



J Adv Ceram 2026, **15**: 9221390

<https://doi.org/10.26599/JAC.2026.9221390>

Research Article

Overcoming the trade-off between electrical conductivity and densification in Al_2O_3 –TiC Ceramics via particle size grading

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Received: 21-Jul-2022; Revised: September 29, 2026; Accepted: October 2, 2026

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Abstract: Electrically conductive composite ceramics face a critical trade-off where increasing the phase domain size ratio between the insulating matrix and the conductive phase theoretically enhances conductivity but practically hinders densification. The resulting porosity inevitably degrades both mechanical and electrical properties. To address this limitation, a matrix particle size grading strategy is proposed using the Al₂O₃–TiC system as a model. The incorporation of nano-Al₂O₃ into a micro-Al₂O₃ matrix triggers a synergistic microstructural evolution. Specifically, nano-Al₂O₃ particles fill interparticle spaces to promote densification and may pin TiC grain boundaries, thereby contributing to the refinement of the conductive phase domains. Concurrently, the micro-Al₂O₃ phase domains coarsen at the expense of nano-Al₂O₃, maximizing the phase domain size disparity and favoring the formation of a continuous three-dimensional TiC network. With 10 vol% nano-Al₂O₃ addition, the calculated effective percolation threshold drops from 15.54 vol% to 10.20 vol%, achieving an electrical conductivity of 4.02×10^4 S/m. Consequently, the composite exhibits enhanced electrical discharge machining (EDM) performance, achieving a material removal rate of 9.50 mm²/min alongside high surface quality. This strategy offers a robust framework for developing advanced conductive ceramics tailored for EDM applications.

Keywords: Al₂O₃–TiC; Particle size grading; Electrical conductivity; Percolation threshold; EDM machinability

1 Introduction

The development of conductive composite ceramics, consisting of an insulating matrix and a dispersed conductive phase, has become a major research frontier in structural–functional ceramics [1–3]. These composites retain the favorable characteristics of traditional structural ceramics, such as high strength and hardness, while exhibiting electrical conductivity [4–6]. This combination enables the efficient machining of highly brittle ceramics via electrical discharge machining (EDM) [7–10], significantly reducing manufacturing costs. Furthermore, conductive ceramics serve as candidates for fabricating electrodes, electrolytic cells, ignition devices, as well as heaters and electromagnetic interference shielding for high-temperature applications [11–15]. They offer superior performance to metallic conductors in extreme environments involving high temperatures, corrosion, and severe wear.

Despite their advantages, a fundamental challenge remains: pursuing low electrical resistivity often compromises mechanical properties. For example, Zhao et al. [16] reported a SiC fiber-derived ceramic with a conductivity of 183 S/m, much higher than 57 S/m for conventional SiC ceramics, but its Vickers hardness was only 17.04 GPa and its flexural strength was 376 MPa. Because both mechanical and electrical performances are governed by the microstructure, a finer matrix grain size generally leads to higher strength, hardness, and fracture toughness. Therefore, ultrafine or nano-sized raw powders are commonly used to achieve a dense, fine-grained microstructure. Fu et al. [17] synthesized nanosized raw powders using a sol gel method and subsequently consolidated them via rapid hot pressing to obtain fully dense ZrO₂–SiO₂ nanocrystalline ceramics.

Conversely, from the perspective of electrical conduction, a larger matrix phase domain size, together with a finer conductive phase domain, favors the preferential segregation of the conductive particles along the matrix phase domain boundaries [18, 19]. This microstructural configuration is favorable for constructing a continuous conductive network and lowering resistivity, which inherently necessitates the use of coarse matrix powders. Specifically, Zhao et al. [20] reported that increasing the initial B₄C particle size from 0.50 μm to 10.29 μm significantly decreased the electrical resistivity of

B₄C–TiB₂ composites (from $\sim 8 \times 10^{-4}$ to $\sim 0.5 \times 10^{-4} \Omega \cdot \text{m}$). Larger matrix phase domains provided smaller total phase domain boundary areas, allowing the dispersed TiB₂ conductive phase to concentrate and interconnect more easily. Meanwhile, Shi et al. [21] demonstrated that completely intergranular-dispersed TiC particles in Al₂O₃–TiC composites, which exhibit a similar size to the matrix grains, effectively lowered the percolation threshold to 10.5% by strengthening interparticle contacts and forming continuous conductive pathways. However, employing large matrix particles severely hinders densification during sintering. The resulting residual pores not only degrade mechanical properties but also act as insulating barriers. Their presence effectively dilutes the volume fraction of the conductive phase, thereby neutralizing the positive effects of the coarse matrix on resistivity reduction. Consequently, a critical trade-off emerges between optimizing electrical conductivity and achieving full densification.

In this study, the Al₂O₃–TiC system is selected as an ideal model to address this limitation. Al₂O₃ is a widely utilized structural ceramic recognized for its excellent mechanical properties [22–24], and its combination with TiC yields composites extensively applied in wear-resistant components and cutting tools [25–30]. Because their extreme hardness makes conventional machining difficult, EDM represents a highly effective processing strategy for Al₂O₃–TiC components [31, 32]. Consequently, the simultaneous optimization of electrical conductivity and densification becomes a critical scientific and industrial objective.

To overcome this dilemma, this study proposes a bimodal matrix design strategy. Specifically, a bimodal Al₂O₃ matrix is designed by incorporating highly active nano-Al₂O₃ particles into a coarse micro-Al₂O₃ skeleton. This architecture provides two synergistic benefits during sintering. First, the high surface energy of the nano-Al₂O₃ significantly promotes the densification of the composite. Second, these nanoparticles may pin the TiC grain boundaries, thereby restricting TiC grain growth and favoring the continuous distribution of the ultrafine conductive phase around the coarse micro-Al₂O₃ phase. Ultimately, this strategy effectively maximizes the size disparity between the insulating

matrix and the conductive phase while simultaneously achieving optimal densification and favorable mechanical performance.

