

MgO· n Ga₂O₃ ($n = 0.98 - 1.2$) Transparent Ceramics with Longer Infrared Cut-Off Edge

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Abstract

In the pursuit of infrared transparent ceramics derived from MgGa₂O₄ with an extended infrared cut-off edge, transparent ceramics of magnesium gallate spinel (MgO· n Ga₂O₃) were synthesized by means of solid-state reactive sintering, followed by hot isostatic pressing (HIP) using MgO and Ga₂O₃ powders as raw materials. This approach yielded novel transparent MgO· n Ga₂O₃ ($0.98 \leq n \leq 1.2$) ceramics, exhibiting an infrared cut-off edge at approximately 7.5 μ m. The impact of composition (n) on the phase formation, densification rates, and microstructure evolution was thoroughly investigated. The in-line transmittance of the ceramics (thickness, 1 mm) was over 80 % in the 3–5 μ m range. Notably, MgO·1.1Ga₂O₃ ceramics achieved 86.9 % transmittance at 4.6 μ m, presenting promising prospects for infrared applications.

Keywords: Magnesium gallate spinel, transparent ceramics, composition control, reactive sintering, microstructure evolution

I. Introduction

Transparent ceramics offer high transmittance, a broad transmittance range, and robustness, rendering them suitable for applications in infrared windows and domes in missile imaging and tracking systems^{1,2}. Traditional infrared optical devices are constrained by atmospheric transmission and are categorized into the 3–5 μ m and 8–14 μ m ranges³. Conventional materials such as sapphire, MgAl₂O₄, and AlON transparent ceramics have been prevalent for mid-infrared windows because of their theoretically high transmittance in the 3–5 μ m wavelength range. However, they typically exhibit short infrared cut-off edges (<6.5 μ m)^{4–6}. Moreover, aerodynamic heating results in a radiation increase, and multiphonon absorption at the mid-infrared cut-off edge increases rapidly with increasing temperature^{7,8}. These factors collectively lead to a significant reduction in transmittance and signal-to-noise ratio, ultimately rendering the window ineffective at the target wavelength. Therefore, there is an urgent need to develop new materials with extended mid-infrared cut-off edges to address this issue.

The intrinsic infrared absorption of materials relies on multi-phonon absorption, as determined based on lattice vibrations^{9,10}. Polycrystalline magnesium gallate (MgGa₂O₄), a member of the AB₂O₄ spinel family, exhibits exceptional mechanical properties and a high melting point (~1930 °C)^{11–13}. Additionally, the replacement of Ga–O bonds results in a significant decrease in the

maximum phonon frequency of MgGa₂O₄ compared to MgAl₂O₄^{9,14}, effectively extending the infrared cut-off edge to 7.5 μ m. This extension demonstrated a notable enhancement in the infrared transmission range compared to MgAl₂O₄. Hence, MgGa₂O₄ ceramics can be ideal candidates for window and dome applications.

Research on MgGa₂O₄ ceramics dates back to the 1960s, primarily focusing on their structure, microwave dielectric properties, and luminescence characteristics based on doping¹⁵. The crystal structure of MgGa₂O₄ exhibits a partially inverted spinel structure with an inversion parameter (x) ranging approximately from 0.67 to 0.9, where $x = 0$ represents a spinel structure and $x = 1$ indicates an inverse spinel structure^{16,17}. Wu *et al.*¹⁸ synthesized magnesium gallate powders using a solid-state reaction method and investigated the impact of sintering temperature on dielectric properties. Yang *et al.*¹⁹ explored the effects of non-stoichiometric ratio on the microwave dielectric properties of magnesium gallate ceramics. However, research on the preparation and properties of transparent MgGa₂O₄ ceramics is limited. Zhang *et al.*²⁰ synthesized magnesium gallate powders and Ni²⁺:MgGa₂O₄ powders by means of nitrate pyrolysis. The in-line transmittance of pure magnesium gallate ceramics and Ni²⁺-doped ceramics, obtained by pulsed electric current sintering (PECS), measured 73.3 % (at 1.2 μ m) and 72.6 % (at 0.85 μ m), respectively. Li *et al.* used MgGa₂O₄ powders synthesized by solid-state reaction, and transparent ceramics were obtained by means of pressureless sintering combined with HIP. The transmittance of the sample was ~50 % in the

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visible region and higher than 75 % in the infrared region. The infrared cut-off edge reached $7.4\ \mu\text{m}$ ²¹.

Spinel crystals typically exhibit a broad solution range, with defects varying according to their composition, which plays a crucial role in ion diffusion rates^{22,23}. In $\text{MgO}\cdot n\text{Al}_2\text{O}_3$ systems, the relationship between defects and components and their impacts on densification rates have been systematically investigated²⁴. For $n > 1$, cation vacancies are the primary defects, with separated vacancies potentially impeding grain boundary movement via solute dragging effects, thereby reducing grain boundary mobility²⁵. Conversely, for $n < 1$, oxygen vacancies are the predominant defects, significantly reducing the activation energy of ion diffusion, often resulting in grain boundary mobility over 100 times greater than that when $n \geq 1$ ^{22,26,27}. According to Chiang *et al.*²⁸, grain boundary mobility is closely linked to composition and plays a vital role in densification and microstructural evolution. Moreover, the alterations in the composition affect the infrared cut-off edge of $\text{MgO}\cdot n\text{Al}_2\text{O}_3$ transparent ceramics²⁹. Therefore, modulating the composition offers the potential to produce $\text{MgO}\cdot n\text{Ga}_2\text{O}_3$ transparent ceramics with superior optical quality.

In this study, transparent ceramics spanning a broad composition range ($\text{MgO}\cdot n\text{Ga}_2\text{O}_3$, $0.98 \leq n \leq 1.2$) were synthesized by means of reactive sintering in air followed by HIP treatment, employing commercial MgO and $\beta\text{-Ga}_2\text{O}_3$ powders as raw materials. The impact of composition on the phase formation process, densification rates, and microstructure evolution of $\text{MgO}\cdot n\text{Ga}_2\text{O}_3$ was investigated in detail. The objective was to achieve a uniform microstructure with minimal porosity, thereby obtaining transparent ceramic samples characterized by outstanding properties through composition design.

