

Dispersion hardening using amorphous nanoparticles deployed via additive manufacturing

Corresponding Author: Professor Di Zhang

This file contains all reviewer reports in order by version, followed by all author rebuttals in order by version.

Version 0:

Reviewer comments:

Reviewer #1

(Remarks to the Author)

Incorporation via additive manufacturing of amorphous B4C nanoparticles into Cu leads to excellent strength ductility balance owing to a good strain-hardening capacity. In addition, a significant reduction of the defect density is found, providing exceptional fatigue resistance. The root causes of these superior properties are elucidated by combining different characterization methods and modelling approaches.

Taken alone, the very good tensile properties would constitute a very nice result but nothing exceptional. But what makes the finding really arresting is the other gain on the lower defectuosity, with enhanced fatigue resistance. This combination gives to the work an excellent degree of novelty. In addition, the study is very complete. For each section I was reading I was trying to find the weak point (to be sufficiently critical with respect to a publication in a journal of the stature of Nature Comm). I was always noting on a sheet of paper what should be done to consolidate the message but immediately after when reading the next section, I was finding a response to my anticipated comment/question. And this has been true till the end of the paper. The authors have really done an enormous amount of work in terms of characterization, testing and modelling to secure the findings and explain the root causes of what they report.

In other words, my appreciation is thus clearly positive.

But this does not mean everything is perfect.

1. First of all, there is a competing Cu-based system that would be nice to mention and analyze deeper (and compare in the charts) : the CuBe alloy. The strength-ductility balance is even better compared to the present system (and this without any addition of nanoparticles). Of course Be is toxic, but it has found several applications in key technologies. Furthermore, maybe the combination of Cu+Be and the amorphous B4C would lead to never attained mechanical performances ?

2. Regarding the mechanical properties, one of the key, if not the main aspect, is the sustained strain hardening capacity explained by the sink in mechanism at the particle interface. It would be worth giving more information about the strain hardening beyond the yield strength model. The starting point would be a Kocks-Mecking hardening rate plot and an attempt to build a closed form model for the strain hardening capacity.

3. Looking at such analysis of the hardening rate evolution, one would confirm through application of the Considere criterion that there is no premature necking, which is often the case in additive manufacturing parts due to the presence of several percents of porosity.

4. In the same spirit, a useful information that would be worth reporting and discussing (in supplementary or extended data) is the rate sensitivity and activation volume, the last being a signature of the dominant plasticity mechanism.

5. One also misses the values of the true fracture strain (from reduction of area at fracture) giving a better quantification of the ductility than the total elongation which is a not a very informative quantity (that depends on the gauge length for instance).

6. The explanation regarding the accumulation of Orowan loops is not totally clear to me when put together with the message that the dislocations sink in at the interface with the amorphous particles. Is it because there is a balance between

the two phenomena and that the sink in only occurs if the stress undergone by the first dislocation loop next to the particle is sufficiently large ? (hence limiting the number of loops accumulating by the Orowan mechanism, acting as form of pile up around the particle) or please justify the difference between a pile up and an accumulation of several Orowan loops.

Reviewer #3

(Remarks to the Author)

Overall this is intriguing work. I applaud the authors for a lot of hard work and a reasonably comprehensive manuscript presenting novel results.

However, there were some points that I believe should be addressed:

