

Effect of C₂S and limestone powder contents on interfacial bond characteristics of high-strength steel fibers embedded in low-carbon mortars

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ABSTRACT

This study explores influence of dicalcium silicate (C₂S) and limestone powder (LP) contents on the interfacial bond characteristics of smooth and hooked steel fibers embedded in low-carbon mortars. Four mortar compositions with varying LP (0–30%) and C₂S (16.4–57.2%) contents were evaluated through single-fiber pullout, compressive tests, SEM–EDS, and nanoindentation (NI) analyses. A mixture containing 10% LP substitution in belite-rich cement (BRC) exhibited optimal performance and sustainability. Although increased C₂S content in BRC reduced early-age strength, it significantly enhanced the long-term mechanical and interfacial properties and improved the 90-day compressive strength, bond strength, and pullout energy by up to 65%. Higher LP levels yielded denser fiber–matrix interfacial zones (FMZs) and reduced porosity (25.0%→18.9%), while elevated C₂S promoted greater C–S–H formation (34.6%→51.5%). The optimized LP–BRC system demonstrated superior durability and interfacial performance with substantially lower CO₂ emissions, offering a viable and sustainable alternative to ordinary Portland cement (OPC).

1. Introduction

The increase in greenhouse gases like CO₂ in the atmosphere is a leading cause of global warming [1]. The production of concrete constitutes a major source of global CO₂ emissions because of cement manufacturing involving the calcination of limestone and significant energy consumption, both of which release large amounts of CO₂ [2–5]. Ordinary Portland cement (OPC) has been widely used in concrete owing to its availability, low cost, and reliable performance; however, its production is energy-intensive and has significant environmental impacts [1,6]. The production of OPC is responsible for approximately half of worldwide CO₂ emissions caused calcination of limestone and combustion of fossil fuels [1,7]. The global annual cement consumption was estimated to be 4.4 billion tons in 2024 and was expected to reach nearly 5.4 billion tons in 2030 [8], while producing one ton of OPC

typically generates approximately 0.87–1.00 tons of CO₂ [9,10]. Therefore, reducing CO₂ emissions during OPC and concrete production has become a core issue in mitigating the global climate crisis [11–13].

Several approaches have been proposed and developed to reduce CO₂ emissions from the cement industry, including (1) the use of alternative fuels and raw materials, (2) the utilization of supplementary cementitious materials (SCMs) for clinker substitution, (3) the implementation of carbon capture, utilization, and storage (CCUS) technologies, (4) the development of alternative low-carbon binders, (5) the enhancement of material efficiency in construction practices, and (6) the promotion of waste reuse and recycling [14–18]. Among these approaches, clinker substitution is one of the most promising strategies for reducing CO₂ emissions [16,17,19]. Clinker substitution can be implemented through one or a combination of three approaches: (1) replacing raw materials, (2) utilizing SCMs and (3) modifying clinker formulation levels [17,18].

Abbreviations: OPC, Ordinary Portland cement; CO₂, Carbon dioxide; C₂S, Dicalcium silicate; BRC, Belite-rich cement; C₃S, Tricalcium silicate; C–S–H, Calcium silicate hydrate; CH, Calcium hydroxide; CS, Compressive strength; LP, Limestone powder; ITZ, Interfacial transition zone; W/b, Water-to-binder ratio; SEM, Scanning electron microscopy; EDS, Energy-dispersive X-ray spectroscopy; NI, Nanoindentation; H, Hardness; E, Young's modulus; FMZ, Fiber–matrix interfacial zone.

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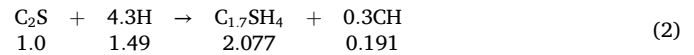
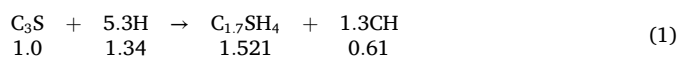
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In this study, a combination of approaches (2) and (3) was employed to produce concrete with reduced CO₂ emissions while maintaining satisfactory mechanical performance.

Low-carbon cement with higher dicalcium silicate (C₂S) content, i.e., belite-rich cement (BRC), is one of the most promising alternatives for OPC to reduce CO₂ emissions [20,21]. Cements with higher C₂S contents generally have lower tricalcium silicate (C₃S) contents. C₂S forms at temperatures above 1100 °C [20,21], while C₃S is generated at higher temperatures of 1300 °C. C₂S can be produced at temperatures 100–300 °C below those necessary for C₃S formation, which corresponds to an approximate 16% reduction in energy consumption [20,22]. Therefore, the CO₂ emissions associated with C₂S production are about 17% lower than those of C₃S [11,13]. Gartner [23] reported that BRC could reduce CO₂ emissions by 10% compared with OPC. Therefore, BRC with high C₂S was selected to develop sustainable concrete with low CO₂ emissions.

However, the practical applications of BRC have been limited [20] because BRC produces lower early strength than OPC owing to the slower hydration rate of C₂S than that of C₃S [20,22,24]. Therefore, it is necessary to accelerate the early hydration and strength development of BRC-based concrete. The incorporation of SCMs has been proposed to enhance both the early-age reactivity and mechanical properties of BRC systems [25]. In addition, the replacement of un-hydrated cement with SCMs, particularly in high-strength or ultra-high-performance concrete, is effective not only in decreasing CO₂ emissions but also in reducing the costs associated with the production of high-strength concrete [19,26]. Among SCMs, limestone powder (LP) is one of the most promising options owing to its very low embodied CO₂ emissions [27] and greater accessibility compared to other SCMs, such as fly ash and blast furnace slag [28,29]. In addition, LP enhances the hydration and strength development of concrete through its nucleation, filler, and chemical effects [30–34]. Therefore, LP was selected in this study to partially replace BRC, not only to reduce the embodied CO₂ but also to increase the early strength of BRC-based concrete.

Fiber-reinforced concrete (FRC) has received considerable attention in civil engineering [35] owing to its enhanced tensile strength, toughness, and energy absorption capacity derived from fiber bridging effects [35,36]. The interfacial bond between fibers and the matrix is a key factor governing the mechanical behavior of FRC [37–39]. To improve this interfacial bond, various approaches have been employed, including the use of finer grains and deformed fibers [37,38] and/or incorporating nanoparticles [39,40]. The use of finer grains improved particle packing density around the fiber surface, resulting in a denser interfacial transition zone. This densification increased both frictional resistance and chemical adhesion at the fiber–matrix interface [37,38]. Deformed fibers enhanced interfacial bonding primarily through mechanical anchorage. In addition, nanoparticles acted as fillers and nucleation sites for hydration products, refining the microstructure at the fiber–matrix interface. Their high surface area of nanoparticles accelerated hydration and reduced interfacial porosity, leading to a stronger and more cohesive interfacial zone [39,40]. Dang and Kim [39] recently reported that the addition of nanoparticles significantly enhanced the interfacial bond strength owing to increased formation of C-S-H in the fiber–matrix interfacial zone (FMZ). The quantity of C-S-H in the FMZ is a critical factor influencing fiber pullout performance. Compared to OPC, BRC containing higher C₂S levels generates a greater amount of C-S-H due to the full hydration of C₂S, producing approximately 26.77% more C-S-H than C₃S, as indicated in Eqs. (1) and (2) [41]. Consequently, the use of BRC is expected to improve the interfacial bond strength between the fibers and mortar matrix while simultaneously lowering the carbon footprint by replacing OPC.



This study investigated the effects of C₂S and LP contents on the interfacial bond characteristics of high-strength steel fibers embedded in low-carbon mortar matrices. The specific objectives were as follows: (1) to investigate the effects of C₂S and LP contents on the development of the compressive strength of low-carbon mortar matrices; (2) to reveal the effects of C₂S and LP contents on the pullout resistance of high-strength steel fibers; (3) to correlate the development of the FMZ and fiber pullout resistance corresponding to the contents of C₂S and LP; and (4) to explore the optimized composition based on environmental evaluation.

2. Experiments

Fig. 1 shows the experimental setup designed to investigate the pullout resistance of high-strength steel fibers embedded in low-carbon mortar incorporating BRC and LP. Two types of steel fiber were investigated: smooth (S) and hooked (H), while four different matrices with different C₂S and LP contents but the same water-to-binder ratio of 0.4 were used. Single-fiber pullout and compressive tests were conducted at ages of 14, 28, and 90 days to explore the development of fiber pullout resistance and compressive strength of the mortar matrices over 90 days. Furthermore, the microstructures and hardness of the FMZ, corresponding to matrices with different C₂S and LP contents, were investigated using SEM-EDS and nanoindentation (NI) tests at 90 days.

2.1. Materials and specimen preparation

2.1.1. Materials

Table 1 lists the physical and chemical properties of OPC (KS L 5201) and BRC while Fig. 2 shows their particle size distribution (PSD). In this study, the PSDs of BRC and LP were measured using “an LA-960 laser scattering particle size distribution analyzer”, whereas that of OPC was obtained from [39]. Both OPC and BRC were supplied by Ssangyong Cement Industry Co., Ltd., while LP was provided by Daejung Chemical Co., Ltd. Table 1 shows that OPC contained 58.3% C₃S, more than double the 26.8% present in BRC. In contrast, the C₂S proportion (16.4%) in OPC was significantly lower than that (57.2%) in BRC.

Table 2 lists the properties of S and H fibers, while Fig. 3 shows their images. The tensile strength was 2580 and 2311 MPa for S and H fibers, respectively, while the elastic modulus of both fibers was 200 GPa. The length of both fibers was 30 mm, whereas their diameters were 0.3 and 0.375 mm, respectively. As shown in Fig. 3, both fibers were brass-coated.

In this study, four different mortar matrices were prepared corresponding to the compositions listed in Table 3. As shown in Table 3, M1 employed only OPC (KS L 5201), while M2 employed only BRC. Both M3 and M4 also used BRC but contained 10% and 30% LP, respectively. All matrices used silica sand with a grain diameter range of 0.1–0.3 mm with an average of 0.25 mm.

Table 4 presents the mass proportions of each phase in the matrices. In real cements, the calcium sulfate is typically added to control the rapid hydration of C₃A [43,44]. However, calcium sulfate is not presented in Table 4 because the phase compositions were determined by the Bogue equations [42,45]. The Bogue equations are well-known for estimating the compound composition of the four main clinker phases based on the oxide chemical composition; however, they do not account for calcium sulfate. Consequently, the compositions of the four main phases do not come out to 100%. M1 and M2 were prepared to examine the effects of C₂S content, while M2, M3, and M4 were used to investigate the effects of LP content on the pullout resistance of steel fibers. The contents of C₂S were 16.4% in M1 and 57.2% in M2. The contents of LP in M2, M3, and M4 were 0%, 10%, and 30%, respectively. As the LP content increased, i.e., cement content decreased, the cement clinker

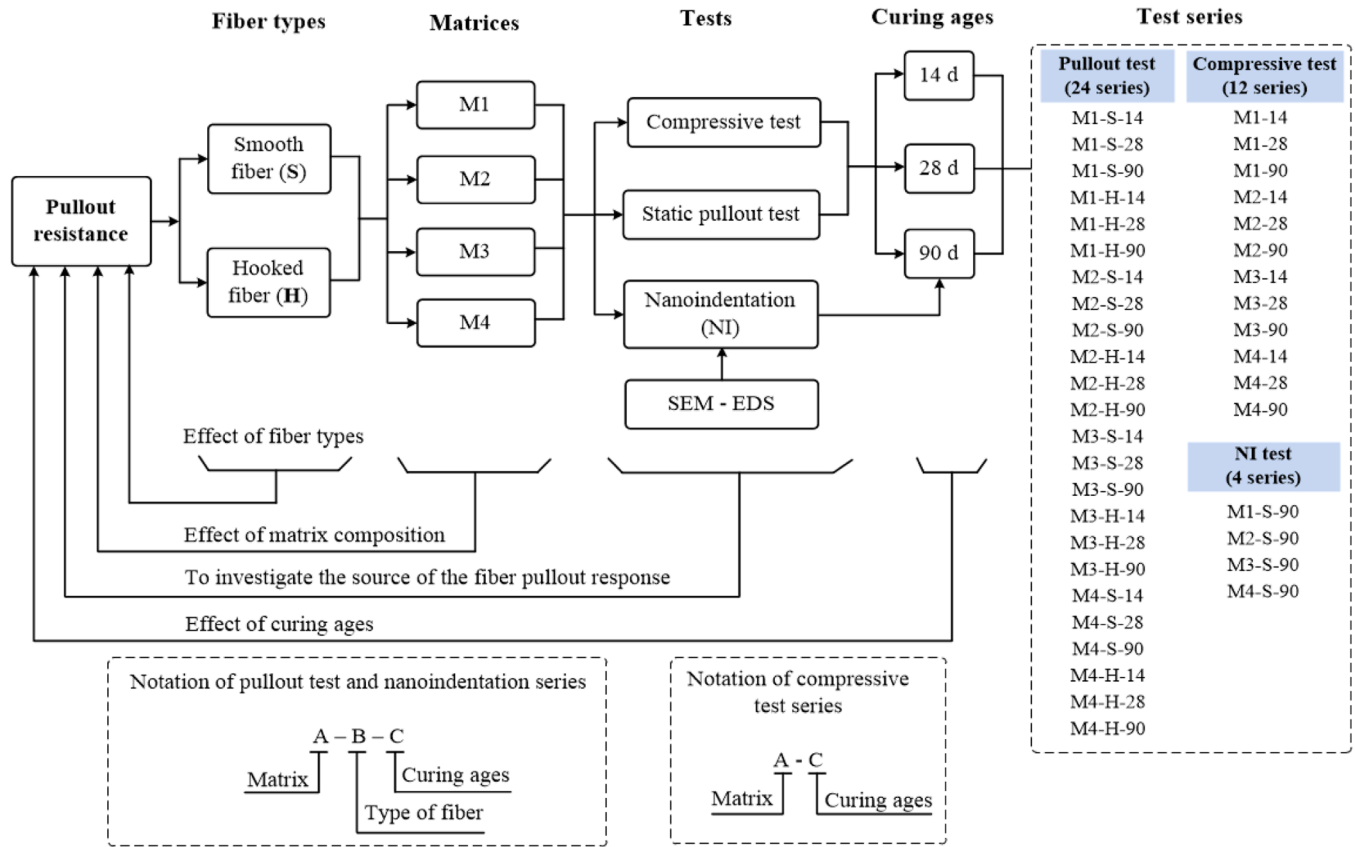


Fig. 1. Experimental program.

Table 1

Properties of the binders.

	OPC [42]	BRC [18,42]
Compound composition (%)		
C ₃ S	58.3	26.8
C ₂ S	16.4	57.2
C ₃ A	8.3	1.40
C ₄ AF	9.7	8.70
Density (g/cm ³)	3.14	3.21

Table 2

Properties of steel fibers.

Name (Notation)	Diameter (mm)	Length (mm)	Density (g/cm ³)	Tensile strength (MPa)	Elastic modulus (GPa)
Smooth (S)	0.3	30	7.9	2580	200
Hooked (H)	0.375	30	7.9	2311	200

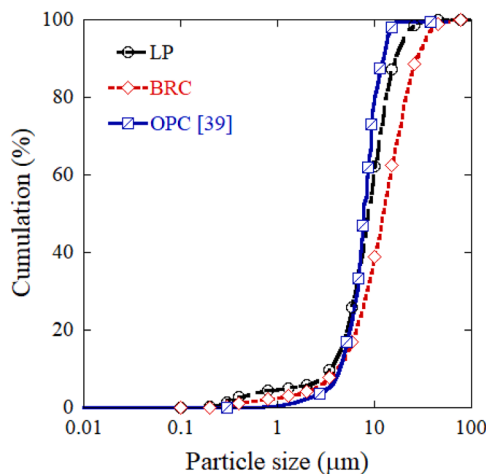


Fig. 2. Particle size distribution of materials.

phases were reduced. For instance, the C₃S content decreased from 26.8% to 18.8%, while C₂S content reduced from 57.2% to 40.0% as LP content increased from 0% to 30% (M2–M4), as shown in Table 4.