II. Experiment and Characterization

(1) $\text{MgO}\cdot n\text{Ga}_2\text{O}_3$ ceramics fabrication

High-purity MgO (99.99 %, 200 nm, TAIMICRON Chemistry Co., Ltd., Japan) and $\beta\text{-Ga}_2\text{O}_3$ (99.99 %, 340 nm, Aladdin Chemical Co., Ltd., China) powders

were employed as raw materials, and isopropanol (AR, Aladdin Chemical Co., Ltd., China) was applied as the solvent. The morphologies of the MgO and Ga_2O_3 raw powders are shown in Fig. 1. The powders were thoroughly mixed in a planetary ball mill, maintaining a molar ratio of $\beta\text{-Ga}_2\text{O}_3/\text{MgO}$ including 0.98, 1, 1.1, 1.2, 1.3 and 1.5. The slurry was dried at $65\ ^\circ\text{C}$ in an oven and subsequently sieved through an 80-mesh screen. After calcination at $800\ ^\circ\text{C}$ for 2 h in air to eliminate residual organic impurities, green bodies were formed by means of dry pressing at 20 MPa followed by cold isostatic pressing (CIP) at 200 MPa. The green bodies ($\varnothing\ 20 \times 4\ \text{mm}$) were then pre-sintered in air at temperatures ranging from 1 200 to 1 400 $^\circ\text{C}$ for 6 h. Subsequently, the pre-sintered bodies underwent HIP treatment at 1 450 $^\circ\text{C}$ for 3 h in argon under a pressure of 180 MPa to yield transparent samples. The volatilization of Ga_2O_3 and MgO was minimized by burying mixed powders during presintering and HIP. After annealing at $950\ ^\circ\text{C}$ for 6 h in air, the samples were ground and polished on both surfaces to a thickness of 1 mm for characterization.

(2) Characterization

X-ray diffraction (XRD, D8, Bruker, Berlin, Germany) was employed to examine the correlation between composition and phase-forming temperature. The density and open porosity of the pre-sintered samples were determined using the Archimedes method. The microstructural changes in the pre-sintered and HIPed samples were assessed by means of scanning electron microscopy (SEM, JSM-6390, JEOL, Tokyo, Japan) based on the observation of the fracture surfaces. The transmittance of the final transparent samples was measured using an ultraviolet-visible near-infrared (UV-VIS-NIR) spectrophotometer (Carry 5000, Varian, Seattle, USA) and Fourier transform infrared spectrometer (FTIR; IRAffinity-1S, SHIMADZU, Kyoto, Japan). Hardness and fracture toughness tests were performed under a 1 kg loading with the Vickers hardness instrument (TUKON-2100B, Instron Co., Norwood, USA).

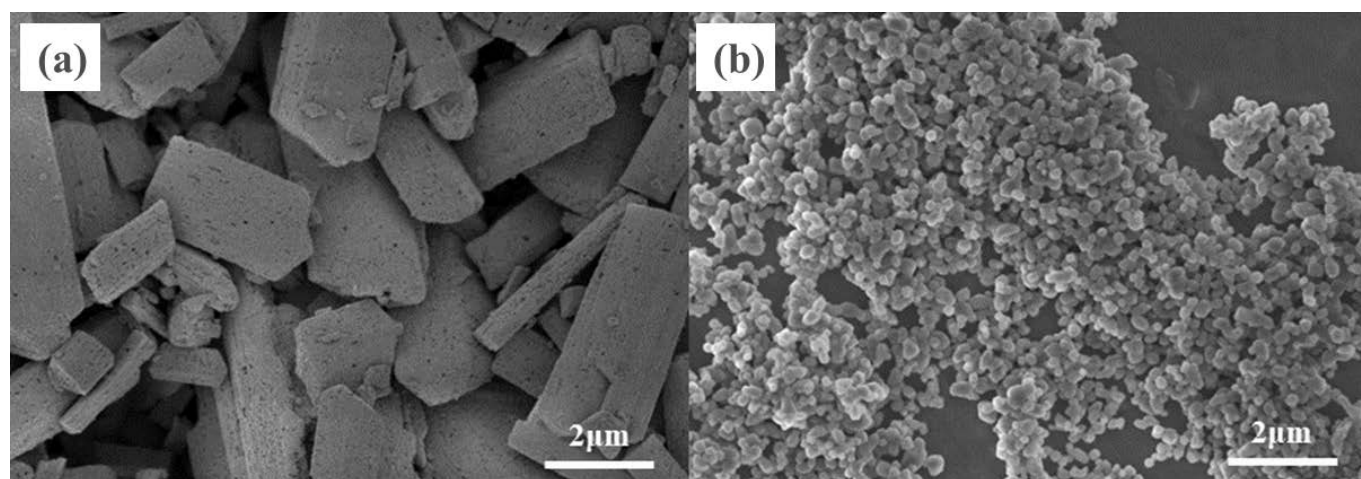


Fig. 1: SEM morphology of high-purity raw material powders of (a) Ga_2O_3 and (b) MgO.

III. Results and Discussion

(1) Effect of compositions on phase formation

The homogeneity of the mixed powders significantly affected the solid-state reaction. Effective solid-state reactions required thorough contact between the initial powder particles with comparable particle sizes. After ball milling, the mixed powders were examined based on backscattered images and energy-dispersive spectroscopy (EDS) testing (Fig. 2), revealing well-mixed initial powder particles and a homogeneous distribution of Mg and Ga elements. The solid-state reaction process adheres to the formula (1).



