



Understanding the thermal behaviour of MSWI bottom ash foamed ternary geopolymer: The fundamental role of pore structure

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ABSTRACT

Porous materials are widely used in refractory applications, but conventional foaming relies on energy-intensive equipment or chemical agents, increasing cost and carbon footprint. Municipal solid waste incineration bottom ash (MSWI BA) provides a low-carbon alternative, as its inherent metallic aluminium can produce foams in the alkali environment. However, uncontrolled gas generation may lead to expansion cracks within the matrix during hardening. In this study, a novel method is proposed to develop a porous Class F fly ash based geopolymer with inhibited crack formation by incorporating steelmaking byproduct ladle slag, which regulates the foaming process induced by MSWI BA. Through adjusting precursors composition, the pore structure is tailored, and systematically characterized to correlate with drying shrinkage, mechanical performance and thermal behaviour. Results indicate that the tortuosity of capillary network is the key factor retarding moisture loss of samples, which in turn affects drying shrinkage, with ultimate values as low as 306 microstrain. Moreover, interconnected macropore structure is found to determine thermal cracking resistance. After 1000 °C exposure, highly connected macropore structure largely maintains the structural integrity of samples, enabling an approximately 200% enhancement in compressive strength relative to room temperature performance. Accordingly, a conceptual model is proposed to understand the influence of pore structure on moisture movement at both ambient and elevated temperatures. This work offers practical insights into waste valorisation and pore structure design, guiding the customization of porous geopolymer to achieve enhanced high temperature performance.

1. Introduction

Fire safety engineering is vital in minimizing the risks associated with fire incidents through integrating a wide range of strategies. These include optimized structural design, implementation of fire detection and suppression systems, effective compartmentation, and safe evacuation planning [1]. While these aspects form the backbone of fire protection protocols, applying thermostable construction materials is increasingly recognized as another essential approach to passive fire resistance. As the most used cementitious building material, though ordinary Portland Cement (OPC) has been testified to exhibit basic thermal stability, its spalling, cracking and strength deterioration above 200 °C raise fire safety concerns [2]. Nowadays, geopolymer has gained considerable attention for its superior performance in high temperature applications. Unlike OPC, geopolymers maintain structural cohesion even at temperatures exceeding 800 °C [3]. This is primarily due to the unique gel composition, special physiochemical properties as well as

self-healing behaviour under elevated temperatures [2]. Furthermore, geopolymers also demonstrate no smoke or toxic gas emissions under fire exposure, adding to their safety profile [4]. Thus, they are suitable for diverse fire-critical applications such as tunnel linings, industrial flooring, structural insulation and protective coatings in energy or defence sectors [2,4]. The utilization of geopolymers as cementitious material in buildings not only enhances thermal protection but also supports sustainability through lower carbon footprint.

Geopolymer offers an eco-friendly and economical route by incorporating aluminosilicate-rich wastes into binder systems, such as fly ash (FA) [5–10], metakaolin (MK) [11,12], etc. Class F FA, a byproduct of coal combustion, is widely used as supplementary cementitious material for its high silicate and aluminate content, making it a promising precursor for geopolymerization. However, its low calcium availability results in slow hardening and unsatisfied mechanical performance, restricting the industrial application [8,9]. In order to further counterbalance the shortages of geopolymer and enhance high temperature

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properties, some researchers tried to develop a hybrid geopolymer system with co-precursors. Wang et al. [13] found admixing desulfurization gypsum into FA/ground granulated blast furnace slag (GGBFS) binary system can significantly improve early strength as well as strength growth rate, and also offset drying shrinkage since the formation of expansion phase ettringite during alkaline reaction. Alternatively, Adediran et al. [14] reported that alkali activated mortar with ladle slag (LS), blast furnace slag (BFS) and fayalite slag (FS) not only exhibits enhanced mechanical performance under ambient conditions but also shows denser microstructure and higher residual strength at 1000 °C, compared to solely based ones. Luo et al. [8] investigated hybrid Class F FA/LS geopolymers and demonstrated that with reactive calcium silicates and alumina-bearing compounds, LS can serve as a good co-precursor in geopolymerization, which significantly expedites early hydration and, moreover, contributes to continuous strength gain at elevated temperatures. Nevertheless, incorporating calcium-rich slags into geopolymers promotes the formation of C-A-S-H which is a source of significant drying shrinkage, generally 2–4 times higher than that of OPC, limiting the large-scale utilization [15]. The promotion of C-A-S-H also results in a co-existence C/N-A-S-H gel system, in which the incompatibility between two gels increases thermal cracking tendency and consequently reduces mechanical stability at high temperatures [7].

In addition to phase regulation, pore structure design of geopolymer based material could also play a crucial role in addressing the aforementioned problems. Collins et al. [16] demonstrated that pore structure governs shrinkage behaviour by comparing OPC and GGBFS systems. They observed that OPC samples show lower drying shrinkage despite experiencing more moisture loss than GGBFS samples. This was attributed to fewer macropores (25–5000 nm) in GGBFS concrete that reduce the ease of water evaporation, coupled with the higher fraction of mesopores (1.25–25 nm), where the moisture release and small menisci form, causing accumulated capillary pressure during drying [17,18]. It can be seen that targeted modification of pore structure provides a viable approach to mitigating drying shrinkage. However, limited research has been focused on pore tailoring to realize shrinkage reduction. Further, while the intrinsic thermostability of geopolymers stems from aluminosilicate framework, the pore structure is equally important in determining the high temperature behaviour [19–21]. Zhang et al. [19] found a highly connected pore structure facilitates vapour release from the matrix during heating, which limits pore pressure buildup and notably mitigates thermal damage. Zhang et al. [21] utilized an air pressure generator to produce pre-fabricated foams, which were then incorporated into a viscous FA based geopolymer paste. This resulting aerated geopolymer exhibits significantly reduced thermal conductivity and negligible thermal cracking, leading to excellent strength retention after 800 °C exposure. While previous studies have established that pore connectivity contributes to mitigating thermal damage, a systematic quantitative analysis linking multiscale pore structure to the thermal deterioration of geopolymers remains lacking.

There are several methods proven to formulate porous geopolymers, such as chemical foaming [22–25], mechanical foaming [12,21,26–29], implanting holey phases [30,31] and replica method [32]. But those aerations require either energy-intensive equipment or additional chemical agents, which increase production costs and exacerbate carbon emissions. In recent years, MSWI bottom ash (BA), often considered a problematic waste due to its heterogeneity and presence of metallic residues, has been engineered as an inherent foaming agent for alkali activated materials [33–37]. By adjusting mixing duration and alkali activator/MSWI BA ratio, Chen et al. [33] developed aerated geopolymers with the porosity up to 37.3%. Despite its great aeration potential, several deficiencies remain to be resolved: its intrinsically low reactivity as a precursor [38,39], the organic matter within BA retarding hydration [36], and the uncontrolled gas production introducing other problems, such as expansion cracks during hardening [35,36]. Furthermore, previous research primarily focused on properties under ambient conditions, while the influence of MSWI BA induced pore

structure on high temperature performance has not been systematically studied.

This study designs an eco-friendly aeration strategy, wherein MSWI BA is applied as an inherent foaming agent in Class F FA geopolymer system, while LS is incorporated as a foaming controller via early hydration promotion. Ternary geopolymers with enhanced volumetric stability and thermal structural integrity are developed. The effectiveness of the foaming process is evaluated through a multi-scale characterization using mercury intrusion porosimetry (MIP) and micro-computed tomography (μ CT). After that, the correlation of pore characteristics with drying shrinkage and mechanical strength under ambient conditions is clarified. More importantly, the mechanism by which pore structure affects the thermal behaviour of geopolymer is elucidated, with emphasis on cracking behaviour, densification and integrity retention. The results provide new insights into future design and tailoring of porous geopolymers, highlighting the impact of pore structure on high temperature performance.

2. Materials and methods

2.1. Materials

Class F FA (Vliegasunie, The Netherlands), MSWI BA (Heros, The Netherlands) and LS (Tata Steel, The Netherlands) served as solid precursors whose chemical compositions were determined by X-ray fluorescence spectrometry (Epsilon 3, PANalytical), and mineralogical phases were qualitatively analysed by X-ray diffraction (D4 PHASER applying Cobalt, Bruker), as presented in Table 1 and Fig. 1 respectively. Those precursors were dried and processed into powders before being used, for practical liquid/powder ratio and mixing homogeneity control. The particle size distribution of each solid was determined by a laser particle sizer (Analysette 22, Fritsch). As shown in Fig. 2, the characteristic values d_{10} , d_{50} and d_{90} across three precursor materials are relatively close, especially d_{50} and d_{90} . Sodium hydroxide (99.0 wt%, VWR), sodium silicate (waterglass with 27.5 wt% SiO_2 and 8.2 wt% Na_2O , Labshop) and deionized water were applied for preparing alkaline activator.

2.2. Sample preparation

The alkaline activator for each sample group was formulated identically with 1.4 of silicate modulus and 6 wt% of equivalent sodium oxide content, and it was prepared 24 h before sample preparation, following [7]. These mixtures contain namely unary system (sole FA based), binary system (FA and LS based), and ternary system (FA, LS and BA based). According to previous studies [7,8], the group G8.00 with 20 wt% LS substitution of FA was set as the representative of binary system. Ternary system samples were prepared as experimental groups to verify the synergistic influence among three precursors on pore structure formation. To ensure the adequate workability during fabrication, the minimum percentage of FA was fixed at 70 wt% with an activator

Table 1
Chemical compositions of precursors (wt%).

Oxide	Ladle slag	Fly ash	MSWI Bottom ash
SiO_2	3.56	51.53	25.95
CaO	43.83	3.39	25.17
Al_2O_3	30.17	24.84	14.55
Fe_2O_3	2.92	8.36	6.82
Na_2O	–	0.84	1.23
K_2O	–	2.16	1.87
MgO	3.19	1.15	1.34
SO_3	1.05	–	2.26
TiO_2	0.40	2.36	2.47
P_2O_5	–	–	1.93
Others	1.77	2.16	2.56
Loss on ignition	13.11	3.21	13.85

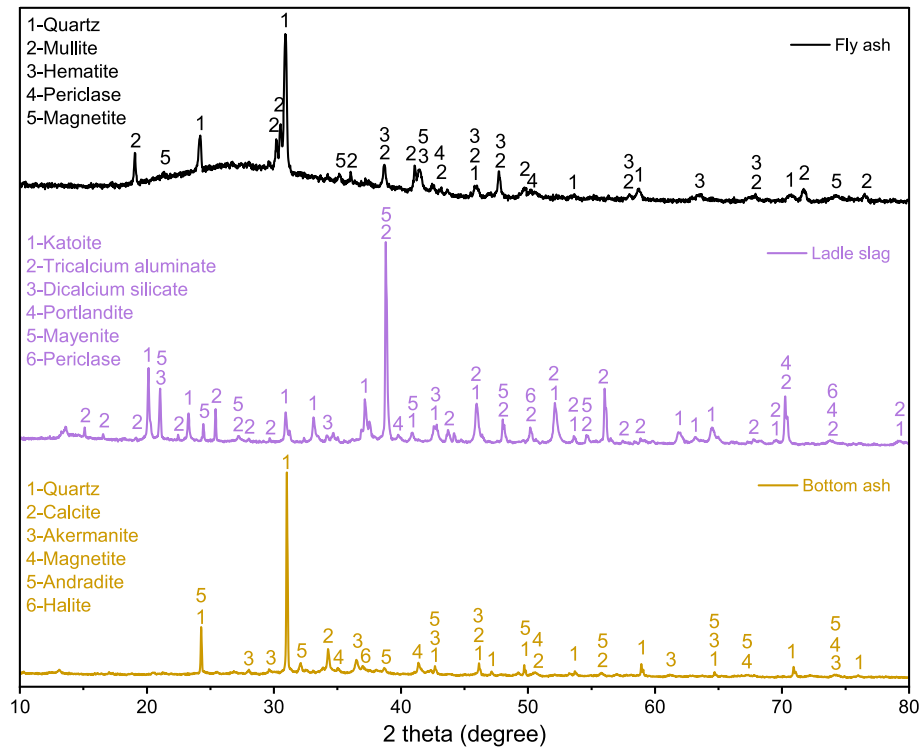


Fig. 1. XRD patterns of precursors.

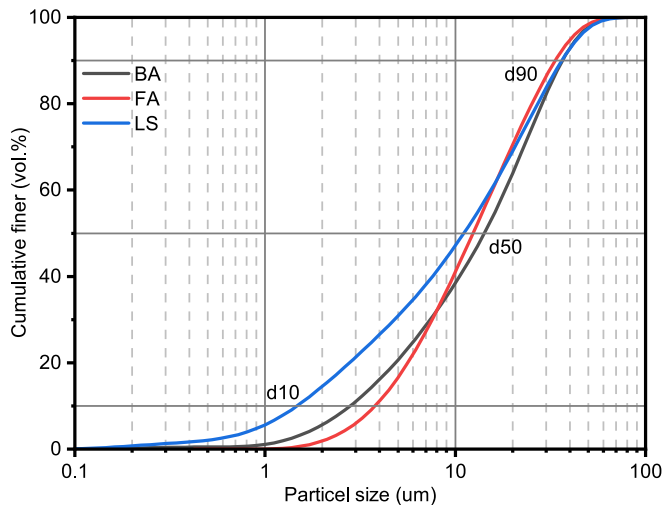


Fig. 2. Particle size distribution of precursors.

Table 2

Mix proportion and corresponding alkali solution of investigated groups.

Sample	Solid			Activator solution		Activator/ solid
	FA (wt %)	BA (wt %)	LS (wt %)	Na ₂ O (wt %)	Si modulus	
G7.10	70	10	20	6	1.6	0.5
G7.15		15	15			
G7.20		20	10			
G8.05	80	5	15			
G8.10		10	10			
G8.15		15	5			
G8.00		0	20			
G10	100	0	0			

where Na₂O (wt%) is a ratio of the mass content of Na₂O in activator solution to the total mass of solid admixed and Si modulus refers to SiO₂/Na₂O molecular ratio of prepared activator solution.

2.3. Experimental methods

2.3.1. Aeration potential test

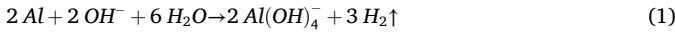
The aeration potential test is indispensable for evaluating MSWI BA as a non-conventional foaming agent, which can be interpreted from three perspectives. Firstly, the measured gas yield provides a quantitative profile of gas generation capacity of as received BA. Secondly, the determined gas yield can also be used to estimate the effective foaming content, which is metallic aluminium in the case of MSWI BA [33,36,37,39,40], facilitating direct comparison of the BA used in this study with similar foaming agents reported in the literature. Thirdly, repeatedly measuring different BA specimens from the same batch enables an assessment of consistency of foaming content within the raw BA, and this helps confirm that the differences in porosity observed among ternary geopolymer mixtures arise primarily from the designed BA dosage rather than the foaming content variation in the raw material.

The foam generation potential of as received BA was characterized by a displacement of water method. The water displacement method is a well-established approach for in-situ quantifying the volume of gas

solution to solid ratio of 0.5. Detailed mix proportions are listed in Table 2.

The sample preparation procedure can be illustrated in the following steps. i) Weighing precursors and then blending them in a mixer for 60 s at a low speed to reach a homogeneous state. ii) Mixing dry precursors with the prepared activator as per designed ratio at a medium speed for another 30 s. iii) Switching to a high speed to mix for another 60 s, and casting paste into plastic moulds afterwards. iv) The pastes within the mould were sealed using plastic foil, following a 60 °C curing for 24 h. As to foamed groups, sealing is conducted after the volume has stabilized. v) Demoulding samples and curing them under sealed conditions at 20 °C for further 27 days. Drying shrinkage specimens were placed into curing chamber after demoulding, as described in Section 2.3.6.

released during chemical reactions [40,41]. In this work, MSWI BA serves as the foaming agent, where metallic aluminium reacts with alkaline activator to generate hydrogen gas. Following the method proposed in Ref. [40], a 3 mol/L NaOH solution, rather than the designed activator, was applied as the reagent to ensure complete oxidation of metallic aluminium (Eq. (1) [37,41]) and thereby accurately estimate the effective Al content available for foaming.



