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Ambient-Atmosphere Ultrasonic Atomization of Low-Melting-Point Wood's Alloy Powders: Process Optimization, Atomization Mechanism, and Solderability

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Relevance Summary

Prepares Wood's alloy powder via ambient ultrasonic atomization, preventing compositional deviation and oxidation.

Optimizes thermal and amplitude parameters to achieve highly spherical, fine powders.

Elucidates a synergistic atomization mechanism driven by capillary waves and acoustic cavitation.

Achieves ~25 MPa joint strength, advancing near-room-temperature electronic packaging

Abstract:

Low-melting-point alloy powders are crucial for temperature-sensitive electronic packaging. However, preparing quaternary Wood's alloy powders containing volatile elements (Pb, Cd) via traditional atomization often causes compositional deviation and oxidation. Herein, 13.3Sn–50Bi–26.7Pb–10Cd (wt%) Wood's alloy powders were prepared using ambient-atmosphere ultrasonic atomization. The effects of ultrasonic amplitude, heating tank temperature, and vibration stage temperature on powder characteristics were systematically investigated. Furthermore, the atomization mechanism was studied using high-speed photography, and the solderability of the optimized powder on 304 stainless steel was evaluated.

Results indicate that increasing ultrasonic amplitude from 20% to 30% effectively regulates powder quality, decreasing the median diameter (D_{50}) from 77.78 μm to 71.07 μm , narrowing the size span, and improving sphericity. An optimal heating tank temperature of 90–150 °C prevented morphological

deterioration caused by severe melt oxidation, while a room-temperature vibration stage ensured high sphericity. Phase analysis confirmed the presence of β -Sn, Bi, Pb, Cd, and Pb_7Bi_3 intermetallics, with Pb_7Bi_3 acting as a solute sink for Bi to lower the melting temperature. High-speed imaging revealed that atomization is driven by the synergistic effect of Faraday capillary waves and acoustic cavitation, triggered by a critical liquid film thickness. Finally, joints soldered with the optimized powder exhibited an average shear strength of ~ 25 MPa, demonstrating excellent suitability for near-room-temperature electronic interconnections.

Keywords: Ultrasonic atomization; Wood's alloy; Pb_7Bi_3 ; Solderability

1. Introduction

The rapid development of miniaturized electronic systems has created an increasingly urgent demand for low-temperature interconnection technologies. Although binary Sn–Bi eutectic alloys possess low melting temperatures, their reliability issues are critical in medical diagnostic devices, aerospace electronics, and military packaging. In such demanding applications, solder joints must endure frequent thermal cycling induced by device switching while maintaining high strength and fatigue resistance throughout long-term service. Taking Sn–58Bi as an example, the Bi-rich phase undergoes preferred orientation and coarsening after thermal cycling, leading to stress concentration at the interface between the Sn-rich and Bi-rich phases, which consequently initiates fatigue microcracks [1]. Zuo et al. [2] discovered that under coupled electric current and thermal cycling, fatigue cracks preferentially nucleate and propagate along Sn/Bi phase boundaries, ultimately causing joint fracture at the interface between the intermetallic compound (IMC) and the solder. Belhadi et al. [3] confirmed that an increased Bi content accelerates the coarsening of interfacial IMC particles during thermal cycling, thereby exacerbating the creep deformation of solder joints. Overall, the coarsening behavior of the Bi phase under conditions such as thermal aging [4] and electromigration [5] restricts the application of binary Sn–Bi eutectic alloys in high-demand fields [6].

To address these challenges, the quaternary eutectic Wood's alloy (13.3Sn–50Bi–26.7Pb–10Cd, wt%) continues to receive attention in these high-reliability applications. This alloy features a eutectic temperature of only $\sim 70^\circ\text{C}$, which is significantly lower than that of conventional Sn–Bi eutectic solders. Although Pb and Cd are hazardous heavy metals subject to Restriction of Hazardous Substances (RoHS) regulations, low-melting-point alloys containing Pb and Cd are still permitted in high-reliability electronic packaging due to their irreplaceable advantages [7, 8].

Upon entering the alloy, Pb preferentially undergoes a peritectic reaction with Bi to form the Pb_7Bi_3 intermetallic phase. During solidification, this phase acts as a solute sink for high-melting-point Bi, consuming excess Bi in the matrix, indirectly lowering the melting temperature by tailoring the solidification path, and enhancing the microstructural homogeneity of the solder [9, 10]. In the molten state, Cd exhibits high surface activity and preferentially migrates to the melt/substrate interface, enhancing the wetting and spreading ability of the solder on common substrates such as steel and copper, thereby ensuring joint quality [8, 11].

To convert bulk alloys into powders that meet the requirements of low-temperature soldering, the

choice of powder fabrication method is critical. The vacuum environment in vacuum induction melting gas atomization (VIGA) causes the volatilization of Cd due to its high vapor pressure, leading to compositional deviation. Electrode induction melting gas atomization (EIGA) requires shaping the feedstock into consumable electrodes; however, the low solidus temperature (~ 70 °C) of Wood's alloy makes it prone to softening and premature melting/breaking [12]. Water atomization (WA) disintegrates the melt using high-pressure water jets, inducing severe oxidation of Sn, Pb, and Cd, while producing irregular particle morphologies with poor flowability, which is unfavorable for soldering applications [13]. In contrast, ultrasonic atomization (UA) requires no vacuum, consumable electrodes, or high-velocity fluids. It utilizes high-frequency mechanical vibrations to excite capillary waves and acoustic cavitation within the liquid film to achieve fragmentation, demonstrating remarkable cost-effectiveness and flexibility for small-batch production of specialty alloys [14, 15].

However, current academic research on UA mostly focuses on pure metals or relatively simple alloy systems, such as pure aluminum [16] and 316L stainless steel [17]. Open literature remains very limited regarding the process parameter–powder property relationships for multi-element quaternary low-melting alloys with large physical property disparities under ambient ultrasonic atomization, as well as the dynamic mechanisms of liquid film spreading, thinning, and fragmentation.

To bridge this gap, this study prepared 13.3Sn–50Bi–26.7Pb–10Cd (wt%) Wood's alloy powders via ambient-atmosphere ultrasonic atomization. We systematically investigated how specific process conditions, including ultrasonic amplitude, melt heating temperature, and vibration stage temperature, affect powder characteristics such as particle size, morphology, and sphericity. Furthermore, high-speed photography was employed to observe the spreading and instability behaviors of the liquid film on the vibration stage. The optimized powder was subsequently utilized as a filler metal to join 304 stainless steel, and the resulting shear strength and fracture morphology features were evaluated.

2. Experimental Materials and Methods

2.1. Feedstock preparation and utilization

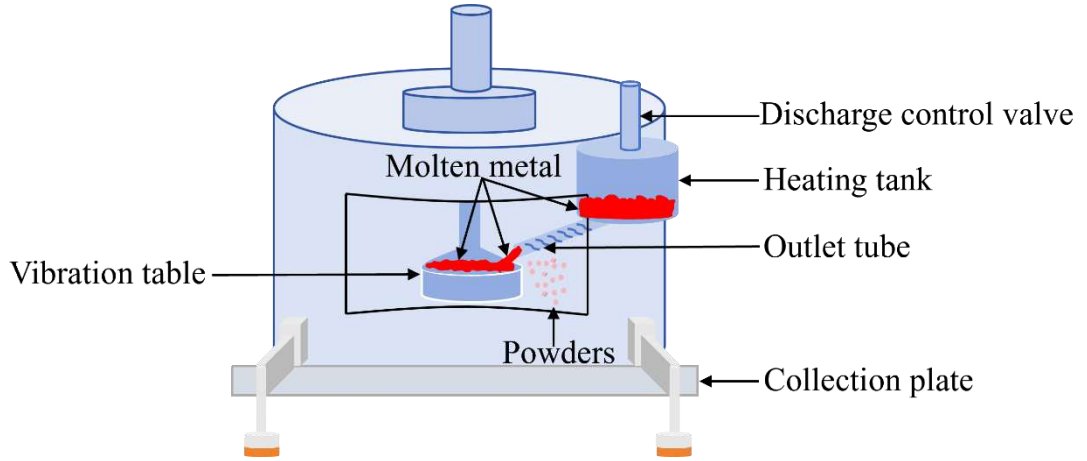


Fig. 1. Schematic diagram of the ultrasonic atomization device.

