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Xiaofeng Yang

University of Science and Technology of China

Haodi Li

University of Science and Technology of China

Dingying Ren

University of Science and Technology of China

Hao Lan

Ganjiang Innovation Academy

Xiaoqiang Hu

Institute of Metal Research

Qian Wang

qianwang@gia.cas.cn

Ganjiang Innovation Academy

Peng Yu

Ganjiang Innovation Academy

Research Article

Keywords: Ultrasonic atomization, Sn-58Bi alloy, Metal powder, Ultrasonic-assisted soldering, Lead-free solder

Posted Date: August 12th, 2026

DOI: <https://doi.org/10.21203/rs.3.rs-10615585/v1>

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Additional Declarations: No competing interests reported.

Version of Record: A version of this preprint was published at Journal of Materials Science: Materials in Electronics on October 18th, 2026. See the published version at <https://doi.org/10.1007/s10854-026-18683-x>.

Effects of Ultrasonic Atomization Parameters on the Properties of Sn-58Bi Powder and Its Application in Ultrasonic-Assisted Soldering

Xiaofeng Yang^{a, b}, Haodi Li^{a, b}, Dingying Ren^{a, b}, Hao Lan^b, Xiaoqiang Hu^c,

Qian Wang^{a, b *}, Peng Yu^{a, b *}

Address correspondence to E-mail:

Qian Wang: qianwang@gia.cas.cn

Peng Yu: pyu@gia.cas.cn

^aSchool of Rare Earths, University of Science and Technology of China, Hefei 230026, China

^bGanjiang Innovation Academy, Chinese Academy of Sciences, Ganzhou 341000, China

^cInstitute of Metal Research, Chinese Academy of Sciences, Shenyang 110000, China

Abstract:

Ultrasonic atomization (UA) leverages high-frequency acoustic waves to excite capillary waves on the melt surface and induce cavitation, facilitating droplet breakup. This technique offers significant advantages in the flexible, efficient, and cost-effective production of metal powders. However, research on the UA parameters for low-melting-point eutectic alloys used in soldering remains limited, and systematic reports on how these parameters affect powder properties are lacking. In this study, the effects of ultrasonic amplitude, melt temperature, sonotrode rotation speed, and argon atmosphere on the morphology, particle size distribution, sphericity, and oxygen content of Sn-58Bi eutectic alloy powders were systematically investigated. Furthermore, the optimized powder was formulated into a solder paste for ultrasonic-assisted soldering experiments on copper substrates. The results indicate that under an amplitude of 30%, a melt temperature of 199 °C (a superheat of ~60 °C), a rotation speed of 0 rpm, and an argon atmosphere, the synthesized powder exhibits a median diameter (D_{50}) of 65.5 μm , a sphericity of 0.87, and a satellite index of 1.07, with a significantly reduced oxygen content compared to that produced in air. In the ultrasonic-assisted soldering experiments, the shear strength of the joint reached a maximum of 44.7 MPa after 8 s of ultrasonic application and 20 min of soldering, corresponding to an intermetallic compound (IMC) layer thickness of 2.68 μm . This study elucidates the influence of UA parameters on the properties of Sn-58Bi powders and highlights the application potential of these engineered powders in ultrasonic-assisted soldering.

Keywords: Ultrasonic atomization; Sn-58Bi alloy; Metal powder; Ultrasonic-assisted soldering;

Lead-free solder

1. Introduction

Metal powders serve as critical feedstocks for technologies such as soldering and additive manufacturing (AM). The properties of these powders, including sphericity, particle size distribution (PSD), and oxygen content, dictate the ultimate performance of the fabricated components [1]. To produce high-quality powders, various atomization techniques have been developed and widely commercialized, including vacuum induction melting gas atomization (VIGA), electrode induction melting gas atomization (EIGA), plasma rotating electrode process (PREP), plasma atomization (PA), and water atomization (WA).

However, these conventional methods present inherent limitations. For instance, VIGA risks impurity introduction due to the direct contact between the melt and the crucible [2]. While EIGA mitigates crucible contamination, it necessitates costly pre-fabricated rod feedstocks and is prone to guide-tube clogging [3]. Furthermore, PREP and PA demand strictly shaped precursors and complex equipment, leading to limited batch yields [4]. Although WA is highly cost-effective, it typically yields severely oxidized and irregularly shaped particles, rendering it limited for applications with stringent sphericity and purity requirements [5].

These limitations have prompted the search for cleaner and more flexible powder production routes. Ultrasonic atomization (UA) [6] technology has received widespread attention owing to advantages such as requiring no crucible during the atomization process and having no special restrictions on raw material shape. This technology utilizes high-frequency acoustic waves to excite capillary waves on the melt surface and induce cavitation internally [7], thereby disintegrating the metallic fluid into droplets [8]. UA has gained increasing attention in metal powder production. For instance, Hinrichs et al. [9] prepared refractory metal alloy powders using ultrasonic atomization. Yankin et al. [10] fabricated 316L stainless steel powder via ultrasonic atomization and evaluated its applicability in additive manufacturing. Priyadarshi et al. [11] observed the effect of ultrasonic amplitude on the morphology of aluminum powders using high-speed photography. Li et al. [12] combined finite element simulations with powder production experiments and found that liquid films of metal melts with high surface tension could hardly thin spontaneously on the ultrasonic sonotrode. The above studies demonstrate that ultrasonic atomization holds potential for producing high-quality metal powders and leaves room for improvement.

However, studies on the ultrasonic atomization process parameters for low-melting-point alloys (such as the Sn-58Bi eutectic alloy) remain limited, and the mechanisms by which parameters such as

ultrasonic amplitude, melt heating temperature, sonotrode rotation speed, and atmospheric conditions affect powder properties are not yet clear. In terms of application, ultrasonic-assisted soldering utilizes the ultrasonic cavitation effect to physically remove oxide films from the base metal surface and promotes liquid metal spreading through acoustic streaming agitation. However, studies on formulating ultrasonically atomized powder into a solder paste and systematically evaluating its ultrasonic soldering performance are rarely reported.

Based on the above background, this study takes the Sn-58Bi eutectic alloy as the research object and investigates the effects of ultrasonic amplitude, melt heating temperature, sonotrode rotation speed, and argon atmosphere on powder properties; and utilizes the optimized powder to formulate solder paste for ultrasonic-assisted soldering, analyzing the effects of soldering holding time and ultrasonic duration on the thickness of intermetallic compounds and the shear strength of joints, and comparing them with commercial solder paste.

2. Experimental Materials and Methods

2.1. Raw materials and surface pretreatment

The raw material for powder production was a commercial Sn-58Bi eutectic alloy ingot. After being sectioned into blocks by wire electrical discharge machining, the blocks were sequentially ground with 180-, 400-, and 800-grit SiC sandpaper, ultrasonically cleaned in absolute ethanol for 20 min, and then blown dry for subsequent use. The base metal for soldering consisted of pure copper plates (purity $\geq 99.9\%$) with dimensions of 10 mm $\times$ 10 mm $\times$ 5 mm. These plates were polished using 280- to 600-grit sandpaper, ultrasonically cleaned in absolute ethanol for 30 min, and then blown dry for subsequent use.

2.2. Ultrasonic atomization powder production and atmosphere control

Powders were prepared using an MSK-UP-300 ultrasonic atomization powder production system with a nominal frequency of 35 kHz. The physical image of the equipment is shown in Fig. 1. The alloy blocks were induction-heated in a heating crucible until completely molten. The melt was then guided through a discharge valve onto the ultrasonic sonotrode, where it was disintegrated into droplets under ultrasonic action and subsequently cooled and solidified to form powder. After the system cooled to room temperature, the powder was collected, sieved through a 60-mesh standard sieve, and vacuum-sealed.

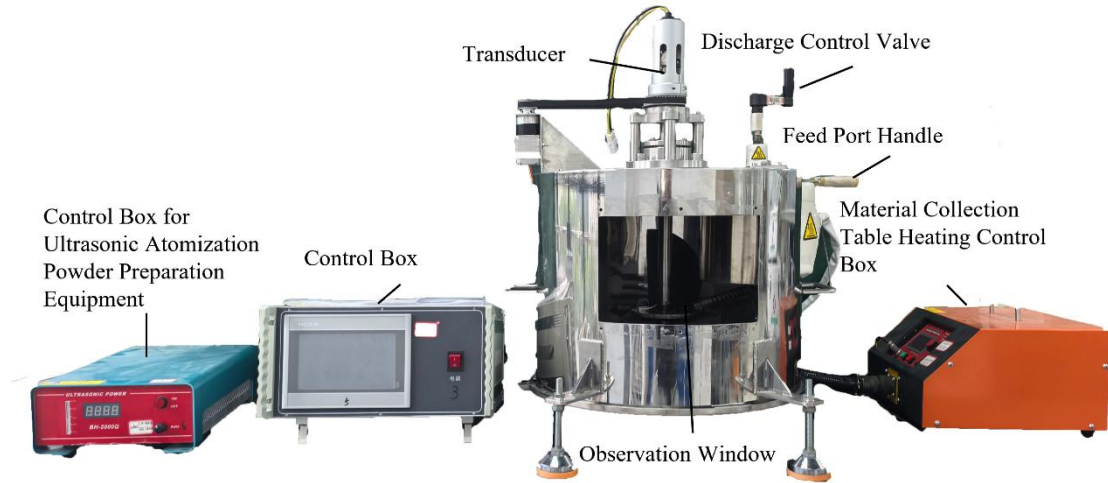


Fig. 1 Physical image of the MSK-UP-300 ultrasonic atomization powder production system

The experimental parameters were set as follows: ultrasonic amplitudes of 30%, 25%, and 20%; melt heating temperatures of 199 °C – 499 °C; and sonotrode rotation speeds of 0 – 20 rpm. To investigate the effect of a protective inert gas atmosphere on powder properties, argon was used as the protective gas. Utilizing the property that argon is denser than air, a bottom-in, top-out gas flow mode was adopted. Prior to the experiments, argon was pre-purged for 30 min to displace the air inside the chamber. The experiments under the argon atmosphere were conducted using the optimal process parameters identified from the experiments conducted in the air atmosphere.

