Research Article

Pei Li, Cunguang Chen*, Chenzeng Zhang, Na Xiao, Haifeng Zhang, Yang Li, and Zhimeng Guo

Effects of ball milling on powder particle boundaries and properties of ODS copper

<https://doi.org/10.1515/htmp-2021-0036>

received May 21, 2021; accepted September 22, 2021

Abstract: Al_2O_3 dispersion-strengthened (ODS) copper has an excellent comprehensive performance due to the strong hindrance of the high concentration nano- Al_2O_3 to the dislocations inside copper grains. However, the processability of ODS copper is seriously deteriorated, which is caused by the presence of unfavorable microlevel Al_2O_3 particles along powder particle boundaries. In this study, a strategy of ball-milling-induced impurity removal is adopted to surmount the dilemma. It was found that the ball milling process can significantly weaken the formation of large Al_2O_3 particles in the primary boundaries. However, due to the activation of the powder particle surface, the metallurgical bonding between the powder particles is strengthened. The results showed that the ball-milled samples exhibited the optimal properties, including the ultimate tensile strength of 488 ± 3 MPa, elongation of $18.7 \pm 0.7\%$, reduction in the area of $46.8 \pm 1.2\%$, 82.2 ± 0.3 Rockwell Hardness measured on the B scale (HRB), and electrical conductivity of $77.2 \pm 0.1\%$ International Association of Classification Societies (IACS).

Keywords: ODS copper, ball milling, Al_2O_3 particles, mechanical property, electrical conductivity

1 Introduction

Copper with face-centered cubic crystal texture exhibits good plasticity, excellent thermal conductivity, and electrical conductivity and thus has been widely used in

power, electronics, energy, transportation, light industry, emerging industries, and other fields [1–3]. However, its significant weakness is low strength, which limits its scope of application [4]. Therefore, numerous investigations have been conducted to improve the strength of copper materials for application requirements [5–8]. Among these efforts, dispersion strengthening and age hardening are regarded as extensively and successfully applied technologies to enhance the mechanical properties of copper alloys [9–11]. Although the age-strengthened alloy has excellent mechanical properties and electrical and thermal conductivity at room temperature, there is a primary disadvantage of poor high-temperature stability because of the growth and redissolution of precipitated phases [12,13]. Since nano- Al_2O_3 particles are very stable at high temperatures, the dispersion-strengthened copper has outstanding elevated temperature properties [14].

In recent years, dispersion-strengthened copper alloys have attracted significant attention because of their application in welding electrode, lead frame, and electronic packaging [15]. The prevalent methods for preparing dispersion-strengthened copper alloys mainly involve mechanical alloying and internal oxidation. In particular, internal oxidation has been currently applied to industrial mass production [16,17]. As the initial raw material, the water-atomized Cu–Al alloy powder is oxidized to obtain nano- Al_2O_3 particles-strengthened copper powder, followed by densification treatment [18]. During the water atomization, most metals react with water or steam to form oxides, following the chemical reaction: $x\text{Me} + y\text{H}_2\text{O} = \text{Me}_x\text{O}_y + y\text{H}_2$ [19]. The surface analysis of collected powder shows that the surface of the metal powder is usually covered with metallic oxides because the metal with high negative free energy and high diffusivity at elevated temperatures is liable to form oxides on the powder particle surface. The thickness of oxide film ranges from tens to thousands of nanometers [20–22]. Unfortunately, the resulting dispersion-strengthened copper has serious adverse effects on powder primary particle boundaries (PPBs) originating from aluminum oxidation, which roots in deoxidizing products in the Cu–Al alloy melting and aluminum segregation oxidation on the surface of powder particles during atomization [23].

