

Supplementary Information for

Dispersion hardening using amorphous nanoparticles deployed via additive manufacturing

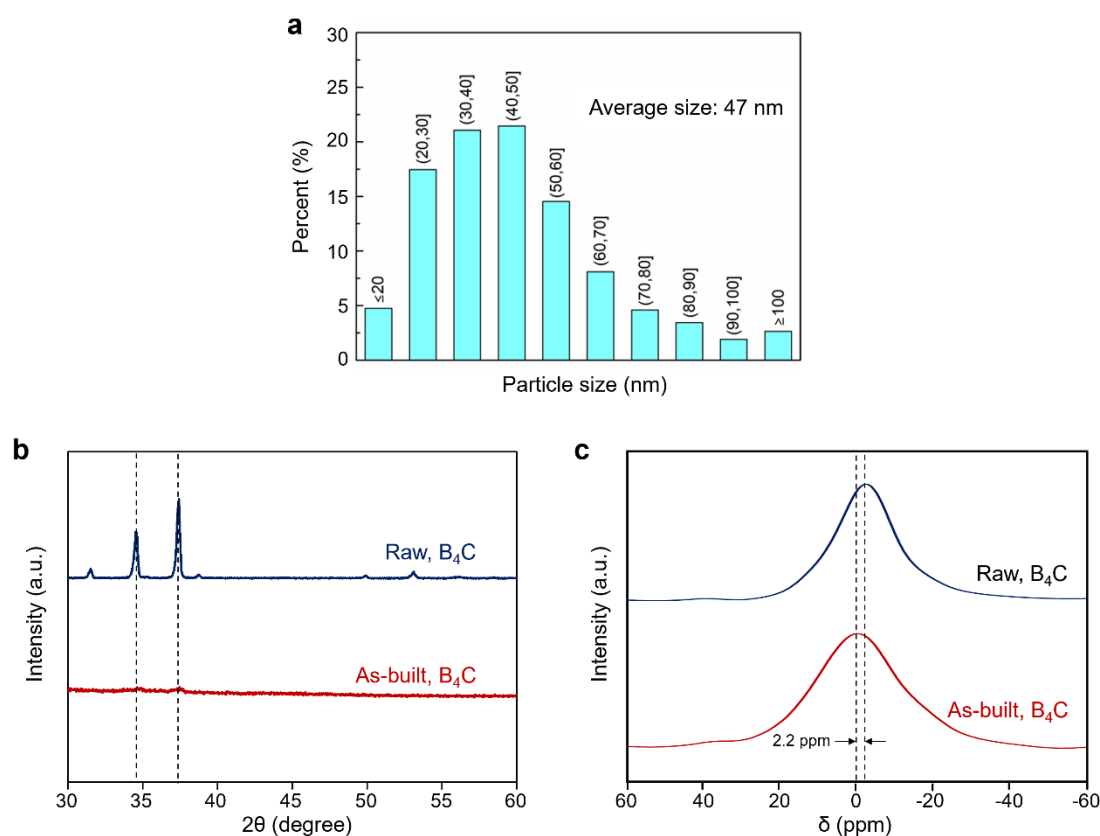
Ge Wang^{1#}, Yin Zhang^{2#}, Jian Liu^{3#}, Wen Chen³, Kang Wang¹, Bo Cui¹, Bingkun Zou¹, Qiubao Ouyang¹, Yanming Zhang⁴, Zhaoyang Hu⁴, Lu Wang⁴, Wentao Yan⁴, Shenbao Jin⁵, Jun Ding⁶, Y. Morris Wang⁷, Ting Zhu⁸, Zan Li^{1*}, Di Zhang^{1*} & Evan Ma^{6*}

¹State Key Laboratory of Metal Matrix Composites, Shanghai Jiao Tong University, Shanghai 200240, China. ²State Key Laboratory for Turbulence and Complex System, College of Engineering, Peking University, Beijing 100871, China. ³Department of Mechanical and Industrial Engineering, University of Massachusetts, Amherst, MA, USA. ⁴Department of Mechanical Engineering, National University of Singapore, Singapore 117575, Singapore. ⁵School of Material Science and Engineering, Hebei University of Technology, Tianjin 300401, China. ⁶Center for Alloy Innovation and Design, State Key Laboratory for Mechanical Behaviour of Materials, Xi'an Jiaotong University, Xi'an 710049, China. ⁷Department of Materials Science and Engineering, University of California, Los Angeles, CA 900095, USA. ⁸Woodruff School of Mechanical Engineering, Georgia Institute of Technology, Atlanta, Georgia 30332, USA.

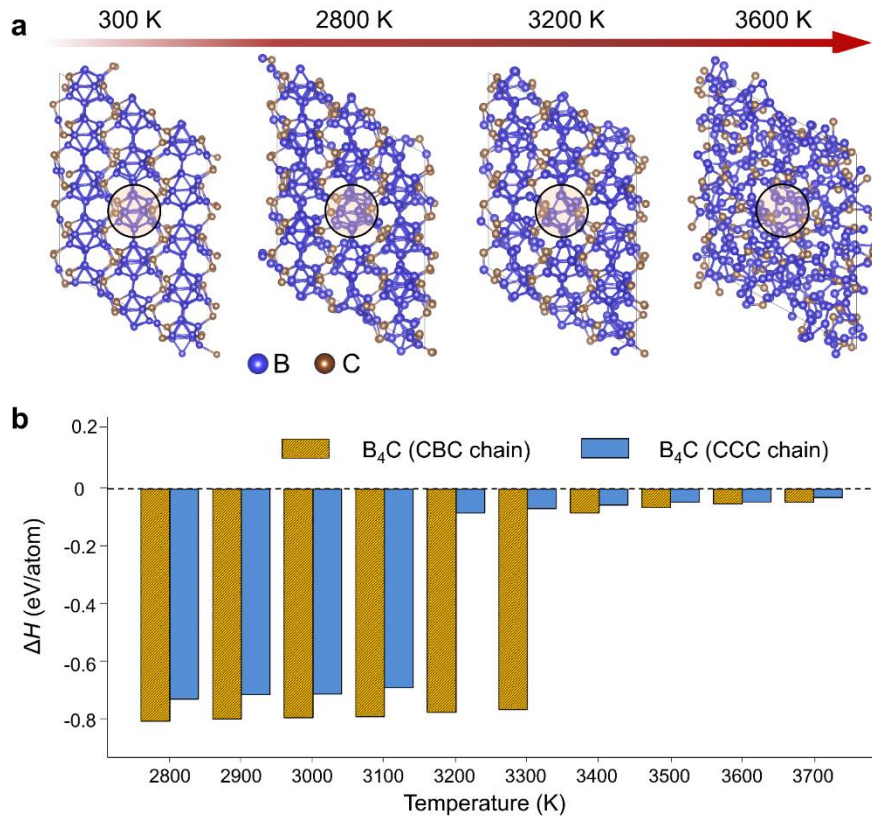
[#]These authors contributed equally.

*Emails: njulizan@sjtu.edu.cn; zhangdi@sjtu.edu.cn; maen@xjtu.edu.cn

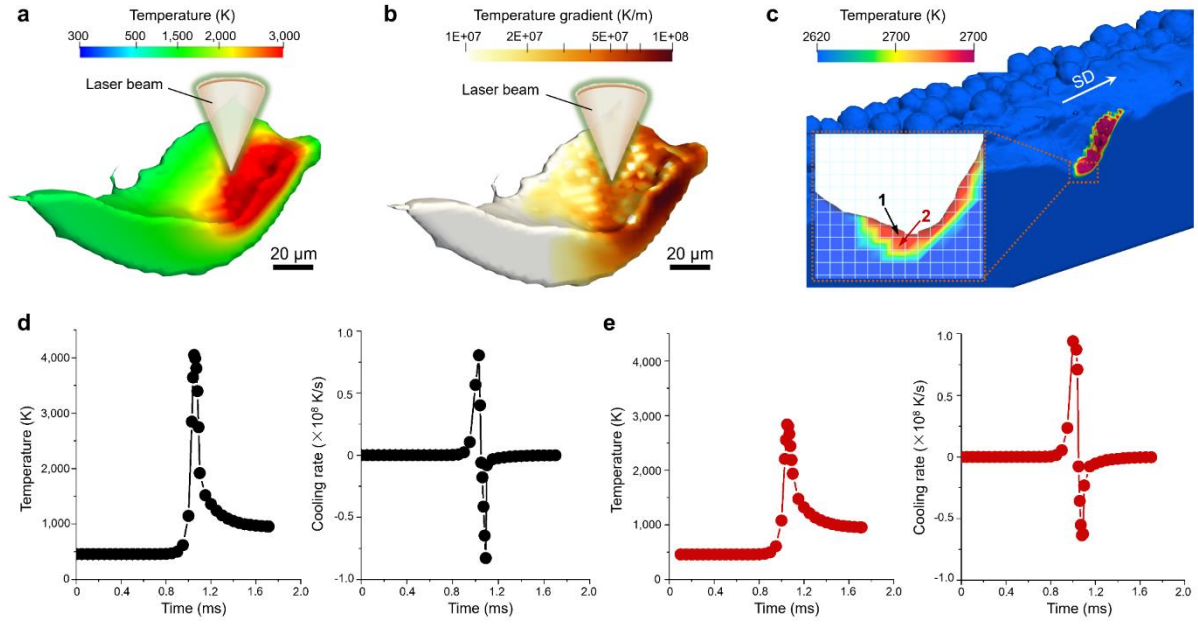
This Supplementary Information (SI) file contains twelve supplementary figures ([Figure S1-S12](#)), six supplementary notes ([Note S1-S6](#)), five supplementary tables ([Table S1-S5](#)), and three supplementary movies ([Movie S1-S3](#))