2.1.2. Specimen preparation

The mortar matrix was mixed using a 20-L Hobart-type mixer. Raw binders such as cement, silica sand, and limestone powder were first dry-mixed for 3 min. Subsequently, water was gradually added while the mixer continued to operate for an additional 3 min [40]. For the single-fiber pullout tests, bell-shaped pullout specimens were used to measure the pullout load versus slip response of fibers in the four mortar matrices, as shown in Fig. 4a. Before casting, the steel fibers were pre-installed in the molds to maintain the embedded length at half of the fiber length (15 mm). For the compressive strength tests, cubic specimens with dimensions of 50 × 50 × 50 mm³ were prepared to determine the compressive strength of plain the mortar matrix without fibers, as shown in Fig. 4b.

For the NI test, cylindrical specimens with diameters of 25 mm and heights of 35 mm were prepared. Prior to casting, nine S steel fibers were vertically fixed in the middle of the mold using a foam sheet [40], as shown in Fig. 4c. After preparing all clean molds, the mortar gradually poured into the molds. During casting, the specimens were vibrated slightly to minimize pores. After casting, the specimens were covered

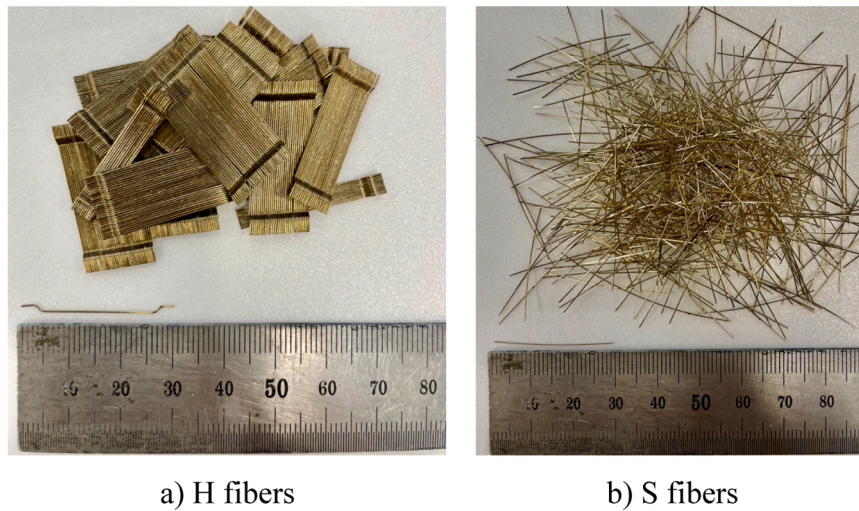


Fig. 3. Images of steel fibers.

Table 3

Matrix composition by weight ratio of cement and LP.

Mixture type	OPC (KS L 5201)	BRC	LP	Silica sand	Water-to-binder ratio (w/b)
M1	1.0	0.0	0.0	1.5	0.4
M2	0.0	1.0	0.0	1.5	0.4
M3	0.0	0.9	0.1	1.5	0.4
M4	0.0	0.7	0.3	1.5	0.4

Table 4

Mass proportions of each phase in matrices.

Mixture ID	C ₃ S (%)	C ₂ S (%)	C ₃ A (%)	C ₄ AF (%)	LP (%)
M1	58.3	16.4	8.3	9.7	0
M2	26.8	57.2	1.4	8.7	0
M3	24.1	51.5	1.3	7.8	10
M4	18.8	40.0	1.0	6.1	30

with plastic sheets and maintained at 20 °C for 24 h prior to demolding. After demolding, the specimens were water-cured at 20 °C for 14, 28, and 90 days until testing. The pullout and compressive specimens were tested at 14, 28, and 90 days under dry conditions, whereas NI specimens were only tested at 90 days.

For pretreatment of the NI and SEM-EDS specimens, the procedure for cutting, grinding, and polishing was as follows: The specimens were

stored in a water tank until the specific age of 90 days, after which they were removed from the tank. Subsequently, these specimens were immersed in acetone for 24 h and stored in a vacuum drying oven at 40 °C for 48 h to prevent further cement hydration. Subsequently, all specimens were sliced to a thickness of 10 mm using “a precision high-speed saw”. The slices were then impregnated with low-viscosity epoxy resin and cured under vacuum to prevent cracking and delamination during abrasive preparation [18,40]. Then, the slices were grinded and polished using “MetPrep 3™ with a PH-4™ power head” [18]. Then, the specimens were followed by final polishing with a cloth containing a 3-μm diamond suspension. During the polishing process, PurpleLube and water-free alcohol-based lubricants were used to enhance the polishing performance [18,40].

2.2. Test setup, procedure, and parameters describing the test results

2.2.1. Single-fiber pullout test

Fig. 5 shows the testing machine and setup used to conduct single-fiber pullout tests. The single-fiber pullout tests were conducted at a constant speed of 1 mm/min. During testing, fiber slip was measured using “a linear variable differential transformer” (LVDT), and the pull-out load was recorded by the load cell attached to the top of the machine cross-head, as shown in Fig. 5. The gap between the fiber grip and specimen should be minimized to eliminate the effects of fiber elongation on the measured slip of fiber from the LVDT.

Fig. 6 illustrates the parameters used to characterize the interfacial

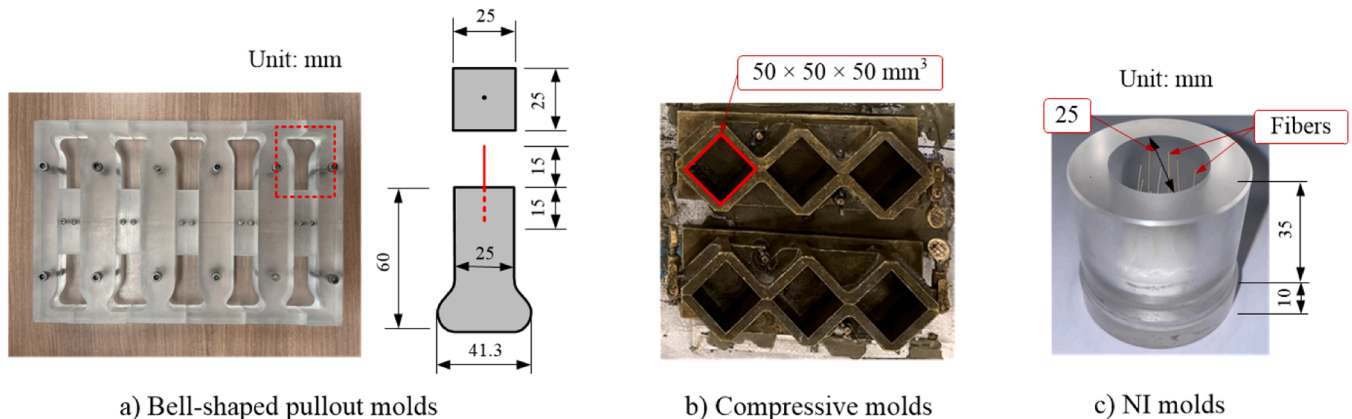


Fig. 4. Geometry of molds.

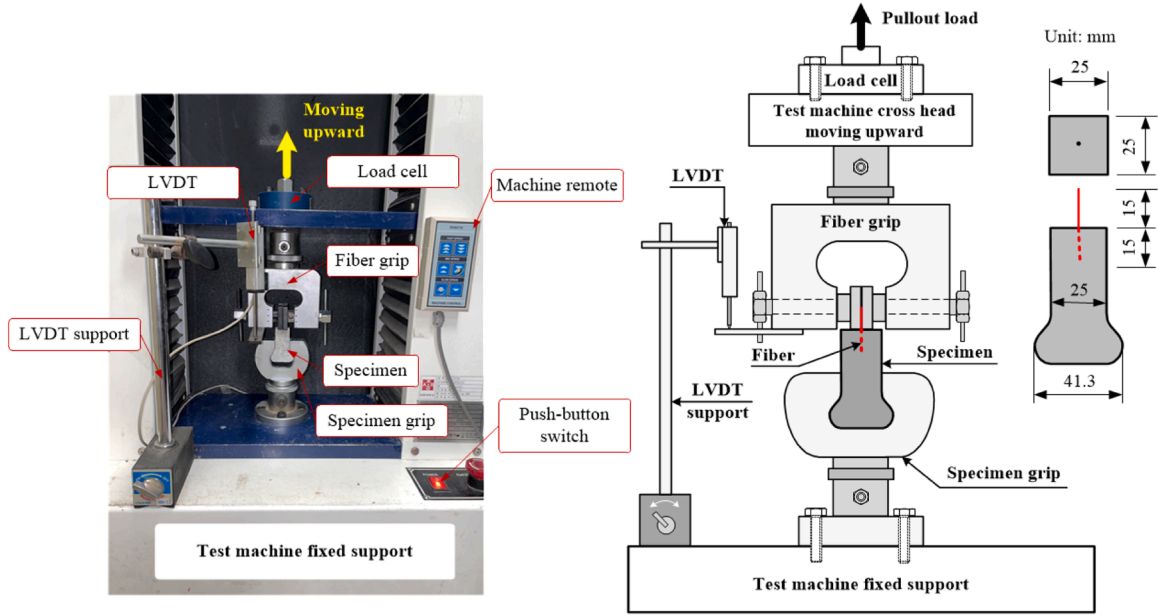


Fig. 5. Pullout test machine.

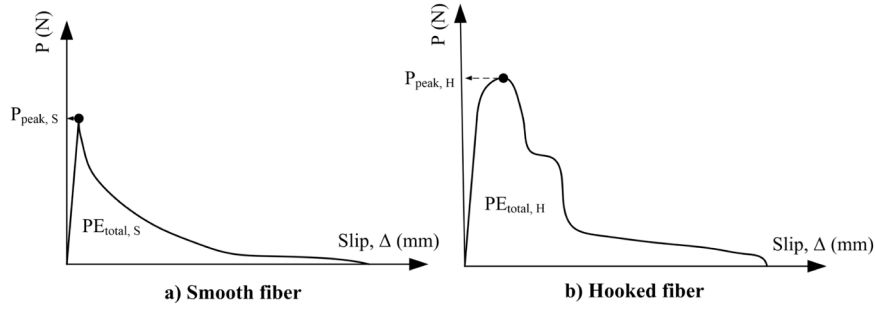


Fig. 6. Pullout load versus slip curves of S and H fibers and their pullout parameters.

bond strengths of the S and H steel fibers. The pullout specimens had a cross-sectional area of $25 \times 25 \text{ mm}^2$, whereas steel fibers were embedded at the center of the specimens with a length of 15 mm. The peak pullout load was identified as P_{peak} , as shown in Fig. 6. In addition, the peak bond strength (τ_{peak}) was determined using Eq. (3), whereas the peak tensile stress of the fiber (σ_{peak}) was calculated using Eq. (4) [39, 46].

$$\tau_{peak} = \frac{P_{peak}}{\pi \times d_f \times L_{em}} \quad (3)$$

$$\sigma_{peak} = \frac{4 \times P_{peak}}{\pi \times d_f^2} \quad (4)$$

where P_{peak} is the peak pullout load, d_f is the diameter of steel fiber, as shown in Table 2, and L_{em} is the embedded length of steel fiber in the matrix. In this study, L_{em} was 15 mm.

To further quantify the pullout resistance of steel fiber, the pullout energy (PE_{total}) and equivalent bond strength (τ_{equi}) were also calculated. PE_{total} was calculated based on the area under the pullout load versus slip curve, as shown in Fig. 6, whereas τ_{equi} was determined based on PE_{total} using Eq. (5) [39,46].

$$\tau_{equi} = \frac{2 \times PE_{total}}{\pi \times d_f \times L_{em}^2} \quad (5)$$

2.2.2. Compressive strength test

Fig. 7 shows the universal testing machine (UTM), with a capacity of 300 kN, used to determine the compressive strength (CS) of the mortar matrices in Table 3. The CS of each series was evaluated at 14, 28, and 90 days. For each series, six specimens were tested under displacement control at a constant rate of 1 mm/min. The average CS value was calculated from at least three measurements.

2.2.3. NI test

In this study, the NI test was conducted using the TI 950 triboindenter system with a load capacity of 10 mN. Fig. 8a shows a typical indentation load (P) versus penetration depth (h) curve, while Fig. 8b illustrates a schematic representation of an indentation test. As shown in Fig. 8a, the typical $P-h$ curve shows an initial constantly increasing load, followed by a short hold and then a constantly decreasing load. Based on the $P-h$ curve, the hardness (H) and Young's modulus (E) of the microstructure within the FMZs was calculated according to Refs. [39,40,47], as shown in Eqs. (6) and (7), respectively:

$$H = \frac{P_{max}}{A_c} \quad (6)$$

$$E = (1 - \nu^2) \left(\frac{1}{E_r} - \frac{1 - \nu_i^2}{E_i} \right)^{-1} \quad (7)$$

where P_{max} is the peak of indentation load (Fig. 8a), and A_c is the projected contact area which can be extrapolated from maximum

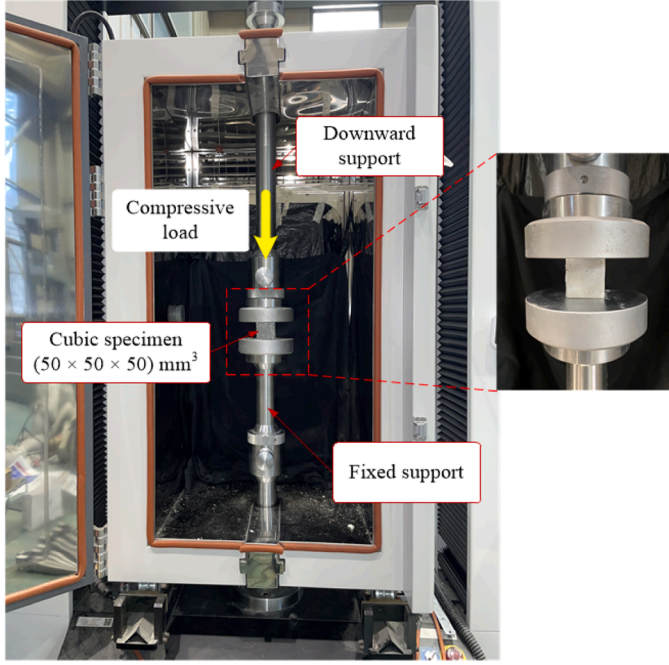


Fig. 7. Universal testing machine.

penetration depth ($h_{\max}$) using the Oliver and Pharr method [47]. ν is Poisson's ratio of the test samples, whereas ν_i and E_i are the Poisson's ratio and Young's modulus of the indenter, respectively. In this study, values of $\nu_i = 0.07$ and $E_i = 1140$ GPa were used for the Berkovich indenter [40,47]. E_r is the reduced modulus, which was calculated using Eq. (8):

$$E_r = \frac{\sqrt{\pi}}{2\beta} \frac{S}{\sqrt{A_c}} \quad (8)$$

where β is the correction factor with a value of 1.034 for the Berkovich indenter [40], and S is the unloading indentation stiffness calculated from the initial slope of the elastic unloading curve at $h_{\max}$, as shown in Fig. (8a).

Fig. 9 shows the NI test setup. Two FMZs were evaluated for each fiber, as shown in Fig. 9a. Each zone contained 40 indentation points arranged in 4 rows and 10 columns, with a consistent spacing of $10 \mu\text{m}$ between rows and columns [39,40]. In each row, two indentation points were in the steel-fiber section, whereas the eight remaining points were in the matrix (Fig. 9a). Consequently, 80 indentation points were assessed for each fiber. Additionally, three fibers were randomly selected from the nine fibers in each specimen. Therefore, 240 indentation points were evaluated for each specimen.

Fig. 9b shows the indentation load versus time curve used to conduct the NI test in this study, following Refs. [37,49,51]. The loading regime was divided into three periods: loading, holding, and unloading. The load increased from 0 to 2 mN over 10 s with a constant rate of 0.2 mN/s (the loading period). The load was then maintained for 5 s (the holding period). Finally, the load decreased from 2 to 0 mN over 10 s (the unloading period), as shown in Fig. 9b.

2.2.4. SEM-EDS

In this study, a Hitachi SU8010 scanning electron microscope was used to conduct SEM-EDS analysis. After the NI measurements, the specimens were analyzed to determine the chemical composition at the indentation points. Prior to the SEM-EDS analysis, the specimens were coated with a thin layer of carbon to enhance their electrical conductivity. Image analysis was conducted using signals from secondary electron-sample interactions to examine the surface topology, with an accelerating voltage of 15 kV.