* **Corresponding author: Cunguang Chen**, Institute for Advanced Materials and Technology, University of Science and Technology Beijing, Beijing 100083, China, e-mail: cgchen@ustb.edu.cn

Pei Li, Chenzeng Zhang, Haifeng Zhang, Yang Li, Zhimeng Guo: Institute for Advanced Materials and Technology, University of Science and Technology Beijing, Beijing 100083, China

Na Xiao: School of Material Science and Engineering, Tsinghua University, Beijing 100084, China

Generally, the PPB problem mainly exists in super-alloys and high-speed steel [24,25]. Recently, increasing attention has been paid to nonferrous metal powder metallurgy materials, such as copper alloys and aluminum alloys [26,27]. In fact, the different types of ceramic particles manifested inside PPBs are the immediate source of adverse effects, including oxides, carbides, and oxycarbides. The weakness of the interpowder bonding, resulting from the particles located in PPBs, thereby significantly influences the powder consolidation and dramatically reduces the elongation-to-failure and fracture toughness of materials [28].

The ceramic particles have been experimentally proved to be derived from the master alloy, the crucible lining, the leaking package, and the flushed nozzle [20]. The PPB problem and inclusions lead to a sharp decline in plasticity. Furthermore, some large oxide particle inclusions directly result in an abrupt fracture during the deformation process. Therefore, many approaches have been introduced to solve the PPB problem, including selecting clean raw materials, vacuum gas atomization, and powder purification [29,30]. However, these methods are hard to implement in the industries owing to their high cost.

In this study, a simple and effective method of ball milling was carried out to eliminate or weaken the PPB problem of the dispersion-strengthened copper alloy. The effects of ball milling on the mechanical and physical properties of dispersion-strengthened copper were studied, and the related mechanism was also interpreted.

2 Experimental

2.1 Material preparation

Figure 1 shows the schematic of the experimental procedure in this study. The Cu-0.25 wt% Al powder was prepared by water atomization, followed by internal oxidation, diffusion, and reduction. Dispersion-strengthened copper powders with internal nano- Al_2O_3 were obtained as starting materials. The initial powders were ground via a high-energy planetary ball mill with a ball-to-powder weight ratio of 3:1 for 1 h. The rotating speed was kept at 200 rpm. The stainless steel vial with the capacity of 1 L and the stainless steel balls (10 mm in diameter) were matched with the planetary ball mill. During the ball milling process, the ethanol was added as the process-control agent to prevent excessive cold welding. Subsequently, the powder reduction was carried out at 450°C for 2 h in hydrogen. Then

Figure 1: Schematic showing the process.

the powders were cold isostatic pressed at room temperature for 5 min with a pressure of 200 MPa, obtaining green bodies with a diameter of 45 and 40 mm in height. Next, the green bodies were sintered at 950°C for 1 h under a decomposing ammonia atmosphere and immediately extruded with a ratio of 36:1.

2.2 Characterization

The morphology of powder particles before and after ball milling was investigated and analyzed using a field-emission scanning electron microscope (FE-SEM; SUPER⁵⁵, ZEISS, Germany) equipped with an energy-dispersive X-ray spectrometer (EDS; Oxford Instruments, Abingdon, UK). The microstructure of specimens for mechanical testing was also identified using FE-SEM after mechanically grinding and polishing. The Rockwell hardness (HRB) was measured with 980 N load using hardmeter (HBRV-187.5, Hua Yin, China). The Archimedes method was applied to measure the density for each sample (ASTM B962-13). Tensile tests were conducted on an MTS810 testing machine with a constant speed of 0.5 mm·min⁻¹ at ambient temperature. At least three tensile samples were prepared to test, and the average values were taken according to the standard of ASTM-E8/E8M-08. The room temperature conductivity was determined with an eddy current device (PZ-60A, MEIYUNKE, China) and calculated according to the standard of ASTM E1004-09. The inclusion analysis was carried out by using the ASPEX inclusion analysis system (Aspex Explorer, FEI, USA). The analyzed area was 3 mm × 3 mm. In this study, only the inclusions with a maximum diameter of more than 0.5 μm were studied.