2.3. Solder paste formulation and ultrasonic-assisted soldering experiments

The powders prepared under the optimal process parameters (amplitude: 30%, heating temperature: 199 °C, argon atmosphere) were uniformly mixed with a commercial flux to formulate the custom-made solder paste. A commercial KZ-1513 Sn-58Bi solder paste was used as the control group. The solder paste was evenly applied onto the pretreated copper surfaces, and the ultrasonic-assisted soldering was performed on a constant-temperature heating platform at 170 °C. The ultrasonic soldering machine is shown in Fig. 2. The soldering holding times were set to 10, 20, and 30 min, while the corresponding ultrasonic durations were 2, 4, 8, and 16 s. Simultaneously, specimens without ultrasonic treatment were prepared as blank controls.



Fig. 2 Handheld ultrasonic soldering machine

2.4. Powder characterization and performance testing

Micro-morphology and elemental analyses were conducted using a CLARA GMH field-emission scanning electron microscope (FE-SEM, Tescan, Czech Republic) to observe the powder morphology and the cross-sectional microstructure of the solder joints. The secondary electron (SE) mode was utilized for powder characterization, the backscattered electron (BSE) mode for interfacial analysis, and the equipped energy-dispersive X-ray spectroscopy (EDS) for elemental distribution mapping. The volumetric particle size parameters (D_{10} , D_{50} , and D_{90}) were determined using a Bettersize 2600 laser diffraction particle size analyzer (Dandong Bettersize, China). Prior to the measurements, the samples were ultrasonically dispersed for 10 min. The sphericity and satellite index of the powders were evaluated using an R-3000 particle size and shape analyzer (Micro-Spectrum Technology, China). The oxygen content was measured with an ONH-3500 oxygen/nitrogen/hydrogen elemental analyzer (NCS Testing Technology, China). The melt contact angles were measured using a DSAHT 1600 high-temperature contact angle meter (KRÜSS, Germany) on a substrate made of the same titanium alloy as the sonotrode. The measurements were performed in both air and argon atmospheres after isothermally holding the samples for 10 min at various temperature points ranging from 199 to 499 °C. The dynamic viscosity of the melt within the temperature range of 199–499 °C was tested using a BCT-1700 high-temperature viscometer (Huanao Technology, China).

2.5. Evaluation of interfacial metallurgy and shear mechanical properties

The soldered joints were cross-sectioned, ground, and polished. Subsequently, they were etched with a 4 wt.% hydrochloric acid-ethanol solution for 5 s. The morphology of the intermetallic compounds (IMCs) was observed, and their thickness was measured using the BSE mode of the FE-SEM. The shear strength of the joints was evaluated using a WDW-50A electronic universal testing machine (Xinguang

Testing Machine, China) at a crosshead speed of 0.5 mm/min. To ensure data reliability, 3 to 5 specimens were tested for each set of parameters.

3. Results and discussion

3.1. Effect of ultrasonic amplitude on powder properties

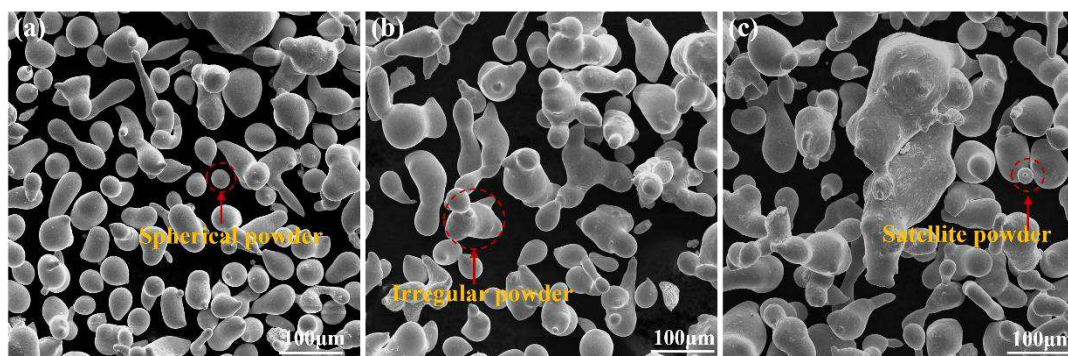


Fig. 3 SEM images of Sn-58Bi powders under different ultrasonic amplitudes: (a) 30%; (b) 25%; (c) 20%

Fig. 3 shows the scanning electron microscopy (SEM) images of the Sn-58Bi powders prepared at different ultrasonic amplitudes. As the ultrasonic amplitude increased from 20% to 30%, the proportion of spherical powders gradually increased, while the amount of irregular powders [13] and satellite particles [8] progressively decreased. At an amplitude of 30%, the powders exhibited the highest quantity of spherical particles with smooth surfaces and few satellites; conversely, at 20% amplitude, abundant irregular particles and satellites emerged.

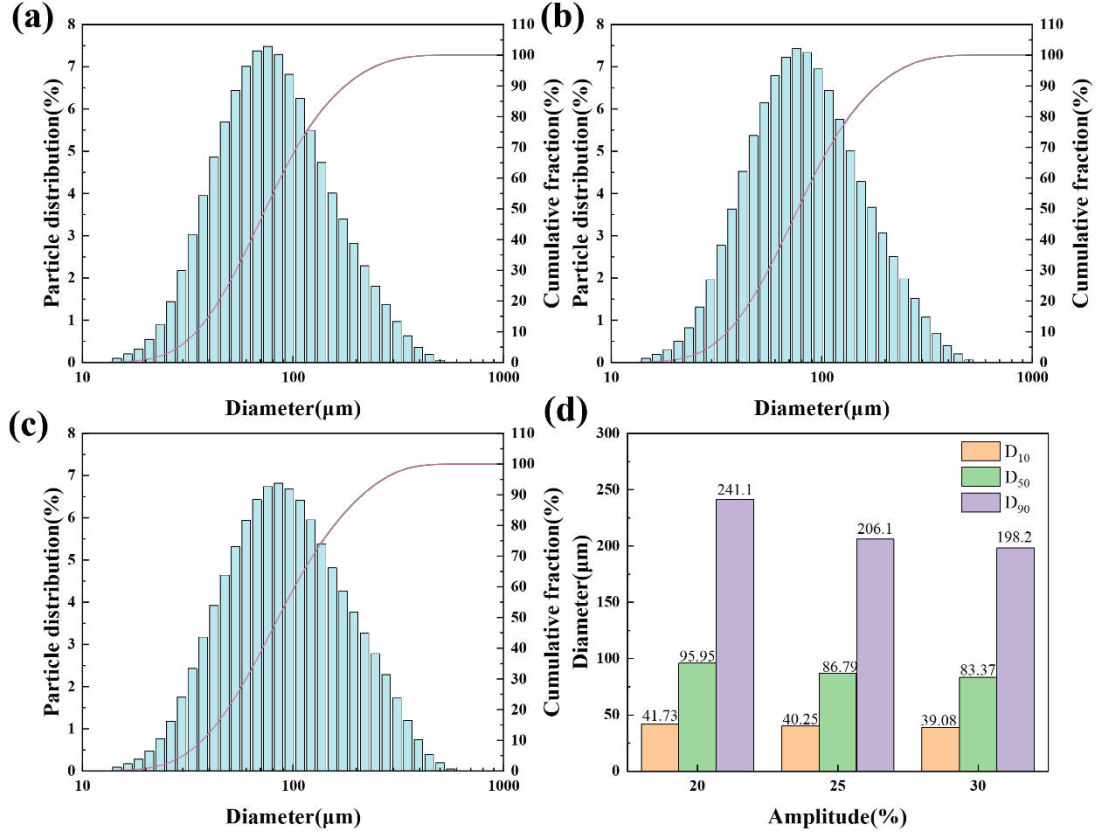


Fig. 4 Particle size distributions of Sn-58Bi powders under different ultrasonic amplitudes: (a) 30%; (b) 25%; (c) 20%; (d) D₁₀, D₅₀, and D₉₀

Fig. 4 illustrates the particle size distributions of the Sn-58Bi powders prepared at different ultrasonic amplitudes. The particle size under each amplitude exhibited a unimodal normal distribution, although the peak position and width varied. When the ultrasonic amplitude was 30%, the D₁₀, volumetric median diameter D₅₀, and D₉₀ of the powders were 39.08 μm, 83.37 μm, and 198.2 μm, respectively, all of which were smaller than the corresponding values at amplitudes of 25% and 20%. The span S was employed to quantify the width of the particle size distribution [10, 14]: $S = \frac{D_{90} - D_{10}}{D_{50}}$. The S values at amplitudes of 30% and 25% were close, being 1.909 and 1.911, respectively. However, at 20%, the S value increased to 2.079, indicating that a lower amplitude leads to a broader particle size distribution.

