. Cross-sectional SEM images of the (a) monomodal and (b) bimodal sintered Cu structures after Berkovich nanoindentation.

fine-particle regions, together with discontinuous pore collapse and localized cracking or separation near particle contacts (Fig. 9b). These features indicate that the sharp stress field preferentially activates structurally weak regions associated with non-uniform particle packing and heterogeneous neck formation.

3.4. Sintering by MD simulation

To elucidate how particle morphology governs the formation and evolution of sintered structures, Fig. 10 presents the atomic configurations of the monomodal, bimodal and flake-shaped Cu structures at different sintering stages. In monomodal structure, the relatively uniform particle size leads to a spatially homogeneous distribution of sintered necks (Fig. 10a1–a3) [86]. During sintering, the residual pores are progressively reduced, while lattice defects and local structural disorder are mainly concentrated around interparticle necks and pore surfaces (Fig. 10a1,a2). Consistently, the local strain maps in Fig. S2 show that deformation is initially concentrated around interparticle necks and progressively extends through the network with increasing compressive strain. Meanwhile, adjacent particles gradually coalesce at contact regions, resulting in the formation of relatively uniform sintered necks throughout the network (Fig. 10a3). This suggests that

sintering in the monomodal structure is primarily accommodated by distributed neck plasticity, gradual pore collapse and relatively homogeneous particle-level load transfer. In the bimodal structure, by contrast, the incorporation of large particles disrupts the relatively homogeneous packing topology observed in the monomodal network and generates a hierarchical sintered structure (Fig. 10b1–b3). The large particles form a coarse load-bearing framework, while the smaller particles occupy interstitial regions and participate in local densification during compression. Consistently, the local strain maps in Fig. S3 reveal heterogeneous strain localization along interparticle necks and large-small particle interfaces, which intensifies with increasing compressive strain. This size-asymmetric arrangement produces a more heterogeneous contact network, with defects preferentially accumulating near large-small particle interfaces, narrow interparticle gaps and incompletely packed regions [87]. The flake-shaped structure exhibits a highly anisotropic lamellar architecture (Fig. 10c1–c3). Its pores are predominantly elongated and aligned along inter-flake regions, rather than being equiaxed and spatially isolated. The pressure-assisted compaction during sintering initially preserves some elongated voids between the aligned platelets. With increasing applied strain, the platelet stacks become progressively flattened, leading to the collapse of layer-parallel pores and the formation of a denser yet anisotropic network (Fig. S4).

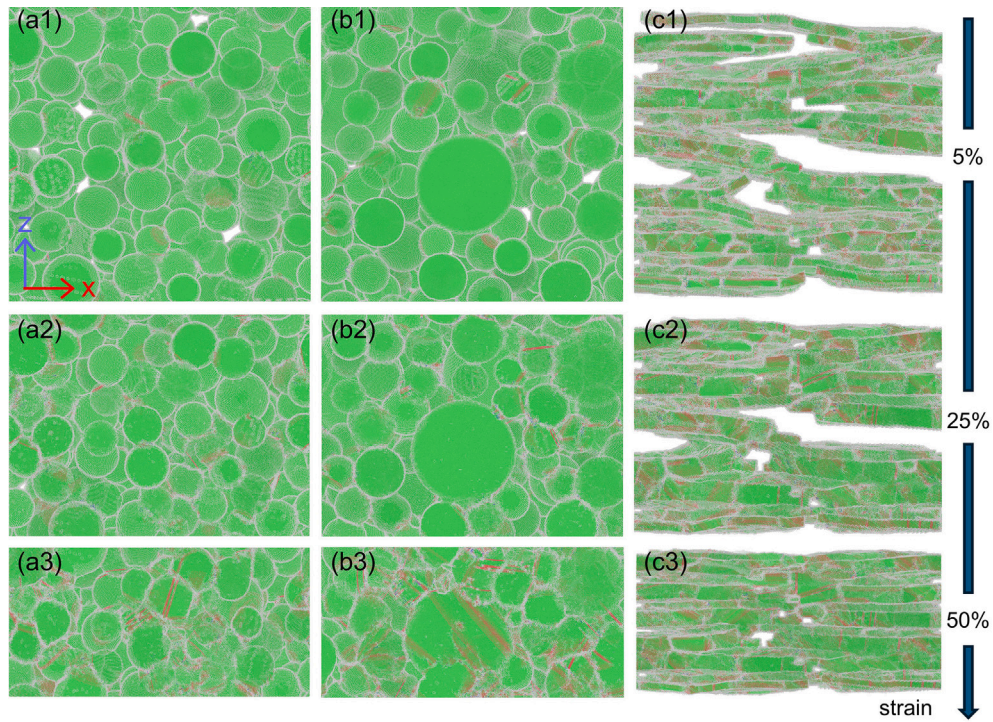


Fig. 10. Side-view atomic configurations of the simulated sintered Cu structures at different compressive strains: (a1–a3) Monomodal, (b1–b3) bimodal and (c1–c3) flake-shaped structures at compressive strains of 5%, 25% and 50%, respectively.