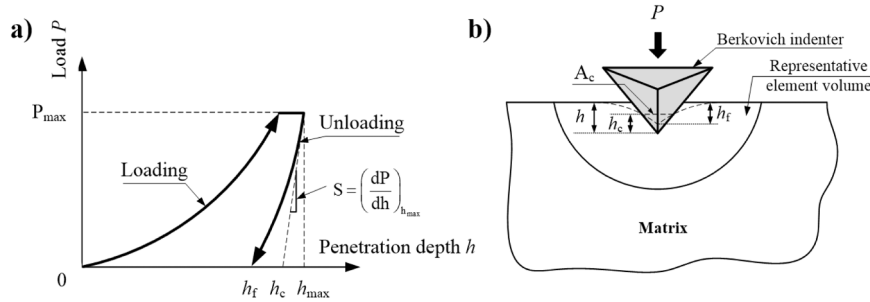


Fig. 8. (a) A typical representation of P-h curve [40,48]; (b) Schematic representation of an indentation test [48–50].

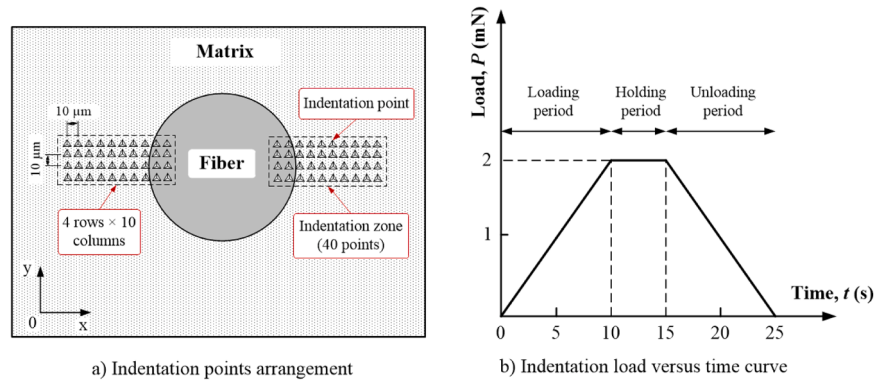


Fig. 9. Setup of the NI test: a) indentation point arrangement adapted from [39,40], and b) indentation load versus time curve adapted from [37,49,51].

3. Experimental results and analysis

3.1. Compressive strength

Fig. 10 shows the CS of mortar matrices (M1, M2, M3, and M4) with different contents of C₂S and LP during curing. In this study, the compressive tests were conducted at three specific curing ages of 14, 28, and 90 days. The specimens were cured in water at 20 °C until testing. M1 was prepared using OPC, whereas M2, M3, and M4 were prepared using BRC. In addition, M2, M3, and M4 were replaced with 0, 10, and 30% LP, respectively. As shown in Fig. 10, the CS increased at longer curing ages. For instance, the CS of M1 at 90 days increased from 57.6 MPa at 14 days to 69 MPa. The increase in CS with longer curing days can be attributed to greater quantity of hydration products in the later ages than in the early ages [18,45,52]. Consequently, the amount of free water was reduced as it was replaced by hydration products, leading to denser microstructures over time [53,54]. Moreover, the hydration products, particularly C-S-H, evolved into increasingly stable and rigid morphologies over time, which was also responsible for the higher CS at a later age than at an earlier age [18].

Although mortars incorporating BRC exhibited lower CS at 14 days compared to those made with OPC, these mortars demonstrated higher CS at 28 and 90 days, as shown in Fig. 10. Specifically, after 14 days of curing, the CS of M1 was 57.6 MPa, compared with 49.6 MPa for M2. In contrast, after 28 days, the CS values for M1 and M2 were 61.3 and 71.6 MPa, respectively. In addition, the CS of mortar matrices decreased with increasing LP content. For example, the CS at 28 days decreased from 71.6 MPa to 67.6 MPa, corresponding to LP's contents increasing from 0 (M2) to 30% (M4). However, the reduction in CS at 10% LP substitution (M3) was less pronounced than that at 30% substitution (M4) when compared with the control mixture (M2) without LP. Specifically, the 90-day CS of M3 decreased by only 7.1%, whereas that of M4 decreased by 29.8% compared to M2. Moreover, the mortar with 10% LP substitution exhibited a higher CS at 14 days than the mixture without LP. The 14-day CS of M3 was 55.7 MPa, whereas that of M2 was 49.6 MPa.

3.2. Pullout resistance

3.2.1. Pullout load (stress) versus slip curves of high-strength fibers

Fig. 11 illustrates the average pullout stress-slip curves for four different mortar matrices (M1, M2, M3, and M4) at curing ages of 14, 28,

and 90 days. The individual pullout load versus slip curves for each specimen, along with the methodology used to determine the average curves, are presented in Appendix A. In each subfigure, the left y-axis represents the pullout load, while the right y-axis shows the corresponding fiber stress calculated from the pullout load. Additionally, the figures in the left column show the pullout behavior of S fibers, whereas those in the right column show that of H fibers.

As shown in Fig. 11, the peak pullout load (P_{peak}) of S fibers was significantly lower than that of H fibers. In addition, the pullout curves of S fibers exhibited a continuous descending branch, whereas those of H fibers displayed a secondary peak, as clearly shown in Fig. 11. Moreover, the slip at P_{peak} was considerably smaller for S fibers (Figs. 11a, 11c, and 11e), and their descending branch (after peak load) exhibited a decrease at a faster rate than that of H fibers (Figs. 11b, 11d, and 11f). These observations are consistent with experimental data from [55–57]. The differences in P_{peak} and the descending branch between the two fibers can be attributed to the fact that S fibers are based on the elastic adhesive-type and frictional bonding for resistance, whereas H fibers benefit from the mechanical interaction between the fiber and matrix provided by the hooked end [38,55,56]. During pullout, the hooked end was straightened to be pulled out, thereby increasing pullout load and reducing the slope of the descending branch [56]. With increasing slip, the pullout load decreased owing to the progressive decay of frictional forces and the reduction of embedded length in both fiber types; however, in H fibers, the partially straightened end continued to provide mechanical anchorage in addition to friction, a mechanism absent in S fibers.

In addition, the pullout load increased with curing age, with the peak load after 90 days being higher than those at 28 and 14 days. This improvement is attributed to the higher CS at later ages, as shown in Section 3.1. More hydration at later ages enhanced matrix densification and interfacial bond strength. For S fibers, the slope of the ascending branch (prior to peak load) was highest in M1 at 14 days (Fig. 11a), whereas similar slopes were observed across all matrices at 28 days (Fig. 11c). At 90 days, M1, M2, and M3 exhibited comparable ascending slopes, whereas M4 displayed the lowest slope (Fig. 11e), likely owing to the reduced cement content at high LP substitution (30%). The descending branch of S fibers showed negligible differences across all matrices at 14 and 28 days (Figs. 11a and 11c); however, at 90 days, clear discrepancies occurred as C₂S content decreased, following the order M2 > M3 > M4 > M1 (Fig. 11e).

A higher pullout load was observed in M2-S-90 at a slip of 12–15 mm

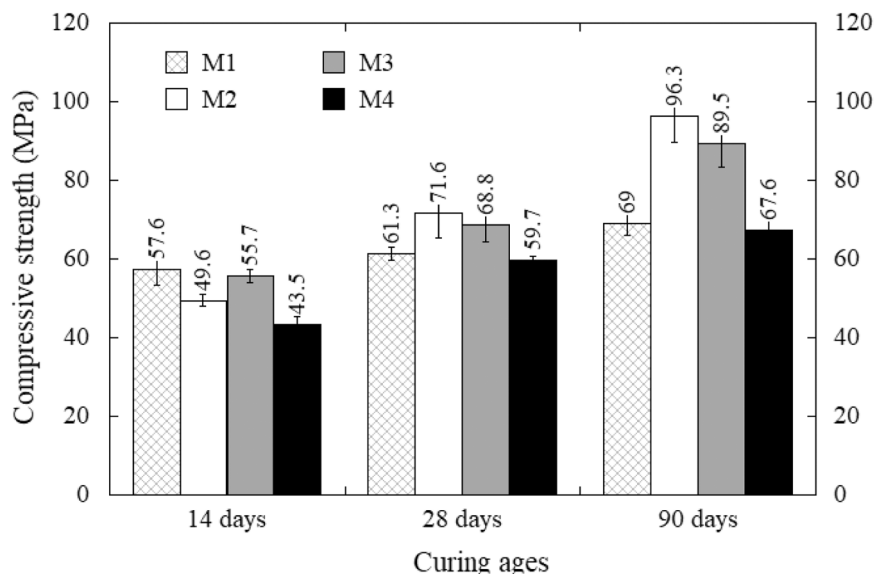


Fig. 10. CS of various matrices at different curing ages.

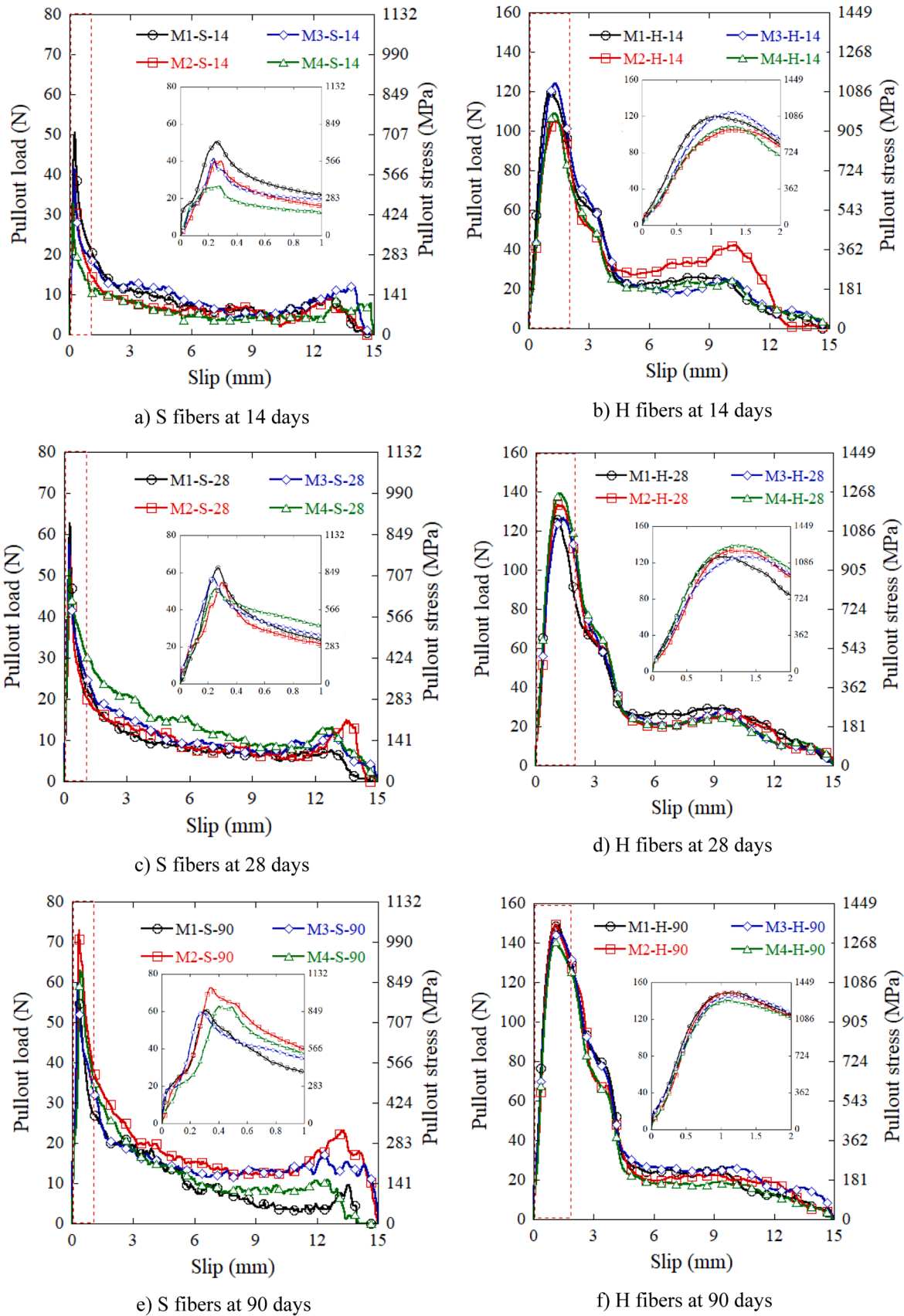


Fig. 11. Pullout load versus slip curves according to various matrices at different curing ages.

compared to M1-S-90, whereas the pullout load within the same slip range decreased with increasing LP substitution ($M2-S-90 > M3-S-90 > M4-S-90$), as shown in Fig. 11e. This phenomenon can be attributed to the deformation of the fiber ends generated during the cutting process (the flattening effect) [58–61]. The flattened fiber ends act as mechanical anchors and induce scratching along the matrix tunnel, thereby increasing the pullout load at a slip of 12–15 mm.

Moreover, the flattening effect progressively increased with curing age in the BRC-based mortar ($M2-S-14 < M2-S-28 < M2-S-90$), whereas it remained almost constant in the OPC-based mortar (M1). The pronounced flattening effect in M2, which became significantly higher than that in M1 at 90 days, can be attributed to the greater accumulation of hydration products—particularly C-S-H—that developed over time in M2 and exceeded those in M1 at later ages [18,45,52]. In addition, BRC-based paste exhibited a finer pore structure than OPC-based paste [53,54], resulting in a denser microstructure in M2 compared with M1. Consequently, the FMZs in M2 became progressively stronger with curing age. This observation is also supported by the compressive strength results, which increased markedly with time in M2 and were significantly higher than those in M1 at 90 days, as shown in Fig. 10. Furthermore, the reduction in flattening effect with increasing LP content is because of the lower cement content associated with higher LP substitution. The reduced cement content led to fewer hydration products, such as C-S-H, thereby producing a less compact microstructure at higher LP levels.

In contrast, the pullout behavior of H fibers was less sensitive to matrix composition (Figs. 11b, 11d, and 11f) because the pullout resistance of H fibers was mainly due to the mechanical interaction between the fiber and matrix. At an age of 14 days, M1 exhibited a slightly higher slope in the ascending branch compared to the other matrices, while no significant differences were observed among the matrices at 28 and 90 days. Moreover, the descending branches of H fibers were well-aligned across all matrices corresponding to the same curing age. Consequently, the matrix differences were mainly reflected in the peak load rather than in the trend of the ascending or descending branches in H fibers, as shown in Fig. 11f.

3.2.2. P_{peak} , τ_{peak} , and σ_{peak}

Figs. 12, 13, and 14 show the peak pullout loads (P_{peak}), peak bond strength (τ_{peak}), and peak tensile stress of fibers (σ_{peak}), respectively, examined at 14, 28, 90 days for both S and H fibers embedded in four different mortar matrices. As illustrated in the figures, P_{peak} , τ_{peak} , and σ_{peak} exhibited similar trends because τ_{peak} and σ_{peak} were directly derived from P_{peak} . The results showed that the P_{peak} of S fibers was

consistently lower (approximately 2–3 times) than that of H fibers. The lower P_{peak} of S fibers than H fibers is because their bond resistance is based on the elastic adhesive-type and friction mechanisms at the fiber–matrix interface, whereas H fibers gain an additional contribution from mechanical anchorage provided by the hooked end. In addition, the H fibers had a larger diameter (0.375 mm) compared to S fibers (0.3 mm), resulting in a greater interfacial area with the surrounding matrix. Consequently, H fibers exhibited significantly higher bond resistance than S fibers.