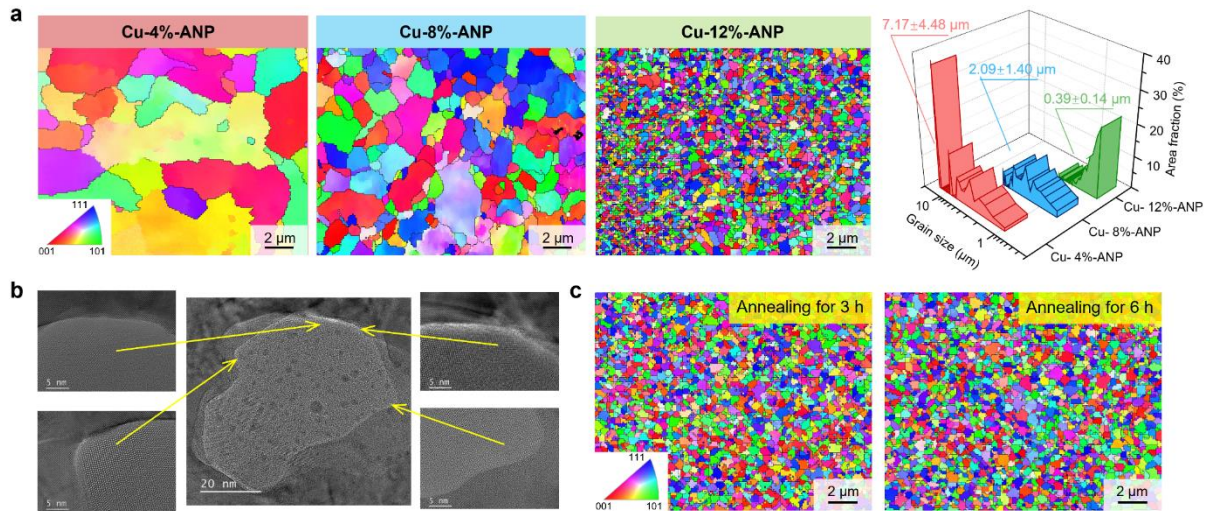
Supplementary Fig. 1 | Structural analyses of in-situ fragmented nanoparticles upon laser powder-bed-fusion (L-PBF). **a**, Statistical calculation of nanoparticle size distributions in the as-built composite, as obtained by bright-field transmission electron microscopy (BF-TEM), showing an average size of ~47 nm. **b**, X-ray diffraction (XRD) of B₄C in the raw and as-printed conditions. Note that the copper matrix was etched in FeCl₃ solution for better characterization of nanoparticles. **c**, The corresponding ¹¹B nuclear magnetic resonance (NMR) spectra, showing a peak shift of nanoparticles in the as-built state relative to the raw material, suggesting an amorphous structure¹.



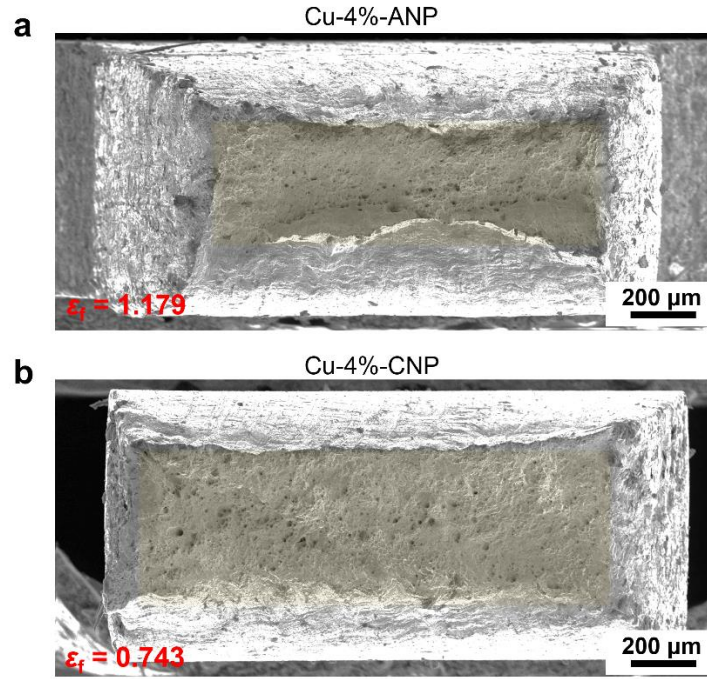
Supplementary Fig. 2 | Atomic structure of B₄C at different temperatures as estimated by *ab initio* molecular dynamics (AIMD) calculations. **a**, The structural evolution of boron carbide icosahedra and chains at different temperatures, ranging from room temperature (300 K) to temperatures above the melting point of boron carbide (~3,300 K). At high temperatures, initially intact carbide icosahedra (as represented by a shaded orange circle) start to disintegrate, but C and B atoms remain locally attached. **b**, The formation enthalpy for two typical boron carbide configurations (CBC chain and CCC chain), ΔH , at high temperatures ranging from 2,800 K to 3,700 K. Note that for both configurations, the formation enthalpy is negative throughout the whole probed temperature range. This suggests that B-C clusters are more stable with respect to the free-atom state in liquid states, facilitating the inconspicuous diffusion of B and C atoms in copper and the formation of amorphous B₄C nanoparticles upon the rapid solidification upon L-PBF.



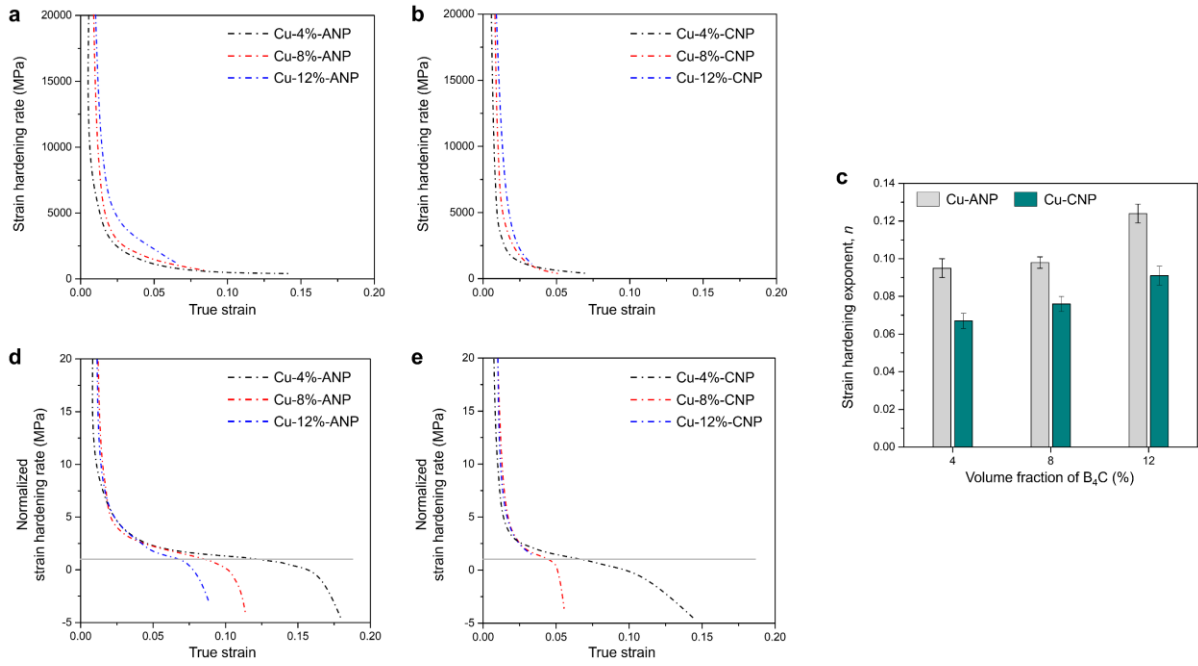
Supplementary Fig. 3 | Modelling of the temperature and temperature gradient of the melt pool during L-PBF of copper composites with a B₄C volume fraction of 12%. **a** and **b** show the temperature and temperature gradient maps of the melt pool, respectively. B₄C is mainly melted by direct laser irradiation and convective heat transfer with high-temperature fluid at the melt pool surface (red region shown by melt pool temperature). The in-situ fragmentation process of micron-sized B₄C mainly happens in this region. The subsequent track-by-track scanning does not induce further refinement of B₄C particles as the melt pool temperature is below the melting point of B₄C ($\sim 3,300$ K). Also, the temperature gradient of B₄C solidification is around 10^7 - 10^8 K/m, leading to the formation of strong Marangoni flow and facilitating the dispersion of nanoparticles. **c**, Three-dimensional temperature maps of powder beds, where two locations in the melt pool are selected for further quantitative analyses. SD stands for scan direction. **d** and **e** show temperature and cooling rate profiles at locations 1 and 2, respectively, as a function of time. The cooling rate of the melt upon solidification can reach 10^8 K/s.