As shown in Fig. 3, the sodium hydroxide was added into the reactor via the pressure-equalizing dropping funnel, allowing controlled addition of reagent. And a Schlenk flask, placed on a magnetic stirring plate, served as the reactor where metallic aluminium oxidation took place. The produced gas would be passing through the rubber tube to the water displacement system and then quantified as the volume change in a graduated cylinder. The air impermeability was assured by Apiezon wax sealing at the connection points.

Before measurement, calibration of the setup was carried out by reacting certain amount of aluminium powder with 3 M solution of NaOH. As plotted in Fig. 4, the volume of hydrogen gas shows a close agreement between theoretical (based on Eq. (1)) and experimental yields that validates the effectivity of designed setup. The measurement was conducted on dry MSWI BA powder at masses of 5, 10 and 15 g. Specimens at each mass were tested in triplicate at room temperature.

2.3.2. Specific densities and porosity

The 28 days age specimens with nominal dimensions of $2 \times 2 \times 2$ cm³ were oven dried at 40 °C to reach a consistent weight before being tested. The bulk density was calculated from the mass and apparent volume of the samples. The apparent volume of each specimen was determined from its external dimensions. Four edges corresponding each dimension, length, width and height, were measured individually using a vernier calliper with a resolution of 0.02 mm. The average values of three dimensions were used for apparent volume calculation. The skeleton density was measured by a pycnometer (AccuPyc II 1340, Micrometrics). The porosity (pore volume) was determined based on Eq. (2). The measurements for each mixture were performed in triplicate.

$$\varphi = 1 - \frac{\rho_b}{\rho_s} \quad (2)$$

where φ represents porosity, ρ_b is bulk density and ρ_s is skeleton density.

2.3.3. Microstructure observation

The morphology of obtained geopolymers, including outer surface and cross-sectional microstructure, was examined by a Zeiss Axio Imager equipped with Zeiss Axiocam 305 colour.

The phase composition of hydration products was selectively characterized by a scanning electron microscopy equipped with energy dispersive spectroscopy (Phenom ProX, ThermoFisher Scientific). The

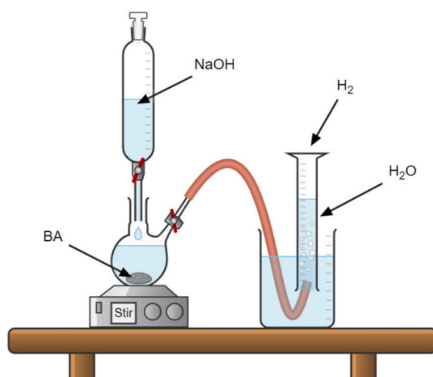


Fig. 3. Metallic Al quantification setup.

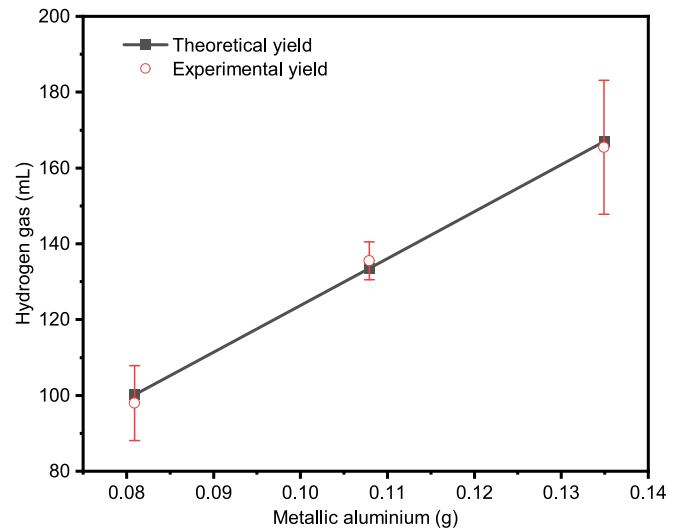


Fig. 4. Calibration curve of testing setup.

tested specimens were derived from crushed samples at 28 days, and embedded into epoxy resin afterwards. Before measuring, the embedded specimens were polished and coated with Pt to improve the quality of obtained images. The measurement was conducted in a backscattered electron mode under an accelerating voltage of 15 kV with a working distance around 7 mm.

2.3.4. MIP

Specimens were sieved from crushed samples at certain curing ages to obtain a particle size around 4 mm for MIP analysis. The test was performed by an AutoPore IV 9500, Micrometrics, including two phases adsorption and desorption. During the adsorption phase, a non-wetting liquid mercury is forced into open pores under increasing pressure, at which mercury entering the pores is inversely proportional to the pore size based on Washburn equation (Eq. (3) [42]). And then mercury is desorbed from the specimen as pressure releases, although typically some would be trapped inside pore network. MIP determines a wide test range from 5 nm to 100 μm for connected pores.

$$r = \frac{-2\gamma\cos\theta}{P} \quad (3)$$

where r refers to pore radius, γ is mercury surface tension, θ is contact angle between mercury and the pore surface and P represents applied pressure.

2.3.5. μ CT

μ CT enables the non-destructive visualization and quantitative characterization of pore structures within the resolvable size range through X-ray attenuation imaging. As X-rays pass through a specimen, the solid phase generally attenuates more X-ray radiation than the pore phase, resulting in relatively higher and lower greyscale intensities in the reconstructed images, respectively. In this study, a μ CT 100, SCANCO Medical AG, was applied to characterize samples that were sectioned into $2 \times 2 \times 4$ cm³ cuboid after curing for 28 days.

The X-ray tube was operated in a cone-beam configuration at 70 kVp and 200 μA, with air used as the scanning medium. During scanning, the specimen was mounted on a rotating stage and incrementally rotated to acquire X-ray projections from different angular positions. The transmitted X-rays were captured using a detector with the matrix size of 2048×2048 pixels. The central region (approximately 1 cm in height) of the cuboid was selected as the scanning range. The integration time was set to 200 ms per projection. Based on the Feldkamp-Davis-Kress method [43], the acquired projection data were reconstructed into a 3D volume comprising 1000 consecutive slices with an isotropic voxel

size of 10 μm .

The reconstructed greyscale images were subsequently processed using a Gaussian filter with a standard deviation of 0.8 voxel and a kernel size of 5 voxels to reduce high frequency noise while preserving the solid-pore interfaces. The processed images were stored as 16-bit greyscale images, as shown in Fig. 5a. A cylindrical volume of interest (VOI), with about 2 cm diameter and 1 cm height, was then defined by applying an inscribed circular region of interest (ROI) consistently across the stack of 1000 slices (see Fig. 5b). A global greyscale threshold was determined using a histogram-based method [44] and applied to all slices within the selected ROI. Thereafter, the greyscale images were segmented into a binary representation, with pore voxels assigned a value of 0 and solid voxels assigned a value of 1. The resulting 3D pore volume was quantitatively analysed by Avizo 2024.2 (ThermoFisher Scientific), as displayed in Fig. 5c.

2.3.6. Drying shrinkage test

After demoulding, the test specimens in size of $4 \times 4 \times 16 \text{ cm}^3$ were exposed in the curing chamber with constant temperature of 20°C and relative humidity of 60%. The original mass and length were recorded after the specimen temperature stabilized at 20°C (about 3 h exposure). The length was determined by a micrometre with $\pm 0.001 \text{ mm}$ uncertainty at certain ages and the mass was measured simultaneously. The expression of drying shrinkage is given by Eq. (4). Based on the sample length and detector accuracy, the measurement error range is 18 microstrain.

$$\mu\epsilon = \frac{L_0 - L_a}{160} \times 10^6 \quad (4)$$

where $\mu\epsilon$ refers to drying shrinkage, for each group measured in triplicate, L_0 is the original length (mm) of demoulded samples and L_a is the sample's length (mm) at certain ages.

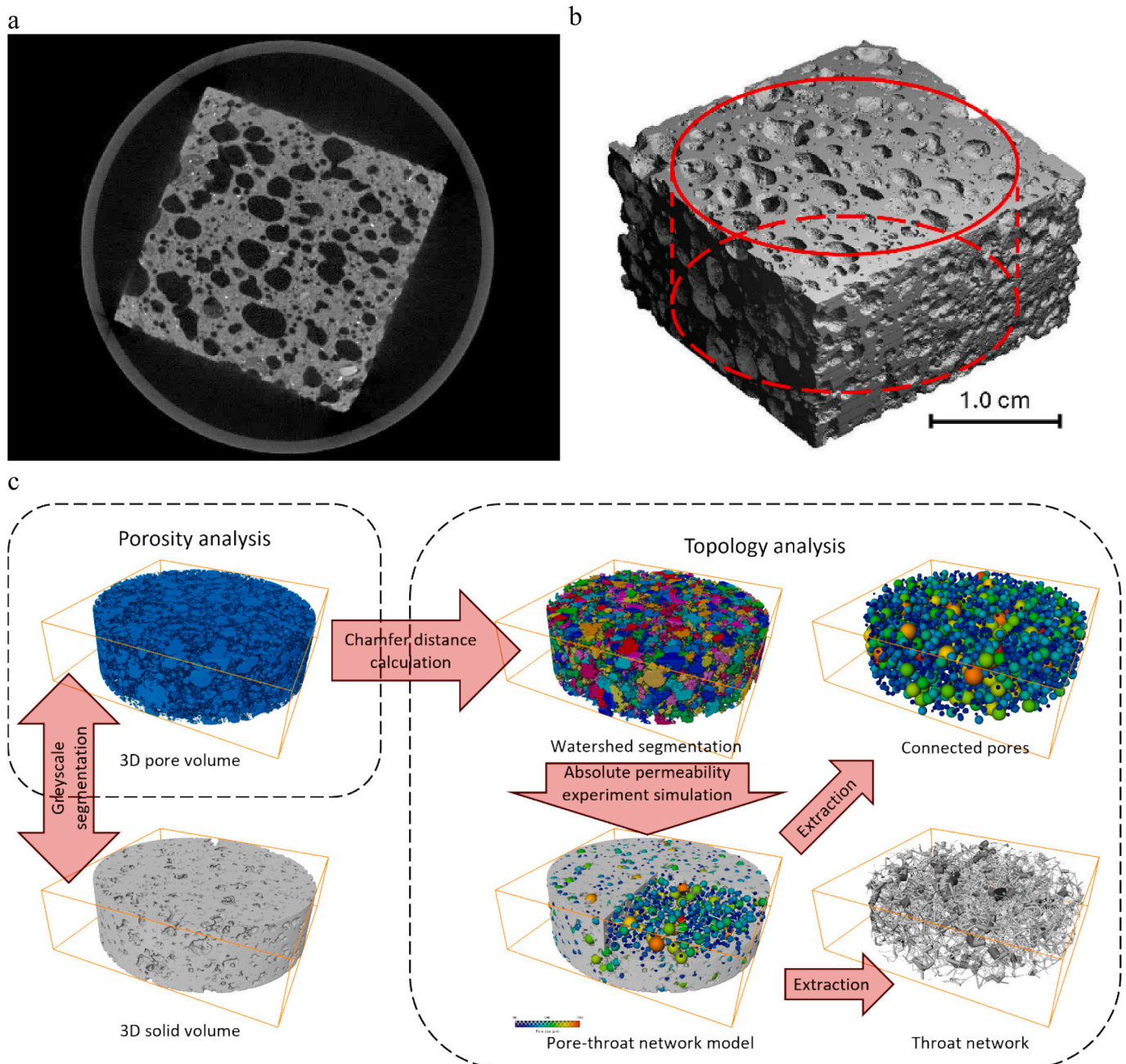


Fig. 5. (a) The radiograph of a slice. (b) VOI definition. (c) Schematic diagram of porosity and topology analyses.

2.3.7. Thermal conductivity

The specimens for thermal conductivity test were processed into $4 \times 4 \times 2 \text{ cm}^3$ slices from 28 days sealing cured prisms. Before being tested, sliced samples were vacuum dried at 40°C for 24 h to eliminate interior humidity. A heat transfer analyser (ISOMET model 2104, Applied Precision) was employed, connected with the one-sided surface probe applying a constant heat source to the sample. The measurement for each specimen was performed at room temperature in triplicate.

2.3.8. Mechanical strength test

The strength test was conducted on 28 d cured samples ($4 \times 4 \times 16 \text{ cm}^3$). According to EN 196-1 [45], flexural strength was determined through the three-point loading method with a loading rate of 50 N/s and then the halved prisms were used for compressive test under a loading rate of 2400 N/s. For each group, the mean flexural strength of three specimens and the mean compressive strength of six specimens were reported.

2.3.9. High temperature behaviour

Under real fire scenarios, the flame temperature can reach $800\text{--}1200^\circ\text{C}$ [46]. To investigate the thermal behaviour, 1000°C was set as the target temperature to simulate extreme but practical fire exposure. Additionally, the intermediate temperatures, 200 and 500°C , were included to reveal the progressive evolution of microstructure during heating. 28 d aged prismatic samples in size of $4 \times 4 \times 16 \text{ cm}^3$ were placed in the furnace and heated from room temperature with an elevating rate of $10^\circ\text{C}/\text{min}$. Once reaching to the set temperature, it was maintained for 1 h to ensure thermal homogenization in cross-sections of samples. Subsequently, the samples were cooled down naturally to ambient temperature for avoiding thermal shock.

The residual performance after high temperature exposure was assessed based on three aspects: morphology, physical properties (mass, residual shrinkage and bulk density) and mechanical properties (strength and microstructure). Similar with drying shrinkage test, high temperature volumetric change was characterized by the axial strain after 1000°C exposure. Residual mechanical strength tests were performed on 1000°C treated samples referring to the protocol in Section 2.3.8.

3. Results

3.1. Aeration potential

The metallic aluminium content of MSWI BA, which determines its intrinsic foaming potential, is estimated from the volume of hydrogen released during reaction with 3 M NaOH solution. Based on the theoretical stoichiometric relationship between the mass of metallic Al and the volume of H_2 generated (Fig. 4), the effective metallic Al contents are quantified for BA specimen with incremental masses of 5, 10, 15 g, as provided in Table 3. The results suggest that the metallic aluminium accounts for about 1.14 wt% of as received BA, corresponding to a gas generation capacity of approximately 14.3 mL per gram of BA. Additionally, the relatively low standard deviations of released H_2 obtained for the different specimen masses demonstrate a good consistency in metallic Al fraction of the BA.

Compared with MSWI BA produced in different regions [39], the MSWI BA sourced in the Netherlands contains a high fraction of reactive

aluminium. As presented in Fig. 6a and b, directly applying BA as the binder substitute would lead to a highly heterogeneous pore structure, accompanied by extensive expansion and internal cracking during paste hardening, consistent with previous studies [35,36]. Modifying the setting behaviour can improve fresh foam stability by enabling the matrix to develop sufficient rigidity, resisting pore ripening and coalescence [27,47]. Accordingly, LS is adequately incorporated into the system and expected to improve the compatibility between matrix stiffening and pore development during the early stages of reaction. This approach is anticipated to benefit from the presence of alumina-bearing compounds in LS, particularly tricalcium aluminate (C_3A) and mayenite (C_{12}A_7), as identified in Fig. 1, which promote early hydration process [8]. Moreover, the precursor content of BA is capped at 20 wt% to moderate aeration pressure, thereby synergistically reducing the tendency of foaming-induced cracking. The cross-sectional image of obtained aerated groups is presented in Fig. 6c.

3.2. Specific densities and porosity

Fig. 7 presents the bulk density and porosity of each group. Overall, G8.05, G8.10 and G8.15 (G8 series) exhibit better foaming effectiveness than G7.10, G7.15 and G7.20 (G7 series). G8.15 shows the highest foamability, achieving the lowest bulk density of $0.83 \text{ g}/\text{cm}^3$ which is approximately half that of unaerated groups. Although its BA addition is the same as that of G8.15, G7.15 displays limited foamability, with 53.68% porosity and $1.16 \text{ g}/\text{cm}^3$ bulk density. Compared with the reference group G8.00, its porosity improved by only 7%.

The disparity in foaming effectiveness between G7 and G8 series is associated with the BA/LS mixing ratio and the LS dosage. As shown in Fig. 8, the porosity in foamed groups increases with the BA/LS ratio. However, when the ratio is fixed at 1, G8.10 displays a porosity 11.3% higher than that of G7.15. This confirms that the accelerated hardening induced by LS may suppress the foaming process despite increased BA addition, thereby limiting porosity development. In this case, a multi-scale pore analysis employing MIP and μCT is applied to investigate the detailed pore structure differences among obtained samples.

