

Article

Circular Reuse of Dryer-Filter Synthetic Microfibres in Geopolymeric Coatings for Architectural Applications

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Abstract

Synthetic microfibres released during textile drying are considered an emerging source of microplastic pollution, yet this waste stream is generally discarded without treatment. This study investigates a valorisation route by incorporating waste dryer-filter microfibres into a potassium-based/metakaolin geopolymeric coating for architectural applications, with the dual objective of preventing environmental release and enhancing material performance. Geopolymer pastes containing 0.1–0.3 wt.% of synthetic microfibres were characterised in terms of physical, mechanical and microstructural behaviour. Microfibre addition produced a marked toughening effect, with flexural strength increasing from about 3 MPa for the unreinforced matrix to 7.5 MPa for the composite containing 0.3 wt.% fibres, while compressive strength decreased moderately due to the presence of a compliant fibrous phase. Microstructural observations confirmed fibre dispersion and fibre–matrix bonding, supporting crack-bridging mechanisms. Density, porosity and water absorption measurements indicated a stable geopolymer gel structure with a connected pore network. Thin-layer applications onto clay brick exhibited satisfactory workability and adhesion, confirmed by pull-off testing (~0.12 MPa) and interfacial microscopy. The results demonstrate that textile-derived microfibres can be effectively immobilised within a potassium geopolymer matrix while improving flexural performance, offering a feasible circular strategy for microfibre waste reuse in mineral coatings.

Keywords: synthetic fibres; microplastics; geopolymer composite; architectural coating; ecodesign; sustainable construction; ceramic cladding



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1. Introduction

Microplastics, commonly defined as synthetic polymer particles smaller than 5 mm, are now ubiquitously detected in aquatic, terrestrial and atmospheric compartments and are recognised as persistent pollutants and vectors for sorbed contaminants and additives. Among them, fibrous microplastics originating from synthetic textiles are frequently reported as one of the dominant morphologies in environmental samples, owing to the extensive use of polyester, polyamide and acrylic fibres in clothing, technical textiles and home furnishings [1–3]. At the industrial scale, the textile and laundry sectors have been identified as major contributors of microfibres to wastewater streams: textile finishing, industrial washing and large-scale laundering continuously generate significant loads of synthetic microfibres and fibre lint, which can be released into process waters or accumulate in filtration residues [4–7]. Recent studies on textile industry effluents and industrial laundry wastewater have quantified high concentrations of microfibres, confirming that

these facilities can markedly increase the microplastic load entering wastewater treatment plants and, ultimately, receiving waters or sludge-amended soils [4–8].

In Europe, the regulatory framework is evolving towards stricter control of microplastic emissions. The EU Zero Pollution Action Plan and the Circular Economy Action Plan set a target of reducing microplastics released into the environment by 30% by 2030, explicitly highlighting textiles as a priority sector [9,10]. In parallel, Commission Regulation (EU) 2023/2055 introduces a progressive restriction on intentionally added microplastics in products, while signalling the need to tackle unintentional releases from sectors such as tyres, textiles and plastic pellet handling [11]. However, beyond wastewater treatments, the safe disposal, destruction and recycling of microplastic-rich solid residues remain poorly regulated and insufficiently addressed [12]. This gap is particularly relevant for large textile laundry and finishing plants, where filter cakes and lint-rich residues are generated at substantial volumes and are typically managed as mixed or hazardous waste, with limited options for material recovery and circular use.

Against this background, a growing body of research has investigated strategies for the recycling, immobilisation and valorisation of plastic particles and fibres in construction materials. Numerous studies have incorporated post-consumer plastic waste into concrete and mortars, using plastic as partial replacement of fine and coarse aggregates, or as discrete fibres, with the dual objective of diverting plastic from landfill and reducing natural aggregate consumption [13–18]. Reviews indicate that plastic waste from packaging and other sources has been employed in a wide range of cement-based products, including structural concretes, masonry units, mortars, lightweight panels and insulating elements [13,14,18]. In many cases, the addition of plastic induces reduced density and thermal conductivity, accompanied by moderate reductions in compressive strength at higher contents, and it can improve toughness, impact resistance and control crack propagation when appropriately dosed [15–18].

Within this broader field, increasing attention is being paid specifically to microplastics and microfibrils. Recent works have examined microplastic-sized particles as lightweight aggregates in ultra-high-performance concrete or as fillers in mortars, reporting significant reductions in density and modified fracture behaviour, while emphasising the need to control dosage, particle size and compatibility with the cementitious matrix [19,20]. A comprehensive review on microplastics in the built environment highlights that microplastic-containing materials already exist in construction (e.g., synthetic fibre-reinforced concretes, polymer-modified mortars, paints and road surfaces), but the amount of dedicated research on deliberately incorporating secondary microplastics as a valorised component remains limited and fragmented [21]. In parallel, studies on recycled plastic fibres from packaging and end-of-life plastic products demonstrate that such fibres can be safely used in mortars and concretes, enhancing tensile and post-cracking behaviour while providing a route for waste utilisation [22–24]. These approaches show that cementitious materials can act as long-term encapsulant for plastic fractions, potentially reducing dispersion pathways if the matrices are sufficiently dense and durable. However, only a few contributions focus on microfibre-rich waste streams directly linked to textile processing, such as raising, shearing or industrial laundry residues, where fibre size distribution, composition and contamination are more complex [6,7,20].

Geopolymers and alkali-activated binders have recently emerged as particularly promising matrices for the inertisation and valorisation of such microplastic wastes. Geopolymers are inorganic aluminosilicate binders formed by alkaline activation of reactive precursors (e.g., metakaolin, fly ash, slag), producing a three-dimensional network of Si–O–Al bonds with ceramic-like properties [25,26]. Compared to ordinary Portland cement, geopolymer systems can offer substantially lower CO₂ emissions, especially when

based on industrial by-products, and exhibit high compressive strength, chemical durability and exceptional resistance to high temperatures and fire [25–28]. Their microstructure, spanning from dense to moderately porous matrices, can be tailored to design coatings and renders that combine mechanical performance with vapour permeability and breathability, key requirements for sustainable envelopes, bio-based construction materials and heritage substrates [27–29]. Geopolymer-based coatings and mortars have been proposed for fire protection, thermal and passive cooling layers, mineral finishing, corrosion protection of steel and surface consolidation, showing in many cases superior fire resistance and durability compared to organic or purely cementitious coatings [29–32].

The addition of fibres into geopolymers is a well-established strategy to mitigate their intrinsic brittleness and low tensile strain capacity. Fibre-reinforced geopolymer composites (FRGCs) have been extensively studied with steel, glass, basalt, carbon, polymeric and natural fibres, leading to improvements in flexural strength, toughness, crack-bridging capacity and impact resistance [33–38]. Polymeric fibres such as polypropylene or polyvinyl alcohol are widely used at the millimetre scale; they can effectively control shrinkage-induced cracking and enhance post-cracking behaviour, though they may slightly decrease compressive strength at high dosages [36–40]. PP and PVA fibres were found to reduce drying shrinkage cracks up to 50–70% for additions at 0.5–2 vol.% thanks to the crack-bridging effect, improving post-crack ductility [36]. They also enhance toughness and flexural strength, but high contents (above 1–2 vol.%) may decrease compressive strength due to fibre agglomeration and void formation [36,40]. Zhang et al. [39] found that PVA fibres were optimally effective at about 2 vol.% and 9 mm in length, increasing geopolymer toughness and flexural strength up to 150%, while compressive strength decreased by more than 6% at the same dosage.

More recently, few works have explored the integration of microplastics into geopolymer matrices, either as aggregates or dispersed particles, to tailor density and thermal properties and to immobilise plastic fractions in a stable ceramic network [41,42]. Madirisha et al. [41] specifically discussed geopolymers as sustainable materials for microplastic remediation, both as filtration media and as matrices capable of trapping and encapsulating microplastics, highlighting their potential as a long-term immobilisation strategy. Previous studies have shown that the incorporation of microplastic particles into geopolymer pastes may lead to reductions in compressive and flexural strength, typically of the order of about 2–24%, depending on dosage and curing conditions [42]. In contrast, fibrous polymeric inclusions can promote crack-bridging and pull-out mechanisms that improve post-cracking behaviour and, in some cases, enhance flexural performance of geopolymer matrices [34,36], highlighting the importance of fibre morphology and aspect ratio. However, most of these studies deal with particulate microplastics rather than fibrous inclusions. The use of synthetic micro- and nanofibres has been explored in other types of construction materials, such as inorganic binders, cementitious composites and polymer concretes [43,44], where they contribute to improved toughness, ductility, flexural strength and tensile strength through mechanisms like crack-bridging. Yet, their incorporation into geopolymer matrices remains largely uninvestigated, with studies focusing mainly on natural or inorganic microfibrils [45–47] rather than waste-based synthetic micro- and nanofibres. In particular, the literature dealing with textile-derived microfibrils or lint-like wastes within geopolymers is still very scarce, and systematic studies that address both the mechanical role of such microfibrils and their effective inertisation within the matrix are largely lacking.