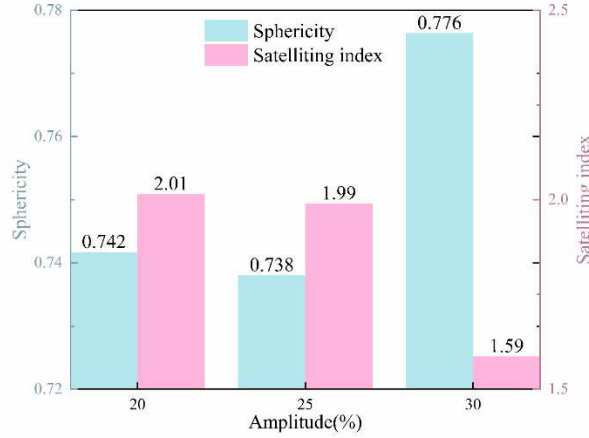


Fig. 5 Sphericity and satellite index of Sn-58Bi powders under different ultrasonic amplitudes

Fig. 5 presents the sphericity and satellite index of the powders prepared at different ultrasonic amplitudes. As the ultrasonic amplitude increased, the sphericity gradually increased, while the satellite index progressively decreased. At an amplitude of 30%, the sphericity reached its maximum and the satellite index its minimum, indicating that this amplitude facilitates the shrinkage of droplets into spheres and suppresses the formation of satellite particles [15].

The above phenomena can be explained by the capillary wave mechanism of ultrasonic atomization [16, 17]. As the driving amplitude increases, the amplitude of the capillary waves also increases. When the wave amplitude exceeds a critical threshold, the wave surface becomes unstable, and droplets detach from the wave crests, initiating atomization [18]. A higher amplitude accelerates the growth of capillary waves and shortens the time required for the liquid film to reach the critical breakup thickness, thereby resulting in the formation of finer droplets from the melt. In addition to the capillary wave effect, cavitation is another crucial mechanism through which ultrasonic amplitude influences powder properties. In an ultrasonic field, tiny gas nuclei within the liquid undergo oscillation, expansion, and contraction under the action of acoustic waves, ultimately leading to implosion [19]. When the ultrasonic amplitude increases, the amplitude of the acoustic pressure injected into the melt correspondingly rises. Consequently, a massive number of cavitation bubbles rapidly expand and aggregate into bubble clusters during the negative pressure phase. Upon collapsing during the high-pressure phase, these bubbles generate micro-shock waves and microjets that tear the liquid film from within. This internal tearing synergizes with the surface capillary wave instability to form a cooperative fragmentation mechanism [20, 21]. An increase in the driving acoustic pressure leads to an elevated bubble number density, which intensifies the coupling interactions among the bubbles and exacerbates the cavitation effect [22].

Therefore, increasing the amplitude not only enhances the capillary wave effect but also intensifies the fragmentation of the liquid film caused by cavitation implosions. Together, these two mechanisms result in droplet refinement and a reduction in powder particle size. Consequently, under the present experimental conditions, an amplitude of 30% is conducive to obtaining finer and more uniformly sized Sn-58Bi powders.

3.2. Effect of melt heating temperature on powder properties

3.2.1. Powder morphology and particle size distribution

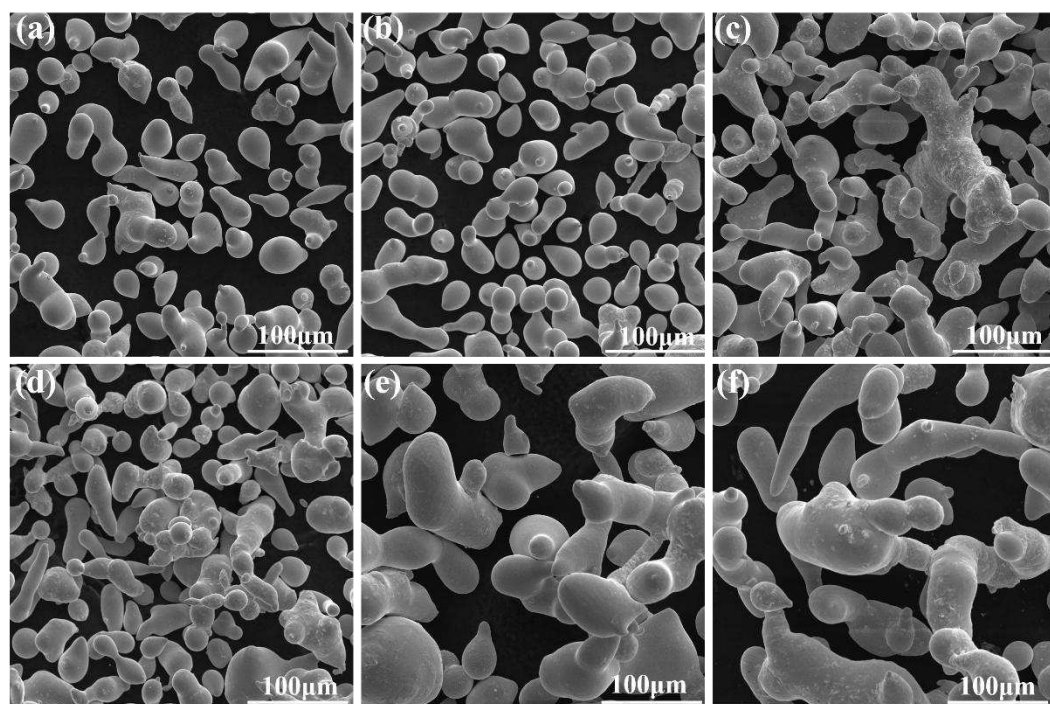


Fig. 6 SEM images of Sn-58Bi powders at different melt heating temperatures: (a) 199 °C; (b) 259 °C; (c) 319 °C; (d) 379 °C; (e) 439 °C; (f) 499 °C

Fig. 6 shows the morphology of the Sn-58Bi powders prepared at different melt heating temperatures. As the heating temperature increased from 199 °C to 499 °C, the proportion of regular spherical particles decreased, while irregular particles and inter-particle adhesion phenomena significantly increased, accompanied by a rise in the number of satellite particles. Overall, the sphericity of fine particles was superior to that of coarse particles.

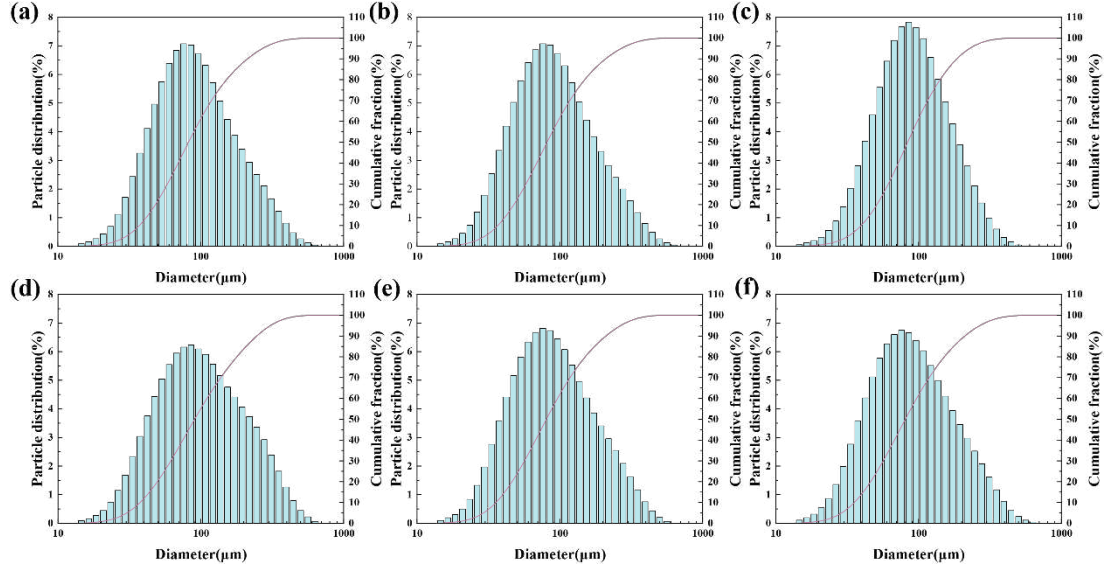


Fig. 7 Particle size distributions of Sn-58Bi powders at different melt heating temperatures: (a) 199 °C; (b) 259 °C; (c) 319 °C; (d) 379 °C; (e) 439 °C; (f) 499 °C

Fig. 7 presents the particle size distribution curves of the powders at different melt heating temperatures, with all samples exhibiting unimodal normal distribution characteristics. The particle size parameters are listed in Table 1. As the heating temperature rose, the median diameter D_{50} initially decreased and then increased, reaching a minimum value of 91.6 μm at 259 °C. The D_{50} was 92.5 μm at 199 °C, increased continuously beyond 319 °C, and rose to 106.3 μm at 499 °C. The span S was generally maintained between 1.7 and 2.2. Specifically, at 319 °C, S reached its minimum of 1.7, indicating the most concentrated particle size distribution. At 439 °C and 499 °C, S rose to 2.2, reflecting a broadened distribution.