3.3. Gel pore and capillary network

In order to analyse the pore-solid network systematically, the detected pores can be classified into different categories with certain width range [48,49], in this study, namely gel pore ($\leq 10 \text{ nm}$), capillary pore ($10\text{--}5000 \text{ nm}$) as well as macropore ($> 5000 \text{ nm}$). The classified porosity, pore distribution characteristics and structure tortuosity are summarized in Table 4.

As a consequence of hydration, gel pores are positively correlated to the degree of reaction and the gel pore volume refers to the quantity of gel generated. Most of LS involved groups show a gel porosity around twice that of sole FA based geopolymer. This can be explained by the effect of LS on the hydration products. As depicted in Fig. 9, the gel phase in the sole FA based system G10, characterized by high silicon, aluminium and sodium but low calcium content that its stoichiometric mass below 5 wt%, is assigned to N-A-S-H. In binary and ternary systems, higher amount of aluminium and calcium are incorporated into the gel phase and multiple gel types are observed, including C/N-A-S-H at Spot 1 of G8.00, N-A-S-H at Spot 1 of G7.20 and C-A-S-H at Spot 4 of G7.20, which is consistent with previous studies [7,9]. As LS dissolves, abundant Ca^{2+} and additional $\text{Al}(\text{OH})_4^-$ are released into the reaction environment, promoting the formation of calcium-enriched gels, which not only increase the overall gel content but also develop diverse gel types [7].

However, increasing BA/LS ratio inhibits gel pore formation in both G7 and G8 series. G8.15 with 5 wt% LS addition even exhibits a slightly lower gel pore fraction than G10. This is possibly due to the synergistic effect of the insufficient calcium source and BA induced retardation. On the one hand, with a low LS addition, the calcium supply is inadequate

Table 3
Metallic aluminium content quantification.

BA Mass (g)	H_2 release (mL)	Standard deviation (mL)	Al content (g)	Al fraction (wt %)
5	70.6	± 1.86	0.057	1.14
10	146.0	± 2.48	0.113	1.13
15	212.4	± 9.28	0.172	1.15

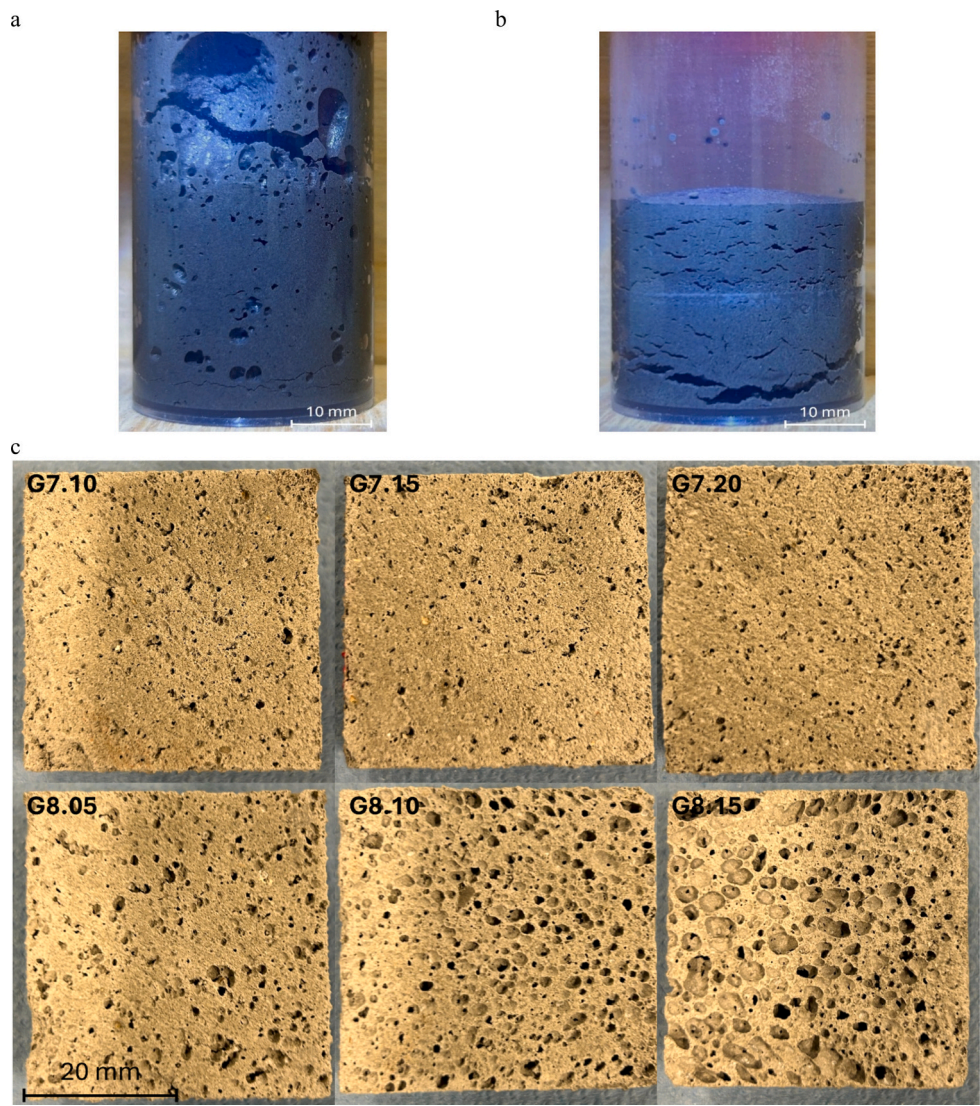


Fig. 6. Side view of sole BA based (a), FA/BA based geopolymers (b) and representative cross-sectional macrograph of ternary geopolymers (c) (apparent colour not representative of actual material colour due to imaging conditions).

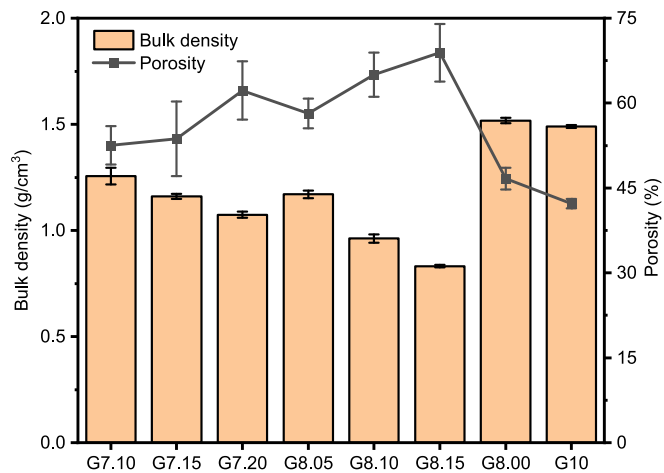


Fig. 7. Bulk density and porosity.

for gel promotion. On the other hand, due to its low pozzolanic reactivity, BA limitedly participates in geopolymerization [38,39], acting as fillers [36] that adsorb part of the activator and reduce its availability to

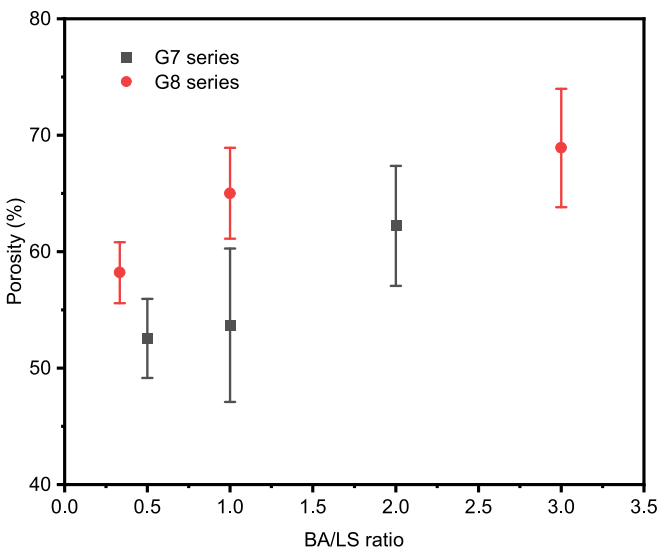


Fig. 8. Relation between porosity and BA/LS ratio.

Table 4

Summary of results from MIP study.

Sample	Porosity (%)				Modal pore width (nm)	Average pore width (nm)	Tortuosity
	Gel pore	Capillary	Macropore	Total			
G7.10	1.08	30.72	4.54	36.34	40.3	45.9	4.18
G7.15	0.78	28.50	4.78	34.06	50.3	48.3	4.20
G7.20	0.73	29.90	8.32	38.95	50.3	53.9	4.00
G8.05	0.94	33.66	3.19	37.80	40.3	44.4	4.18
G8.10	0.68	31.70	3.40	35.78	50.3	45.2	4.43
G8.15	0.40	30.73	2.40	33.52	50.0	45.1	4.51
G8.00	0.82	31.65	0.83	33.29	40.3	37.1	3.77
G10	0.44	30.65	0.74	31.83	32.4	33.8	2.97

reactive particles, thereby retarding precursor dissolution and gel formation.

The capillary porosity ranges from 28 to 34% across three systems, as displayed in Table 4. Although similar capillary porosities are observed in 8 groups, the capillary width distribution differs markedly. In order to comprehensively learn the capillary network, capillary pores are divided into small capillaries (10–50 nm), medium capillaries (50–100 nm) and large capillaries (100–5000 nm), as shown in Fig. 10. It is obvious that adding LS into unary system significantly increases medium capillary fraction. Compared with G10, the proportion of medium capillaries in G8.00 doubles to 28% representing the secondary major pore type. This redistribution is also illustrated in Fig. 11 that the bimodal curve transitions into a unimodal distribution and the modal pore width expands from 32.4 nm of G10 to 40.3 nm of G8.00. The increase in medium capillary volume is attributed to the conversion of metastable hydrates (C_2AH_8 and CAH_{10}), formed from mayenite in LS, into more stable phases such as hydrogarnet (C_3AH_6) and gibbsite (AH_3), which occupy less space than the original crystalline solids [50]. It should be noted that this conversion may occur considerably later than the foaming event in the aerated groups.

With the incorporation of BA, a higher fraction of large capillaries is observed. Fig. 10 gives a normalized fraction of pores that large capillaries account for approximately 30% of total porosity in the foamed groups, which is about 10% higher than that of the unaerated groups. Owing to aforementioned retardation effect of BA, the content of free water adsorbed on inert particles increases in ternary system, resulting in additional water-filled spaces left behind during hydration and thereby promoting large capillaries formation [51]. While, it is noteworthy that although MIP has well-known limitations in characterising pore structure, such as ink-bottle effect and possible pore damage induced by high pressure [52], the results remain suitable for comparative analysis among the investigated groups. Besides, the macroporosity determined by MIP excludes pores with the size larger than 100 μm and therefore the results should not represent the full macropore structure.

The geometrical tortuosity is the ratio of the actual flow path to straight-line length with the same entrance and exit point, which could reflect the complexity and indirectness of flow pathways. In this work, the tortuosity of samples is characterized using the corrugated pore structure model (CPSM) based on the determined MIP intrusion/extrusion cycle [53]. It was successfully applied for cementitious materials, as calculated by Eq. (5) [54–56].

$$V_T = 1 - \frac{4.996}{1 + \exp\left(\frac{\tau + 5.8032}{4.6242}\right)} \quad (5)$$

where V_T is the fraction mercury entrapment, as demonstrated in Appendix Fig. A, of total intrusion and τ is the tortuosity value.

It is found in Table 4 that the incorporation of LS significantly increases pore tortuosity, consistent with previous studies [54,55]. And ternary geopolymers generally have a more tortuous pore structure than unaerated groups. The increase of tortuosity can be linked to the

changes in pore size distribution. As illustrated in Fig. 11, adding LS into unary system broadens the pore width distribution, leading to a more heterogeneous capillary network. Consequently, the tortuosity increases from 2.97 in G10 to 3.77 in G8.00. Upon the addition of BA, the pore width distribution becomes further dispersed and the tortuosity of ternary groups exceeds 4. As shown in Fig. 10, a clear trend is observed within G8 series that the fraction of small capillaries increases with BA/LS ratio, which contributes to greater pathway complexity. As a result, G8.15 exhibits the highest tortuosity of 4.51 among foamed groups. However, in G7.20, the influence of capillary network on tortuosity is partially offset by the high content of macropores present in the specimen, which provide preferential flow pathways and thus lower overall tortuosity to 4.

3.4. Macropore structure

The MIP test reveals the network of gel pores and capillaries across different groups. However, due to restricted measurement range and MIP specimens' size, it shows limitations in detecting macropore structure, particularly in the aerated groups. Additionally, mercury can only intrude into connected pores, making MIP less effective for identifying isolated pores. Therefore, μCT is employed to provide complementary insight into macroporosity, enabling a more comprehensive analysis of the full-scale pore structure.

In previous studies [8,28,54], the result of μCT is commonly presented by pore fraction with respect to pore diameter. However, that is not an intuitive way to understand the characteristics of a highly porous structure, as shown in Appendix Fig. B. Thus, the log differential volume (Fig. 12) is computed to characterize pore width distribution of each group. According to the obtained skewed distribution, pores are concentrated at two ranges, namely Region I and Region II [28], and their structures are assessed respectively based on the 3D model as presented in Fig. 13. Region I signifies pores whose size smaller than 100 μm which can be considered as the entrapped air during sample mixing or freshly produced foams, and they are poorly interconnected in the matrix. Region II contains pores whose size larger than 100 μm resulting from ripening and coalescence of foams. The amount and size distribution of Region II pores are originally influenced by the slurry yield stress [27,47]. Based on categorised macropore model, no cracks were observed in Region II across all groups, evincing that the controlled foaming process used in this study does not introduce cracks (wider than 100 μm) into the solid phase. It should be noted that the pore width distribution of large pores involves uncertainties, which may be associated with the presence of metallic iron introduced by LS. Strong attenuation contrasts between dense metal particles and air-filled macropores can induce local greys value fluctuations and increase reconstruction artifacts in the segmentation large pores [57], as presented in Appendix Fig. C. Nevertheless, all samples were scanned and reconstructed under identical conditions, and therefore the relative comparison of pore characteristics among different groups remains valid.

