

Article

Fabrication of High-Performance Porous Thermal Insulating Ceramics via High-Proportion Utilization of Industrial Solid Wastes

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Abstract

Large-scale high-value utilization of industrial solid waste is urgently required worldwide. Porous ceramics prepared from single-type solid waste face great limitations in balancing mechanical properties and thermal insulation performance. In this work, various types of industrial solid wastes—red mud, waste electric porcelain, and coal gangue—were fully adopted as primary raw materials, while SiC was employed as a foaming agent to prepare porous ceramics. Results show that at 3% SiC addition and 1140 °C sintering temperature, the apparent porosity of the ceramics reaches to 21.8%, with thermal conductivity of 0.08 W/(m·K). Moreover, favorable pore-size distribution and desirable crystalline phases (Mg–Fe–Al spinel and sodium calcium feldspar) are obtained under this optimal condition, accompanied by a favorable compressive strength of 2.91 MPa, making it a promising low-cost, high-performance high-temperature insulation material.

Keywords: porous ceramics; coal gangue; waste electric porcelain; red mud; SiC content; sintering temperature

1. Introduction

As industry develops, both the categories and output of solid waste have been increasing constantly, resulting in growing severity of environmental pollution. Red mud, waste electric porcelain, and coal gangue are three typical bulk industrial solid wastes [1–6]. Red mud is a strongly alkaline industrial by-product generated during alumina extraction from bauxite, featuring high iron content and abundant aluminosilicate components, but suffers from strong alkalinity [2]. Waste electric porcelain is a high-aluminum and high-silicon waste ceramic material with outstanding high-temperature resistance, outstanding mechanical strength, and excellent chemical inertness [7]. Coal gangue is massively discharged during coal mining and coal washing, which is rich in silica and alumina raw materials and low in cost [8]. All these are restricted by large compositional fluctuation and inconsistent physicochemical properties, making it a challenge to realize their large-scale and efficient utilization.

The large-scale and high-value utilization of industrial solid wastes is mainly focused on building materials, high-temperature refractory materials, and subgrade filling geotechnical engineering [2,9,10]. Among them, porous ceramics, which act as a kind



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of multifunctional high-temperature resistant thermal insulation material, represent a key route for effectively massive disposal and high-value utilization of industrial solid wastes [11–13].

Li [14] prepared closed-pore foam ceramics by a high-temperature microfoaming method using coal gangue solid waste as the main raw material at a dosage of 75 wt.%, which has a low density of 0.22 g/cm^3 and exhibits a superior thermal conductivity of $0.11 \text{ W/(m}\cdot\text{K)}$ with a high closed porosity of 83.95%. Liu [15] proposed a porous thermal insulation material made from waste electric porcelain at a content of 40 wt.%. This porous lightweight refractory material has a bulk density of 0.99 g/cm^3 and an apparent porosity of 59%. Ren [16] utilized coal gangue as the primary raw material at a content of 75% to prepare porous ceramics. The bulk density reached as low as 0.414 g/cm^3 , with a compressive strength of 1.64 MPa and a closed porosity as high as 74.86%. These reported studies have verified that industrial solid wastes can be effectively utilized as raw materials to fabricate porous ceramics, whose comprehensive performances satisfy the basic application requirements for high-temperature kilns, high-temperature catalyst carriers, high-temperature phase-change, heat storage support materials, etc. Nevertheless, to sustain the fundamental performance, the dosage of individual industrial solid waste is constrained. In addition, the introduction of solid waste generally leads to a significant decline in mechanical properties, especially compressive strength. Combined utilization of multi-component solid wastes with different phases represents a viable alternative approach.

Ge et al. [17] used chromium slag and coal bottom ash as raw materials to prepare closed-cell foam ceramics with a bulk density of 0.191 g/cm^3 and a compressive strength of 1.3 MPa. Recently, more and more attention has been paid to all-solid waste based on the intrinsic chemical reactions of multiple solid wastes. Liu [18] proposed a self-foaming ceramic by using secondary aluminum ash, spodumene flotation tailings, and kaolin tailings as raw materials, exhibiting porosity of 83.29%, bulk density of 0.43 g/cm^3 , compressive strength of 2.46 MPa, and thermal conductivity of $0.076 \text{ W/(m}\cdot\text{K)}$. Zhao [19] and Weng [20] prepared porous ceramics by using fly ash, red mud, and silica fume as raw materials, together with SiC such as a foaming agent, demonstrating excellent mechanical properties and thermal insulation. However, solid wastes such as coal gangue and waste electric porcelain contain low contents of reactive silica and alumina glass phases, feature large particle size and inferior specific surface area, making them exhibit inert reactivity than fly ash in the solid-phase reaction system of porous ceramics, thereby severely restricting the resource disposal efficiency of such inert solid wastes [21–23].

Moreover, foaming agents play an irreplaceable role in pore formation. The variations in the quantity and viscosity of the liquid phase during the high-temperature foaming stage were identified as primary factors affecting pore structure and performance. Additionally, the size and morphology of the crystalline phases significantly influenced pore growth behavior [17]. In reported porous ceramic preparation methods, SiC serves as a superior foaming agent. During sintering, it reacts with oxygen/oxides in the ceramic matrix to release gases, thereby forming uniform and stable pore structures [16,24]. The unreacted SiC micropowder significantly enhances the mechanical properties of the porous ceramic. Furthermore, SiC endows the product with stable chemical properties, excellent thermal insulation performance, and strong thermal shock resistance, effectively extending its service life [23,25,26].

In this work, all-solid-waste-based porous ceramics were fabricated by leveraging the complementary chemical compositions and mineral characteristics of red mud, waste electric porcelain, and coal gangue. The reasonable compounding of the three solid wastes can effectively balance component proportions, optimize sintering behavior, and compensate for the performance deficiencies of single raw materials. Meanwhile, light-burned

magnesium oxide is added as an additive to modulate the liquid-phase behavior during the sintering process, promote the formation of the Mg–Fe–Al spinel phase, and optimize the mechanical strength and thermal stability of the porous ceramic. Furthermore, it was systematically investigated how silicon carbide (SiC) as a foaming agent and sintering temperature affect the microstructure, mechanical properties, and thermal insulation performance of the prepared porous ceramics. The results demonstrate an effective strategy that can remarkably lower the production costs for porous ceramics and provide a feasible pathway for the high-value resource fabrication of high-performance all-solid-waste-based porous ceramics.

2. Materials and Methods

2.1. Raw Materials

The raw materials were composed of coal gangue (particle size < 200 mesh; Lingshou Jinchang Mineral Products Processing Co., Ltd., Lingshou, China), waste electric porcelain (particle size < 200 mesh; Zhonglian Jiaotai Co., Ltd., Beijing, China), red mud (Lingshou Yixin Mineral Products Processing Co., Ltd., Lingshou, China), and light-burned magnesium oxide (particle size < 200 mesh, laboratory reagent, LR; Shenyang Tianyi Refractory Materials Co., Ltd., Shenyang, China). Polyvinyl alcohol (PVA, Aladdin, Shanghai, China) was applied as the binder, while silicon carbide (Tianjin Yetai/Sanhui, Tianjin, China) was used as the foaming agent.

The detailed chemical compositions of raw materials were summarized in Table 1. Coal gangue contains comparable amounts of Al_2O_3 (34.05 wt.%) and SiO_2 (32.55 wt.%), together with 14.45 wt.% Fe_2O_3 . The Al_2O_3 – SiO_2 system constitutes the main ceramic-forming oxides, while Fe_2O_3 acts as a flux to reduce sintering temperature. Waste electric porcelain is rich in Al_2O_3 (52.13 wt.%) and SiO_2 (28.46 wt.%), which serve as the primary raw materials for mullite formation, endowing the product with excellent mechanical strength and thermal stability. Moderate K_2O (5.87 wt.%) further facilitates sintering by lowering the system's melting temperature. Red mud is characterized by a high content of Fe_2O_3 (59.28 wt.%), together with Al_2O_3 (13.11 wt.%) and CaO (10.94 wt.%). The abundant Fe_2O_3 acts as an effective flux to promote low-temperature liquid-phase sintering, while Al_2O_3 and CaO contribute to rigid ceramic skeletons and improved mechanical strength. Light-burned MgO is dominated by MgO (73.59 wt.%), with minor CaO (7.59 wt.%) and SiO_2 (6.44 wt.%). MgO reacts with Al_2O_3 to form magnesium aluminate spinel, enhancing the material's thermal stability and alkali resistance, while CaO serves as an auxiliary flux to promote sintering densification [25,27–30].

Table 1. Chemical compositions of coal gangue, waste electric porcelain, red mud, and light-burned MgO (wt.%).