The feedstock used was Wood's alloy with a nominal composition of 13.3Sn–50Bi–26.7Pb–10Cd (wt%) and a melting point of 70 °C. The ultrasonic atomization was performed using an MSK-UP-300 atomization system, as schematically illustrated in Fig. 1. The alloy was melted in a heating tank to a predefined temperature via an induction coil. Once melted, the liquid metal was continuously delivered onto the top surface of the ultrasonic vibration stage under gravity, controlled by a flow valve and a guide tube. The atomization process was conducted in an ambient air atmosphere. Upon completion and subsequent cooling of the equipment to room temperature, the resultant powder was carefully swept from the collection platform using a soft brush and subsequently sealed in a vacuum sealer for preservation. We systematically investigated how specific process conditions, including ultrasonic amplitudes from 20% to 30%, melt heating temperatures between 90 °C and 180 °C, and vibration stage temperatures ranging from 28 °C to 120 °C, affect key powder characteristics such as particle size, morphology, and sphericity. The powder produced under optimized conditions was further utilized as a filler metal to solder 304 stainless steel at 172 °C.

2.2. Powder and joint characterization

The morphological and microstructural features of the powders were characterized using a high-resolution field-emission scanning electron microscope (FE-SEM, CLARA GMH). The particle size distribution was measured employing a laser particle size analyzer (Bettersize 2600). Morphological indices, including sphericity, satellite index, and bluntness, were quantified using a dynamic image analyzer (R-3000 Powder Size and Shape Analyzer System). The apparent density and tap density of the powders were evaluated utilizing a multi-functional powder physical property tester (HYL-1001). The flowability of the powders was assessed using a standard Hall flowmeter. The dynamic behavior of the

ultrasonic atomization process was captured using a high-speed camera (VEO610L). Finally, the shear strength of the soldered joints was evaluated using a universal testing machine (WDW-50A, China) equipped with a customized shear fixture.

3. Results and discussion

3.1 Effect of ultrasonic amplitude on powder properties

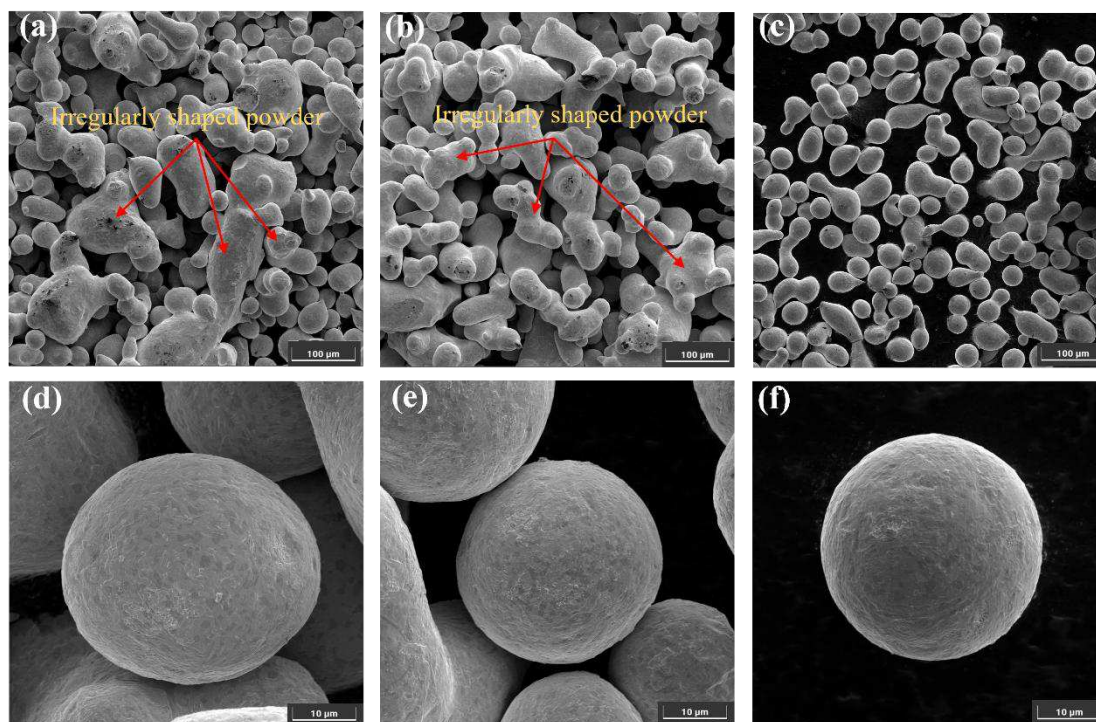


Fig. 2. SEM images of powder at different amplitudes: (a) and (d) 20%, (b) and (e) 25%, (c) and (f) 30%.

Fig. 2 presents the SEM images of the powders produced at ultrasonic amplitudes of 20%, 25%, and 30%, under a constant heating tank temperature of 120 °C and a vibration stage temperature of 28 °C. At an amplitude of 20%, a high proportion of irregular flake-like and agglomerated particles was observed. Many droplets solidified before adequate spheroidization, resulting in irregularly shaped powders with rough surfaces, as shown in Fig. 2(d). As revealed by Fig. 2(b) and (e), increasing the amplitude to 25% led to a higher number of spherical particles; however, satellite droplets were still observed adhering to the surfaces of some particles. Furthermore, Fig. 2(c) and (f) demonstrate that when the amplitude was further increased to 30%, the proportion of near-spherical particles increased significantly. Concurrently, irregular and satellite powders decreased markedly, leading to smoother particle contours. This trend indicates that increasing the ultrasonic amplitude promotes the uniformity

of melt fragmentation and enhances droplet spheroidization.

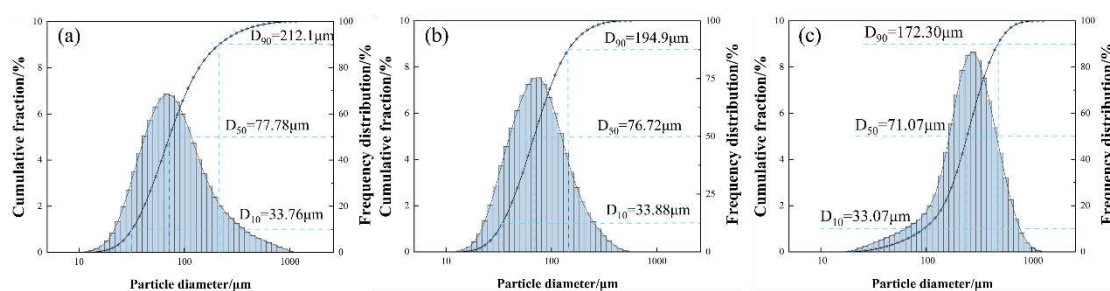


Fig. 3. Particle size distributions of the powders at different ultrasonic amplitudes: (a) 20%; (b) 25%; (c) 30%.

The particle size distribution (PSD) results were consistent with the morphological observations. Fig. 3 and Table 1 summarize the D_{10} , D_{50} , D_{90} , and particle size span (S) at different amplitudes. The span S is defined as follows [18]: $S = \frac{D_{90}-D_{10}}{D_{50}}$. At 20% amplitude, the D_{50} was 77.78 μm , the D_{90} reached a high of 212.1 μm , and the span S was 2.29. When the amplitude was 25%, the D_{50} decreased slightly to 76.72 μm , the D_{90} dropped to 194.9 μm , and the span S was reduced to 2.10. At 30% amplitude, the D_{50} further decreased to 71.07 μm , the D_{90} decreased to 172.30 μm , and the span S narrowed to 1.96. The D_{10} values showed minimal variation across the three sets of data. This indicates that the suppressive effect of amplitude on coarse powders is more pronounced. As the amplitude increases, the sizes of medium and large particles decrease, leading to an overall refinement of the powder and a more concentrated particle size distribution.