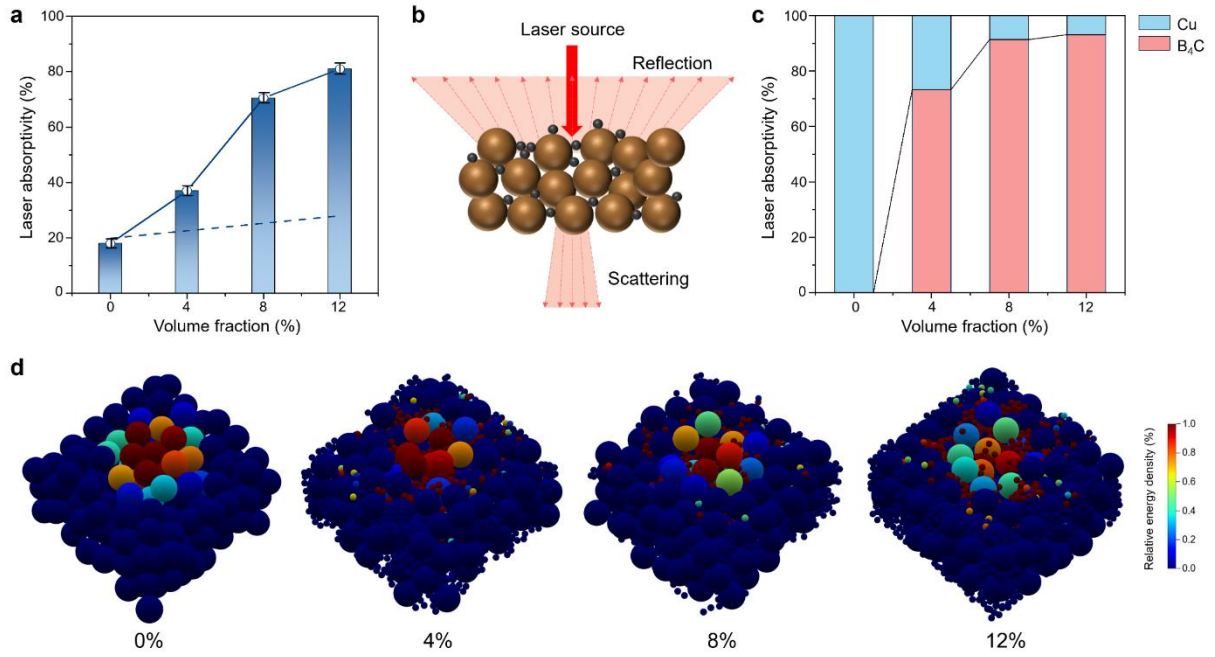
Supplementary Fig. 4 | Microstructure of L-PBF copper nanocomposites with different volume fractions of reinforcing particles and the microstructure after annealing. a, Representative electron backscatter diffraction (EBSD) images showing microstructures of L-PBF copper composites containing different B₄C volume fractions (4%, 8%, and 12%). The statistical calculation of grain sizes in different samples is also shown. **b,** HAADF-STEM images of L-PBF copper nanocomposites after annealing at 1,173K for 6 h. The amorphous nanoparticles generated upon AM are crystallized after annealing, as revealed by the enlarged view of different locations. **c,** Representative EBSD images of L-PBF copper nanocomposites before and after annealing (the volume fraction of B₄C is 12%). Grain sizes of copper in these two samples are similar (i.e., $\sim$ 400 nm).



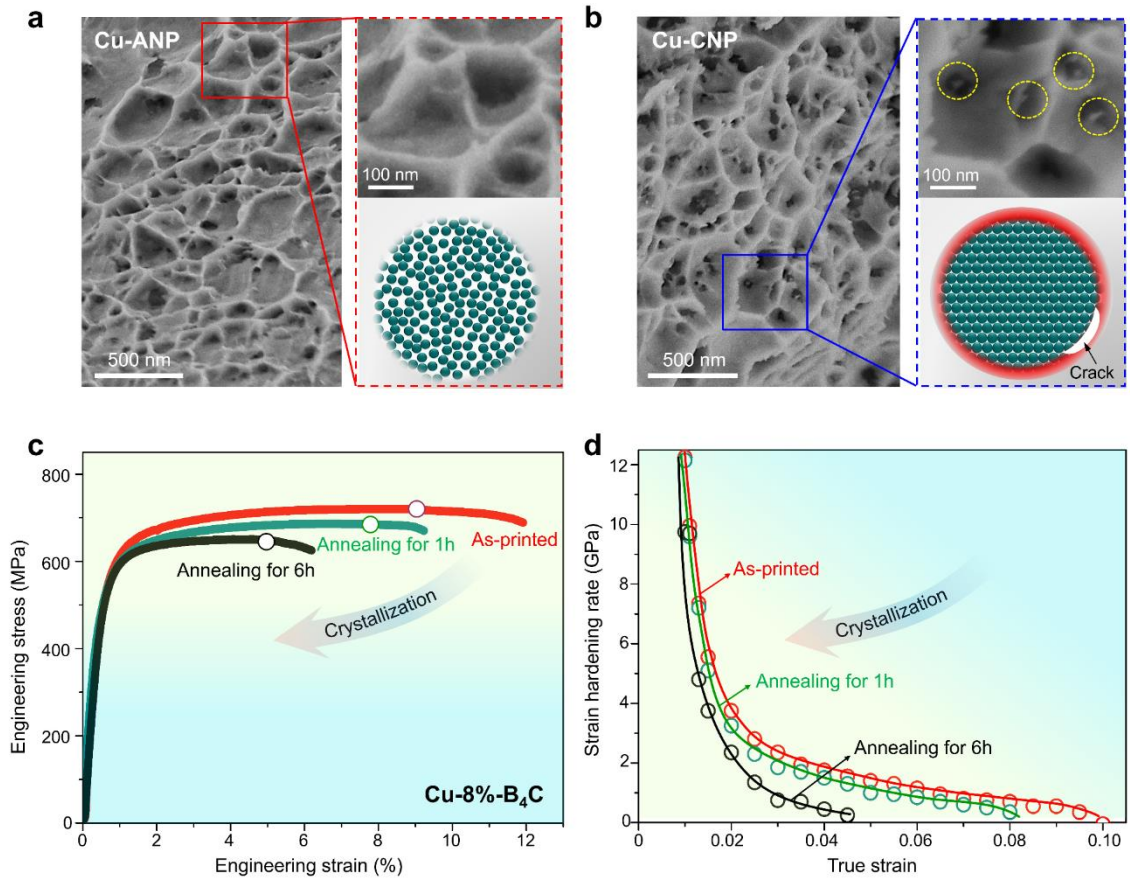
Supplementary Fig. 5 | Representative fracture surfaces of L-PBF nanocomposites after tensile testing. a, Cu-4%-ANP nanocomposite; b, Cu-4%-ANP nanocomposite. The reduction in area, defined as $q = (A_0 - A_f)/A_0$, where A_0 and A_f are the initial and fracture cross-sectional areas, respectively, can be estimated. The true fracture strain (ϵ_f) can then be estimated as $\epsilon_f = \ln(1/1-q)$. Note that ϵ_f of Cu-4%-ANP is significantly higher than that of Cu-4%-CNP.



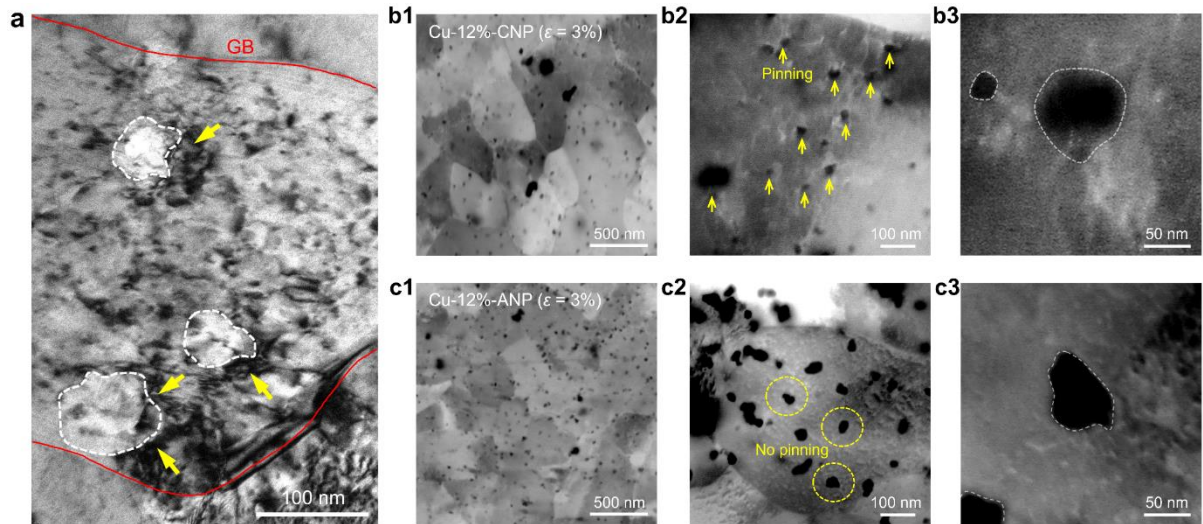
Supplementary Fig. 6 | Strain hardening behaviour analyses of L-PBF copper nanocomposites. **a-b**, Strain hardening rate vs. true strain plots of L-PBF Cu-ANP and Cu-CNP nanocomposites with different volume fractions of nanoparticles (4%, 8% and 12%), respectively, before deformation instability occurs (necking or sample fracture). **c**, The work-hardening exponent n of different L-PBF samples, as estimated from the Hollomon equation². The values are calculated for applied strain ranging from 1% to the uniform elongation of different samples. **d-e**, The normalized strain hardening rate–true strain curves of L-PBF Cu-ANP and Cu-CNP nanocomposites, respectively. The strain at which deformation instability (i.e., necking) occurs for different nanocomposites is indicated by the intersection of the gray line with each curve, based on the Considère criterion³.