- 1) On line 78, the authors claim that "the B4C microparticles spontaneously transformed into nanoparticles (size ~47 nm)..." This statement seems misleading in two ways: i) the transformation is not spontaneous, but is a driven process depending on energy density and scanning time (Fig 2). It is the driven fragmentation process that is the novelty of the work and only occurred under precise circumstances that required excess energy input. ii) The claim that the particles are ~47 nm is also an oversimplification that might lead to misinterpretations. Fig 2 suggests that these are not monosized particles but they have a distribution of sizes. Extended data Fig 7 also seems to support that there is a distribution of sizes. What is the distribution? Is 47 nm a median size, or what is the median size?
- 2) The manuscript describes the particles as amorphous but actually the results show that they are actually an amorphous-nanocrystalline nanocomposite (Fig 1). Perhaps a more appropriate term would be quasi-amorphous, or I see some people use an acronym like amorphous-crystalline nanocomposite (ACN). It's not clear to me how critical the presence of crystallites are. The authors mention that the amorphous nature of the particles helps prevent cracking at the particle-matrix interface a result that is inferentially supported herein. But what would happen if the particles were fully amorphous?
- 3) On a related note, the authors argue that the amorphous nature is due to the rapid solidification avoiding thermodynamically-preferred crystallization of the B4C. This seems plausible qualitatively. But to get an amorphous phase with the onset of crystallization in the form of internal nm-scale crystallites seems to suggest a very precise solidification rate. Can the authors comment on what the solidification rate is, and how sensitive the particle size and crystallinity is on the precision of the solidification rate?
- 4) The impressive fatigue curve is shown for 4% ANP. What about the other cases up to 12% ANP? Does the fatigue curve shift up or down? I could argue both ways, but the current work seems incomplete with regard to fatigue results.

Version 1:

Reviewer comments:

Reviewer #1

(Remarks to the Author)

The authors have convincingly answered all my queries/suggestions. When needed they indeed added the requested extra information and new data such as the stress relaxation experiments, extraction of activation volume, Kocks Mecking plot, true fracture strain. These new data contribute to position better the potential of the proposed class of materials.

I just have one minor point to be treated for the final version :

- Be is indeed toxic but when Be is in Cu it is not truly toxic anymore (so please write Be is toxic not CuBe); note that CuBe was for instance used for the brakes of Ferrari cars, such as some formula 1. It was giving them a big advantage (this is partly explained in the book of MF Ashby on material selection)!

Reviewer #3

(Remarks to the Author)

I find that the authors have addressed most of the questions with reasonable improvements to the text.

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Response letter

We thank all the reviewers for their in-depth review and constructive comments. Below, we address each point raised, first repeating the review comments and then providing our responses in blue font. The manuscript has been revised accordingly, with major changes highlighted in red font.

Reviewer #1:

Incorporation via additive manufacturing of amorphous B₄C nanoparticles into Cu leads to excellent strength ductility balance owing to a good strain-hardening capacity. In addition, a significant reduction of the defect density is found, providing exceptional fatigue resistance. The root causes of these superior properties are elucidated by combining different characterization methods and modelling approaches.

Taken alone, the very good tensile properties would constitute a very nice result but nothing exceptional. But what makes the finding really arresting is the other gain on the lower defectuosity, with enhanced fatigue resistance. This combination gives to the work an excellent degree of novelty. In addition, the study is very complete. For each section I was reading I was trying to find the weak point (to be sufficiently critical with respect to a publication in a journal of the stature of Nature Comm). I was always noting on a sheet of paper what should be done to consolidate the message but immediately after when reading the next section, I was finding a response to my anticipated comment/question. And this has been true till the end of the paper. The authors have really done an enormous amount of work in terms of characterization, testing and modelling to secure the findings and explain the root causes of what they report.

In other words, my appreciation is thus clearly positive.

But this does not mean everything is perfect.

Response: We sincerely thank the reviewer for his positive and constructive comments. Below, we provide a point-by-point response to the concerns raised.

1. First of all, there is a competing Cu-based system that would be nice to mention and analyze deeper (and compare in the charts): the CuBe alloy. The strength-ductility balance is even better compared to the present system (and this without any addition of nanoparticles). Of course Be is toxic, but it has found several applications in key technologies. Furthermore, maybe the combination of Cu+Be and the amorphous B₄C would lead to never attained mechanical performances?

Response: Thank you for the reviewer's suggestion. As the reviewer pointed out, the strength-ductility combination of Cu-Be alloy is indeed excellent. The toxic nature and relatively low electrical conductivity of this alloy may, however, limit the large-scale applications. We apologize for overlooking this specific alloy system and have now included its tensile property

data in the revised property comparison figure (Fig. 3c).