Table 1. PSD parameters of powders at different amplitudes.

Ultrasonic amplitude/%	$D_{10}/\mu\text{m}$	$D_{50}/\mu\text{m}$	$D_{90}/\mu\text{m}$	S
20	33.76	77.78	212.10	2.29
25	33.88	76.72	194.90	2.10
30	33.07	71.07	172.30	1.96

Similar results were obtained for the particle shape parameters. As shown in Fig. 4(a), when the amplitude increased from 20% to 30%, the sphericity increased while the satellite index decreased. Specifically, the highest sphericity and the lowest satellite index were achieved at an amplitude of 30%.

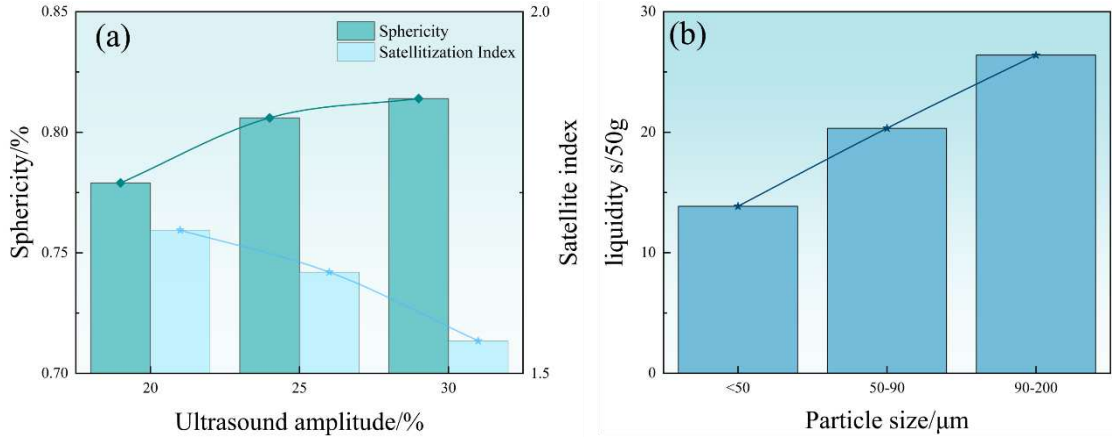
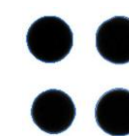


Fig. 4. (a) Sphericity and satellite index of the powders at different amplitudes; (b) Flowability and density characteristics of the powder produced at 30% amplitude.

Furthermore, Fig. 4(b) demonstrates that the powder produced at 30% amplitude exhibited excellent Hall flowability, with an apparent density of 4.505 g/cm^3 and a tap density of 5.519 g/cm^3 . To evaluate the powder characteristics more comprehensively, the bluntness index was introduced [19]. A bluntness value of 100% corresponds to a perfectly smooth circle, whereas lower values, particularly those below 50%, indicate highly irregular and rough particle profiles. The typical 2D projections of powder particles with varying bluntness values are illustrated in Table 2.

Table 2. Typical 2D projections of powder particles exhibiting different bluntness values.

Bluntness/%	<50	50-70	70-90	90-100
Powder Projection				

The effect of ultrasonic amplitude on powder bluntness is presented in Fig. 5. Fig. 5(a) indicates that the majority of the powder particles exhibit a bluntness value greater than 50%. Furthermore, Fig. 5(b) illustrates that an increase in ultrasonic amplitude leads to higher powder bluntness. Specifically, as the amplitude increases from 20% to 30%, the average bluntness of the powder rises from 60.5% to 66.4%. Combined with the visual references in Table 2, this implies that the contour features of the powder particles transitioned from relatively irregular to more regular shapes, generally approaching an ideal sphere. This morphological evolution is consistent with the observed trends of increasing sphericity and decreasing satellite index.

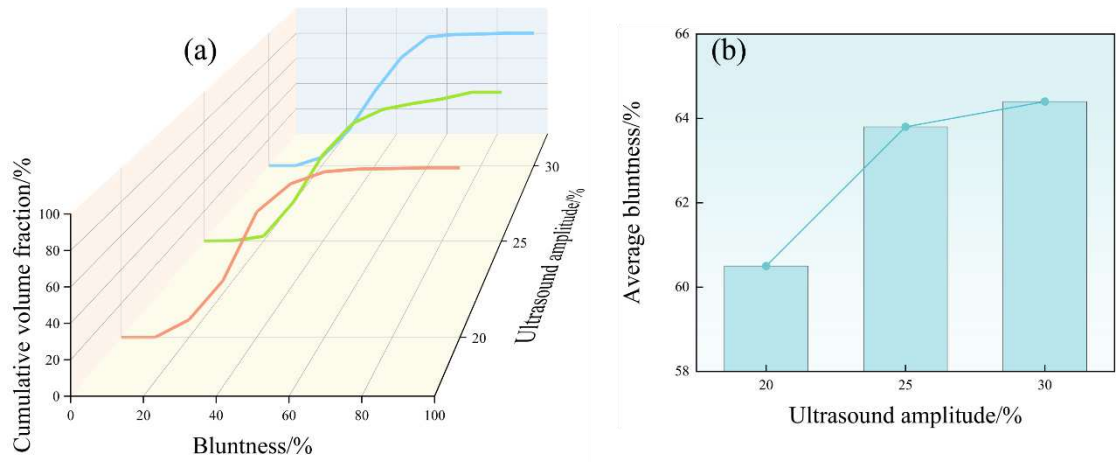


Fig. 5. Effect of ultrasonic amplitude on powder bluntness: (a) Cumulative volume fraction of bluntness; (b) Average bluntness vs. ultrasonic amplitude.

The influence of amplitude on powder size and morphology can be attributed to the enhanced input of ultrasonic energy. According to the droplet size prediction model proposed by Rajan et al. [20]: $d_p = \left(\frac{\pi\sigma}{\rho f^2}\right)^{0.33} [1 + A(We)^{0.22}(Oh)^{0.166}(I_N)^{-0.0277}]$, where σ is surface tension (dyne/cm), ρ is density (g/ml), f is excitation frequency (Hz), and d_p is the droplet size (cm), We is the Weber number, Oh is the Ohnesorge number, I_N is intensity number. It can be deduced that, under the condition of relatively constant liquid surface tension, density, and excitation frequency, the powder particle size correlates negatively with the amplitude. Increasing the amplitude enhances the acoustic energy density absorbed per unit volume of the liquid film. Consequently, the amplitude of the capillary waves is amplified, and smaller liquid droplets are generated after the wave crests are sheared off. Simultaneously, more intense acoustic cavitation activity promotes more violent collapse microjets from the bubbles within the liquid film. This facilitates the tearing of the liquid film and suppresses the formation of large droplets, thereby narrowing the particle size distribution [16, 21]. Therefore, ultrasonic amplitude is a highly effective process parameter for regulating powder quality.