2 Experimental procedure

2.1 Raw materials

Commercially available powders were used as starting materials, including micro- Al_2O_3 (99.99% purity, average particle size 10.81 μm ; Shanghai Chaoqin New Material Technology Co., Ltd., China), nano- Al_2O_3 (99.99% purity, particle size 20–30 nm; Shanghai Chaoqin New Material Technology Co., Ltd., China), and TiC (99.99% purity, average particle size 0.92 μm ; Changsha Langfeng Metal Material Co., Ltd., China). The morphological features of the raw micro- Al_2O_3 and nano- Al_2O_3 powders are shown in the scanning electron microscopy (SEM) images in Fig. S1. In the overall composite system, the TiC content, acting as the conductive phase, was fixed at 25 vol%. The remaining Al_2O_3 matrix was formulated as a bimodal mixture of micro- Al_2O_3 and nano- Al_2O_3 . Specifically, the volume fraction of the nano- Al_2O_3 was systematically set at 0, 5, 10, and 15 vol% relative to the total Al_2O_3 content. Here, 0 vol% corresponds to no addition of nano- Al_2O_3 and serves as the experimental control [33].

2.2 Materials preparation

Given that the electrical percolation threshold of TiC in an Al_2O_3 matrix typically ranges from 20 to 25 vol%, a constant TiC fraction of 25 vol% was adopted in this work to guarantee reliable electrical conductivity for the composites. A schematic illustration of the overall fabrication route is provided in Fig. S2.

For the mixing process, the raw Al_2O_3 and TiC powders were placed into a polyethylene bottle, using Al_2O_3 milling balls and ethanol as the dispersion medium. The mixture was homogenized via ball mixing at 60 rpm for 24 h. Afterward, the resulting slurry was dried under vacuum using a rotary evaporator to rapidly remove excess ethanol. The powder mixture was further dried in an oven at 60 °C for 24 h and subsequently passed through a 200-mesh sieve to yield a uniformly blended dry powder.

The mixed powder was loaded into a graphite die with an inner diameter of 40 mm. Graphite foil was placed between the powder and the inner wall of the die to facilitate sample ejection after sintering. The die assembly was then transferred into a spark plasma sintering furnace (SPS; SPS-20T-10, Shanghai Chen Hua Science and Technology Co., Ltd., Shanghai, China). Sintering was conducted under a vacuum environment, with a uniaxial pressure of 50 MPa applied from the very beginning of the process. The system was heated to 1650 °C at a rate of 100 °C/min and maintained for 5 min. Following the isothermal holding stage, the applied pressure was completely released within 5 min, and the sintered samples were allowed to cool naturally to room temperature within the furnace chamber.

2.3 Characterization

The theoretical density of the composites was calculated according to the rule of mixtures. The bulk density was measured using the Archimedes method, from which the relative density was determined. The microstructural morphology of the composites was observed using a field-emission SEM (Mira, TESCAN, Czech Republic) at an accelerating voltage of 5 kV after sputter-coating with gold. To quantitatively determine the size of phase domains (i.e., contiguous clusters comprising multiple grains of the same phase), the linear intercept method was applied to SEM images using Nano Measurer software. For each sample, at least 200 randomly selected domains were measured to obtain a statistically representative mean size.

The room-temperature electrical resistivity was measured using a four-point probe tester (FPP; HPS2662, Helpass Electronic Technology Co., Ltd., China). For each sample, ten measurements were taken at different positions in the central region, and the average value was calculated. Rectangular bar specimens with dimensions of 1.5 mm × 2 mm × 25 mm were cut from the sintered bodies using wire EDM (SYJ-7720, Hefei Kejing Material Technology Co., Ltd., China). The flexural strength was measured by three-point bending tests on a universal testing machine (UTM; AGS-X, Shimadzu Instruments Manufacturing Co., Ltd., Suzhou, China) with a support span of 20 mm and a crosshead

speed of 0.5 mm/min^[34]. For each composition, five rectangular bar specimens were tested, and the measured values were averaged to obtain the final flexural strength. One of these bars, pre-cut through the center of the sintered disc, was selected for Vickers hardness (H_V) and fracture toughness (K_{IC}) measurements after bending failure. The tests were performed using a Vickers hardness tester (HVS-10, Shanghai Shangcai Testing Machine Co., Ltd., China), and K_{IC} was calculated from indentation crack lengths using the empirical Evans formula^[35]. A load of 19.6 N was applied to the polished sample surfaces with a dwell time of 15 s. The reported H_V and K_{IC} values are arithmetic averages of ten uniformly distributed indentation points, reflecting property variations along the radial direction of the disc.

To assess the EDM machining efficiency, the material removal rate (MRR) was evaluated. The EDM parameters were set as follows: machining voltage of 36 V, pulse-on time of 32 μ s, pulse-off time of 224 μ s, molybdenum wire diameter of 0.18 mm, and a water-based emulsion as the dielectric coolant. The MRR was calculated using the following equation:

$$\text{MRR} = \frac{lh}{t} \quad (1)$$

where l is the length of the wire-cut path, h is the thickness of the machined sample, and t is the machining time.

3 Results and discussion

3.1 Densification

Fig. 1 presents the relative density of the Al_2O_3 -TiC composites with varying nano- Al_2O_3 additions. As the nano- Al_2O_3 content increases up to 10 vol%, the relative density rises from 97.21% to 99.02%. However, upon further increasing the nano- Al_2O_3 addition to 15 vol%, the relative density declines to 97.90%. This non-monotonic trend suggests that a moderate nano- Al_2O_3 addition is beneficial for densification, whereas an excessive addition adversely affects the relative density.

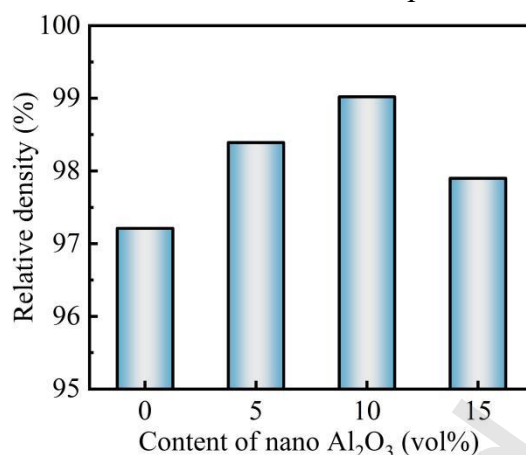


Fig. 1 Effect of nano-Al₂O₃ content on the relative density of the Al₂O₃–TiC composites.