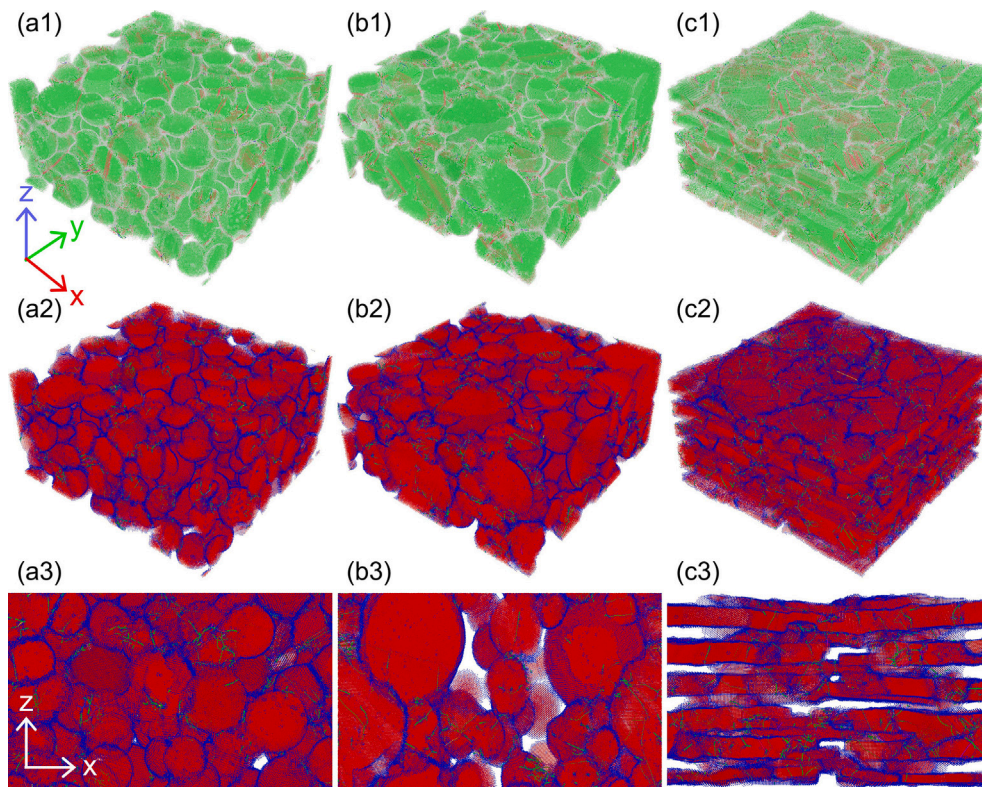


Fig. 11. (a1–c1) CNA maps of the monomodal, bimodal and flake-shaped structures after sintering, respectively; (a2–c2) corresponding distributions of dislocations in the three structures; (a3–c3) cross-sectional views showing the internal defect distribution and strain localization within the sintered networks.

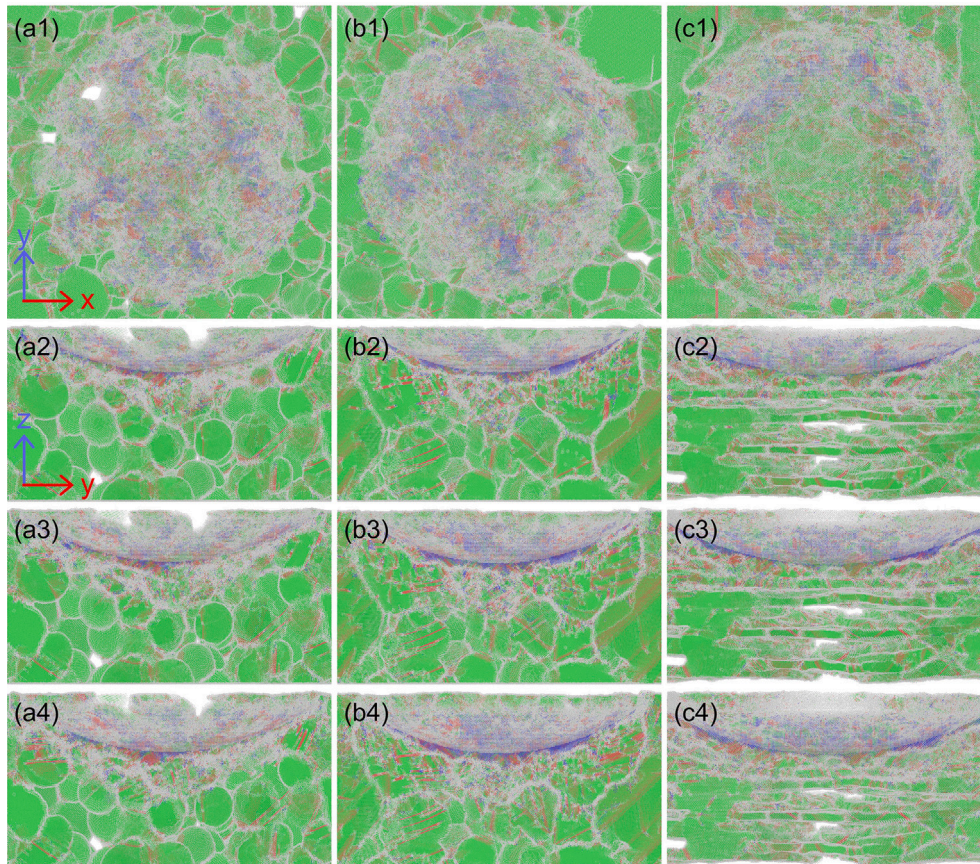


Fig. 12. (a1–c1) Top-view configurations of the indented structures under spherical indentation. (a2–c2), (a3–c3), and (a4–c4) are corresponding side-view configurations at indentation depths of 50, 60, and 70 Å, respectively, showing the morphology-dependent evolution of subsurface deformation.

Fig. 11 further reveals the morphology-dependent dislocation activity and defect distribution within the sintered Cu networks. In the monomodal structure, the high number density of small particles produces abundant interparticle necks and pore-adjacent boundaries, where dislocations are extensively generated and accumulated (Fig. 11a1). The closely spaced necks and particle interfaces act as effective barriers to dislocation motion, promoting dislocation pinning and the formation of a highly interconnected defect network (Fig. 11a2) [88]. This high dislocation density around the necked particle framework provides an atomistic basis for the enhanced plastic resistance of the monomodal structure (Fig. 11a3). Compared with the monomodal structure, the bimodal structure exhibits a more heterogeneous packing topology, characterized by coarse particles embedded within a fine-particle-rich matrix and separated by irregular interstitial pores (Fig. 11b1). This structural partitioning is reflected in the dislocation distribution: dislocations do not uniformly accumulate around all sintered necks, but are preferentially generated within large-particle interiors and near large-small particle interfaces, where curvature mismatch and deformation incompatibility are more pronounced (Fig. 11b2,b3). Meanwhile, the fine-particle-rich regions mainly accommodate local compaction, rearrangement and interface adjustment. Therefore, plastic deformation in the bimodal structure is governed by coordinated coarse-fine particle interactions, combining skeleton-supported load transfer, interfacial defect accumulation and fine-particle-assisted densification [89]. The flake-shaped structure exhibits a distinct dislocation pattern, with defects preferentially aligned along platelet interiors, interlamellar interfaces and layer-parallel pores (Fig. 11c1). Such a quasi-planar defect distribution reflects the dominance of platelet bending, interlayer sliding and anisotropic pore collapse, rather than homogeneous dislocation