3.2. Effect of heating tank temperature on powder characteristics

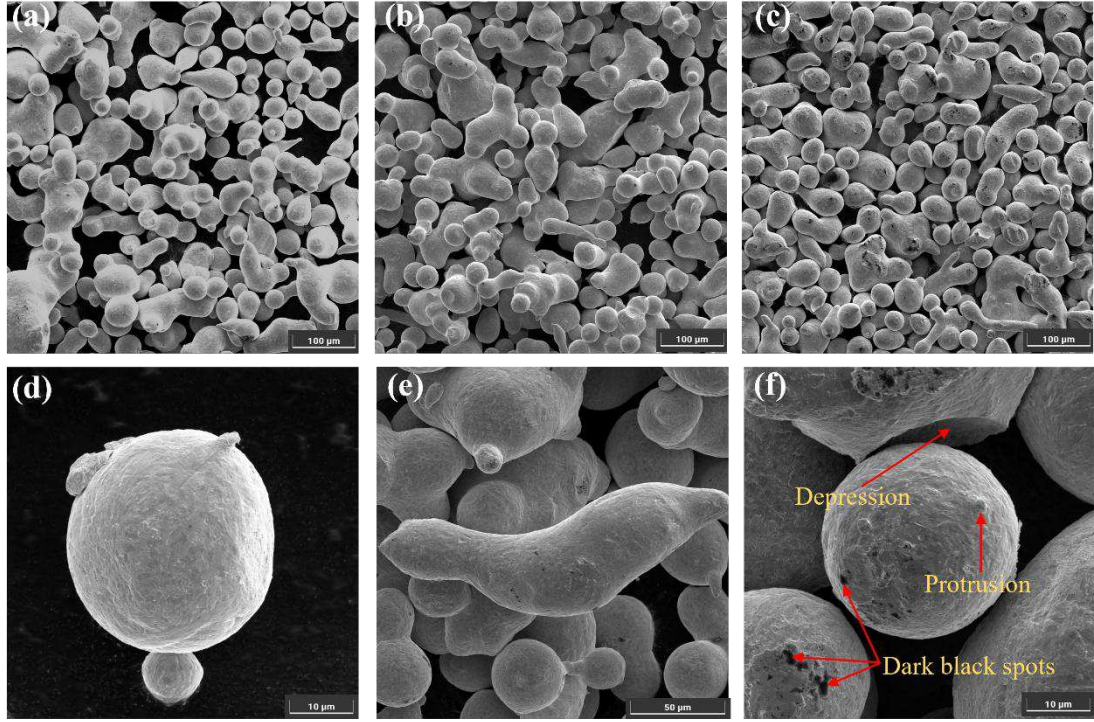


Fig. 7. Particle size distributions of the powders at different heating tank temperatures: (a) 90 °C; (b) 150 °C; (c) 180 °C.

Fig. 6 presents the SEM images of the powders produced at heating tank temperatures of 90 °C, 150 °C, and 180 °C, under a constant ultrasonic amplitude of 30% and a vibration stage temperature of 28 °C. As shown in Fig. 6(a) and (d), when the heating tank temperature was set to 90 °C, the proportion of spherical particles reached its maximum. These particles exhibited relatively smooth surfaces with a minimal number of satellite powders. As the temperature increased to 150 °C, spherical particles remained dominant, although a small number of irregularly shaped particles and satellite droplets began to emerge, as depicted in Fig. 6(b) and (e). Further elevating the temperature to 180 °C significantly deteriorated the powder morphology, evidenced by a marked increase in irregular particles and satellite powders, as revealed in Fig. 6(c) and (f). Furthermore, dark spots were observed on some particle surfaces, indicating the presence of surface pores or oxidation defects. The satellite powders presented in Fig. 6(d) were formed by the in-flight collision of droplets with varying sizes and velocities [22]. Conversely, the irregular particle in the center of Fig. 6(e) resulted from the incomplete spheroidization of a droplet that underwent only shape deformation during fragmentation [23]. Additionally, the dark spots on the powder surface in Fig. 6(f) denote the existence of surface pores, which were formed by the shrinkage of liquid metal during the final stages of solidification [24].

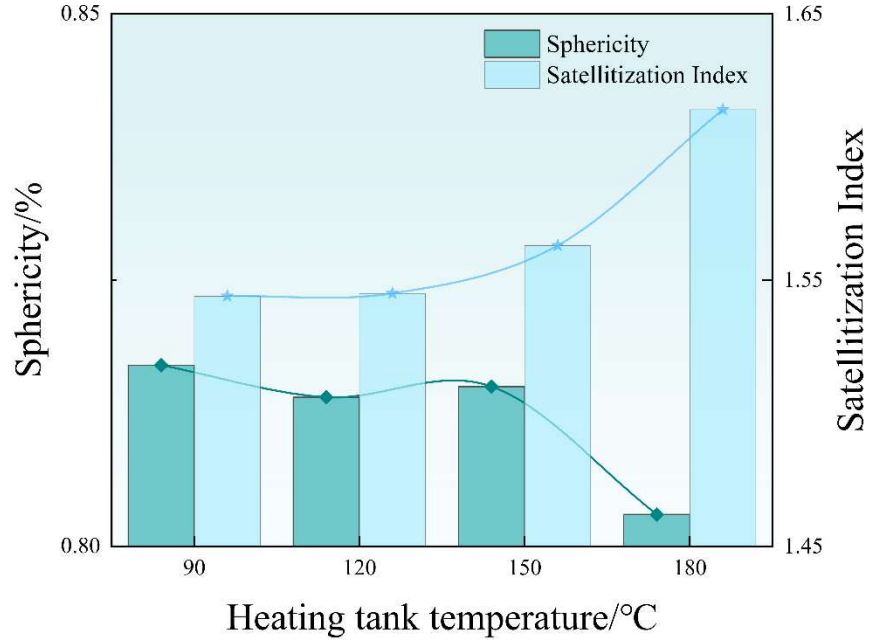


Fig. 8. Sphericity and satellite index of the powders at varying heating tank temperatures.

Incorporating the data at 120 °C from Fig. 3(c), the median particle size (D_{50}) across the four temperature conditions follows an ascending order: 120 °C (71.07 μm) < 90 °C (72.54 μm) < 150 °C (74.26 μm) < 180 °C (76.72 μm). The variation in the span S across the four temperatures was relatively minor, ranging from 1.82 to 1.96. Specifically, the minimum S (1.82) occurred at 150 °C, while the maximum S (1.96) was observed at 120 °C. This marginal difference of approximately 7.14% indicates that the heating tank temperature exerts a weaker influence on the width of the particle size distribution compared to the ultrasonic amplitude.

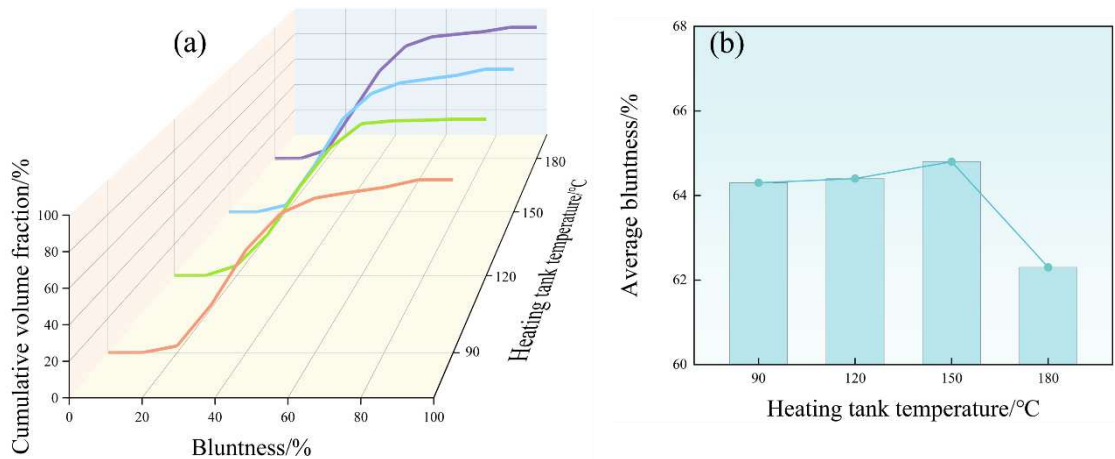


Fig. 9. Effect of heating tank temperature on powder bluntness: (a) Cumulative volume fraction of bluntness; (b) Average bluntness vs. heating tank temperature.