3 Results and discussion

3.1 Characteristics of powders

Figure 2(a) shows the SEM morphology of the water-atomized raw powders after reduction. It can be observed that the powders are displayed in an irregular shape with slightly sintered agglomeration. As shown in Figure 2(b), SEM observations with higher magnification reveal some spherical and white particles on the rough surface of the dispersion-strengthened powders. Further EDS analysis shows that these ~500 nm particles in size are basically spherical oxide composites mainly containing Al and O elements, as shown in Figure 3(a) and (b). These results are probably attributed to the aluminum segregation on the powder particle surface and oxidation during the water atomization process. According to previous research, the oxide particles are the fracture origin, which negatively influence the plasticity and fatigue properties. Therefore, it is essential to remove the oxide particles.

The SEM image of the cross-section of non-milling Al_2O_3 dispersion-strengthened (ODS) Cu powder is shown in Figure 3(c). There are white particles and lines on the grain gap. The concentration of Al and O on the grain boundary according to the EDS line-scan, as shown in Figure 3(d), is consistent with the findings obtained by Sabard and Low [31,32]. The concentration of Al and O on

the grain boundary causes the PPB problem and results in poor plasticity, which is hard to be well coordinated with the matrix during the deformation process, so these large-sized Al_2O_3 particles would become a potential crack source and induce crack propagation. There are extraordinary concerns that the continuous distribution of Al_2O_3 particles along the grain boundaries and particle boundaries weakens the interface bonding in the ODS Cu alloy, which dramatically affects the plastic deformation capacity [33].

The SEM morphology of powders after ball milling is shown in Figure 2(c) and (d). Obviously, the powder surface morphology exhibits a significant difference before and after ball milling. In Figure 2(c), the powder surface is rugged, whereas all the particles present a more uniform shape with a smooth surface after ball milling shown in Figure 2(d). The morphology of the powders after ball milling has been ground into a flake shape.

3.2 Microstructure and density

Figure 4 shows the back scattering diffraction (BSD) result of sintered ODS Cu at 950°C for 1 h under a decomposing ammonia atmosphere. During sintering, Al_2O_3 cluster on particle boundaries weakens the interface between the two Al_2O_3 particles because Al_2O_3 can hardly

Figure 2: Low-magnification (a and c) and high-magnification (b and d) SEM images of raw powder and milled powder.

Figure 3: (a) SEM image of the ODS Cu powder surface, (b) the EDS of white spot shown in (a), (c) cross-section of ODS powder, and (d) EDS scan over the powder boundaries in (c).

be sintered at this temperature. The sintering process also has been seriously hindered [34].

The black regions in Figure 4 present such residual pores in sintered bulks. Because of these defects, many pores are observed in the microstructure of sintered bulks of ODS Cu with poor sintered properties. The powder particles shown in Figure 4(b) are smaller than those shown in Figure 4(a) because ball milling process has opened the original agglomerated particles. Figure 5 shows the relative density of sintered samples. The density of non-milled and milled samples is 83.2 ± 0.3 and $80.5 \pm 0.6\%$, respectively. The hardness and yield strength of powder particles were improved by ball milling, so the

relative density of milled powder compact was lower than non-milled compact. Therefore, the sintering density of milled samples is lower. The density of non-milled and milled samples is 99.5 ± 0.4 and $99.7 \pm 0.3\%$ after extrusion. The main reason for this phenomenon is that the densification was strongly inhibited by Al_2O_3 in the copper matrix [35]. Figure 6 is the BSD microstructure of the extruded specimens. As observed from the cross-section of Figure 6(a) and (b), Al_2O_3 appears in the shape of a continuous net in the copper matrix. In the longitudinal section of the non-milled and milled samples shown in Figure 6(c) and (d), there were elongated oxide particles. In the cross-section and longitudinal sections, it could be found that the ball milling

Figure 4: BSD morphology of sintered samples by (a) raw powder and (b) milled powder.

Figure 5: Relative density changes after sintering and extrusion.

process reduced the number of inclusions and made them evenly distributed.