Component	Mass Fraction/wt.%									
	K_2O	MgO	Al_2O_3	SiO_2	Fe_2O_3	P_2O_5	CaO	MoO_3	TiO_2	Others
Coal gangue	3.618	1.157	34.047	32.545	14.446	0.1	6.442	2.535	1.841	3.269
Waste electric porcelain	5.870	1.157	52.126	28.462	4.458	0.039	1.739	1.496	2.041	2.612
Red mud	1.104	2.304	13.113	7.553	59.281	0.142	10.944	0.511	3.797	1.251
Light-burned MgO	0.263	73.589	1.447	6.440	5.262	0.055	7.590	2.858	0.561	1.935

2.2. Experimental Procedure

All raw materials were subjected to crushing and screening pretreatment prior to use. Specifically, coal gangue and waste electric porcelain were first coarsely crushed via a jaw crusher (JZGJ-100-1, Hebi Jingzhong Technology Co., Ltd., Hebi, China). After sieving

through a 200-mesh sieve, homogeneous raw material powders with uniform particle size were obtained. Commercial light-burned magnesium oxide and red mud were used directly after 200-mesh sieving without additional crushing treatment. The particle size of the final mixed powders was controlled below 75 μm to guarantee uniform component distribution and adequate sintering reactions.

The base matrix was formulated with waste electric porcelain, red mud, coal gangue, and light-burned magnesium oxide at a fixed mass ratio of 4:4:1:1. Based on this matrix formulation, SiC was introduced as a foaming agent at mass fractions ranging from 1 wt.% to 5 wt.% of the matrix, in increments of 1 wt.%. All powder mixtures were homogenized in a mortar. Subsequently, the 5% PVA solution was added to the powder mixture at 8 wt.% of the matrix. The mixture was fully ground in an agate mortar for 10–15 min to form a uniform PVA coating on the particle surface.

A total of 7 g of the homogenized composite powder was weighed and loaded into a cylindrical stainless steel mold with an inner diameter of 20 mm. Cylindrical green samples with a diameter of 20 mm and a height of approximately 10 mm were fabricated using a powder tablet press (YP-30T, Hefei Kejing Material Technology Co., Ltd., Hefei, China) under a uniaxial pressure of 15 MPa for a holding time of 90 s. Five parallel samples were prepared for each formula to ensure the reliability and reproducibility of experimental results. The as-pressed green samples were dried in an electric constant-temperature blast drying oven (101-2AB, Shanghai Peiyin Laboratory Instrument Co., Ltd., Shanghai, China) at 80 °C for 10 h. The dried green bodies were cooled naturally in a dryer and then sealed for standby storage.

A gradient heating sintering procedure was adopted for sample treatment. The temperature was first increased from room temperature to 800 °C at a heating rate of 10 °C/min and maintained for 30 min to eliminate organic residues and decompose carbonate components. Subsequently, the temperature was raised to 1000 °C at 5 °C/min with a 30 min holding stage. Finally, the temperature was elevated to the target sintering temperature (1135 °C, 1140 °C, 1145 °C) at 5 °C/min and held for 3 h. After sintering completion, the samples were cooled to 800 °C at a cooling rate of 5 °C/min, followed by furnace cooling to room temperature. The sintering temperature of all SiC-modified samples was fixed at 1140 °C, and the SiC content of the temperature series samples was fixed at 3 wt.%. The main preparation process of this experiment is shown in Figure 1. The detailed compositions and corresponding labels of samples with different SiC contents are listed in Table 2.

Table 2. Specimen formulations.

Sample Number	Waste Electric Porcelain/wt.%	Red Mud/wt.%	Coal Gangue/wt.%	Light-Burned MgO	SiC (Addition)/wt.%	PVA (Addition)/wt.%
S1	40	40	10	10	1	8
S2	40	40	10	10	2	8
S3	40	40	10	10	3	8
S4	40	40	10	10	4	8
S5	40	40	10	10	5	8

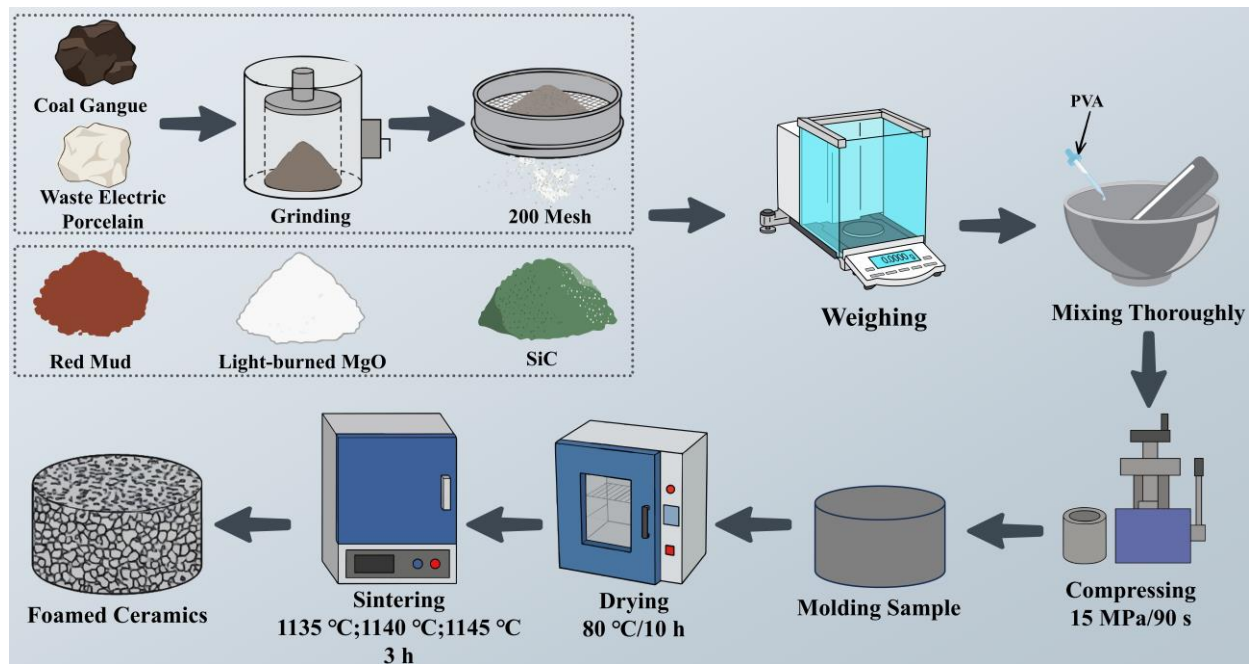


Figure 1. The preparation schematic diagram of porous ceramics.

2.3. Characterization

2.3.1. Bulk Density, Apparent Porosity, and Water Absorption

According to GB/T 1966-2024 [31], the bulk density and apparent porosity of the porous ceramics were determined using the boiling method. Furthermore, the bulk density (ρ_b), apparent porosity (P_a) and water absorption (W) of the porous ceramics can be calculated using Equations (1), (2), and (3), respectively [9,32,33].

$$\rho_b = \frac{m_1}{m_3 - m_2} \times \rho_1 \quad (1)$$

$$P_a = \frac{m_3 - m_1}{m_3 - m_2} \times 100\% \quad (2)$$

$$W = \frac{m_3 - m_1}{m_1} \times 100\% \quad (3)$$

where m_1 is the mass of dried porous ceramics, g; m_2 is the suspended mass of the saturated sample, g; m_3 is the mass of the saturated sample in air, g; ρ_1 is the density of the impregnating solution at the testing temperature.

2.3.2. Linear and Vertical Expansion Rates

Loss on ignition reflects the mass loss of samples during sintering, which originates from organic decomposition, carbon combustion, carbonate decomposition, and foaming reaction. The loss on ignition was calculated according to Equation (4) [20].

$$S = \frac{m_1 - m_2}{m_1} \quad (4)$$

where m_1 is the mass of the dried green body, g; m_2 is the mass of the cooled sintered specimen, g.

2.3.3. Room Temperature Compressive Strength Test

Compressive strength of specimens was tested using a universal high-low temperature testing machine (WDW-3T, Jinan Puye Electromechanical Technology Co., Ltd., Jinan, China), with the loading speed set at 0.5 mm/min. The calculation formula is as follows [3]:

$$\sigma = \frac{4P}{\pi d^2} \quad (5)$$

where σ is the compressive strength, MPa; P is the specimen breaking load, N; d is the specimen diameter, mm.

2.3.4. Room Temperature Thermal Conductivity Test

The room-temperature thermal conductivity of porous ceramics was measured using the laser flash method. The testing instrument was a laser thermal conductivity meter (DRX-II-JG, Xiangtan Huafeng Instrument Manufacturing Co., Ltd., Xiangtan, China), and the test standard was GJB 1201.1A-2021 [34]. Prior to testing, the sintered samples were ground into circular specimens with parallel, smooth upper and lower surfaces. The test surfaces of the specimens were sprayed with graphite to ensure uniform absorption of laser energy. The tests were conducted at room temperature under vacuum conditions.

2.3.5. Analysis of Raw Material Composition

X-ray fluorescence spectrometry (XRF, EDX4500H, Jiangsu Skyray Instrument Co., Ltd., Kunshan, China) was utilized to analyze the chemical composition and elemental proportion of prepared samples. A 50 mm beryllium window and 4 kW end-window X-ray tube were adopted for measurement. P10 gas consisting of 10% methane and 90% argon served as the detector working gas.

2.3.6. Phase Analysis

X-ray diffraction (XRD, Rigaku SmartLab, Tokyo, Japan) was adopted to identify crystalline phases of specimens. A copper target was applied during testing, with the tube voltage and current set at 40 kV and 200 μ A, respectively. The divergence/scatter slits (DS/SS) were fixed at 1°, and the receiving slit (RS) was 0.3 mm. The scanning rate was controlled at 10°/min, with a scanning range of 5° to 80°.