The matrix with lower C_2S content produced higher P_{peak} , τ_{peak} , and σ_{peak} values after 14 days of curing, as shown in Figs. 12, 13, and 14. For example, as C_2S contents increased from 16.4% in M1–57.2% in M2, the P_{peak} , τ_{peak} , and σ_{peak} at 14 days decreased from 56.4 to 47.7 N (15.4%), 4.1–3.6 MPa (12.2%), and 798–675.3 MPa (15.4%), respectively. In addition, although lower P_{peak} , τ_{peak} , and σ_{peak} were observed in matrices with higher C_2S content at 14 days, significantly higher values were obtained at 90 days. These results are clearly shown for S fibers. As the C_2S content increased from 16.4% (M1) to 57.2% (M2), the P_{peak} , τ_{peak} , and σ_{peak} of S fibers at 90 days increased, on average, from 64.6 to 85.1 N (37.7%), 4.8–6.2 MPa (29.2%), and 913.2–1204.2 MPa (31.9%), respectively.

Although higher LP contents reduced pullout performance, the substitution of 10% LP yielded comparable results. As shown in Fig. 12, the P_{peak} of the S fiber at 14 days decreased from 47.7 N for BRC without LP (M2) to 46.6 N for 10% LP (M3) and further to 33.6 N for 30% LP (M4). The higher pullout performance of matrices with lower LP can be attributed to their higher cement content, which produced more hydration products. Furthermore, Fig. 14 shows that the peak stresses of both S and H fibers were much lower than their tensile strengths (2580 and 2311 MPa, respectively), indicating that no fiber rupture occurred during testing.

3.2.3. PE_{total} and τ_{equi}

Figs. 15 and 16 show the pullout energy (PE_{total}) and equivalent bond strength (τ_{equi}), respectively, examined at 14, 28, 90 days for both S and H fibers embedded in matrices M1, M2, M3 and M4. Similar trends were observed between PE_{total} and τ_{equi} because τ_{equi} values were derived from PE_{total} . The results show that substituting 10% LP into BRC (M3) produced higher PE_{total} and τ_{equi} than OPC (M1) across all curing ages, even at 14 days for S fibers. For S fibers, PE_{total} values of M1 and M3 were 126.8 and 143.8 N·mm at 14 days, 144.0 and 177.6 N·mm at 28 days, and 175.7 and 227.1 N·mm at 90 days, respectively. These results confirm the beneficial role of LP in improving both PE_{total} and τ_{equi} .

At early ages, increasing LP content notably enhanced fiber pullout

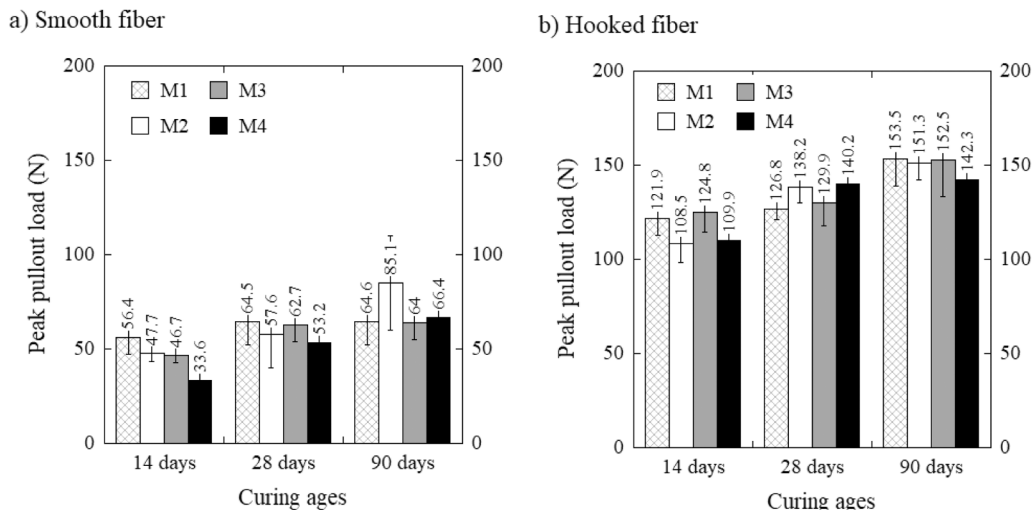
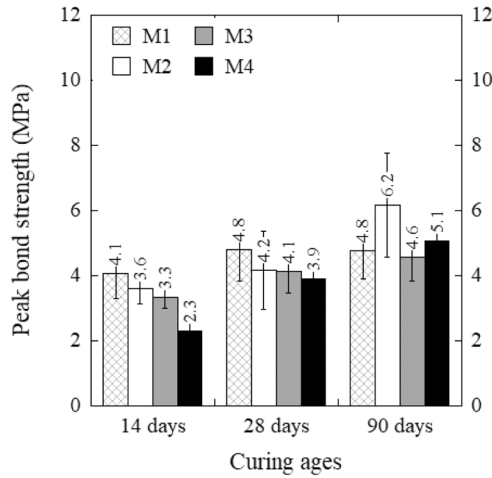


Fig. 12. Peak pullout loads.

a) Smooth fiber



b) Hooked fiber

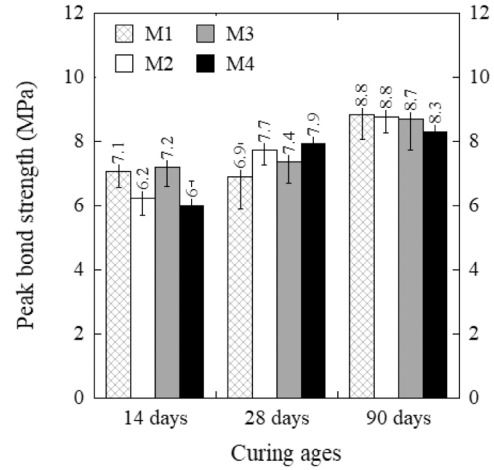
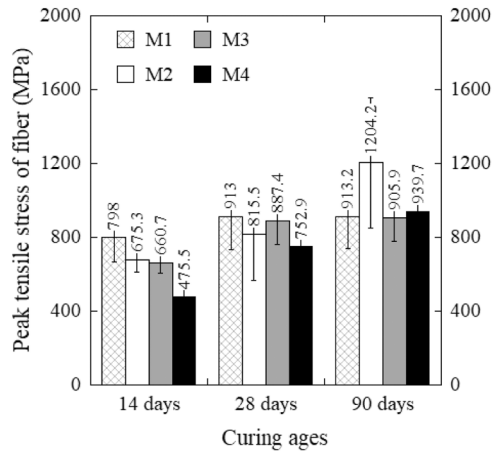


Fig. 13. Peak bond strength.

a) Smooth fiber



b) Hooked fiber

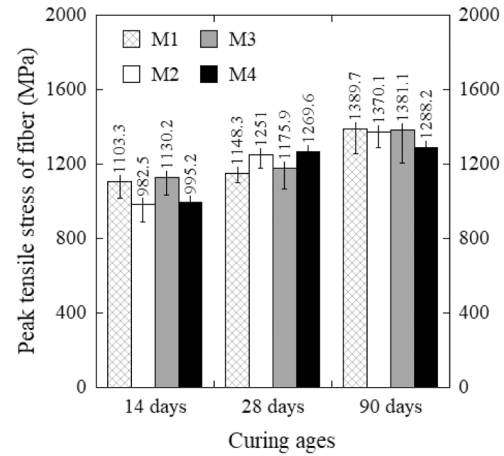
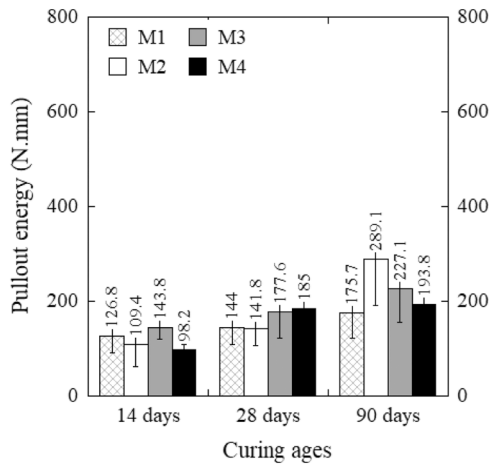


Fig. 14. Peak tensile stress of fibers.

a) Smooth fiber



b) Hooked fiber

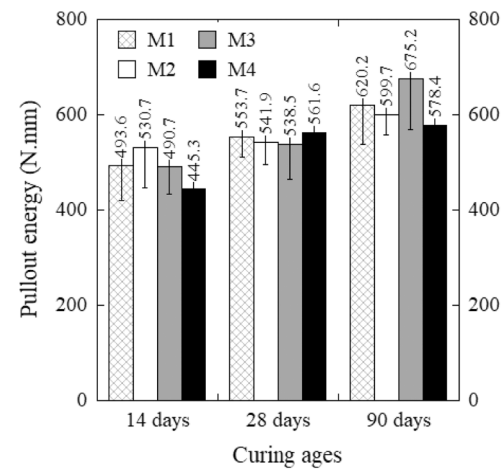


Fig. 15. Pullout energy.

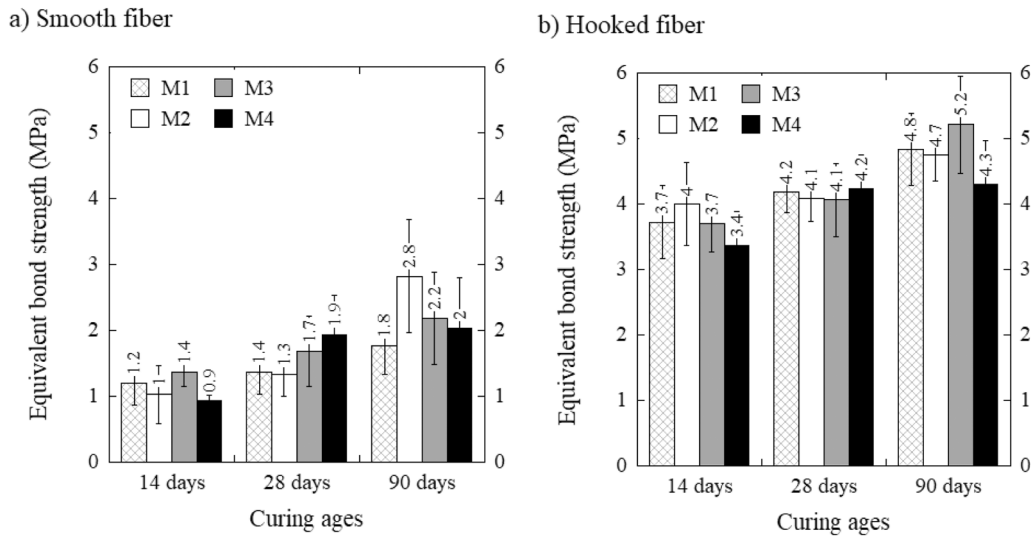


Fig. 16. Equivalent bond strength.

resistance: as LP increased from 0% to 30%, PE_{total} at 28 days increased from 141.8 to 185.0 N-mm and τ_{equi} from 1.3 to 1.9 MPa (Figs. 15a and 16a). However, at later ages, the trend was reversed: matrices with lower LP (and thus higher C_2S) showed superior values. At 90 days, PE_{total} and τ_{equi} decreased from 289.1 to 193.8 N-mm and 2.8–2.0 MPa, respectively, as LP increased from 0% to 30%. These results indicate that LP mainly contributed at early ages through nucleation effects, whereas C_2S hydration dominated long-term performance.

Furthermore, Figs. 15 and 16 show that M1 (16.4% C_2S) and M2 (57.2% C_2S) exhibited only slightly different PE_{total} and τ_{equi} at 14 and 28 days; however, at 90 days, the matrix with higher C_2S content (M2) produced significantly greater values for S fibers compared to M1. This improvement was due to the long-term hydration of C_2S , which generates larger amounts of hydration products [18,45,52] and results in a denser M2 microstructure [53,54]. Moreover, the hydration of C_2S consumes more water than that of C_3S (Eqs. (1) and (2)), and M2 had a significantly higher C_2S content than M1 (Table 4). Consequently, the free water was reduced in M2 at later ages. Thus, the shrinkage value increased with curing age [52,53,62], thereby enhancing the frictional bond at the FMZ. Wang et al. [54] reported that the shrinkage of BRC-based concrete was up to 42% lower than that of OPC-based concrete at 3 days but only approximately 26% lower at 28 days.

The results show that PE_{total} and τ_{equi} of H fibers were approximately 2–4 times higher than those of S fibers, depending on the matrix type and curing age. This enhancement is attributed to the mechanical anchorage provided by the hooked ends [55,56,58,63], which is absent in S fibers. At the initial stage of fiber pullout, the adhesion between the fiber and matrix was fractured, and then friction at the interface between the fiber and matrix was activated; once these mechanisms were fully engaged, the hooked end started to contribute additional mechanical resistance based on the mechanical interaction between the fiber and matrix to be pulled out. The higher PE_{total} of H fibers compared with that of S fibers can be explained by the slip-hardening pullout response. During pullout, the hooked end should be straightened, which increased slip at P_{peak} and resulted in slip-hardening compared with the slip-softening response of S fibers [56]. Consequently, the pullout load was maintained over a larger slip range after P_{peak} .

In addition, the pullout resistance of H fibers was less sensitive to variations in C_2S and LP contents than that of S fibers, particularly for PE_{total} and τ_{equi} at 90 days. For S fibers, pullout resistance closely followed changes in compressive strength, whereas H fibers showed a slight variation. For example, at 28 days, clear differences in PE_{total} (Fig. 15) and τ_{equi} (Fig. 16) were observed for S fibers, while H fibers

showed similar values regardless of matrix composition. Furthermore, at 90 days, S fibers exhibited significant increases in P_{peak} (from 64.6 to 85.1 N, Fig. 12) and PE_{total} (from 175.7 to 289.1 N-mm, Fig. 15), corresponding to an increase in compressive strength (from 69.0 to 96.3 MPa, Fig. 10) as the C_2S content increased from 16.4% (M1) to 57.2% (M2). In contrast, this trend was not clear for H fibers.

3.3. Fiber–matrix interface zone

3.3.1. Microstructural images

Fig. 17 shows the SEM images and EDS mapping at the FMZs for four matrices after 90 days of curing, whereas Table 5 presents their wt% chemical elements. The results showed that the FMZs of the BRC-based matrix (M2) exhibited a higher calcium (Ca) content than those of the OPC-based matrix (M1), as evidenced by the greater number of blue points in Figs. 17b and 17a. The corresponding Ca contents were 10.59% and 7.31% for M2 and M1, respectively, as shown in Table 5. This phenomenon is due to the higher amount of C-S-H produced in the matrix with a greater C_2S content (M2) at later ages compared with the matrix containing lower C_2S content (M1). In addition, fewer and finer cracks were observed in M2 than in M1. This can be attributed to the lower chemical shrinkage in M2 due to the slower hydration rate of C_2S [45,53,62].

Furthermore, the results revealed higher carbon (C) contents in matrices with greater LP substitution, as evidenced by the increasing density of blue points in M4 compared with those in M3 and M2 (Figs. 17d, 17c, and 17b, respectively). Correspondingly, Table 5 shows that the C contents were 18.61%, 21.47%, and 42.14% for M2, M3, and M4, respectively. These results indicate that LP effectively filled the pores, leading to denser microstructures in the FMZs. However, excessive LP substitution reduced the calcium (Ca) content in the FMZs. As shown in Figs. 17b and 17d, more green points (Ca) were observed in M2 than in M4. In contrast, the incorporation of an appropriate LP content enhanced the Ca concentration in the FMZs. Specifically, at 10% LP substitution (M3, Fig. 17c), more green points were observed than in M2 (Fig. 17b) and M4 (Fig. 17d). The corresponding Ca contents were 10.59%, 10.74%, and 6.12% for M2, M3, and M4, respectively.