Fig. 3 illustrates the phase-forming process of the ball-milled powders (with compositions $n = 0.98, 1, 1.1, 1.2$ and 1.3) after heating at varying temperatures for 2 h. In the MgO–Ga₂O₃ system phase diagram, MgGa₂O₄ served as the sole intermediate phase. Previous studies have indicated that the formation of the spinel phase begins at 850 °C when MgO and Ga₂O₃ react¹⁸. For the Mg-rich composition ($n = 0.98$), both unreacted MgO and Ga₂O₃ coexisted with the spinel phase at 1 000 °C. The complete disappearance of the MgO and Ga₂O₃ phases and termination of the phase transition occurred at 1 100 °C. The stoichiometric composition ($n = 1$) was entirely converted to the spinel phase at 950 °C, the lowest temperature among all the compositions in this study. Conversely, for Ga-rich compositions, the phase forming temperature increased with increasing Ga₂O₃ content (n value). At lower temperatures, MgO and Ga₂O₃ initially formed a near-stoichiometric spinel phase. With increasing temperature, the residual Ga₂O₃ gradually dissolved into the spinel lattice, ultimately yielding a Ga-rich spinel phase. The phase formation temperatures of the compositions with $n = 1.1$ and $n = 1.2$ are 1 000 °C and 1 100 °C, respectively. However, when n increases to 1.3, the residual Ga₂O₃ phase persists even after heated above 1 500 °C, surpassing the phase-forming temperature that was indicated in the phase diagram. The phase-forming reaction was influenced by both thermodynamic and kinetic factors. Although the mixed

powders could form a pure spinel phase at elevated temperatures, the excess Ga₂O₃ could segregate to form a second phase during cooling. Consequently, obtaining a pure spinel phase was challenging when $n \geq 1.3$. Therefore, the compositions n ranging from 0.98 to 1.2 could primarily be considered in subsequent experiments. The phase-formation temperatures of MgO·nGa₂O₃ with various compositions are listed in Table 1. The reported solid solubility of MgO·nAl₂O₃ ranges from 0.6 to 9.1³⁰, with transparent samples achievable when $0.98 \leq n \leq 3$ ³¹. The solid-solution range of MgO·nGa₂O₃ was narrower, possibly because of its intricate structure.

(2) Effect of compositions on densification

The densification process of samples was characterized by their density, open porosity, and microstructures. Fig. 4(a) and 4(b) present the density curves and temperatures required to eliminate open pores for samples (the open porosity was less than 0.1 % measured with the Archimedes method) with varying Ga₂O₃/MgO ratios. Generally, the densities of the samples with different compositions exhibited an increasing trend with the increasing temperature. Owing to the elevated density of Ga₂O₃ and the lattice distortion induced by surplus Ga₂O₃ assimilation into the spinel lattice, the increased Ga₂O₃ content resulted in higher sample densities. The densification rates followed the order of stoichiometric > Ga-rich > Mg-rich. The sample with $n = 1$ achieved a closed-pore status (open pore-free) at approximately 1 270 °C. For the Mg-rich composition ($n = 0.98$), the most notable density increase occurred between 1 180 and 1 400 °C, with an open pore-free status achieved at 1 350 °C. This could be attributed to the limited capacity of the magnesium gallate spinel to accommodate excess MgO and its higher dissolution temperature. Conversely, in the Ga-rich composition ($n > 1$), densification typically began at lower temperatures than complete phase formation. Consequently, the extra Ga₂O₃ dissolved into the spinel phase concurrently with densification. The temperature at which the sample achieved a closed pore status increased with increasing n . Additionally, for $n \geq 1.3$, although the samples reached a closed-pore status,

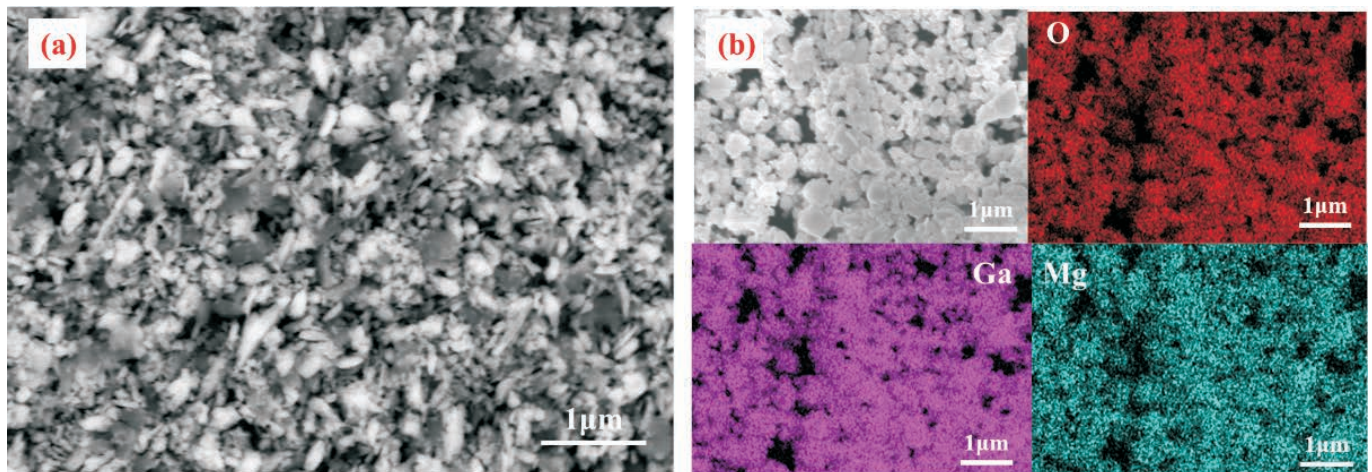


Fig. 2: (a) Backscattering images of the mixed powder and (b) EDS mapping profiles.

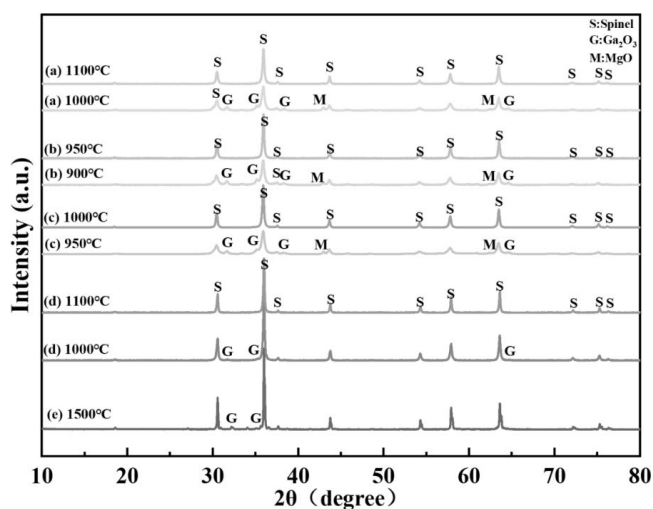


Fig. 3: X-ray patterns of $\text{MgO} \cdot n\text{Ga}_2\text{O}_3$ samples with (a) $n = 0.98$, (b) $n = 1$, (c) $n = 1.1$, (d) $n = 1.2$, and (e) $n = 1.3$ heated at different temperatures.

residual Ga_2O_3 coexisted with magnesium gallate spinel, as shown in Table 1. The temperature required for reactive sintering ($\sim 1300^\circ\text{C}$) was notably lower than that reported by Wu *et al.* (sintered at $1285\text{--}1435^\circ\text{C}$) and Li *et al.* (presintered at 1450°C and HIPed at 1575°C)^{18,21}. Their method involved the synthesis of MgGa_2O_4 powders via solid-state reaction before obtaining ceramic samples by means of pressureless sintering. Factors such as the initial

particle size and sintering activity may affect the sintering temperature, while the heat energy generated during the reaction sintering process can effectively enhance densification.