2.3.7. Microstructure and Elemental Distribution Characterization

Field emission scanning electron microscope (FE-SEM, Hitachi Regulus 8100, Hitachi High-Tech Corporation, Tokyo, Japan) combined with energy dispersive X-ray spectroscopy (EDS) was used to observe the microscopic morphology and elemental distribution of samples. Surface microstructure was captured under an accelerating voltage of 5 kV, and elemental distribution analysis was carried out at 15 kV. All observations were completed with a SE2 secondary electron detector (Hitachi Regulus 8100, Hitachi High-Tech Corporation, Tokyo, Japan).

3. Results and Discussion

3.1. Crystalline Phase of Porous Ceramics

The effects of foaming agent dosage and sintering temperature on the phase composition of as-prepared samples are shown in Figure 2. The XRD patterns in Figure 2a reveal that samples sintered at different temperatures present characteristic diffraction peaks well matched with magnesium iron aluminum spinel (PDF#71-1237) and sodium calcium feldspar (PDF#86-1650). When the sintering temperature increased from 1135 °C to 1145 °C, the intensities of the characteristic diffraction peaks of magnesium iron aluminum

spinel (especially for $2\theta = 30\text{--}40^\circ$) increased significantly. Meanwhile, the full width at half maximum (FWHM) of diffraction peaks decreased and the peak profiles became sharper. According to the Scherrer Equation (6), a smaller FWHM corresponds to a larger grain size and higher crystallinity [35].

$$D = \frac{0.89\lambda}{\beta \cos\theta} \quad (6)$$

where D is the grain size, nm; λ is the X-ray wavelength, nm; θ is the diffraction angle, $^\circ$; β is the FWHM, radian.

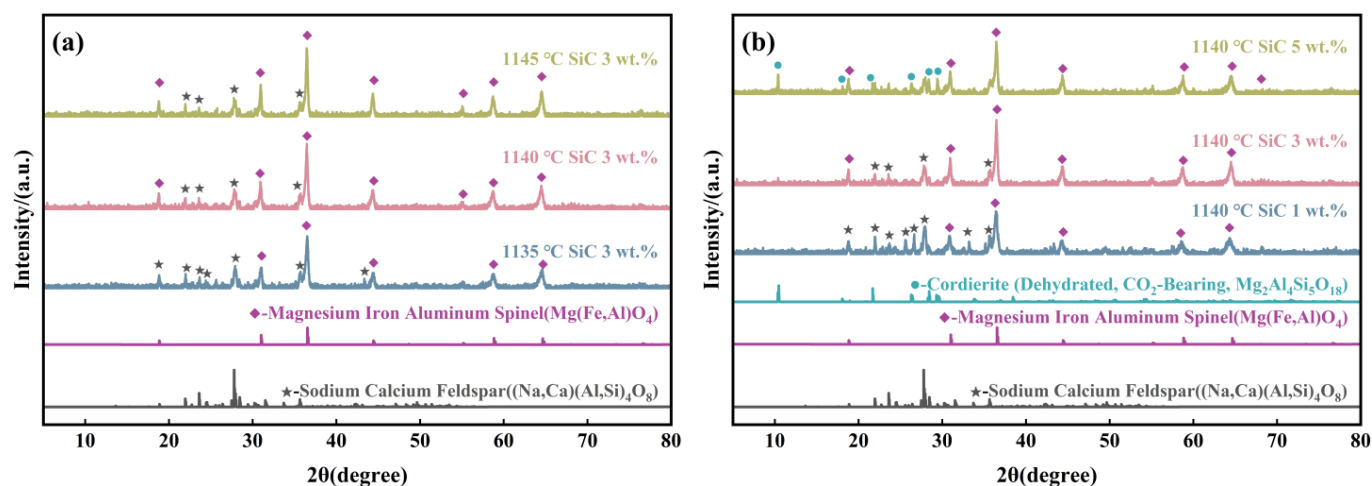
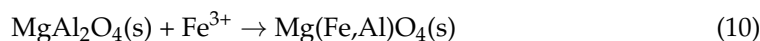
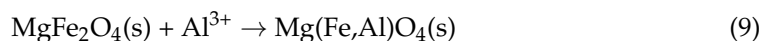
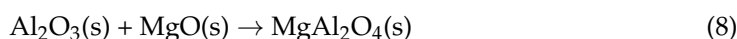
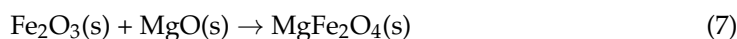


Figure 2. XRD patterns of samples (a) sintered at 1135 $^\circ\text{C}$, 1140 $^\circ\text{C}$, and 1145 $^\circ\text{C}$; (b) with different SiC contents of 1 wt.%, 3 wt.%, and 5 wt.% SiC.

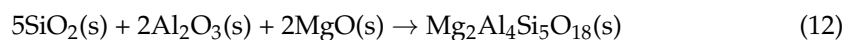
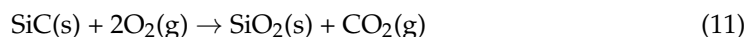
These results demonstrate that elevated sintering temperature facilitates grain growth and crystallinity improvement. At higher temperatures, the accelerated atomic diffusion promotes nucleation and grain growth, thus forming more perfect crystal structures.

At 1000–1100 $^\circ\text{C}$, impurities within the samples generate a liquid phase and accelerate the interdiffusion of Mg^{2+} , Al^{3+} and Fe^{3+} . At $\text{MgO}/\text{Fe}_2\text{O}_3$ interfaces, Mg^{2+} and Fe^{3+} interdiffuse to form a magnesium ferrite solid solution (Equation (7)). At $\text{Al}_2\text{O}_3/\text{MgO}$ interfaces, the interdiffusion of Mg^{2+} and Al^{3+} produces MgAl_2O_3 (Equation (8)). When the sintering temperature was raised from 1135 $^\circ\text{C}$ to 1145 $^\circ\text{C}$, cation interdiffusion is further enhanced by the increased liquid phase content and reduced melt viscosity, which in turn promotes the formation of Mg–Fe–Al spinel (Equations (9) and (10)) [36–38].



From Figure 2b, elevating SiC dosage from 1 wt.% to 3 wt.% narrows the FWHM and sharpens the characteristic peaks of magnesium iron aluminum spinel, confirming that moderate SiC incorporation favors crystalline development of the spinel phase. Meanwhile, sodium calcium feldspar suffers gradual peak attenuation, accompanied by the initial formation of cordierite (Dehydrated, CO_2 -Bearing) with weak diffraction responses. Further increasing the SiC content to 5 wt.% leads to an obvious decrease in the characteristic peak intensity of Mg–Fe–Al spinel. The excessive SiO_2 produced by SiC oxidation consumes partial MgO and Al_2O_3 to form more cordierite, which reduces the reactants available for

spinel crystallization and dilutes the relative content of the spinel phase, thus weakening its diffraction response. The formation pathway of cordierite is discussed as follows [20,39,40]:



Semi-quantitative phase analysis was performed based on the integral area of XRD characteristic peaks, and the relative mass fractions of each crystalline phase under different preparation parameters are summarized in Table 3. All data are normalized to the total mass of detectable crystalline phases, excluding the amorphous glass phase generated during liquid-phase sintering.

Table 3. Relative mass fraction of crystalline phases in porous ceramics under different preparation parameters.

Temperature/°C	SiC/wt. %	Mg-Fe-Al Spinel/wt. %	Sodium Calcium Feldspar/wt. %	Cordierite/wt. %
1135	3	61.9	38.1	0
1140	1	66.6	33.4	0
1140	3	86.3	13.7	0
1140	5	73.4	10.3	16.3
1145	3	85.7	14.3	0

For samples with fixed 3 wt.% SiC, increasing the sintering temperature from 1135 °C to 1140 °C significantly raises the relative content of Mg–Fe–Al spinel from 61.9 wt.% to 86.3 wt.%, while sodium calcium feldspar decreases sharply from 38.1 wt.% to 13.7 wt.%. This confirms that elevated temperature accelerates ion diffusion in the liquid phase and drives the transformation from feldspar to spinel. Further heating to 1145 °C causes negligible change in phase proportions, indicating the spinel formation reaction basically reaches equilibrium at 1140 °C.

For samples sintered at 1140 °C, raising SiC dosage from 1 wt.% to 3 wt.% increases spinel content from 66.6 wt.% to 86.3 wt.%, as the exothermic oxidation of SiC promotes spinel crystallization. When SiC rises to 5 wt.%, spinel content drops to 73.4 wt.%, and cordierite emerges with a relative content of 16.3 wt.%. This is because excessive SiC oxidation introduces extra SiO₂, which reacts with MgO and Al₂O₃ to form cordierite and consumes partial precursors for spinel synthesis. These quantitative results are highly consistent with the peak intensity trends in Figure 2.

3.2. Microstructure and Thermal Property of Porous Ceramics

Figure 3 compares the macroscale appearance, fractured cross-sectional microstructure, and EDS elemental mapping of samples sintered at a constant 1140 °C with SiC additions of 1 wt.%, 3 wt.%, and 5 wt.% from top to bottom.