accumulation (Fig. 11c2). Consequently, the planar, interface-guided defect evolution inhibits the development of a three-dimensionally interconnected dislocation-pinning network, limiting effective dislocation storage and reducing the capacity for sustained plastic resistance (Fig. 11c3).

3.5. Nanoindentation by MD simulation

To provide an atomistic interpretation of the morphology-dependent deformation trends observed experimentally, MD simulations were performed to examine dislocation activity, pore collapse and sintered-neck deformation. Fig. 12 presents the atomistic configurations of the monomodal, bimodal and flake-shaped Cu structures under spherical indentation, showing the top-view residual deformation morphology and the subsurface deformation at indentation depths of 50, 60 and 70 Å. In the monomodal structure, the spherical particles beneath the indenter undergo progressive neck deformation, pore closure and local compaction (Fig. 12a1–a4). This deformation pattern is consistent with the SEM cross-sections in Fig. 8(d) and (g). With increasing indentation depth, defect activity initiates mainly around sintered necks, particle boundaries and pore surfaces, and then spreads laterally and downward (Fig. 12a2,a3). The progressive expansion of the defect-rich zone with increasing depth is further illustrated in Fig. S5. The motion of these dislocations is repeatedly impeded and pinned, promoting local dislocation storage and the development of an interconnected defect network (Fig. 12a4). This pinning effect suppresses unrestricted long-range dislocation glide and enables the applied deformation to be accommodated through distributed LSPD rather than through a single dominant collapse path [90]. The bimodal structure shows a more heterogeneous deformation field (Fig. 12b1–b4). The fine particles in

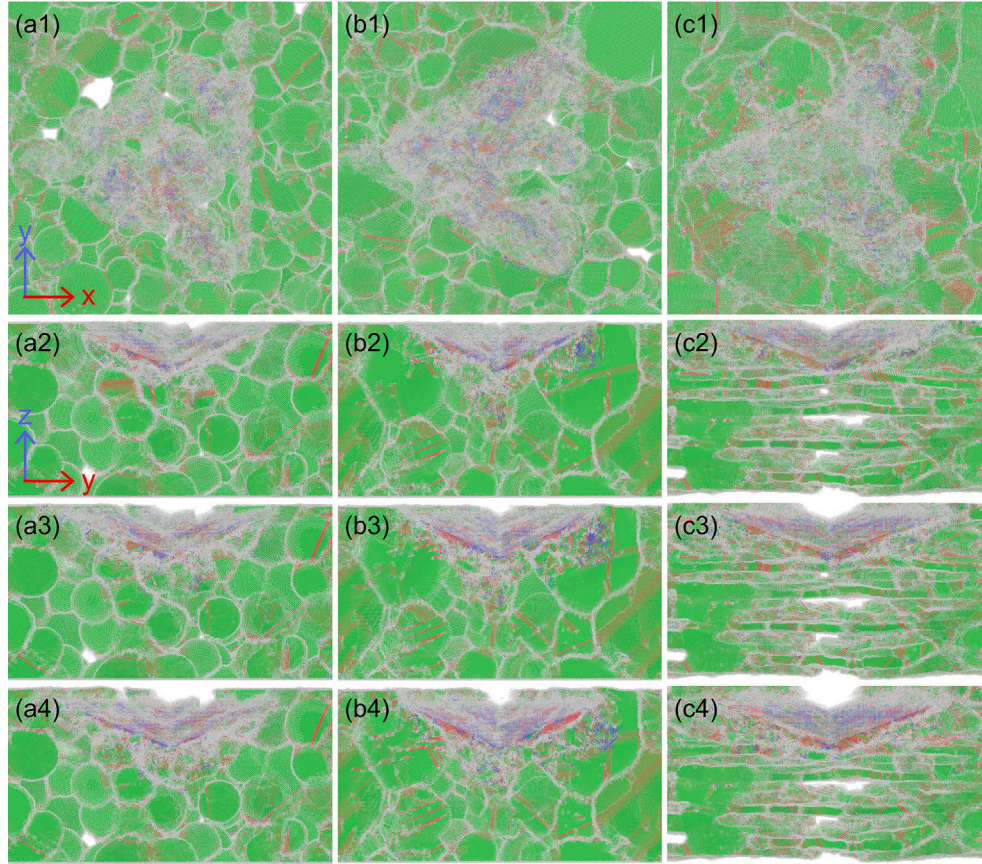


Fig. 13. (a1–c1) Top-view configurations of the indented structures under Berkovich indentation. (a2–c2), (a3–c3), and (a4–c4) are the corresponding side-view configurations at indentation depths of 50, 60, and 70 Å, respectively, showing the morphology-dependent evolution of subsurface deformation.