As to unaerated groups G8.00 and G10, pores are fully concentrated in Region I, accounting for 14.53% and 12.8% respectively in Table 5. In contrast, MIP counterparts show less than 1% macroporosity, indicating

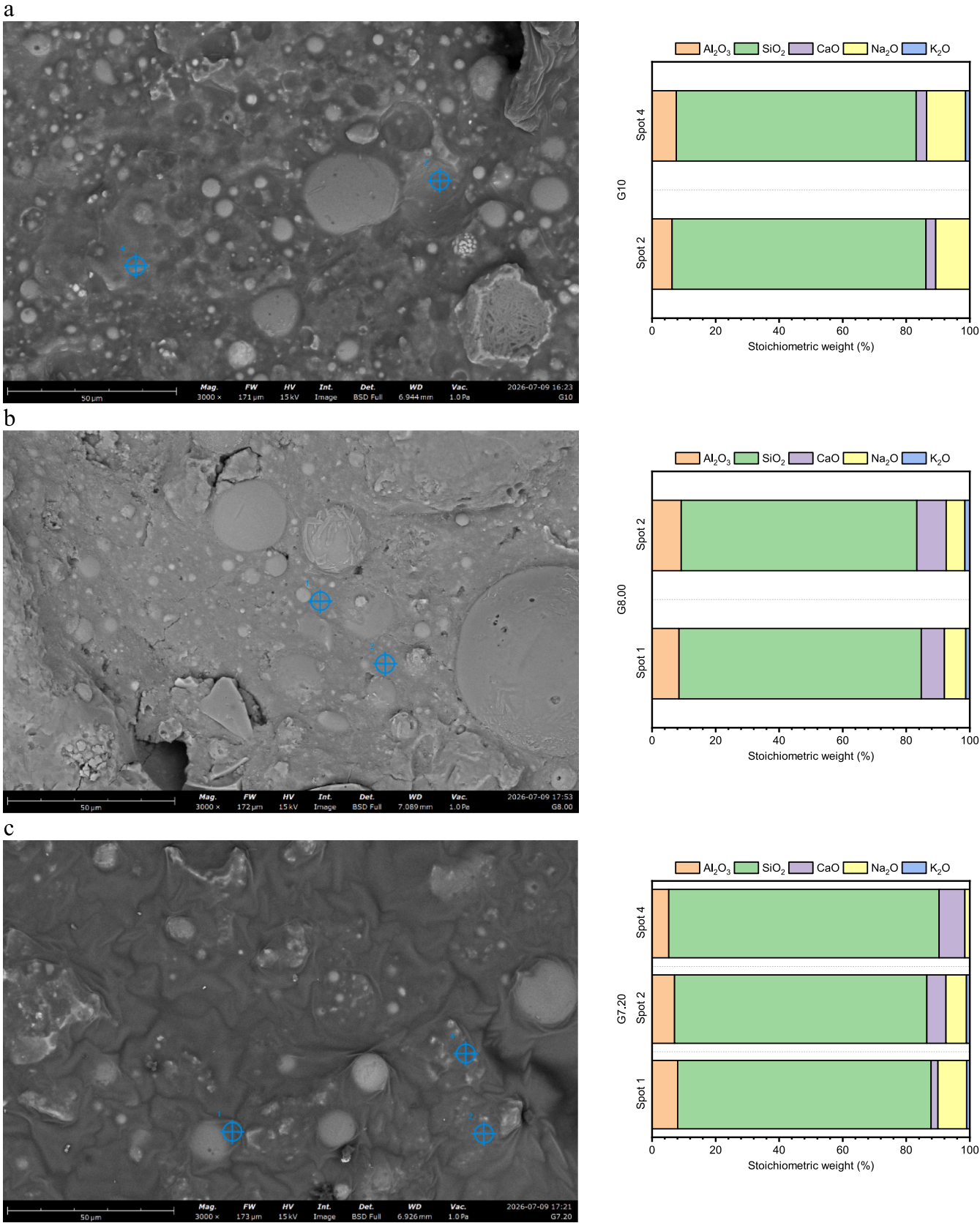


Fig. 9. Scanning electron microscope image of representative samples, unary system G10 (a), binary system G8.00 (b) and ternary system G7.20 (c), with local composition analysis.

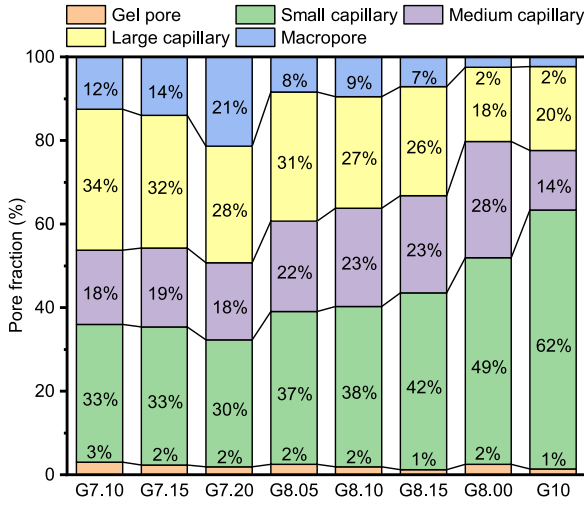


Fig. 10. Normalized fraction of pores.

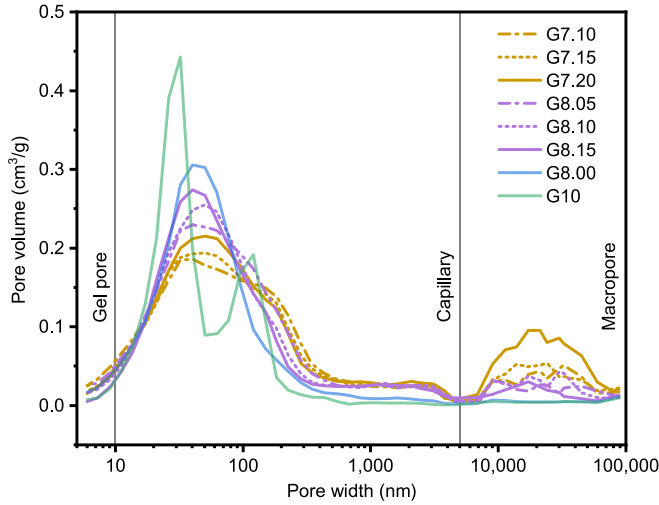


Fig. 11. Pore width distribution from intrusion curve.

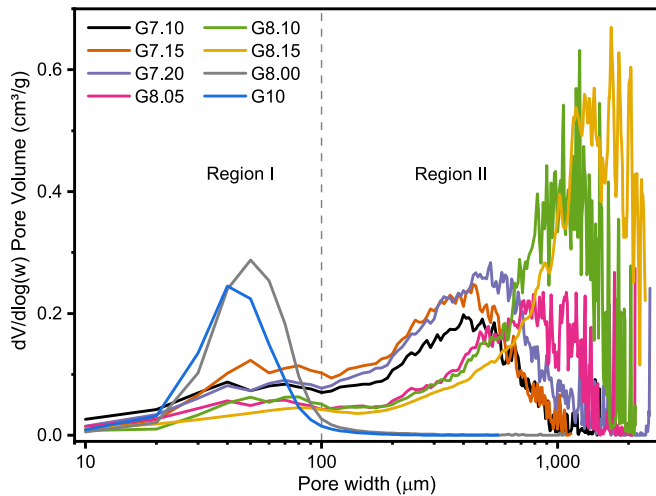


Fig. 12. Pore width distribution.

that Region I pores in unaerated groups are poorly connected. Compared with G10, the total porosity of G8.00 slightly increases by 1.98%. Due to the addition of LS, the tricalcium aluminate (C_3A) and mayenite ($C_{12}A_7$)

present in LS would intensively react with water that rapidly develops hydrate bonds between precursors particles, thus notably reducing mixture flowability [7]. As a result, more entrapped voids are detected in G8.00. In ternary system, a clear trend can be observed in Table 5 that increasing the BA/LS ratio leads to a higher total porosity as well as a greater fraction of Region II pores. Interestingly, increasing BA/LS ratio reduces the porosity in Region I, implying that enhanced foamability promotes the coalescence of smaller pores into larger ones. It is obvious that G8.15 has the highest macroporosity of 44.18% and the smallest specific surface area of $25.69 \text{ cm}^2/\text{g}$, as over 90 vol% of macropores are formed with the size larger than $100 \mu\text{m}$.

When the BA/LS ratio is fixed at 1, G8.10 exhibits a macroporosity of 36.89% with 85.5 vol% of macropores larger than $100 \mu\text{m}$, but G7.15 shows the macroporosity 10.34% lower than that of G8.10, and only 67.5 vol% porosity is constituted by Region II pores. Besides, both Region II modal pore width and mean pore width of G7.15 are significantly smaller than those of G8.10. It could be attributed to the increased LS accelerating the setting process [8] that facilitates the stabilization of fresh foams during the semisolid stage of slurry [27,47], which not only inhibits pore ripening but also limits pore interconnection, thus restricting macropore structure development. Correspondingly, as depicted in Fig. 12, the pore width distribution of G7.15 shifts leftward compared with G8.10 and, moreover, Region I area becomes wider, indicating more foams are stabilized at an early stage in G7.15.

To further investigate the connectivity and transport properties of developed macropore structures, a pore-throat network model is established from the extracted 3D pore volume in Avizo 2024.2. The evaluated network characteristics include pore connectivity, coordination number, throat width, tortuosity and permeability. Specifically, average coordination number describes the average number of neighbouring pores connected to a pore, while volume-weighted coordination number accounts for the contribution of pores with different volumes and reflects the connectivity of pores that occupy most of the pore volume, as calculated by Eq. (6):

$$CN_{\text{volume-weighted}} = \frac{\sum_{i=1}^n V_i CN_i}{\sum_{i=1}^n V_i} \quad (6)$$

where CN refers to the coordination number to a pore, n is the number of pore bodies, and V_i being the volume of the pore i . Connectivity density depicts how many throats presenting per unit volume. A lower connectivity density indicates a less interconnected pore network, which may reduce the number of available transport pathways. Both hydraulic tortuosity and absolute permeability are computed through numerical flow simulations based on the generated pore-throat network [58], following the Eq. (7) and (8), respectively.

$$k = \frac{\sum q_{ij} \mu L}{\Delta P A} \quad (7)$$

where k refers to the absolute permeability, assuming that the network is completely saturated with defined single phase fluid, ΔP is the pressure difference applied between the top and bottom boundaries of VOI, q_{ij} demonstrates the volumetric flow rate through the throat connecting pore i and pore j and the summation is performed on all throats across the network, μ is the viscosity of defined fluid, L represents the length of VOI in the flow direction, and A is the cross-sectional area of VOI.

$$\tau = \frac{\sum_{i=1}^n \|v_i\|}{\sum_{i=1}^n \|vx_i\|} \quad (8)$$

where n is the number of throats, v_i refers to the velocity of the fluid passing through the throat i and vx_i is the projection of v_i along the flow direction. It is noteworthy that the generated pore-throat network

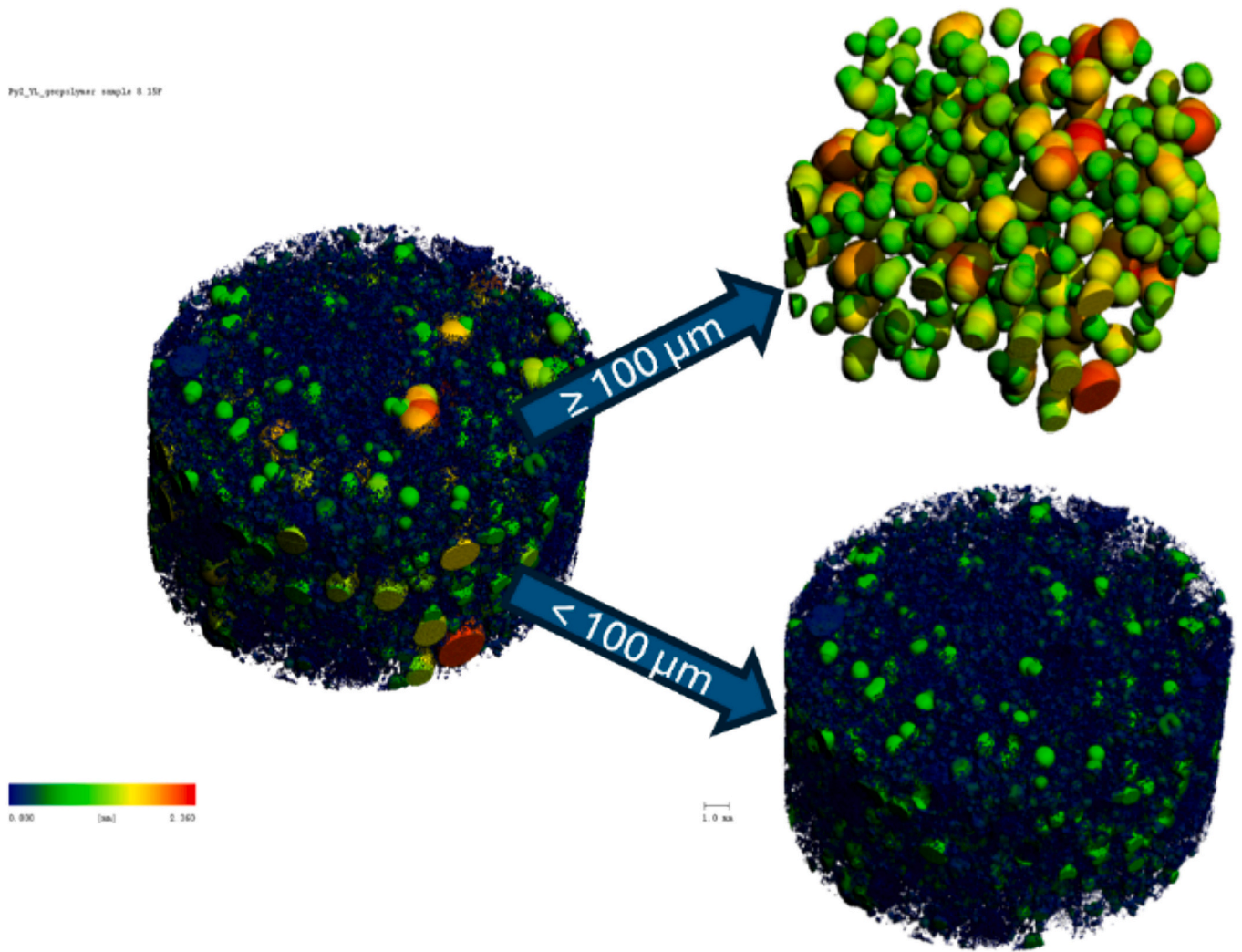


Fig. 13. Categorical analysis of macropores.

Table 5

Summary of results from porosity analysis.

Sample	Region I			Region II			Total porosity (%)	Mean pore width (μm)	Specific surface area (cm ² /g)
	Porosity (%)	Modal width (μm)	Dev. (μm)	Porosity (%)	Modal width (μm)	Dev. (μm)			
G7.10	8.21	40	27	15.97	400	244	24.18	284	44.33
G7.15	8.63	50	25	17.92	450	213	26.55	266	50.23
G7.20	7.74	70	30	23.92	520	352	31.66	387	43.18
G8.05	5.65	70	31	18.89	870	444	24.54	525	28.39
G8.10	5.36	80	38	31.53	1240	573	36.89	760	27.99
G8.15	4.26	80	38	39.92	1690	759	44.18	1098	25.69
G8.00	14.53	50	19	0.50	–	–	15.02	60	42.87
G10	12.80	40	22	0.24	–	–	13.04	51	45.53

where Dev. is the standard deviation to modal width.

represents a topological simplification of the continuous pore space, in which pore bodies and throats are algorithmically defined based on the distance transform and watershed segmentation [59,60]. Thus, the reported characteristics of connected pore network, as listed in Table 6, should be interpreted as descriptors of the network topology and transport property rather than exact geometrical measurements.

Aligning with porosity analysis, unaerated groups exhibit an isolated macropore structure with no continuous transport passways. In G7 series, increasing BA/LS ratio results in higher connected porosity, accompanied by improved pore connectivity, as reflected by increased

coordination number and connectivity density. Consequently, the absolute permeability of pore network rises from 0.13 μm² for G7.10 to 0.38 μm² for G7.20. Nevertheless, the hydraulic tortuosity also increases from 1.61 for G7.15 to 1.70 for G7.20. This can be attributed to the ink-bottle effect arising from the large difference between throat width and pore size. As shown in Table 5, the mean pore width of developed macropore structure is 387 μm in G7.20, whereas its estimated average throat width is only 73 μm. The pronounced mismatch creates flow constriction at the pore-throat junctions that reduces the local flow velocity within throats. However, increasing BA/LS ratio in G8 series leads

Table 6

Characteristics of connected pore network.

Sample	Connected porosity (%)	Average CN	Volume-weighted CN	Connectivity density (mm^{-3})	Average throat width (μm)	Hydraulic tortuosity	Absolute permeability (μm^2)
G7.10	21.95	3.42	6.03	2.39	98	1.72	0.13
G7.15	24.95	3.80	6.87	3.91	80	1.61	0.27
G7.20	30.73	4.97	9.27	7.84	73	1.70	0.38
G8.05	22.45	3.84	7.55	2.52	136	1.78	0.09
G8.10	36.07	3.69	10.89	4.09	142	1.81	0.64
G8.15	42.13	3.21	7.07	1.88	183	1.35	1.45
G8.00	–	–	–	–	–	–	–
G10	2.76	–	–	0.08	–	–	–

where CN refers to the coordination number to a pore.

to a reduction in coordination number. Especially, the average coordination number decreases from 3.69 for G8.10 to 3.21 for G8.15, while the corresponding volume-weighted coordination number drops to 7.07, the lowest value among G8 series. Unlike G7 series, pore coalescence becomes more frequent and obvious in G8 series, owing to the relatively low LS addition, as illustrated in Fig. 14. In G8.10, some interconnections between neighbouring pores evolve into merged pore structures engaging 2 or 3 pores. Compared with G8.10, more extensive pore coalescence and ripening produce substantially large pores in G8.15, with some large pores continuously coalescing with adjacent ones. This connection evolution would significantly shrink coordination number and connectivity density but effectively enlarge throat width. Therefore, the resulting interconnected macropore structure provide less resistance to fluid flow, obviously increasing absolute permeability of the network.