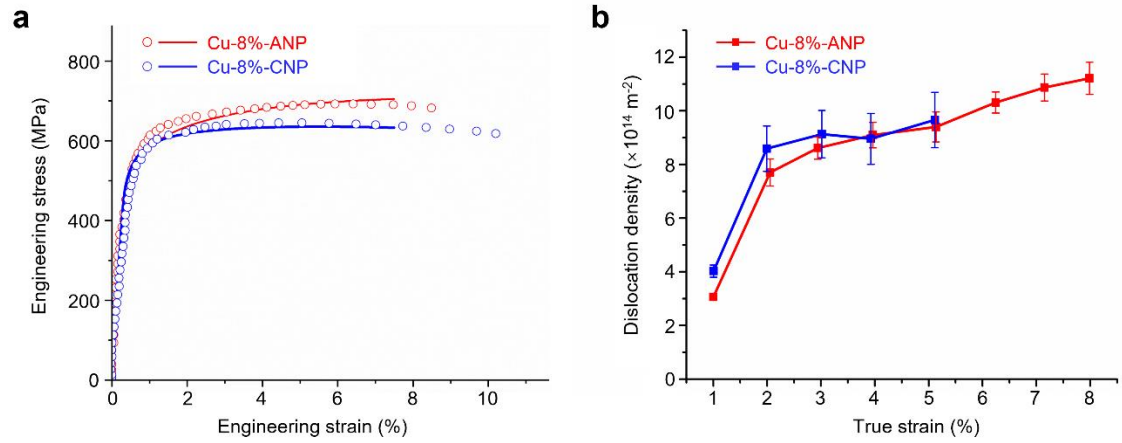
Supplementary Fig. 7 | Laser absorptivity analyses. **a**, Laser absorptivity of Cu-B₄C composite powders as a function of B₄C volume fraction at a laser wavelength of 1064 nm, as measured by a double-beam UV-Visible-NIR spectrophotometer. The dashed line represents the calculated laser absorptivity of composite powders using the rule-of-mixtures. Three measurements were conducted to obtain the statistical values. **b**, Schematic of laser-composite powder bed interactions upon full laser ray tracing modelling. **c**, The relative percentage of absorbed laser energy in Cu (coloured light blue) and B₄C powders (coloured pink), respectively. Note that the reinforcing particles account for the majority of the laser absorption process, even at a low volume fraction (e.g., 4%). **d**, The normalized surface energy density distribution of composite powders with different B₄C contents upon laser irradiation (wavelength: 1064 nm). Remarkably, B₄C particles exhibit the maximum surface energy density, and the surface energy density of Cu decreases as B₄C content increases due to the extensive and uniform laser absorption process.



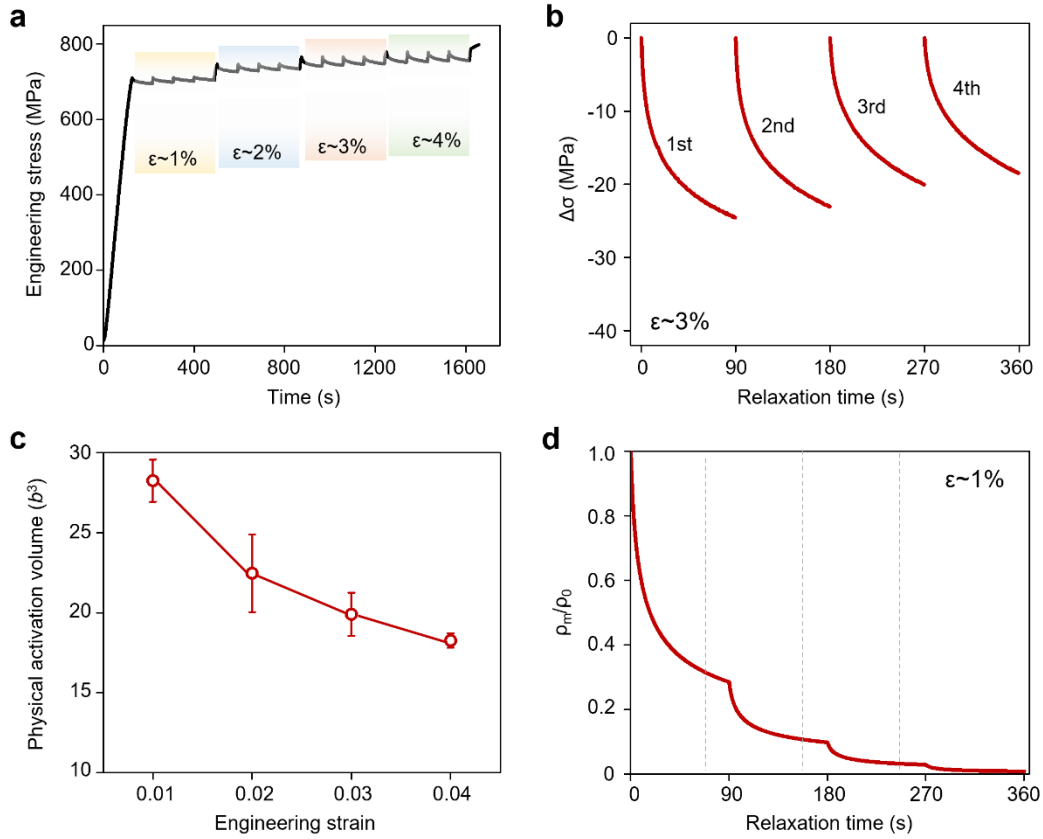
Supplementary Fig. 8 | Fractography analyses of L-PBF copper nanocomposites and tensile properties of composites after heat-treatment. **a** and **b** are SEM images of fracture surfaces of Cu-12%-ANP and Cu-12%-CNP, respectively. Clearly, Cu-12%-CNP shows numerous nanoparticles at the dimple bottom, suggesting that the crystallized nanoparticles tend to nucleate cracks upon deformation. **c** and **d** are tensile stress-strain curves and strain hardening rate plots of Cu-8%-ANP nanocomposites before and after annealing at 1,173K for different time. The points at which necking appears are indicated.



Supplementary Fig. 9 | TEM and electron channeling contrast imaging (ECCI) of different L-PBF copper nanocomposites after a tensile strain of 3%. a, Representative two-beam bright-field TEM image of the deformed Cu-12%-CNP nanocomposites (3% strain). Dislocations are found to accumulate near the CNPs, as indicated by yellow arrows. GB stands for grain boundary. **b1-b3,** Representative ECC images of deformed microstructures of Cu-12%-CNP. Dislocations are pinned by CNPs, as indicated by yellow arrows in **b2**. **b3** is an enlarged view of an ECC image, showing dislocation blockage and accumulation near nanoparticles. **c1-c3,** Representative ECC images of deformed microstructures of Cu-12%-ANP. Dislocations are distributed uniformly inside metal grains, with no discernable pinning at the ANPs.