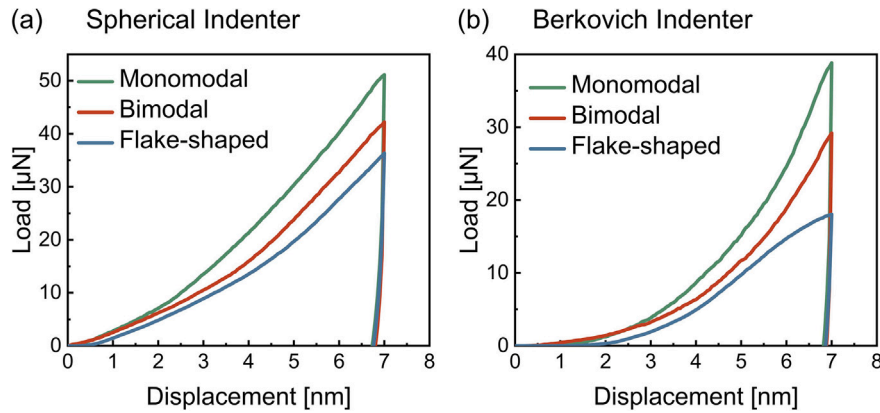


Fig. 14. Load–penetration depth responses of sintered Cu structures with different particle morphologies under MD indentation tests: (a) spherical indentation curves for the monomodal, bimodal and flake-shaped structures, respectively. (b) Berkovich indentation curves for the corresponding structures.

the interstitial regions are preferentially compacted and rearranged, whereas the large particles remain the primary load-bearing units. This agrees with the SEM observation in Fig. 8(e) and (h). In the atomistic configurations, defect accumulation is enhanced near large/fine particle interfaces, where curvature mismatch and local packing heterogeneity generate stronger stress concentrations [91]. The corresponding cross-sections in Fig. S6 further show the concentrated and spatially irregular evolution of this defect-rich region. Therefore, compared with the monomodal structure, the bimodal network develops sharper strain gradients and more localized defect nucleation. The flake-shaped structure

exhibits the most anisotropic deformation mode (Fig. 12c1–c4). Defect activity is concentrated along platelet interfaces. As further shown in Fig. S7, defects propagate preferentially along platelet interfaces and layer-parallel regions, indicating that plasticity is governed primarily by morphology-controlled interlayer deformation.

Fig. 13 shows the atomistic deformation of the three sintered Cu structures under Berkovich indentation. Owing to its sharper tip geometry, the Berkovich indenter produces a more localized contact field than the spherical indenter, leading to more concentrated defect nucleation and accumulation beneath the indentation region. This localized

response is more pronounced in the bimodal and flake-shaped structures, where non-uniform neck geometry, particle-size mismatch and interlamellar contacts promote spatially heterogeneous deformation. Figs. (S8–S10) further show a comparatively uniform defect distribution in the monomodal structure, stronger localization around coarse-fine particle interfaces in the bimodal structure, and interface-guided defect evolution in the flake-shaped structure. The morphology-dependent deformation behavior is also reflected in the simulated load-penetration responses shown in Fig. 14. For both spherical and Berkovich indentation, the indentation load increases continuously with penetration depth, indicating progressive resistance associated with neck deformation, pore collapse and local densification. Overall, the simulations demonstrate that particle morphology governs deformation localization within the porous network. Deformation is more distributed in the monomodal structure, whereas the bimodal and flake-shaped structures concentrate strain near heterogeneous necks, particle interfaces, and weak interlayer regions. Differences in absolute load between simulation and experiment arise primarily from the limited scale, reduced contact volume, and idealized geometry of the MD models.

4. Conclusion

This study elucidates the role of particle morphology in the mechanical response of pressure-assisted sintered Cu structures. By integrating spherical and Berkovich nanoindentation, SEM characterization and MD simulations, it reveals how monomodal, bimodal and flake-shaped structures influence deformation localization and indentation resistance. The main findings are summarized as follows:

1. Particle morphology strongly governs the mechanical response of sintered Cu networks. The monomodal structure exhibits the highest hardness and indentation yield strength, owing to its relatively uniform particle-neck network, distributed load-transfer paths and enhanced dislocation pinning within the finer particle/grain domains. The bimodal structure shows the highest elastic modulus, which can be attributed to the stiff load-bearing framework of coarse particles and the filling of interstitial regions by fine particles, thereby improving network connectivity and constraining elastic deformation. In contrast, the flake-shaped structure exhibits the lowest modulus and strength because of weak connectivity, interflake sliding and elongated pore collapse.
2. Both indenters reproduce the same morphology-dependent ranking of mechanical properties. Spherical indentation yields slightly higher elastic moduli and lower scatter than Berkovich indentation, reflecting a more spatially averaged response of the porous network. In contrast, the sharper Berkovich tip probes a more confined volume and is therefore more sensitive to local microstructural heterogeneity.
3. Smooth loading curves and the absence of distinct pop-in events indicate a continuous elastic–plastic transition governed by neck deformation, pore closure, local densification, and particle rearrangement. SEM observations of residual impressions, collapsed pores, and morphology-dependent deformation zones further show that the deformation is mainly accommodated by irreversible collapse of the pores and localized plastic flow in mechanically weak regions.
4. MD simulations provide atomistic insight into the morphology-dependent deformation mechanisms underlying the experimental trends. The monomodal structure accommodates indentation through relatively distributed neck deformation and defect accumulation, whereas the bimodal structure exhibits load sharing between coarse-particle frameworks and fine-particle regions. In the flake-shaped structure, deformation is mainly concentrated around weak interlamellar contacts and elongated pore regions.

In summary, this study demonstrates that particle morphology is a key design parameter governing the local mechanical response of porous

sintered Cu. By establishing direct links among particle packing, pore-neck topology, indenter-dependent deformation, and atomistic plastic mechanisms, this work provides a mechanistic framework for understanding local stiffness, strength, and damage evolution in sintered metallic networks. These insights are particularly relevant to the mechanical design of sintered Cu interconnects and die-attach layers in power electronic packaging, where morphology-dependent stiffness and deformation accommodation can influence local stress transfer under thermomechanical loading, while further practical optimization should also consider thermal and electrical transport requirements.