It is important to note that the B₄C-reinforced pure copper was initially employed as a model to demonstrate the basic principle of dispersion hardening with amorphous nanoparticles. However, we anticipate that nanocomposites with unprecedented mechanical properties will be achieved when this strategy is extended to typical high-strength copper alloys, such as Cu-Be alloys. In response to the reviewer's comment, we have added the following sentences to the revised manuscript in line 25, page 4:

“Our copper nanocomposites also demonstrate strength–ductility combinations that surpass previous types of copper-based alloys and are comparable to the well-known high-strength copper-beryllium alloy (which is toxic and has relatively low electrical conductivity) (Fig. 3c; see Supplementary Note 2 for referential data). It is worth mentioning that ANP-reinforced pure copper is used here to demonstrate the basic principle of dispersion hardening with amorphous nanoparticles. Nanocomposites with unprecedented mechanical properties are highly anticipated when this strategy is applied to typical high-strength copper alloys.”

2. Regarding the mechanical properties, one of the key, if not the main aspect, is the sustained strain hardening capacity explained by the sink in mechanism at the particle interface. It would be worth giving more information about the strain hardening beyond the yield strength model. The starting point would be a Kocks-Mecking hardening rate plot and an attempt to build a closed form model for the strain hardening capacity.

Response: We thank the reviewer for the constructive suggestion. In response to the comment, we have provided the corresponding strain hardening rate (Θ)-true strain plot in Supplementary Figs. 6a-6b, where a notably higher Θ is observed for various Cu-ANP nanocomposites throughout the uniform deformation process (i.e., before necking occurs). Additionally, the strain hardening exponent n (calculated for applied strain ranging from 1% to the uniform elongation, using the Hollomon equation) is also provided and compared in Supplementary Fig. 6c, demonstrating the improved strain hardening capacity of Cu-ANP nanocomposites. This enhanced strain hardening coefficient, in conjunction with the Kocks-Mecking hardening rate plot, underscores the critical role of the amorphous nanoparticle strategy in achieving excellent mechanical properties. To clarify this point, we have added the following sentence to the revised manuscript (line 37, page 4):

“The superiority of ANP in tuning mechanical properties is primarily manifested in its ability to enhance strain hardening, as shown by the Kocks-Mecking hardening rate plot and the increased strain hardening exponent based on the Hollomon equation (Supplementary Figs. 6a-6c). As a result, Cu-ANP exhibits twice the total elongation of Cu-CNP with the same nanoparticle content.”

3. Looking at such analysis of the hardening rate evolution, one would confirm through application of the Considere criterion that there is no premature necking, which is often the

case in additive manufacturing parts due to the presence of several percents of porosity.

Response: Thanks for the reviewer's suggestion. We have plotted the normalized strain hardening rate–true strain curves to evaluate the initiation of deformation instability (i.e., necking) for different nanocomposites, based on the Considère criterion. These curves are presented in Supplementary Figs. 6d-6e.

As shown, necking in Cu-ANP nanocomposites basically follows the typical Considère criterion, indicating the porosity-free nature of the samples and the absence of premature failure mechanisms. In contrast, Cu-12%-CNP nanocomposite fractures before necking appears, which is attributed to the strong deformation incompatibility between the CNPs and the copper matrix. To clarify the different failure mechanisms, and following the reviewer's suggestion, we have added the following sentence to the revised manuscript (line 37, page 5):

“The absence of premature failure in Cu-ANP nanocomposites is consistent with the presence of necking, as predicted by the Considère criterion (see Supplementary Fig. 6d); however, the Cu-12%-CNP nanocomposite fractures before necking occurs, indicating significant stress concentration.”

4. In the same spirit, a useful information that would be worth reporting and discussing (in supplementary or extended data) is the rate sensitivity and activation volume, the last being a signature of the dominant plasticity mechanism.

Response: This is an excellent suggestion. We fully agree with the reviewer's comment that the activation volume, particularly the physical activation volume (V^*), can provide significant insights into the deformation mechanisms of our L-PBF copper nanocomposites. Using the Cu-8%-ANP nanocomposite as a demonstration, we conducted multiple stress relaxation experiments to investigate the underlying hardening mechanisms (new Supplementary Fig. 11).