The analysis results for sphericity and satellite index are presented in Fig. 8. The powders produced

at 90 °C, 120 °C, and 150 °C exhibited similar values for sphericity and satellite index, with the 90 °C condition yielding the highest sphericity and the lowest satellite index. However, when the heating tank temperature was elevated to 180 °C, the sphericity plummeted to its lowest point, accompanied by a surge in the satellite index to its peak. This trend is highly consistent with the variation in powder bluntness demonstrated in Fig. 9 and particle size distribution demonstrated in Table 3. The average bluntness across the 90 °C, 120 °C, and 150 °C groups remained approximately 64.5%, whereas it dropped to 62.3% at 180 °C. Additionally, Fig. 9(a) reveals that the vast majority of the powder particles possessed a bluntness value exceeding 50%.

Table 3. particle size distribution parameters of powders at different heating tank temperatures.

Heating tank temperature/°C	D ₁₀ /μm	D ₅₀ /μm	D ₉₀ /μm	S
90	33.79	72.54	171.20	1.89
120	33.07	71.07	172.30	1.96
150	34.81	74.26	169.70	1.82
180	35.48	76.72	180.00	1.88

The aforementioned results can be elucidated by the influence of melt superheating on the droplet solidification behavior. The Wood's alloy utilized in this study has a melting point of approximately 70 °C; hence, all investigated heating tank temperatures provided a certain degree of superheating. Within an appropriate superheating range (e.g., 90–150 °C), the solidification time of the melt exceeds the structural relaxation time required for spheroidization. Driven by surface tension, the droplets have ample time to contract into spherical shapes prior to complete solidification, thereby yielding powders with high sphericity, a low satellite index, and elevated bluntness [25]. Conversely, excessively high heating tank temperatures drastically accelerate the oxidation rate of the melt in air. The resulting oxide film on the surface impedes the spheroidization process of the droplets, consequently diminishing the sphericity and bluntness while elevating the satellite index. Therefore, an optimal processing window exists for the heating tank temperature.

3.3. Effect of vibration stage temperature on powder characteristics

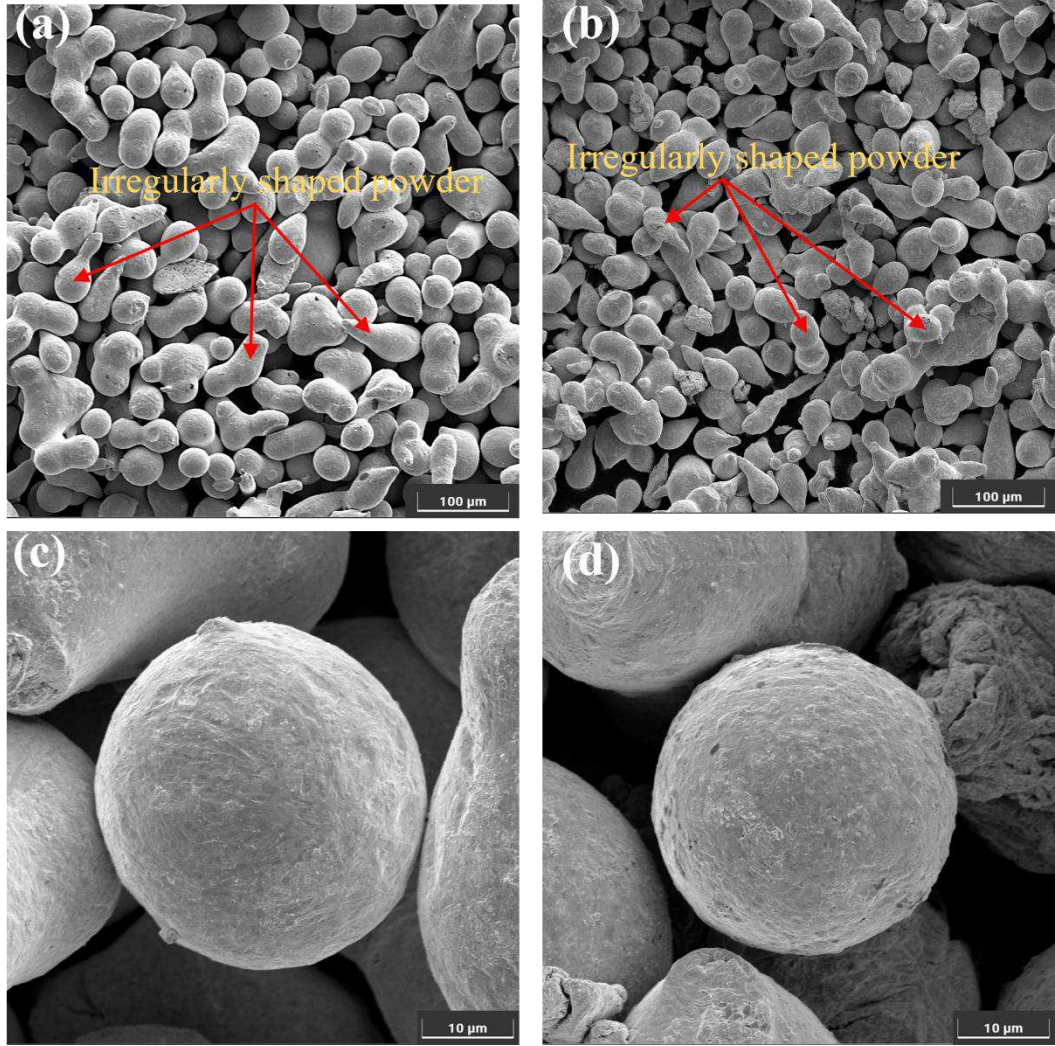


Fig. 10. SEM images of the powders produced at varying vibration stage temperatures: (a) and (c) 70 °C; (b) and (d) 120 °C.

Fig. 10 displays the SEM images of the powders produced at vibration stage temperatures of 70 °C and 120 °C, under a constant ultrasonic amplitude of 30% and a heating tank temperature of 120 °C. By comparing these with the room-temperature results presented in Fig. 2(c) and (f), it is evident that the optimal sphericity is achieved at a vibration stage temperature of 28 °C, where near-spherical particles dominate and the numbers of irregular particles and satellite powders are minimized. As the vibration stage temperature increases to 70 °C, Fig. 10(a) and (c) reveal a decline in the proportion of spherical particles and the emergence of more irregularly shaped particles. Upon further heating to 120 °C, the powder morphology severely deteriorates, as demonstrated in Fig. 10(b) and (d), marked by the presence of abundant irregular particles and the lowest sphericity.

This morphological evolution can be attributed to the competing mechanisms of undercooling and

secondary heating. When the vibration stage is at room temperature, the melt film acquires a certain degree of undercooling upon contacting the cooler surface, which promotes the rapid solidification of the droplets. In contrast, at a vibration stage temperature of 70 °C (comparable to the alloy's melting point), the liquid film experiences neither significant undercooling nor effective secondary heating. When the temperature is elevated to 120 °C, the high-temperature stage provides continuous heat input to the spreading liquid film, effectively acting as a secondary heating source. This process exacerbates melt oxidation and forms an oxide film that impedes surface-tension-driven spheroidization, ultimately leading to reduced sphericity and an increased satellite index.