CRedit authorship contribution statement

Tianxing Du: Writing – original draft, Methodology, Investigation. **Chenshan Gao:** Methodology, Investigation. **Lai Wei:** Methodology, Investigation. **Huaiyu Ye:** Writing – review & editing, Supervision. **Guoqi Zhang:** Validation, Supervision. **Leiming Du:** Writing – review & editing, Supervision.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Acknowledgements

The authors acknowledge financial support from the 2023TQ07C475 program, the Shenzhen Major Science and Technology Projects (KJZD20240903102309013), the State Key Laboratory of Electrical Insulation and Power Equipment (EIPE26205), and the Shenzhen Polytechnic University Research Fund (Grant No. 6026310016K). This work was also supported by the ECSEL Joint Undertaking (JU) under grant agreement No. 876659, with support from the European Union's Horizon 2020 research and innovation programme and the participating countries.

Appendix A. Supplementary data

Supplementary data to this article can be found online at doi: [10.1016/j.matdes.2026.117159](https://doi.org/10.1016/j.matdes.2026.117159).

Data availability

Data will be made available on request.

References

- [1] A. Wang, Y. Ma, D. Zhao, Pore engineering of porous materials: effects and applications, *ACS Nano* 18 (34) (2024) 22829–22854.
- [2] R. Luque, A. Ahmad, S. Tariq, M. Mubashir, M.S. Javed, S. Rajendran, R.S. Varma, A. Ali, C. Xia, Functionalized interconnected porous materials for heterogeneous catalysis, energy conversion and storage applications: recent advances and future perspectives, *Mater. Today* 73 (2024) 105–129.
- [3] Z.A. Sandhu, M.A. Raza, N.S. Awwad, H.A. Ibrahim, U. Farwa, S. Ashraf, A. Dildar, E. Fatima, S. Ashraf, F. Ali, Metal–organic frameworks for next-generation energy storage devices; a systematic review, *Mater. Adv.* 5 (1) (2024) 30–50.
- [4] S. Aziz, A.R. Mazhar, A. Ubaid, S.M.H. Shah, Y. Riaz, T. Talha, D.-W. Jung, A comprehensive review of membrane-based water filtration techniques, *Appl. Water Sci.* 14 (8) (2024) 169.
- [5] M.U. Shahid, T. Najam, M. Islam, A.M. Hassan, M.A. Assiri, A. Rauf, A. Ur Rehman, S.S.A. Shah, M.A. Nazir, Engineering of metal organic framework (MOF) membrane for waste water treatment: synthesis, applications and future challenges, *J. Water Process Eng.* 57 (2024) 104676.
- [6] Y. Yang, Z. Zhou, X. Chu, X. Tang, M. Zhou, W. Zhou, T. Fu, 3D-printed zeolite with combined structure for xylene isomerization, *Mater. Des.* 219 (2022) 110744.
- [7] Z. Alsudairy, N. Brown, A. Campbell, A. Ambus, B. Brown, K. Smith-Petty, X. Li, Covalent organic frameworks in heterogeneous catalysis: recent advances and future perspective, *Mater. Chem. Front.* 7 (16) (2023) 3298–3331.
- [8] P. Zhu, P.R. Slater, E. Kendrick, Insights into architecture, design and manufacture of electrodes for lithium-ion batteries, *Mater. Des.* 223 (2022) 111208.
- [9] D. Gastol, M. Capener, C. Reynolds, C. Constable, E. Kendrick, Microstructural design of printed graphite electrodes for lithium-ion batteries, *Mater. Des.* 205 (2021) 109720.