As shown in Supplementary Fig. 11c, the V^* of the Cu-ANP nanocomposite is $28.2 \pm 1.3 b^3$ at an engineering strain of 0.01 and decreases with increasing strain, reaching $18.3 \pm 0.4 b^3$ at a strain of 0.04. This V^* is comparable to or even smaller than those observed in nanograined metals, suggesting that the deformation behaviour is dominated by interfacial interactions induced by ANPs. Furthermore, our results reveal that mobile dislocations are exhausted at a very high rate, with only 28% of the initial dislocation density remaining after the first relaxation cycle at a strain of 0.01 (Supplementary Fig. 11d). Considering the absence of strong dislocation accumulation at interface (Fig. 4a of the main text) and the minimal local misorientation across the ANPs, this rapid exhaustion is attributed to the uniform dislocation accumulation throughout the deformed region, which aligns with the material's remarkable strain hardening ability.

Based on the reviewer's suggestion, we have included the experimental details in Method section, the activation volume results in the new Supplementary Fig. 11, and added the following sentence to the revised manuscript (line 17, page 7):

“In addition, the dislocation segments that interact locally with ANP at the interface

undergo a thermally activated process with small activation volumes. This behaviour is reflected in multiple stress relaxation tests, which revealed a physical activation volume (V^*) of $\sim 20 b^3$ for the representative Cu-8%-ANP nanocomposite (Supplementary Fig. 11). This value is comparable to, or even smaller than, those reported in nanograined metals (on the order of tens of b^3), highlighting the unique dislocation manipulation behaviour induced by ANPs.”

5. One also misses the values of the true fracture strain (from reduction of area at fracture) giving a better quantification of the ductility than the total elongation which is a not a very informative quantity (that depends on the gauge length for instance).

Response: Thank you for the valuable suggestion. In the revised manuscript, we have included scanning electron microscopy (SEM) images of the fracture surfaces of the representative L-PBF nanocomposites with 4 vol.% of B₄C. Based on these images, we measured the reduction of area, defined as $q = \frac{A_0 - A_f}{A_0}$, where A_0 and A_f are the initial and fracture cross-sectional areas, respectively.

Using this data, the true fracture strain (ϵ_f) was estimated as:

$$\epsilon_f = \ln \left(\frac{1}{1-q} \right) \quad (R1)$$

The results, as provided in Supplementary Fig. 5, clearly highlight the advantage of the representative Cu-ANP nanocomposite in sustaining large plastic deformation and resisting fracture as compared with Cu-CNP (ϵ_f to be 1.179 and 0.743, respectively). We have added the sentence *“This enhanced tensile ductility of Cu-ANPs is further evidenced by the improvement in true fracture strain, estimated from the reduction in area at fracture (Supplementary Fig. 5).”* In line 19, page 4 of the revised manuscript, and have provided the fracture morphology analyses as a new supplementary figure (Supplementary Fig. 5).

6. The explanation regarding the accumulation of Orowan loops is not totally clear to me when put together with the message that the dislocations sink in at the interface with the amorphous particles. Is it because there is a balance between the two phenomena and that the sink in only occurs if the stress undergone by the first dislocation loop next to the particle is sufficiently large? (hence limiting the number of loops accumulating by the Orowan mechanism, acting as form of pile up around the particle) or please justify the difference between a pile up and an accumulation of several Orowan loops.

Response: This is an excellent question, and thanks for the reviewer for the insightful comment. As inferred from both experimental and modeling results in this work, dislocation in Cu-ANP nanocomposites can bypass the ANPs following the Orowan-strengthening mechanism. However, the traditional dislocation loops that would normally form after the bypassing process are no longer present (or are largely absorbed). These loops, along with the GNDs that

would otherwise pile up near the interfaces, are accommodated via shear transformations at the surface of the ANPs. Notably, this absorption process is readily activated because, in our molecular dynamics (MD) simulations, we do not observe any dislocation loops being stored at the ANPs after the first dislocation bypass (Fig. 4e in the main text). The lack of dislocation accumulation at the Cu-ANP interface is evident in the low stress concentration and misorientation observed at the interface after deformation (Fig. 4a-4b in the main text).