3.3.2. Phase analysis by deconvolution technique

3.3.2.1. Deconvolution technique. Cementitious materials always have a multiphase nature. Therefore, the key objective of the grid-indentation technique is to determine the mechanical properties of the individual

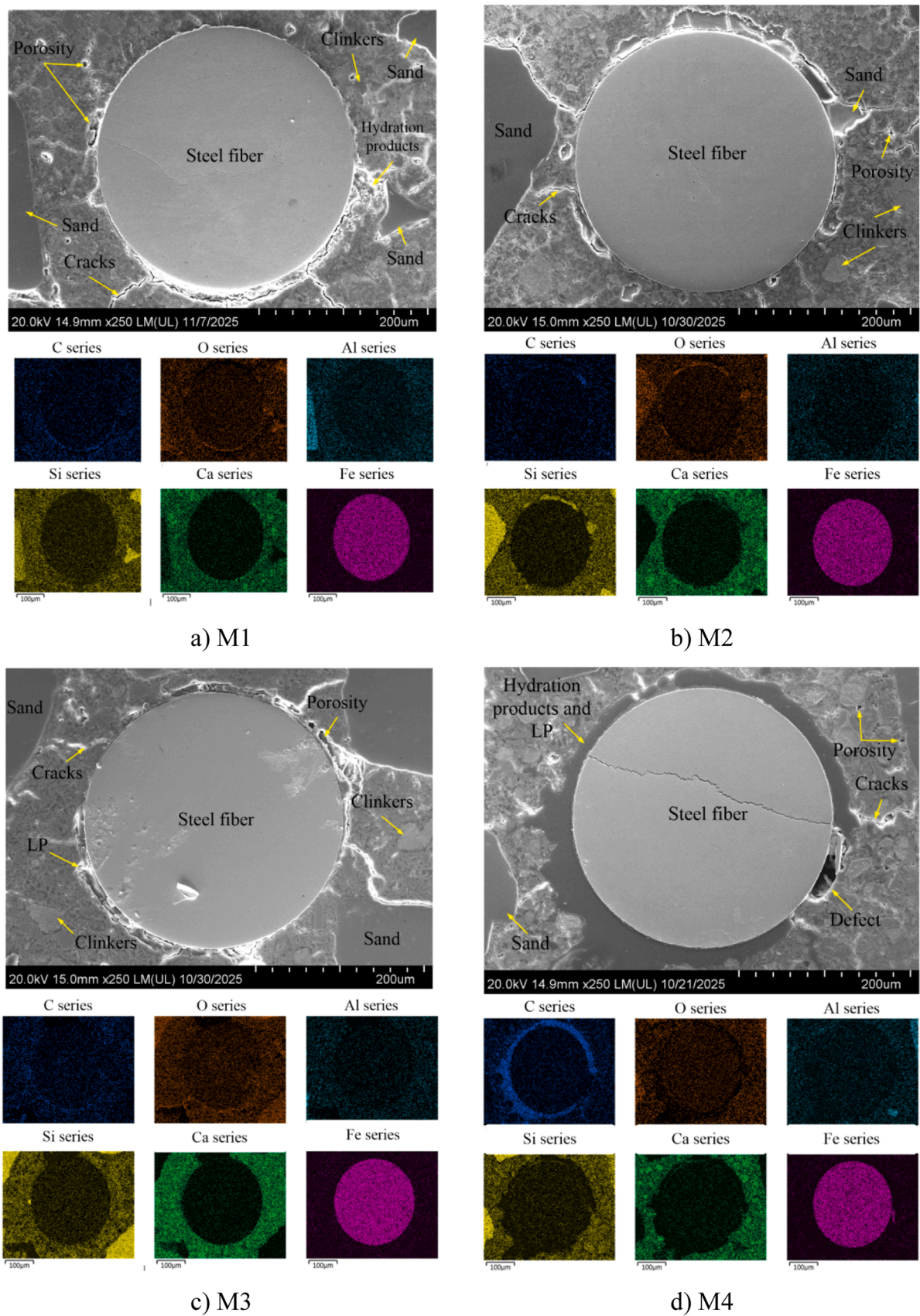


Fig. 17. SEM images and EDS mapping at the FMZ.

Table 5
wt% of chemical elements at FMZs.

Elements	M1	M2	M3	M4
C	26.77	18.61	21.47	42.14
O	24.43	27.69	28.22	20.17
Al	1.50	0.69	0.55	0.66
Si	5.47	9.04	6.97	4.12
S	0.20	0.24	0.10	0.16
Ca	7.31	10.59	10.74	6.12
Fe	34.30	33.14	31.95	26.63

Note: C – Carbon; O – Oxygen; Al – Aluminum; Si – Silicon; S – Sulfur; Ca – Calcium; Fe – Iron

phases [64]. However, composite materials are generally more complex, making direct analysis more challenging. Consequently, statistical distributions of the individual phases must be employed to interpret indentation data [65]. This requires the deconvolution of heterogeneous data, hence the name “deconvolution technique”. The deconvolution technique was originally developed by Constantinides et al. [65–67]. The key to this deconvolution technique is fitting several probability density functions (PDF) to the experimental frequency plot (normalized histogram) of the measured quantity [48,49,64].

Assuming that the objective materials contain n phases and that the indentation data corresponding to each phase of the materials conform to a Gaussian normal distribution, the distribution of the mechanical property $x = (E, H)$ for each phase J is expressed as follows:

$$p_J(x) = \frac{1}{\sqrt{2\pi s_J^2}} \exp\left(-\frac{(x - \mu_J)^2}{2s_J^2}\right) \quad (9)$$

where μ_J and s_J are the mean and standard deviation of $x = (E, H)$ of phase $J = 1$ to n .

The overall frequency distribution ($P(x)$) of the mechanical properties $x = (E, H)$ was calculated by summing n theoretical PDFs, that is,

$$P(x) = \sum_{J=1}^n f_J p_J(x) \quad (10)$$

where f_J is the surface fraction of phase J . f_J is calculated and subjected to the constraints following Eqs. (11) and (12), respectively:

$$f_J = \frac{N_J}{N} \quad (11)$$

$$\sum_{J=1}^n f_J = 1 \quad (12)$$

where N_J and N are the numbers of indentations on phase J and total number of experiments, respectively.

The unknown parameters (f_J , μ_J , and s_J) can be determined by minimizing the standard error between the theoretical and experimental PDF:

$$\min \sum_{i=1}^m \frac{(P_{\text{exp}} - P(x_i))^2}{m} \quad (13)$$

where P_{exp} and $P(x_i)$ represent the experimental and theoretical PDF, respectively. m is the number of bins chosen to construct the histogram. To avoid overlapping adjacent phases, the deconvolution result must satisfy the following constraint conditions:

$$\mu_J + s_J \leq \mu_{J+1} + s_{J+1} \quad (14)$$

3.3.2.2. Phase analysis. Fig. 18 presents the experimental and best-estimated PDFs for hardness (H) and modulus (E) of phases in the FMZs of four different mortar matrices (M1, M2, M3, and M4) after 90 days of curing, obtained using the deconvolution technique. Fig. 18a–d

correspond to the H distributions, while Fig. 18e–h show the E distributions for M1, M2, M3, and M4, respectively. In each case, the x-axis represents H (or E), while the y-axis represents the frequency density calculated using Eqs. (9) and (10). From left to right along the x-axis, H increases in the following order: Porosity < low-density C-S-H (LD C-S-H) < high-density C-S-H (HD C-S-H) < CH < Clinker < Sand. As illustrated in Fig. 18, the following observations can be made: (1) for porosity, the peaks in H and E were centered around 0.2–0.3 GPa and 9–12 GPa, respectively, which were close to the modulus value of approximately 10 GPa reported in Ref. [48]. These low values can be attributed to voids, cracks, or poorly hydrated regions in the matrix. (2) For LD C-S-H and HD C-S-H, peaks in H and E were observed in the ranges of 0.4–1.5 GPa and 15–32 GPa, respectively. As expected, LD C-S-H exhibited lower values than HD C-S-H due to its more loosely packed microstructural configuration [48]. (3) For CH, peaks in H and E appeared at 2.5–3.2 GPa and 40–50 GPa, respectively, which are consistent with the values reported in Refs. [48,66]. (4) For sand, the peaks were centered around 8.8–10.0 GPa for H and 75–100 GPa for E , in agreement with previous studies [39,49]. (5) For cement clinker, peaks in H and E were observed at 6.5–7.5 GPa and 150–170 GPa, respectively. The existence of clinker particles even after long hydration periods can be explained by the use of a low water-to-cement ratio ($w/c < 0.42$), which limited complete hydration [48,49].

Table 6 summarizes the corresponding parameters from the deconvolution analysis, including f_J , μ_J , and s_J . The comparison between M1 and M2 highlighted the effects of varying C_2S content, while the comparison among M2, M3, and M4 illustrated the effect of changing the LP content in the mixtures. As shown in Table 6, the C-S-H phases (LD and HD) were the most abundant in all matrices, contributing approximately 40–50% of the total composition. This is because C_2S and C_3S are the major constituents of cement clinker, and their primary hydration products are C-S-H. In addition, a higher C-S-H content was observed in FMZ with increasing C_2S content at 90 days. Specifically, the total C-S-H fraction increased from 34.6% (M1) to 51.5% (M2) as the C_2S content increased from 16.4% (M1) to 57.2% (M2). In M1, the clinker fraction remained relatively high (18.5%), whereas in M2, it decreased to 8.1%, indicating more complete hydration in M2. Less residual clinker corresponds to more hydration products, thereby explaining the higher C-S-H content measured in M2. In contrast, the porosity and C-S-H content in the FMZs decreased when adding LP. Specifically, porosity decreased from 25.0% (M2) to 18.9% (M4), while the C-S-H fraction decreased from 51.5% (M2) to 40.9% (M3). The reduction in C-S-H observed in matrices with LP contents can be attributed to their decreased cement content. However, at a higher replacement level of 30% LP (M4), the C-S-H content in the FMZs was slightly higher than that of M3. While a 10% LP substitution reduced the C-S-H fraction, an increase of LP to 30% resulted in a slight improvement of C-S-H (47.3%). This trend can be attributed to the enhanced filler and nucleation effects of LP, which improved the hydration efficiency of the remaining cement. The increased number of LP particles provided additional nucleation sites for C-S-H precipitation and facilitated water access to C_2S and C_3S , leading to slightly higher C-S-H content in the FMZs of M4 compared with M3.

3.4. Environmental evaluation

This study evaluated the efficiency of balancing the environmental impact and mechanical performance of mortar matrices with varying C_2S and LP contents. The eco-efficiency index (EI) was calculated based on the total carbon dioxide emissions (CO_{2-e}), CS, P_{peak} , and PE_{total} at three curing ages of 14, 28, and 90 days. Because the H fiber showed limited sensitivity to variations in the C_2S and LP contents, only P_{peak} and PE_{total} values of the S fiber were considered in the determination of EI. The CO_{2-e} values were calculated according to Eq. (15), while the indices of EI_{CS} , EI_P , and EI_{PE} , corresponding to CS, P_{peak} , and PE_{total} , were calculated using Eqs. (16), (17), and (18), respectively.

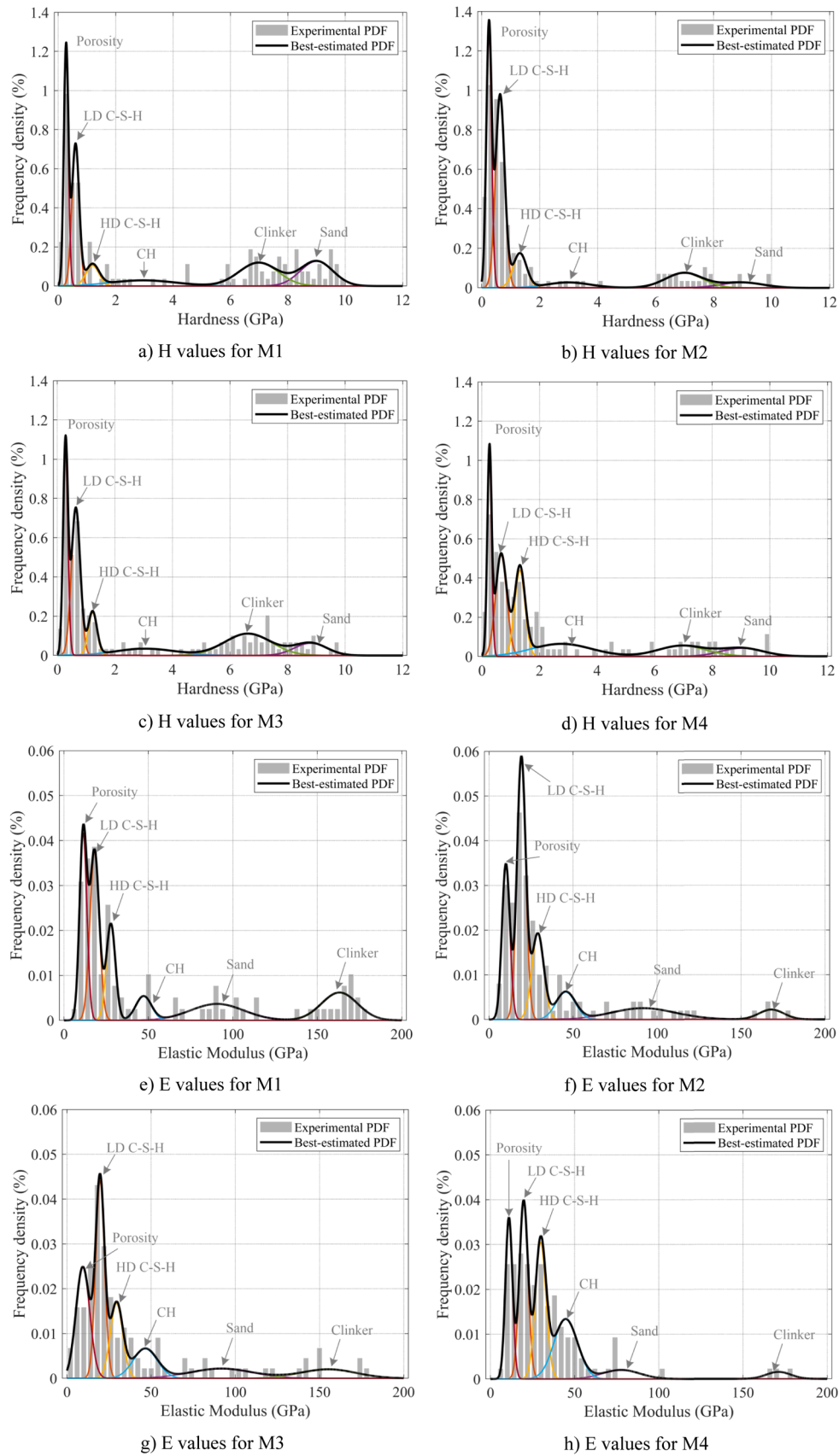


Fig. 18. Experimental and best-estimated PDFs for H and E of indents in the FMZs for different matrices with different C_2S and LP contents at 90 days after curing.

Table 6

Deconvolution results of FMZs after 90 curing days.

Matrix	Phases	E		H		Mean f_J (%)	f_{C-S-H} (%)
		$\mu_J \pm s_J$ (GPa)	f_J (%)	$\mu_J \pm s_J$ (GPa)	f_J (%)		
M1	Porosity	11.4 ± 2.2	22.7	0.27 ± 0.08	24.2	23.5	
	LD C-S-	18.1 ± 2.7	25.8	0.60 ± 0.13	22.7	24.3	34.6
	H						
	HD C-S-	27.8 ± 2.7	14.4	1.19 ± 0.23	6.1	10.3	
	H						
	CH	47.1 ± 4.6	6.2	2.96 ± 1.15	8.3	7.3	
	Sand	90.7 ± 15.7	14.4	9.00 ± 0.57	18.2	16.3	
	Clinker	163.1 ± 10.6	16.5	6.98 ± 0.68	20.5	18.5	
M2	Porosity	10.0 ± 2.3	20.2	0.25 ± 0.09	29.8	25.0	
	LD C-S-	19.2 ± 2.3	37.9	0.63 ± 0.16	38.3	38.1	51.5
	H						
	HD C-S-	29.0 ± 3.5	16.9	1.32 ± 0.23	9.9	13.4	
	H						
	CH	45.6 ± 5.7	8.9	2.99 ± 0.74	5.0	7.0	
	Sand	91.7 ± 19.0	12.1	8.97 ± 0.72	5.0	8.6	
	Clinker	168.1 ± 7.3	4.0	6.98 ± 0.64	12.1	8.1	
M3	Porosity	9.4 ± 4.1	25.5	0.29 ± 0.08	22.6	24.1	
	LD C-S-	19.7 ± 2.6	29.1	0.65 ± 0.15	28.8	29.0	40.9
	H						
	HD C-S-	29.4 ± 3.7	15.5	1.23 ± 0.15	8.2	11.9	
	H						
	CH	46.5 ± 7.2	11.8	3.07 ± 1.03	8.9	10.4	
	Sand	91.6 ± 18.4	10.0	8.80 ± 0.56	8.9	9.5	
	Clinker	155.3 ± 16.3	8.2	6.62 ± 0.81	22.6	15.4	
M4	Porosity	11.0 ± 2.1	18.7	0.26 ± 0.07	19.1	18.9	
	LD C-S-	19.7 ± 2.5	25.2	0.67 ± 0.19	25.2	25.2	47.3
	H						
	HD C-S-	29.8 ± 3.2	24.3	1.32 ± 0.18	19.8	22.1	
	H						
	CH	44.7 ± 7.0	23.4	2.76 ± 1.04	16.8	20.1	
	Sand	77.6 ± 11.1	5.6	9.03 ± 0.74	7.6	6.6	
	Clinker	171.1 ± 7.2	2.8	6.95 ± 0.83	11.5	7.2	

Table 7CO_{2-e} of the raw materials.