3.5. Drying shrinkage

The drying shrinkage behaviour of all specimens is studied over a period of 65 days, beyond which no significant changes are observed. As presented in Fig. 15, there are three stages, namely deformation, inflexion and stable stage. i) Deformation stage: the samples during the first 5 days are not sufficiently rigid, and exhibit a relatively free deformation. It is clear that foamed samples show a significant lower shrinkage compared to unary system. G10 shrinks linearly to 594 microstrain while G8.10 and G8.15 swell 36, 72 microstrain at day 4

respectively. The rest specimens have experienced a strain retention to different extents. ii) Inflexion stage: the evolution of shrinkage presents an inflexion, implying interior drying dominates the process [17]. The dry shrinkage of all groups obviously increases from the order of several tens microstrain at day 5 to a few hundred microstrain at day 13, except for G10. iii) Stable stage: after 14 days, the drying shrinkage reaches a plateau and afterwards there is no development or only a very slight increment. In general, G8 series perform better than G7 series and unaerated groups.

The moisture exchange takes place when freshly unsealed prisms are exposed to the curing chamber, which can be demonstrated by the mass loss curve, as given in Fig. 16. Overall, the stabilized moisture loss follows the order G8 series < G7 series < G8.00 < G10 which can be attributed to the different pore structure. In the first 5 days, the loss ratio sharply elevates to around 5 wt% of all groups since the evaporation of free water that adsorbed on the exposure surface, including the wall of open pores. It is evidenced by the positive relationship shown in Fig. 17 between water loss during deformation stage and specific surface area. As drying continues, water in capillaries begins to evaporate, corresponding to the aforementioned inflexion stage. The increase in water loss during the inflexion stage is governed by capillary tortuosity. The highly tortuous pore structure, such as G8.15, increases the effective path length for moisture movement that the vapour takes longer time to escape from the matrix at a given humidity difference. As depicted in Fig. 17, the moisture loss during inflexion stage shows a strongly inverse relationship with capillary tortuosity and largely determines the

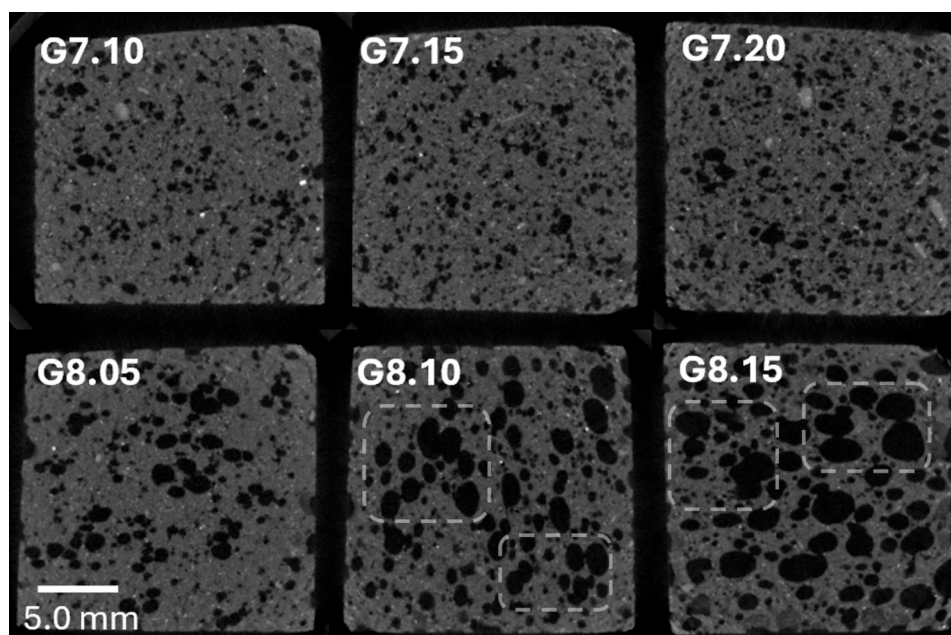
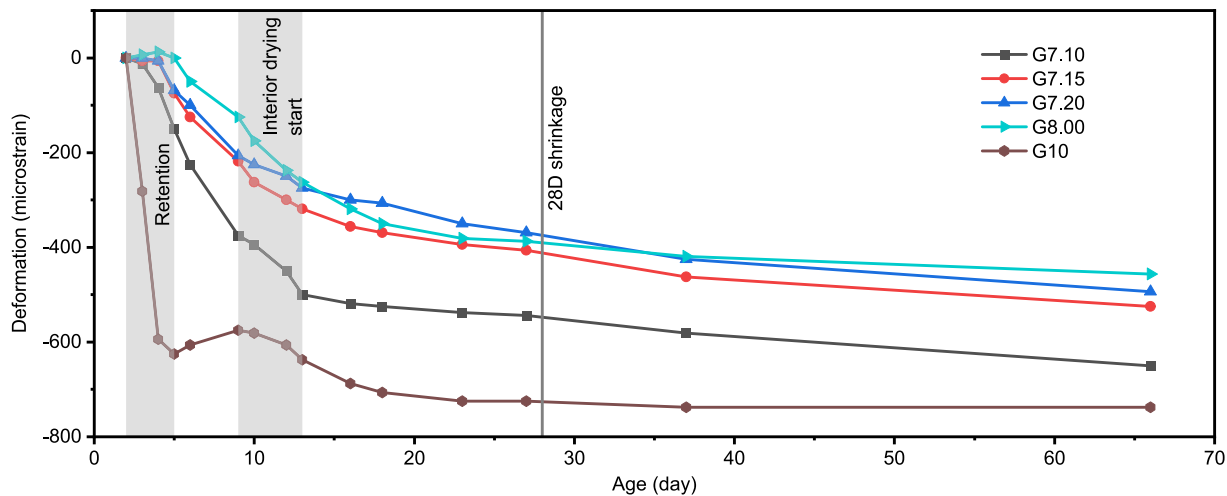


Fig. 14. Representative radiograph of foamed geopolymers. Merged pore structures are highlighted with the dashed rectangle.

a



b

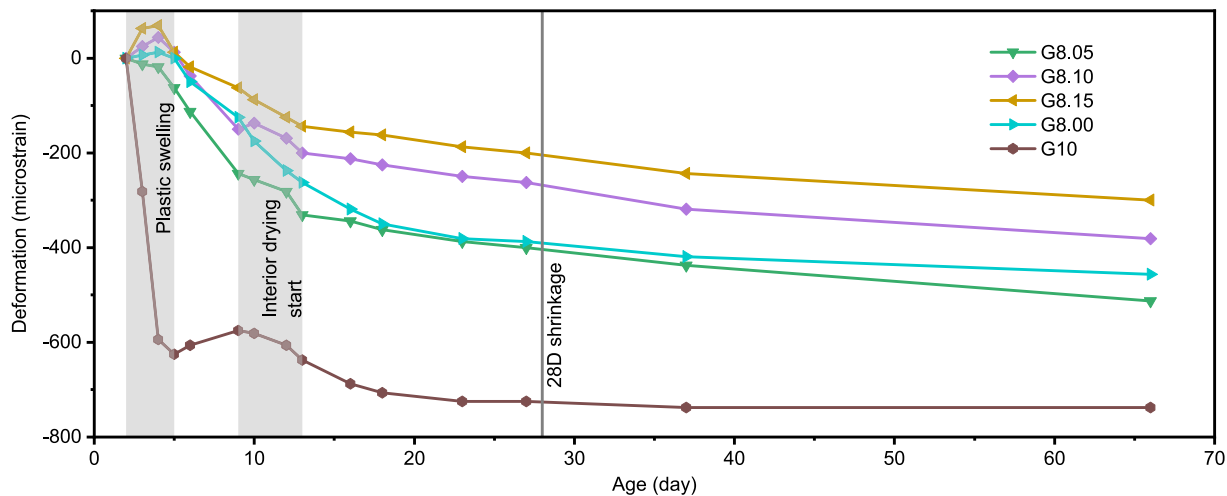


Fig. 15. Development of drying shrinkage with age. Reference groups with G7 series (a) and G8 series (b).

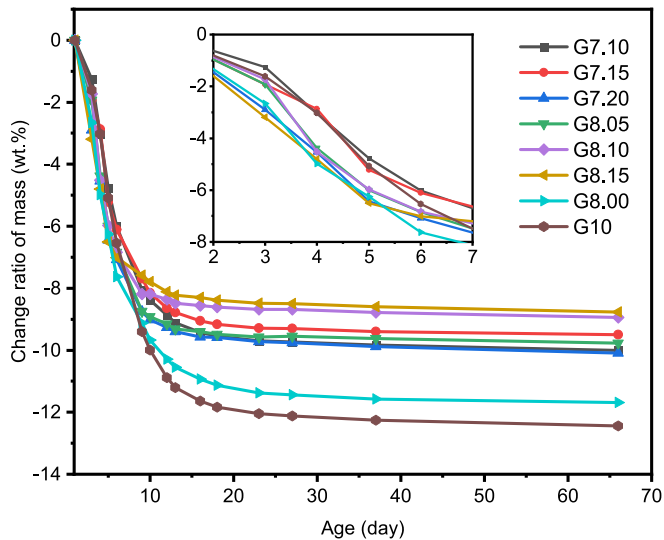


Fig. 16. Mass loss ratio at certain ages.

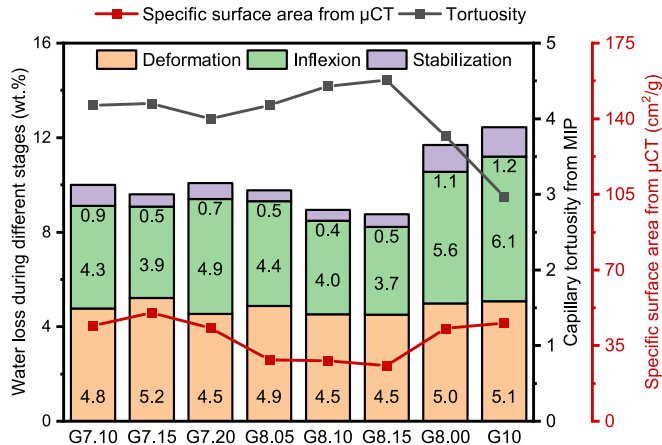


Fig. 17. Relation among mass loss during different stages, specific surface area and capillary tortuosity.

ultimate moisture loss ratio. After 14 days, the mass of all samples shows the tendency toward stabilization, implying the surface water layer is removed and the interior humidity reduces in a gentle manner. The

reference groups exhibit the most significant mass change at the end of measuring period. G7 series show the mass loss exceeding 9.3 wt% at day 28, which further increased to over 9.5 wt% by 65 days. G8.10 and G8.15 change the least with the mass loss of 8.68 wt% and 8.50 wt% at day 28, stabilizing at 8.94 wt% and 8.76 wt% at day 65, respectively.

The drying shrinkage can be predominantly attributed to the formation of menisci within the capillary pores during evaporation [51]. The meniscus would exert negative pressure in the remaining water that pulls the solid matrix inward, causing tensile stresses and volume contraction. The finer capillary pores create smaller meniscus, thus leading to higher tensile force and shrinkage. Based on the proposed micromechanical model [16,18], the statistical meniscus radius at different curing ages can be approximately resolved using MIP results, as shown in Appendix Fig. D. The test results and computed meniscus radii are listed in Table 7. Due to the tortuous capillary network, the moisture release is obviously restricted among foamed groups, resulting in larger meniscus radii at 28 d compare with unaerated groups. It is found that drying shrinkage is inversely proportional to meniscus radius for porous geopolymer, as plotted in Fig. 18. The statistical meniscus radius of G8.15 is 103.93 nm and correspondingly shrinkage is barely 198 microstrain at 28 d that much lower than other groups.

Moreover, the expansive admixture is also able to change shrinkage behaviour. The metallic aluminium oxidation is constantly ongoing along with hydration process during drying. As BA particles are being dissolved gradually, embedded Al is further exposed to the alkaline solution and generates expansion phase, H_2 gas that would not greatly reshape hitherto formed pore structure but counteract the tensile stress caused by moisture loss and thus reduce early shrinkage. As depicted in Fig. 15a and b, G7 series initially experience a volumetric stable period. G8.10 and G8.15 even exhibit a plastic expansion at early ages.

Furthermore, the skeleton strength can confine drying shrinkage due to its resistance effect. It can be noticed in Fig. 18 that the drying shrinkage for unaerated samples is lower than the expectation based on capillary stress. G8.00 remains volumetrically stable in the initial 5 days, as evidenced by the meniscus radius of 64.65 nm at shrinkage onset, indicating that entire withdrawal of water from large capillaries causes mass loss without deformation. This can be mainly attributed to the promotion effect of C_3A present in LS, enhancing early strength of the matrix [8] and significantly resisting contraction. On the contrary, G10 deforms distinctly at early ages due to its rather low hydration rate that the skeleton strength is not sufficient to resist moisture-induced deformation. As geopolymerization processes, N-A-S-H gel is being formed accumulatively, developing adequate strength to withstand tensile stress, so that delayed hardening leads to a similar shrinking behaviour as aerated groups exhibit.

Based on the above analysis, the volumetric change responding to increasing moisture loss in different groups can be clearly interpreted. As displayed in Fig. 19, G10 is the most sensitive group to moisture reduction at the beginning of drying process due to its lack of early strength, after which its deformation stabilizes. Simultaneously, the formation and growth of N-A-S-H gel produces a small increase in the

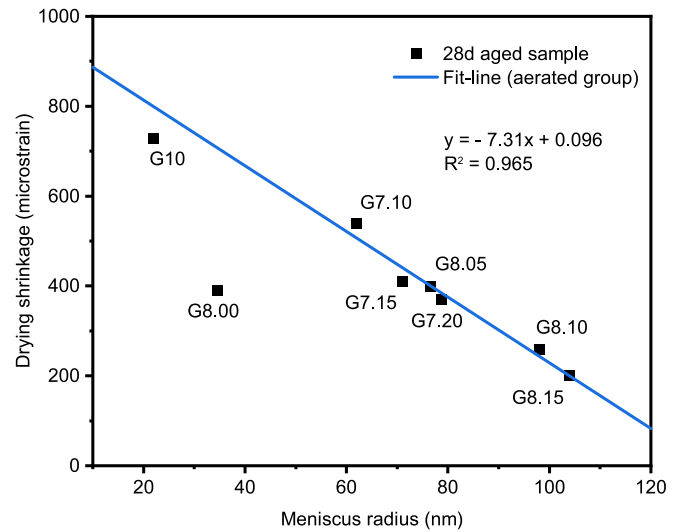


Fig. 18. The impact of meniscus radius on 28 d drying shrinkage.

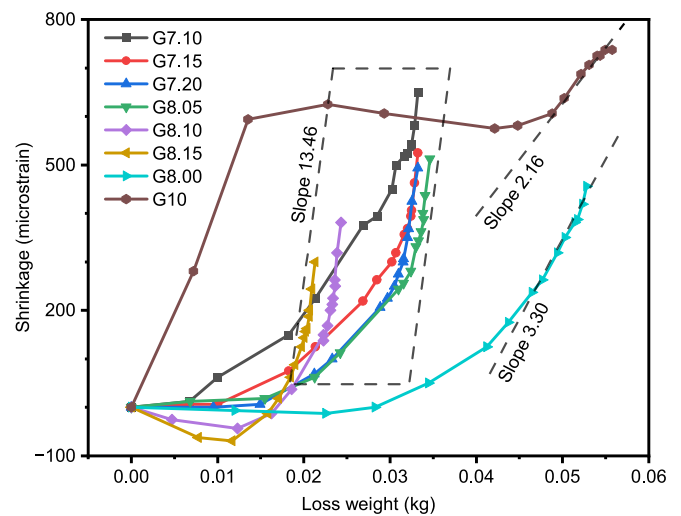


Fig. 19. Relation between drying shrinkage and mass loss.

apparent volume of the skeleton [61]. Foamed groups show the certain volumetric stability before the removal of water in large capillaries, referring to the counteraction of expansive phase. Nonetheless, the high initial strength makes G8.00 the most inert group to early shrinkage that almost entire withdrawal of water from large capillaries causes mass loss without shrinkage. As hydration slows down, moisture loss has a more direct impact on shrinkage, steepening the slope. After inflexion point, the relation between drying shrinkage and mass loss tends to be linear

Table 7

Summary of results from shrinkage measurement and computed meniscus radii at certain ages.