Unlike the engineered polymeric fibres discussed above, which are manufactured with controlled geometry, composition and surface properties to achieve predictable reinforcement, textile-derived microfibrils originate from post-consumer textile wear and

laundering processes and are intrinsically heterogeneous in polymer-type, length and surface condition. In addition, their surface chemistry is often altered by repeated washing cycles and the presence of residual detergents, finishing agents and environmental contaminants, which may affect wettability and interfacial interaction with alkaline geopolymer pastes. Their incorporation into geopolymer matrices is therefore primarily motivated by stabilisation and immobilisation of a mixed waste stream, with reinforcement acting as a secondary and less predictable effect. In this context, fibre–matrix interaction is expected to be governed mainly by physical interlocking and mechanical anchoring within the evolving gel network rather than by tailored chemical adhesion, although local variations in surface roughness and composition may still influence interfacial behaviour at the micro scale.

Against this background, the present study investigates the circular reuse of dryer-filter synthetic microfibres as an additive in geopolymeric coatings intended for architectural applications, with the dual objective of microplastic mitigation and material performance enhancement. Specifically, the work aims to develop metakaolin-based geopolymeric coating formulations incorporating different microfibre contents collected from household dryer filters, and evaluate their fresh, mechanical and durability-related properties relevant to building envelopes. The geopolymer matrix used in this study was developed as a potassium polysialate, selected to minimise efflorescence phenomena compared to sodium-based systems and to ensure improved chemical and dimensional stability in thin-layer applications. By demonstrating the feasibility and performance implications of this approach, the study aims to establish dryer-filter synthetic microfibres as a novel secondary raw material for geopolymeric coatings, contributing to circular-economy strategies in both the textile and construction sectors.

2. Materials and Methods

A potassium-based geopolymer slurry (FA-0) was produced following a standard preparation route adopted in previous studies [48,49], using a planetary centrifugal mixer (ARE-500, Thinky Corporation, Tokyo, Japan). The formulation was designed to obtain a consolidated geopolymer with nominal molar ratios of $\text{SiO}_2/\text{Al}_2\text{O}_3 = 4.0$ and $\text{K}_2\text{O}/\text{Al}_2\text{O}_3 = 0.8$. Metakaolin (Argical™ M1200S, Imerys, Paris, France, median particle size $d_{50} \approx 1.5 \mu\text{m}$) was combined with an alkaline activating solution based on potassium disilicate, obtaining a slurry containing respectively the 39.7 wt.% and 60.3 wt.% of the two components. The activator was formulated as reported in [48,49] by dissolving potassium hydroxide pellets (purity > 85%, Sigma-Aldrich, Merck KGaA, Darmstadt, Germany) and high-purity fumed silica (99.8%, Sigma-Aldrich, Merck KGaA, Darmstadt, Germany) in deionised water to obtain a solution with final molar ratios $\text{H}_2\text{O}/\text{K}_2\text{O} = 13.5$ and $\text{SiO}_2/\text{K}_2\text{O} = 2.0$. The geopolymer slurry without fibres (FA-0) was homogenised in the planetary centrifugal mixer for 3 min at 1000 rpm to ensure complete wetting and dispersion of the precursor particles prior to fibre incorporation.

Mixed synthetic microfibres were collected from dryer filters from household laundry activities (Figure 1). The lint originated from heterogeneous laundered textiles and was assumed to consist of a blend of common apparel-related polymers in variable proportions, mainly including polyester, polyamide, acrylic fibres and polypropylene. Based on a preliminary visual assessment, the recovered microfibres showed an estimated average length of approximately 4–5 mm, with a broad distribution ranging from about 1 to 12 mm. The collected fibres were manually cleaned to remove impurities and subsequently characterised by electron microscopy (E-SEM FEI Quanta 200, FEI, Hillsboro, OR, USA) to assess their morphology, focusing on average diameter and surface texture. In addition, Fourier transform infrared spectroscopy analyses were performed to qualitatively investigate the average chemical composition of the microfibre mixture, evaluated through attenuated total

reflection (ATR) measurements using a Thermo Scientific Nicolet iS5 FTIR Spectrometer equipped with an iD7 ATR accessory and diamond crystal, by ThermoFisher, Waltham, MA, USA. Each spectrum was accumulated from 32 individual measurements, within the range 4000–400 cm^{-1} .



Figure 1. Synthetic mixed microfibres collected from dryer filters.

The synthetic microfibres were added to the plain geopolymers slurry at three different weight percentages of 0.1%, 0.2%, and 0.3%. The microfibres were introduced gradually during mixing to minimise agglomeration and ensure uniform dispersion. After fibre addition, each mixture was further mixed for one minute at 800 rpm to deagglomerate fibre clusters and achieve a homogeneous distribution within the geopolymer matrix. The three formulations (labelled as FA-1, FA-2, and FA-3 and containing 0.1, 0.2 and 0.3 wt.% of microfibres respectively), alongside the plain unreinforced material FA-0, were visually inspected for homogeneity before casting in prismatic silicone moulds of $10 \times 10 \times 60$ mm, or cylindrical moulds 16 mm in diameter and 32 mm in height, which were left unsealed and cured for 24 h in laboratory conditions, at ambient temperature and humidity ($T: 20 \pm 2$ °C, RH: 40–60%), followed by further 24 h in the same conditions after demolding to promote the complete removal of residual moisture. The samples obtained were then used for subsequent characterisation and coating application.

After 28 days, consolidated samples were tested for density, open porosity and water absorption using a hydrostatic balance. Total open porosity and bulk density were additionally quantified by mercury intrusion porosimetry (Thermo Finnigan Pascal 140 and Pascal 240, Thermo Fisher Scientific, Waltham, MA, USA), along with pore size distribution in the range of 0.0058–100 μm . Mechanical properties were evaluated using a universal testing machine (Z050, Zwick-Roell, Ulm, Germany) operated in displacement-controlled mode, with a crosshead speed of 1 mm min^{-1} . Prismatic specimens with nominal dimensions of approximately $9 \times 9 \times 60 \text{ mm}^3$ were tested in four-point bending after surface grinding to remove asperities and ensure uniform loading conditions. Compressive strength tests were performed on the cylindrical specimens. For each formulation, flexural and compressive strength values were calculated as the average of at least three specimens.

External and fracture surfaces were examined using environmental scanning electron microscopy (E-SEM FEI Quanta 200, FEI, Hillsboro, OR, USA) to investigate the material finishing, fibre–matrix interfacial bonding, crack-bridging mechanisms, potential fibre pull-out and the dispersion homogeneity of the incorporated microfibres.

In addition to the consolidated specimens produced for physical–mechanical characterisation, further samples were prepared to evaluate the adhesion behaviour of the

developed materials on different architectural substrates, namely solid clay brick, glazed porcelain stoneware, and laminated wood. These tests aimed to assess the applicability of the geopolymer-based formulations as thin finishing layers (1.0–2.0 mm) on vertical surfaces or architectural components. Since the formulation containing 0.3 wt.% of waste microfibres (FA-3) exhibited the highest flexural strength, only this composite was selected for adhesion trials on all substrates. In addition, the fibre-free benchmark geopolymer (FA-0) was also applied on solid clay brick as a control coating to assess the effect of microfibre addition on adhesion performance. Prior to coating, the substrates were cleaned, lightly abraded and, where appropriate (brick and wood), moistened to prevent excessive absorption of the aqueous phase of the geopolymer, which could otherwise induce premature drying, debonding or micro-cracking. The fresh mixture was applied by spatula, levelled and smoothed under a plastic sheet, and subsequently cured under ambient laboratory temperature and humidity conditions for 28 days.

The FA-3 geopolymer coating–substrate interfaces were examined by optical microscopy (HIROX RH-2000, Hirox Co., Ltd., Tokyo, Japan) on cross-sections of the coated specimens. This analysis was carried out to assess the uniformity of the applied layer, the regularity of the coating thickness on the different substrate types, and the quality of the interfacial contact. Attention was paid to the identification of possible interfacial gaps or detachment zones between the geopolymer layer and the substrate, as well as to the evaluation of coating penetration and mechanical interlocking in the case of porous substrates.