Besides, there was a significant change in the total inclusions and inclusions per unit area for each sample (shown in Table 1). The total number of non-milled inclusions is 7,151, whereas the number of inclusions after 1 h milling is 2,956. The number of inclusions per unit area decreases with ball milling, and the values of non-milled and milled are 58 and 36 mm^{-2} , respectively. The total number of inclusions decreases by 38% through ball milling.

Table 1: Total inclusions and inclusions per unit area for each sample

Milling time (h)	Numbers of total inclusions	Number of inclusions per unit area (mm^{-2})
0	7,151	58
1	2,956	36

Based on the rating criteria for PPBs reported by other researchers [36], the rating for PPBs of the extruded specimen without ball milling is Grade 3. With ball milling, the fine Al_2O_3 particles become discontinuous, indicating that the PPB problem decreases accordingly. After ball milling, the distribution of Al_2O_3 particles changes from the previously continuous network structure to a discontinuous condition along the boundary of the powders.

It is noticeable that the existence of a continuous network of Al_2O_3 particles along the boundary would cause a decrease in mechanical properties, especially in plastic-processing property. There are many factors affecting the PPB problem on the Cu–Al powder surface. However, water is highly in contact with Cu–Al alloy powders at high pressure during water atomization, and since the affinity of H_2O and Al is greater than Cu during water atomization, the elemental segregation of Al and O would occur on the powder surface [23].

Figure 6: BSD morphology of as-extruded samples: (a and b) cross-section and (c and d) longitudinal section of (a and c) non-milled sample and (b and d) milled sample.

Table 2: Mechanical and physical properties of dispersion-strengthened copper alloy before and after ball milling

Milling time (h)	Tensile strength (MPa)	Yield strength (MPa)	Elongation (%)	Section shrinkage (%)	HRB ($\text{N}\cdot\text{mm}^{-2}$)	Electrical conductivity (%)
0	435 ± 5	362 ± 3	14.3 ± 1.2	25.1 ± 0.8	72.0 ± 0.9	77.2 ± 0.3
1	488 ± 3	417 ± 4	18.7 ± 0.7	46.8 ± 1.2	82.2 ± 0.3	77.2 ± 0.1

However, during the smelting of Cu–Al alloys, Al readily reacts with oxygen to form alumina. The atomizing nozzle is generally made of alumina. Under the repeated scour action of the high-temperature liquid flow, the inner hole of the nozzle gradually becomes bigger so that the alumina flushed away from the nozzle is mixed with the powders. In addition, the raw powders and impurities, which dropped from furnace lining during smelting, would lead to the presence of micron-sized Al_2O_3 particles [36].

3.3 Mechanical properties

The values of mechanical and physical properties are shown in Table 2 and Figure 7. The tensile strength is 435 ± 5 MPa and increases to 488 ± 3 MPa with an enhancement of 12% after ball milling. The yield strength of the milled sample is 417 ± 4 MPa, with 15% increase compared to the non-milled sample. As shown in Table 2, the elongation and section shrinkage of ball milling specimen are much higher than that without ball milling, an increase of 30 and 88%, respectively, indicating the beneficial effect of ball milling on enhancing the plasticity of ODS Cu alloy. These promising results may be attributed to the fact that ball milling weakens the impact of PPBs

and improves the interface bonding. Figure 8 shows the schematic of ball milling process about the variation of oxides on the powder particle surface. Raw powders with PPBs were ball-milled in alcohol. Due to the violent collision during ball milling, part of the ceramic inclusions on the surface of the powder particles fell from the surface of the powder particles and entered into the alcohol. After ball milling, they were filtered out together with the alcohol. Under the interaction between the copper powder and the steel ball, the soft powder particles were squashed to produce layered composite powder, so that part of the ceramic particles originally on the surface of the powder was incorporated inside the powder [37]. Through the ball-milling treatment, the total content of ceramic inclusions is reduced, and at the same time, the related PPBs effect is weakened.