At 1 wt.% SiC (Figure 3a), the light-brown sample exhibits the densest fracture structure with homogeneously distributed fine pores and an inconspicuous foaming effect. Increasing SiC loading to 3 wt.% (Figure 3b) darkens the sample and markedly raises porosity, accompanied by large pores exceeding 500 µm. Extra CO₂ released from SiC decomposition elevates internal gas pressure, which overcomes melt resistance to induce pore expansion; adjacent small pores are coalesced into large voids to construct a hierarchical pore architecture. Further raising SiC to 5 wt.% produces excessive internal gas. The fraction of large pores rises unevenly, and interconnected through-holes form in localized regions [24,41,42].

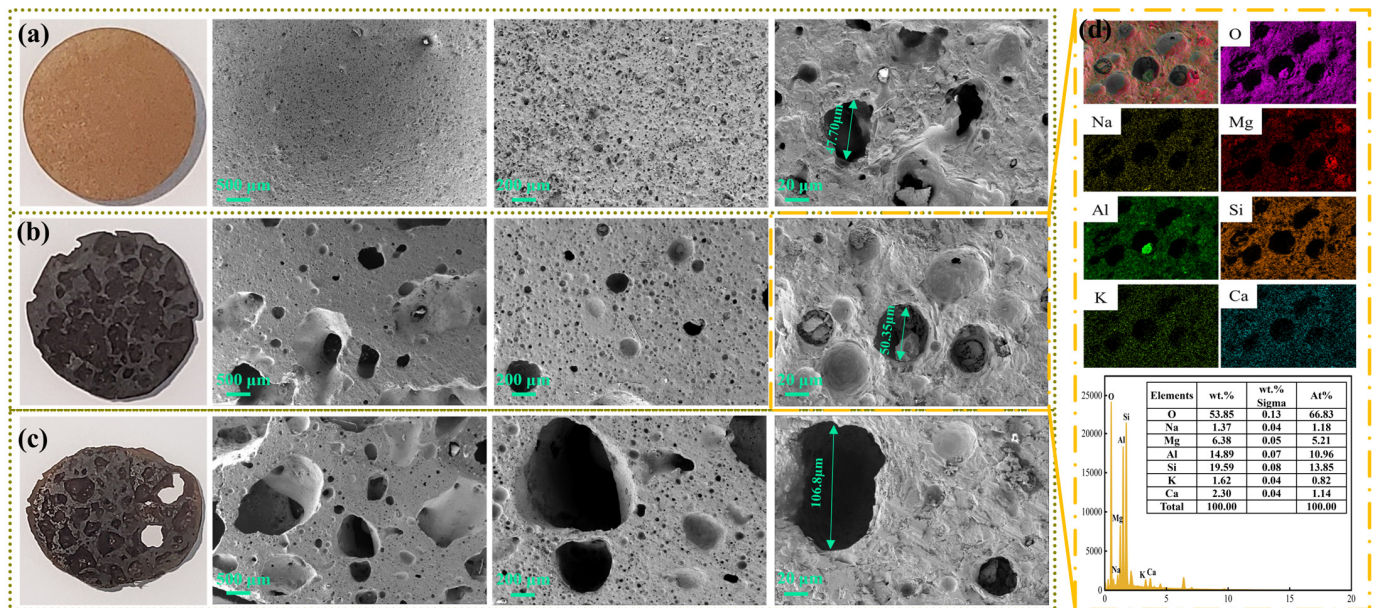


Figure 3. SEM and EDS images of samples sintered at 1140 °C with varying SiC contents at 1140 °C: (a) 1%, (b) 3 wt.%, and (c) 5 wt.%; (d) the EDS of the sample with 3 wt.% SiC.

EDS characterization on the sample with 3 wt.% SiC reveals dominant Si, Al, Mg, Ca, K, and Na elements, consistent with the XRD patterns. Apart from locally enriched Al and Mg stemming from residual unreacted Al_2O_3 and MgO , the remaining elements distribute homogeneously. This demonstrates sufficient sintering driving force at the set temperature to facilitate elemental migration and thorough solid-state reactions [10].

Figure 4 illustrates the microstructural evolution of specimens with fixed 3 wt.% SiC sintered at 1135 °C, 1140 °C, and 1145 °C (from top to bottom).

At 1135 °C (Figure 4A), the dark-colored sample contains abundant fine, densely packed micropores. Loose intergranular bonding and blurred grain boundaries originate from inadequate liquid-phase fraction and suppressed solid-state atomic diffusion, leading to incomplete densification and high microporosity at the premature sintering stage [43]. Upon heating to 1140 °C (Figure 4B), enhanced liquid generation drives particle rearrangement via capillary force; fine irregular voids gradually coalesce and grow into macropores accompanied by partial pore spheroidization and closure. Further elevating the temperature to 1145 °C (Figure 4C) induces significant microstructural evolution: the overall pore population declines, while isolated oversized cavities emerge. Smooth rounded pore peripheries and obvious viscous creep of the matrix verify robust liquid-phase sintering. The abundant liquid phase partially fills interstitial micropores, accelerates grain coarsening via grain boundary migration, and increases local bulk density, whereas the reduced melt viscosity facilitates Ostwald ripening and pore coalescence [44–46]. Quantitative pore-size distribution profiles (Figure 4a–c) calculated by Image-Pro Plus 6.0 match the morphological evolution revealed by SEM characterization.

Changes in thermal conductivity are determined by both the true porosity and the pore structure. As the sintering temperature increases from 1135 °C to 1145 °C, the sample's bulk density continues to decrease, while the true density changes very little because no fundamental phase changes have occurred. Based on Equation (13) [47], it can be inferred that the sample's true porosity will continue to increase, and the thermal conductivity drops from 0.142 W/(m·K) to 0.066 W/(m·K). The gaseous constituent (~0.026 W/(m·K)) possesses markedly inferior thermal conduction capacity relative to the ceramic solid skeleton. High-temperature foaming expands the total pore volume and breaks the continuity of solid-state heat conduction pathways [48–50]. Discontinuous pore frameworks further aggravate

phonon scattering and restrain lattice thermal transport [51,52]. The specimen sintered at 1140 °C realizes an optimal trade-off between porosity and thermal insulation performance, which is favorable for fabricating lightweight high-strength thermal-insulating porous ceramics.

$$P_t = \left(1 - \frac{\rho_b}{\rho_t}\right) \times 100\% \quad (13)$$

where P_t is true porosity; ρ_b is bulk density, g/cm³; ρ_t is true density, g/cm³.