We apologize for not clarifying this point more explicitly earlier. To address this, we have made improvements to the manuscript to better explain the evolution of dislocation loops. These revisions include the following:

- (1) *“Traditionally, dislocations bypassing obstacles could accumulate in the form of Orowan loops...”* (line 26, page 5)
- (2) *“...They themselves dynamically work harden while simultaneously eliminating excessive dislocation loops and stress concentrations at the interfaces...”* (line 15, page 7)
- (3) *“As a result, the local nonhomogeneous deformation at the interfaces and the loops after dislocation bypassing can be effectively accommodated in the Cu-ANP nanocomposites, leaving minimal stress concentrations at the interfaces (Fig. 4e, right panel).”* (line 33, page 7)

We hope these revisions address the raised issues, thank you!

Reviewer #3:

Overall this is intriguing work. I applaud the authors for a lot of hard work and a reasonably comprehensive manuscript presenting novel results.

However, there were some points that I believe should be addressed:

Response: We thank the reviewer for his appreciation on our work. Below, we provide a point-to-point reply to the issues raised.

1) On line 78, the authors claim that "the B₄C microparticles spontaneously transformed into nanoparticles (size ~47 nm)..." This statement seems misleading in two ways: i) the transformation is not spontaneous, but is a driven process depending on energy density and scanning time (Fig 2). It is the driven fragmentation process that is the novelty of the work and only occurred under precise circumstances that required excess energy input. ii) The claim that the particles are ~47 nm is also an oversimplification that might lead to misinterpretations. Fig 2 suggests that these are not monosized particles but they have a distribution of sizes. Extended data Fig 7 also seems to support that there is a distribution of sizes. What is the distribution? Is 47 nm a median size, or what is the median size?

Response: We sincerely thank the reviewer for pointing this out and apologize for the improper description. As the reviewer mentioned, the B₄C microparticles were melted, driven to fragment into nanoparticles, and subsequently solidified. This process was not spontaneous but rather induced by the extremely non-equilibrium nature of the L-PBF process. Furthermore, the nanoparticle size is non-uniformly distributed in the as-built samples, with an average size of 47 nm based on Gaussian fitting (the size distribution is provided in the new Supplementary Fig. 1). In line with the reviewer's constructive comment, we have corrected the sentence in the revised manuscript as follows:

"...the B₄C microparticles were melted and transformed into nanoparticles with an average size of ~47 nm (see Supplementary Fig. 1a for particle size distributions)..."

This correction has been made in line 35, page 2 of the revised manuscript.

2) The manuscript describes the particles as amorphous but actually the results show that they are actually an amorphous-nanocrystalline nanocomposite (Fig 1). Perhaps a more appropriate term would be quasi-amorphous, or I see some people use an acronym like amorphous-crystalline nanocomposite (ACN). It's not clear to me how critical the presence of crystallites are. The authors mention that the amorphous nature of the particles helps prevent cracking at the particle-matrix interface a result that is inferentially supported herein. But what would happen if the particles were fully amorphous?

Response: This is an excellent question. We agree with the reviewer's suggestion that the fragmented B₄C nanoparticle is more appropriately termed as "quasi-amorphous" nanoparticle. We have revised the terminology accordingly when describing the microstructure of the nanoparticle in Fig. 1 (line 8, page 3 of the revised manuscript), while retaining the

abbreviation “ANPs” for “quasi-amorphous nanoparticles” thereafter to facilitate a clearer understanding of the microstructural characteristics.

We fully understand the reviewer’s concern that the presence of copper nanocrystals within the amorphous B₄C nanoparticles (i.e., the ultranano-in-nano structure) may play an important role. As indicated in our main text, the copper nanocrystals are by-products of B₄C nanofragmentation. This occurs because the strong chemical affinity between B and C, along with their robust atomic bonding, prevents the diffusion of individual elements even under molten conditions. This phenomenon is supported by our ab initio molecular dynamics (AIMD) calculations of the formation enthalpy of B₄C at various temperatures (Supplementary Fig. 2). Conversely, copper atoms diffuse into the B₄C nanoparticles and form ultra-nano crystallites.

However, 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 materials [R1]. 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.