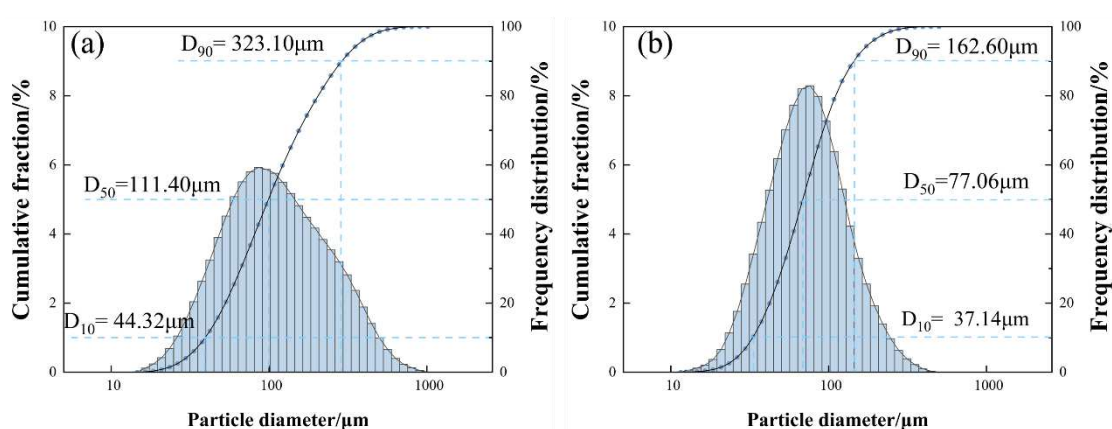


Fig. 11. Particle size distributions of the Sn-Bi powders at varying vibration stage temperatures: (a) 70 °C; (b) 120 °C.

Fig. 11 and Table 4 summarize the values of D_{10} , D_{50} , D_{90} , and the particle size span (S) obtained at different vibration stage temperatures. Incorporating the 28 °C data from Fig. 3(c), the median particle size (D_{50}) under the three temperature conditions follows an ascending order: 28 °C (71.07 μm) < 120 °C (77.06 μm) < 70 °C (111.40 μm). Notably, at 70 °C, the D_{50} anomalously surges to 111.40 μm, the D_{90} reaches a staggering 323.10 μm, and the span S peaks at 2.50, representing the maximum values among all experimental groups. Conversely, at 120 °C, the D_{50} drops to 77.06 μm, and the span S narrows to 1.63, marking the minimum value across all groups. The substantial variation in the span S among the three temperatures presents a relative difference of approximately 35.48%, which far exceeds the 7.14% variation observed for the heating tank. This significant difference demonstrates that the vibration stage temperature exerts a more profound impact on the width of the particle size distribution than the heating tank temperature.

Table 4. Particle size distribution parameters of the powders produced at different vibration

stage temperatures.

vibrating table temperature/°C	D ₁₀ /μm	D ₅₀ /μm	D ₉₀ /μm	S
28	33.07	71.07	172.30	1.96
70	44.32	111.40	323.10	2.50
120	37.14	77.06	162.60	1.63

The analytical results for sphericity and satellite index are presented in Fig. 12. At a vibration stage temperature of 28 °C, the sphericity reached its maximum while the satellite index was minimized. As the temperature of the stage increased, the sphericity gradually declined, accompanied by a progressive rise in the satellite index. The trend in powder bluntness, as illustrated in Fig. 13, aligns completely with these observations. The average bluntness peaked at 64.4% at 28 °C, decreased to 62.1% at 70 °C, and further dropped to 56.8% at 120 °C. Additionally, Fig. 13(a) reveals that the majority of the powder particles possess bluntness values concentrated around 50%, which corroborates the SEM observations of abundant irregular particles formed at this elevated temperature.

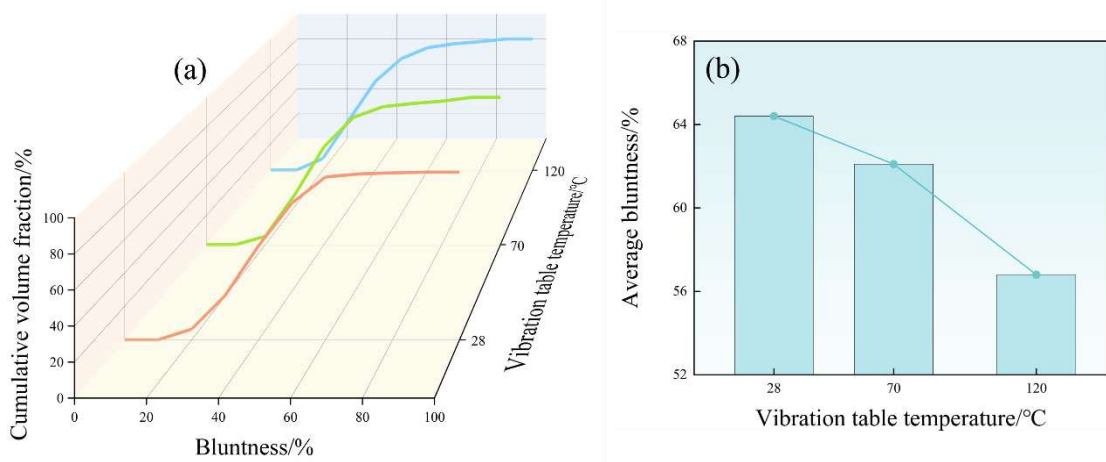


Fig. 13. Effect of vibration stage temperature on powder bluntness: (a) Cumulative volume fraction of bluntness; (b) Average bluntness vs. vibration stage temperature.

In summary, the overarching influence of the vibration stage temperature on powder characteristics can be attributed to the competing mechanisms of undercooling and secondary heating. The moderate undercooling provided by the room-temperature stage facilitates rapid droplet solidification, promoting the formation of high-quality powders. Conversely, when the stage temperature approaches or exceeds the alloy's melting point, the undercooling effect diminishes, and secondary heating may even dominate, thereby exacerbating oxidation, hindering spheroidization, and ultimately deteriorating powder

morphology.

3.4. Phase composition and microstructure

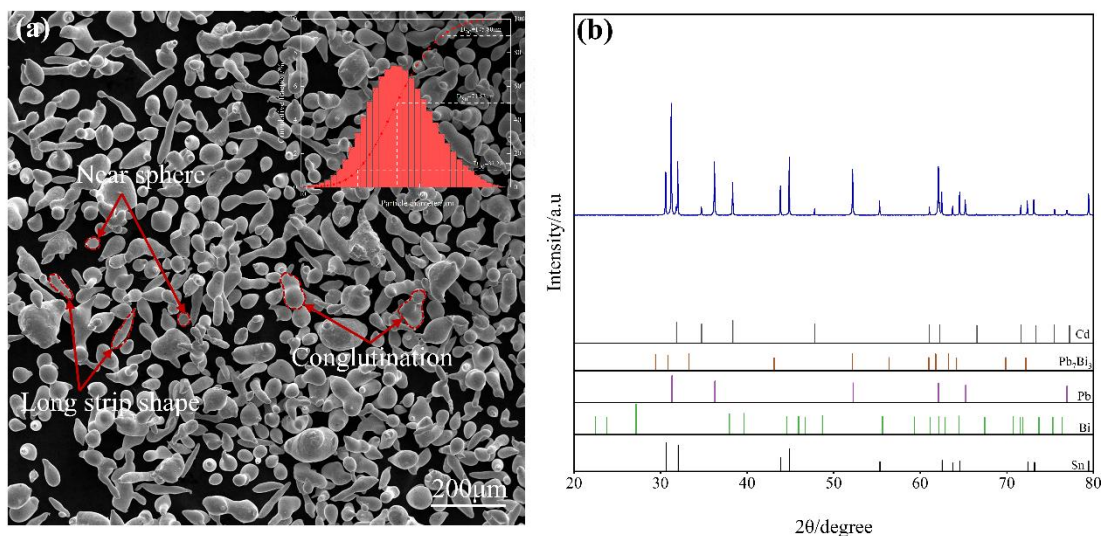


Fig. 14. (a) Powder morphology and particle size, (b) X-ray diffraction

Fig. 14(a) displays the SEM morphology and particle size distribution of the powder produced under the optimal process parameters (ultrasonic amplitude of 30%, heating tank temperature of 120 °C, and vibration stage temperature of 28 °C). The powder overall exhibits polydisperse characteristics, comprising predominantly near-spherical particles alongside a minor fraction of agglomerates and individual elongated particles, with an average sphericity of 0.704 and a D_{50} of 71.8 μm. The corresponding XRD pattern of the powder is presented in Fig. 14(b). The indexing of the diffraction peaks reveals the coexistence of β-Sn (body-centered tetragonal), Bi (rhombohedral), Pb (face-centered cubic), Cd (hexagonal), and the intermetallic compound Pb_7Bi_3 phase. Notably, no binary Sn–Bi intermetallic compounds were detected. This absence is attributed to the nature of the Sn–Bi eutectic system, wherein Bi precipitates from the β-Sn matrix as an elemental phase rather than forming intermetallic compounds with Sn [26]. Given the rapid solidification characteristic of the ultrasonic atomization process, the Pb_7Bi_3 phase acts as a solute sink for the high-melting-point Bi element during solidification. It consumes excess Bi in the matrix within localized regions, thereby indirectly lowering the melting temperature of the alloy by modulating the solidification path [9, 10].