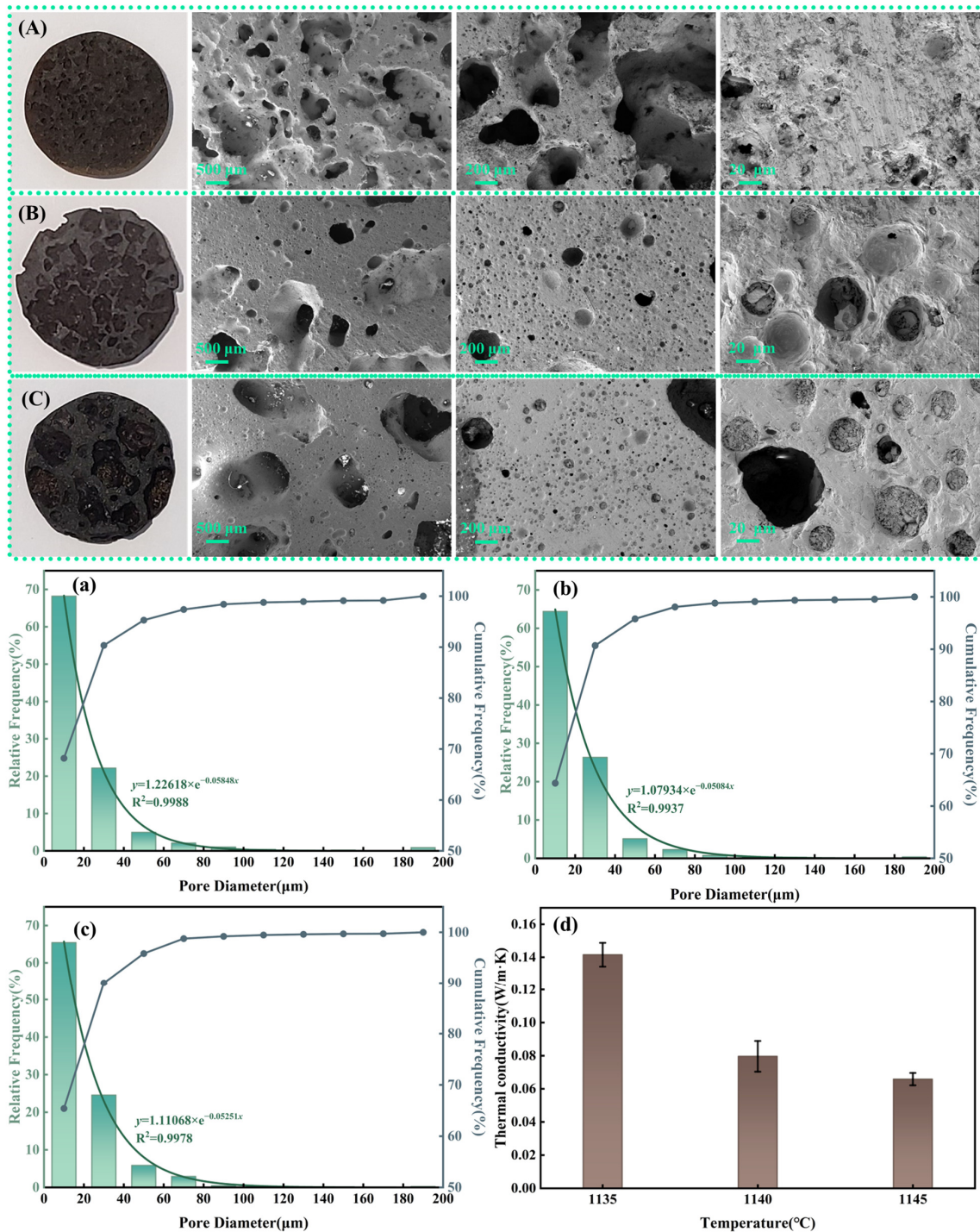


Figure 4. Macroscopic photographs and SEM micrographs of samples added with 3 wt.% SiC sintered at (A) 1135 °C, (B) 1140 °C, and (C) 1145 °C; (a–c) Pore diameter distributions of the corresponding samples, statistically derived from SEM images with 500 μm scale bar; (d) Thermal conductivity of samples sintered at different temperatures.

3.3. Physical Properties of Porous Ceramics

Loss on ignition quantitatively reflects the release of volatile components and the degree of solid-phase reaction during sintering, while apparent porosity and water absorption are critical indicators for evaluating the development of interconnected open-pore structures in porous ceramics [22,53,54]. Accordingly, the effects of SiC dosage (Figure 5a) and sintering temperature (Figure 5b) on the loss on ignition, apparent porosity, and water absorption of sintered specimens were systematically investigated in this section.

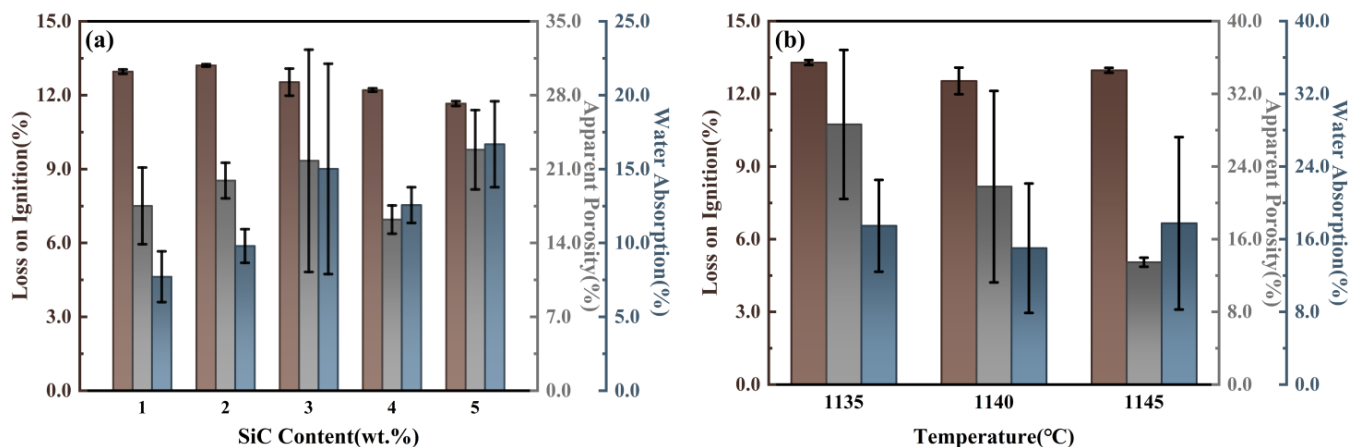


Figure 5. Loss on ignition, apparent porosity, and water absorption of porous ceramics: (a) with different SiC contents (1%, 2 wt.%, 3 wt.%, 4 wt.%, 5 wt.%) at 1140 °C, (b) sintered at different temperatures (1135 °C, 1140 °C, 1145 °C) with 3 wt.% SiC.

As illustrated in Figure 5a, all samples exhibit a stable loss on ignition range of 11–13%, which is primarily attributed to the oxidation of residual carbon, sulfide decomposition, and structural water removal from coal gangue, as well as the dehydration of crystalline water and thermal decomposition of minor carbonates in red mud. High-temperature oxidation of SiC generates solid SiO_2 (Equation (11)), which produces a mass-compensation effect. With increasing SiC dosage, such solid-phase mass accumulation effectively offsets the mass loss caused by volatile emission, thereby reducing the overall loss on ignition [55–57].

When the SiC content increases from 1 wt.% to 5 wt.%, the apparent porosity rises gradually from 17.5% to 22.8%, and water absorption presents a highly consistent variation trend. This phenomenon verifies the effective pore-forming capability of SiC. The underlying mechanism is dominated by two synergistic effects. On the one hand, SiC particles undergo intense interfacial oxidation under high-temperature oxidative atmospheres, and the in situ generated CO_2 gas is partially trapped within the high-viscosity liquid phase, forming abundant isolated and interconnected open pores [12,58–60]. On the other hand, unreacted SiC acts as a rigid secondary phase that restricts particle rearrangement and suppresses sintering densification, thereby preserving the inherent interstitial pores within the ceramic matrix [59,61,62]. The combined effects of gas foaming and second-phase pinning significantly promote the evolution of open-pore structures with increasing SiC incorporation.

As depicted in Figure 5b, all specimens maintain a stable loss on ignition value ranging from 12.5% to 13.3% without significant discrepancies across different sintering temperatures. This indicates that nearly all thermally induced mass-loss and mass-compensation physical and chemical reactions have been fully accomplished at 1135 °C. As the sintering temperature elevates from 1135 °C to 1145 °C, the apparent porosity decreases considerably from 28.6% to 13.5%, accompanied by a synchronous reduction in water absorption. This

evolution is primarily governed by the enhanced liquid-phase generation and intensified viscous flow behavior at elevated temperatures [63].

Alkali metal oxides (Na_2O , CaO) in red mud and low-melting components in waste electric porcelain gradually melt above 1135°C , forming fluid silicate liquid phases. Further temperature increments increase the liquid phase fraction and reduce melt viscosity, thereby facilitating sintering densification through two dominant mechanisms. On the one hand, capillary force derived from the fluid liquid phase drives particle rearrangement and interfacial bonding, effectively eliminating interstitial pores within the matrix [64]. On the other hand, the high-fluidity melt encapsulates partial open pores and converts them into closed pore structures; meanwhile, residual trapped gas dissolves in the liquid phase and escapes continuously during sustained sintering [65]. The synergistic effect of pore closure, gas discharge, and matrix densification collectively accounts for the continuous decline in apparent porosity and water absorption with increasing sintering temperature.

Bulk density and compressive strength are essential mechanical and structural indicators that determine the practical application potential of porous ceramics. Figure 6a,b illustrate the variations in bulk density and compressive strength under varying SiC dosages and sintering temperatures, respectively. As shown in Figure 6a, increasing the SiC content from 1 wt.% to 5 wt.% induces a prominent degradation in both bulk density and mechanical performance. Specifically, the bulk density sharply decreases from 2.28 g/cm^3 to 1.31 g/cm^3 , while the compressive strength drops drastically from approximately 15.7 MPa to 1.94 MPa. Combined with the aforementioned SEM microstructural analysis, this performance deterioration is primarily attributed to the formation of highly interconnected macropores induced by excessive SiC incorporation ($>3\text{ wt.}\%$). The emergence of continuous large pore networks disrupts the integrity of the ceramic skeleton. The increased porosity reduces the effective load-bearing cross-sectional area and exacerbates internal stress concentration, thereby resulting in a severe decline in macroscopic mechanical properties [45,55].