- [10] Y. Li, Y. Liu, H. Chen, A. Zhang, Y. Zhang, J. Zhang, B. Chen, Q. Han, J. Wang, From clinic to lab: advances in porous titanium-based orthopedic implant research, *J. Mater. Res. Technol.* 30 (2024) 3780–3806.
- [11] C. Song, L. Liu, Z. Deng, H. Lei, F. Yuan, Y. Yang, Y. Li, J. Yu, Research progress on the design and performance of porous titanium alloy bone implants, *J. Mater. Res. Technol.* 23 (2023) 2626–2641.
- [12] X. Ji, L. Du, S. He, H. van Zeijl, G. Zhang, Microstructural and micromechanical characterization of sintered nano-copper bump for flip-chip heterogeneous integration, *Microelectron. Reliab.* 150 (2023) 115180.
- [13] L. Du, W. Jiao, O. Bäcke, M.H. Colliander, R.H. Poelma, J. Fan, W.D. van Driel, X. Fan, G. Zhang, Interface strength and crack propagation mechanisms in sintered copper nanoparticles, *Acta Mater.* 296 (2025) 121187.
- [14] L. Du, K. Liu, D. Hu, O. Bäcke, X. Hu, X. Ji, J. Fan, R.H. Poelma, M.H. Colliander, G. Zhang, Microstructural and mechanical anisotropy in pressure-assisted sintered copper nanoparticles, *Acta Mater.* 287 (2025) 120772.
- [15] L. Du, G. Schaffar, W. Jiao, K. Liu, R.H. Poelma, V. Maier-Kiener, J. Fan, D. Kiener, W.D. van Driel, X. Fan, et al., Tailoring microstructure and mechanical properties of sintered Cu nanoparticles, *Acta Mater.* (2025) 121501.
- [16] G. Kasperovich, J. Haubrich, J. Gussone, G. Requena, Correlation between porosity and processing parameters in TiAl6V4 produced by selective laser melting, *Mater. Des.* 105 (2016) 160–170.
- [17] S. Ahmadi, R. Hedayati, R.A.K. Jain, Y. Li, S. Leeftang, A. Zadpoor, Effects of laser processing parameters on the mechanical properties, topology, and microstructure of additively manufactured porous metallic biomaterials: a vector-based approach, *Mater. Des.* 134 (2017) 234–243.
- [18] T. Ronneberg, C.M. Davies, P.A. Hooper, Revealing relationships between porosity, microstructure and mechanical properties of laser powder bed fusion 316 L stainless steel through heat treatment, *Mater. Des.* 189 (2020) 108481.
- [19] C. Chen, D. Kim, Y. Liu, T. Sekiguchi, Y. Su, X. Long, C. Liu, C. Liu, K. Suganuma, Development of micron-sized Cu–Ag composite paste for oxidation-free bare Cu bonding in air condition and its deterioration mechanism during aging and power cycling tests, *J. Mater. Res. Technol.* 24 (2023) 8967–8983.
- [20] Q. Jia, G. Zou, W. Wang, H. Ren, H. Zhang, Z. Deng, B. Feng, L. Liu, Sintering mechanism of a supersaturated Ag–Cu nanoalloy film for power electronic packaging, *ACS Appl. Mater. Interfaces* 12 (14) (2020) 16743–16752.
- [21] D. Mondal, M. Patel, H. Jain, A. Jha, S. Das, R. Dasgupta, The effect of the particle shape and strain rate on microstructure and compressive deformation response of pure Ti-foam made using acrowax as space holder, *Mater. Sci. Eng. A* 625 (2015) 331–342.
- [22] C. Chen, K. Suganuma, Microstructure and mechanical properties of sintered Ag particles with flake and spherical shape from nano to micro size, *Mater. Des.* 162 (2019) 311–321.
- [23] P. Suri, R.P. Koseski, R.M. German, Microstructural evolution of injection molded gas-and water-atomized 316 L stainless steel powder during sintering, *Mater. Sci. Eng. A* 402 (1–2) (2005) 341–348.
- [24] A.Y. Kumar, J. Wang, Y. Bai, S.T. Huxtable, C.B. Williams, Impacts of process-induced porosity on material properties of copper made by binder jetting additive manufacturing, *Mater. Des.* 182 (2019) 108001.
- [25] A. Wijaya, B. Eichinger, F. Chamaseamani, B. Sartory, R. Hammer, V. Maier-Kiener, D. Kiener, M. Mischitz, R. Brunner, Multi-method characterization approach to facilitate a strategy to design mechanical and electrical properties of sintered copper, *Mater. Des.* 197 (2021) 109188.
- [26] S. Ghouse, S. Babu, R.J. Van Arkel, K. Nai, P.A. Hooper, J.R. Jeffers, The influence of laser parameters and scanning strategies on the mechanical properties of a stochastic porous material, *Mater. Des.* 131 (2017) 498–508.
- [27] Y. Li, X. Wang, M. Liu, T. Gu, X. Long, Orientation-induced fluctuations of elastic modulus by Berkovich nanoindentation, *Int. J. Mech. Sci.* (2026) 111367.
- [28] D.A. Lucca, K. Herrmann, M. Klopstein, Nanoindentation: measuring methods and applications, *CIRP Ann.* 59 (2) (2010) 803–819.
- [29] O. Pshyk, O. Nakonechna, E. Coy, Current view of nanoindentation: recent developments and application in material characterization, *ACS Appl. Mater. Interfaces* 17 (25) (2025) 36275–36300.
- [30] K. Durst, V. Maier, Dynamic nanoindentation testing for studying thermally activated processes from single to nanocrystalline metals, *Curr. Opin. Solid State Mater. Sci.* 19 (6) (2015) 340–353.
- [31] V. Maier-Kiener, K. Durst, Advanced nanoindentation testing for studying strain-rate sensitivity and activation volume, *Jom* 69 (11) (2017) 2246–2255.
- [32] X. Zhang, S. Gholami, M. Khedmati, B. Cui, Y.-R. Kim, Y.-J. Kim, H.-S. Shin, J. Lee, Spark plasma sintering of a lunar regolith simulant: effects of parameters on microstructure evolution, phase transformation, and mechanical properties, *Ceram. Int.* 47 (4) (2021) 5209–5220.
- [33] X. Yan, L. Du, C. Gu, T. Tian, C. Gao, G. Zhang, J. Fan, Microscale mechanical properties in sintered copper nanoparticles, *Mater. Sci. Eng. A* (2025) 148684.
- [34] H. Besharatloo, M. de Nicolás-Morillas, M. Chen, A. Mateo, B. Ferrari, E. Gordo, E. Jiménez-Piqué, J.M. Wheeler, L. Llanes, Micropillar compression of Ti (C, N)-FeNi cermets: microstructural, processing, and scale effects, *J. Eur. Ceram. Soc.* 43 (7) (2023) 2826–2833.
- [35] D. Hu, Z. Guo, N. Grilli, A. Tay, Z. Lu, W. Yan, Understanding the strain localization in additively manufactured materials: micro-scale tensile tests and crystal plasticity modeling, *Int. J. Plast.* 177 (2024) 103981.
- [36] J. Stinville, M. Charpagne, R. Maaß, H. Proudhon, W. Ludwig, P. Callahan, F. Wang, I. Beyerlein, M. Echlin, T. Pollock, Insights into plastic localization by crystallographic slip from emerging experimental and numerical approaches, *Annu. Rev. Mater. Res.* 53 (1) (2023) 275–317.