The observed evolution in relative density is hypothesized to be governed by a trade-off between the beneficial packing effect of nano-Al₂O₃ and its potential adverse impact on powder characteristics. At moderate additions, nano-Al₂O₃ fills interstices, improves bimodal packing, and its high surface area accelerates diffusion, lowering the densification activation energy^[36]. Conversely, when the nano-Al₂O₃ content reaches 15 vol%, the optimal particle size gradation is likely disrupted. Excessive nanoparticles increase interparticle friction and attractive forces^[37]. This disruption would hinder particle rearrangement during compaction and early sintering stages, potentially introducing microstructural inhomogeneities that constrain diffusion and reduce the final densification.

3.2 Microstructures

The microstructural evolution and phase domain size distributions of the Al₂O₃–TiC composites are illustrated in Fig. 2. Based on the atomic number contrast in the backscattered electron (BSE) mode, the bright phase is identified as TiC, while the dark region corresponds to the Al₂O₃ matrix (the absence of impurity phases was further confirmed by XRD, Fig. S3).

In the baseline composite without nano-Al₂O₃ (Fig. 2(a–c)), the TiC phase tends to form coarse clusters at the Al₂O₃ grain boundary junctions. The limited phase domain size disparity between Al₂O₃ (7.08 μm) and TiC (3.33 μm) restricts the effective spreading of the TiC phase along the boundaries. Consequently, the initial spatial distribution of TiC is relatively discontinuous and coarse.

As the nano- Al_2O_3 content increases from 5 to 10 vol%, the microstructure undergoes a distinct evolution, appearing to transition into a continuous three-dimensional (3D) skeleton. At 5 vol% (Fig. 2(d–f)), the TiC phase begins to spread along the Al_2O_3 grain boundaries, developing a chain-like morphology. This spreading is driven by an amplified size disparity where the Al_2O_3 phase domain size coarsens to $8.90\ \mu\text{m}$, while the TiC phase domain size refines to $3.21\ \mu\text{m}$. When the nano- Al_2O_3 content reaches 10 vol% (Fig. 2(g–i)), the Al_2O_3 phase domain size further increases to $9.52\ \mu\text{m}$ and the TiC phase domain size decreases to $2.89\ \mu\text{m}$.

The progressive coarsening of Al_2O_3 phase domains with increasing nano- Al_2O_3 addition is attributed to curvature-induced grain growth via the Gibbs–Thomson effect. Due to their high specific surface area and chemical potential, the highly active nano- Al_2O_3 particles are expected to create a strong potential gradient during sintering. This gradient drives rapid solid-state mass transport, mainly via grain boundary and surface diffusion. As a result, the nano- Al_2O_3 particles are assimilated into the coarser micro- Al_2O_3 grains to minimize interfacial free energy^[17, 38]. Conversely, the refinement of TiC phase domains is primarily attributed to the Zener pinning. Nano- Al_2O_3 particles dispersed along TiC grain boundaries are expected to exert a pinning pressure that counteracts grain boundary migration, thereby restricting TiC coarsening^[39, 40]. It should be noted that the SEM observations are two-dimensional (2D), whereas the proposed 3D conductive network and the associated mechanisms (e.g., Zener pinning and grain-boundary connectivity) are inferred from these observations and the measured electrical properties.

However, increasing the nano- Al_2O_3 content to 15 vol% (Fig. 2(j–l)) disrupts this optimal architecture. The TiC phase domain size further refines to $2.68\ \mu\text{m}$ due to the enhanced pinning by the abundant nano- Al_2O_3 . These refined TiC grains also pin the Al_2O_3 grain boundaries, limiting the Al_2O_3 phase domain size to $7.71\ \mu\text{m}$. In addition, excessive nano- Al_2O_3 causes severe agglomeration, forming Al_2O_3 grain clusters after sintering (Fig. S4). This agglomeration reduces the number of well-dispersed nanoparticles available to fill the gaps between micro- Al_2O_3 particles, degrading the particle

size gradation and lowering packing efficiency. Consequently, sinterability decreases, leading to the observed reduction in relative density.