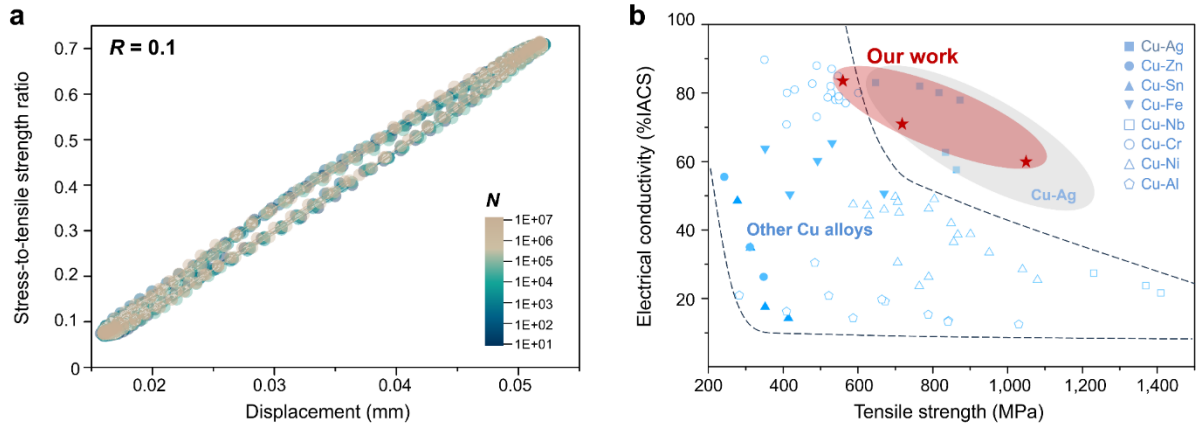


Supplementary Fig. 10 | In-situ synchrotron diffraction analysis of deformation mechanisms. **a**, Experimental and crystal plasticity finite-element (CPFE) modelling results of the macroscopic stress-strain curves of different L-PBF copper nanocomposites (8 vol.% B₄C). Experimental and simulation results are represented by symbols and solid lines, respectively. **b**, Dislocation density versus strain of copper in Cu-8%-ANP and Cu-8%-CNP samples, as estimated from the modified Williamson–Hall method. Error bars represent the standard deviation.



Supplementary Fig. 11 | Dislocation kinetics analyses of L-PBF copper nanocomposites.

a, Multiple stress-relaxation tests of L-PBF Cu-8%-ANP at various specific engineering strains (1%, 2%, 3%, and 4%). At each strain, four repeated stress-relaxation cycles with a time interval of 90 s were conducted. **b**, Typical stress-relaxation series of the L-PBF nanocomposite at a strain of 3%. **c**, Physical activation volume (V^*) evolution as a function of strain for Cu-8%-ANP. Statistical calculations of the activation volumes were obtained from the multiple stress-relaxation cycles. **d**, Evolution of the mobile dislocation density of Cu-8%-ANP as a function of relaxation time, at a representative strain of 1%.



Supplementary Fig. 12 | Fatigue and electrical properties of L-PBF copper nanocomposites. **a**, Cyclic hysteresis loops of the representative Cu-4%-ANP nanocomposite with different cycle numbers. The ratio of the minimum stress to the maximum stress experienced during a cycle, R , is 0.1. **b**, Electrical conductivity versus tensile strength of L-PBF Cu-ANP nanocomposites compared with conventional copper-based alloys. Note that this combination of Cu-ANP rivals Cu-Ag alloys, but the tensile ductility is much higher (high-strength Cu-Ag alloys usually demonstrate tensile ductility of less than 3 percent). Sources of the references are described in [Supplementary Note 6](#).

Supplementary Note 1 | A summary of mechanical properties, including ultimate tensile strength (σ_u), uniform elongation (ϵ_u) of the reported representative high-performance additive manufacturing (AM) metals. For better comparisons of different metallic materials, tensile strength is normalized by the Young's modulus of different metal matrix, which are 110 GPa, 115 GPa, 200 GPa, 70 GPa, and ~220 GPa for copper-, titanium-, steel-, aluminium-, and high-entropy alloy (HEA)-based materials, respectively.

		σ_u (MPa)	$\sigma_u / E (\times 10^{-3})$	ϵ_u (%)	Symbol
Cu-ANP (This work)		1033	9.4	7.1	
		721	6.6	9.0	
		557	5.1	13.3	
Ti alloys	Ti-Cu ⁴	1180	10.3	2.1	
		1073	9.3	5.5	
		860	7.5	10.4	
	Ti-O-Fe ⁵	1271	11.1	3.3	
		1205	10.5	3.0	
		1152	10.0	3.0	
		1054	9.2	6.0	
		892	7.8	5.0	
		547	4.8	8.6	
	Fe19Ni5Ti ⁶	1300	6.5	9.0	
	17-4 PH ⁷	1070	5.1	7.7	
		1172	5.9	4.8	
		1188	5.9	3.5	
		1319	6.6	4.8	
		1444	7.2	2.9	
		1358	6.8	6.5	
Steel	AF9628 ⁸	1320	6.6	6.6	
		1420	7.1	6.1	

	HY100 ⁹	1200	6.0	6.0	
	7075+Zr ¹⁰	411	5.9	5.3	
	AlSi10Mg+TiB ₂ ¹¹	537	7.7	7.1	
Al alloys		452	6.5	3.8	◀
	Al-Cu-Mg-Zr ¹²	408	5.8	6.2	
		399	5.7	7.9	
	Al _x CoCrFeNi ¹³	886	4.1	10.1	
		673	3.1	11.2	
HEA (fcc)	CoCrFeMnNi ¹⁴	651	3.0	27.9	◆
	FeCoCrNi ¹⁵	745	3.5	17.0	
	CoCrFeMnNi ¹⁶	665	3.1	24.0	
		604	2.8	16.0	
		480	4.4	13.6	
		453	4.1	11.0	
		341	3.1	15.5	
		295	2.7	33.2	
		275	2.5	27.0	
Cu alloys ¹⁷		280	2.5	23.4	■
		321	2.9	25.0	
		305	2.8	26.0	
		466	4.2	12.0	
		425	3.9	9.0	

Supplementary Note 2 | A summary of mechanical properties, including ultimate tensile strength (σ_b), total elongation (ϵ_t) of the reported precipitation/dispersion strengthened copper-based materials.

	σ_b (MPa)	ϵ_t (%)	Symbol
	1033	9.4	
Cu-ANP (This work)	721	11.9	★
	557	20.5	
	911	3.2	
Cu-CNP (This work)	650	6.2	★
	513	16.7	
	670	7	
Cu-2Ni-0.5Si- x Ti ¹⁸	710	6	
	630	6	
	1102	2.4	
Cu-Ni	1049	3.3	□
Cu- x Ni- y Si- z Co- u Cr- v Al ¹⁹	880	4.1	
	857	7	
	817	3	
Cu-1.01Ni-0.14Co-0.28Si-0.45Cr ²⁰	810	3.4	
	440	13	
	500	12	
Cu-0.5Cr- x Ag- y Zr ²¹	490	12	
	530	11	
Cu-Cr	570	11	■
	350	21	
Cu-0.3Cr-0.1Zr-0.05Mg ²²	410	6.4	
	478	10	

		530	5.4	
		549	6.1	
		612	9.7	
		546	10.5	
	Cu-0.7Cr-0.19Zr ²³	391	23.4	
		409	23.9	
		594	9	
	Cu-1Cr-0.2Zr-Mg ²⁴	601	8.1	
		629	6.5	
		947	3.8	
		900	5.3	
		831	5.7	
		1132	1.4	
Cu-Ti	Cu-3Ti-2Ni ²⁵	805	2.7	○
		1075	1.7	
		789	2.8	
		937	3.1	
		586	6.4	
		406	9.6	
	Cu-2.5Fe-xCr-ySi ²⁶	435	8.4	
Cu-Fe		415	4.9	●
	Cu-10Fe-xZr ²⁷	556	5.6	
		490	10.5	
		538	13.5	
Cu-Zr	Cu-0.5Zr-0.065B ²⁸	552	9	△
		551	11.8	

		568	7.7	
		575	13.4	
		523	17.3	
		609	12	
		679	4.7	
		734	2.32	
	Cu- x Al ₂ O ₃ ²⁹	526	13.8	
Cu-Al ₂ O ₃		606	10.5	▲
		693	2.45	
		753	3.08	
		674	9.3	
	Cu-2.03Al ₂ O ₃ - x Ti ³⁰	655	5.7	
		1148	10.4	
		1110	12.7	
	Cu-2Be-0.3Ni-0.2Co ³¹	756	21.6	
Cu-Be		705.8	15.6	▽
		698.6	18	
		698.1	21	
	Cu-2.7Be ³²	679.3	21.5	

Supplementary Note 3 | Additional discussion on the effect of copper nanocrystals on the deformation mechanisms of L-PBF composites.

We consider the impact of copper nanocrystals on the overall deformation mechanisms of the nanocomposites to be marginal for the following two reasons.

(1) **Limited volume fraction:** The volume fraction of the copper nanocrystals within the amorphous B₄C nanoparticle is relatively small, approximately 13%, as estimated from the STEM images. As revealed by the simulation results, the local structural configurations within the ANPs evolve during straining, with bond breaking and reforming (with neighbouring atoms) helping to relieve stress concentrations at metal–ANP interfaces. These deformation characteristics are intrinsic to the amorphous structure and are not significantly altered by the incorporation of a small fraction of copper nanocrystals.

(2) **Intrinsic suppression of shear banding in amorphous nanoparticles.** It has been reported that collective atomic rearrangements, known as shear transformation zones (STZs), are the fundamental plastic flow units in amorphous materials. According to the Griffith criterion for crack propagation, there is a critical sample size of approximately 100 nm, above which the spatial percolation of STZs leads to the development of localized shear bands and strain softening in amorphous materials³³. In contrast, for amorphous samples smaller than ~100 nm (such as the quasi-amorphous B₄C nanoparticles obtained in this work), shear banding instability is intrinsically suppressed, resulting in homogeneous plastic flow.