The adhesion of the FA-3 geopolymer coating to the solid clay brick substrate was further evaluated by a non-standardised micro pull-off test adapted from ISO 4624 [50]. The same procedure was also applied to the fibre-free benchmark coating (FA-0) deposited on solid clay brick to provide a control reference. Aluminium dollies with a diameter of 12 mm were bonded to the surface of the consolidated geopolymer layer using a two-component structural epoxy adhesive. Prior to bonding, the dolly surfaces were mechanically roughened using sandpaper to enhance adhesion, while the brick substrate surface was lightly abraded to remove surface irregularities and ensure consistent contact conditions. Given the very limited thickness of the geopolymer coating (≈ 2 mm), pre-cutting around the dollies was intentionally avoided. In such thin layers, perimeter scoring is known to induce local damage and micro-cracking, which can compromise the validity of pull-off measurements. The resulting dolly diameter/coating thickness ratio ($\approx 6:1$) is consistent with typical geometries adopted for adhesion testing of thin mineral renders and skim coats, and it was considered adequate to ensure effective load transfer through the full coating thickness without the need for edge isolation. The pull-off tests were carried out using a universal testing machine (Z050, Zwick-Roell, Ulm, Germany) under displacement-controlled conditions, with a loading rate of 0.5 mm min^{-1} . Five dollies were tested, and the maximum pull-off force was recorded for each specimen. The failure mode (adhesive or cohesive) was determined by visual inspection of the dolly imprint and the corresponding fracture surface on the substrate.

Finally, the water vapour transmission of the FA-3 formulation was evaluated using an adapted dry-cup experimental setup adapted from ISO 12572 [51]. Three circular specimens of FA-3 geopolymer composites were cut from consolidated plates and ground to ensure uniform thickness (3.5 ± 0.2 mm). Cylindrical glass vessels (internal diameter 50 mm) were filled with ~ 20 g of anhydrous silica gel ($\text{RH} \approx 0\%$) and covered with a single FA-3 disc placed on each vessel opening. The specimens were sealed at the perimeter using a non-permeable plastic film and self-sealing tape to prevent lateral vapour diffusion. The assemblies were then placed inside a climatic chamber at $(23 \pm 2)^\circ\text{C}$ and $(50 \pm 5)\%$ RH. The mass of the sealed systems was recorded every 90 min using a precision balance (± 0.001 g) until a steady linear mass-gain regime was reached. The slope of mass increase,

dm/dt [$g\ h^{-1}$], was obtained by linear regression of the final quasi-steady portion of the curve. The water vapour transmission rate per unit area, W [$kg\ m^{-2}\ s^{-1}$], was calculated as

$$W = \frac{dm}{dt} \frac{1}{A} \quad (1)$$

where A (m^2) is the exposed area of the specimen. The water vapour permeability coefficient, δ ($kg\ m^{-1}\ s^{-1}\ Pa^{-1}$), was then derived from

$$\delta = \frac{(W \cdot d)}{\Delta p} \quad (2)$$

where d is the specimen thickness (m) and Δp is the water vapour pressure difference across the specimen. Under the dry-cup configuration used here ($\approx 0\%$ RH inside the vessel maintained by silica gel and 50% RH outside the vessel), the external vapour pressure was calculated as $p_{ext} = RH_{ext} \times p_{sat}(23\ ^\circ C) = 0.50 \times 2.81 \times 10^3\ Pa \approx 1.41 \times 10^3\ Pa$, while the internal vapour pressure was assumed negligible ($p_{int} \approx 0\ Pa$). Therefore, Δp was approximated as $p_{ext} - p_{int} \approx 1.41 \times 10^3\ Pa$. Finally, the water vapour diffusion resistance factor μ was estimated as

$$\mu = \frac{\delta_{air}}{\delta} \quad (3)$$

taking the water vapour permeability of stagnant air at $23\ ^\circ C$ as $\delta_{air} = 2.0 \times 10^{-10}\ kg\ m^{-1}\ s^{-1}\ Pa^{-1}$.

3. Results and Discussion

3.1. Chemical and Morphological Characterisation of the Microfibre Lint

The collected dryer-filter microfibres were preliminarily characterised to assess their morphology and chemical nature prior to incorporation into the geopolymer matrix. As observed by ESEM (Figure 2), the recovered lint consisted of a highly entangled and heterogeneous network of filaments forming loose agglomerates at the millimetre scale. Higher-magnification micrographs highlighted fibres with variable diameters ($8\text{--}30\ \mu m$) and surface textures, including smooth regions and longitudinal striations typical of synthetic textile filaments, confirming the mixed and irregular nature of the waste-derived microfibre fraction.

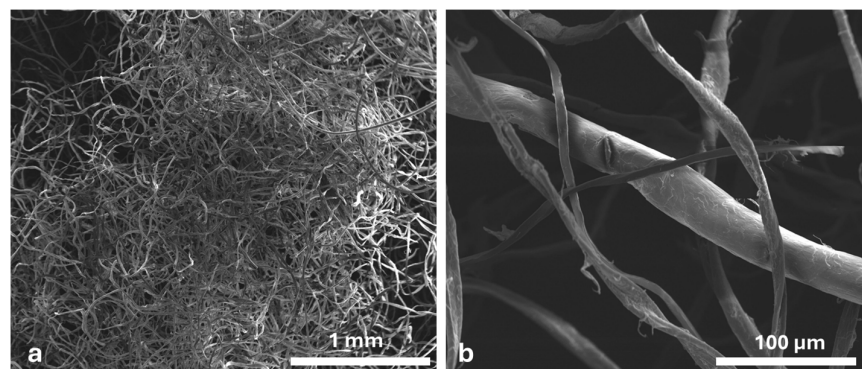


Figure 2. ESEM micrographs of the synthetic microfibre lint: low magnification (a), high magnification (b).

ATR-FTIR analyses were performed on three different lint subsamples batches (Figure 3). Since these batches contained fibres deriving from textile blends with different compositions, the spectra contained peaks related to functional groups of several fibre types that are often superimposed. A broad band at around $3300\ cm^{-1}$ due to the O-H stretching

vibration could be attributed to cellulose-type fibres but also to polyacrylic fibres [52]. Polyacrylic fibres present evident peaks at around 2920 cm^{-1} , 2850 cm^{-1} and 2240 cm^{-1} due to C-H₂ asymmetric stretching, C-H₂ symmetric stretching and C≡N stretching, respectively [53]. Polyester fibres showed typical bands at ~ 1730 and 1230 cm^{-1} due to C=O and C-O stretching, while the band at $\sim 1640\text{ cm}^{-1}$, is compatible with polyamide components (amide-related absorption) as nylon, but also in combination with the band at ~ 1530 to amide vibration in protein fraction of wool-type fibres, present in textile blends with mixed composition [53].

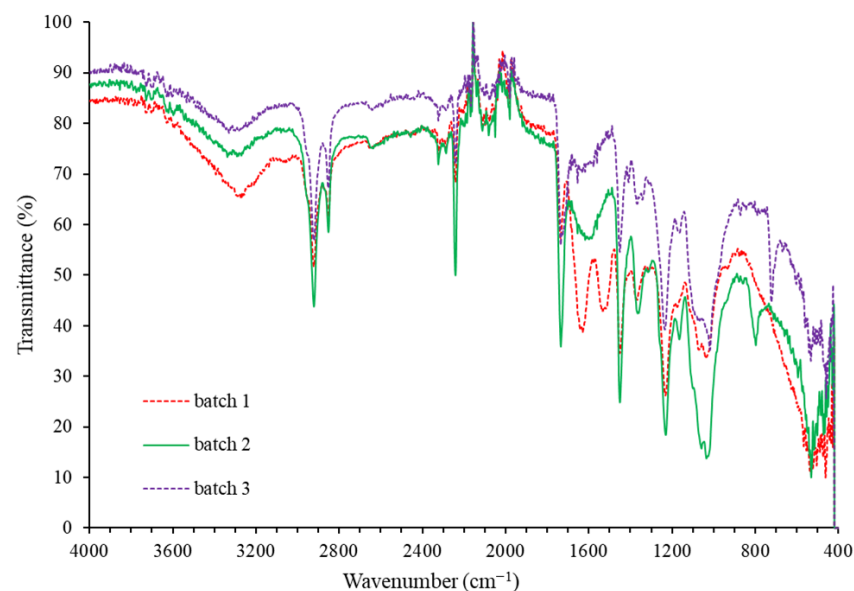


Figure 3. ATR-FTIR spectra of different dryer-filter lint subsamples.