As the strength of the copper alloy is greatly affected by inclusions in the matrix, the larger the size and number of inclusions, the lower the yield strength and tensile strength of the material. Numerous reports have shown that cracks are apt to initiate around inclusions [39–42].

Herein, the number of inclusions has a greater influence on the section shrinkage than the elongation. Inclusions in copper reduce the transverse ductility of copper. Therefore, the tensile strength, yield strength, and elongation of copper material have been significantly improved by ball milling [38,39].

As shown in Figures 7 and 9(a) and (b), it is noted that the fracture surface contains typical dimples considered as the ductile fracture at low magnification. However, the substantial macroscopic difference in the fracture aspect is observed between non-milled and milled

Figure 7: Room temperature engineering strain vs stress curves of as-extruded samples. **Figure 8:** Schematic of ball milling on powder changes.

Figure 9: Fracture morphologies of non-milled sample (a, c and e) and milled sample (b, d and f) after extrusion.

samples. In Figure 9(c), the mixed fracture surface consisted of fine dimples, and a few particle boundary facets can be found. These PPBs mostly contain fine powders with severe surface oxidation, which are difficult to be sintered and easily maintain the original state even after the extrusion process [24,43]. The oxide on the surface of the powder along PPBs can inhibit boundary migration and lead to interparticle fracture. In Figure 9(d), the phenomenon of PPBs could hardly be observed, and the fracture surface consists of fine dimples and some secondary cracks. Figure 9(e) and (f) in the high magnification of non-milled and milled samples shows small dimples due to grain tearing.

As seen in Table 2, the variation trend of the HRB is the same as the tensile strength. The hardness value of the sample without ball milling is 72.0 ± 0.9 HRB. The hardness reaches 82.2 ± 0.3 HRB after ball milling. Moreover, the electrical conductivity of the fabricated dispersion-strengthened copper remains nearly unchanged after ball

milling. The electrical conductivity is unaffected before and after ball milling.

4 Conclusion

In this study, the effect of ball milling of dispersion-strengthened copper on its microstructure and properties was investigated. Conclusions are as follows:

- (1) The ball milling process significantly weakens the PPB problem of ODS Cu alloy compared with the non-milling samples.
- (2) After ball milling, the elongation increases by 30%, and the section shrinkage sharply increases by 88% compared with non-milling specimens.
- (3) The ultimate tensile strength and hardness of specimens reached to 488 ± 3 MPa and 82.2 ± 0.3 HRB with an enhancement of 12 and 14%, respectively.

Acknowledgments: The authors gratefully acknowledge financial support from the institution of the National Natural Science Foundation of China (Grant No. 92066205) and the Fundamental Research Funds for the Central Universities (Grant No. FRF-MP-20-52).

Funding information: This study was supported by the National Natural Science Foundation of China (Grant No. 92066205) and the Fundamental Research Funds for the Central Universities (Grant No. FRF-MP-20-52).

Author contributions: Pei Li: methodology, validation, investigation, resources, writing – original draft, visualization. Cunguang Chen: conceptualization, methodology, validation, writing – review and editing, supervision, funding acquisition. Chenzeng Zhang: resources, formal analysis, data curation. Na Xiao: resources and formal analysis. Haifeng Zhang and Yang Li: data processing and article corrections. Zhimeng Guo: methodology, resources, and funding acquisition.

Conflict of 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 article.

Data availability statement: The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.