Table 1. Particle size distributions of Sn-58Bi powders at different melt heating temperatures.

Melt heating temperature(°C)	D_{10} (μm)	D_{50} (μm)	D_{90} (μm)	S
199	41.9	92.5	238.9	2.1
259	41.4	91.6	235.3	2.1
319	44.7	94.2	208.6	1.7
379	42.0	101.9	254.3	2.1
439	44.4	103.9	268.7	2.2
499	45.5	106.3	282.7	2.2

The relationship between the surface tension of a metal melt and temperature follows a semi-

empirical equation [23, 24]: $\sigma_m \left(\frac{M}{\rho_m} \right)^{\frac{2}{3}} = K(T_c - T)$, where σ is the surface tension, M is the molar mass of the melt, ρ_m is the melt density, T_c is the melting point, and K is an experimental constant. According to this equation, the surface tension of the metal melt is negatively correlated with temperature. Furthermore, based on the droplet size estimation formula [25]: $D_p = 0.34 \left(\frac{8\pi\sigma_m}{\rho f^2} \right)^{\frac{1}{3}}$, where D_p is the droplet size, σ_m is the surface tension, ρ is the liquid density, and f is the frequency. By combining these two equations, it can be deduced that as the temperature rises and surface tension decreases, the atomized droplet size should theoretically continue to decrease, and the droplets should find it more difficult to maintain a spherical morphology. However, Table 1 reveals that the D_{50} exhibits an anomalous trend of initially decreasing and subsequently increasing. This suggests that, in addition to surface tension, other regulating factors are at play [26].

3.2.2. Melt viscosity and powder oxygen content

To elucidate the anomalous variation in particle size, the dynamic viscosity of the Sn-58Bi melt and the oxygen content of the corresponding powders were measured at different temperatures, as shown in Fig. 8(a) and (b). The viscosity test results indicate that the melt viscosity exhibits a declining trend as the temperature rises. It should be noted that the overall viscosity values measured in the air atmosphere are somewhat high. This is attributed to the oxide film formed on the melt surface, which hinders the rotation of the viscometer rotor [27]. Nevertheless, the fundamental rule that viscosity decreases with increasing temperature remains unaffected.

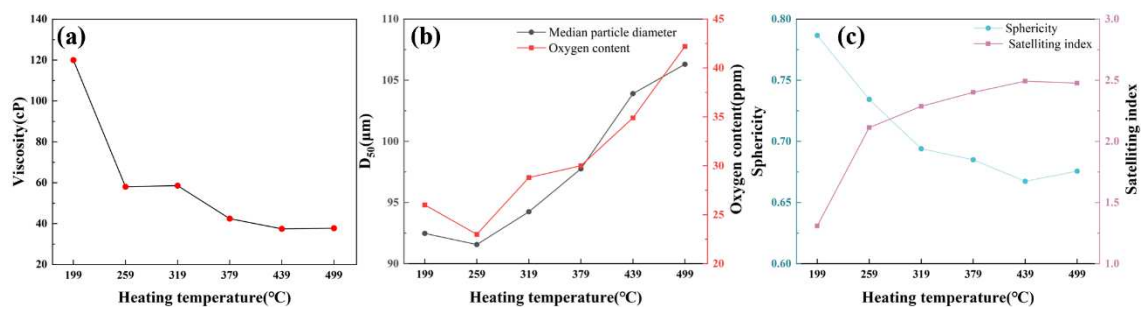


Fig. 8 Effect of melt temperature on Sn-58Bi powder characteristics: (a) dynamic viscosity of the melt; (b) oxygen content distribution; (c) sphericity and satellite index

The powder oxygen content also exhibits a trend of initially decreasing and then increasing with the heating temperature. The extremum point coincides with that of the particle size, both occurring at 259 °C. This variation likely stems from the competition between two mechanisms: (1) When the temperature is below 259 °C, the viscosity reduction induced by the temperature rise plays a dominant role. The lower-

viscosity melt spreads more readily on the sonotrode to form a thin liquid film, thereby enhancing the ultrasonic atomization efficiency. Consequently, the powder particle size gradually refines with increasing temperature, which is consistent with theoretical predictions. (2) When the temperature exceeds 259 °C, the high-temperature oxidation effect gradually becomes the dominant factor. Surface oxidation of the melt intensifies, forming an oxide film. During ultrasonic fragmentation, this oxide film breaks up alongside the droplets and adheres to the particle surfaces, leading to an increase in particle size [28]. Simultaneously, the reduced surface tension prolongs the spheroidization time of the droplets. Droplets that have not fully solidified are more prone to colliding and coalescing during flight, further resulting in a higher proportion of coarse particles and an overall increase in particle size.

3.2.3. Sphericity and satellite index

The sphericity and satellite index of the powders at different temperatures are shown in Fig. 8(c). As the melt heating temperature rises, the sphericity decreases, and the satellite index increases. At 199 °C, the sphericity reaches its peak, and satellite particles are minimized. This trend is consistent with the relationship between surface tension and temperature. Surface tension acts as the driving force for the droplet spheroidization process, functioning to minimize the surface area of the liquid mass. However, the elevated temperature lowers the surface tension, resulting in insufficient spheroidization driving force. Consequently, it becomes more difficult for the droplets to shrink into regular spheres, making the formation of irregular powders more likely at higher temperatures.

3.3. Effect of inert protective atmosphere on powder properties

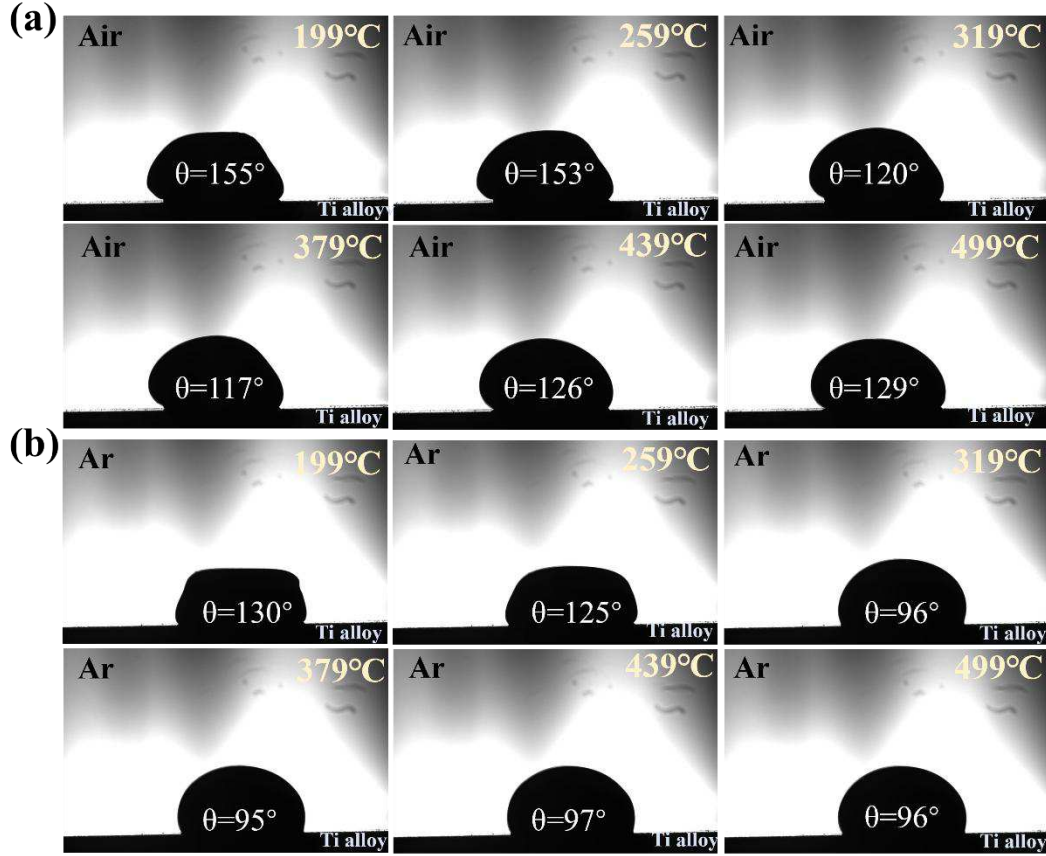


Fig. 9 High-temperature contact angles of Sn-58Bi at different temperatures in air and argon atmospheres: (a) in air; (b) in argon

The contact angle morphologies of the Sn-58Bi melt on the titanium alloy substrate under different temperatures and atmospheres are shown in Fig. 9, and their quantitative variation trends are illustrated in Fig. 10. Overall, the melt contact angles in the argon atmosphere were consistently lower than those in the air atmosphere across all tested temperatures, and their variation trends with temperature exhibited distinct differences. In the air atmosphere, the contact angle initially decreased and subsequently increased with rising temperature. This variation was also governed by the combined effects of surface tension and oxidation; specifically, the viscosity reduction induced by heating initially played a dominant role, whereas high-temperature oxidation gradually became dominant once the temperature exceeded a certain threshold.