3.2. Physical and Microstructural Characterisation of the Geopolymer Materials

The physical properties of the benchmark geopolymer and the three fibre-reinforced geopolymer formulations are reported in Table 1. The physical properties of the three geopolymer composites revealed that the fibre content influenced bulk density and, to a lesser extent, total porosity, through a combination of water-release dynamics during curing and volumetric effects associated with the fibre phase. FA-0 provided a reference bulk density of 1.48 g cm^{-3} and a total porosity of 42.7%. Compared to FA-0, all fibre-reinforced formulations exhibited lower bulk density and a slightly reduced total porosity, suggesting that microfibre addition might have both influenced the consolidation behaviour by modifying water redistribution and evaporation kinetics during curing, both affected the volumetric balance between the dense inorganic matrix and the low-density fibrous phase (Table 1).

Table 1. Physical properties of the investigated formulations.

Formulation	Fibre Content [wt.%]	Bulk Density ¹ [g cm ⁻³]	Total Porosity ¹ [%]	Av.ge Pore Diameter ¹ [μm]	Water Absorption [%]
FA-0	0.0	1.48	42.7	0.015	35.7 ± 0.5
FA-1	0.1	1.33	40.0	0.012	31.0 ± 0.5
FA-2	0.2	1.40	34.7	0.012	30.6 ± 0.5
FA-3	0.3	1.25	39.4	0.014	33.7 ± 0.5

¹ By MIP (instrument accuracy of 4%).

In particular, FA-1 showed a bulk density of 1.33 g cm^{-3} and a total porosity of 40.0%, while FA-2 reached a higher density (1.40 g cm^{-3}) and the lowest porosity (34.7%), indicating the most compact structure among the reinforced formulations. FA-3 presented a lower density (1.25 g cm^{-3}) and a porosity value close to that of FA-1, despite the higher fibre content. The observed reduction in bulk density relative to FA-0, combined with a simultaneous decrease in total porosity in FA-1, FA-3 and especially FA-2, suggested the interplay of two competing mechanisms. On the one hand, the incorporation of a very low-density fibrous phase introduced a volumetric effect, which tended to reduce the overall bulk density of the composites even at low weight fractions. On the other hand, the presence of dispersed microfibrils appeared to influence water redistribution and evaporation during early curing, locally moderating drying rates and promoting a more gradual consolidation of the geopolymer gel. This second mechanism may account for the reduction in total porosity observed in the fibre-reinforced formulations compared with FA-0, with FA-2 representing the most favourable balance between volumetric fibre contribution and consolidation of the matrix. At higher fibre loading (FA-3), the volumetric effect of the low-density phase became dominant, leading to a further decrease in bulk density without a corresponding reduction in porosity.

Regarding pore size distributions, reported in Figure 4, all samples presented an average pore diameter in a narrow range between approximately 0.012 and $0.014 \text{ }\mu\text{m}$, confirming the predominant presence of a fine porous structure in the nano-mesoporous range typical of well-reacted metakaolin-based polysialate systems [48] and consistent with the reference geopolymer matrix FA-0 [49]. FA-0 showed some pores in the range $10\text{--}100 \text{ }\mu\text{m}$, probably formed by air bubbles trapped during casting and subsequent consolidation. The fibre-reinforced samples do not show the presence of pores in this range, but a minor contribution of pores in the $0.1\text{--}10 \text{ }\mu\text{m}$ range. Such small variations are not considered significant and indicate that fibre addition did not modify the intrinsic pore size distribution of the geopolymer gel, nor did it promote the formation of fibre-induced macropores.

Water absorption values remained relatively similar across the reinforced formulations (30.6–33.7%), indicating that fibre addition did not substantially alter pore accessibility or connectivity. This is consistent with MIP results, showing comparable average pore diameters and suggesting that the microfibrils mainly affected the volumetric balance and drying dynamics rather than the intrinsic pore-network structure of the geopolymer gel.

SEM observations were performed to assess the morphology and homogeneity of the geopolymer matrix and to identify possible microstructural differences related to fibre content. Representative micrographs of the benchmark geopolymer and the three fibre-reinforced formulations are shown in Figure 5, reporting surface regions at lower magnification and corresponding high-magnification views of the matrix. At high magnification, all samples displayed a highly homogeneous and well-developed geopolymer gel, characterised by the typical nanostructured morphology of metakaolin-based polysialate systems.

The benchmark formulation FA-0 (Figure 5a) showed a homogeneous and uniform geopolymer microstructure, characterised by a relatively dense gel texture at the micro scale and the formation of typical geopolymer nanoprecipitates. Occasional porosity, likely related to air bubbles trapped during casting, are visible on the surface, in agreement with the higher total porosity measured by MIP and pore size distribution (Figure 4).

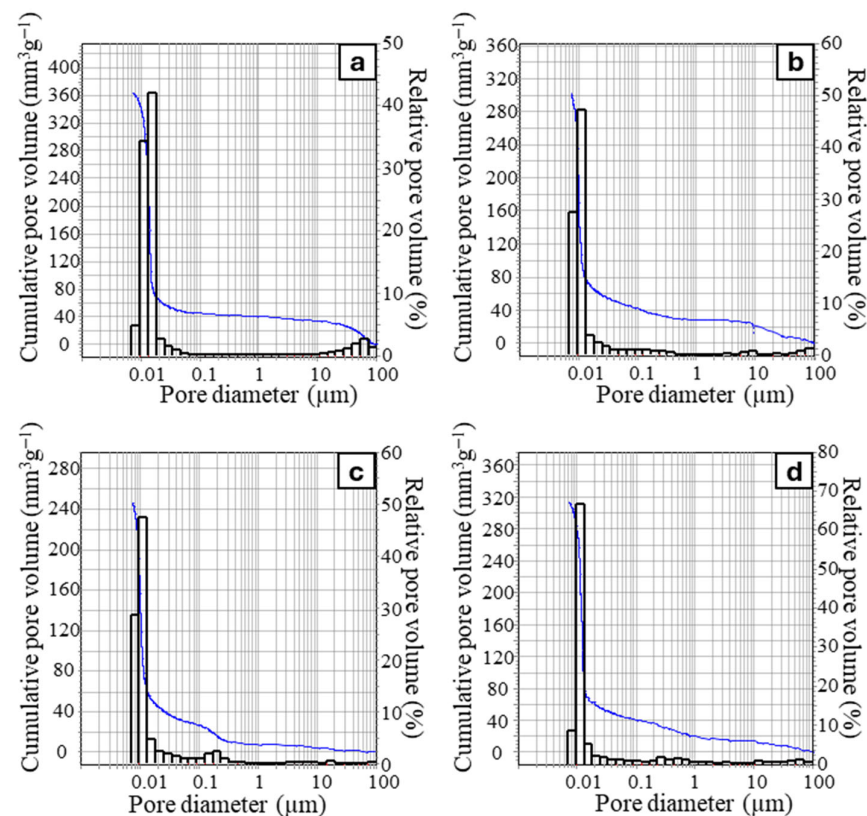


Figure 4. Pore size distribution of the investigated formulations: (a) FA-0; (b) FA-1; (c) FA-2; (d) FA-3. Blue line: cumulative pore volume; bars: relative pore volume for each pore size class.

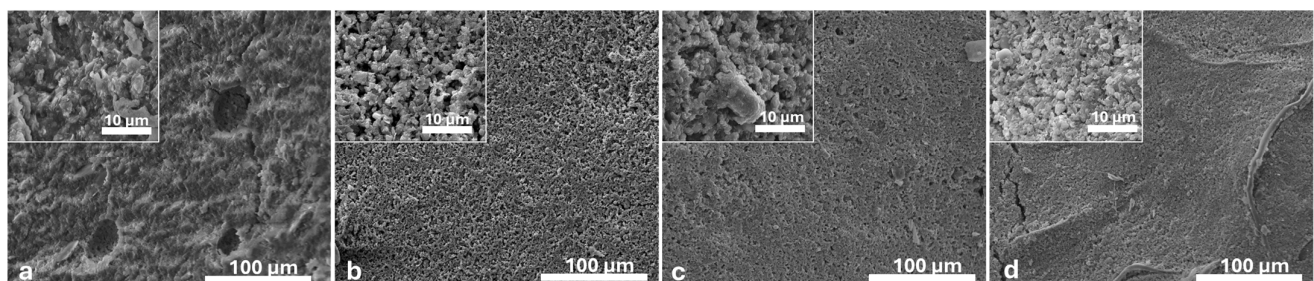


Figure 5. ESEM micrographs of the investigated formulations (external surface): (a) FA-0; (b) FA-1; (c) FA-2; (d) FA-3.