Item	OPC [18, 70]	BRC [18, 23,70]	LP [18, 70]	Silica sand [68,69]	Water [18]
CO _{2-e} (kg/ton)	860	774	8	100	0.196

considering raw material consumption, energy demand, fuel and electricity mixes, transportation, and raw material-derived CO₂ emissions. The system boundary followed a cradle-to-gate approach, covering stages from raw material extraction to cement production, with a functional unit of 1000 kg of material. Production data were evaluated at regional levels and combined using weighted averages to obtain global emission factors. Multiple datasets were integrated to capture key inputs such as kiln efficiency and regional production volumes. The Getting the Numbers Right (GNR) database provided regional kiln distribution and energy consumption (MJ/t clinker), which were used to estimate clinker production energy. Fuel mixes were derived from GNR and International Energy Agency (IEA) Cement Roadmap data, while fuel-related CO₂ emissions were calculated using the Intergovernmental Panel on Climate Change emission factors for stationary combustion in manufacturing and construction sectors.

Cement consisting of clinker and gypsum was predicted by including additional processing and 5% gypsum by mass. According to Miller et al. [70], the production of 1 ton of cement emitted 901 kg CO₂, with contributions from calcination (51%), fuel combustion (42%), clinker cooling (1%), milling (3%), raw material acquisition (1%), and grinding (2%). However, CO₂ emissions differed regionally because of the differences in regional kiln efficiency and fuel types, transportation considerations and electricity mixes. The reported CO₂ emission values were 842 kg/ton (Africa), 860 kg/ton (Europe), 865 kg/ton (China), 839 kg/ton (Japan), and 910 kg/ton (USA), etc., providing a weighted global average of 860 kg/ton. This value was therefore adopted for OPC in this study. Similarly, CO₂ emissions for BRC vary by region, including 768 kg/ton (Africa), 783 kg/ton (Europe), 791 kg/ton (China), 721 kg/ton (Japan), and 826 kg/ton (USA), etc., with a weighted average of 787 kg/ton. Gartner et al. [23] reported that BRC production emitted approximately 10% less CO₂ than OPC, corresponding to 774 kg/ton [18]. As the difference between reported values is minor, a value of 774 kg/ton was used for BRC in this study. For LP, Miller et al. [70] also indicated regional variation, with a weighted average emission factor of 8 kg/ton, which was adopted here.

Fig. 19 shows the parameters used for environmental evaluation, including CO_{2-e}, EI_{CS} , EI_P and EI_{PE} . Lower values of CO_{2-e} indicate a more environmentally friendly performance, while lower values of EI_{CS} , EI_P , and EI_{PE} represent a better sustainability–performance balance. The results show that CO_{2-e} decreased with increasing C₂S and LP contents, as shown in Fig. 19a. When the C₂S content increased from 16.4% to 57.2%, CO_{2-e} was reduced from 726.7 to 664.8 kg/m³. Similarly, increasing LP content from 0% to 30% further decreased CO_{2-e} from 664.8 to 499.5 kg/m³. Higher C₂S and LP contents reduced CO_{2-e} because BRC and LP have lower embodied CO₂ than OPC, as shown in Table 7. In addition, the values of EI_{CS} , EI_P , and EI_{PE} decreased with increasing curing age because a higher performance was obtained at later ages than at an early age. For example, the CS of M1 at 90 days was 69 MPa, whereas that at 14 days was only 57.6 MPa, as shown in Fig. 10.

The results of EI_{CS} indicate that the use of BRC alone was not beneficial in balancing sustainability and CS at an early age, although it became more advantageous at later ages, as observed in the comparison between M1 and M2 (Fig. 19b). In contrast, the incorporation of LP consistently reduced EI_{CS} across all curing ages, as observed by comparing M3 and M4 with M1. Among the mixtures, the matrix containing BRC with 10% LP substitution (M3) exhibited the most promising balance between CO_{2-e} and CS. Specifically, M3 achieved the lowest EI_{CS} values at 14 and 90 days. For example, the EI_{CS} values of M1

$$CO_{2-e} = \sum_{i=1}^n (M_i \times CO_{2-i}), \quad (15)$$

$$EI_{CS} = \frac{CO_{2-e} \text{ (kg/m}^3\text{)}}{CS \text{ (MPa)}} \quad (16)$$

$$EI_P = \frac{CO_{2-e} \text{ (kg/m}^3\text{)}}{P_{peak} \text{ (N)}} \quad (17)$$

$$EI_{PE} = \frac{CO_{2-e} \text{ (kg/m}^3\text{)}}{PE_{total} \text{ (N} \bullet \text{mm)}} \quad (18)$$

where M_i is the mass of each material per unit volume of mortar (kg/m³), and CO_{2-i} represents the CO₂ emissions of each raw material, as summarized in Table 7.

In this study, the carbon footprint of raw materials was adopted from Refs. [18,23,68–70] with detailed calculations based on the life-cycle assessment (LCA) methodology described in Ref. [70]. The assessment quantified CO₂ emissions associated with material production,

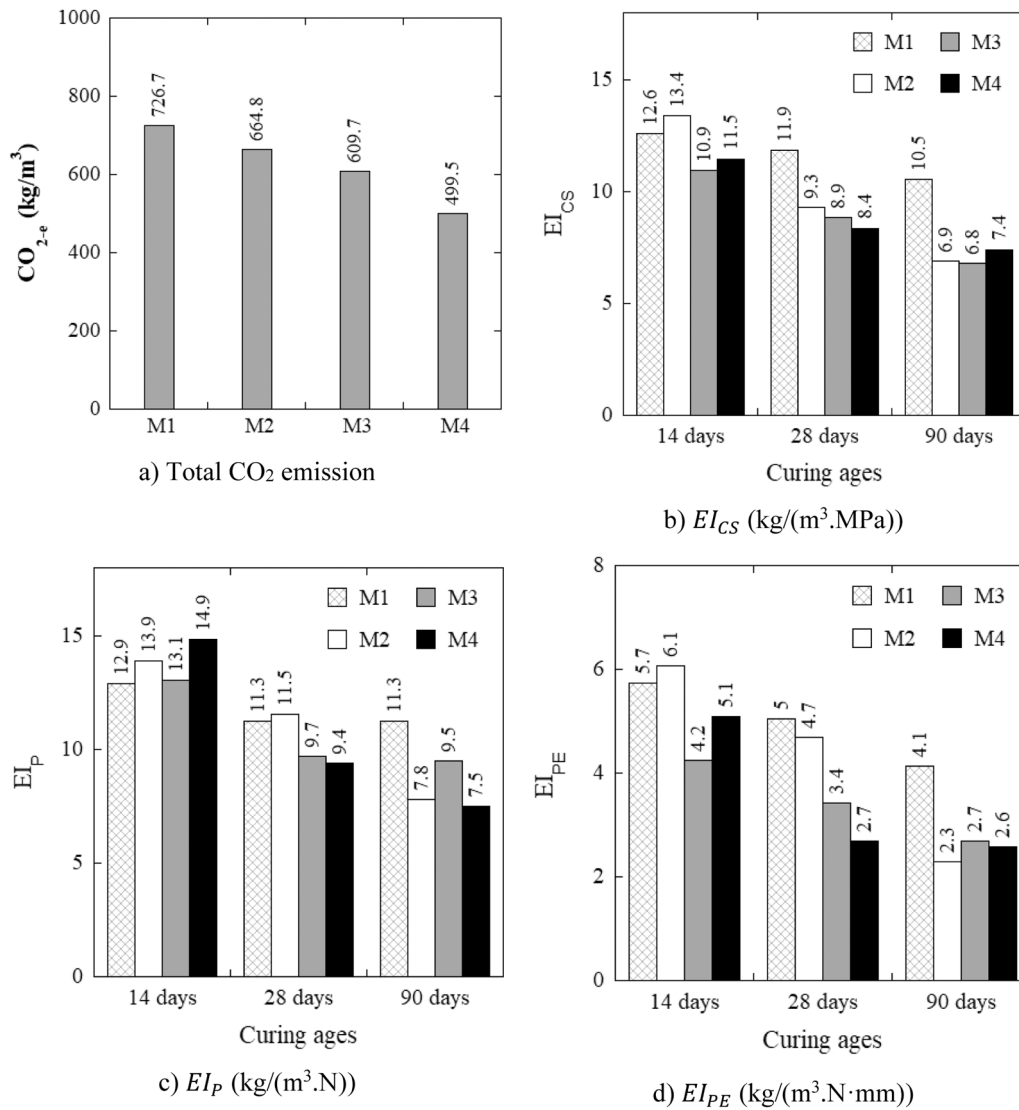


Fig. 19. Parameters for environmental evaluation: a) CO_{2-e}, b) EI_{CS}, c) EI_P, and d) EI_{PE}.

at 14, 28, and 90 days were 12.6, 11.9, and 10.5, respectively, whereas those of M3 were substantially lower at 10.9, 8.9, and 6.8, respectively. Therefore, the substitution of LP provided a better balance between sustainability and compressive strength than plain BRC at both early and later curing ages.

For pullout resistance, matrices incorporating BRC showed higher EI_P and EI_{PE} at 14 days, but significantly lower values were observed at 90 days. For instance, at 14 days, the EI_P value for M1 was 12.9, compared to 13.9 for M2. In addition, the EI_{PE} value for M1 (5.7) was slightly lower than that for M2 (6.1). In contrast, at 90 days, the EI_P value for M1 (11.3) was significantly higher than that for M2 (7.8), while the EI_{PE} value for M1 (4.1) was remarkably higher than that for M2 (2.3). Lower EI values indicate a better sustainability balance. Thus, the results suggest that the mixtures with BRC achieved a more favorable balance between pullout resistance and sustainability at later ages than at early ages.

Moreover, the substitution of LP into BRC (M3 and M4) resulted in lower values of EI_P from 28 days onward, whereas the EI_{PE} values were reduced as early as 14 days, compared with the matrix prepared with OPC (M1), as shown in Figs. 19c and 19d. This indicates that LP substitution can effectively compensate for the reduced pullout resistance observed in BRC alone relative to OPC.

4. Discussions

4.1. Effects of C₂S and LP contents on compressive strength

The lower early-age CS of M2 than that of M1 can be attributed to the slower hydration kinetics of C₂S, which is the dominant phase in BRC. Scrivener et al. [2] showed that the hydration of C₂S was insignificant during the first 10 curing days. In this study, M2 contained 57.2% C₂S, compared to only 16.4% in M1 (Table 4). However, a significantly higher CS was observed in M2 than in M1 at 90 days. The higher long-term strength observed in M2 was attributed to significant hydration of C₂S over time. The continued hydration of C₂S decreased the available water. Therefore, denser microstructures were obtained at later ages [53,54]. In addition, the continued hydration of C₂S resulted in greater cumulative formation of C-S-H in M2 than in M1 [18,45,52] because the complete hydration of C₂S generates more C-S-H than that of C₃S [18,54,71]. As C-S-H is the primary hydration product responsible for strength development [18], its increased formation significantly contributed to the enhanced long-term CS in mortars with a higher C₂S content.

In addition, the CS of mortar matrices decreased with increase in LP content, even though the porosity at the FMZs decreased, as shown in Table 6. This discrepancy can be explained by the different roles of

porosity and reactive binder content in governing CS. While the addition of LP improved particle packing and refined the pore structure through filler and nucleation effects, LP was largely inert and replaced part of the reactive binder. At higher LP contents, the reduction in cement led to a lower volume of hydration products, which outweighed the beneficial effect of the reduced porosity. Consequently, despite a denser microstructure, the overall load-bearing capacity of the matrix decreased, resulting in reduced CS. In addition, it should be noted that the porosity values obtained from NI represented local porosity characteristics measured in the FMZ, rather than the bulk porosity of the entire compressive specimen. Therefore, the porosity derived from the FMZ cannot be considered fully representative of the overall porosity of the compressive specimens. As a result, the CS cannot be directly correlated with the FMZ porosity alone.

The reduction in CS at 10% LP substitution (M3) was insignificant compared to that at 30% substitution (M4), relative to the control mixture (M2) with no LP. The insignificant reduction in CS observed in M3 compared to that in M4 can be attributed to the beneficial effect of LP in accelerating strength development. LP acts as a nucleation site for hydration products, thereby compensating for the reduced cement content at 10% LP substitution. This finding is consistent with the results reported by Nguyen et al. [18]. They found that a 10% LP substitution produced a CS comparable to that of OPC-based paste. However, substituting 30% LP caused a marked decrease in CS, as the positive influence of LP was insufficient to compensate for the considerable reduction in cement content. At 30% LP replacement, dilution dominated rather than nucleation [18,72]. Therefore, a 30% LP substitution should be avoided in the preparation of high-performance concrete.

Moreover, the mortar with 10% LP substitution exhibited a higher CS at 14 days than the mixture without LP, although it became lower at 90 days. This behavior can be attributed to the nucleation effect of LP. LP served as a nucleation site for C-S-H and CH formed during C_2S and C_3S hydration [18]. These products precipitated on LP-particle surfaces rather than only on the reactant surfaces in matrices without LP. Consequently, clinker particles in LP-containing mixtures had greater access to water, thereby accelerating hydration at early ages. However, at later ages, once all LP particles were covered by hydrated products, their beneficial effect was reduced. Consequently, the hydration of C_2S becomes the dominant mechanism at later ages. Thus, the CS of BRC without LP exceeded that of cement with 10% LP substitution at 90 days. These findings indicate that the use of an optimal LP content can effectively accelerate early-age compressive strength development through its nucleation effect, whereas the hydration of C_2S plays a more significant role in later-age strength gain.

4.2. Effects of C_2S and LP contents on pullout resistances

The lower pullout resistances of matrices with higher C_2S contents were observed at an early age. This phenomenon can be attributed to the slow hydration kinetics of C_2S . Generally, the hydration rate of C_2S was lower than that of C_3S by approximately 5–8 times due to the lower solubility of C_2S [43,73]. M2 prepared by BRC exhibited a higher C_2S content than M1 prepared by OPC. Specifically, M2 contained 57.2% C_2S , whereas M1 contained only 16.4% C_2S . The slower hydration of C_2S produced fewer hydration products at an early age. Several studies have investigated the evolution of hydration products through extensive experimental methods, including thermogravimetric analysis (TGA) and X-ray diffraction (XRD) [45,54,74–78], as well as three-dimensional (3D) hydration simulations [18]. Nguyen et al. [18] reported, based on TGA results, that bound water—representing the evolution of C-S-H gel, ettringite, and hemi- and/or mono-carboaluminate phases at 7 days—decreased from 13.11% to 8.75% as the C_2S content increased from 21.9% to 57.2%. These findings are consistent with those of Meng et al. [74] observed that bound water content increases over time but tended to decrease with increasing BRC content. In addition, several studies reported that larger amounts of C-S-H and CH were produced at

early ages in OPC-based matrices than in BRC-based matrices [45,54,74–78]. The lower amount of hydration products and chemically bound water in BRC-based matrices can be attributed to two main reasons. First, BRC exhibits a lower reaction rate than OPC [45,54,74–78]. Second, the complete hydration of C_3S produces more chemically bound water than that of C_2S . Specifically, hydration of 1 g of C_3S generates 0.237 g of chemically bound water, whereas 1 g of C_2S produces 0.209 g [74].