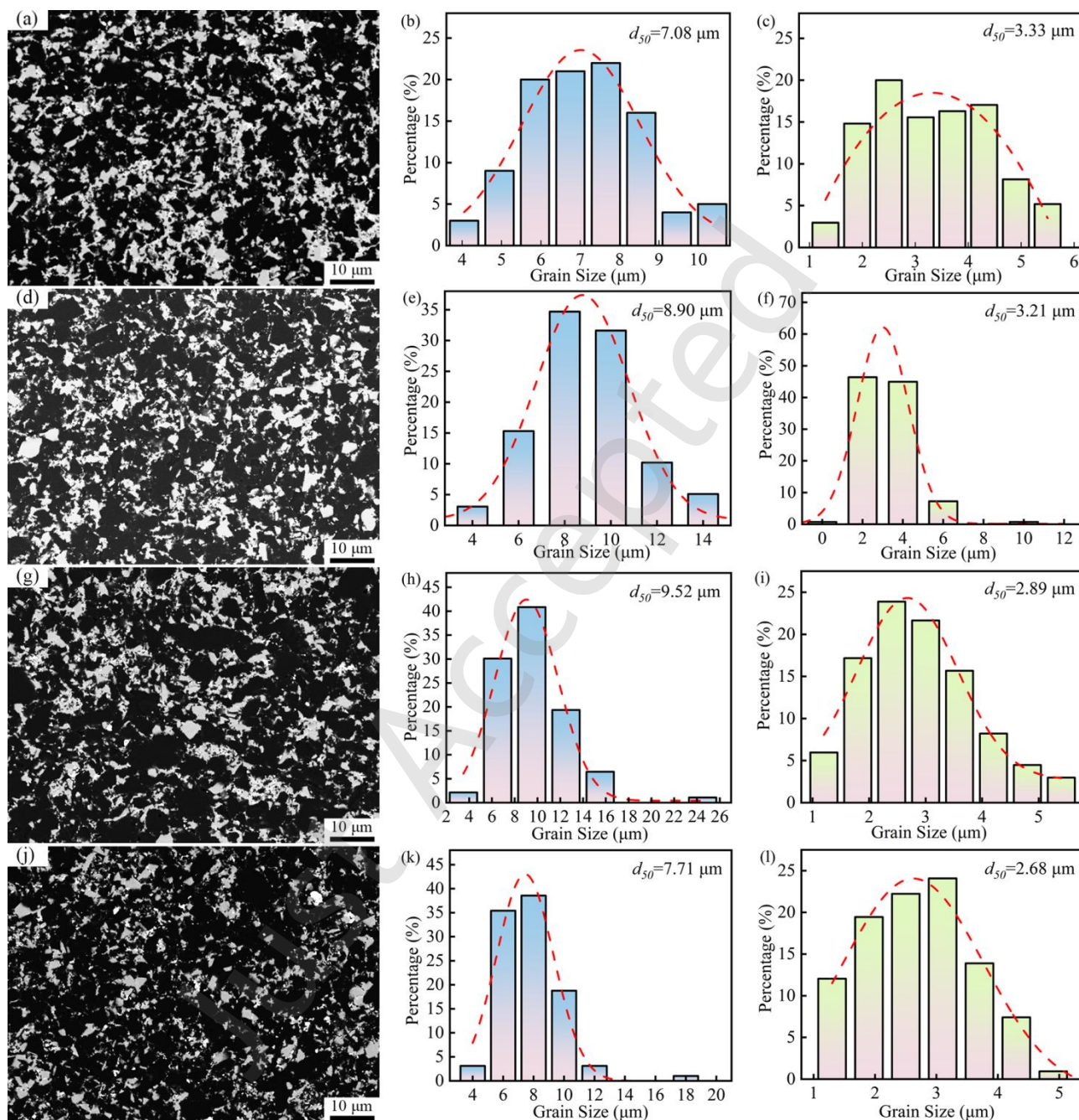


Fig. 2 (a, d, g, j) BSE images of polished surfaces and the corresponding phase domain size distributions of (b, e, h, k) Al₂O₃ and (c, f, i, l) TiC for the composites containing (a–c) 0 vol%, (d–f) 5 vol%, (g–i) 10 vol%, and (j–l) 15 vol% nano-Al₂O₃.

3.3 Mechanical properties

Fig. 3 presents the flexural strength, Vickers hardness, and fracture toughness of the Al_2O_3 –TiC composites. The variations in these mechanical properties are closely related to the densification behavior of the materials. As the nano- Al_2O_3 content increases from 0 to 10 vol%, the flexural strength rises from 391 to 450 MPa, and the fracture toughness reaches a maximum of $3.50 \text{ MPa}\cdot\text{m}^{1/2}$, consistent with improved relative density. However, when the nano- Al_2O_3 addition is further increased to 15 vol%, both the strength (417 MPa) and hardness (18.49 GPa) decline, which may be attributed to residual pores caused by reduced densification.

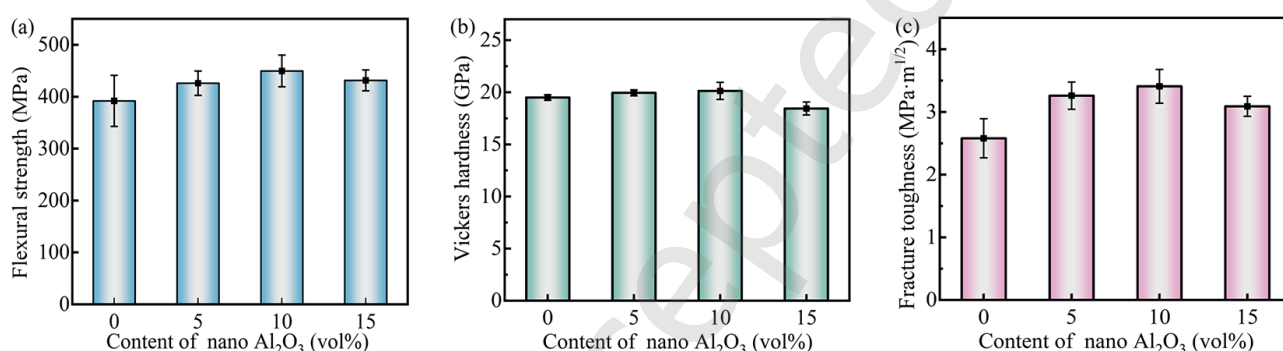


Fig. 3 Mechanical properties of the Al_2O_3 –TiC composites as a function of nano- Al_2O_3 content. (a) Flexural strength. (b) Vickers hardness. (c) Fracture toughness.

The underlying toughening mechanisms were investigated by examining fracture surface morphologies (Fig. 4(a–c)) and crack propagation paths (Fig. 4(d)). In the baseline composite without nano- Al_2O_3 (Fig. 4(a)), the Al_2O_3 grains exhibit smooth cleavage facets typical of transgranular fracture, while the TiC particles display a sugar-like morphology characteristic of intergranular fracture. Upon nano- Al_2O_3 incorporation, the highly active nanoparticles are considered to promote intergranular bonding and effectively bridge adjacent micron-sized Al_2O_3 grains during sintering. As revealed in Fig. 4(b) and its high-magnification counterpart (Fig. 4(e)), the Al_2O_3 grains exhibit two distinct transgranular fracture morphologies: smooth facets and stepped cleavage planes. The crack propagation path in Fig. 4(d) reveals the main toughening mechanisms. When a crack meets TiC and Al_2O_3 grain clusters, it bends along their internal interfaces, leading to clear crack deflection and

bridging. As shown in Fig. 4(f), cracks mainly grow through the Al_2O_3 matrix by transgranular fracture, but switch to intergranular fracture when encountering TiC and Al_2O_3 clusters, which agrees with the SEM observations in Fig. S4. The synergistic activation of these fracture modes substantially increases the crack path tortuosity and dissipates additional fracture energy, thereby yielding a marked improvement in the fracture toughness.