Compared to FA-0, the fibre-reinforced formulations (Figure 5b–d) generally exhibited a less dense and continuous surface layer, despite a reduced occurrence of open surface defects, especially for FA-2 and FA-3. This behaviour could be attributed to the interaction between the microfibres and the fresh geopolymer paste, which may locally retain and redistribute free water, moderating evaporation kinetics and influencing surface consolidation during curing. The aluminosilicate network appeared continuous and uniformly distributed, with no evidence of unreacted metakaolin platelets, secondary crystalline phases or phase separation, albeit superficial interparticle porosity at the micro–nano scale. This confirmed that the geopolymerisation process proceeded effectively in all formulations and that fibre addition did not interfere with the development of the inorganic matrix. At the lowest fibre percentage (FA-1, Figure 5b), the surface showed a homogeneous texture with visible micro voids. As fibre content increased (Figure 5c), the surface morphology progressively evolved toward a more compact and continuous appearance, with a reduction in the number and size of open surface voids, reflecting a more controlled con-

solidation of the matrix (Figure 5c,d). At the highest fibre content (Figure 5d), the surface remained dense but locally intersected by shallow fissures and few micro-cracks, which are attributed to shrinkage stresses developing during drying in the presence of a larger fibre volume fraction. In this case, fibres were more frequently observed at or near the surface, although they remained fully embedded within the geopolymer matrix (Figure 5d). The fibre addition did not appear to generate fibre-induced macroporosity, as no elongated voids or fibre-shaped cavities are observed after fracture. Instead, the differences in bulk density among formulations could be attributed mainly to the volumetric contribution of the low-density fibrous phase and to subtle variations in gel packing and drying kinetics. The slightly more continuous surface observed for FA-2, compared with FA-1 and FA-3, is consistent with the lower total porosity measured by MIP and may reflect a more balanced redistribution of water during early curing.

Moreover, fracture surfaces (Figure 6) revealed that, regardless of fibre content, the internal microstructure of the composites is uniform and well consolidated. The geopolymer gel exhibited a fine, nanostructured morphology throughout the section, and no evidence of fibre sedimentation, clustering or preferential orientation was observed. The fibres were evenly distributed and fully surrounded by the matrix, indicating good physical compatibility between the organic synthetic phase and the inorganic gel. Importantly, no radial voids or interfacial gaps were detected around the fibres, suggesting that differences in bulk density and porosity are not associated with interfacial defects or incomplete geopolymerisation. Fibre incorporation, indeed, primarily affected the composite at the meso scale through volumetric and drying-related effects, rather than by introducing interfacial defects or discontinuities in the geopolymer gel [54]. The absence of fibre-induced macroporosity and the continuity of the matrix supported the physical and mechanical results, where the reduction in compressive strength coexists with improved flexural behaviour, indicating that the fibres act mainly as crack-bridging inclusions without compromising overall matrix integrity. Observations of fracture surfaces revealed several fibres intersecting the crack path and locally bridging adjacent matrix regions. Notably, individual fibres were observed to remain anchored across partially separated fragments or to protrude from the fracture plane after debonding, often with residual geopolymer gel adhering to their surface (Figure 6d).

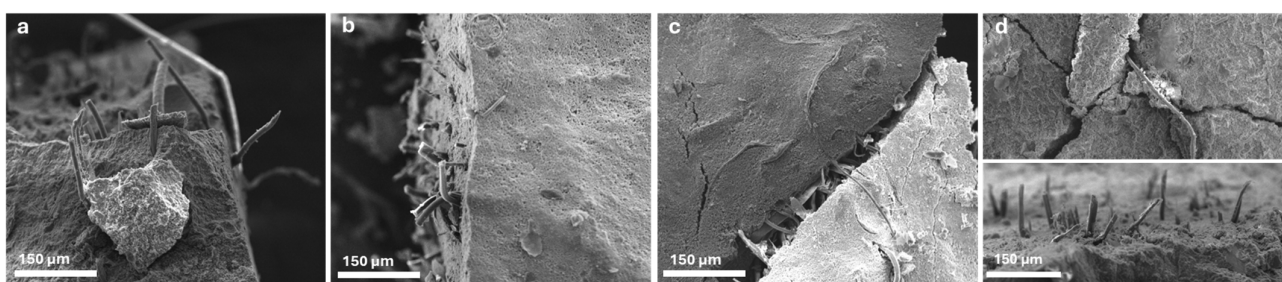


Figure 6. ESEM micrographs of the investigated formulations (fracture surfaces): (a) FA-1; (b) FA-2; (c,d) FA-3.

3.3. Mechanical Characterisation

The flexural strength results (Table 2) showed that FA-1 and FA-2 exhibited comparable average values of 5.7 ± 1.3 MPa and 5.3 ± 0.8 MPa, respectively, while FA-3 reached a higher value of 7.5 ± 0.4 MPa, corresponding to an increase of approximately 30–40% compared to the formulations with less fibre. The flexural strength of the unreinforced geopolymer FA-0 was 3.0 ± 0.6 MPa, allowing the toughening contribution of microfibre addition to be directly quantified as an overall increase of approximately 75–150% in flexural strength relative to the unreinforced matrix, depending on fibre dosage. This trend

indicates that, within the investigated range, microfibre incorporation did not act as an inert filler but contributed to crack-bridging (Figure 6d) and stress redistribution mechanisms, leading to a more damage-tolerant behaviour of the composite under bending. In particular, the fibre addition produced an overall improvement in flexural strength compared to the unreinforced matrix: while FA-1 and FA-2 showed similar values within experimental scatter, the FA-3 formulation exhibited a clearly higher strength even when variability was considered, indicating that the reinforcing effect becomes more significant at the highest fibre dosage.

Table 2. Mechanical properties of the investigated formulations.

Formulation	Fibre Content [wt.%]	Compressive Strength [MPa]	Flexural Strength [MPa]
FA-0	0.0	42.8 ± 5.0	3.0 ± 0.6
FA-1	0.1	25.6 ± 5.4	5.7 ± 1.3
FA-2	0.2	24.7 ± 5.6	5.3 ± 0.8
FA-3	0.3	22.0 ± 3.2	7.5 ± 0.4

Compared with values reported in the literature for metakaolin-based potassium polysialate matrices of similar composition [55], the flexural strengths measured for FA-1–3 are higher, highlighting the effect of the specific synthetic microfibre addition used here, suggesting that the mixed textile-derived microfibres provided a distinct reinforcing contribution, likely related to their morphology and dispersion within the matrix.

Also, the analysis of fracture surfaces (Figure 6) provided useful insight into the mechanical behaviour of the composites and supports the trends observed in flexural testing. The fracture morphology highlighted the role of fibre content in governing crack propagation mechanisms and energy dissipation during bending.

At low fibre content (Figure 6a), fibres were sparsely distributed and did not significantly influence the overall fracture morphology. Crack propagation was therefore largely controlled by the geopolymer matrix, and the fracture surface exhibited the typical brittle character of a ceramic-like material. Nevertheless, the fibres were well embedded within the matrix, and no macroscopic debonding, interfacial gaps or voids were observed, indicating good fibre–matrix adhesion despite the limited reinforcement fraction.

As fibre content increased (Figure 6b,c), the interaction between the fibres and matrix became more evident along the fracture surfaces. In these formulations, fibres were more frequently involved in the fracture process, and the crack path increasingly interacted with the reinforcing phase. In FA-3, and to a lesser extent in FA-2, fracture surfaces clearly revealed the activation of fibre pull-out mechanisms, with fibres partially extracted from the matrix rather than fractured (Figure 6d). This behaviour indicated that crack propagation involved frictional sliding and interfacial debonding, contributing to increased energy dissipation during fracture. There, fibres bridging the crack path could be observed to maintain partial cohesion between fractured regions, preventing abrupt separation and resulting in a less brittle failure mode. This transition from a predominantly matrix-controlled fracture behaviour, observed in FA-1, toward a more damage-tolerant response in FA-3 was consistent with the higher flexural strength and reduced scatter measured for the latter formulation. Overall, the fracture surface analysis confirmed that increasing fibre content promoted the activation of distributed toughening mechanisms without altering the integrity or consolidation of the geopolymer matrix.