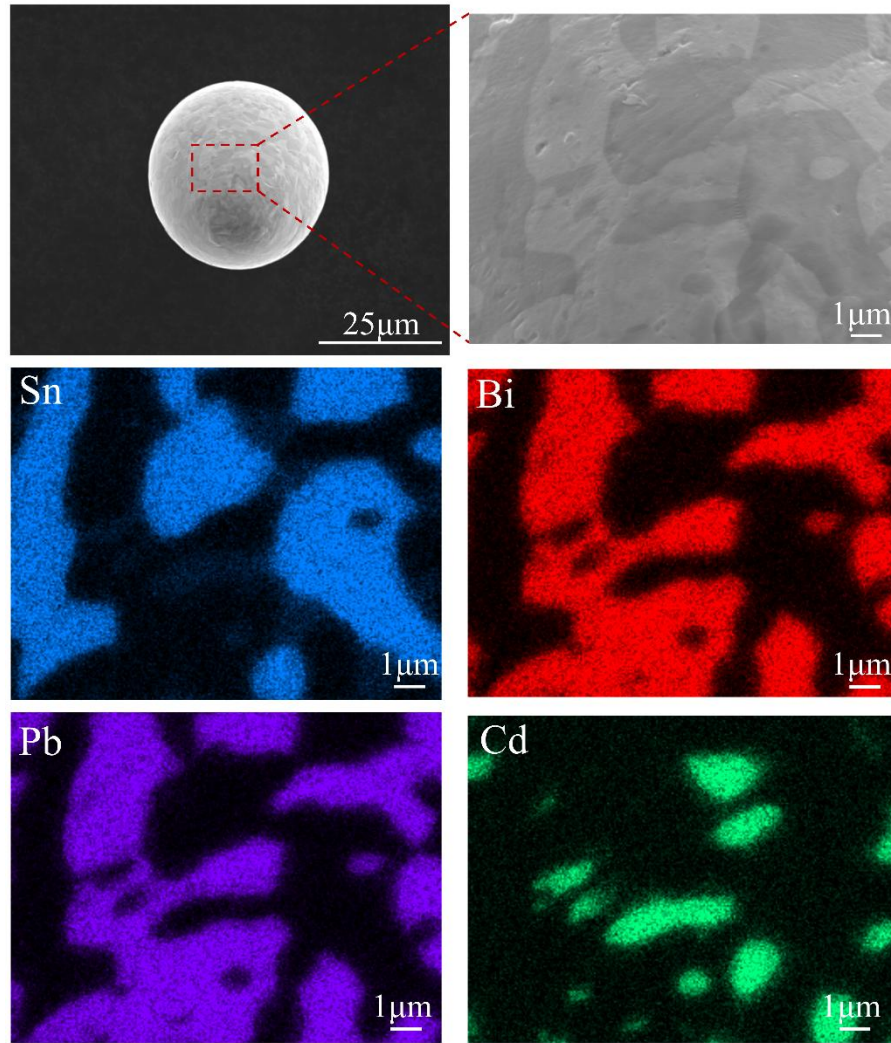


Fig. 15. Alloy powder characterized by SEM and EDS

Fig. 15 illustrates the backscattered electron (BSE) image of the powder particles and the corresponding EDS elemental mapping results. Two distinct contrast regions, namely light gray and dark gray areas, can be clearly distinguished in the BSE image. The EDS mapping reveals that Pb and Bi elements exhibit significant co-segregation within the light gray regions, thereby corroborating the formation of the Pb_7Bi_3 phase. Conversely, the Sn and Cd elements are predominantly distributed across the dark gray matrix. This elemental distribution characteristic indicates the formation of a typical phase-separated microstructure within the powder: the Pb_7Bi_3 intermetallic compounds are embedded in the Sn-rich and Cd-rich matrix in the form of discrete islands and network-like morphologies.

3.5. Shear strength and fracture morphology of soldered joints

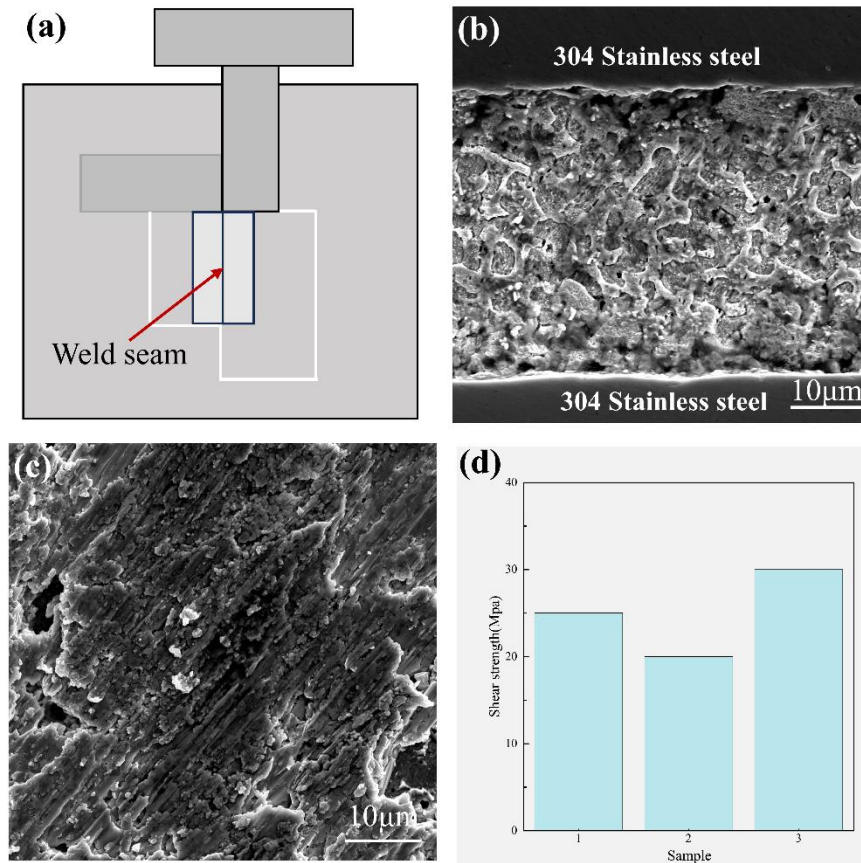


Fig. 16. Mechanical evaluation of the soldered joints: (a) Schematic of the customized shear fixture; SEM micrographs of (b) the cross-section of the soldered joint and (c) the fracture surface; (d) Shear strength measurement results.