References

- [1] Jones, W. E., J. Chiguma, W. Johnson, A. Pachamuthu, and S. Santos. Electrically and thermally conducting nanocomposites for electronic applications. *Materials*, Vol. 3, 2010, pp. 1478–1496.
- [2] Guo, T. B., S. Wei, C. wang, Q. Li, and Z. Jia. Texture evolution and strengthening mechanism of single crystal copper during ECAP. *Materials Science and Engineering: A*, Vol. 759, 2019, pp. 97–104.
- [3] Lu, X. D., X. C. Lu, X. Zhang, M. X. Shi, F. Roters, G. Z. Kang, et al. Dislocation mechanism based size-dependent crystal plasticity modeling and simulation of gradient nano-grained copper. *International Journal of Plasticity*, Vol. 113, 2019, pp. 52–73.
- [4] Afifeh, M., S. Hosseinipour, and R. Jamaati. Nanostructured copper matrix composite with extraordinary strength and high electrical conductivity produced by asymmetric cryorolling. *Materials Science and Engineering: A*, Vol. 763, 2019, id. 138146.
- [5] Zhang, X. H., Y. Zhang, B. Tian, K. Song, P. Liu, Y. L. Jia, et al. Review of nano-phase effects in high strength and conductivity copper alloys. *Nanotechnology Reviews*, Vol. 8, No. 1, 2019, pp. 383–395.
- [6] Lemougna, P. N., J. Yliniemi, E. Adesanya, P. Tanskanen, P. Kinnunen, J. Roning, et al. Reuse of copper slag in high-strength building ceramics containing spodumene tailings as fluxing agent. *Minerals Engineering*, Vol. 155, 2020, id. 106448.
- [7] Sarkar, S., C. Srivastava, and K. Chattopadhyay. Development of a new class of high strength copper alloy using immiscibility route in Cu-Fe-Si system: evolution of hierarchical multi-scale microstructure. *Materials Science and Engineering: A*, Vol. 723, 2018, pp. 38–47.
- [8] Zaidi, K. A., S. Ram, and M. K. Gautam. Utilisation of glass powder in high strength copper slag concrete. *Advances in Concrete Construction*, Vol. 5, No. 1, 2017, id. 065.
- [9] Gan, G., C. H. Zhang, D. H. Yang, and M. B. Yang. Effect of TiB₂ on microstructure of 7075 Al alloy semi-solid slurry at different solid fraction. *Materials Transactions*, Vol. 58, No. 5, 2017, pp. 734–738.
- [10] Correia, J. B., H. A. Davies, and C. M. Sellars. Strengthening in rapidly solidified age hardened Cu-Cr and Cu-Cr-Zr alloys. *Acta Materialia*, Vol. 45, No. 1, 1997, pp. 177–190.
- [11] Fu, Y. B., Q. F. Pan, Z. Q. Cao, S. F. Li, and Y. Q. Huo. Strength and electrical conductivity behavior of nanoparticles reaction on new alumina dispersion-strengthened copper alloy. *Journal of Alloys and Compounds*, Vol. 798, 2019, pp. 616–621.
- [12] Thompson, S. Interrelationships between yield strength, low-temperature impact toughness, and microstructure in low-carbon, copper-precipitation-strengthened, high-strength low-alloy plate steels. *Materials Science and Engineering: A*, Vol. 711, 2018, pp. 424–433.
- [13] Sevostyanov, M. A., A. G. Kolmakov, K. V. Sergiyenko, M. A. Kaplan, A. S. Baikin, and S. V. Gudkov. Mechanical, physical-chemical and biological properties of the new Ti-30Nb-13Ta-5Zr alloy. *Journal of Materials Science*, Vol. 55, No. 29, 2020, pp. 14516–14529.
- [14] Daoud, A., J. B. Vogt, E. Charkaluk, L. Zhang, and J. C. Biasci. Effect of temperature on the low cycle fatigue behavior of Glidcop Al-15. *Procedia Engineering*, Vol. 2, No. 1, 2010, pp. 1487–1495.
- [15] Broyles, S. E., M. Zhang, and J. C. Gibeling. Influence of annealing on the creep behavior of GlidCop Al-15. *Materials Science and Engineering: A*, Vol. 779, 2020, id. 139112.
- [16] Zhou, X. Y., D. Q. Yi, L. Nylorg, Z. Hu, J. Huang, and Y. Cao. Influence of Ag addition on the microstructure and properties of copper-alumina composites prepared by internal oxidation. *Journal of Alloys and Compounds*, Vol. 722, 2017, pp. 962–969.
- [17] Aghamiri, S. M. S., N. Oono, S. Ukai, R. Kasada, H. Noto, Y. Hishinuma, et al. Microstructure and mechanical properties of mechanically alloyed ODS copper alloy for fusion material application. *Nuclear Materials and Energy*, Vol. 15, 2018, pp. 17–22.
- [18] Li, Z. D., C. G. Lin, and S. Cui. Development of research and application of copper alloys with high strength and high conductivity. *Advanced Materials Research*, Vol. 1053, 2014, pp. 61–68.