Compressive tests showed that strength values decreased progressively with increasing fibre content by approximately 40–50% compared with the unreinforced material FA-0 (Table 2). This reduction can be attributed to the presence of a low-elastic-modulus organic phase within the otherwise rigid geopolymer matrix and to local stress concentrations at

the fibre–matrix interface, rather than to an increase in void content. This behaviour is consistent with the literature data [36–40], which report reductions in compressive strength in synthetic-fibre-reinforced geopolymers, often associated with relatively weak fibre–matrix interactions related to the intrinsically hydrophobic nature of polymeric fibres.

3.4. Thin-Layer Application and Substrate Compatibility

Based on the results of physical, microstructural and mechanical characterisation, formulation FA-3 was selected for further investigation of its applicability as a thin, continuous geopolymer coating for building materials, as it provided the best balance between mechanical performance and workability while also enabling the highest incorporation and immobilisation of waste microfibres among the tested formulations. This choice was motivated not only by its enhanced flexural performance and reduced data scatter, but also by the activation of microfibre-related toughening mechanisms observed at fracture surfaces, while preserving a homogeneous and well-consolidated geopolymer matrix. In addition, the physical properties of FA-3, namely bulk density, total open porosity, pore size distribution and water absorption, fell within the ranges commonly reported for mineral-based finishing mortars and rendering systems [56–58]. Total porosity values of approximately 39% and water absorption around 34 wt.% indicated a pore network capable of storing and redistributing moisture, while mercury intrusion porosimetry revealed that most of the pore volume was concentrated in the nanometric range (~10–14 nm). Such a microstructural profile, typical of metakaolin-based geopolymers with limited macrovoid formation [48], suggested that FA-3 matrix could engage in moisture interaction with porous substrates and participate in capillary liquid transport and moisture buffering at the macro scale. These properties are thought to be relevant for mineral surface finishings intended for building envelopes, where mechanical robustness, crack tolerance and moisture interaction are required alongside adequate adhesion. FA-3 was therefore selected as the reference formulation for thin-layer coating trials (Figure 7). A fibre-free benchmark coating (FA-0) was also prepared on clay brick as a control reference for adhesion testing.

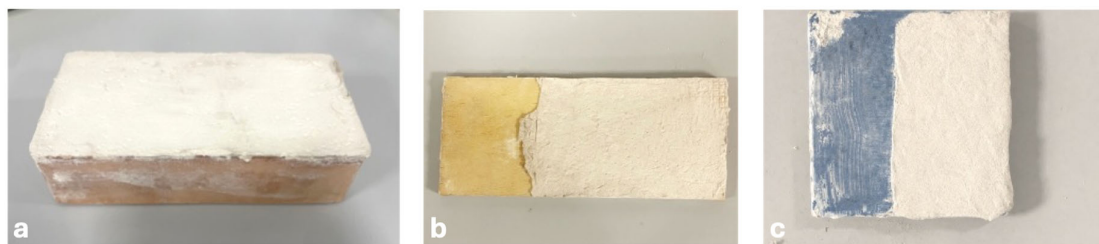


Figure 7. Thin-coated architectural surfaces by deposition of FA-3 layers: (a) clay brick; (b) ply wood; (c) glazed porcelain tile.

The selected substrates were chosen to span a wide range of surface chemistry, roughness and porosity, and thus to provide a realistic evaluation of coating–substrate compatibility under different conditions. A standard solid clay brick was used as a porous mineral substrate (250 × 120 × 60 mm), representative of conventional masonry elements. A laminated wood panel of comparable dimensions was selected as an organic and hygroscopic substrate, while a glazed porcelain stoneware tile surface (100 × 100 mm) was employed as a dense, non-porous and low-roughness substrate, representing a particularly demanding condition in terms of adhesion and interfacial compatibility. The resulting coated specimens were cured under ambient laboratory conditions and subsequently analysed to assess adhesion and interface quality.

3.4.1. Assessment of Interfacial Adhesion by Optical Microscopy

Optical microscopy observations of the FA-3 geopolymer coating surfaces and cross-sections (Figure 8) highlighted that the applied layer exhibited a certain variability in thickness, mainly attributable to the manual application procedure adopted for the deposition. Differences in local thickness and surface regularity could be observed across the coated areas, reflecting the intrinsic limitations of hand-applied thin layers. Measurements across the three investigated substrates indicated an average coating thickness of approximately 180 μm . On the clay brick substrate (Figure 8a), however, the application of light pressure using a plastic film immediately after spatula deposition proved effective in significantly levelling the surface, resulting in a more uniform thickness and a smoother profilometric profile of the coating layer.

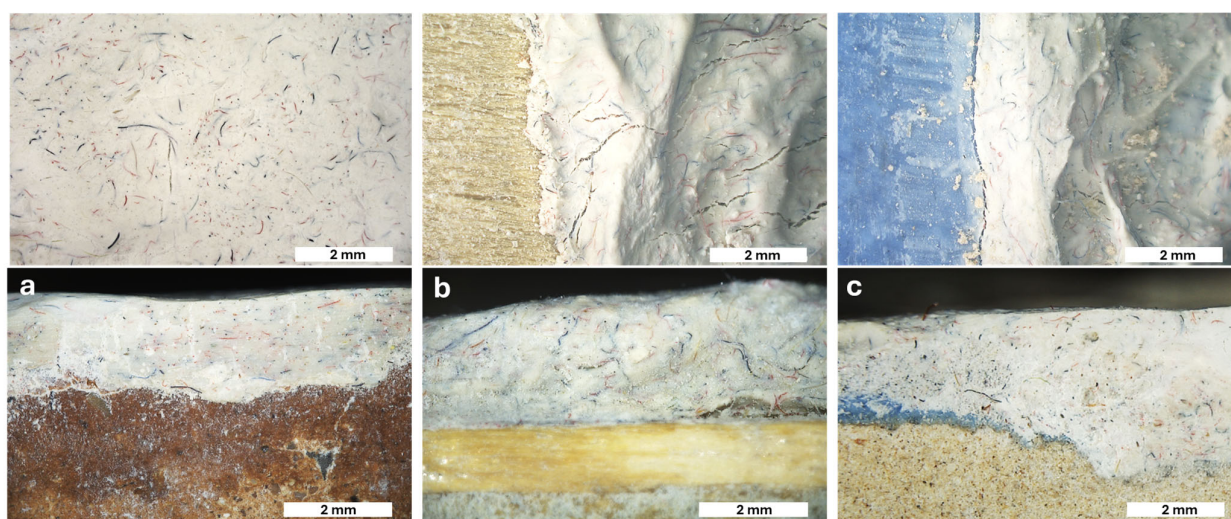


Figure 8. Optical microscopy images of the FA-3 thin layers onto different substrates. ((a), upper panel: top view; lower panel: side view) clay brick; ((b), upper panel: top view; lower panel: side view) ply wood; ((c), upper panel: top view; lower panel: side view) glazed porcelain tile.

Despite the relatively high fibre content, the reinforcement appeared homogeneously distributed and fully embedded within the geopolymer matrix in all cases. No fibre protrusion or preferential exposure at the surface was observed, confirming that the inorganic matrix effectively encapsulated the microfibres and prevented their potential release or re-dispersion into the surrounding environment once the coating is consolidated.

On the plywood substrate (Figure 8b), optical microscopy revealed the presence of superficial micro-cracking and a tendency toward localised interfacial detachment. These features were attributed to insufficient substrate pre-treatment prior to coating application. Limited pre-wetting of the wooden support resulted in partial absorption of free water from the fresh geopolymer paste, promoting local shrinkage during drying and the formation of micro-cracks at the surface and interface. This behaviour did not preclude the use of FA-3 as a coating for porous substrates, but rather indicated the necessity of appropriate substrate conditioning, including adequate pre-wetting to achieve near-saturation conditions and to minimise differential shrinkage effects.

In contrast, the pre-wetting procedure adopted for the clay brick substrate proved sufficient to ensure dimensional stability of the coating, with no evidence of micro-cracking or interfacial detachment across the examined sections.

For the glazed porcelain stoneware substrate (Figure 8c), the coating–substrate interface appeared continuous and well-bonded, with no observable gaps, voids or entrapped air pockets, despite the low porosity and smooth surface of the ceramic support. A slight

local lifting of the coating was observed only at the extreme edge of the applied layer, as visible in the upper-right micrograph. This feature could be attributed to unconfined shrinkage occurring at the termination of the coating, where the layer thickness locally reached approximately 2.7 mm due to the end of the spatula stroke. The effect remained highly localised, limited to a few micrometres from the coating edge, and did not affect the integrity of the coating in the central regions.