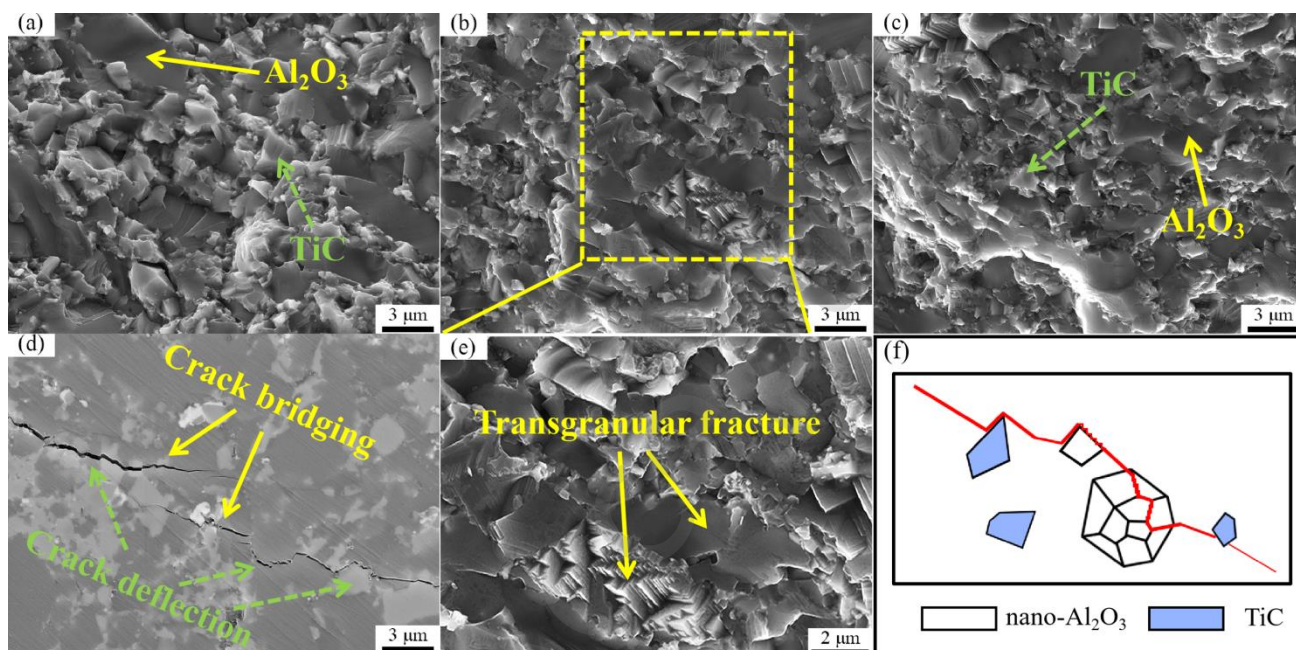


Fig. 4 Fracture behavior and toughening mechanisms of the Al_2O_3 –TiC composites. Fracture surfaces of composites with (a) 0 vol%, (b) 10 vol%, and (c) 15 vol% nano- Al_2O_3 . (d) Crack propagation path and (e) magnified view of (b) for the 10 vol% composite. (f) Schematic of crack deflection through Al_2O_3 grain clusters.

3.4 Electrical conductivity

Fig. 5(a) presents the measured electrical conductivity of the Al_2O_3 –TiC composites as a function of nano- Al_2O_3 content. As the nano- Al_2O_3 addition increases from 0 vol% to 10 vol%, the electrical conductivity gradually increases from 2.95×10^4 S/m to 4.02×10^4 S/m. However, when the nano- Al_2O_3 addition is further increased to 15 vol%, the conductivity drops to 3.27×10^4 S/m.

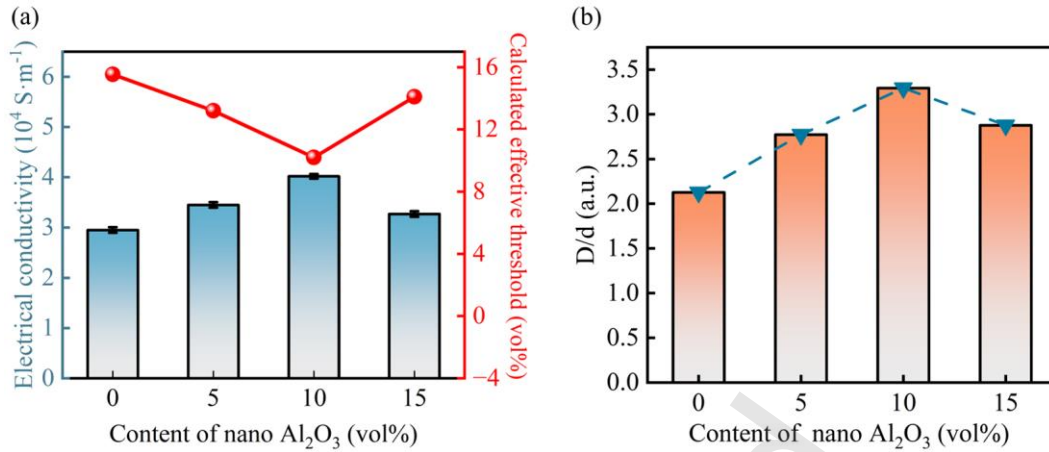


Fig. 5 (a) Measured electrical conductivity and calculated effective threshold, and (b) statistical size ratio (D/d) as a function of nano-Al₂O₃ content.

As shown in Fig. 5(a), the electrical conductivities of all composites exceed 10⁴ S/m, which confirms that the conductive phase content is already well above the calculated effective threshold. To verify whether the conductivity variations were caused by density fluctuations, the porosity-corrected TiC volume fraction was calculated as $V_p^* = V_p \times \rho$, where V_p represents the nominal volume fraction of the conductive phase and ρ is the relative density [41]. The actual TiC contents for composites with different nano-Al₂O₃ additions are 24.30%, 24.60%, 24.76%, and 24.48%, respectively. Because conductivity is insensitive to such minor content changes (<0.5%) in the post-percolation regime, density fluctuations cannot explain the observed conductivity differences. Therefore, the enhanced conductivity is likely to stem from a reduction in the calculated effective threshold driven by microstructural evolution.