Supplementary Note 4 | A summary of fatigue properties, represented by fatigue life and fatigue strength-to-tensile strength ratio Cu-4%-ANP nanocomposite, along with references for the reported state-of-the-art copper materials, including nanograined, gradient and nanotwinned microstructures. Fatigue properties of the well-annealed coarse-grained copper is also shown. All of comparison samples possess a stress ratio R of 0.1.

	Fatigue life (cycles)	fatigue-to-tensile strength ratio	Symbol
Cu-4%-ANP (This work)	1.2×10^5	0.95	★
	1.3×10^5	0.90	
	3.7×10^5	0.85	
	6.3×10^5	0.80	
	3.2×10^6	0.75	
	$>1 \times 10^7$	0.70	
	$>1 \times 10^8$	0.60	
Cu-12%-ANP (This work)	7.2×10^4	0.90	★
	4.0×10^5	0.80	
	3.8×10^6	0.70	
	$>2.4 \times 10^7$	0.65	
	$>6.7 \times 10^7$	0.60	
NT-Cu-Al ^a ³⁴	4.8×10^3	0.78	□
	9.1×10^3	0.72	
	1.2×10^4	0.72	
	4.0×10^4	0.57	
	2.2×10^5	0.57	
	2.2×10^6	0.45	
NT-Cu-Al ^b ³⁴	2.6×10^3	0.78	■
	3.8×10^3	0.78	
	6.3×10^4	0.45	
	2.4×10^5	0.45	
NT-Cu ^a ³⁵	3.3×10^3	0.64	●
	1.2×10^4	0.53	
	5.8×10^4	0.39	
	1.1×10^6	0.27	
	1.4×10^6	0.23	
	6.5×10^6	0.24	
NT-Cu ^b ³⁶	5.7×10^4	0.83	○
	1.2×10^5	0.72	
	5.2×10^5	0.66	
	8.0×10^5	0.61	
	9.4×10^5	0.61	
	1.4×10^6	0.59	
	1.7×10^6	0.67	
	4.3×10^6	0.54	

	6.1×10 ⁶	0.58	
	7.0×10 ⁶	0.54	
	1.0×10 ⁷	0.50	
	1.3×10 ⁷	0.57	
	3.5×10 ⁷	0.54	
	7.3×10 ⁷	0.50	
GNG-Cu ³⁷	1.4×10 ⁴	0.90	
	3.2×10 ⁵	0.76	
	4.8×10 ⁵	0.76	□
	1.3×10 ⁶	0.70	
	4.1×10 ⁶	0.70	
	9.8×10 ⁶	0.63	
CG-Cu ³⁶	1.4×10 ⁵	0.82	
	2.6×10 ⁵	0.74	
	9.0×10 ⁵	0.65	
	4.4×10 ⁶	0.61	
	5.2×10 ⁶	0.58	□
	6.9×10 ⁶	0.56	
	7.8×10 ⁶	0.56	
	1.0×10 ⁷	0.48	
	1.8×10 ⁷	0.43	
	2.2×10 ⁷	0.48	
UFG-Cu ³⁸	4.2×10 ³	0.88	
	7.9×10 ³	0.88	
	1.6×10 ⁴	0.84	
	5.4×10 ⁴	0.62	
	5.4×10 ⁴	0.59	
	7.0×10 ⁴	0.54	▲
	5.8×10 ⁵	0.46	
	5.8×10 ⁵	0.42	
	7.0×10 ⁵	0.44	
	1.9×10 ⁶	0.39	
Cu-Cr composite ³⁹	1.9×10 ⁵	0.80	
	3.9×10 ⁵	0.74	
	4.7×10 ⁵	0.69	
	1.2×10 ⁶	0.63	▲
	4.0×10 ⁶	0.56	
	9.6×10 ⁶	0.52	
GS-Cu ⁴⁰	3.2×10 ³	0.85	
	6.9×10 ³	0.81	
	3.1×10 ⁴	0.77	△
	5.8×10 ⁴	0.73	

	1.8×10^5	0.69
	2.7×10^5	0.51
	5.5×10^5	0.39
	6.9×10^5	0.47
	4.0×10^6	0.31

Supplementary Note 5 | A summary of strength (hardness) retention of L-PBF Cu-ANP after annealing at different temperatures for 1h, as compared with the conventional copper-based alloys.

		Temperature (K)	Strength retention	Symbol
Cu- 4, 8, 12vol.%ANP (This work)		623	1; 1; 1	★ ★ ★
		673	1; 1; 1	
		723	1; 1; 1	
		773	0.98; 1; 1	
		823	0.98; 1; 1	
		873	0.96; 0.99; 0.99	
		923	0.95; 0.99; 0.99	
		973	0.94; 0.98; 0.98	
		1023	0.93; 0.97; 0.98	
		1073	0.92; 0.93; 0.97	
		1123	0.91; 0.91; 0.96	
		1173	0.90; 0.90; 0.94	
Cu-Ag	Cu-0.1wt.%Ag ⁴¹	673	0.99	■
		723	0.71	
		773	0.57	
	Cu-24wt.%Ag ⁴²	568	0.96	
		683	0.72	
		831	0.44	
Cu-Zn	Cu-30at.%Zn ⁴³	573	0.62	●
		673	0.55	
		773	0.47	
		873	0.40	
	Cu-5wt.%Zn ⁴⁴	673	0.63	
Cu-Sn	Cu-0.04wt.%Sn ⁴¹	773	0.58	▲
		823	0.57	
	Cu-0.032wt.%Sn-0.023wt.%Te ⁴¹	673	0.96	
		723	0.97	
		773	0.69	
		823	0.59	
Cu-Fe	Cu-2.0wt.%Fe-0.7%wt.%Cr ⁴⁵	673	0.92	▼
		723	0.90	
		773	0.92	
		823	0.88	
		873	0.84	
		923	0.8	
		973	0.75	
		1023	0.6	
		1073	0.49	

Cu-Nb	Cu-0.7wt.%Fe- 0.15wt.%P ⁴⁶	673	0.92	□
		723	0.92	
		773	0.85	
	Cu-2.35wt.%Fe- 0.03wt.%P- 0.1wt.%Cr ⁴⁷	673	0.95	
		723	0.92	
		773	0.89	
		823	0.84	
		873	0.81	
	Cu-10wt.%Nb ⁴⁸	673	0.96	
		873	0.92	
		973	0.89	
	Cu-5at.%Nb ⁴⁹	1073	0.89	
		1173	0.79	
	Cu-7at.%Nb ⁴⁹	873	0.98	
		973	0.93	
		1073	0.88	
		1173	0.70	
	Cu-10at.%Nb ⁴⁶	973	0.97	
		1073	0.84	
		1173	0.79	
Cu-Cr	Cu- 0.535wt.%Cr- 0.1wt.%Mg ⁵⁰	673	0.99	○
		773	0.94	
		823	0.85	
		873	0.74	
		923	0.63	
		973	0.45	
	Cu-1.41wt%Cr- 0.05wt.%Y ⁵¹	673	0.99	
		723	0.98	
		773	0.96	
		823	0.81	
		873	0.66	
		923	0.56	
	Cu-0.67wt.%Cr- 0.23wt.%Sn ⁵²	693	0.95	
		733	0.89	
		773	0.80	
		813	0.63	
		853	0.52	
Cu-Ni	Cu-0.78wt.%Ni- 0.32wt.%Mn- 0.21wt%.P ⁵³	673	0.97	△
		723	0.90	
		773	0.79	
		823	0.63	
		873	0.46	

Cu-Al	Cu-8at.%Al ⁵⁴	665	0.45	◻
		760	0.42	
	Cu-8at.%Al ⁵⁵	665	0.60	

Supplementary Note 6 | A summary of electrical conductivity versus tensile strength of L-PBF Cu-ANP nanocomposites as compared with conventional copper-based alloys.