To examine the microstructural evolution during the sintering process, SEM characterization was performed on pre-sintered samples with various compositions. Figs. 5(a) – 5(c) illustrate the fracture surfaces of the samples with $n = 0.98$ composition sintered at 1330°C , 1350°C , and 1380°C for 6 h, revealing a uniform grain size distribution. Concurrently, as the temperature increased, both the density and the grain size of the samples increased. At 1350°C , intra-crystalline pores started to emerge, and at 1380°C , the average grain size sharply increased from approximately $1\ \mu\text{m}$ to $8\ \mu\text{m}$, resulting in numerous pores being encapsulated within the grains, which proved hard to eliminate even with HIP treatment. Figs. 5(d) – 5(f) present the fracture surfaces of the samples with $n = 1$ composition sintered at 1250°C , 1300°C , and 1350°C for 6 h. At 1250°C , the grain size remained below $1\ \mu\text{m}$, with a significant presence of residual pores, consistent with the density curve. As the temperature increased, the pores were gradually eliminated, without apparent intra-crystalline pores observed. However, thermodynamically stable large pores with coordination numbers ≥ 6 emerged in samples pre-sintered above 1350°C , proving challenging to eliminate even at elevated temperatures³², thus reducing the transmittance of the sample even after HIP treatment.

Table 1: Phase formation temperature for spinels with different compositions.

$\text{MgO} \cdot n\text{Ga}_2\text{O}_3$	0.98	1	1.1	1.2	1.3
Phase formation temperature/ $^\circ\text{C}$	1100	950	1000	1100	>1500

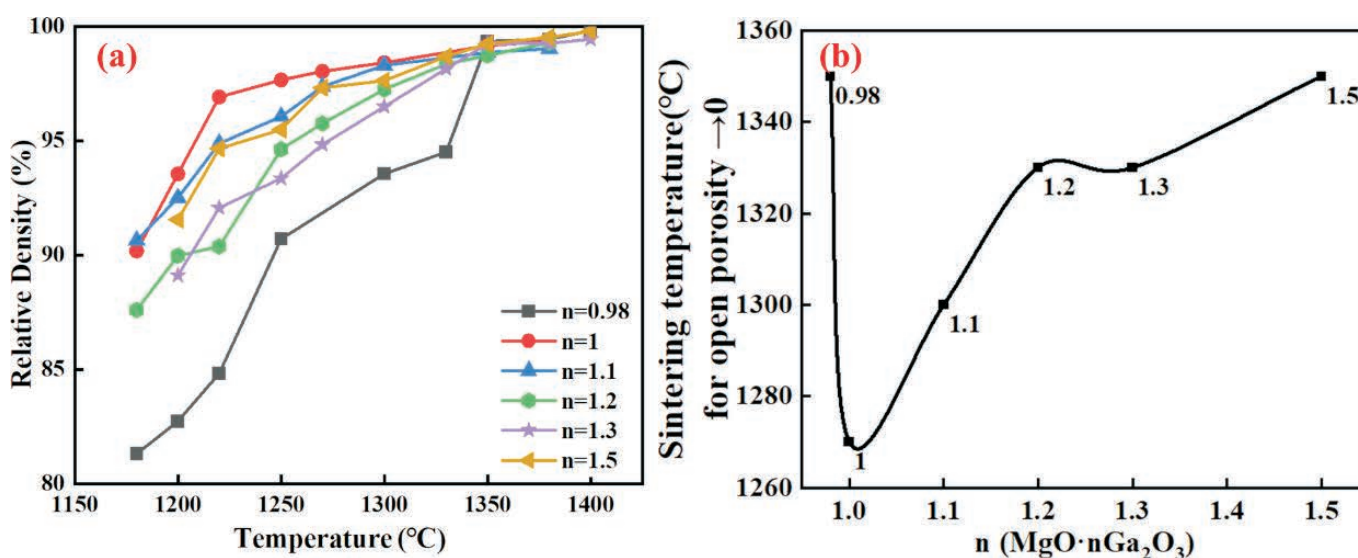


Fig. 4: (a) Relative densities of samples as a function of pre-sintering temperature, and (b) pre-sintering temperatures for open pore-free samples as a function of composition.