- [19] Gany, A. Innovative concepts for high-speed underwater propulsion. *International Journal of Energetic Materials and Chemical Propulsion*, Vol. 17, 2018, pp. 83–109.
- [20] Simonelli, M., C. Tuck, N. Aboulkhair, I. Maskey, I. Ashcroft, R. D. Wildman, et al. A study on the laser spatter and the oxidation reactions during selective laser melting of 316L stainless steel, Al-Si10-Mg, and Ti-6Al-4V. *Metallurgical and Materials Transactions A*, Vol. 46, 2015, pp. 3842–3851.
- [21] Qiu, T. T., M. Wu, Z. Y. Du, G. Chen, L. Zhang, and X. H. Qu. Microstructure evolution and densification behaviour of powder metallurgy Al-Cu-Mg-Si alloy. *Powder Metallurgy*, Vol. 63, No. 1, 2020, pp. 54–63.
- [22] Ichikawa, Y., R. Tokoro, M. Tanno, and K. Ogawa. Elucidation of cold-spray deposition mechanism by auger electron spectroscopic evaluation of bonding interface oxide film. *Acta Materialia*, Vol. 164, 2019, pp. 39–49.
- [23] Gao, Z. J., G. Q. Zhang, Z. Li, H. Yuan, W. Y. Xu, and Y. Zhang. Surface segregation and oxidation behavior of superalloy powders fabricated by argon atomization. *Materials Science Forum*, Vol. 747, 2013, pp. 518–525.
- [24] Chang, L., W. Sun, Y. Cui, and R. Yang. Preparation of hot-isostatic-pressed powder metallurgy superalloy Inconel 718 free of prior particle boundaries. *Materials Science and Engineering: A*, Vol. 682, 2017, pp. 341–344.
- [25] Sergi, A., R. H. U. Khan, and M. M. Attallah. The role of powder atomisation route on the microstructure and mechanical properties of hot isostatically pressed Inconel 625. *Materials Science and Engineering: A*, Vol. 808, 2021, id. 140950.
- [26] Wang, H. S., H. G. Chen, J. W. Gu, C. E. Hsu, and C. Y. Wu. Improvement in strength and thermal conductivity of powder metallurgy produced Cu-Ni-Si-Cr alloy by adjusting Ni/Si weight ratio and hot forging. *Journal of Alloys and Compounds*, Vol. 633, 2015, pp. 59–64.
- [27] Chen, C. G., W. W. Wang, Z. M. Guo, C. B. Sun, A. A. Volinsky, and V. Paley. Annealing effects on microstructure and mechanical properties of ultrafine-grained Al composites reinforced with nano- Al_2O_3 by rotary swaging. *Journal of Materials Engineering and Performance*, Vol. 27, 2018, pp. 1738–1745.
- [28] Zhao, L., X. H. Zhang, T. Q. Deng, and J. Jiang. Develop an effective oxygen removal method for copper powder. *Advanced Powder Technology*, Vol. 29, 2018, pp. 1904–1912.
- [29] MacDonald, J. E., R. H. U. Khan, M. Aristizabal, K. E. A. Essa, M. J. Lunt, and M. M. Attallah. Influence of powder characteristics on the microstructure and mechanical properties of HIPped CM247LC Ni superalloy. *Materials and Design*, Vol. 174, 2019, id. 107796.
- [30] Shu, D. L., S. G. Tian, N. Tian, J. Xie, and Y. Su. Thermodynamic analysis of carbide precipitation and effect of its configuration on creep properties of FGH95 powder nickel-based superalloy. *Materials Science and Engineering: A*, Vol. 700, 2017, pp. 152–161.