Consequently, the stiffness and strength of the interfacial transition zone (ITZ) decreased in M2 compared to that in M1 at 14 days. This trend was further confirmed by the lower compressive strength of M2 relative to M1 at 14 days, as shown in Fig. 10. Furthermore, the frictional bond between the fiber and matrix strongly depends on the shrinkage of the matrix [58,79], whereas the shrinkage decreased with increased water content [58,80]. Owing to the slower hydration of C_2S , less water was consumed in matrices with higher C_2S contents at an early age. Therefore, a higher amount of water was present in the matrix with higher C_2S content. This smaller shrinkage reduced the frictional bond at the FMZ, explaining the lower bond resistance observed in matrices with higher C_2S content at early ages. The shrinkage value of BRC-based concrete was lower than that of OPC-based concrete under the same environment due to the slower hydration rate of C_2S [45,53,62]. Specifically, the shrinkage of BRC-based concrete was up to 42% lower than that of OPC-based concrete at 3 days [54].

However, matrices with higher C_2S content showed superior performance at 90 days. This improvement is attributed to the long-term hydration of C_2S . Nguyen et al. [18] reported that BRC-based matrices (M2) produced higher amounts of C-S-H and lower amounts of CH after 90 days of hydration compared with OPC-based matrices (M1). At 90 days, the mass fractions of C-S-H were 61.1% for M2 and 53.5% for M1, while those of CH were 12.88% and 17.9%, respectively [18]. In addition, bound water—representing the evolution of C-S-H gel, ettringite, and hemi- and/or mono-carboaluminate phases—was also higher in M2 than in M1 at 90 days. These findings agreed with previous studies [45,54,74–78], which indicated that the BRC reaction became more pronounced at later ages, producing more C-S-H and less CH. The higher C-S-H and lower CH contents in M2 were attributed to the hydration characteristics of C_3S and C_2S , which generated more C-S-H and less CH [74–77]. Meng et al. [74] reported that completed hydration of 1 g of C_3S produced 0.75 g of C-S-H, whereas that of 1 g of C_2S generated 0.994 g. Furthermore, continued C_2S hydration in BRC at later ages reduces porosity and shifted the pore size distribution toward finer pores [77], thereby improving the microstructure of the matrices. Overall, later-age hydration of BRC resulted in higher C-S-H and lower CH and AFm contents [77]. Since C-S-H is the primary hydration product in cement-based materials, matrices with higher C-S-H content exhibit better performance [77].

Furthermore, the superior performance in matrices with higher C_2S content at later ages can be explained by three main factors. First, significant amounts of C_2S were consumed after 90 days of hydration. This is evidenced by the lower residual clinker content in M2 (8.1%) than in M1 (18.5%), as shown in Table 6. The extensive hydration of C_2S reduced the amount of available water in the system, leading to the formation of denser microstructures at later ages compared to early ages [53,54]. Second, the hydration of C_2S generated more C-S-H than C_3S , with 1 vol of C_2S producing 2.077 volumes of C-S-H, compared to only 1.521 volumes from 1 vol of C_3S (Eqs. (1) and (2)), resulting in a denser microstructure and stiffer ITZ. This finding is further supported by the NI results of the FMZs, where the C-S-H contents at 90 days were 34.6% for M1 and 51.5% for M2. Third, hydration of C_2S consumed more water than that of C_3S (1.49 vs. 1.34 volumes per unit volume), as shown in Eqs. (1) and (2), leading to reduced free water content at later ages. Less free water contributed to greater shrinkage [53,58,80], thereby enhancing the frictional bond at the fiber-matrix interface. Consequently, the pullout performance significantly improved in matrices with higher C_2S content at later ages. Wang et al. [54] reported that the

shrinkage of BRC-based concrete was up to 42% lower than that of OPC-based concrete at 3 days but only 26% and 24% lower at 28 and 180 days, respectively.

In addition, pullout performance decreased with increasing LP content. This phenomenon is because of the lower cement content in matrices with higher LP substitution. Nguyen et al. [18] reported that bound water, representing the evolution of C-S-H, ettringite, and carboaluminate phases, decreased with the addition of LP. Although LP accelerated the hydration rate, this effect did not compensate for the reduced cement content. Consequently, these matrices produced fewer hydration products, resulting in a less compact microstructure. However, the reduction observed in the 10% LP matrix (M3) was insignificant. This finding indicates that the nucleation effect of LP compensates for the reduction in cement content by accelerating hydration. Furthermore, M3 exhibited a higher P_{peak} at 14 days for H fibers and at 28 days for S fibers, whereas the matrix without LP (M2) showed a significantly higher P_{peak} at 90 days. These results suggest that LP improves early-age strength and pullout performance, while C_2S contributes more effectively at later ages. However, excessive LP addition led to significant reductions in both pullout performance and compressive strength; for example, M4 (30% LP) showed a 42% decrease in P_{peak} compared with M2 at 14 days. Therefore, a 10% LP substitution is recommended for practical applications.

Furthermore, the lower sensitivity of H fibers to variations in matrix compositions was observed in Figs. 15 and 16. This phenomenon can be attributed to their different pullout mechanisms [55,57]. Specifically, the pullout resistance of H fibers is primarily governed by mechanical anchorage effects [55,57]. During pullout, the fibers were straightened through rotation of their cross-sections at the embedded location. In contrast, the pullout resistance of S fibers, which depends primarily on adhesive and frictional bonding at the ITZ [55,81,82], demonstrated greater sensitivity to matrix composition, particularly at later ages. From this observation, mortars prepared with BRC and 10% LP are recommended for both S and H fibers, while those with 30% LP are suitable primarily for H fibers, as their performance was comparable to that of OPC. Both mixtures offer a favorable balance between sustainability and mechanical performance, comparable to that of OPC.

4.3. Correlation of the development of FMZ and fiber pullout resistance

The FMZ with a higher C-S-H content exhibited better pullout performance than that with a lower C-S-H content. The matrices with higher C_2S content exhibited greater amounts of C-S-H in the FMZ after 90 days of curing. The higher C-S-H fraction in FMZs of mixtures rich in C_2S caused by the sustained and progressive hydration of C_2S over time. Unlike C_3S , which hydrates rapidly at early ages, C_2S reacts more slowly but continues to contribute to the hydration products at later ages. Additionally, the full hydration of 1 unit volume of C_2S produced higher volumes of C-S-H than C_3S . Therefore, a higher C_2S content ultimately leads to the formation of a larger volume of C-S-H. The higher C-S-H content in M2 than that in M1 can be confirmed by the NI results presented in Section 3. Specifically, NI analysis showed that the C-S-H content in M2 was 51.5%, whereas that in M1 was only 34.6%. Moreover, the lower residual clinker content observed in M2 compared with that in M1 indicated that more clinkers hydrated. The remaining clinkers for M2 was 8.1%, whereas that for M1 was 18.5% after the curing for 90 days, as shown in Table 6. Consequently, M2 produced more hydration products than M1. Because C-S-H is the principal hydration product and the primary contributor to strength development, the increased C-S-H formation improved pullout performance of M2 compared to that of M1 at 90 days.

In contrast, the porosity in FMZs decreased with increasing LP. The reduction in porosity was attributed to the nucleation, filler, and chemical effects of LP. Several studies reported that LP accelerated the hydration process and produced a denser microstructure through

nucleation and chemical effects [18,72,83–85]. Regarding the nucleation effect, LP, composed of $CaCO_3$, exhibited a surface atomic arrangement similar to that of C-S-H. As a result, LP allowed C-S-H (and possibly CH) to nucleate and grow on its surface [86,87]. Therefore, LP particles provided nucleation sites for the growth of C-S-H and CH [83, 85]. Consequently, LP promoted faster precipitation from the pore solution, increasing the rate of hydration reactions [88] and enhancing cement hydration [72,89,90]. Concerning the filler effect, LP also acted as a filler, leading to reduced capillary porosity [72]. In addition, LP particles filled the voids between coarser cement particles, thereby increasing packing density, reducing the total pore volume of the cement system, and contributing to microstructural densification [18, 90].

For the chemical effect, reactions occurred in the presence of C_3A in the matrix, where C_3A reacted with LP to form monocarbonate and hemihydrate phases [18,89,91]. The formation of monocarbonate contributed to microstructural densification by acting as an additional solid phase within the pore structure [72]. Furthermore, the presence of LP stabilized ettringite and inhibited its transformation into monosulphate. As a result, the total volume of hydration products increased [91–93]. This increase in hydration product volume could reduce porosity to an extent that compensated for the partial replacement of PC with LP. The stabilization of ettringite in the presence of LP therefore led to a higher solid volume of hydration products, as supported by chemical shrinkage results, resulting in reduced porosity [91–93] and a denser microstructure in the FMZs.

In general, LP particles acted as nucleation sites during hydration and physically filled voids between cement grains, thereby refining the pore structure and lowering the overall porosity of the FMZ. In contrast, a decrease in C-S-H content was observed in matrices with LP substitution. As LP replaced a portion of the cement, the total available clinker phases decreased, leading to reduced C-S-H formation. However, at a higher substitution level of 30%, a slight increase in C-S-H content at 90 days was observed. This can be attributed to the enhanced filler and nucleation effects of LP particles, which provided additional sites for C-S-H precipitation and facilitated water access to C_2S and C_3S . Although the addition of LP improves the packing density and reduces porosity, it lowers the total C-S-H content due to the lower cement content in the system. Despite the densification effect, the pullout resistance decreased with increasing LP content. This finding suggests that the beneficial effect of LP could not compensate for the reduction in hydration products when a significant portion of cement was replaced by LP.

4.4. Optimized composition based on environmental evaluation

Based on the results of the environmental evaluation, several important trends can be identified. First, plain BRC (M2) did not provide a favorable balance between sustainability and mechanical properties at early ages. This was due to its slow hydration kinetics, leading to lower compressive strength and pullout resistance despite reduced CO_2 emissions. However, at later ages (90 days), the continued hydration of C_2S improved strength development, which in turn made BRC more eco-efficient. Second, incorporating LP proved to be consistently beneficial. Although higher LP contents reduced CS because of lower cement content, LP substitution improved the eco-efficiency indices by compensating for the lower pullout resistance of BRC. In particular, a 10% LP substitution in BRC (M3) emerged as the optimal composition. It not only lowered CO_2 emissions but also provided comparable or superior eco-efficiency indices relative to both OPC (M1) and plain BRC (M2). Overall, these findings demonstrate that the strategic combination of BRC with moderate LP substitution yields the best balance between environmental performance and mechanical properties. This balance is critical for practical applications where both sustainability and structural performance must be achieved simultaneously.

4.5. Effects of composition, fineness and hydration of clinkers on mechanical performances

In this study, three main variables including C_3S , C_2S , and LP contents were investigated to evaluate their effects on the interfacial bond characteristics of high-strength steel fibers embedded in mortar. These variables were controlled by blending commercial materials, including OPC and BRC supplied by Ssangyong Cement Industry Co., Ltd., and LP provided by Daejung Chemical Co., Ltd. The PSDs of OPC, BRC, and LP are presented in Fig. 2, while the phase compositions of C_3S , C_2S , C_3A , C_4AF , and LP in each matrix were summarized in Table 4. C-S-H, generated from the hydration of C_3S and C_2S , is the primary hydration product in cement-based materials (CBMs) and is mainly responsible for their mechanical performance [18,94]. C_3S and C_2S together account for a significant portion of cement clinkers, approximately 80%, as shown in Table 1. Therefore, variations in C_3S and C_2S contents in the matrices primarily governed the performance of the CBMs. In addition to the dominant influence of C_3S and C_2S , minor changes in CBM performances could also occur from other phases, such as C_3A , C_4AF , gypsum, and differences in PSD. The effects of these factors were discussed below.

For matrices without LP (M1 and M2), OPC exhibited a slightly finer PSD than BRC (Fig. 2), whereas the significant difference in composition of C_3S , C_2S and C_3A between OPC and BRC, as shown in Table 4. Therefore, the higher mechanical performance at early ages observed in the OPC-based matrix (M1) compared with the BRC-based matrix (M2) could be partially attributed to the finer PSD. However, the influence of PSD alone was insignificant because the differences in PSD between OPC and BRC were insignificant. In addition, although BRC particles were coarser than OPC, and finer particles were generally associated with faster hydration, M2 exhibited higher CS and pullout resistance. This behavior was attributed to the continued hydration of C_2S at later ages. Therefore, the primary factor governing the performance difference between M1 and M2 was the difference in phase composition rather than PSD. The proportions of C_3S , C_2S , C_3A , and C_4AF in M1 were 58.3%, 16.4%, 8.3%, and 9.7%, respectively, whereas those in M2 were 26.8%, 57.2%, 1.4%, and 8.7%, respectively. The difference in C_4AF content between M1 and M2 was negligible, and although the C_3A content in M1 was slightly higher than in M2, the most significant variation occurred in C_3S and C_2S contents. As C_3S increased, C_2S decreased correspondingly. Therefore, the analysis of mechanical performance in this study primarily focused on the effect of C_2S content in M1 and M2.

For the matrices containing LP (M2, M3, and M4), LP was added only to the BRC-based system. Because LP particles were finer than those of BRC, LP accelerated hydration through nucleation and dilution effects, in addition to chemical effects, as discussed in Section 4.3. As shown in Table 4, the differences in C_3A and C_4AF contents among these matrices were insignificant, whereas notable variations occurred in the C_3S , C_2S , and LP contents. As the LP content increased, the proportions of C_3S and C_2S decreased accordingly. Therefore, the decrease in CS and pullout resistance observed with increasing LP content were attributed primarily to the decreased C_3S and C_2S content. Consequently, LP content was selected as a key variable for analyzing the changes in mechanical performance among M2, M3, and M4.

Overall, the differences in PSDs among OPC, BRC, and LP were insignificant. OPC exhibited slightly finer particles than LP, while LP was finer than BRC. The primary differences among the matrices arose from variations in the proportions of C_3S , C_2S , and LP. The scope of applicability of this study was limited to systems using commercial OPC, BRC, and LP with similar proportions and PSDs.

5. Summary and conclusions

Herein, the interfacial bond characteristics of S and H steel fibers embedded in low-carbon mortar matrices with a w/b of 0.4 were investigated. The low-carbon matrices were prepared using OPC, BRC, and LP. The LP contents ranged from 0% to 30%, whereas C_2S contents

varied from 16.4% to 57.2%. In addition, the properties of FMZs were examined to clarify the sources of bond resistance. To achieve this, single-fiber pullout and compressive tests were conducted at curing ages of 14, 28, and 90 days, while SEM-EDS and NI were performed at 90 days. Based on the findings, the following conclusions can be drawn:

- LP enhanced early-age strength, whereas long-term strength development was governed by C_2S hydration. A 10% LP substitution enhanced early-age strength (57.7 MPa compared to 49.6 MPa without LP). However, at later ages (28 and 90 days), LP had limited effects, and the effects of C_2S on hydration dominated. Consequently, the matrix without LP showed the highest compressive strength.
- A 10% LP substitution in BRC is recommended for practical applications because it demonstrated a strength comparable to that of OPC at 14 days and significantly higher strength at 28 and 90 days. Additionally, this composition demonstrated a significant balance between sustainability and mechanical performance, achieving optimal results in terms of both compressive strength and pullout resistance at both early and later ages.
- Mortars with higher C_2S contents exhibited lower compressive strength at 14 days but significantly improved strength at 28 and 90 days owing to the long-term hydration of C_2S . At 14 days, matrices with 16.4% and 57.2% C_2S reached 57.6 and 49.6 MPa, respectively, whereas at 90 days, the trend reversed, with strengths of 69 and 96.3 MPa, respectively.
- Although higher C_2S contents decreased the peak pullout load and peak bond strength, the highest pullout energy and equivalent bond strength for S fibers were obtained with BRC containing 10% LP at 14 days. Mortars prepared with BRC without LP substitution (57.2% C_2S) exhibited lower pullout resistance at 14 days; however, they achieved the highest values across all pullout parameters at 90 days when compared with OPC (16.4% C_2S).
- Although increasing LP substitution in BRC reduced pullout energy and equivalent bond strength, the performance still remained superior to that of OPC.
- Microstructural analysis revealed denser FMZs with increasing LP content, with porosity decreasing from 25% to 18.9% as LP increased from 0% to 30%. NI also confirmed higher C-S-H contents in high- C_2S matrices, increasing from 34.6% to 51.5% as C_2S rose from 16.4% to 57.2%.
- H fibers consistently demonstrated higher pullout resistance than S fibers owing to additional mechanical-anchorage, whereas S fibers were more sensitive to variations in matrix composition.