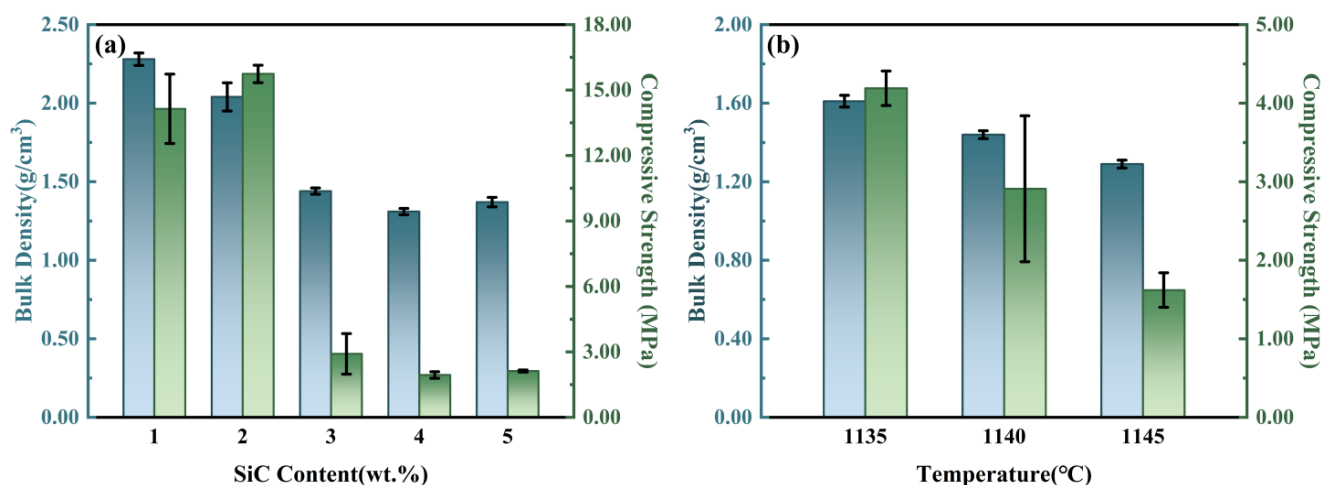


Figure 6. Bulk density and compressive strength of porous ceramics: (a) with different SiC contents (1%, 2 wt.%, 3 wt.%, 4 wt.%, 5 wt.%) at 1140°C , (b) sintered at different temperatures (1135°C , 1140°C , 1145°C) with 3 wt.% SiC.

As presented in Figure 6b, synchronous reductions in bulk density and compressive strength are observed as the sintering temperature increases from 1135°C to 1145°C , with density decreasing from 1.61 g/cm^3 to 1.29 g/cm^3 and strength declining from 4.19 MPa to 1.62 MPa. At 1145°C , substantial liquid-phase generation originating from the melting of low-melting components in red mud and waste electric porcelain significantly reduces the melt viscosity [52,66]. The residual internal gas and in situ generated CO_2 from SiC

oxidation readily overcome the melt surface tension, promoting rapid bubble expansion and pore coalescence. Although abundant open pores are encapsulated and transformed into closed pores by the sufficient liquid phase (consistent with the decreased apparent porosity), intensive high-temperature foaming induces significant macroscopic volume expansion, which dominates over the densification effect and ultimately reduces the bulk density [16,43].

From the perspective of pore structural evolution, high-temperature-induced bubble coarsening and merging increase the average pore size and thin the pore wall thickness, which is consistent with the SEM microscopic characteristics. Such coarsened pore architecture weakens the load transfer and stress dispersion capacity of pore walls, consequently reducing the compressive strength [15]. In essence, a competitive mechanism exists between sintering densification and foaming-induced pore coarsening within the investigated temperature range. The pore coarsening effect dominates the structural evolution from 1135 °C to 1145 °C, thereby governing the final physical and mechanical properties of the porous ceramics.

Figure 7 shows a comparison of the samples in this study with typical porous thermal insulation ceramics based on solid waste reported in previous studies in terms of compressive strength and thermal conductivity [14,16,18,28,67–71]. All data conform to the following pattern: as thermal conductivity decreases, compressive strength decreases accordingly, and this decrease in thermal conductivity is determined by an increase in true porosity. The specimen containing 3 wt.% SiC, sintered at 1140 °C, exhibited a good balance between low thermal conductivity and high compressive strength, demonstrating highly competitive overall performance among similar thermal insulation materials.

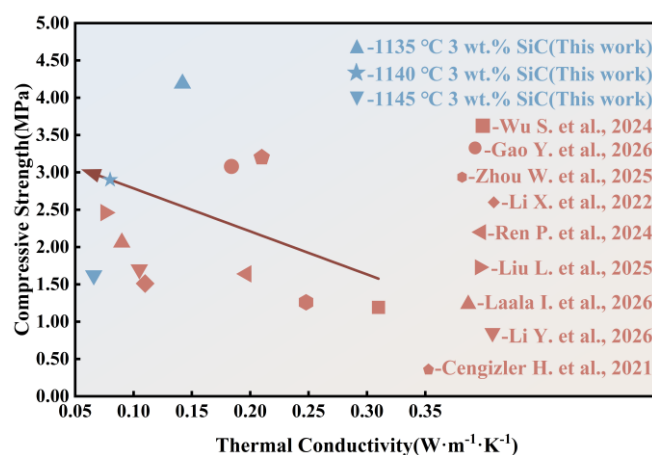


Figure 7. Performance comparison of compressive strength and thermal conductivity [14,16,18,28,67–71]. The arrow indicates the direction of better comprehensive performance, corresponding to lower thermal conductivity and higher compressive strength.

4. Conclusions

High-performance thermal-insulating porous ceramics were successfully prepared from red mud, coal gangue, and waste electric porcelain using SiC as a foaming agent. The coupled effects of SiC dosage and sintering temperature on phase evolution, pore structure, mechanical properties, and thermal conductivity were systematically explored. The results indicate that Ca, Fe, and alkali metal oxides in red mud effectively enhance the sintering reactivity of the composite system. These components react with Si–Al phases derived from coal gangue and waste electric porcelain, as well as incorporated light-burned magnesia, to form Mg–Fe–Al spinel and sodium calcium feldspar. Appropriate SiC addition further

improves the crystallinity of spinel and promotes the generation of cordierite, thereby stabilizing the microstructure of the ceramic matrix.

SiC oxidation induces continuous pore formation and interconnectivity, which reduces bulk density and thermal conductivity but gradually degrades compressive strength with increased dosage. Sintering temperature presents a competitive effect between liquid-phase densification and pore coarsening. Excessively high temperature dominates bubble expansion and merging, thinning pore walls, and deteriorating mechanical performance. The optimized sample obtained at 3 wt.% SiC and 1140 °C achieves a balanced pore structure, with an apparent porosity of 21.8%, a thermal conductivity of 0.08 W/(m·K) and a compressive strength of 2.91 MPa. Compared to previously reported porous insulating ceramics made from solid waste that offer the same level of thermal insulation, this material has superior mechanical load-bearing capacity and demonstrates highly competitive overall performance. In terms of economic benefits, since industrial solid waste accounts for 90% of the raw materials, this material not only significantly reduces production costs compared to commercially available pure mineral-based insulating ceramics but also enables the resource recovery of large volumes of industrial solid waste. It is expected that this product will be widely adopted and applied in medium-temperature insulation applications, such as industrial furnace linings and building energy conservation.

This work provides a facile and sustainable approach for high-value recycling of solid wastes and controllable fabrication of high-performance thermal-insulating porous ceramics.

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References

1. Zhou, L.; Liu, J.; Tu, T.; Wu, Y.; Zhang, G. Fast grain growth phenomenon in high-entropy ceramics: A case study in rare-earth hexaaluminates. *J. Adv. Ceram.* **2023**, *12*, 111–121. [[CrossRef](#)]
2. Xenidis, A.; Harokopou, A.D.; Mylona, E.; Brofas, G. Modifying alumina red mud to support a revegetation cover. *JOM* **2005**, *57*, 42–46. [[CrossRef](#)]
3. Wang, R.; Liu, X.; Wei, H.; Xu, Y.; Zhou, R.; Huang, Z.; Zhan, H.; Ma, S.; Yang, C. Study on phase behavior and mechanical properties of high temperature resistant materials prepared from waste electric porcelain. *Ceram. Int.* **2023**, *49*, 11537–11543. [[CrossRef](#)]
4. Giergiczny, Z. Fly ash and slag. *Cem. Concr. Res.* **2019**, *124*, 105826. [[CrossRef](#)]
5. Snehasree, N.; Nuruddin, M.; Moghal, A.a.B. Critical Appraisal of Coal Gangue and Activated Coal Gangue for Sustainable Engineering Applications. *Appl. Sci.* **2025**, *15*, 9649. [[CrossRef](#)]
6. Hertel, T.; Pontikes, Y. Geopolymers, inorganic polymers, alkali-activated materials and hybrid binders from bauxite residue (red mud)—Putting things in perspective. *J. Clean. Prod.* **2020**, *258*, 120610. [[CrossRef](#)]