A schematic illustration of the customized shear fixture used for the test is provided in Fig. 16(a). Fig. 16(b) displays the cross-sectional SEM morphology of the soldered joint. It is evident that a metallurgical bond was successfully formed between the solder layer and the 304 stainless steel substrates on both sides, with no obvious macroscopic pores or un-bonded defects observed. The shear strength measurement results for multiple specimens are summarized in Fig. 16(d), yielding an average shear strength of approximately 25 MPa. This value demonstrates the promising application potential of the powder for demanding, low-stress electronic interconnection scenarios operating near room temperature [27, 28]. The SEM morphology of the shear fracture surface is depicted in Fig. 16(c). The fracture surface is predominantly characterized by flat cleavage facets and mutually parallel river patterns, which are the typical microstructural features of brittle transgranular cleavage fracture. Simultaneously, a small number of micro-dimples can be observed in localized regions of the fracture surface, indicating

the occurrence of ductile tearing during the fracture process. This mixed fracture mode can be attributed to the presence of the hard and brittle Pb_7Bi_3 intermetallic phase within the solder seam. Under shear loading, the rigid Pb_7Bi_3 phase acts as a stress concentration point and a crack initiation site. Once initiated, the brittle cracks rapidly propagate along the cleavage planes, forming flat cleavage facets and river patterns. Conversely, the relatively softer $\beta\text{-Sn}$ matrix phase absorbs a portion of the strain energy through the nucleation, growth, and coalescence of micro-voids. This process leaves micro-dimple features on the fracture surface and, to a certain extent, delays the complete failure of the joint.

3.6. Analysis of the ultrasonic atomization mechanism

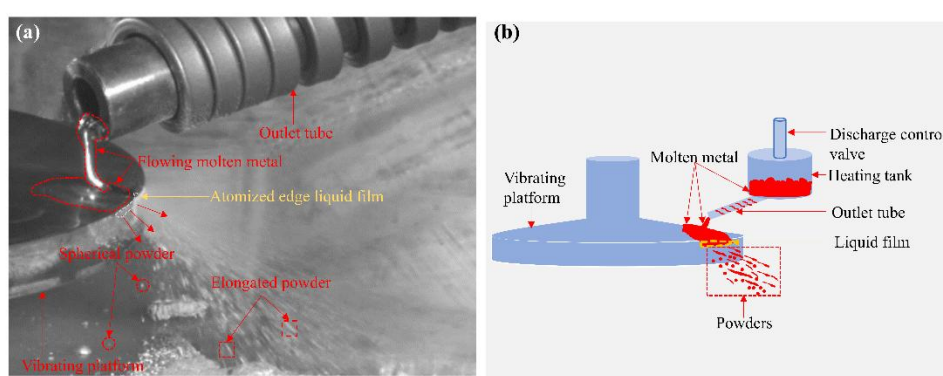


Fig. 17. Ultrasonic atomization mechanism: (a) High-speed photographic image of the atomization process; (b) Schematic illustration of the corresponding physical mechanism.

Fig. 17(a) captures the high-speed photographic image of the Wood's alloy melt undergoing atomization on the ultrasonic vibration stage, while the corresponding schematic of this process is illustrated in Fig. 17(b). High-speed imaging revealed that the atomization was delayed until the molten metal spread to the periphery of the ultrasonic horn. At the initial central contact zone, the liquid film was excessively thick to allow adequate acoustic energy coupling. As gravity-driven flow induced film thinning at the edges, the ultrasonic energy became concentrated within a diminishing volume, thereby triggering rapid atomization. Two distinct droplet morphologies, namely near-spherical droplets and filamentary or ligament-shaped droplets, were identified. These morphologies are characteristic of the dual mechanisms governing ultrasonic atomization. The near-spherical droplets are primarily generated by the Faraday capillary wave mechanism [29]. The high-frequency vibration of 35 kHz excites capillary waves on the surface of the liquid film. When the inertial force at the wave crests exceeds the confinement of surface tension, the wave crests are pinched off to form near-spherical droplets. The characteristic wavelength of these capillary waves is determined by the Kelvin dispersion relation [30], which depends

on the surface tension, liquid density, and ultrasonic frequency. Conversely, the filamentary droplets originate from the acoustic cavitation mechanism. Cavitation nuclei within the liquid film grow during the negative pressure phase of the ultrasonic field and collapse during the subsequent positive pressure phase. The localized high pressure and microjets released during this collapse tear the liquid film into elongated filaments [14, 21]. The simultaneous coexistence of these two droplet morphologies in this experiment aligns with the predictions of the cavitation–capillary wave hypothesis [31, 32]. The collapse of cavitation bubbles generates intense hydraulic shocks within the liquid film. As these shock waves propagate to the film surface, they amplify the instability of the Faraday capillary waves, subsequently leading to the fragmentation of the liquid film and the ejection of droplets.

Furthermore, it was observed during the experiment that a portion of the melt solidified directly on the vibration stage without undergoing atomization. This portion of the liquid flow failed to reach the edge of the vibration stage to form a sufficiently thin liquid film due to excessive viscosity or premature flow cessation. In this scenario, the hydraulic shock waves generated by the bubble collapse were attenuated by high viscous dissipation, rendering them unable to drive the instability of the surface capillary waves. This phenomenon highlights the controlling effect of the liquid film thickness on the atomization process. Effective atomization can only be achieved when the liquid film is sufficiently thin and the acoustic energy coupling efficiency is adequately high. Under these conditions, the synergistic effect of capillary waves and cavitation can overcome the barriers of viscous dissipation and surface tension.

4. Conclusions

In this study, low-melting-point Wood's alloy powders with a nominal composition of 13.3Sn–50Bi–26.7Pb–10Cd (wt%) were successfully prepared from bulk alloys via ambient-atmosphere ultrasonic atomization. We systematically investigated the effects of process parameters on powder characteristics, elucidated the atomization mechanism synergistically driven by capillary waves and acoustic cavitation, and evaluated the applicability of the powders for low-temperature interconnections. The main conclusions are drawn as follows:

(1) Ultrasonic amplitude serves as a highly effective process parameter for regulating powder quality. As the amplitude increases from 20% to 30%, the median particle size (D_{50}) decreases from 77.78 μm to 71.07 μm , the size span (S) narrows from 2.29 to 1.96, the sphericity increases monotonically, and the

average bluntness improves from 60.5% to 66.4%. Regarding the thermal conditions, an optimal processing window of 90–150 °C exists for the heating tank temperature, beyond which excessive temperatures exacerbate oxidation and deteriorate the particle morphology. Furthermore, maintaining the vibration stage at room temperature is most conducive to obtaining highly spherical powders. Elevated stage temperatures diminish the undercooling effect and even trigger secondary heating, thereby degrading the overall powder quality.

(2) The synthesized powder consists of β -Sn, Bi, Pb, Cd, and the intermetallic compound Pb_7Bi_3 . During the solidification process, the Pb_7Bi_3 phase acts as a solute sink for the Bi element, consuming excess Bi in the matrix and indirectly lowering the effective melting temperature of the alloy.

(3) The atomization process is synergistically driven by Faraday capillary waves and acoustic cavitation, with the liquid film thickness acting as the critical prerequisite for triggering atomization. Only when the liquid film is sufficiently thin and the acoustic energy coupling efficiency is adequately high can the synergistic effect of capillary waves and cavitation overcome the barriers of viscous dissipation and surface tension to accomplish effective atomization.

(4) Joints soldered on 304 stainless steel at 172 °C utilizing the optimized powder exhibited an average shear strength of approximately 25 MPa. The fracture surface displayed a mixed fracture mode, predominantly characterized by brittle transgranular cleavage and accompanied by localized micro-dimples. This strength level thoroughly fulfills the application requirements for near-room-temperature and low-stress electronic interconnections.

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Statements & Declarations

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Competing interests

The authors have no relevant financial or non-financial interests to disclose.

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Hao Lan: Writing – review & editing, Conceptualization. **Qian Wang:** Writing – review & editing, Supervision, Project administration, Funding acquisition, Conceptualization. **Peng Yu:** Writing – review & editing, Supervision, Conceptualization.

Data availability

The datasets generated during and/or analysed during the current study are available from the corresponding author on reasonable request.

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Consent to participate: Not applicable

Consent to publish: Not applicable.