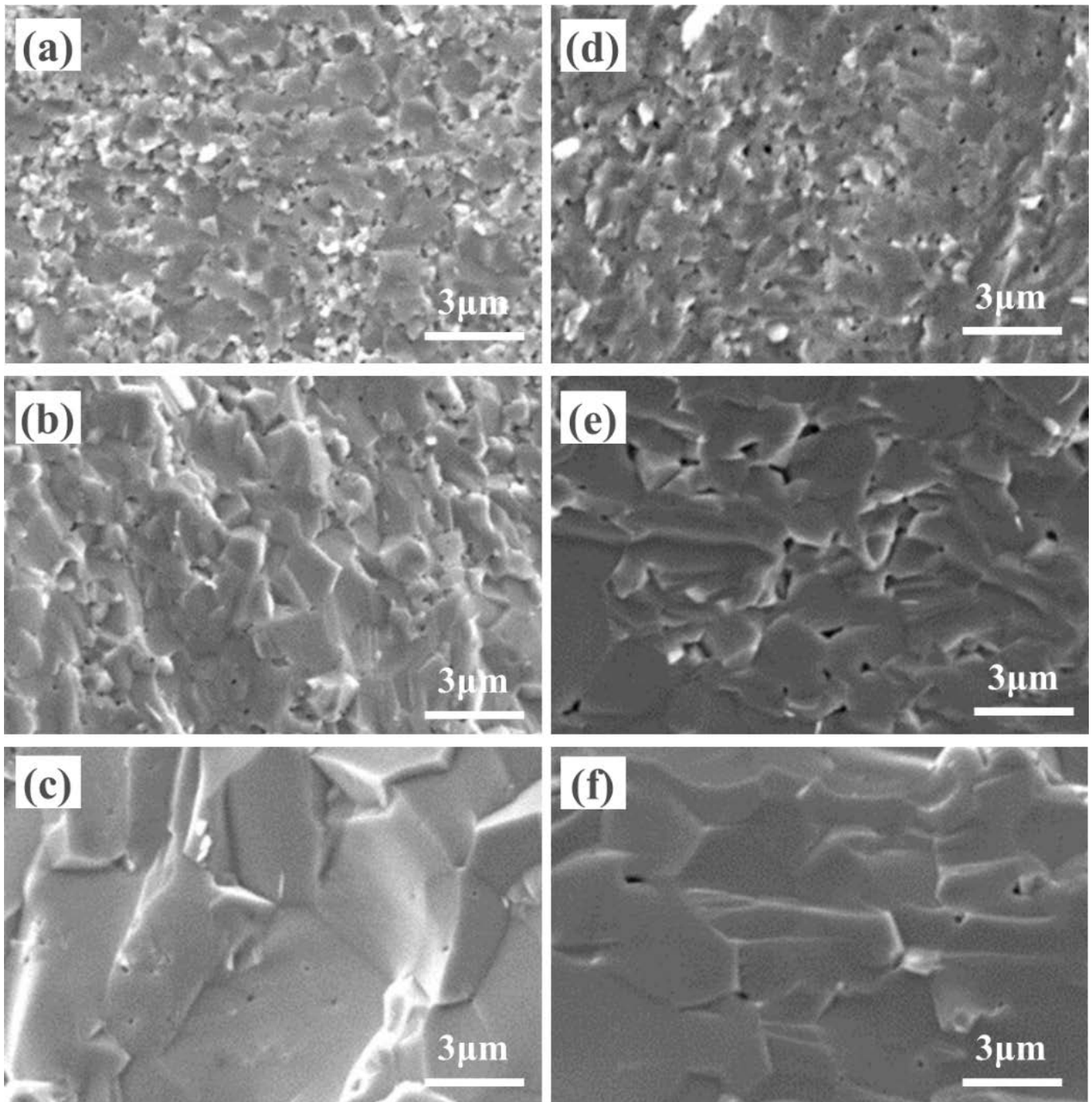


Fig. 5: SEM fracture surfaces of $\text{MgO} \cdot n\text{Ga}_2\text{O}_3$ samples. (a – c: $n = 0.98$, pre-sintered at 1330 °C, 1350 °C, and 1380 °C; d – f: $n = 1$, pre-sintered at 1250 °C, 1300 °C, and 1350 °C, respectively.).

The microstructures of the Ga-rich samples pre-sintered in air are shown in Fig. 6. A notable reduction in grain size and a decrease in pore size was observed with increasing Ga_2O_3 content at the same temperature, compared to the samples with $n = 1$. As the sintering temperature increased from 1250 °C (Figs. 6(a) and 6(d)) to 1350 °C (Figs. 6(c) and 6(f)), the grain size increased modestly, from 1 μm to approximately 4 μm. At 1300 °C (Figs. 6(b) and 6(d)), the grain size was approximately 2 μm, with most pores concentrated along the grain boundaries. A further temperature increase to 1350 °C resulted in the appearance of intra-crystalline pores in both composition samples.

In general, the densification rate of the sample followed the order of stoichiometric > Ga-rich > Mg-rich. The rapid densification rate in the stoichiometric sample led to pore aggregation and eventual formation of large pores. The Ga-rich compositions effectively mitigated the densification rate, facilitating pore elimination and preventing the formation of large pores. The Mg-rich samples required a high sintering temperature above 1380 °C to completely eradicate the open pores. At high temperatures, rapid grain growth occurred, resulting in the formation of intra-crystalline pores. Hence, the Ga-rich compositions were advantageous for producing pre-sintered samples with small pore size and no intra-crystalline pores.