For comparison purposes, a fibre-free benchmark geopolymer coating (FA-0) was also applied on solid clay brick under the same surface preparation and curing conditions. Unlike FA-3, FA-0 developed pronounced early-age shrinkage micro-cracking over the coated surface, with a typical mud-cracking pattern and local partial detachment (Figure 9). This different behaviour was attributed to the absence of microfibre reinforcement, which, in FA-3, contributed to increasing the fresh paste consistency and to mechanically limiting shrinkage crack formation through micro-bridging effects, whereas the more fluid FA-0 matrix was more prone to drying-related deformation even though the substrate surface was pre-wetted to saturation.



Figure 9. FA-0 thin layer applied on clay brick substrate.

3.4.2. Adhesion Strength by Micro Pull-Off Test on Clay Brick Substrates

The adhesion of the FA-3 geopolymer thin coating to the solid clay brick substrate was evaluated by a micro pull-off test. A total of four (out of five measurements) valid peak loads were recorded. Using a 12 mm dolly (bonded area of 113.1 mm²), the four values recorded for the FA-3 thin layer on solid brick corresponded to an average tensile bond strength of 0.122 ± 0.036 MPa. In all these valid tests, failure occurred at the coating/brick interface (Figure 10), indicating that the measured value was mainly governed by interfacial adhesion and near-interface microstructure rather than by the bulk cohesive strength of the geopolymer composite.

The irregular and jagged perimeter of the detached imprint, rather than a perfectly planar detachment, was consistent with mixed micro-mechanisms during separation, as local tearing and micro-mechanical interlocking. The brick surface underneath remained partially soiled by residual material, suggesting that the geopolymer layer penetrated superficial roughness and porosity, and that debonding required partial disruption of the interfacial zone. The microfibrils did not appear to drive the debonding process directly; however, their presence could help the detached layer to remain coherent without powdering or fragmentation, which is beneficial for serviceability, limiting spalling during local damage.



Figure 10. Dolly imprints (diameter 12 mm) showing interface failure after pull-off test on clay brick substrate.

Conversely, the pull-off adhesion test could not be performed on the fibre-free benchmark formulation FA-0 applied on the clay brick, since the coating spontaneously detached from the substrate in small fragments, preventing proper dolly bonding and therefore meaningful pull-off measurements (Figure 9). These observations suggested that the microfibre incorporation provided also a functional contribution to the integrity and handling resistance of thin geopolymer finishings.

From a comparative perspective, the measured adhesion values for FA-3 (0.09–0.17 MPa) fell within the typical range reported for thin mineral-based coatings, lime-based plasters and finishing layers applied on porous masonry substrates, including earthen plasters defined in DIN 18947 and mineral rendering mortars classified according to EN 998-1, where pull-off strengths are strongly dependent on substrate preparation, curing conditions and test geometry [59–61]. According to DIN 18947, tensile bond strength values for earthen plasters are typically below ~0.1 MPa, whereas mineral rendering mortars generally exhibit tensile bond strengths above ~0.08–0.10 MPa on porous masonry substrates, depending on formulation and surface preparation [62,63].

Although the average pull-off strength obtained for FA-3 (~0.12 MPa) is lower than that typically reported for high-performance protective or repair coatings on concrete, the material is intended as a thin architectural finishing layer rather than for structural use. In such applications, adhesion is largely controlled by substrate conditions and pull-off values are interpreted together with the failure mode. In the present case, failure consistently occurred at the coating/brick interface, indicating that the measured strength was governed mainly by the near-surface cohesion of the brick and the interfacial zone rather than by the cohesive strength of the geopolymer layer. This failure mode and adhesion level are consistent with those commonly reported for mineral rendering mortars rather than earthen plasters, despite the absence of fine aggregates in the geopolymer formulation. In this context, the adhesion performance of FA-3 was compatible with that of conventional mineral finishing and lining systems and is particularly notable given the very small coating thickness and the absence of primers or surface treatments beyond standard pre-wetting. When considered together with the comparatively high flexural strength measured for FA-3 (~7.5 MPa), these results indicate a mechanical response closer to mineral mortars, with flexural behaviour influenced by fibre-related toughening mechanisms rather than aggregate interlocking.

3.4.3. Water Vapour Permeability

The water vapour transmission properties of FA-3 were evaluated using an adapted dry-cup configuration based on gravimetric monitoring under controlled climatic conditions. A steady linear mass-gain regime was reached after approximately 21 h, consistent with the limited specimen thickness (3.5 ± 0.2 mm). In this regime, the mass of the sealed

cup–specimen assembly increased linearly as a function of time, and the final quasi-steady portion of the mass–time curve was fitted by linear regression (Figure 11). For the representative specimen considered, the regression yielded

$$m = 0.0052 \cdot t + 85.49 \quad (R^2 = 0.99) \quad (4)$$

where m is the mass in grams and t the time in hours. The exposed area of the specimen, excluding the sealed rim, was 40 mm in diameter ($A = 1.26 \times 10^{-3} \text{ m}^2$). The water vapour transmission rate per unit area (WVTR) was obtained as the ratio between the slope of the regression line and the specimen area, resulting in $\text{WVTR} = 4.15 \text{ g m}^{-2} \text{ h}^{-1}$. Using the measured WVTR, the specimen thickness and the vapour-pressure gradient imposed by the dry-cup configuration at 23°C ($\Delta p \approx 1.41 \times 10^3 \text{ Pa}$), the permeability δ of the composite was calculated as $\delta = 2.87 \times 10^{-12} \text{ kg m}^{-1} \text{ s}^{-1} \text{ Pa}^{-1}$. The corresponding water vapour diffusion resistance factor μ , assuming the vapour permeability of still air at 23°C , was $\mu = 73.2$.

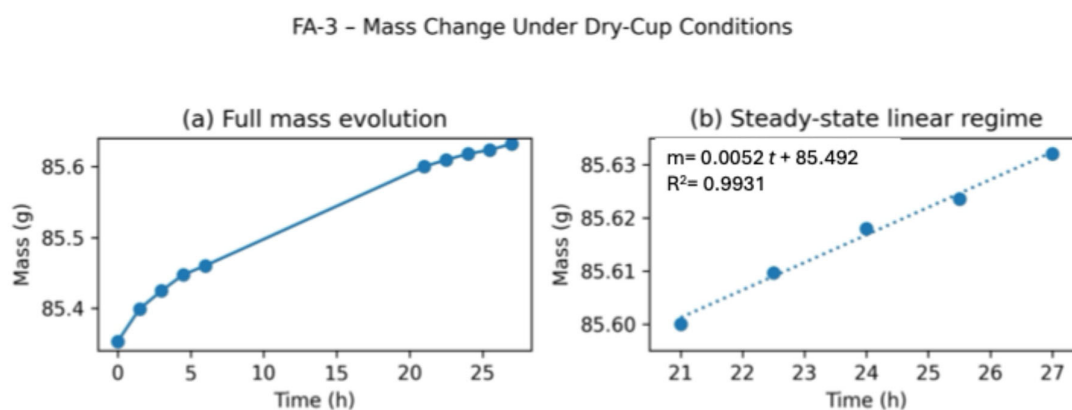


Figure 11. FA-3 mass evolution under adapted dry-cup conditions at $(23 \pm 2)^\circ\text{C}$ and $(50 \pm 5)\%$ RH: (a) full mass–time curve; (b) steady-state linear regime.

Measurements on three nominally identical FA-3 specimens produced μ values within $\pm 10\%$ of this mean. This variability is attributed to experimental uncertainties inherent to the non-standard laboratory setup, including incomplete desiccant dehydration, potential minor multidirectional flows or edge leakage at the seal, and slight thickness inhomogeneity of the discs. These limitations do not affect the interpretation of the results but constrain the precision of the absolute values obtained.

Despite the relatively high total porosity and water absorption, the measured μ indicated that vapour transport through FA-3 was hindered when compared to other highly porous mineral or clay-based finishing materials, such as natural lime plasters or earthen plasters, which typically show significantly lower diffusion resistance factors (dry-cup μ in the range of ~ 5 – 15 for lime and unfired clay materials) under comparable test conditions [64]. This behaviour might be explained by the intrinsic nanometric scale of the geopolymer pore network, dominated by gel pores with an average diameter of $\sim 0.014 \mu\text{m}$ (Figure 4). Under these conditions, the characteristic pore size became much smaller than the molecular mean free path of water vapour in air at ambient temperature and pressure ($\lambda \approx 0.1$ – $0.2 \mu\text{m}$), placing the composite in a regime where vapour transport is partially governed by Knudsen diffusion [65–67]. In this regime, molecule–wall collisions dominate over intermolecular collisions, reducing effective diffusivity even in materials with substantial void volume and connectivity [67,68].