- [37] J. Jiang, T.B. Britton, A.J. Wilkinson, The orientation and strain dependence of dislocation structure evolution in monotonically deformed polycrystalline copper, *Int. J. Plast.* 69 (2015) 102–117.
- [38] D. Hughes, N. Hansen, The microstructural origin of work hardening stages, *Acta Mater.* 148 (2018) 374–383.
- [39] Y. Wang, D. Raabe, C. Klüber, F. Roters, Orientation dependence of nanoindentation pile-up patterns and of nanoindentation microtextures in copper single crystals, *Acta Mater.* 52 (8) (2004) 2229–2238.
- [40] E. Andrews, L.J. Gibson, M. Ashby, The creep of cellular solids, *Acta Mater.* 47 (10) (1999) 2853–2863.
- [41] N. Huber, R. Viswanath, N. Mameka, J. Markmann, J. Weißmüller, Scaling laws of nanoporous metals under uniaxial compression, *Acta Mater.* 67 (2014) 252–265.
- [42] M. Huang, Z. Li, C. Wang, Discrete dislocation dynamics modelling of microvoid growth and its intrinsic mechanism in single crystals, *Acta Mater.* 55 (4) (2007) 1387–1396.
- [43] J. Segurado, J. Llorca, An analysis of the size effect on void growth in single crystals using discrete dislocation dynamics, *Acta Mater.* 57 (5) (2009) 1427–1436.
- [44] J. Segurado, J. Llorca, Discrete dislocation dynamics analysis of the effect of lattice orientation on void growth in single crystals, *Int. J. Plast.* 26 (6) (2010) 806–819.
- [45] H.-J. Guo, C. Ling, E.P. Busso, Z. Zhong, D.-F. Li, Crystal plasticity based investigation of micro-void evolution under multi-axial loading conditions, *Int. J. Plast.* 129 (2020) 102673.
- [46] U.B. Asim, M.A. Siddiq, M.E. Kartal, A CPFEM based study to understand the void growth in high strength dual-phase titanium alloy (Ti-10V-2Fe-3Al), *Int. J. Plast.* 122 (2019) 188–211.
- [47] S. Ha, K. Kim, Void growth and coalescence in fcc single crystals, *Int. J. Mech. Sci.* 52 (7) (2010) 863–873.
- [48] M. Dittmann, F. Aldakheel, J. Schulte, F. Schmidt, M. Krüger, P. Wriggers, C. Hesch, Phase-field modeling of porous-ductile fracture in non-linear thermo-elasto-plastic solids, *Comput. Methods Appl. Mech. Eng.* 361 (2020) 112730.
- [49] V. Kumar, Z. Fang, P. Fife, Phase field simulations of grain growth during sintering of two unequal-sized particles, *Mater. Sci. Eng. A* 528 (1) (2010) 254–259.
- [50] L. Du, W. Jiao, J. Fan, G. Zhang, Tailoring sintering pressure to optimize nanoparticle coarsening and mechanical strength in Cu-based interconnects, in: 2025 IEEE 75th Electronic Components and Technology Conference (ECTC), IEEE, 2025, pp. 1223–1229.
- [51] R. Wu, X. Zhao, Y. Liu, Atomic insights of Cu nanoparticles melting and sintering behavior in CuCu direct bonding, *Mater. Des.* 197 (2021) 109240.
- [52] W. Lv, J. Lv, J. Liu, C. Chen, Y. Kang, Sintering dynamic evolution and enhancement mechanism of nano-Cu/boron nitride composite matrix with excellent mechanical properties from the atomic perspective, *Compos. Struct.* 354 (2025) 118756.
- [53] W. Lv, S. Wei, J. Liu, J. Lv, J. Yang, Y. Zhang, C. Chen, Y. Kang, Trimodal mediated superior sintering and fracture resistance mechanism in nano-metal interconnection, *Mater. Des.* (2026) 116221.
- [54] M. Diehl, M. Wicke, P. Shanthraj, F. Roters, A. Brueckner-Foit, D. Raabe, Coupled crystal plasticity–phase field fracture simulation study on damage evolution around a void: pore shape versus crystallographic orientation, *Jom* 69 (5) (2017) 872–878.
- [55] R. Hari Krishnan, J.-B. le Graverend, A creep-damage phase-field model: predicting topological inversion in Ni-based single crystal superalloys, *Mater. Des.* 160 (2018) 405–416.
- [56] F. Roters, P. Eisenlohr, L. Hantcherli, D.D. Tjahjanto, T.R. Bieler, D. Raabe, Overview of constitutive laws, kinematics, homogenization and multiscale methods in crystal plasticity finite-element modeling: theory, experiments, applications, *Acta Mater.* 58 (4) (2010) 1152–1211.
- [57] O. Jamond, R. Gatti, A. Roos, B. Devincere, Consistent formulation for the discrete-continuous model: improving complex dislocation dynamics simulations, *Int. J. Plast.* 80 (2016) 19–37.
- [58] L. Du, X. Ji, K. Liu, J. Fan, G. Zhang, Investigating coalescence kinetics in Cu nanoparticle by molecular dynamics and 3D reconstruction, in: 2024 25th International Conference on Electronic Packaging Technology (ICEPT), IEEE, 2024, pp. 1–4.
- [59] W. Lv, S. Wei, J. Liu, J. Lv, C. Chen, Y. Kang, Deformation mediated densification and bonding strengthening driven through pressure-assisted sintering of two phase nano-metal composites, *Compos. Struct.* (2026) 120180.
- [60] J. Liu, W. Lv, Y. Mou, C. Chen, Y. Kang, Coalescence behavior of Cu nanoparticles during sintering: based on atomic scale to macro scale, *J. Mater. Res. Technol.* 27 (2023) 2490–2507.
- [61] J.C. Petit, M. Sperl, Pressure model and scaling laws in jammed bidisperse granular packings, *J. C. Petit, M. Sperl, Granul. Matter* 27 (1) (2025) 23.
- [62] W.C. Oliver, G.M. Pharr, An improved technique for determining hardness and elastic modulus using load and displacement sensing indentation experiments, *J. Mater. Res.* 7 (6) (1992) 1564–1583.
- [63] S. Plimpton, Fast parallel algorithms for short-range molecular dynamics, *J. Comput. Phys.* 117 (1) (1995) 1–19.
- [64] A.P. Thompson, H.M. Aktulga, R. Berger, D.S. Bolintineanu, W.M. Brown, P.S. Crozier, P.J. In't Veld, A. Kohlmeyer, S.G. Moore, T.D. Nguyen, et al., LAMMPS—a flexible simulation tool for particle-based materials modeling at the atomic, meso, and continuum scales, *Comput. Phys. Commun.* 271 (2022) 108171.
- [65] A. Stukowski, Visualization and analysis of atomistic simulation data with OVITO—the open visualization tool, *Model. Simul. Mater. Sci. Eng.* 18 (1) (2010) 015012.
- [66] P.L. Williams, Y. Mishin, J.C. Hamilton, An embedded-atom potential for the Cu–Ag system, *Model. Simul. Mater. Sci. Eng.* 14 (2006) 817–833.
- [67] J. Li, K.J. Van Vliet, T. Zhu, S. Yip, S. Suresh, Atomistic mechanisms governing elastic limit and incipient plasticity in crystals, *Nature* 418 (6895) (2002) 307–310.