The experimental data were thus fitted using the general effective medium (GEM) equation proposed by McLachlan et al. [42]

$$V_p \frac{\sigma_p^{1/t} - \sigma_c^{1/t}}{\sigma_p^{1/t} + \frac{1-\varphi_c}{\varphi_c} \sigma_c^{1/t}} + (1 - V_p) \frac{\sigma_m^{1/t} - \sigma_c^{1/t}}{\sigma_m^{1/t} + \frac{1-\varphi_c}{\varphi_c} \sigma_c^{1/t}} = 0 \quad (2)$$

where σ denotes electrical conductivity, with subscripts p , m and c referring to the conductive phase, matrix and composite, respectively; φ_c is the calculated effective threshold and t is the critical

percolation exponent. In the classical 3D random percolation theory, $t = 2.0$ is indeed the universal value for a statistically distributed 3D network [43]. This universal value is strictly valid only for idealized systems where the conducting phase is randomly dispersed throughout the insulating matrix and the conductive pathways are formed through statistical contacts between isolated particles. However, in the present experiment, the conductive TiC phase preferentially segregates along the Al_2O_3 grain boundaries. This specific distribution may naturally lead to a reduction in the t value.

By fitting the experimental data within a physically reasonable range of the critical exponent [44, 45], $t = 0.8$ yielded the best statistical agreement among the tested values, with the lowest mean relative error (MRE) and the highest coefficient of determination (R^2), as detailed in Tables S1 and S2. However, the differences in fitting statistics are very small, and the available data do not necessarily establish $t = 0.8$ as the unique physical exponent for this system. In composites, critical exponents deviating from the universal value ($t = 2.0$) are well documented, and values of $t < 1.0$ have been attributed to the interplay between the connectivity and the structural variations that follow the increase of the fractional volume content of the conducting phase [46, 47]. Physically, the selected $t = 0.8$ provides a plausible rationale consistent with the microstructural characteristics of the present Al_2O_3 –TiC system. Rather than being isolated spheres dispersed randomly, the conductive TiC grains are predominantly segregated along the matrix grain boundaries (Fig. 2 and Fig. S4). This microstructural arrangement may facilitate the rapid establishment of conductive pathways once the percolation threshold is exceeded, so that the conductivity rises sharply with only a small increment in conductive phase content near the percolation threshold, thereby possibly contributing to a low critical exponent t .

The model-derived effective threshold φ_c is plotted in Fig. 5(b). Since all specimens contain 25 vol% TiC, a content above the estimated percolation threshold, the reported φ_c values are obtained from conductivity data at a single TiC concentration and should not be taken as direct experimental measurements of the percolation transition. The theoretical results reveal that as the nano- Al_2O_3

addition increases to 10 vol%, φ_c decreases from 15.54 vol% to 10.20 vol%, before rising to 14.10 vol% at 15 vol% addition. To contextualize these results, the obtained φ_c range (10.20–15.54 vol%) is in good overall agreement with previously reported percolation thresholds for the Al₂O₃–TiC system. For instance, Ahmad^[30] fabricated Al₂O₃–TiC composites via SPS and reported a percolation threshold of ~ 12.5 vol%. Furthermore, Shi et al.^[21] systematically compared the effects of different processing routes: the in situ reaction synthesis using carbon nanotubes as the carbon source yielded a low threshold of 11.2 vol%, owing to the complete grain-boundary dispersion of TiC particles and their similar size to the Al₂O₃ grains. In contrast, conventional powder mixing with graphite resulted in higher thresholds of 14.6 vol% and 16.5 vol%, respectively. Notably, in the present study, tailoring the microstructure with an optimal 10 vol% nano-Al₂O₃ effectively reduces φ_c to 10.20 vol%, which is lower than the lowest threshold reported to date for this system^[21].

As shown in Section 3.2, the addition of nano-Al₂O₃ causes matrix coarsening and TiC refinement, which enlarges the phase domain size disparity. This disparity is one of several factors that may influence percolation behavior. To quantify this structural effect, the phase domain size ratio (D/d , where D is the matrix phase domain size and d is the conductive phase domain size) is plotted against the calculated effective percolation threshold. As summarized by Lux^[48], this relationship can be generally expressed as $V_c \propto 1/(1+D/d)$, suggesting that a larger size disparity may contribute to lowering the critical percolation concentration. As shown in Fig. 5(b), the D/d ratio increases with nano-Al₂O₃ up to 10 vol%, which is correlated with the decreased φ_c (Fig. 5(a)). The larger size disparity may facilitate finer TiC phase domains to segregate along coarsened Al₂O₃ phase domain boundaries, thereby possibly contributing to a continuous conductive path at a lower critical concentration. However, it should be emphasized that density, TiC distribution, TiC phase domain morphology, pore population, and matrix microstructure also change simultaneously with D/d . Therefore, the observed D/d – φ_c correlation should be regarded as a qualitative interpretation rather than independent proof that D/d alone controls the percolation threshold.

The intrinsic correlation between the aforementioned evolution of electrical properties and microstructural characteristics is visually depicted by the schematic illustrations presented in Fig. 6. In the composite without nano- Al_2O_3 addition (Fig. 6(a)), the limited phase domain size disparity may prevent the TiC phase from forming a continuous conductive network along the Al_2O_3 phase domain boundaries. Furthermore, without the filling effect of nano- Al_2O_3 , the microstructure contains a noticeable amount of residual pores. These pores may act as physical barriers that separate the TiC phase domains, leaving many of them isolated or disconnected. According to percolation theory, such a randomized distribution, compounded by pore-induced discontinuities, means that a higher volume fraction of TiC is required to bridge the gaps and form an interconnected network. This structural isolation leads to frequent interruptions in the conductive pathways, resulting in a high calculated effective threshold and relatively low electrical conductivity.