Sample	28 d μ e	28 d mass loss ratio (wt%)	WL at 28 d (vol %)	r_m at 28 d (nm)	WL at inflexion point (vol%)	r_m at inflexion point (nm)	WL at shrinkage onset (vol%)	r_m at shrinkage onset (nm)
G7.10	540	9.730	17.263	61.91	11.817	110.88	2.527	639.96
G7.15	414	9.303	14.382	70.96	12.004	91.35	3.922	313.41
G7.20	378	9.755	14.057	78.70	11.935	96.76	5.844	212.77
G8.05	396	9.550	14.937	76.43	12.679	93.18	6.055	194.23
G8.10	270	8.681	9.989	97.97	8.707	113.01	6.347	155.17
G8.15	198	8.497	8.579	103.93	7.718	115.25	6.187	142.13
G8.00	396	11.446	22.742	34.50	18.169	42.64	11.043	64.65
G10	720	12.126	26.613	21.96	19.089	30.06	–	–

where r_m refers to meniscus radius and WL is water loss ratio.

for all specimens. It implies that the completely hardened geopolymer undergoes elastic deformation in response to the tensile stresses distributed on capillary walls. G8.00 exhibits a steeper curve with a slope of 3.3, compared to the 2.16 slope of G10. According to the water loss ratio at inflexion point in Table 7 and capillary porosity listed in Table 4, the particular fraction of pores involved in drying process could be estimated. In contrast to G10 ($30.65-19.089 = 11.561\%$), a higher number of small capillaries ($31.65-18.169 = 13.481\%$) participate in drying shrinkage of G8.00, intensifying deformation under tensile stress. From the same perspective, aerated geopolymers generally present a sharper incline with the 13.46 slope than reference groups since more capillaries are involved into drying process. They also show a close agreement on the final tendency, implying the similarity in apparent elasticity of the solid-pore structure.

Overall, the reduction of drying shrinkage for investigated geopolymers can be interpreted from three perspectives. Drying shrinkage is predominantly driven by the moisture loss for porous geopolymer. Highly tortuous capillary network and small specific surface area prevent rapid moisture loss and encourage forming menisci with large radius during drying, resulting in a relatively slight tensile force on solid phase. Moreover, early-age shrinkage can be relieved by the swelling effect of expansive admixture. Generated expansive phase, such as H_2 , would counteract contraction stress and delay volume change onset. In addition, drying shrinkage development is synthetically impacted by skeleton strength and involved capillary porosity. Fast hardening can significantly delay shrinkage onset and slow down shrinkage growth. And solid-pore structure tends to deform easily as moisture evaporates when more capillaries participate in interior drying process.

3.6. Thermal conductivity

As shown in Table 8, the thermal conductivity of geopolymer samples ranges from 0.135 to 0.263 W/m·K, indicating notable variation in heat transfer performance across the investigated formulations. The highest value observed among foamed samples is G7.10 (0.224 W/m·K) since its poor foamability results in a denser microstructure. In contrast, G8.15 exhibits the lowest conductivity 0.135 W/m·K, about half that of reference group G8.00, due to the high porosity, making it more suitable for thermal insulation applications.

Pearson correlation coefficient is applied to understand the relationship between the measured pore structure characteristics and thermal conductive properties, which ranges from -1 (perfect positive relationship) to 1 (perfect negative relationship). It is clear in Fig. 20 that the λ and $C_{vol.}$ are closely related to macropores larger than $100\ \mu\text{m}$ (Region II). The overall porosity, obviously, negatively correlates to thermal conductivity where Region II macropores contribute mostly. As discussed in Section 3.4, Region I macropores are generally influenced by slurry setting process and foaming intensity, leading them to stabilize or coalesce into large ones, which implies an opposing trend between Region I and Region II porosities. Therefore, Region I macroporosity

Table 8
Summary of thermal conductive properties.

Sample	λ (W/m·K)	$C_{vol.}$ (MJ/m ³ ·K)	α (mm ² /s)	$C_{wt.}$ (J/g·K)	Mean temperature during test (°C)
G7.10	0.224	1.190	0.188	0.979	28.16
G7.15	0.203	1.290	0.158	1.100	28.16
G7.20	0.182	1.070	0.170	1.001	28.44
G8.05	0.199	1.210	0.165	1.050	28.12
G8.10	0.148	0.881	0.168	0.898	28.28
G8.15	0.135	0.816	0.166	0.975	28.39
G8.00	0.263	1.500	0.176	0.998	28.40
G10	0.234	1.540	0.152	1.029	28.33

where λ refers to conductivity, $C_{vol.}$ represents volumetric heat capacity, α is thermal diffusion and $C_{wt.}$ is specific heat capacity.

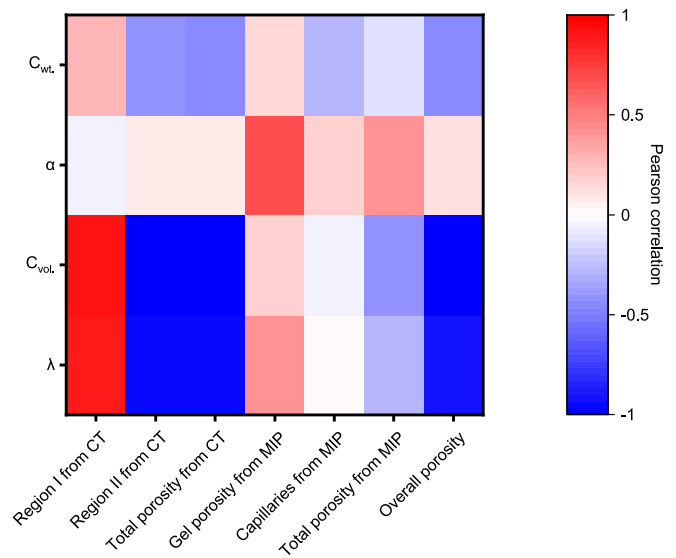


Fig. 20. Correlation coefficient between specific porosities and thermal conductive properties.

shows an apparent positive correlation with thermal conductivity.

$C_{wt.}$ and α depend on solid geopolymer phase, e.g., gel content and molecular structure, rather than the pore structure, although there is a mediumly positive correlation appearing between thermal diffusion and gel porosity. In fact, a higher gel pore content indicates more gel bridges, forming continuous thermal pathways that enhance heat transfer throughout the solid phase, thereby improving thermal diffusion. This relationship is clearly demonstrated by unaerated groups G8.00 and G10. The addition of LS promotes the formation of gel bridges, which improves the thermal diffusion from 0.152 to 0.176 mm²/s.

3.7. Mechanical performance

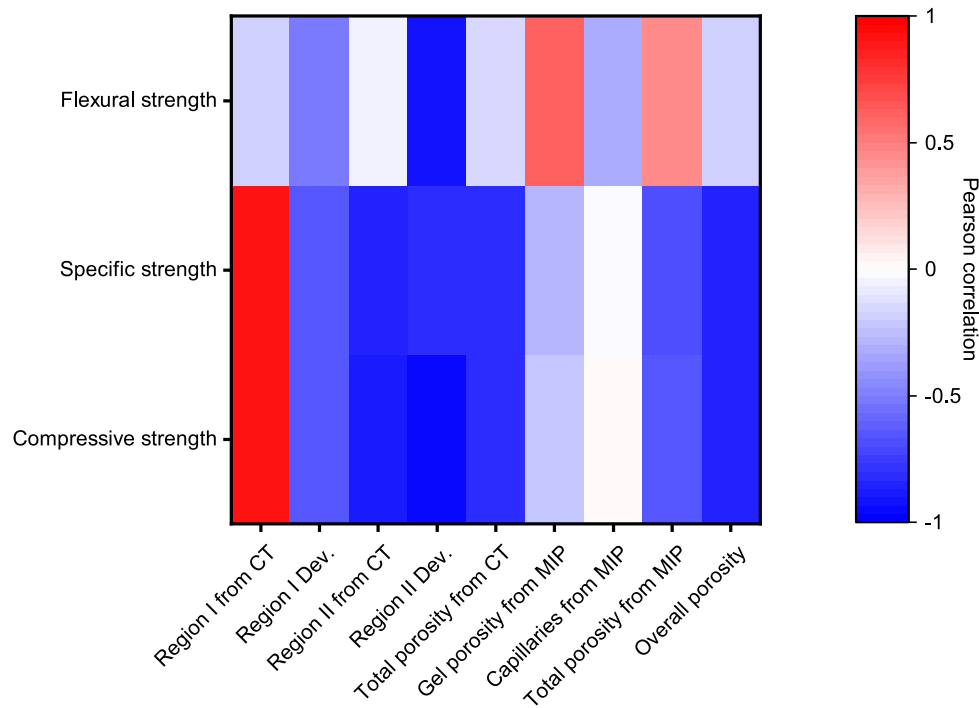
As regards aerated materials, the relatively low mechanical strength remains a persistent challenge. Variations in foaming method and binder system can lead to distinct mechanical behaviours. In this section, the factors influencing mechanical strength are analysed from the perspective of pore structure, thereby providing clear guidance for future optimization.

As shown in Table 9, foamed samples present an obvious reduction in compressive strength at room temperature, compared with unaerated groups. The 28 d compressive strength of G8 series, which exhibit the relatively high foamability, is lower than 3 MPa and their specific strength (compressive strength/bulk density ratio) is approximately 2.4 Pa·m³/g, corresponding to 25% that of unaerated groups. This significant reduction is attributed not only to the increased porosity but also to the high pore size deviation [28]. To further elucidate the relationship between pore structure characteristics and mechanical performance, Pearson correlation analysis is applied, and the results are demonstrated in Fig. 21. It can be found that compressive and specific strengths are inversely related to overall porosity at room temperature, particularly with respect to Region II macropores. On the one hand, a higher macroporosity would lead to thinner interstitial films between foams, which act as stress concentration zones under compressive loading. This causes premature failure of the sample before the entire solid phase can be effectively engaged, thereby lowering the specific strength of the matrix and reducing its stress tolerance. On the other hand, the compressive strength is also influenced by thickness consistency and spatial distribution of these films, which rely on the foam size uniformity. Among the aerated groups, G7.15 demonstrates the highest specific strength since its relatively low Region II Dev., as revealed by μCT analysis, which indicates a more uniform spatial arrangement of solid films. As a result,

Table 9

Mechanical strength at room temperature/after high temperature exposure.

Sample	28 d compressive strength				28 d flexural strength (MPa)	
	Room temperature (MPa)	Specific strength (Pa·m ³ /g)	1000 °C treated (MPa)	Specific strength (Pa·m ³ /g)	Room temperature	1000 °C treated
G7.10	4.02	3.20	12.67	10.09	1.38	0.37
G7.15	3.91	3.37	10.97	9.45	1.17	0.41
G7.20	2.96	2.76	8.71	8.11	1.03	0.39
G8.05	2.96	2.53	7.55	6.45	1.03	0.23
G8.10	2.23	2.32	7.12	7.40	0.82	0.39
G8.15	2.12	2.55	6.78	8.15	0.76	0.47
G8.00	14.10	9.29	23.45	15.45	1.97	0.62
G10	15.87	10.65	12.30	8.25	3.14	–

**Fig. 21.** Correlation coefficient between specific porosities and mechanical strengths.

stress is more evenly distributed throughout the solid phase and the compressive capacity is further enhanced.

It can also be observed in Fig. 21 that the apparent positive correlation between Region I porosity and compressive capacity. This results from the opposing trend between Region I and Region II macroporosities mentioned in Section 3.6, rather than from Region I pores themselves, as evidenced by the unaerated groups which contain only Region I pores. The mechanical performance of G8.00 is inferior to G10's, despite the positive effect of LS on gel formation. More voids are entrapped in G8.00, introducing additional stress concentration points and making the specimens more prone to fracture.

The flexural strength at room temperature is mainly influenced by macropore size deviation instead of porosity, as depicted in Fig. 21. Similar to the compressive behaviour, foamed samples with a high Region II Dev. would have more localized stress concentration zones which cause weakness under bending and reduce flexural capacity. G8.15 with the highest Region II Dev. shows the lowest 28 d flexural strength of 0.76 MPa. Beyond matrix homogeneity, the tensile strength of solid phase is another factor in bending resistance, which is inherently correlated with gel content and gel porosity.

In summary, the mechanical strength of porous geopolymers is governed by three primary factors. First, macroporosity, particularly pores larger than 100 μm , predominantly determines the strength level of the matrix, as it dictates how much of the solid phase can be

potentially engaged in load resistance. Second, the efficiency with which the solid phase participates in load-bearing strongly depends on macropore size deviation; a lower deviation promotes more uniform stress distribution and thereby improves the overall specific strength. Third, the intrinsic strength of the solid phase is the foundation of mechanical strength, which originates from the binder system and relies on the polymerization process.

3.8. Thermal performance

3.8.1. Morphology

It can be seen from Fig. 22 that all samples show a colour change after exposure to high temperature. Binary and ternary systems turn to a light brown coloration, whereas unary system appears reddish brown, owing to increased iron oxidation [62] with higher FA addition. Additionally, cracks are developed during heating in all samples to different extents. It needs to be noted that the real-time morphology at 1000 °C cannot be accurately determined due to equipment limitations (first cooling needs to be applied). As a result, the width and length of the cracks shown in Fig. 22 may differ from their original dimensions before the natural cooling process. Nevertheless, the measurement should still be valid for comparison purposes. As shown in Fig. 22, G10 exhibits transverse cracks with apparently wide widths on the surface. In contrast, G8.00 shows a lower density of split cracks, with partial crack

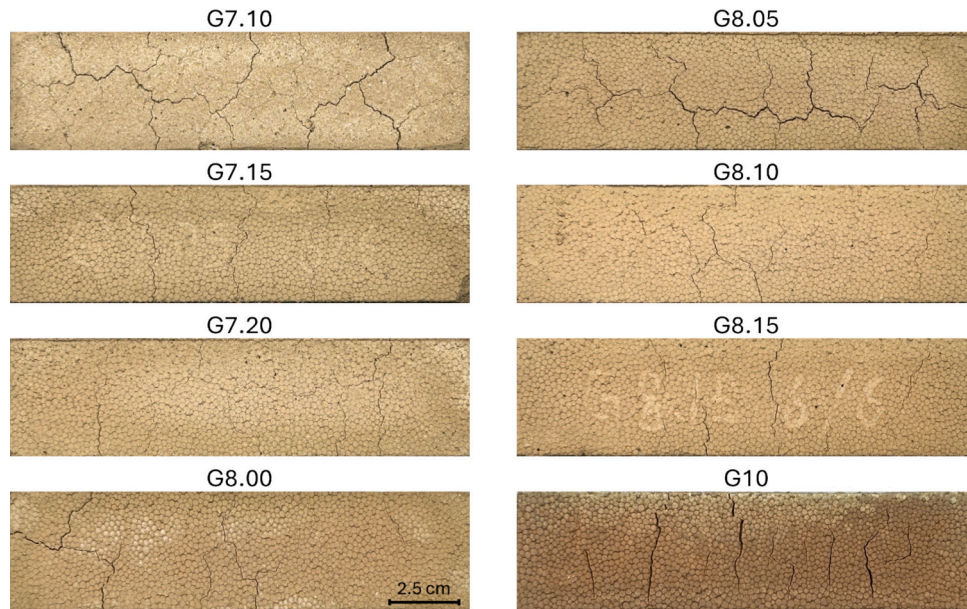


Fig. 22. Morphology of samples after 1000 °C exposure.

connectivity. For ternary system, notably, increasing BA/LS ratio in both G7 and G8 series leads to less thermal cracks. G7.20 and G8.15 present only few partial-depth cracks. But G7.10 and G8.05 demonstrate the poorest integrity among foamed groups that crazing cracks distribute across the surface.