To evaluate whether the measured μ was compatible with vapour transport controlled by nanoporous flow, theoretical estimates were derived using the porosity-based relation

proposed by Hall et al. [64] and corrected for Knudsen effects, where the Knudsen number is defined as $Kn = \frac{\Lambda}{\lambda}$, with λ as the molecular mean free path and Λ the characteristic pore diameter of the material. For FA-3, the total porosity fraction from MIP was $f = 0.39$, the average pore diameter was $\Lambda \approx 1.4 \times 10^{-2} \mu\text{m}$, and a representative mean free path of water vapour molecules in air at ambient conditions was taken as $\lambda \approx 1.2 \times 10^{-1} \mu\text{m}$ [64]. Using these values a Knudsen number $Kn \approx 9$ was obtained, indicating a flow regime dominated by molecule–wall interactions ($Kn \gg 1$). Incorporating this correction into the porosity-based diffusion resistance model of Hall et al. produced a theoretical diffusion resistance factor μ of the order of 65–70, which is in close agreement with the experimentally determined value of $\mu \approx 73$ for FA-3. This consistency supported the interpretation that the resistance to vapour transport in FA-3 arose primarily from pore-scale effects rather than from insufficient accessible porosity and confirmed that the measured values aligned with current theoretical descriptions of vapour mobility in nanoporous inorganic matrices.

Despite not achieving the very low diffusion resistance factors typically associated with highly breathable lime- or clay-based plasters, the physical profile of FA-3 depicted a material that is able to retain moisture-exchange capacity while limiting the rate of vapour transport. In this context, relatively high μ can also be regarded as functionally advantageous, since a finer and less permeable pore network may reduce transport pathways and limit the potential migration of entrapped dryer-filter microfibres through the coating thickness. Nevertheless, FA-3 cannot be considered a fully vapour-blocking barrier (as typical synthetic or resin-based coatings), because the matrix remains intrinsically porous and still allows a non-zero vapour flux. This intermediate behaviour may still be functionally relevant for architectural surface treatments where a controlled, rather than maximised, vapour exchange is acceptable or desirable, and where excessive drying or rapid moisture fluxes could be detrimental to the substrate or finishing system.

4. Conclusions

The present study explored the reuse of dryer-filter synthetic microfibres as a reinforcing and immobilised waste fraction in metakaolin–potassium geopolymer coatings intended for architectural applications. Among the tested compositions, FA-3 (0.3 wt.% fibres) offered the most favourable balance between microstructural integrity and mechanical performance and was therefore retained as the representative formulation for coating assessment. A fibre-free benchmark formulation was also produced and tested for reference, confirming the functional contribution of the microfibre addition to coating integrity and adhesion.

FA-3 preserved the characteristic gel-based nanostructure of metakaolin-derived geopolymers and maintained a high degree of consolidation without evidence of fibre clustering or interfacial defects. Total accessible porosity (by MIP) remained in the range of $\sim 40\%$, and water absorption reached ~ 34 wt.%, confirming that fibre addition did not significantly alter the connectivity of the pore network or hinder matrix formation. Fracture surface observations indicated the activation of distributed toughening mechanisms at higher fibre content, including crack-bridging and fibre pull-out, which directly corresponded with the increased flexural strength (≈ 7.5 MPa) and reduced scatter measured for FA-3 compared with lower fibre dosages, albeit there was a reduction in compressive strength of $\sim 40\%$ compared to the plain geopolymer. Adhesion testing on clay brick substrates yielded values around 0.12 MPa under non-standard pull-off conditions, consistent with the performance range expected for thin mineral finishing systems. From an application standpoint, these results support the use of FA-3 as a thin mineral-based coating where moderate mechanical robustness, cohesive cracking behaviour and compatibility with porous substrates are required. The material retained adequate workability for manual

spreading, produced continuous layers at about 2 mm thickness and showed uniform encapsulation of fibres without surface exposure, which is relevant for serviceability and material integrity.

Beyond reinforcement, FA-3 exhibited a pore structure characterised by substantial open void volume and accessible moisture pathways, but with a vapour diffusion resistance factor in the moderate range ($\mu \approx 73$), attributable to the predominance of nanometre-scale gel pores. This intermediate behaviour indicates that FA-3 does not achieve the very low μ values typical of highly breathable lime- or clay-based coatings yet still retains a measurable moisture-exchange capacity. Such a balance between open porosity and moderated vapour diffusivity may also contribute to limiting the migration of entrapped microfibrils through the pore network while preserving non-zero vapour transfer in thin-layer applications. In such contexts, a mineral coating combining open pore volume with moderated vapour diffusivity can contribute to hygric buffering without acting as an unrestricted vapour conduit, supporting its potential use as a thin finishing or levelling layer in multi-coat mineral systems. Nevertheless, the vapour diffusion resistance performances measured here should be regarded as a preliminary estimation, and further, more systematic studies will be performed. Indeed, the versatility of geopolymer matrices and their hierarchically tunable pore network offer opportunities for future optimisation of such finishing systems, tailoring vapour diffusivity while maintaining mechanical performance and fibre immobilisation.

Another key outcome of this work is the demonstration that FA-3 provided an effective immobilisation pathway for a volumetrically significant textile waste fraction. Dryer-filter lint is characterised by a very low density, such that comparatively small masses correspond to large loose volumes that are difficult to stabilise and routinely discarded. Considering a 3 mm coating thickness and the measured bulk density of FA-3 ($1.25 \text{ g}\cdot\text{cm}^{-3}$), the proposed formulation would immobilise approximately 0.4 L of loose lint per m^2 , suggesting a potentially relevant waste diversion when applied at large surface areas in building and indoor applications. In this configuration, the coating acts not only as a functional architectural surface but also as an encapsulation medium, reducing the dispersive potential of mixed plastic microfibre residues. Overall, the results confirmed that dryer-filter microfibrils can be structurally integrated within geopolymer matrices without compromising consolidation or handling and can contribute simultaneously to composite performance and waste sequestration. Dryer-filter lint, in fact, is currently typically disposed of through incineration or landfilling: incineration may cause atmospheric emissions and results in the complete loss of material value, whereas landfilling may lead to long-term environmental release through leaching and water pathway contamination. In contrast, geopolymer encapsulation could offer a reuse-oriented route by physically immobilising microfibrils within a stable inorganic matrix, although potential microfibre dispersion during handling and mixing operations should also be considered and controlled through appropriate processing and safety measures.

The proposed approach should be therefore interpreted as a complementary circular strategy for a heterogeneous waste stream that is currently difficult to recycle: while prevention of fibre release remains the preferred option within the waste hierarchy, the incorporation of dryer-filter lint into a stable geopolymer matrix offers a downstream pathway for reuse and physical immobilisation of microfibrils already collected, potentially contributing to their diversion from residual waste and limiting environmental dispersion. At end of life, the fibre-containing geopolymer coating would fall within conventional mineral construction and demolition waste streams and could, in principle, be mechanically processed and reused as a secondary mineral fraction in new binders. Nevertheless, the present findings are based on laboratory-scale formulations using microfibrils collected from domestic dryer filters, which may not fully represent industrial textile waste streams;

future studies will therefore focus on long-term durability under service conditions and on validating the approach with fibres collected at larger and more controlled scales.

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Abbreviations

The following abbreviations and symbols are used in this manuscript:

RH	Relative humidity
WVTR	Water vapour transmission rate
A	Exposed specimen area (m ²)
D	Specimen thickness (m)
dm/dt	Mass change rate (g h ^{−1})
Kn	Knudsen number
m	Mass (g)
p_ext, p_int	External and internal water vapour pressure (Pa)
p_sat	Saturation vapour pressure (Pa)
W	Water vapour transmission rate per unit area (kg m ^{−2} s ^{−1})
WA	Water absorption (%)
δ	Water vapour permeability coefficient (kg m ^{−1} s ^{−1} Pa ^{−1})
Δp	Water vapour pressure difference across the specimen (Pa)
λ	Mean free path of water vapour molecules in air (m)
μ	Water vapour diffusion resistance factor (−)

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