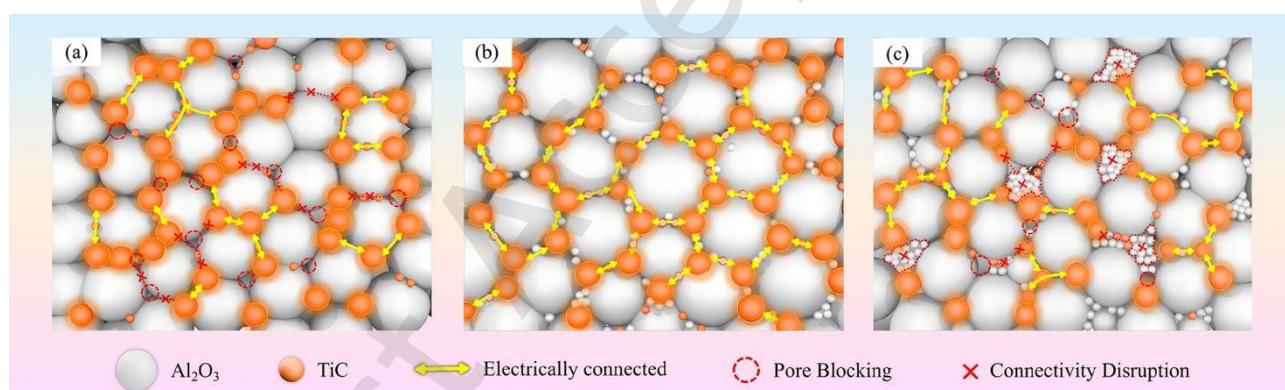


Fig. 6 Schematic illustrations of the conductive networks in composites containing (a) 0 vol%, (b) 10 vol%, and (c) 15 vol% nano- Al_2O_3 .

Upon the incorporation of an optimal amount (10 vol%) of nano- Al_2O_3 (Fig. 6(b)), the microstructural topology appears to evolve from a fragmented state into an interconnected network. The TiC phase develops a nearly continuous distribution along the Al_2O_3 phase domain boundaries. The 3D interconnectivity of the conductive network is considered to be substantially improved, providing abundant, low-resistance channels for electron transport. This is consistent with the mechanism discussed in Section 3.2: the consumption of active nano- Al_2O_3 promotes matrix

coarsening and enlarges D/d , which may favor phase domain boundary segregation of TiC. In addition, the filling effect of nano- Al_2O_3 improves densification and reduces residual pores, further perfecting the conductive network.

However, when the nano- Al_2O_3 addition exceeds the optimal threshold, the electrical properties deteriorate. Excess nano- Al_2O_3 tends to agglomerate during sintering, forming micro-sized insulating agglomerates (Fig. 6(c)). These agglomerates are embedded within the TiC conductive network and act as high-resistivity “insulating islands” that disrupt the continuous TiC pathways. In addition, excessive nano- Al_2O_3 may cause suboptimal particle packing and residual pores. Together with the agglomerates, these insulating voids compromise the 3D interconnectivity of the TiC phase. Electrons must detour or tunnel through these barrier regions, which increases transport impedance and degrades the macroscopic electrical conductivity.

3.5 EDM Machinability

Fig. 7(a) illustrates the effect of nano- Al_2O_3 content on the MRR of the composites. As the nano- Al_2O_3 addition increases from 0 to 15 vol%, the MRR first increases and then decreases. It rises from 8.78 mm^2/min to a peak of 9.50 mm^2/min at 10 vol% before decreasing to 8.93 mm^2/min . This enhancement is consistent with the improved conductive network discussed in Section 3.4. The continuous TiC pathway along the Al_2O_3 phase domain boundaries is expected to stabilize discharge channels, may increase the effective discharge frequency and pulse energy utilization, and may help suppress localized arc concentration, thereby contributing to the higher MRR ^[49].

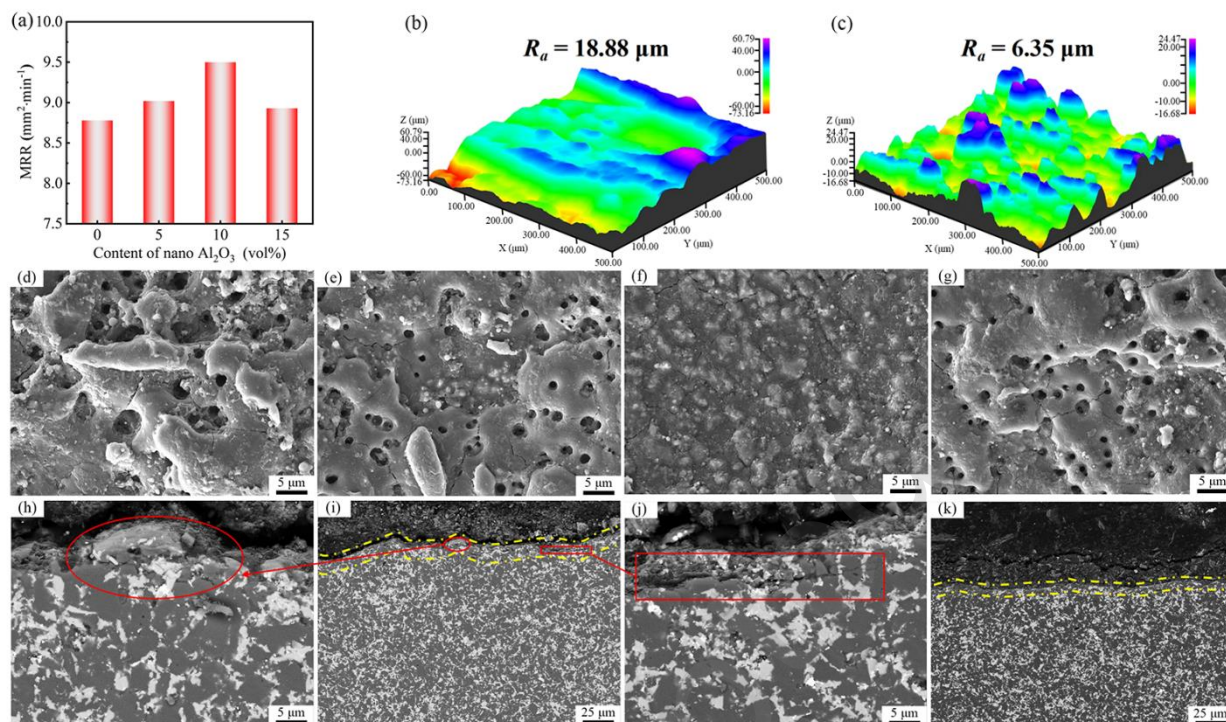


Fig. 7 (a) MRR; (b, c) 3D surface profiles of the 0 and 10 vol% composites; (d–g) SEM surface morphologies of the 0, 5, 10, and 15 vol% composites; (i) cross-sectional morphology of the 0 vol% composite and (h, j) its local magnified views; (k) cross-sectional morphology of the 10 vol% composite.