Thermal cracking may originate from the incompatibility within the hybrid gel matrix. The incorporation of LS promotes the formation of Ca-enriched gels, resulting in the co-existence of multiple gel phases, as evidenced in Fig. 9. The mismatch in the thermal behaviour of these different gels increases the tendency of sever cracking at high temperatures [7,8]. Nevertheless, this mechanism is unlikely to be the governing factor in the present study. Based on the cracking behaviour observed across three systems, the variations in thermal cracking are more consistently correlated with differences in the pore structure than with the composition of the hybrid gel matrix.

Thermal cracking can also result from the thermo-hydra process in which the vapour pressure builds up in pores and induces internal tensile stress, as well as the thermo-mechanical process where the stress originates in the gradients of thermal deformation within the material [63,64]. The latter generally occurs in all specimens. Thermal conduction establishes a temperature gradient from exposed area to the core of prisms when subjected to elevated temperatures. Due to thermal gradient, the physicochemical transformation proceeds asynchronously between the surface and core regions, causing differential volumetric change which increases the risk of integrity deterioration [65]. However, thermal-mechanical process is not supported to be the governing mechanism of thermal cracking, since samples with lower thermal conductivity, e.g. G7.20 and G8.15, exhibit better residual integrity.

The thermo-hydra process is related with moisture movement during heating which can be influenced by the pore structure [19,66]. As analysed in Section 3.4, G10 and G8.00 have the similar macropore structure, characterized by pore sizes predominantly below 100 μm and poor interconnectivity, which facilitates vapour accumulation and promotes moisture clogging within the specimens, resulting in vapour pressure buildup and crack formation during heating [63,64]. Moreover, compared with G8.00, the less tortuous capillary network in G10 provides relatively 'efficient' vapour pathways, which accelerate moisture transport via capillaries during early heating stage (prior to irreversible pore structure deforming) [67]. This would help vapour accumulation and aggravate clogging effect, further deteriorating structural integrity, as reflected by its more extensive cracking than

G8.00's. Conversely, the crack density of foamed groups decreases significantly with increasing BA/LS ratio. This improvement is attributed to the enhanced foaming process, which promotes macropore coalescence in the fresh state, resulting in an interconnected macropore structure with the hydraulic tortuosity below 1.81 as listed in Table 6. It enables rapid vapour release and effectively mitigates moisture clogging, thereby largely limiting the development of cracks within the skeleton during heating.

3.8.2. Physical properties

It is notable in Table 10 that foamed samples with higher BA/LS ratio, such as G8.15, G8.10, exhibit reduced mass loss but greater residual shrinkage, ultimately resulting in a slight densification after 1000 °C exposure. On the one hand, the mass loss of samples is mainly contributed by physical and chemical water removal, including the adsorbed free water in pores and the combined water existing in formed hydrates. As depicted thermogravimetry analysis in Appendix Fig. E, for ternary system, dehydration peak of C/N-A-S-H gels occurs at around 100 °C and the C_3AH_6 phase introduced by LS decomposes between 200 and 400 °C. Increasing BA/LS ratio inhibits both gel formation and C_3AH_6 production, thereby reducing total mass loss. On the other hand, increasing BA/LS ratio promotes thermal shrinkage because enhanced macropore connectivity limits thermo-hydra cracks formation, allowing thermal stress to be transmitted through the integrated pore-solid matrix rather than being relieved by crack opening.

Additional mechanisms, such as viscous sintering and recrystallization [7–9], may also contribute to the overall densification of

Table 10
Structural change of samples after high temperature exposure.

Sample	Mass loss (wt%)	Residual shrinkage (microstrain)	Bulk density (g/cm^3)	
			Room temperature	1000 °C treated
G7.10	13.89	40,626	1.26	1.22
G7.15	13.46	44,910	1.16	1.15
G7.20	13.10	49,770	1.07	1.09
G8.05	12.59	38,826	1.17	1.15
G8.10	12.02	45,486	0.96	0.97
G8.15	11.69	48,582	0.83	0.85
G8.00	12.85	43,614	1.52	1.51
G10	11.01	51,246	1.49	1.55

investigated groups. However, their effects can be partially offset by thermal cracking, as evidenced by the slight decrease in bulk density observed in samples exhibiting extensive cracks.

3.8.3. Mechanical properties

Fig. 23 displays the mechanical change compared to room temperature values (Table 9). The strength evolution varies notably among different systems after 1000 °C exposure. It is found that sole FA based geopolymer experiences a drastic reduction in mechanical strength at 1000 °C. Specifically, the compressive strength of G10 is decreased by 23%, and its flexural capacity is completely lost. This can be attributed to the thermal deterioration of structural integrity [2]. Due to the thermo-hydra process, G10 is susceptible to extensive cracking and even spalling (if higher heating rate is applied) [19] when heated to 1000 °C. Similarly, G8.00 undergoes noticeable cracking and loses over 60% flexural strength after 1000 °C exposure.

In contrast, all ternary groups show an exceptional improvement in compressive strength and maintain partial flexural strength to certain degrees after high temperature treatment. G8.15, the most porous group, shows the highest improvement of more than twice its original compressive strength, while its flexural strength decreases by less than 40%. This mainly results from its macropore structure that promotes moisture release during heating and largely preserves structural integrity at 1000 °C. On top of that, further hydration of unreacted precursor particles and recrystallization of hydrates during heating also contribute to its strength enhancement [7]. Therefore, G8.15 exhibits an approximately 200% increase in compressive strength, and its specific strength is improved from 2.55 to 8.15 Pa·m³/g after exposure to 1000 °C as shown in Table 9. However, not all aerated samples demonstrate good structural stability. G8.05 shows the lowest specific strength of 6.45 Pa·m³/g and experiences the largest flexural strength loss, approaching 80%, after thermal treatment. Owing to the limited foaming, the macropore structure of G8.05 consists of relatively large pores but limited interconnected with an absolute permeability of only 0.09 μm² as shown in Table 6 that provides insufficient vapour-release pathways. Besides, the semi-connected large pore bodies, located along moisture movement paths, temporarily accommodate transported vapours [68,69], thus enabling vapour pressure buildup within pores at elevated temperatures. Consequently, cracks tend to form around those macropores and propagate through the skeleton as observed microstructure evolution in Fig. 24, ultimately resulting in weakened structural bonding. In this case, the insufficient foaming instead exacerbates high temperature deterioration. It is noteworthy that the thermostability of integrity is critical in determining residual mechanical performance. Although G8.05 undergoes a similar phase transformation after 1000 °C exposure,

compared with G8.15, as depicted in Appendix Fig. F, the loss of structural integrity significantly limits the strength gain after 1000 °C exposure.

4. Discussion

4.1. Afore-heating moisture movement

Under curing conditions, the moisture content within the pore network decreases with exposure time due to the humidity difference between sample matrix and surrounding atmosphere [51]. It is found that the moisture movement is strongly constrained by the pore structure. In order to further clarify the role of pore structure during drying process, the mass change rate of G8.15, the representative of ternary system, is compared with two other binder systems', as shown in Fig. 25. In addition, the conceptual pore structure model of ternary system is proposed in Fig. 26 to aid in understanding moisture movement.

During the initial drying stage, moisture loss is primarily attributed to the evaporation of water adsorbed on sample exposure surface, including the wall of open macropores, as illustrated in Fig. 26a. As shown in Fig. 25, G8.15 demonstrates the highest initial mass change rate of over 1.5 wt% per day and maintains this rate until day 4. Owing to its highly interconnected macropore structure, internal moisture adsorbed on macropore walls releases rapidly, resulting in a consistent high mass loss rate. In contrast, the unaerated groups display a gradual increase in moisture loss rate, reaching a maximum before day 5. This delayed response is due to the higher fraction of isolated macropores (Region I pores) which indirectly exposed to the environment, leading sample's external surface evaporation as the main contributor of early water removal. It can also explain the comparable water loss ratio among G8.15, G8.00 and G10 during deformation stage, despite G8.00 and G10 having about twice specific surface area of G8.15, as displayed in Fig. 17.

As drying progresses, evaporation in capillary pores becomes significant. Simultaneously, water in isolated macropores is gradually transported through capillary network to surrounding environment, as depicted in Fig. 26b. The moisture loss rate during this stage is mainly governed by capillary tortuosity. A highly tortuous capillary structure, such as G8.15, increases the effective movement path for vapour diffusion, thereby slowing moisture escape under a given humidity gradient. Conversely, G10 and G8.00, which exhibit lower capillary tortuosity, perform higher water release rates after day 5 in Fig. 25.

These differences in moisture movement not only determine drying shrinkage behaviour but also provide insight into high temperature behaviour of different systems.

4.2. The relationship between pore structure and thermal deterioration

Based on the cracking behaviour and mechanical performance of different groups upon 1000 °C exposure, the extent of thermal deterioration follows the order ternary system < binary system < unary system. Thermo-hydra process induced cracking is identified as a major factor governing structural integrity of heated samples which fundamentally determines residual mechanical strength.

The macropore structure plays a critical role in thermal cracking resistance of foamed geopolymer. Analogous to moisture removal at early curing stage, an interconnected macropore structure enables rapid vapour ventilation during heating, thus preventing moisture clogging and effectively inhibiting thermal cracks. As a result, the retained structural integrity allows the beneficial effects of densification and phase evolution to be translated into a significant gain in compressive strength after 1000 °C exposure. In contrast, porous samples with insufficient macropore interconnectivity show exacerbated high temperature cracking. On the one hand, the semi-connected macropore structure provides ineffective vapour-release paths. On the other hand, the macropores in relatively tortuous pathways temporarily

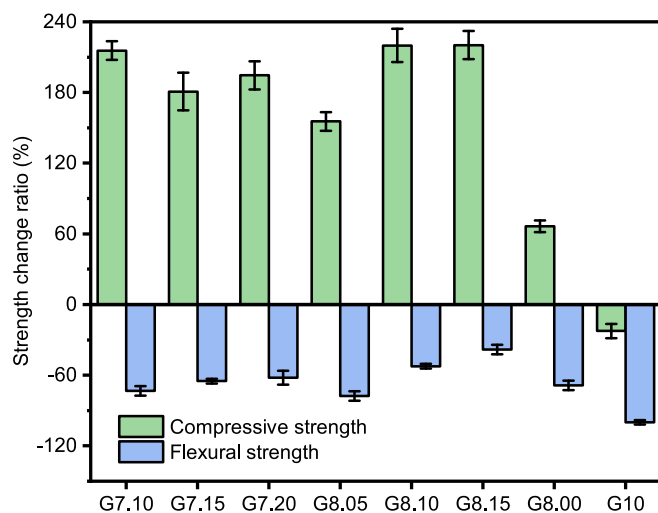


Fig. 23. Mechanical strength change after 1000 °C exposure.

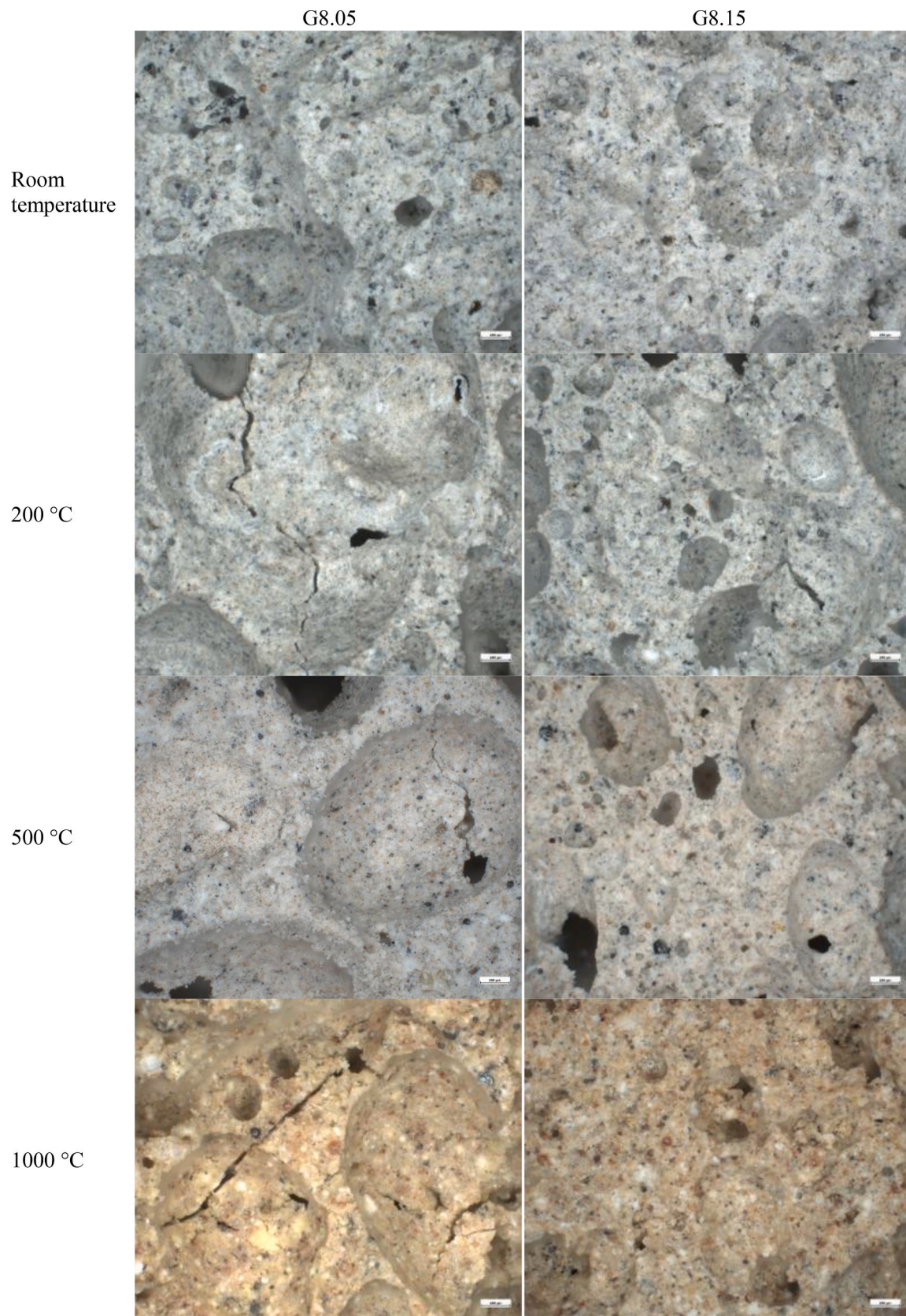


Fig. 24. Microstructure evolution of two distinct macropore structures (G8.05 and G8.15) at elevated temperatures.

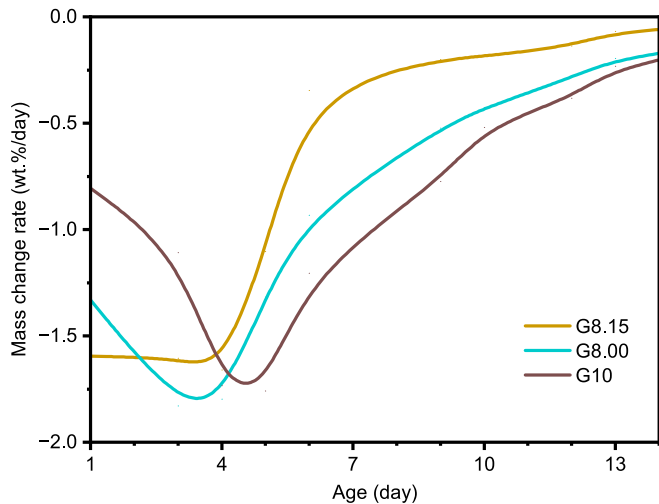


Fig. 25. Mass change rate at certain ages.

accommodate transported vapours [68,69], facilitating vapour pressure buildup within pores at elevated temperatures and ultimately weakening structural integrity.