- [68] N. Huber, I. Ryl, Y. Wu, M. Hablitzel, B. Zandersons, C. Richert, E. Lilleodden, Densification of nanoporous metals during nanoindentation: the role of structural and mechanical properties, *J. Mater. Res.* 38 (3) (2023) 853–866.
- [69] I. Zarudi, L. Zhang, W. Cheong, T. Yu, The difference of phase distributions in silicon after indentation with Berkovich and spherical indenters, *Acta Mater.* 53 (18) (2005) 4795–4800.
- [70] S. Pathak, S.R. Kalidindi, Spherical nanoindentation stress–strain curves, *Mater. Sci. Eng. R Rep.* 91 (2015) 1–36.
- [71] V. Králík, J. Němeček, Comparison of nanoindentation techniques for local mechanical quantification of aluminium alloy, *Mater. Sci. Eng. A* 618 (2014) 118–128.
- [72] G.J. Schaffar, D. Tscharnuter, V. Maier-Kiener, Exploring the high-temperature deformation behavior of monocrystalline silicon—an advanced nanoindentation study, *Mater. Des.* 233 (2023) 112198.
- [73] A. Leitner, V. Maier-Kiener, D. Kiener, High temperature flow behavior of ultra-strong nanoporous Au assessed by spherical nanoindentation, *Nanomaterials* 8 (6) (2018) 366.
- [74] H. Bei, S. Shim, G.M. Pharr, E.P. George, Effects of pre-strain on the compressive stress–strain response of MO-alloy single-crystal micropillars, *Acta Mater.* 56 (2008) 4762–4770.
- [75] N. Khiara, M. Coulombier, J.-P. Raskin, Y. Bréchet, T. Pardoen, F. Onimus, Helium nano-bubbles in copper thin films slows down creep under ion irradiation, *Acta Mater.* 288 (2025) 120854.
- [76] H. Ledbetter, E. Naimon, Elastic properties of metals and alloys. II. copper, *J. Phys. Chem. Ref. Data* 3 (4) (1974) 897–935.
- [77] S. Agnew, J. Weertman, The influence of texture on the elastic properties of ultrafine-grain copper, *Mater. Sci. Eng. A* 242 (1–2) (1998) 174–180.
- [78] Y.-C. Kim, E.-J. Gwak, S.-M. Ahn, J.-I. Jang, H.N. Han, J.-Y. Kim, Indentation size effect in nanoporous gold, *Acta Mater.* 138 (2017) 52–60.
- [79] D. Farkas, J. Stuckner, R. Umbel, B. Kuhr, M.J. Demkowicz, Indentation response of nanoporous gold from atomistic simulations, *J. Mater. Res.* 33 (10) (2018) 1382–1390.
- [80] A. Mostafaei, P.R. De Vecchis, I. Nettlehip, M. Chmielus, Effect of powder size distribution on densification and microstructural evolution of binder-jet 3D-printed alloy 625, *Mater. Des.* 162 (2019) 375–383.
- [81] G. Dirras, J. Gubicza, S. Ramtani, Q. Bui, T. Szilágyi, Microstructure and mechanical characteristics of bulk polycrystalline Ni consolidated from blends of powders with different particle size, *Mater. Sci. Eng. A* 527 (4–5) (2010) 1206–1214.
- [82] S. Lefebvre, B. Devincere, T. Hoc, Simulation of the hall–petch effect in ultra-fine grained copper, *Mater. Sci. Eng. A* 400 (2005) 150–153.
- [83] A. Jaeggi, A.-M. Eckel, R. Pini, A.K. Rajagopalan, M. Mazzotti, Exploring disordered packing of non-equant particles: insights from computed tomography and Monte Carlo simulations, *Powder Technol.* 452 (2025) 120468.
- [84] T. Du, C. Gao, O. Bäcké, L. Wei, H. Ye, G. Zhang, M.H. Colliander, L. Du, Particle morphology-dependent deformation mechanisms in sintered Cu structures, *Mater. Sci. Eng. A* (2026) 150688.
- [85] S. Jakob, A. Leitner, A. Lorch, M. Eidenberger-Schober, W. Knabl, R. Pippan, H. Clemens, V. Maier-Kiener, Influence of crystal orientation and Berkovich tip rotation on the mechanical characterization of grain boundaries in molybdenum, *Mater. Des.* 182 (2019) 107998.
- [86] B. Cheng, A.H. Ngan, The sintering and densification behaviour of many copper nanoparticles: a molecular dynamics study, *Comput. Mater. Sci.* 74 (2013) 1–11.
- [87] S. Martin, R. Parekh, M. Guessasma, J. Léchelle, J. Fortin, K. Saleh, Study of the sintering kinetics of bimodal powders. A parametric DEM study, *Powder Technol.* 270 (2015) 637–645.
- [88] A. Lange, A. Samanta, H. Majidi, S. Mahajan, J. Ging, T. Olson, K. Van Benthem, S. Elhadj, Dislocation mediated alignment during metal nanoparticle coalescence, *Acta Mater.* 120 (2016) 364–378.
- [89] F. Abdeljawad, D.S. Bolintineanu, A. Cook, H. Brown-Shaklee, C. DiAntonio, D. Kammler, A. Roach, Sintering processes in direct ink write additive manufacturing: a mesoscopic modeling approach, *Acta Mater.* 169 (2019) 60–75.
- [90] N. Beets, Y. Cui, D. Farkas, A. Misra, Mechanical response of a bicontinuous copper–molybdenum nano-composite: experiments and simulations, *Acta Mater.* 178 (2019) 79–89.
- [91] Y. Wang, M. Wang, X. Fang, F. Guo, H. Liu, R. Scattergood, C. Huang, Y. Zhu, Extra strengthening in a coarse/ultrafine grained laminate: role of gradient interfaces, *Int. J. Plast.* 123 (2019) 196–207.