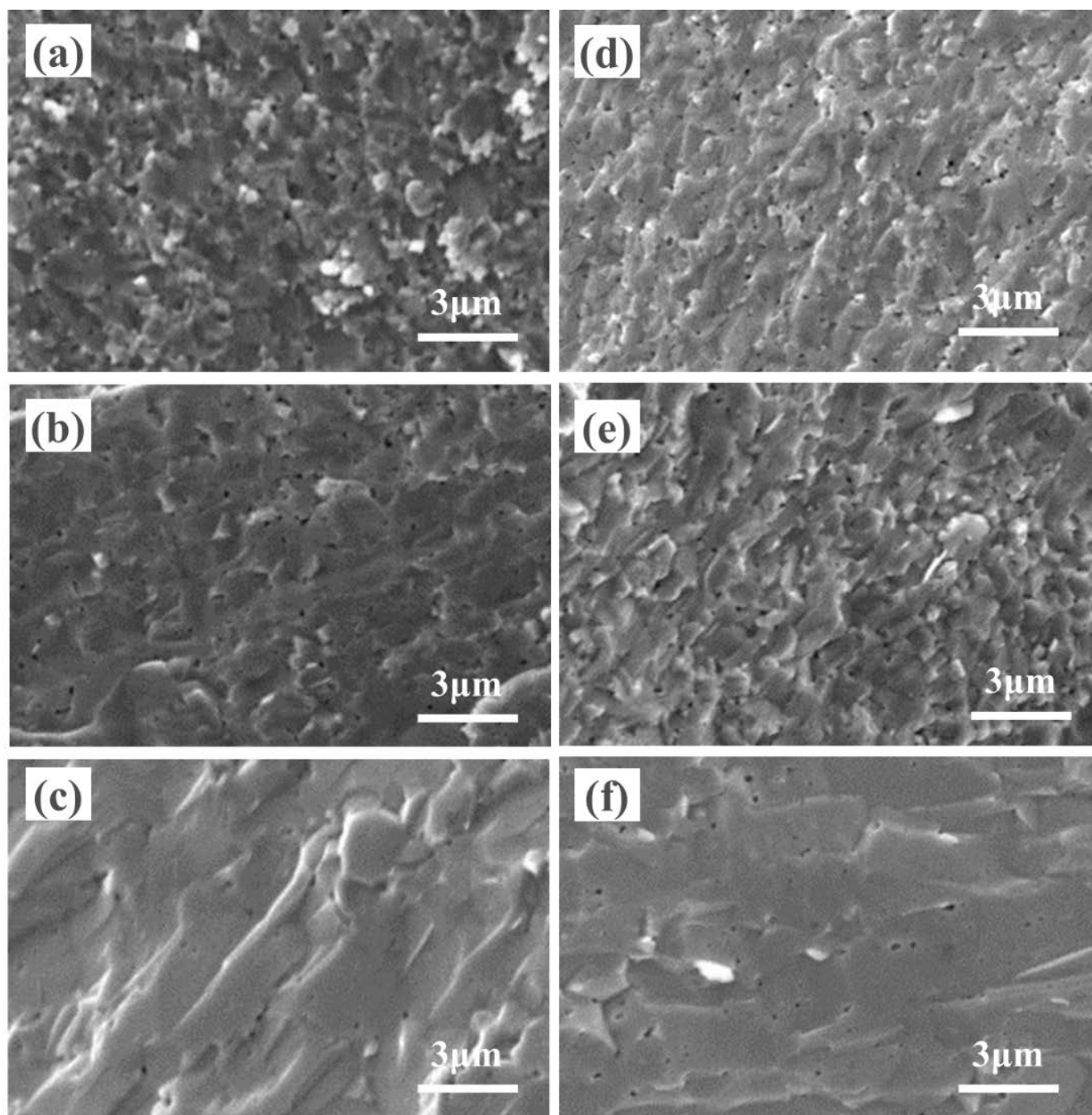


Fig. 6: SEM fracture surfaces of $\text{MgO} \cdot n\text{Ga}_2\text{O}_3$ samples. (a – c: $n = 1.1$, d – f: $n = 1.2$, pre-sintered at 1 250 °C, 1 300 °C, and 1 350 °C, respectively).

(3) Effect of compositions on properties of HIPed samples

Hot isostatic pressing (HIP) is an effective method for eliminating the residual pores and is commonly employed to produce transparent ceramics with high optical quality. Key considerations for HIP treatment include the relative density, open porosity, and microstructure of the pre-sintered samples. First, pre-sintered samples should reach a certain density level to ensure substantial elimination of open pores. Second, the pore size should be relatively small, and intra-crystalline pores should be avoided. The microstructures of the samples after HIP treatment at 1 450 °C are depicted in Fig. 7. Compared to the pre-sintered samples, the majority of inter-crystalline pores vanished, and the grain sizes significantly increased after

HIP treatment. For the HIPed samples with $n = 0.98$, the average grain sizes reached approximately 15 μm . Currently, the sample continues to display intergranular fracture, with a clear observation of the second-phase segregation at the grain boundaries in the backscattered image (Fig. 7(a)). Scattering from the second phases poses a significant challenge for transparent ceramics, leading to opacity. EDS tests conducted on the segregated second-phase grains revealed Mg enrichment, which is corroborated by the phase diagram¹⁵. For the stoichiometric ($n = 1$) and Ga-rich ($n > 1$) samples, the microstructure after HIPing was homogeneous. As n increased from 1 to 1.2 (Figs. 7(b) – 7(d)), the average grain size decreased from approximately 40 μm to 10 μm . This trend may be attributed to the smaller grain size of the Ga-rich samples

after pre-sintering coupled with the less pronounced grain growth of Ga-rich spinel during HIP process. In addition, certain nanosized pores were observed within the grains in samples with $n = 1 - 1.2$ (as shown in the red circles of Fig. 7(e) and (f)), potentially leading to a low in-line transmittance. The irregularly shaped pores in the $n = 1$ sample (Fig. 7(e)) likely stemmed from thermodynamically stable pores persisting at the grain boundaries during pre-sintering. The intra-crystalline pores with $n = 1.1$ and 1.2 exhibited similar microstructures, with sizes in the hundreds of nanometers (Fig. 7(f)), likely resulting from residual intra-crystalline pores in pre-sintered bodies. These pores can scatter light, particularly at wavelengths comparable to the

diameter of pores, potentially reducing the transmittance in the corresponding bands. This observation aligned with our predictions, which will be discussed later.

After HIPing, the samples appeared dark because of oxygen vacancies and graphite contamination. To recover the normal color, the samples were annealed at approximately 950°C for 6 h in air. Pores within the grains regrew during annealing, and exhibited strong light scattering at wavelengths comparable to their size³³. However, their impacts on the mid-infrared wavelengths can be ignored. The images of the annealed samples are shown in Fig. 8 (thickness, 1 mm). In-line transmittance testing was conducted