In the protective argon atmosphere, the oxidation reaction was effectively suppressed, and the contact angle exhibited a continuous downward trend with increasing temperature: it was 130° at 199 °C, dropped rapidly to below 100° beyond 319 °C, and stabilized at 95°–97° in the high-temperature range. Devoid of the constraint from an oxide film, the surface tension of the melt decreased with rising

temperature, leading to a continuously enhanced spreading ability. Consequently, the contact angle steadily decreased upon heating.

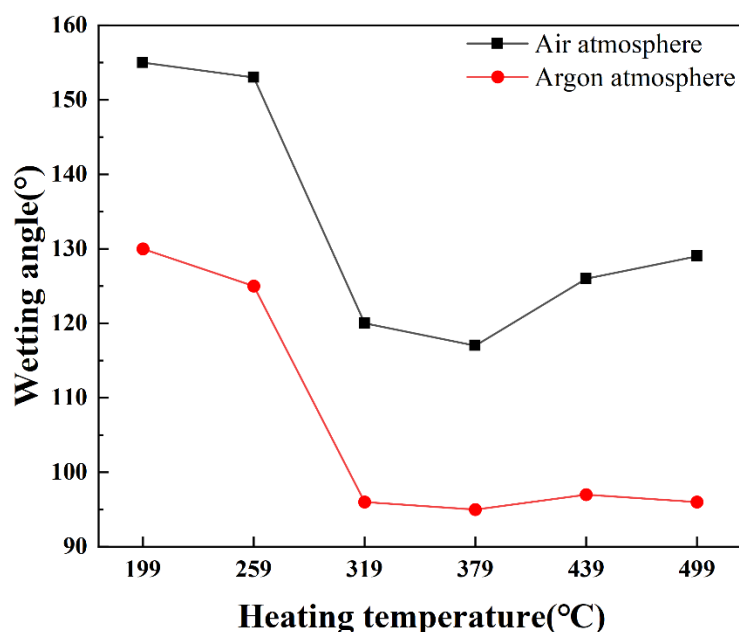


Fig. 10 Variation of high-temperature contact angles of Sn-58Bi at different temperatures in air and argon atmospheres

The oxide film on the melt surface in the air atmosphere consumed a portion of the ultrasonic energy, hindering the spheroidization and shrinkage of the droplets during flight. Meanwhile, fragmented oxide films promoted adhesion between droplets, thereby increasing the probability of satellite powder formation. In contrast, under the argon atmosphere, the melt was free from the hindrance of an oxide film. Thus, the ultrasonic energy could act more efficiently on melt fragmentation, allowing the droplets to fully shrink and spheroidize driven by surface tension during flight.

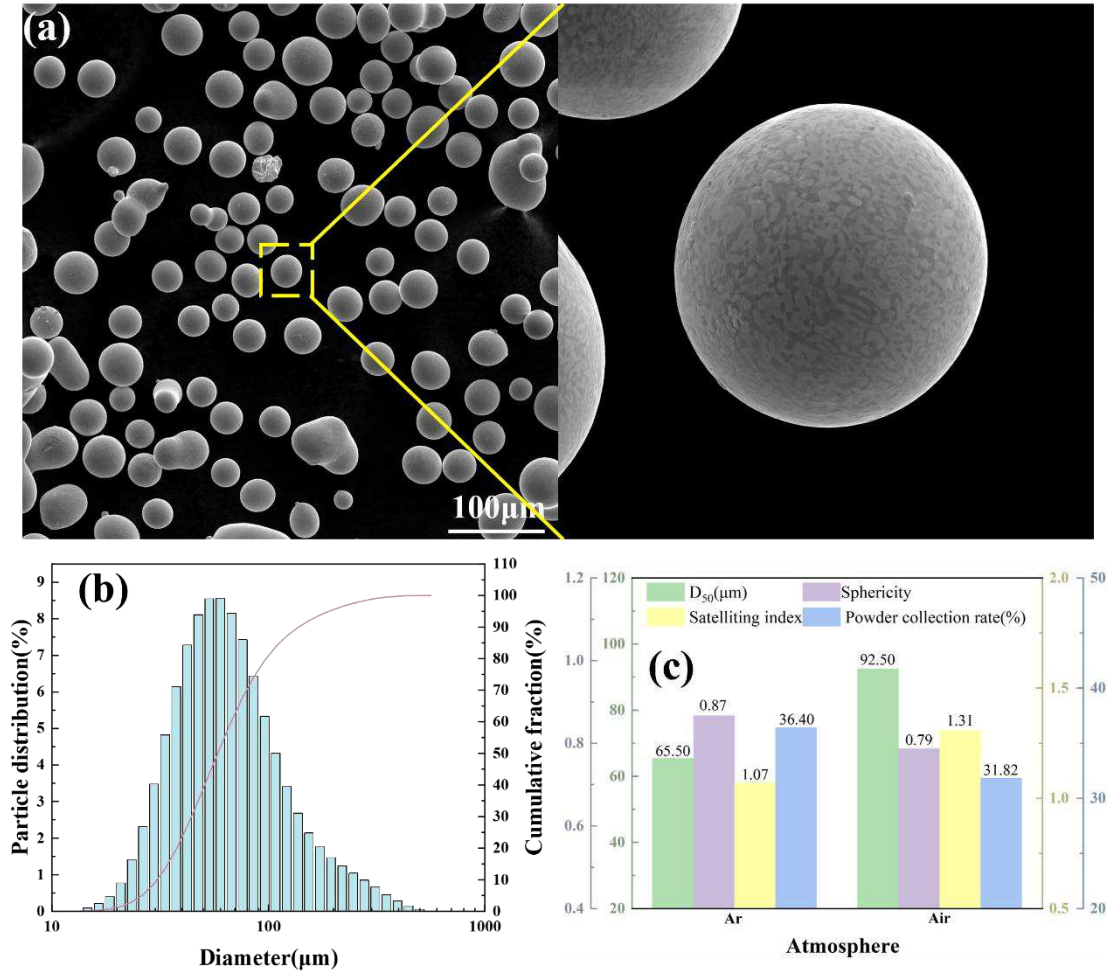


Fig. 11 Characteristics of Sn-58Bi powders: (a) SEM morphology and (b) particle size distribution in an argon atmosphere; (c) property comparison between argon and air atmospheres

Under the optimal process parameters (amplitude: 30%, heating temperature: 199 °C), the morphology and particle size distribution of the Sn-58Bi powders prepared with argon protection are shown in Fig. 11(a) and (b). The powders generally exhibited a regular spherical shape with an extremely low proportion of irregular particles and a uniform particle size distribution. Moreover, the overall particle size was significantly finer than that prepared in the air atmosphere. Fig. 11(c) summarizes the comparison of selected performance indicators of the powders under the two atmospheres. Compared to the air atmosphere, argon protection achieved an overall improvement in powder quality: the median diameter D_{50} decreased from 92.50 μm to 65.50 μm, representing a reduction of 29.19%; the sphericity increased from 0.79 to 0.87; the satellite index dropped from 1.31 to 1.07; and concurrently, the powder collection rate improved from 31.82% to 36.40%. This demonstrates that argon protection ultimately optimizes powder particle size, morphology, and yield by suppressing oxidation and improving melt

spreading and atomization efficiency.