7. Cao, Y.; Lv, Z.; Zhang, K.; Lan, C.; Fan, M.; Lu, X.; Ke, Y.; Guo, W.; Yang, Y.; Wang, X.; et al. Waste electric porcelain-based refractory bricks with significantly enhanced mechanical properties: Preparation, characterization and mechanism. *Ceram. Int.* **2024**, *50*, 49698–49711. [\[CrossRef\]](#)
8. Leroy, C.; Ferro, M.C.; Monteiro, R.C.C.; Fernandes, M.H.V. Production of glass-ceramics from coal ashes. *J. Eur. Ceram. Soc.* **2001**, *21*, 195–202. [\[CrossRef\]](#)
9. Yoon, S.; Yun, Y. An advanced technique for recycling fly ash and waste glass. *J. Mater. Process. Technol.* **2005**, *168*, 56–61. [\[CrossRef\]](#)
10. Xu, W.; Shen, K.; Cao, Z.; Liu, F.; Zhang, Y.; Zhang, T.; Wu, N.; Ouyang, S. Crystallization and thermal stability effects on tailings glass-ceramics by various heat treating processes. *Mater. Chem. Phys.* **2021**, *263*, 124334. [\[CrossRef\]](#)
11. Luo, Y.; Zheng, S.; Ma, S.; Liu, C.; Wang, X. Preparation of sintered foamed ceramics derived entirely from coal fly ash. *Constr. Build. Mater.* **2018**, *163*, 529–538. [\[CrossRef\]](#)
12. Li, X.; Yang, Q.; Sun, D.; Lv, C. Foaming mechanism and performance of closed-pore foamed ceramics prepared using calcium carbonate as a foaming agent. *Ceram. Int.* **2021**, *47*, 29162–29173. [\[CrossRef\]](#)
13. Ohji, T.; Fukushima, M. Macro-porous ceramics: Processing and properties. *Int. Mater. Rev.* **2013**, *57*, 115–131.
14. Li, X.; Pan, M.; Tao, M.; Liu, W.; Gao, Z.; Ma, C. Preparation of high closed porosity foamed ceramics from coal gangue waste for thermal insulation applications. *Ceram. Int.* **2022**, *48*, 37055–37063. [\[CrossRef\]](#)
15. Liu, Y.; Xu, Y.; Zhou, R.; Wei, H.; Zhang, W.; Zhan, H.; Ma, S.; Yang, C.; Huang, Z. Study on Preparation of Porous Light Refractory from Waste Electric Porcelain. *ChemistrySelect* **2024**, *9*, e202303743. [\[CrossRef\]](#)
16. Ren, P.; Zhou, R.; Liu, H.; Huo, Y.; Wang, Y. Preparation of coal gangue based foamed ceramics with SiC as blowing agent and study on its thermal insulation performance. *Ceram. Int.* **2024**, *50*, 31680–31690. [\[CrossRef\]](#)
17. Ge, X.; Sun, W.; Zhou, M.; Wang, Q.; Pei, L.; Zhang, Y. Influence of crystalline phases on the properties of foamed ceramics from spodumene flotation tailings. *Ceram. Int.* **2025**, *51*, 58234–58247. [\[CrossRef\]](#)
18. Liu, L.; Wang, W.; Sun, J.; Liu, C. Full tailings-based foamed ceramics: Preparation and characterization using secondary aluminum ash as self-foaming agent. *Ceram. Int.* **2025**, *51*, 19188–19198. [\[CrossRef\]](#)
19. Zhao, Y.; Ding, X.; Zhou, Y.; Qian, H.; Deng, X.; Ran, S. Influence of the foaming process on the pore structure and properties of fly ash-based foamed ceramics. *Ceram. Int.* **2025**, *51*, 14398–14407. [\[CrossRef\]](#)
20. Weng, J.; Shen, X.; Yang, Y.; Zhang, X.; Fan, M.; Gao, G.; Guo, Z.; Lv, Z.; Feng, X. Low-Cost Foamed Ceramics with Enhanced Mechanical Performance and Uniform Pore Size Structure. *Crystals* **2025**, *15*, 180. [\[CrossRef\]](#)
21. Li, C.; Gao, R.; Ouyang, H.; Shen, T.; Chen, Z.; Li, Y. Hierarchical porous (Ta_{0.2}Nb_{0.2}Ti_{0.2}Zr_{0.2}Hf_{0.2})C high-entropy ceramics prepared by a self-foaming method for hermal insulation. *J. Adv. Ceram.* **2024**, *13*, 956–966. [\[CrossRef\]](#)
22. Dry, C.; Meier, J.; Bukowski, J. Sintered coal ash/flux materials for building materials. *Mater. Struct.* **2004**, *37*, 114–121. [\[CrossRef\]](#)
23. Chen, G.-Z.; Shen, C.-F.; Zhou, L.-J. Design and performance analysis of wireless sensor network location node system for underground mine. *Min. Sci. Technol.* **2009**, *19*, 813–818. [\[CrossRef\]](#)
24. Heon Kim, C.; Go, E.B.; Tae Kim, H. Thermophysical Properties of Zirconia Toughened Alumina Ceramics with Boron Nitride Nanotubes Addition. *Heat Transf. Eng.* **2019**, *41*, 1354–1364. [\[CrossRef\]](#)
25. Bhatnagar, A.; Vilar, V.J.P.; Botelho, C.M.S.; Boaventura, R.a.R. A review of the use of red mud as adsorbent for the removal of toxic pollutants from water and wastewater. *Environ. Technol.* **2011**, *32*, 231–249. [\[CrossRef\]](#)
26. Chen, Y.; Yu, C.; Cheng, X.; Wang, R.; Deng, C.; Ding, J.; Liu, Z.; Ma, B.; Zhu, H.; Hu, J.; et al. Advances in Sintering Technologies for SiC Ceramics: Mechanisms, Challenges, and Industrial Applications. *High-Temp.* **2025**, *2*, 10013. [\[CrossRef\]](#)
27. Carlson, C.L.; Adriano, D.C. Environmental Impacts of Coal Combustion Residues. *J. Environ. Qual.* **1993**, *22*, 227–247. [\[CrossRef\]](#)
28. Cengizler, H.; Koç, M.; Şan, O. Production of ceramic glass foam of low thermal conductivity by a simple method entirely from fly ash. *Ceram. Int.* **2021**, *47*, 28460–28470. [\[CrossRef\]](#)
29. Chai, Z.; Chen, X.; Fang, M.; Min, X.; Huang, Z.; Wang, Q.; Liu, C.; Liu, Y.; Huang, F. Fabrication and properties of high-thermal-storage RTO ceramics using bauxite tailings and red mud. *Ceram. Int.* **2023**, *49*, 31342–31350. [\[CrossRef\]](#)
30. Sun, Y.; Li, S.; Zhao, Q.; Cong, Z.; Xia, Y.; Jiao, X.; Chen, D. Recent Advancements in Alumina-Based High-Temperature Insulating Materials: Properties, Applications, and Future Perspectives. *High-Temp.* **2025**, *2*, 10001. [\[CrossRef\]](#)
31. GB/T 1966-2024; Porous Ceramics-Determination of Apparent Porosity and Bulk Density. State Administration for Market Regulation: Beijing, China, 2024.
32. Zhang, X.; Chen, X.; Min, W.; Liang, G.; Zhang, W.; Yao, S.; Zhong, X. Preparation of multifunctional ceramic foams for sound absorption, waterproofing, and antibacterial applications. *RSC Adv.* **2024**, *14*, 1009–1017. [\[CrossRef\]](#) [\[PubMed\]](#)
33. Xu, X.; Zhang, J.; Wu, J.; Chen, Z.; Wang, Y. Preparation of foamed ceramics from steel slag with high calcium and iron content. *Int. J. Appl. Ceram. Technol.* **2023**, *20*, 2993–3003. [\[CrossRef\]](#)
34. GJB 1201.1A-2021; Test Method for Thermal Diffusivity of Solid Materials at High Temperature Part 1: Laserpulse Method. Equipment Development Department of People's Republic of China Central Military Commission: Beijing, China, 2021.