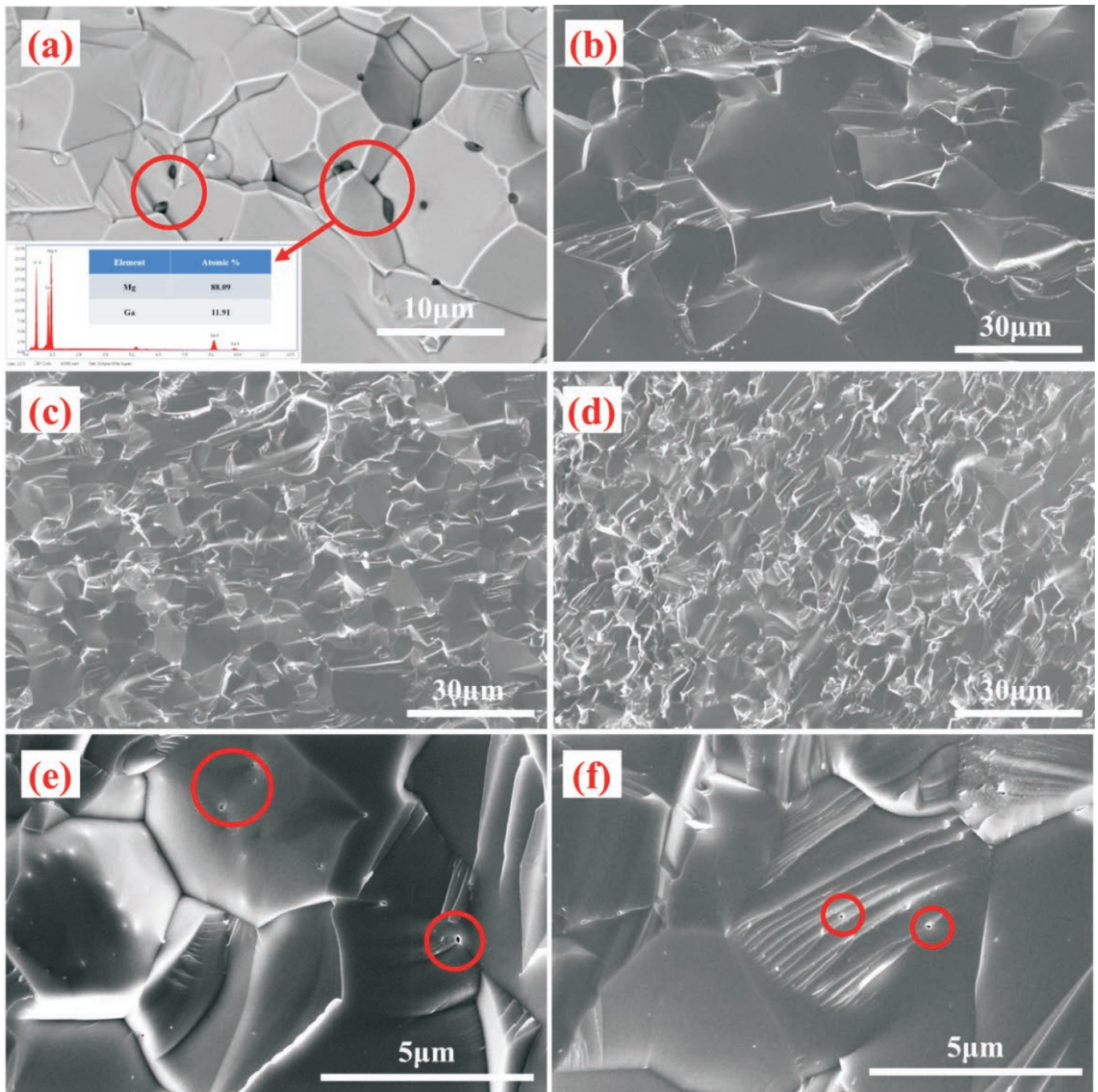


Fig. 7: SEM fracture surfaces of the samples HIPed at 1450°C . (a) $n = 0.98$; (b) $n = 1$; (c) $n = 1.1$; (d) $n = 1.2$; (e, f) larger magnification for (b) and (d), respectively. (The red circles represent intra-crystalline pores).

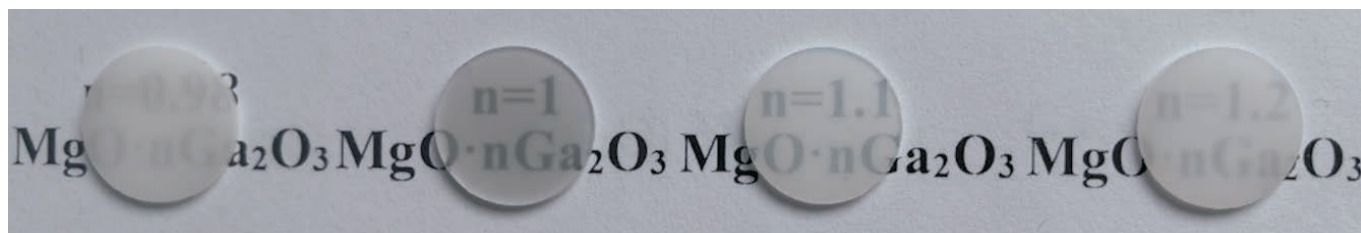


Fig. 8: Photographs of $\text{MgO} \cdot n\text{Ga}_2\text{O}_3$ transparent ceramics. (From left to right: $n = 0.98, 1, 1.1$, and 1.2 , respectively).

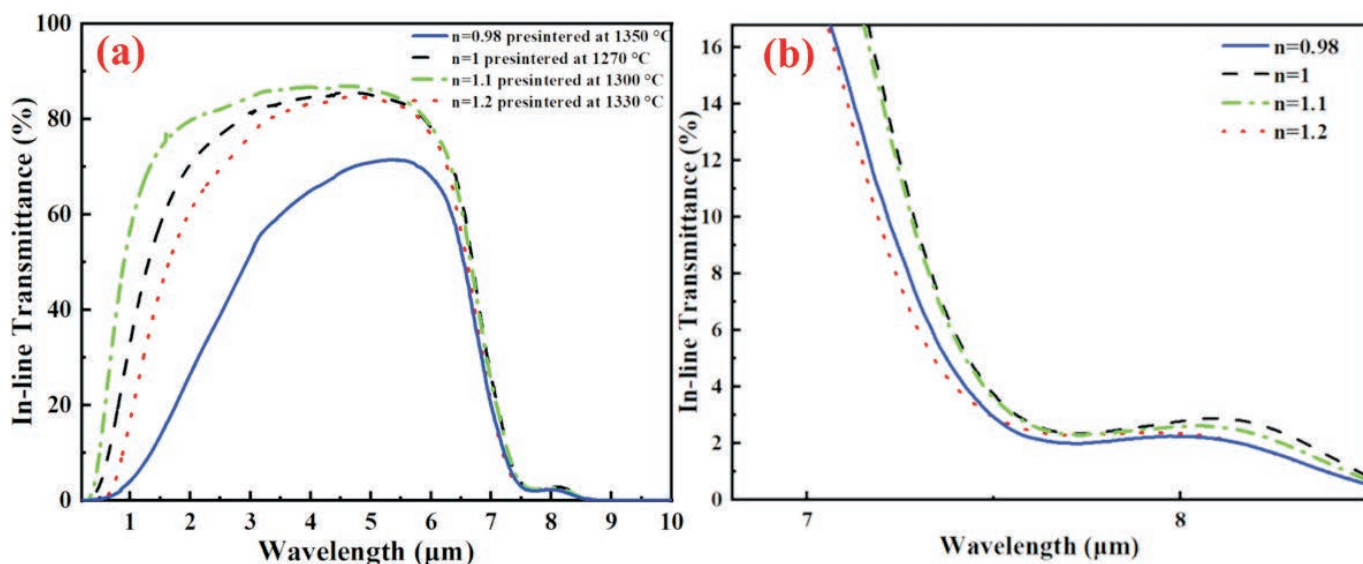


Fig. 9: (a) Transmittance and (b) infrared cut-off wavelength of $\text{MgO} \cdot n\text{Ga}_2\text{O}_3$ ($n = 0.98 - 1.2$) transparent ceramics.