- [31] Sabard, A., H. L. Lovelock, and T. Hussain. Microstructural evolution in solution heat treatment of gas-atomized Al alloy (7075) powder for cold spray. *Journal of Thermal Spray Technology*, Vol. 27, No. 1, 2018, pp. 145–158.
- [32] Lou, X. Y., P. L. Andresen, and R. B. Rebak. Oxide inclusions in laser additive manufactured stainless steel and their effects on impact toughness and stress corrosion cracking behavior. *Journal of Nuclear Materials*, Vol. 499, 2018, pp. 182–190.
- [33] Nehila, A., W. Li, N. Gao, X. X. Xing, H. Q. Zhao, P. Wang, et al. Very high cycle fatigue of surface carburized CrNi steel at variable stress ratio: failure analysis and life prediction. *International Journal of Fatigue*, Vol. 111, 2018, pp. 112–123.
- [34] Wagih, A. and A. Fathy. Experimental investigation and FE simulation of spherical indentation on nano-alumina reinforced copper-matrix composite produced by three different techniques. *Advanced Powder Technology*, Vol. 28, No. 8, 2017, pp. 1954–1965.
- [35] Ashby, M. F., S. Bahk, J. Bevk, and D. Turnbull. The influence of a dispersion of particles on the sintering of metal powders and wires. *Progress in Materials Science*, Vol. 25, No. 1, 1980, pp. 1–34.
- [36] Liu, J. T., Y. W. Zhang, Y. Tao, and Y. Zhang. Microstructure of FGH97 P/M superalloy. *Transactions of Materials and Heat Treatment*, Vol. 32, No. 3, 2011, pp. 47–51.
- [37] Suryanarayana, C. Mechanical alloying: a novel technique to synthesize advanced materials. *Research*, Vol. 2019, 2019, id. 19.
- [38] Tang, K., W. J. Wang, H. H. Ding, L. Zhou, J. Guo, W. He, et al. Influence of laser shock peening on fatigue performance of LZ50 axle steel for railway wheel set. *Fatigue and Fracture of Engineering Materials and Structures*, Vol. 43, No. 9, 2020, pp. 1938–1948.
- [39] Niclas, A., A. Karasev, and P. G. Jonsson. The effect of different non-metallic inclusions on the machinability of steels. *Materials*, Vol. 8, No. 2, 2015, pp. 751–783.
- [40] Cervellon, A., S. Hemery, P. Kurnsteiner, B. Gault, P. Kontis, and J. Cormier. Crack initiation mechanisms during very high cycle fatigue of Ni-based single crystal superalloys at high temperature. *Acta Materialia*, Vol. 188, 2020, pp. 131–144.
- [41] Gu, C., J. Lian, Y. Bao, and S. Munstermann. Microstructure-based fatigue modelling with residual stresses: prediction of the microcrack initiation around inclusions. *Materials Science and Engineering: A*, Vol. 751, 2019, pp. 113–141.
- [42] Bergsmo, A. and F. P. E. Dunne. Competing mechanisms of particle fracture, decohesion and slip-driven fatigue crack nucleation in a PM nickel superalloy. *International Journal of Fatigue*, Vol. 135, 2020, id. 105573.
- [43] Cai, C., K. Pan, Q. Teng, X. Gao, B. Song, J. Liu, et al. Simultaneously enhanced strength and ductility of FGH4097 nickel-based alloy via a novel hot isostatic pressing strategy. *Materials Science and Engineering: A*, Vol. 760, 2019, pp. 19–25.