35. Lv, Z.; Shen, X.; Guo, J.; Cao, Y.; Lan, C.; Ke, Y.; Yang, Y.; Qi, J. Microstructure and Magnetic Properties Dependence on the Sputtering Power and Deposition Time of TbDyFe Thin Films Integrated on Single-Crystal Diamond Substrate. *Processes* **2022**, *10*, 2626. [\[CrossRef\]](#)
36. Marchant, D.D.; Lambertbates, J. Preparation and Properties of Iron-Magnesium-Aluminum Spinel. *J. Am. Ceram. Soc.* **2006**, *66*, 99–102. [\[CrossRef\]](#)
37. Sickafus, K.E.; Wills, J.M.; Grimes, N.W. Structure of Spinel. *J. Am. Ceram. Soc.* **2004**, *82*, 3279–3292.
38. Li, S.; Yin, Z.; Wang, A.; Yu, C.; Feng, R.; Liu, G.; Tian, X.; Chen, L.; Ma, C. Characterization, Exploitation and Application of Tibetan Microcrystalline Magnesite. *High-Temp. Mater.* **2024**, *1*, 10007. [\[CrossRef\]](#)
39. Lv, Z.; Fan, M.; Cao, Y.; Wang, X.; Lan, C.; Wang, Q.; Yang, C.; Shen, X.; Feng, X. Ca(OH)₂-assisted preparation of mullite-corundum foam ceramics based on waste electrical porcelain and gangue: Phase transformation, structural homogenization, and foaming mechanism. *Ceram. Int.* **2025**, *51*, 47482–47491. [\[CrossRef\]](#)
40. Wang, H.; Wang, S.; Meng, Z.; Chen, Z.; Liu, L.; Wang, X.; Qian, D.; Xing, Y. Mechanism of cordierite formation obtained by high temperature sintering technique. *Ceram. Int.* **2023**, *49*, 20544–20555. [\[CrossRef\]](#)
41. Zhang, L.; Hu, D.; Snetkov, I.L.; Balabanov, S.; Palashov, O.; Li, J. A review on magneto-optical ceramics for Faraday isolators. *J. Adv. Ceram.* **2023**, *12*, 873–915. [\[CrossRef\]](#)
42. Yu, G.; Gao, W.; Yao, Y.; Zhang, T.; Fu, Y.; Kong, X. Recycling of Silicomanganese Slag and Fly Ash for Preparation of Environment-Friendly Foamed Ceramics. *Materials* **2023**, *16*, 6724. [\[CrossRef\]](#) [\[PubMed\]](#)
43. Lv, Z.; Fan, M.; Cao, Y.; Lan, C.; Wang, Q.; Yang, C.; Shen, X.; Feng, X. Effect of potassium feldspar dosage and sintering temperature on the properties of saggar derived from waste ceramic for calcination of ternary lithium-ion batteries cathode materials. *Ceram. Int.* **2025**, *51*, 59239–59251. [\[CrossRef\]](#)
44. Shao, J.; Bai, C.; Li, X.; Yang, K.; Zheng, T.; Qiao, Y.; Zhang, L.; Li, H.; Colombo, P. Open-cell mullite ceramic foams derived from porous geopolymer precursors with tailored porosity. *J. Adv. Ceram.* **2023**, *12*, 279–295. [\[CrossRef\]](#)
45. Huang, J.Y.; Yuan, J.C.; Zhu, T.T.; Zhong, T.; Xu, Y.F.; Luo, S.N. Dynamic compressive strength of alumina ceramics. *Ceram. Int.* **2022**, *48*, 36371–36382. [\[CrossRef\]](#)
46. Nanev, C.N. Thermodynamic and molecular-kinetic considerations of the initial growth of newly born crystals; crystal size distribution; Dissolution of small crystals during Ostwald ripening due to temperature changes. *Prog. Cryst. Growth Charact. Mater.* **2023**, *69*, 100604. [\[CrossRef\]](#)
47. Ali, M.S.; Hanim, M.a.A.; Tahir, S.M.; Jaafar, C.N.A.; Norkhairunnisa, M.; Matori, K.A. Preparation and characterization of porous alumina ceramics using different pore agents. *J. Ceram. Soc. Jpn.* **2017**, *125*, 402–412. [\[CrossRef\]](#)
48. Xia, F.; Cui, S.; Pu, X. Performance study of foam ceramics prepared by direct foaming method using red mud and K-feldspar washed waste. *Ceram. Int.* **2022**, *48*, 5197–5203. [\[CrossRef\]](#)
49. Wu, Z.; Zhao, Y.; Li, H.; Zeng, X.; Lin, J.; Wu, J.; Zhou, Y.; Chen, G.; Lai, X. Multifunctional ceramifiable silicone foam for smart fire fighting. *Chem. Eng. J.* **2024**, *496*, 154149. [\[CrossRef\]](#)
50. Huo, W.; Zhang, X.; Chen, Y.; Wang, D.; Liu, J.; Yan, S.; Yang, J. Mechanical strength of highly porous ceramic foams with thin and lamellate cell wall from particle-stabilized foams. *Ceram. Int.* **2018**, *44*, 5780–5784. [\[CrossRef\]](#)
51. Brunori, C.; Cremisini, C.; Massanisso, P.; Pinto, V.; Torricelli, L. Reuse of a treated red mud bauxite waste: Studies on environmental compatibility. *J. Hazard. Mater.* **2005**, *117*, 55–63. [\[CrossRef\]](#) [\[PubMed\]](#)
52. Chen, H.; Shih, N.; Wu, C.; Lin, S. Effects of the Loss on Ignition of Fly Ash on the Properties of High-Volume Fly Ash Concrete. *Sustainability* **2019**, *11*, 2704. [\[CrossRef\]](#)
53. Das, D.; Rout, P.K. A Review of Coal Fly Ash Utilization to Save the Environment. *Water Air Soil Pollut.* **2023**, *234*, s11270. [\[CrossRef\]](#)
54. Dong, X.; Han, F.; Hua, W.; Liu, T.; Zheng, J.; An, C.; Li, M. Preparation and properties of microcrystalline foam ceramics from silicon manganese smelting slag. *Ceram. Int.* **2024**, *50*, 2073–2082. [\[CrossRef\]](#)
55. Fu, F.; Hu, N.; Yang, S.; Li, Q.; Ye, Y.; Tang, R.; Chen, G. Effect of green gel system upon the properties of phosphorus tailings foamed ceramics. *Ceram. Int.* **2024**, *50*, 38748–38760. [\[CrossRef\]](#)
56. Fu, F.; Hu, N.; Ye, Y.; Chen, G.; Jia, J. Production of lightweight foam ceramics by adjusting sintering time and heating rate. *Constr. Build. Mater.* **2023**, *394*, 132063. [\[CrossRef\]](#)
57. Huang, Y.; Hu, N.; Ye, Y.; Ou, Z.; Shi, X. Preparation and pore-forming mechanism of MgO–Al₂O₃–CaO-based porous ceramics using phosphorus tailings. *Ceram. Int.* **2022**, *48*, 29882–29891. [\[CrossRef\]](#)
58. Li, H.; Pan, Z.; Chen, X.; Zhao, J.; Tang, L.; Liu, J.; Li, P.; Zhai, J. Stable relaxor ferroelectric phase of NaNbO₃-based ceramic with superb energy storage performances. *Mater. Today Phys.* **2023**, *38*, 101208. [\[CrossRef\]](#)
59. Li, Y.; Liu, C.; Luan, Z.; Peng, X.; Zhu, C.; Chen, Z.; Zhang, Z.; Fan, J.; Jia, Z. Phosphate removal from aqueous solutions using raw and activated red mud and fly ash. *J. Hazard. Mater.* **2006**, *137*, 374–383. [\[CrossRef\]](#) [\[PubMed\]](#)
60. Li, X.; Sun, D.; Yang, Q.; Lv, C. Foaming performance of the closed-pore foamed ceramics using silicon nitride as a foaming agent and sodium tetraborate as a modifier of foaming. *J. Eur. Ceram. Soc.* **2021**, *41*, 6330–6337. [\[CrossRef\]](#)

61. Ke, Y.; Lv, Z.; Fan, M.; Gao, J.; Cao, Y.; Zhang, X.; Wang, Y.; Shi, S.; Shen, X. Achieving high-volume waste ceramic incorporation in refractory materials: Synergistic improvement of mechanical properties and slag corrosion resistance. *Ceram. Int.* **2026**, *52*, 26852–26866. [[CrossRef](#)]
62. Tovbin, Y.K. Theory of absolute rates of gas/solid interface reactions. *Russ. J. Electrochem.* **2009**, *45*, 960–982. [[CrossRef](#)]
63. Zhang, X.; Lv, Z.; Fan, M.; Weng, J.; Wang, Y.; Lu, X.; Zhang, Z.; Wang, Q.; Shi, S.; Shen, X. Enhanced corrosion resistance of electrical porcelain saggers using a gradient-structured layer derived from industrial solid wastes. *Ceram. Int.* **2026**, *52*, 35337–35350.
64. Niu, A.; Lin, C. Trends in research on characterization, treatment and valorization of hazardous red mud: A systematic review. *J. Environ. Manag.* **2024**, *351*, 119660. [[CrossRef](#)]
65. Shi, X.; Liao, Q.; Liu, L.; Deng, F.; Chen, F.; Wang, F.; Zhu, H.; Zhang, L.; Liu, C. Utilizing multi-solid waste to prepare and characterize foam glass ceramics. *Ceram. Int.* **2023**, *49*, 35534–35543. [[CrossRef](#)]
66. Sahoo, M.K.; Mandal, A. Design and Analysis of Three-Dimensional Foams: A Review. *Arch. Comput. Methods Eng.* **2024**, *31*, 2265–2293. [[CrossRef](#)]
67. Wu, S.; Zhou, Y.; Gao, W.; Zhang, Z.; Liu, A.; Wu, C.; Peng, X.; Li, S.; Li, C.; Huang, Z. Microstructure and properties of SiC porous ceramics prepared by carbothermal reduction of iron tailings. *Constr. Build. Mater.* **2024**, *420*, 135647. [[CrossRef](#)]
68. Gao, Y.; She, A.; Wang, Y.; Li, H.; Lei, H.; Zhang, L.; Huang, Z.; Hou, Z.; Xu, Y.; Zhang, H.; et al. Tailings foam ceramics with controllable pore structures for high-performance thermal insulation. *J. Build. Eng.* **2026**, *118*, 114993. [[CrossRef](#)]
69. Zhou, W.; Ma, Z.; Ye, G.; Yuan, B.; Zhang, Z.; Zhao, D. Optimizing the pore structure and properties of alumina-based foamed ceramics through curing condition prioritization. *J. Alloys Compd.* **2025**, *1022*, 180065. [[CrossRef](#)]
70. Laala, I.; Bouzidi, N.; Pérez-Villarejo, L.; Bouzeriba, H.; Eliche-Quesada, D. Physico-mechanical, microstructural and thermal properties of foamed ceramics based on Algerian phosphate wastes. *Eur. J. Environ. Civ. Eng.* **2026**, *30*, 2620441. [[CrossRef](#)]
71. Li, Y.; Yan, C.; Zhang, J.; Li, J.; Zhao, J.; Cao, F. Experiment and prediction on performance of all-solid-waste foam ceramics. *J. Mater. Cycles Waste Manag.* **2026**, *28*, 2913–2929. [[CrossRef](#)]

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