The 3D surface morphologies and roughness values are presented in Fig. 7(b, c). The composite without nano- Al_2O_3 addition (Fig. 7(b)) exhibits deep discharge craters and sharp ridges, yielding a high surface roughness of $18.88 \mu\text{m}$. In contrast, the 10 vol% nano- Al_2O_3 composite (Fig. 7(c)) displays a smoother profile, with the roughness reduced to $6.35 \mu\text{m}$. This reduction is likely related to the enhanced electrical conductivity, which may promote homogeneous spark discharges and may reduce the formation of deep localized craters. In addition to the electrical contribution, the improved mechanical properties also play a role. The near-full density (99.02%) and the combination of good flexural strength (450 MPa) and fracture toughness ($3.50 \text{ MPa} \cdot \text{m}^{1/2}$) are expected to enhance the material's resistance to thermal shock and mechanical impact during spark discharges, reducing the

incidence of premature localized fracture that would otherwise interrupt the continuous removal process.

The SEM surface morphologies are presented in Fig. 7(d–g). For the samples with 0 and 5 vol% nano- Al_2O_3 (Fig. 7(d, e)), the surfaces exhibit typical EDM defects, including spherical or flake-shaped recast debris, overlapping arc craters, and microcracks. These defects are likely caused by insufficient densification and low conductivity. The low relative density (97.21%) introduces residual pores that act as stress concentration sites under the thermal shock of discharge, leading to localized fragmentation and spalling [50]. Poor conductivity may also impede stable energy transfer, preventing the molten material from adequately filling the eroded craters [51, 52]. In addition, the low fracture toughness provides limited resistance to crack propagation under thermal stress. Together, these factors may result in severely degraded surface quality.

Conversely, the 10 vol% composite (Fig. 7(f)) presents a relatively smooth and flat machined surface. Its densified microstructure is expected to resist localized thermal fragmentation, allowing material removal to proceed in a controlled melting and vaporization manner. The high flexural strength and fracture toughness further contribute to suppressing crack initiation and propagation under the cyclic thermal stress of pulsed discharges [53]. The improved conductivity may sustain continuous spark discharges, which may enable the molten phase to uniformly spread and heal micro-defects [2]. The synergistic effect of superior electrical conductivity and mechanical properties is considered to be responsible for the best surface quality among all compositions. However, when the nano- Al_2O_3 content reaches 15 vol% (Fig. 7(g)), nanoparticle agglomeration deteriorates both the sintering densification and the conductive network continuity. Consequently, the diminished spark stability and reduced mechanical integrity cause the machined surface to revert to a rough morphology with deteriorated surface quality.

The cross-sectional SEM morphologies of the base material after EDM cutting further highlight the benefits of nano- Al_2O_3 modification. As shown in Fig. 7(i), the 0 vol% composite exhibits a tortuous

cross-sectional profile, indicative of unstable material removal. Local magnified views of this profile (Fig. 7(h, j)) reveal severe thermal damage, including arc-shaped spalling edges and cracks propagating parallel to the machining direction. These defects act as critical stress concentrators that severely degrade the structural reliability of the base material [54].

In contrast, the 10 vol% composite (Fig. 7(k)) exhibits a straight and uniform cross-sectional profile. The optimized conductive network within this sample is expected to provide stable energy input and may reduce localized overheating, thereby contributing to suppressed thermal shock damage. Meanwhile, the high fracture toughness and flexural strength are expected to suppress crack initiation and propagation under thermal cycling, preserving the structural integrity of the machined subsurface [55]. Consequently, the optimal incorporation of nano- Al_2O_3 enhances the precision and stability of the EDM process, resulting in high machined surface quality and structural integrity.

4 Conclusion

In this study, the trade-off between electrical conductivity and densification in Al_2O_3 –TiC ceramics was overcome via a bimodal particle size grading strategy. At the optimal 10 vol% nano- Al_2O_3 addition, the relative density of the composite was elevated from 97.21% to 99.02%. While near-full densification ensures high mechanical strength, the enhancement in electrical properties is mainly attributed to the microstructural evolution induced by nano- Al_2O_3 addition, which enlarges the matrix-to-conductive phase domain size ratio (D/d) from 2.13 to 3.29 and favors the formation of a continuous 3D conductive network. Consequently, the calculated effective threshold decreased from 15.54 vol% to 10.20 vol%, and the electrical conductivity increased from 2.95×10^4 S/m to 4.02×10^4 S/m. Ultimately, this combined reduction in porosity and enhancement in electrical conductivity resulted in favorable EDM machinability, with the material removal rate increasing from 8.78 mm²/min to 9.50 mm²/min and the surface roughness decreasing from 18.88 μm to 6.35 μm .

Electronic Supplementary Material: Supplementary material is available in the online version of this article

Acknowledgements

This work was supported by the National Natural Science Foundation of China (Nos. 52472061 and 52072003), the Natural Science Foundation of Anhui Provincial Education Department (No. 2024AH040031).

Author contributions

Liangliang Liu: Investigation, Data curation, Writing-original draft, Formal analysis, Validation, Visualization. Xiaofeng Jia: Investigation, Formal analysis, Data curation. Lei Zhao: Investigation, Formal analysis. Dong Wang: Methodology, Resources. Xing Jin: Methodology, Resources. Songlin Ran: Conceptualization, Resources, Methodology, Writing-review & editing, Project administration, Supervision. Zhaoyang Wu: Conceptualization, Validation, Resources, Writing-review & editing.

Availability of data and materials

The data that support the findings of this study are available from the corresponding author upon reasonable request.

Competing interests

The authors have no competing interests to declare that are relevant to the content of this article.

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Just Accepted