3.4. Effect of sonotrode rotation speed on powder properties

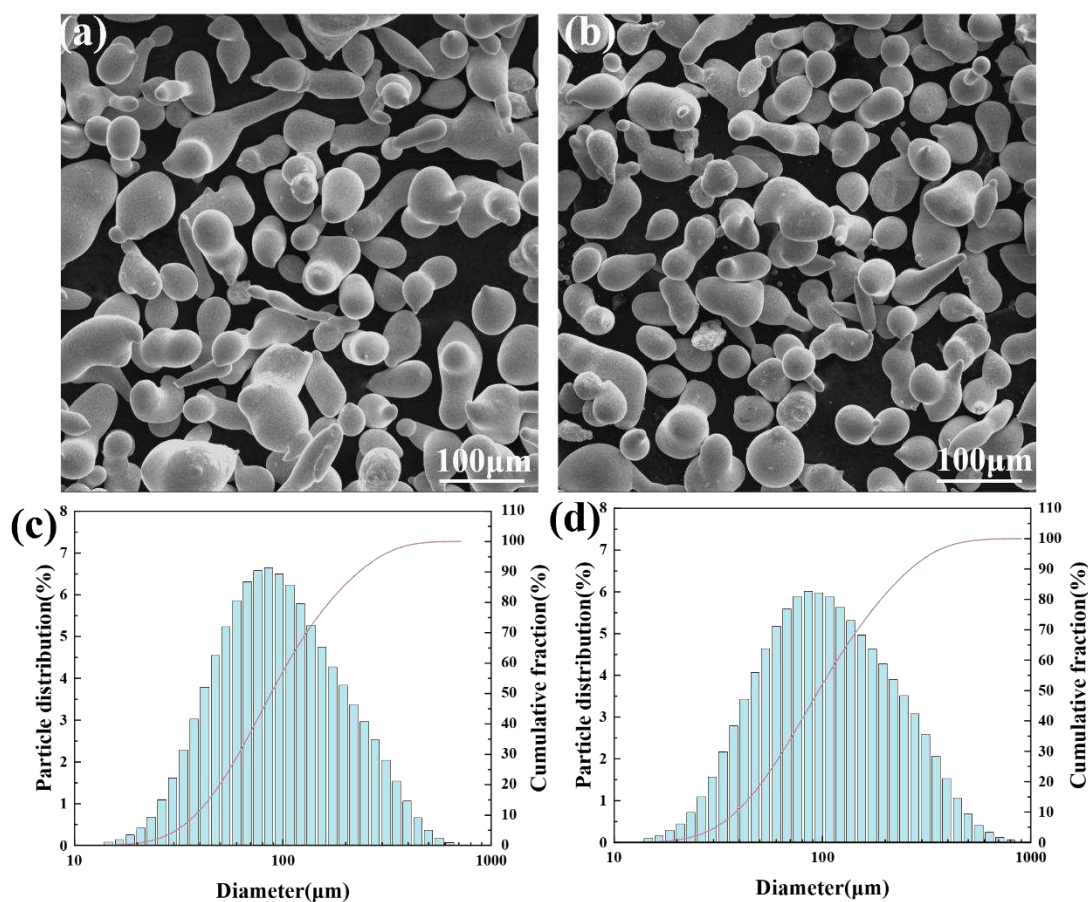


Fig. 12 Morphology and particle size distributions of Sn-58Bi powders at different rotation speeds: (a, c) 10 rpm; (b, d) 20 rpm

The morphology and particle size distributions of the powders prepared at different rotation speeds are shown in Fig. 12. Compared with the stationary state of the sonotrode, the introduction of rotation reduced the proportion of regular spherical particles while significantly increasing the number of irregular and satellite particles. Furthermore, this trend progressively intensified as the rotation speed increased from 0 to 20 rpm. The particle sizes of all samples exhibited a unimodal normal distribution. The specific particle size distributions are listed in Table 2:

Table 2. Particle size distributions of Sn-58Bi powders at different sonotrode rotation speeds.

Sonotrode rotation speed (rpm)	D ₁₀ (μm)	D ₅₀ (μm)	D ₉₀ (μm)	S
0	41.9	92.5	238.9	2.1
10	42.7	98.7	260.3	2.2
20	43.2	107.9	296.8	2.4

As the rotation speed increased, the median diameter D_{50} of the powder increased: it was 92.5 μm in the stationary state, increased to 98.7 μm at 10 rpm, and reached 107.9 μm at 20 rpm. Simultaneously, the span S also increased from 2.1 in the stationary state to 2.4 at 20 rpm, indicating a gradually broadening particle size distribution. The variation patterns of powder sphericity, satellite index, and powder collection rate with rotation speed are presented in Fig. 13.

These patterns align with the trends observed in morphology and particle size: As the rotation speed increased, the powder sphericity gradually decreased while the satellite index increased, leading to a progressive deterioration in morphological quality. Concurrently, the powder collection rate increased with the elevated rotation speed. This phenomenon occurs because the centrifugal force drives the melt to spread radially along the sonotrode, leading to an uneven distribution of the liquid film thickness. Regions with a thin liquid film undergo premature atomization, whereas regions with a thick film are flung outwards by the centrifugal force before being sufficiently fragmented by the ultrasound. This results in the formation of large-sized droplets, ultimately causing an increase in the overall D_{50} and a broader particle size distribution. Furthermore, the centrifugal force imparts additional radial and tangential velocities to the droplets, making their flight trajectories more complex. This elevates the probability of incompletely solidified droplets colliding and adhering in mid-air, thereby increasing the number of satellite particles and reducing sphericity. On the other hand, the centrifugal force expands the spreading area of the melt on the sonotrode and reduces the proportion of melt lost directly due to gravity. This enables a larger volume of melt to participate in the atomization process, ultimately enhancing the powder collection rate.

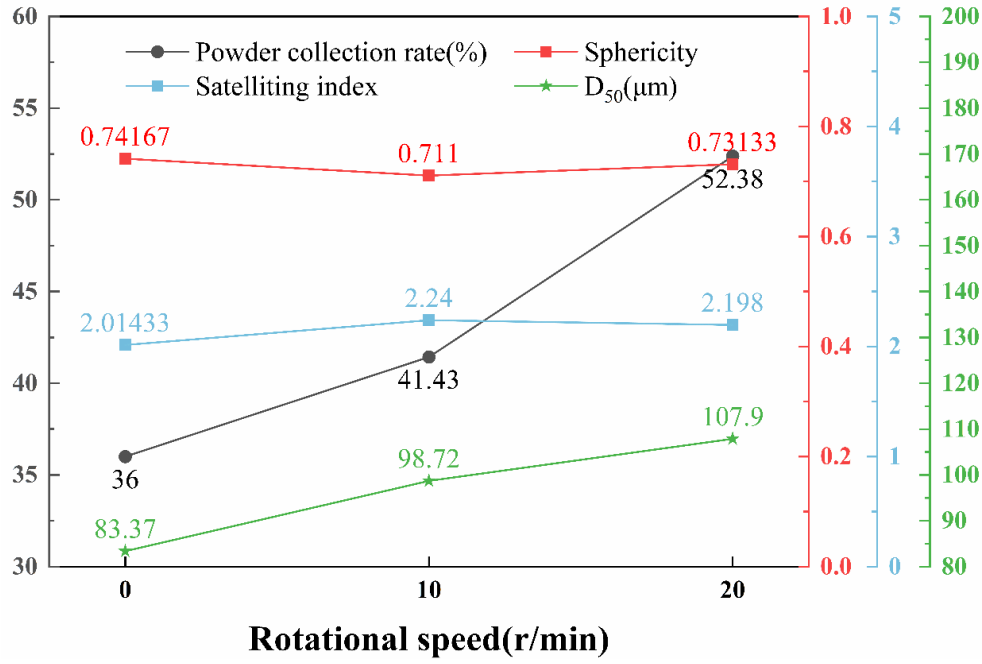


Fig. 13 Effect of sonotrode rotation speed on the sphericity, satellite index, and powder collection rate of Sn-58Bi powders

Overall, while sonotrode rotation effectively boosts powder production capacity, it sacrifices the particle size uniformity and morphological quality of the powder to some extent.

3.5. Effects of ultrasonic-assisted soldering process on IMC layer thickness and solder seam width

The test results of IMC layer thickness and solder seam width under different ultrasonic soldering processes are shown in Fig. 14. At a constant soldering holding time, the IMC layer thickness initially increased and then decreased with prolonged ultrasonic duration, peaking at an ultrasonic duration of 8 s across all groups. Under the same ultrasonic duration, the IMC layer thickness continuously increased with the extension of the soldering holding time. The maximum IMC layer thickness of 2.78 μm was obtained under the condition of 30 min soldering holding time and 8 s ultrasonic duration, whereas the minimum thickness of 1.64 μm occurred at 10 min holding time and 16 s ultrasonic duration.