on these samples and the results are shown in Fig. 9(a). The infrared cut-off edges for all the compositions were approximately 7.5 μm. The transmittance of the sample with $n = 0.98$ was notably low owing to the presence of the MgO second phase. Stoichiometric and Ga-rich samples exhibited excellent transmittance in the infrared range, albeit with a significant decrease in the visible range, which was attributed to residual pores. The $\text{MgO} \cdot 1.1\text{Ga}_2\text{O}_3$ demonstrated the highest in-line transmittance, consistently over 80 % between 3 and 5 μm, peaking at 86.9 % at 4.6 μm. As can be seen from Fig. 9(b), the variation of composition affects the infrared cut-off wavelength of the materials. The stoichiometric samples have the longest infrared cut-off wavelength. The Mg-rich samples have a slightly shorter cut-off wavelength, and the Ga-rich samples blue-shifted with the increase of n values. This may be due to the high number of defects and lattice distortions in the samples with non-stoichiometric ratios, which intensify the phonon vibrations and ultimately lead to the blue-shift of the infrared cut-off wavelength. The transmittance curves for MgAl_2O_4 (prepared by our group), $\text{MgO} \cdot 1.1\text{Ga}_2\text{O}_3$ (this work), MgGa_2O_4 (in ref. 21) and $\text{MgAl}_{0.5}\text{Ga}_{1.5}\text{O}_4$ (in ref. 32) are collected in Fig. 10^{21,34}. Consistent with prior discussion, the samples in this study exhibited lower transmittance in the visible region than that of MgAl_2O_4 , MgGa_2O_4 and $\text{MgAl}_{0.5}\text{Ga}_{1.5}\text{O}_4$, which was due to residual pores within the grains. However, the transmittance of $\text{MgO} \cdot 1.1\text{Ga}_2\text{O}_3$ in the infrared region was higher than that of MgGa_2O_4 and $\text{MgAl}_{0.5}\text{Ga}_{1.5}\text{O}_4$. The infrared cut-off

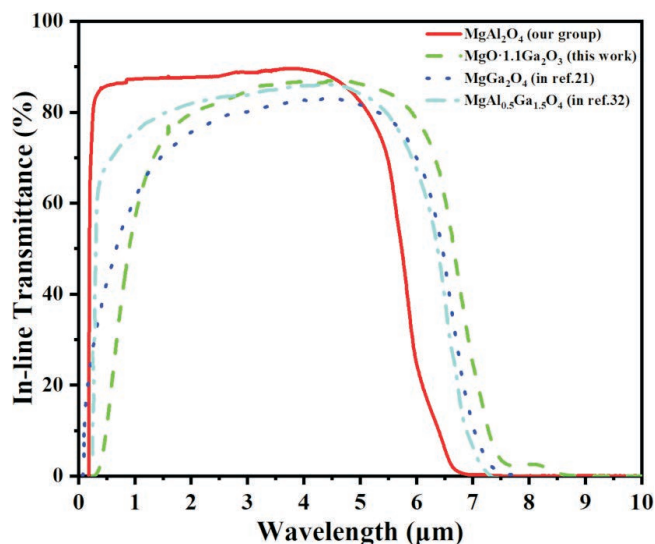


Fig. 10: Transmittance curves of MgAl_2O_4 (prepared by our group), $\text{MgO} \cdot 1.1\text{Ga}_2\text{O}_3$ (this work), MgGa_2O_4 (in ref. 21) and $\text{MgAl}_{0.5}\text{Ga}_{1.5}\text{O}_4$ (in ref. 32) transparent ceramics.

edge of $\text{MgO} \cdot n\text{Ga}_2\text{O}_3$ was longer than that of MgAl_2O_4 , with higher transmittance observed at wavelengths over 4.6 μm. Moreover, hardness and fracture toughness testing of the samples revealed that the sample with $n = 1.1$ had a hardness of 11.9 MPa and a fracture toughness of 1.14 $\text{MPa} \cdot \text{m}^{1/2}$, which was slightly better than that of the sample with $n = 1$ (hardness 11.6 MPa²¹, fracture toughness 1.04 $\text{MPa} \cdot \text{m}^{1/2}$). These characteristics render it an ide-

al material for mid-infrared domes and windows in high-speed aircraft.

As we know, several factors, including the composition, powder properties, and sintering parameters, contribute to the formation of secondary phases and intra-crystalline pores. Understanding the origin of these defects during reactive sintering and/or HIP in various compositions is crucial. By gaining deeper insights, we can identify more appropriate sintering regimes to mitigate defects such as intra-crystalline pores. Ultimately, this approach should yield transparent ceramics with enhanced optical qualities.

IV. Conclusions

MgO·nGa₂O₃ (0.98 ≤ n ≤ 1.2) polycrystalline transparent ceramics were successfully synthesized by means of solid-state reactive sintering combined with hot isostatic pressing using high-purity magnesium oxide and gallium oxide powders. The stoichiometric samples (n = 1) exhibited the lowest densification temperature (1 270 °C) during reactive sintering. Increased gallium oxide content effectively suppressed grain growth. However, excess gallium oxide complicated densification, leading to intra-crystalline pores formation and reduced transmittance in the final HIPed samples. The optimal transmittance was achieved for n = 1.1, reaching up to 86.9 % in-line transmittance at 4.6 μm, which was close to the theoretical value. Compared with MgAl₂O₄, the infrared cut-off was significantly broadened (~7.5 μm). Nevertheless, further optimization and exploration of the raw material selection, reactive sintering, and hot isostatic pressing conditions need further study for gallium-rich compositions.

Acknowledgements

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