		Tensile strength (MPa)	Electrical conductivity (%I _{ACS})	Symbol
Cu- ANP (This work)	4 vol.%	557	83.0	★
	8 vol.%	721	70.2	
	12 vol.%	1033	61.1	
Cu-Ag ⁵⁶⁻⁵⁸	Cu-3wt.%Ag	647	83.0	■
		765	82.0	
	Cu-5wt.%Ag	817	80.1	
		873	77.9	
	Cu-6wt.%Ag	834	62.7	
		863	57.6	
Cu-Zn ⁵⁹	Cu-5wt.%Zn	243	55.5	●
	Cu-15wt.%Zn	311	35.0	
	Cu-30wt.%Zn	348	26.3	
Cu-Sn ⁵⁹	Cu-1wt.%Sn	278	48.5	▲
	Cu-2wt.%Sn	313	34.8	
	Cu-5wt.%Sn	352	17.5	
	Cu-8wt.%Sn	414	14.2	
Cu-Fe ^{59,60}	Cu-Fe-P	352	63.8	▼
	Cu-Fe-Co-P-Sn	418	50.4	
	Cu-Fe-P	492	60.2	
	Cu-Fe-P-Zn	531	65.4	
	Cu-Fe-P-Co-Sn	670	50.7	
Cu-Nb ⁵⁹	Cu-5at.%Nb	1235	27.4	□
	Cu-7at.%Nb	1370	23.7	
	Cu-10at.%Nb	1410	21.6	
Cu-Cr ⁶¹	Review	540	78	○
		567	77	
		529	80	
		550	78	
		549	79	
		431	81	
		527	82	
		490	88	
		530	87	
		409	70.8	
		410	80	
		601	80	
		490	73	
		519.3	78.7	

		350	89.7	
		478	82.7	
		546	85	
		625	47.1	
		587	47.5	
		700	49.7	
		706	48.2	
		670	45.9	
		710	45.1	
	Review	630	44.2	
		804	49	
		1080	25.4	
Cu-Ni ^{61,62}		1040	28.5	△
		951	33.4	
		901	38.8	
		867	38.6	
		849	42	
	Cu-10Ni-2Si	789	23.6	
	Cu-10Ni-2Si	856	36.5	
	Cu-10Ni-2Si	789	46.2	
	Cu-10Ni-2Si-2Ti	764	23.6	
	Cu-10Ni-2Si-2Ti	706	30.5	
	Cu-4at.%Al	485	30.4	
	Cu-8at.%Al	673	19.1	
	Cu-12at.%Al	788	15.3	
	Cu-15at.%Al	842	13.5	
	Cu-15at.%Al-3at.%Mg	1030	12.4	
Cu-Al ^{43,63}	Cu-3.1wt.%Al	283	20.9	◇
	Cu-3.1wt.%Al	522	20.8	
	Cu-3.1wt.%Al	664	19.7	
	Cu-6.8wt.%Al	409	16.2	
	Cu-6.8wt.%Al	587	14.3	
	Cu-6.8wt.%Al	841	13.2	

Supplementary Table 1 | Optimized laser processing parameters for Cu-ANP nanocomposites with different volume fraction of B₄C (4, 8 and 12%).

Samples	Hatch spacing (μm)	Layer thickness (μm)	Laser powder (W)	Scan speed (mm/s)
Cu-4%-ANP	40	20	165	1000
Cu-8%-ANP	40	20	158	1000
Cu-12%-ANP	40	20	130	900

Supplementary Table 2 | Thermophysical properties of Cu in the simulation^{64,65}.

Properties	Value
Solidus temperature (T_s)	1357.45 K
Liquidus temperature (T_l)	1357.85 K
Specific latent heat of melting (L_m)	2.087×10^5 J/kg
Specific latent heat of evaporation (L_{vap})	4.727×10^6 J/kg
Saturated vapor pressure (P_0) at $T=2835$ K	1.013×10^5 Pa
Emissivity factor (ε)	0.4
Surface tension coefficient (σ_0) at T_l	1.33 N/m
Temperature sensitivity of surface tension ($\frac{d\sigma}{dT}$)	2.3×10^{-4} N/(m·K)
Heat transfer coefficient (h)	40 W/(K·m ²)

Supplementary Table 3 | Temperature-dependent thermophysical properties of copper used for simulation⁶⁵.

Properties	Temperature (K)	Value
Specific heat, c_p (J/(K·kg))	373.15	397
	473.15	408
	873.15	441
	1273.15	464
	1357.45	469
	1357.85	495
	1473.15	495
Density at liquidus state, ρ_l kg/m ³	298.15	8930
	1357.45	8295
	1357.85	7960
	1473.15	7873
	1573.15	7797
Thermal conductivity, k W/(K·m)	298.15	400
	573.15	382
	773.15	370
	1357.45	330
	1357.85	146
	1373.15	147
	1473.15	150
	1573.15	153
Viscosity, μ (Pa·s)	1357.85	0.00437
	1473.15	0.00427
	1573.15	0.00328
	1673.15	0.00293
	1773.15	0.00266

Supplementary Table 4 | Thermophysical properties of B₄C in the simulation^{64,66}.

Properties	Value
Density (ρ_s)	2520 kg/m ³
Specific heat (c_s)	1000 J/(K·kg)
Thermal conductivity (k)	40 W/(K·m)
Solidus temperature (T_s)	2645.15 K
Liquidus temperature (T_l)	2780.15 K
Specific latent heat of melting (L_m)	1.650×10^6 J/kg
Specific latent heat of evaporation (L_{vap})	4.966×10^7 J/kg
Saturated vapor pressure (P_0) at $T=3773$ K	1.013×10^5 Pa
Interface energy of B ₄ C – Gas	1.06 N/m
Interface energy of B ₄ C – Liquid Cu	0.02 N/m

Supplementary Table 5 | Parameters for the CPFЕ constitutive model.

System	C_{11}^{Cu} (GPa)	C_{12}^{Cu} (GPa)	C_{44}^{Cu} (GPa)	E_{np} (GPa)	ν_{np}	$\dot{\gamma}_p^0$ (s ⁻¹)
Cu-ANP	168.4	121.4	75.4	700	0.18	0.001
Cu-CNP	168.4	121.4	75.4	450	0.18	0.001
	s_0 (MPa)	h_0 (MPa)	s_{sat} (MPa)	σ_y^0 (MPa)	h_{NP} (MPa)	m
Cu-ANP	170	100	180	3150	25	0.001
Cu-CNP	180	10	190	1900	25	0.001

References

1. Pallier, C. et al. Structure of an amorphous boron carbide film: An experimental and computational approach. *Chem. Mater.* **25**, 2618-2629 (2013).
2. Hollomon, J. H. Tensile deformation. *Transactions of the Metallurgical Society of AIME*, **162**, 268-290 (1945).
3. Wang, Y., Chen, M., Zhou, F. & Ma, E. High tensile ductility in a nanostructured metal. *Nature* **419**, 912-915 (2002).
4. Zhang, D. et al. Additive manufacturing of ultrafine-grained high-strength titanium alloys. *Nature* **576**, 91-95 (2019).
5. Song, T. et al. Strong and ductile titanium–oxygen–iron alloys by additive manufacturing. *Nature* **618**, 63-68 (2023).
6. Kürsteiner, P. et al. High-strength Damascus steel by additive manufacturing. *Nature* **582**, 515-519 (2020).
7. LeBrun, T., Nakamoto, T., Horikawa, K. & Kobayashi, H. Effect of retained austenite on subsequent thermal processing and resultant mechanical properties of selective laser melted 17-4 PH stainless steel. *Mater. Des.* **81**, 44-53 (2015).
8. Seede, R. et al. An ultra-high strength martensitic steel fabricated using selective laser melting additive manufacturing: Densification, microstructure, and mechanical properties. *Acta Mater.* **186**, 199-214 (2020).
9. Dilip, J. J. S., Ram, G. D. J., Starr, T. L. & Stucker, B. Selective laser melting of HY100 steel: Process parameters, microstructure and mechanical properties. *Addit. Manuf.* **13**, 49-60 (2017).
10. Martin, J. H. et al. 3D printing of high-strength aluminium alloys. *Nature* **549**, 365-369 (2017).
11. Dan, C. et al. Achieving ultrahigh fatigue resistance in AlSi10Mg alloy by additive manufacturing. *Nat. Mater.* **22**, 1182-1188 (2023).
12. Zhang, H. et al. Effect of Zirconium addition on crack, microstructure and mechanical behavior of selective laser melted Al-Cu-Mg alloy. *Scr. Mater.* **134**, 6-10 (2017).
13. Sun, Z. et al. Reducing hot tearing by grain boundary segregation engineering in additive manufacturing: example of an Al_xCoCrFeNi high-entropy alloy. *Acta Mater.* **204**, 116505 (2021).
14. Melia, M. A. et al. Mechanical and corrosion properties of additively manufactured CoCrFeMnNi high entropy alloy. *Addit. Manuf.* **29**, 100833 (2019).
15. Brif, Y., Thomas, M. & Todd, I. The use of high-entropy alloys in additive manufacturing. *Scr. Mater.* **99**, 93-96 (2015).
16. Liu, Y. et al. Microstructure and mechanical behavior of additively manufactured CoCrFeMnNi high-entropy alloys: Laser directed energy deposition versus powder bed fusion. *Acta Mater.* **250**, 118884 (2023).
17. Tang, X. et al. The current state of CuCrZr and CuCrNb alloys manufactured by additive manufacturing: A review. *Mater. Des.* **224**, 111419 (2022).
18. Watanabe, C., Takeshita, S. & Monzen, R. Effects of small addition of Ti on strength and microstructure of a Cu-Ni-Si alloy. *Metall. Mater. Trans. A* **46**, 2469-2475 (2015).
19. Ban, Y. et al. Properties and precipitates of the high strength and electrical conductivity Cu-Ni-Co-Si-Cr alloy. *J. Mater. Sci. Technol.* **93**, 1-6 (2021).
20. Zhao, Z. et al. Microstructure and properties of a Cu-Ni-Si-Co-Cr alloy with high strength and high conductivity. *Mater. Sci. Eng. A* **759**, 396-403 (2019).
21. Watanabe, C., Monzen, R. & Tazaki, K. Mechanical properties of Cu-Cr system alloys with and