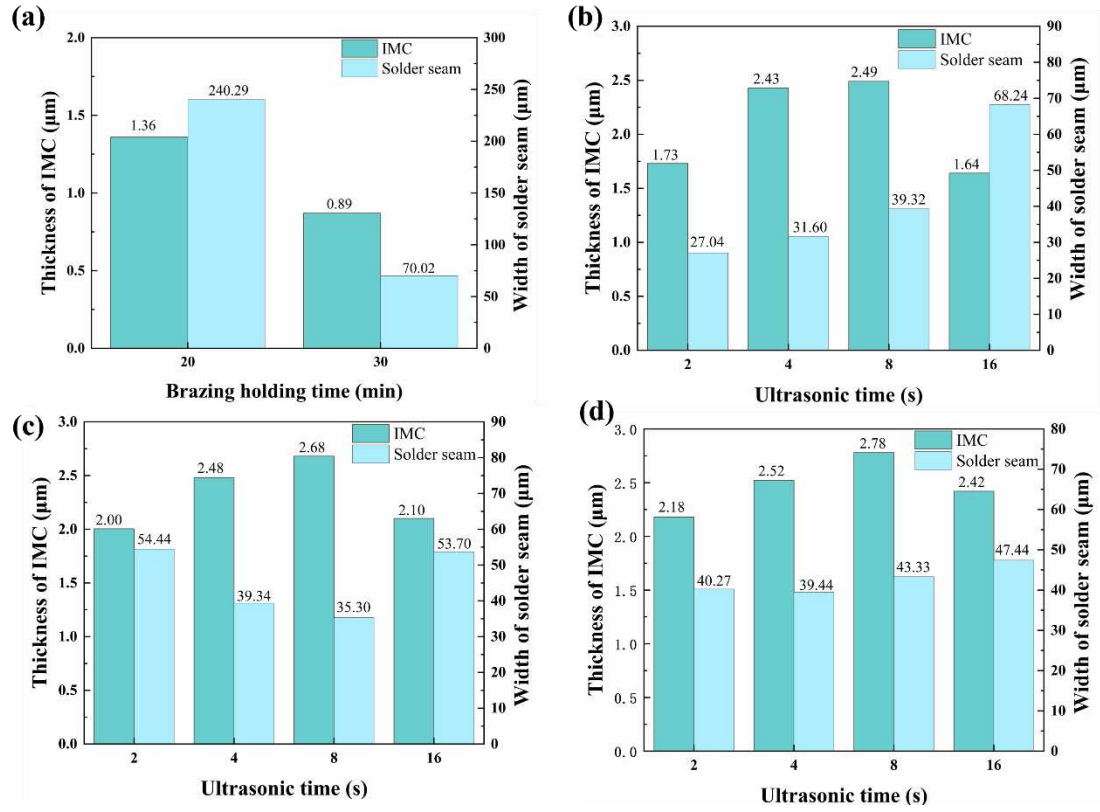


Fig. 14 Effects of ultrasonic soldering parameters on IMC layer thickness and solder seam width:
(a) without ultrasound; and with varying ultrasonic durations at holding times of (b) 10 min, (c) 20 min, and (d) 30 min

Taking the group with 20 min soldering holding time as a representative example, the cross-sectional interfacial morphologies of the joints under different ultrasonic durations are shown in Fig. 15. At 2 s of ultrasound, the IMC layer was relatively thin, displaying unsatisfactory overall continuity. As the ultrasonic duration extended to 4 s and 8 s, the IMC layer continuously grew and thickened, becoming progressively continuous and planar, which gradually enhanced the degree of interfacial metallurgical bonding. When the ultrasonic duration was further extended to 16 s, the IMC layer experienced fragmentation and lost its continuity. Some IMC particles detached from the interface and dispersed into the interior of the solder seam, demonstrating that excessive ultrasound exerts a destructive effect on the IMC layer.

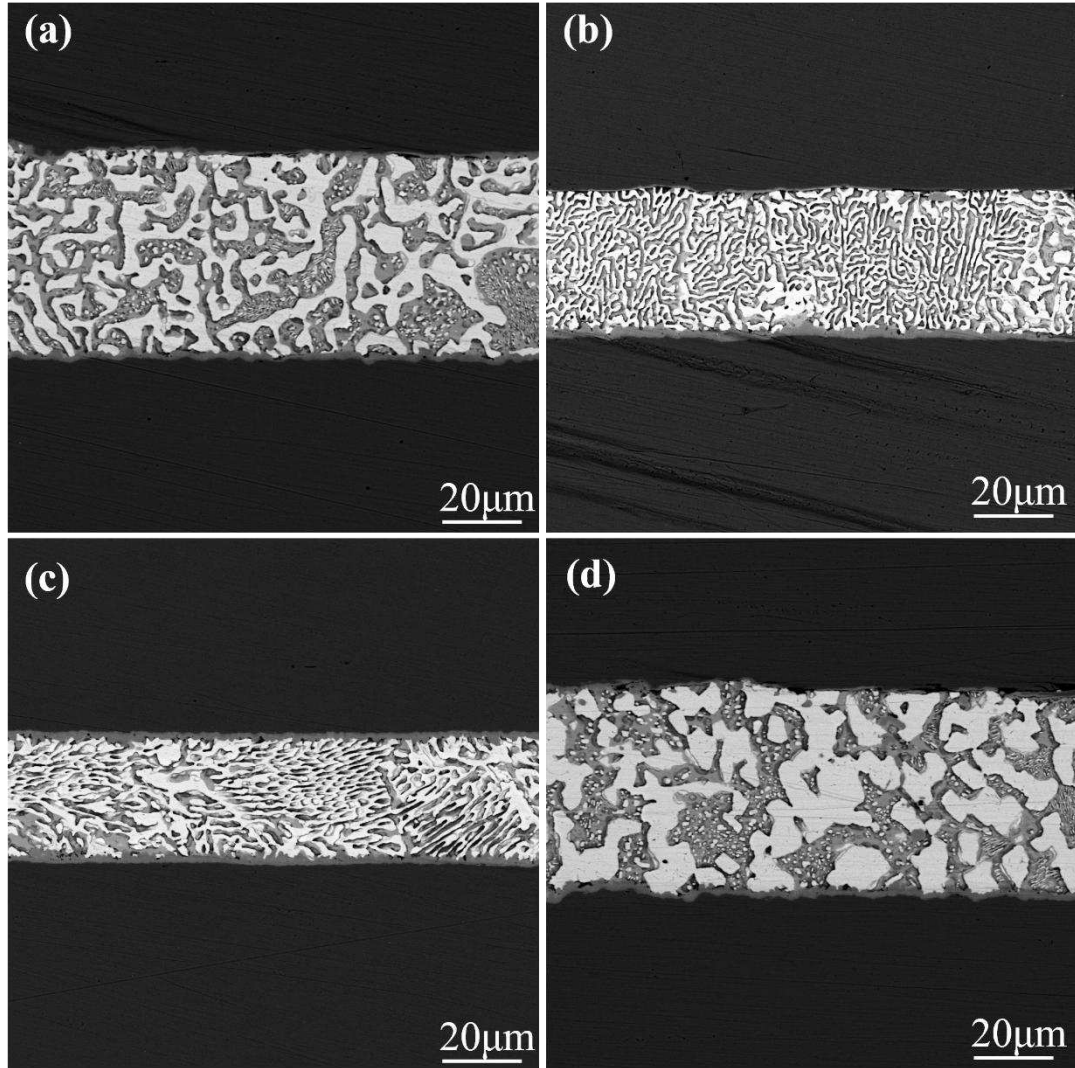


Fig. 15 Interfacial microstructures of joints under 20 min soldering holding time with different ultrasonic durations: (a) 2 s; (b) 4 s; (c) 8 s; (d) 16 s

These findings reveal a critical threshold in the ultrasonic regulation of IMC growth. When the ultrasonic duration does not exceed 8 s, cavitation and acoustic streaming break the oxide film on the base metal surface and clean the interface, thereby improving the wetting and spreading capabilities of the molten solder. Concurrently, the stirring effect in the molten solder accelerates the mass transfer of Cu and Sn atoms toward the interface, providing favorable conditions for the formation and growth of IMCs such as Cu_6Sn_5 [29]. However, when the ultrasonic duration is extended to 16 s, the sustained intense acoustic streaming impact induces a crushing/fragmentation effect on the newly formed IMC layer [30].

Coupled with the EDS results in Table 3, Cu elements were detected in the solder seam region far from the interface. This confirms that a fraction of the fragmented IMC particles detached from the

interface, became entrained in the liquid solder, and ultimately remained trapped upon solidification. Consequently, this led to a decrease in the IMC layer thickness. The non-ultrasonic specimens exhibited an anomalous thickness variation: the IMC thickness at 30 min holding time was unexpectedly smaller than that at 20 min. This anomaly is presumably attributed to the poor spreading uniformity of the solder on the substrate without ultrasound; the melted solder paste was easily squeezed out of the joint region, resulting in an insufficient Sn supply for interfacial reaction and thereby suppressing IMC growth.

Table 3. EDS elemental analysis results in the solder seam region away from IMCs under 16 s ultrasonic duration at different soldering times.

Soldering time (min)	Sn (At. %)	Bi (At. %)	Cu (At. %)
10	50.9	12.0	37.1
20	8.9	79.6	11.5
30	13.9	76.6	9.5

Furthermore, Figs. 14 and 15 demonstrate the regulatory effect of ultrasound on the solder seam width: ultrasonic application significantly reduced the solder seam width, with all ultrasonically assisted samples showing smaller seam widths than the unassisted ones. Under the holding conditions of 10 min and 30 min, the solder seam width gradually increased with prolonged ultrasonic duration; whereas under the 20 min holding condition, the solder seam width exhibited a trend of initially decreasing and then increasing, with the minimum seam width achieved at an 8 s ultrasonic duration.