Therefore, the influence of the copper nanocrystal on the deformation mechanisms and deformation stability of ANPs is considered insignificant. In response to the reviewer’s comment and the above discussion, we have revised the manuscript by adding a sentence “*We further note that the copper nanocrystals in ANPs could play an insignificant role in the deformation mechanisms (see additional discussion in Supplementary Note 3 for more details).*” in line 17, page 8, and included an additional discussion (Supplementary Note 3) to further elaborate on the role of copper nanocrystals in ANPs.

[R1] 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).

3) On a related note, the authors argue that the amorphous nature is due to the rapid

solidification avoiding thermodynamically-preferred crystallization of the B₄C. This seems plausible qualitatively. But to get an amorphous phase with the onset of crystallization in the form of internal nm-scale crystallites seems to suggest a very precise solidification rate. Can the authors comment on what the solidification rate is, and how sensitive the particle size and crystallinity is on the precision of the solidification rate?

Response: Thank you for the comment. We would like to emphasize that the crystalline phases within the amorphous B₄C nanoparticles are copper nanocrystals, formed due to the diffusion of Cu atoms into the B₄C melt. This process had been elucidated through both physical modeling (including AIMD simulations and multiphysics modeling) and experimental results (STEM and APT) presented in the main text. As noted in our response to Comment 2, the diffusion of Cu atoms in the melt is more extensive than that of B or C because of the strong chemical affinity between B and C, which prevents the diffusion of individual elements under molten conditions. The Cu atoms subsequently re-precipitate to form copper nanocrystals within the amorphous B₄C structure due to the low solubility of Cu in both B and C.

The cooling rate during solidification in L-PBF Cu-B₄C, as estimated by the multiphysics modeling (Supplementary Fig. 4), is on the orders of 10⁷-10⁸ K/s. This extremely high cooling rate is likely the key factor responsible for the formation of the amorphous structure of B₄C. However, such a high cooling rate is insufficient to vitrify face-centred cubic (fcc) monatomic metals like copper [R2]. This is due to the heterogeneous nucleation and rapid crystal growth of supercooled liquids, which results in the formation of ultra-nano copper crystals (with an average size of ~2.9 nm) embedded with the amorphous B₄C.

[R2] Tong, X. et al. Breaking the vitrification limitation of monatomic metals. *Nat. Mater.* **23**, 1193-1199 (2024).

4) The impressive fatigue curve is shown for 4% ANP. What about the other cases up to 12% ANP? Does the fatigue curve shift up or down? I could argue both ways, but the current work seems incomplete with regard to fatigue results.

Response: This is an excellent question and we thank the reviewer for pointing it out. Previously, we conducted fatigue experiment to verify the reduced propensity for damage initiation and accumulation in L-PBF Cu-ANP nanocomposites, using a sample with 4 vol.% of ANP as a demonstration. Following the reviewer's suggestion, we have now examined the fatigue properties of a Cu-12%-ANP sample, and the results have been added to Fig. 3d in the main text. Remarkably, the general trend of fatigue life versus the fatigue strength-to-tensile strength ratio in the Cu-12%-ANP nanocomposite is consistent with the Cu-4%-ANP sample, despite a lower fatigue strength-to-tensile strength ratio observed during low-cycle fatigue testing. It is worth mentioning that the primary focus of this study is to demonstrate the novel dispersion hardening enabled by amorphous nanoparticles through additive manufacturing. An in-depth investigation into the fatigue properties of L-PBF Cu-ANP nanocomposites, as well

as the mechanisms responsible for their remarkable fatigue resistance, is currently underway and will be presented in future work.

We hope these revisions address the raised issues, thank you!

Response letter

Reviewer #1:

The authors have convincingly answered all my queries/suggestions. When needed they indeed added the requested extra information and new data such as the stress relaxation experiments, extraction of activation volume, Kocks Mecking plot, true fracture strain. These new data contribute to position better the potential of the proposed class of materials.

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Response: We sincerely appreciate the reviewer's suggestion. Accordingly, we have revised the sentence to: *"Our copper nanocomposites also demonstrate strength–ductility combinations that surpass previous types of copper-based alloys and are comparable to the well-known high-strength copper-beryllium alloy (noting that beryllium is toxic) (Fig. 3c; see Supplementary Note 2 for referential data)."* This revision can be found on line 25, page 4 of the revised manuscript.