- without Zr and Ag. *J. Mater. Sci.* **43**, 813-819 (2008).
22. Huang, A. H. et al. Optimizing the strength, ductility and electrical conductivity of a Cu-Cr-Zr alloy by rotary swaging and aging treatment. *Mater. Sci. Eng. A* **746**, 211-216 (2019).
 23. Hu, J. et al. Optimizing strength and electrical conductivity of Cu-Cr-Zr alloy by two-stage aging treatment. *Mater. Lett.* **315**, 131937 (2022).
 24. Li, L. et al. Microstructure and properties of Cu-Cr-Zr (Mg) alloys subjected to cryorolling and aging treatment. *J. Alloys. Compd.* **938**, 168656 (2023).
 25. Li, R., Bi, Y., Li, C., Wang, H. & Cao, G. Precipitation behavior and properties change of Cu-3Ti-2Ni alloy during aging process. *Mater. Sci. Eng. A* **901**, 146510 (2024).
 26. Yu, X. et al. Effect of dual-scale FeCrSi phase on the strength and conductivity of Cu-2.5Fe-0.3Cr-0.2Si alloy. *Mater. Sci. Eng. A* **861**, 144340 (2022).
 27. Ren, L. et al. Grain refinement and mechanical properties enhancement of Cu-10 wt%Fe alloys via Zr addition. *Mater. Sci. Eng. A* **846**, 143309 (2022).
 28. Ye, Y. et al. Enhanced strength and electrical conductivity of Cu-Zr-B alloy by double deformation-aging process. *J. Alloys. Compd.* **615**, 249-254 (2014).
 29. Chen, F. et al. Achieving a better combination of strength and electrical conductivity of Cu-Al₂O₃ nanocomposites by accumulative roll-bonding to ultrahigh cycles. *Mater. Charact.* **178**, 111205 (2021).
 30. Zhang, X. et al. Enhanced tensile properties in a Cu-Al₂O₃ alloy via trace Ti addition. *J. Alloys. Compd.* **862**, 158687 (2021).
 31. Tang, Y., Kang, Y., Yue, L. & Jiao, X. The effect of aging process on the microstructure and mechanical properties of a Cu-Be-Co-Ni alloy. *Mater. Des.* **85**, 332-341 (2015).
 32. Liu, Y. et al. Microstructure and mechanical evolution of Cu-2.7Be sheets via annealing. *Metals* **10**, 241 (2020).
 33. Jang, D. & Greer, J. R. Transition from a strong-yet-brittle to a strong-and-ductile state by size reduction of metallic glasses. *Nat. Mater.* **9**, 215-219 (2010).
 34. Heckman, N. M., Berwind, M. F., Eberl, C. & Hodge, A. M. Microstructural deformation in fatigued nanotwinned copper alloys. *Acta Mater.* **144**, 138-144 (2018).
 35. Shute, C. J. et al. Detwinning, damage and crack initiation during cyclic loading of Cu samples containing aligned nanotwins. *Acta Mater.* **59**, 4569-4577 (2011).
 36. Pan, Q. S., Lu, Q. H. & Lu, L. Fatigue behavior of columnar-grained Cu with preferentially oriented nanoscale twins. *Acta Mater.* **61**, 1383-1393 (2013).
 37. Jing, L. J. Fatigue property and cyclic deformation mechanism of gradient nanograined Cu. (University of Science and Technology of China, 2020).
 38. Zhang, K. & Weertman, J. Microstructural changes produced by fatigue in high-purity copper severely deformed by cryogenic rolling. *Metall. Mater. Trans. A* **40**, 2255-2263 (2009).
 39. Masuda, C. & Tanaka, Y. Fatigue properties of Cu-Cr in situ composite. *Int. J. Fatigue* **28**, 1426-1434 (2006).
 40. Murata, N., Tamakawa, K., Suzuki, K. & Miura, H. Fatigue strength of electroplated copper thin films under uni-axial stress. *J. Solid Mech. Mater.* **3**, 498-506 (2009).
 41. Kim, K. T., Jung, W. J. & Choi, C. S. Influence of alloying elements on thermal conductivity and high temperature strength of copper based alloys. *Mater. Sci. & Technol.* **17**, 455-458 (2001).
 42. Han, K. et al. The fabrication, properties and microstructure of Cu-Ag and Cu-Nb composite conductors. *Mater. Sci. & Eng. A* **267**, 99-114 (1999).

43. Maki, K., Ito, Y., Matsunaga, H., & Mori, H. Solid-solution copper alloys with high strength and high electrical conductivity. *Scr. Mater.* **68**, 777-780 (2013).
44. Sarma, V. S., Sivaprasad, K., Sturm, D. & Heilmaier, M. Microstructure and mechanical properties of ultra fine grained Cu-Zn and Cu-Al alloys produced by cryorolling and annealing. *Mater. Sci. & Eng. A* **489**, 253-258 (2008).
45. Fernee, H., Nairn, J., Atrens, A. Cold worked Cu-Fe-Cr alloys. *J. Mater. Sci.* **36**, 5497-5510 (2001).
46. Dong, Q. et al. Effect of thermomechanical processing on the microstructure and properties of a Cu-Fe-P alloy. *J. Mater. Eng. and Perform.* **24**, 1531-1539 (2015).
47. Guo, F. A. et al. Study of rare earth elements on the physical and mechanical properties of a Cu-Fe-P-Cr alloy. *Mater. Sci. and Eng. B* **147**, 1-6 (2008).
48. Lei, R., Wang, M., Xu, S., Wang, H. & Chen, G. Microstructure, hardness evolution, and thermal stability mechanism of mechanical alloyed Cu-Nb alloy during heat treatment. *Metals* **6**, 194 (2016).
49. Botcharova, E., Freudenberger, J., Schultz, L. Mechanical and electrical properties of mechanically alloyed nanocrystalline Cu-Nb alloys. *Acta Mater.* **54**, 3333-3341 (2006).
50. Sun, Y. et al. Effects of Mg addition on the microstructure and softening resistance of Cu-Cr alloys. *Mater. Sci. Eng. A* **776**, 139009 (2020).
51. Wang, Y., Qu, J., Wang, X., Jie, J. & Li, T. Effects of Y addition on the microstructure, properties and softening resistance of Cu-Cr alloy. *J. Alloys Compd.* **902**, 163816 (2022).
52. Li, J. et al. Effect of Cr and Sn additions on microstructure, mechanical-electrical properties and softening resistance of Cu-Cr-Sn alloy. *Mater. Sci. Eng. A* **802**, 140628 (2021).
53. Huang, J., Xiao, X., Xiong, S., Wan, J. & Guo, C. Effect of Ni and Mn contents on the microstructure and properties of Cu-Ni-Mn-P alloy. *J. Alloys Compd.* **901**, 163636 (2022).
54. An, X. H., Qu, S., Wu, S. D. & Zhang, Z. F. Effects of stacking fault energy on the thermal stability and mechanical properties of nanostructured Cu-Al alloys during thermal annealing. *J. Mater. Res.* **26**, 407-415 (2011).
55. An, X. H. et al. Enhanced strength-ductility synergy in nanostructured Cu and Cu-Al alloys processed by high-pressure torsion and subsequent annealing. *Scr. Mater.* **66**, 227-230 (2012).
56. Benghalem, A. & Morris, D. G. Microstructure and strength of wire-drawn Cu-Ag filamentary composites. *Acta Mater.* **45**, 397-406 (1997).
57. Zhang, B. B., Tao, N. R. & Lu, K. A high strength and high electrical conductivity bulk Cu-Ag alloy strengthened with nanotwins. *Scr. Mater.* **129**, 39-43 (2017).
58. Lim, M. S., Song, J. S. & Hong, S. I. Microstructural and mechanical stability of Cu-6 wt.% Ag alloy. *J. Mater. Sci.* **35**, 4557-4561 (2000).
59. Davis, J. R. *Copper and copper alloys*. (ASM International, 2001).
60. Lu, D. P. et al. Study on high-strength and high-conductivity Cu-Fe-P alloys. *Mater. Sci. & Eng. A* **421**, 254-259 (2006).
61. Yang, K. et al. Recent development of advanced precipitation-strengthened Cu alloys with high strength and conductivity: A review. *Prog. Mater. Sci.* **138**, 101141 (2023).
62. Yang, Y. H. et al. Microstructure and properties of high-strength Cu-Ni-Si-(Ti) alloys. *Rare Metals* **40**, 3251-3260 (2021).
63. Lee, S., Im, Y. D., Matsumoto, R. & Utsunomiya, H. Strength and electrical conductivity of Cu-Al alloy sheets by cryogenic high-speed rolling. *Mater. Sci. & Eng. A* **799**, 139815 (2021).
64. Keene, B. J. Review of data for the surface tension of pure metals. *Int. Mater. Rev.* **38**, 157-192

- (1993).
65. Chase, M. W. NIST-JANAF thermochemical tables. *J. Phys. Chem. Ref. Data Monogr.* **9**, 1929-1564 (1998).
 66. Cotton, J. Boron carbide (B_4C) - Properties and Information about Boron Carbide. *AZO Materials* <https://www.azom.com/properties.aspx?ArticleID=75> (2001).