3.6. Effect of ultrasonic-assisted soldering process parameters on the shear strength of solder joints

The shear strength test results of the soldered joints under different process parameters are shown in Fig. 16. As depicted in Fig. 16(a), under conditions without ultrasonic assistance, the overall shear strength of the joints remained at a relatively low level. Notably, the specimen soldered for 10 min failed to form an effective metallurgical bond, making it nearly impossible to measure a valid shear strength.

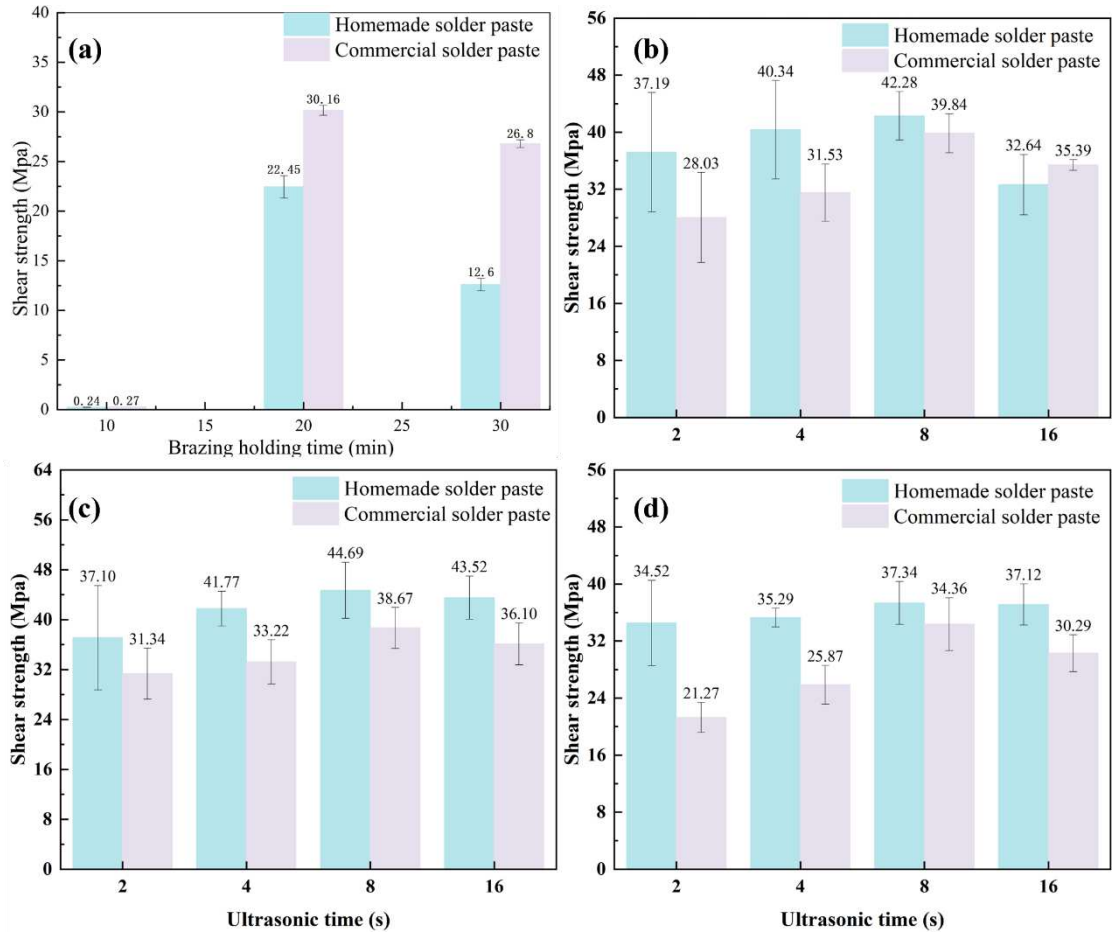


Fig. 16 Shear strength of soldered joints: (a) effect of holding time without ultrasound; effect of ultrasonic duration at holding times of (b) 10, (c) 20, and (d) 30 min

Figs. 16(b)–(d) illustrate the variations in shear strength of the joints fabricated with the custom-made and commercial Sn-58Bi solder pastes versus ultrasonic duration at soldering holding times of 10, 20, and 30 min, respectively. Under all three sets of process conditions, the shear strength of the joints for both solder pastes exhibited a trend of initially increasing and subsequently decreasing with prolonged ultrasonic duration, both peaking at 8 s. Specifically, under the condition of 20 min holding time and 8 s ultrasonic duration, the joint made with the custom-made paste achieved the highest strength of 44.69 MPa among all process parameters, outperforming the commercial paste (38.67 MPa) under identical conditions. The peak strength of the commercial paste was 39.84 MPa, which corresponded to the process combination of 10 min holding time and 8 s ultrasonic duration. In overall comparison, except for the condition of 10 min holding time and 16 s ultrasonic duration, where the custom-made paste was slightly weaker than the commercial one, the shear performance of the joints utilizing the custom-made paste was superior across all other parameters. The shear strength of the joints exhibited a clear correlation

with the IMC layer thickness. In this study, the IMC thickness corresponding to the peak strength was approximately 2.68 μm , which closely approached the maximum IMC thickness measured in the experiments. Conversely, the minimum strength corresponded to the thinnest IMC layer of 1.64 μm . Previous studies [31] indicate that when the IMC thickness falls within the range of 1–3 μm , the joint can achieve a balance between adequate interfacial metallurgical bonding and the avoidance of brittle fracture. If the layer is overly thin, the interfacial reaction is insufficient, resulting in inadequate bonding strength. Conversely, an excessively thick layer makes the brittle IMC prone to the initiation of micro-cracks and voids internally, thereby degrading the fracture toughness of the joint [32, 33]. In this study, with the exception of the specimen soldered for 30 min without ultrasound, the IMC thicknesses of all other joints fell within the 1–3 μm range. This demonstrates that the selected process parameters exert a favorable regulatory effect on IMC growth, ensuring stable and reliable mechanical properties for the joints.

4. Conclusions

In this study, the effects of ultrasonic amplitude, melt temperature, sonotrode rotation speed, and argon atmosphere on the properties of Sn-58Bi powders prepared by ultrasonic atomization were investigated. Furthermore, the performance of the optimized powder in ultrasonic-assisted soldering was evaluated.

(1) As the ultrasonic amplitude increased from 20% to 30%, the capillary wave and cavitation effects were enhanced. Consequently, the median diameter (D_{50}) of the powder decreased, sphericity improved, and the satellite index dropped. Higher amplitudes are conducive to obtaining finer powders with superior sphericity.

(2) When the melt heating temperature increased from 199 °C to 499 °C, the D_{50} of the powder exhibited a trend of initially decreasing and then increasing. The optimal comprehensive powder quality was achieved at a heating temperature of 199 °C, yielding a D_{50} of 92.5 μm and a relatively low oxygen content. Excessive temperatures exacerbated oxidation, leading to the formation of an oxide film that hindered spheroidization and promoted the generation of satellite particles.

(3) The centrifugal force introduced by the sonotrode rotation increased the powder collection rate from 36% in the stationary state to 52.38% at 20 rpm. However, the D_{50} increased from 83.37 μm to 107.9 μm , accompanied by a decrease in sphericity and an increase in the satellite index. While rotation can boost production capacity, it sacrifices particle size uniformity and morphological quality.

(4) The argon protective atmosphere effectively suppressed melt oxidation. Compared to the air atmosphere, the D_{50} of the powder decreased by 29.19%, the sphericity increased from 0.79 to 0.87, the satellite index dropped from 1.31 to 1.07, and the oxygen content was significantly reduced.

(5) In the ultrasonic-assisted soldering experiments, the shear strength of the joints initially increased and then decreased with prolonged ultrasonic duration. A maximum shear strength of 44.69 MPa was achieved at an ultrasonic duration of 8 s and a soldering holding time of 20 min, corresponding to an intermetallic compound thickness of 2.68 μm . Extending the ultrasonic duration to 16 s led to the fragmentation of the IMC layer and a subsequent decline in strength. Under identical process parameters, the joint strength of the custom-made solder paste was generally superior to that of the commercial solder paste.

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Statements & Declarations

Funding

This work was supported by the 2023 Ganzhou “Jiebangguashuai” Program (Grant No. 2023ULGX0002) and the Jiangxi Double Thousand Plan (Grant No. jxsq2023101058).

Competing interests

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

Acknowledgements

This work was supported by grant from the 2023 Ganzhou "Jiebangguashuai" Program (NO.2023ULGX0002) and JIANGXI DOUBLE THOUSAND PLAN (No. jxsq2023101058).

Author contributions

Xiaofeng Yang: Writing – review & editing, Writing – original draft, Methodology, Formal analysis.
Haodi Li: Writing – review & editing, Methodology **Dingying Ren:** Writing – review & editing, Validation, Supervision. **Xiaoqiang Hu:** Writing – review & editing, Resources, Project administration.
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.

Ethical approval: Not applicable

Consent to participate: Not applicable

Consent to publish: Not applicable.