The effect of capillary tortuosity on moisture regulation under high temperatures is more pronounced in unaerated samples. As observed in Fig. 9, LS introduces Ca-enriched gels into the binder system that forms a hybrid gel matrix. Due to the thermal incompatibility between different gels, the hybrid matrix is prone to microcrack formation which lowers mechanical strength after 800 °C exposure [7]. However, in this study, the binary system demonstrates superior mechanical performance compare with unary system after exposure to 1000 °C. This observation indicates that gel chemistry alone is insufficient to explain the high temperature behaviour [2]. Alternatively, it can be attributed to the pore structure. At the macroscale, unaerated samples exhibit a relatively dense and continuous solid structure, in which moisture clogging easily occurs [64]. At the mesoscale, water existing in pores and hydration products gradually evaporates via capillary network during heating [69]. Compared with G8.00, the capillary network of G10 has lower tortuosity which enables more direct vapour transport within the matrix, as illustrated in Fig. 26b. This would accelerate internal vapour accumulation under high temperatures and aggravates moisture clogging effect, thereby leading to more extensive thermal cracking and more severe mechanical deterioration.

Overall, pore topology is not simply evaluated as beneficial or detrimental in this study. Instead, its role is distinguished according to the pore scale. For foamed geopolymers, the topology of the macropore network governs the retention of structural integrity and consequently the residual mechanical performance after 1000 °C exposure. In contrast, for unaerated geopolymers, the tortuosity of the capillary

network primarily controls moisture transport and therefore determines the extent of thermal damage. Based on these observations, the unified moisture transport mechanism, as demonstrated in Fig. 26, is proposed to explain the thermal response of both porous and dense geopolymer systems.

4.3. Limitation and further works

This study mainly focuses on the final structural and mechanical responses of the porous geopolymer after high temperature exposure. Although the residual properties and microstructural characterization provide insights into thermal deterioration mechanisms, the dynamic evolution of moisture migration, thermal strain and crack initiation during heating was not directly captured. Future studies incorporating in-situ monitoring techniques, such as pore pressure evolution [64] and real-time acoustic emission [63], are suggested to further elucidate the transient deterioration process.

Compared with peer-reported results, the overall trend observed between 28 days compressive strength and bulk density in this work is consistent with them, as illustrated in Fig. 27. However, within the present dataset, only G7.10 and G7.15 exhibit comparable performance in terms of strength-density balance. This divergence can be attributed to two primary factors, the applied foaming approach and the reactivity of the used precursors. Firstly, the foaming methodology plays a pivotal

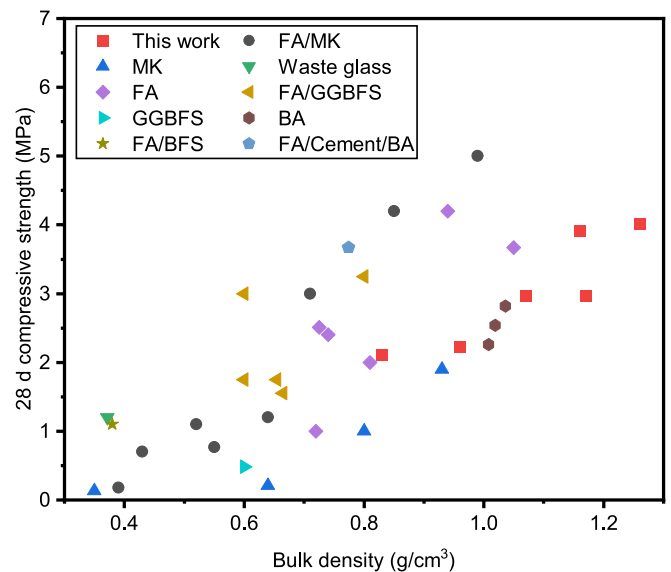


Fig. 27. Comparison of the relationship between compressive strength and bulk density of foamed geopolymers at 28 days with relevant studies [10,12,21–26,29,33,37,70,71].

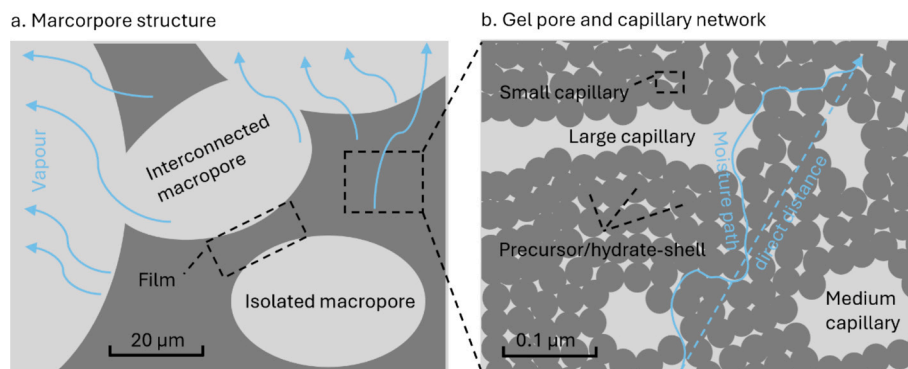


Fig. 26. Schematic concept of pore structure and water movement.

role in defining the homogeneity of the pore structure, which directly affects mechanical strength. In this study, LS is incorporated to regulate MSWI BA-induced aeration by promoting matrix stiffening, thereby stabilizing fresh foams. Nevertheless, further foaming optimization remains possible to achieve a more uniform pore size distribution and enhanced mechanical performance. For instance, Klima et al. [25] demonstrated that the introduction of various non-ionic and ionic surfactants can modify the interfacial properties between foam and paste at fresh stage in alkaline activated system, leading to a more homogenous pore structure. Secondly, the intrinsic reactivity of precursor materials determines the strength of solid skeleton which is the foundation of mechanical performance for porous geopolymers. High-performance foamed geopolymers documented in previous studies (see Fig. 27) often incorporate highly reactive materials such as MK [10], GGBFS [21,29] or OPC [37] as partial replacements. Compared with ternary system in this work, those precursors accelerate gel formation more effectively, promoting a denser microstructure and stronger interfacial bonding within the matrix which contribute supplementary mechanical strength. Future studies are therefore suggested to explore the combined use of more reactive precursors to further improve the mechanical properties.

5. Conclusion

This study systematically investigates the porosity and pore topology of FA/BA/LS based geopolymers and elucidates their influence on drying shrinkage, thermal conductivity and mechanical performance under ambient conditions. Beyond the conventional emphasis on gel chemistry and phase evolution, pore structure is found to play a decisive role in governing the thermal response of porous geopolymers following exposure to 1000 °C. The developed ternary geopolymers generally exhibit enhanced volumetric stability, thermal insulation and resistance to thermal deterioration, indicating its potential for lightweight refractory materials. The main conclusions are as follows:

- The porous geopolymer is developed using MSWI BA as an inherent foaming agent with the incorporation of LS. The metallic aluminium present in BA provides significant aeration potential, while LS accelerates early hydration, enabling controlled porosity development and effectively preventing foaming-induced cracking. The foamability is tailored by the relative addition of BA and LS. Increasing BA/LS ratio enhances foam generation and pore interconnection, achieving 44.18% macroporosity without expansion cracks.
- The drying shrinkage of geopolymers can be mitigated by dedicated capillary network tailoring. During drying, moisture movement is strongly governed by capillary complexity where a highly tortuous capillary network, as exemplified by G8.15, effectively retards water removal from the matrix after 5 days. This restrained moisture loss suppresses the formation of small capillary menisci, thereby limiting capillary stress buildup and resulting in final drying shrinkages as low as 306 microstrain.

- Macroporosity, particularly pores larger than 100 µm, predominantly determines thermal conductivity and compressive strength of foamed geopolymers at ambient temperature. Moreover, foam uniformity is the key to enhance both compressive and flexural strengths as a homogeneous pore structure enables the load being evenly distributed throughout the solid phase, thereby improving overall load-bearing efficiency.
- The topology of foams is recognized as the key factor governing thermal deterioration at elevated temperatures, rather than gel chemistry or phase evolution alone. An interconnected macropore structure with high permeability facilitates the release of evaporated water during heating and alleviates moisture clogging within the matrix. Consequently, the propagation of thermal cracks is significantly inhibited, preserving overall structural integrity of geopolymer samples and enabling an over 200% gain in compressive strength after exposure to 1000 °C for the optimal group.
- The proposed moisture transport mechanism provides an additional perspective for understanding the thermal deterioration of unaerated geopolymers. A dense matrix can become more susceptible to thermal damage during heating due to insufficient open porosity promoting vapour release. The vulnerability would be further intensified when its capillary network provides direct moisture pathways that intensify internal vapour accumulation within the matrix and ultimately lead to extensive thermal cracking and severe mechanical deterioration, regardless of the thermal compatibility of gel system.

CRediT authorship contribution statement

Yanshuo Liu: Writing – original draft, Investigation, Formal analysis, Data curation, Conceptualization. **Yan Luo:** Writing – original draft, Investigation. **Jiale Yuan:** Investigation. **H.J.H. Brouwers:** Supervision, Funding acquisition. **Qingliang Yu:** Writing – review & editing, Supervision, Project administration, Funding acquisition, Conceptualization.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Appendix A

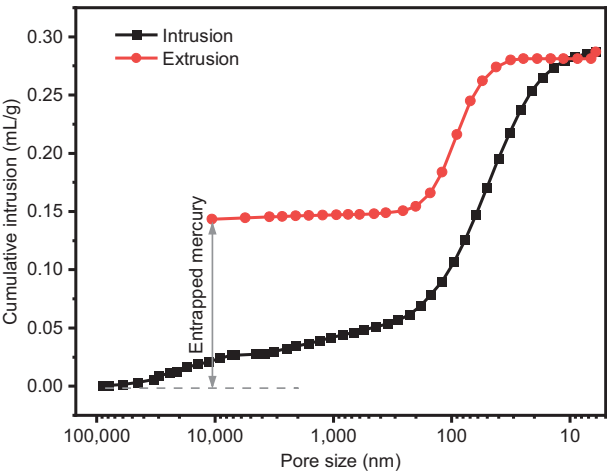


Fig. A. Intrusion/extrusion cycle.

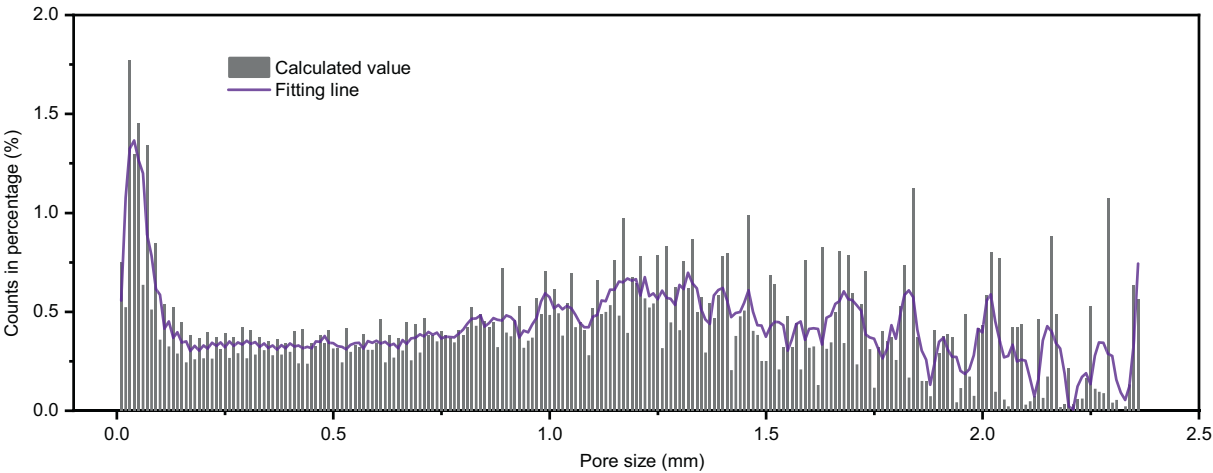


Fig. B. Relative frequency of G8.15 pore size distribution.

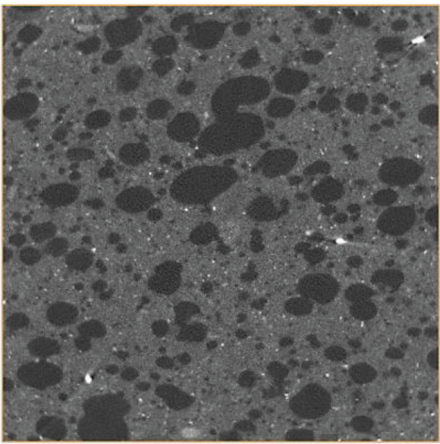


Fig. C. A radiograph of ternary geopolymers with metal artifacts.

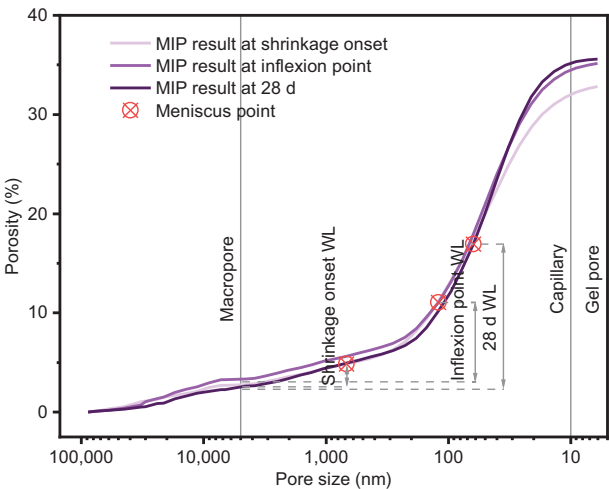


Fig. D. Computation of meniscus radius at certain stages for one group.

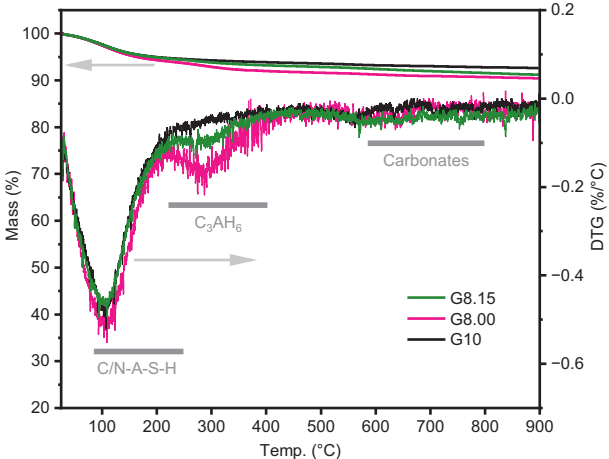


Fig. E. Thermogravimetric analysis of selected binder systems (grind powder specimen).

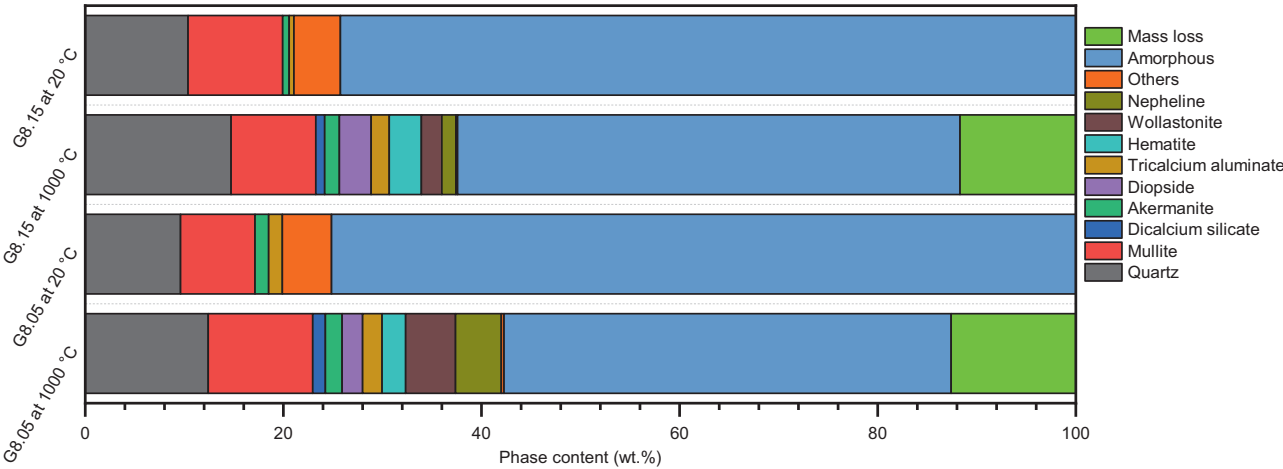


Fig. F. Phase compositions evolution of G8.05 and G8.15 after 1000 °C exposure.

Data availability

Data will be made available on request.

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