Hence, replacing 10% LP in BRC is recommended as an effective alternative to OPC, as it balances early-age performance, long-term strength, and reduced CO_2 emissions. Although the use of BRC without LP showed superior performance in terms of both compressive strength and pullout resistance compared to OPC at 90 days, it still has disadvantages owing to later-age strength development. Therefore, this composition can only be applied to structures or components where high early strength is not required, such as mass concrete, pavements, foundations, and precast elements with extended curing periods. In contrast, BRC substituting 10% LP can overcome this limitation. Over 14 days, compressive strength and pullout resistance were slightly lower than those of plain OPC owing to the nucleation effect of LP, while superior performance was observed at 28 and 90 days owing to the continued hydration of C_2S . Furthermore, although replacing 30% LP in BRC resulted in significantly lower strength at 14 days, the differences with OPC became insignificant at 28 and 90 days. Although this mixture exhibited the highest reduction in CO_2 emissions, it remains a promising option for applications where sustainability is prioritized alongside moderate mechanical performance.

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CRediT authorship contribution statement

Van Thong Nguyen: Writing – original draft, Writing – review &

editing, Investigation, Validation, Methodology, Formal analysis, Data curation. **Van Phi Dang:** Writing – review & editing, Investigation. **Hyeon Woo Noh:** Writing – review & editing, Investigation. **Dong Joo KIM:** Writing – review & editing, Funding acquisition, Project administration, Supervision.

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.

Appendix A

The pullout behaviors of single S and H steel fibers after 14, 28, and 90 days of curing for four different mortar matrices are presented in Figs. A1–A6. A total of 10 specimens were tested for each series. The average pullout curves were obtained by averaging the responses of at least three of the best-performing specimens within each series. To generate the average curves, the slip was incremented from 0 to 15 mm in 0.01 mm steps. For each slip increment, the corresponding pullout load was determined by linear interpolation of the experimental data. The pullout loads were then averaged at each slip value to obtain the final average curves.

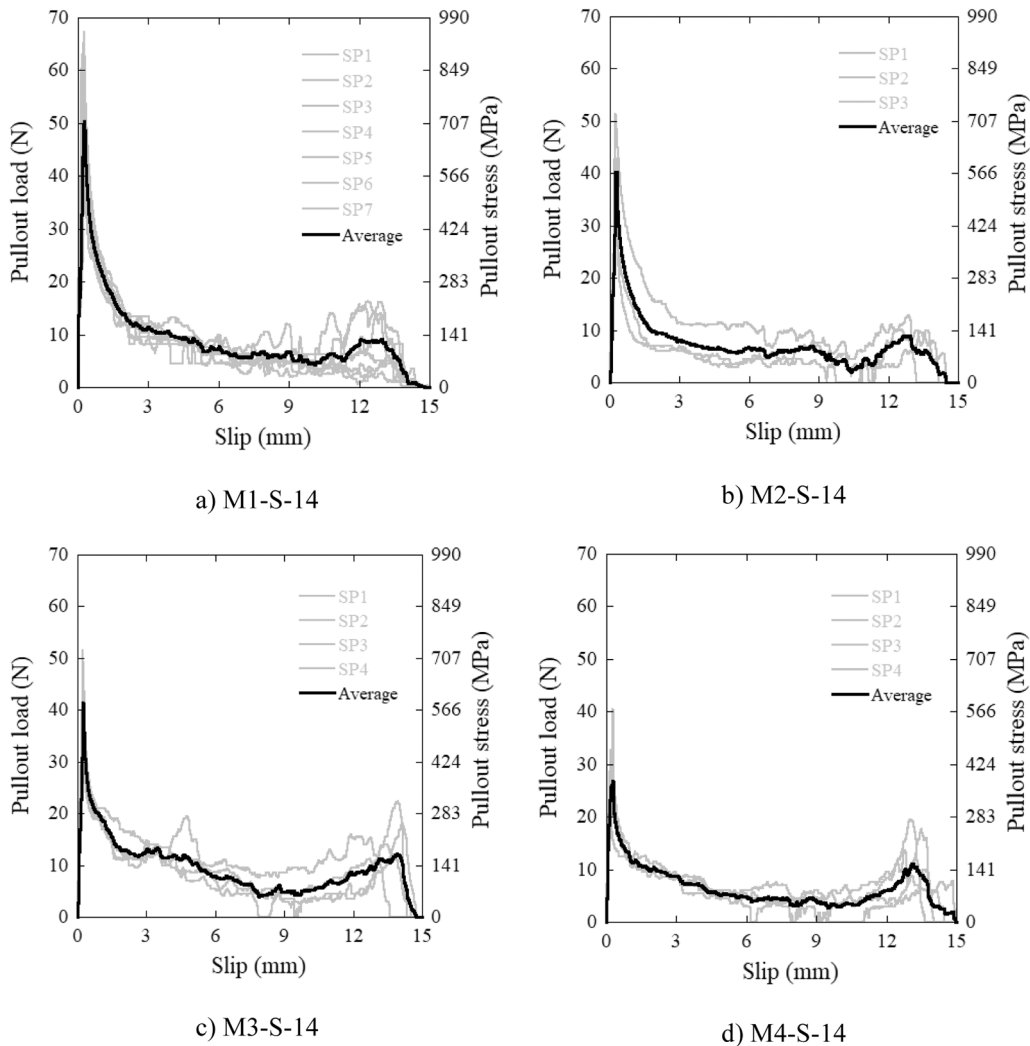


Figure A1. Pullout behavior of single-S fibers at 14 days

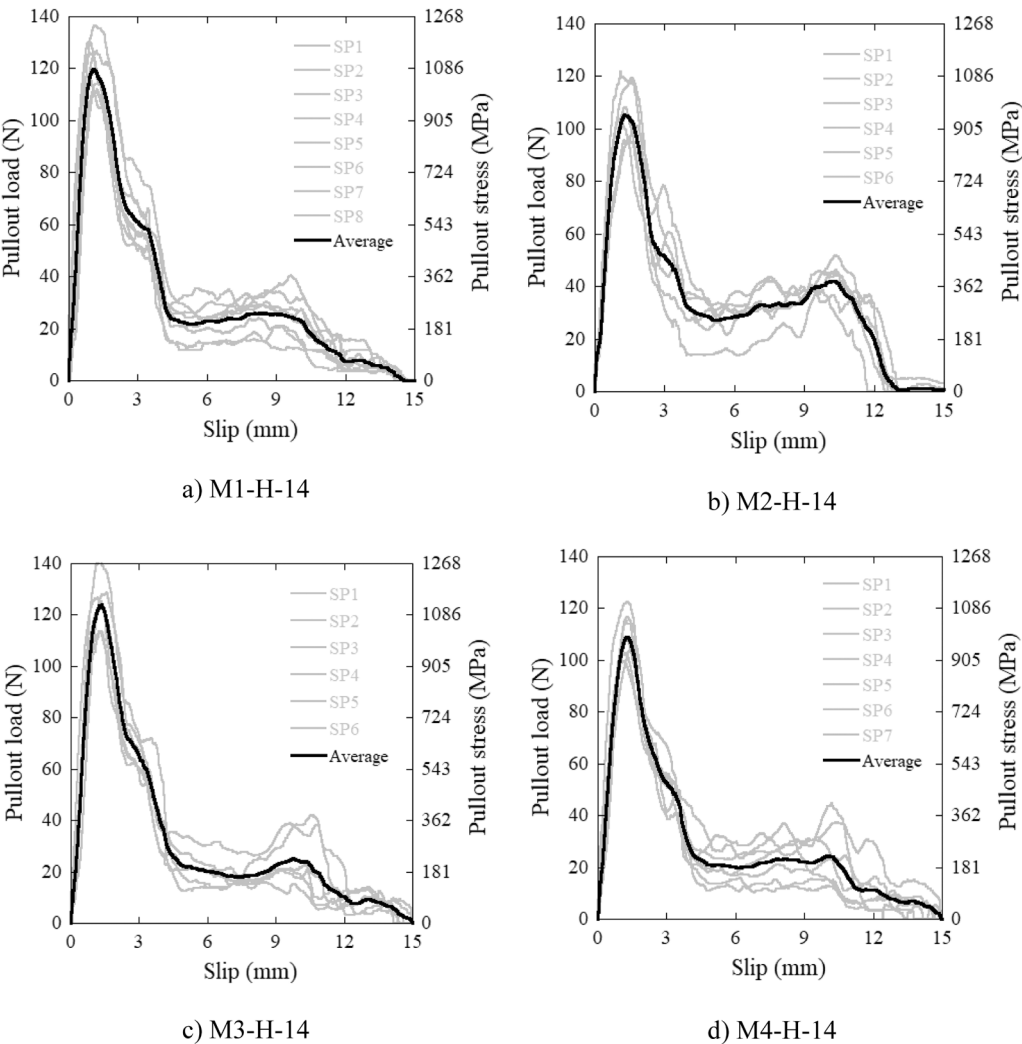


Figure A2. Pullout behavior of single-H fibers at 14 days

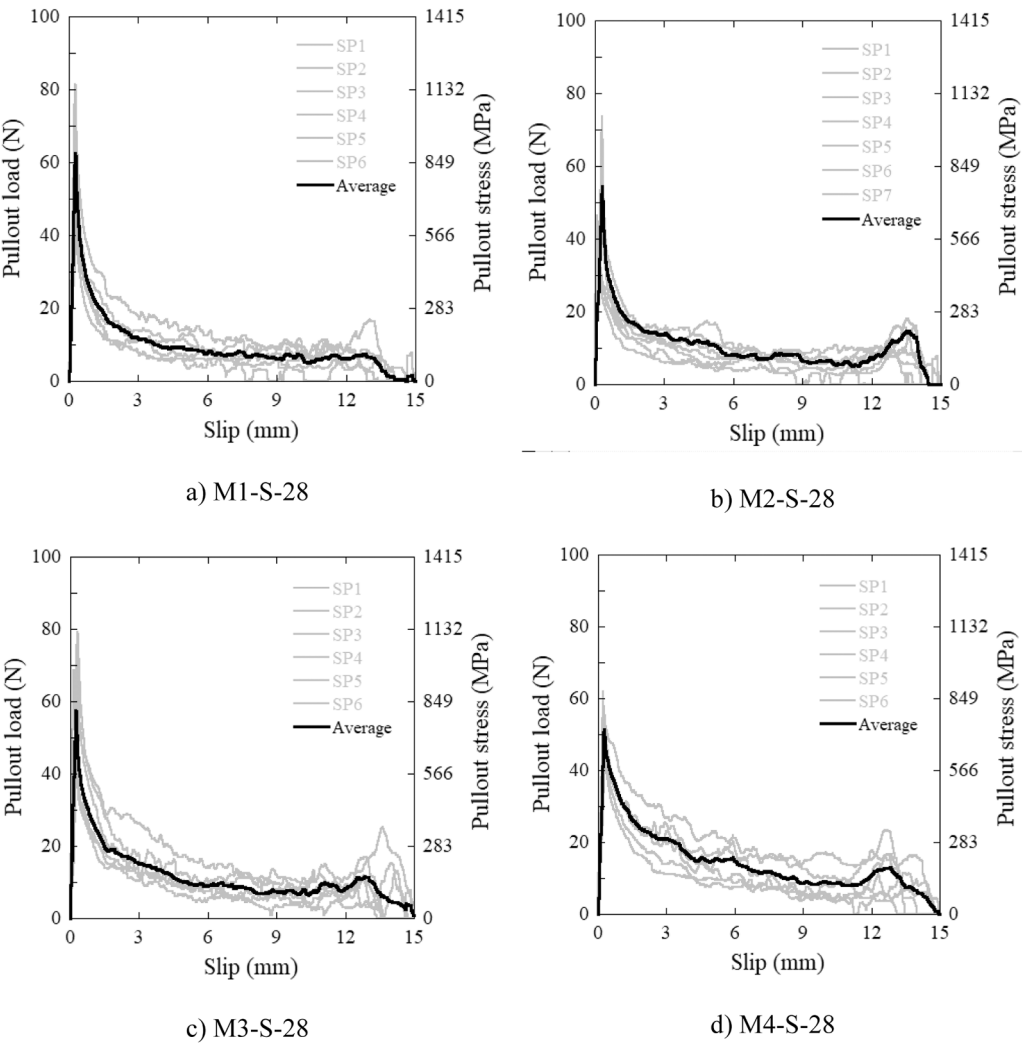


Figure A3. Pullout behavior of single-S fibers at 28 days

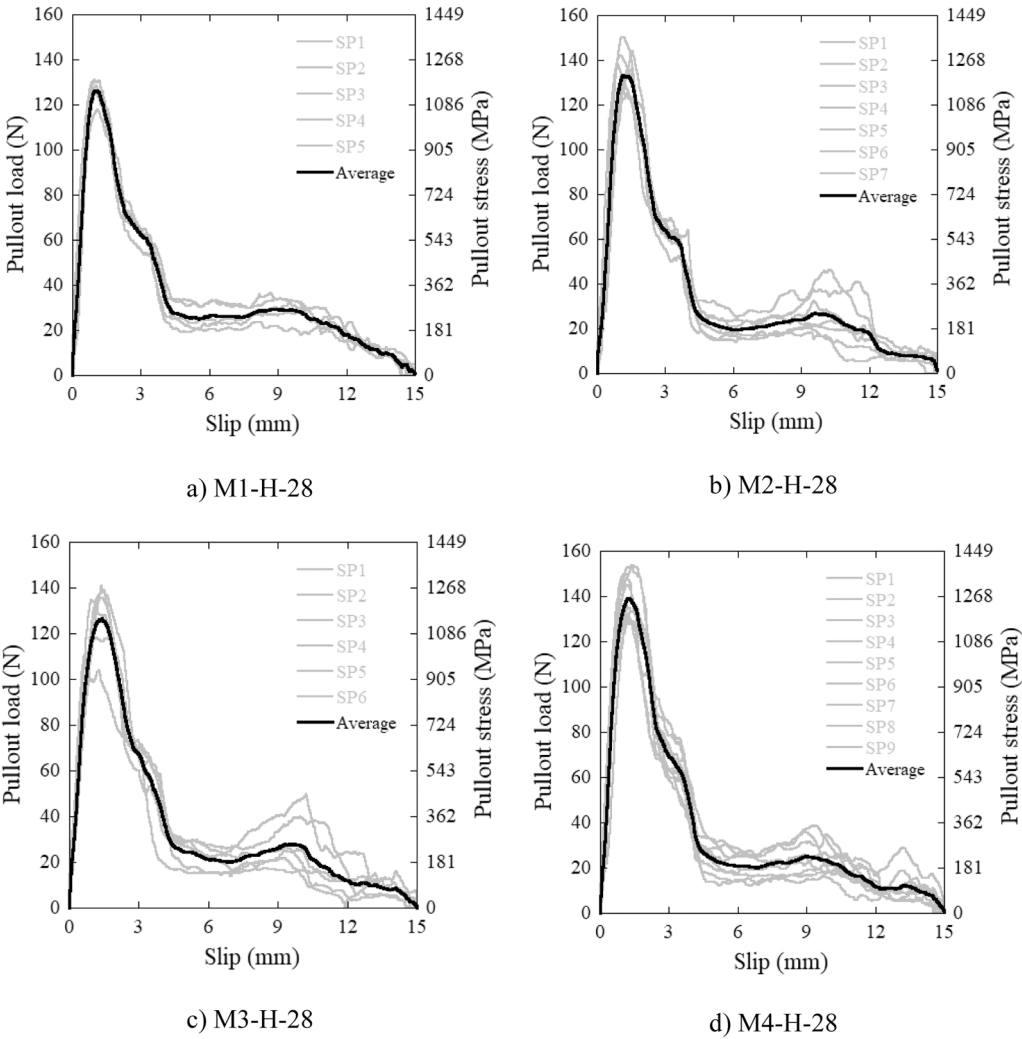


Figure A4. Pullout behavior of single-H fibers at 28 days

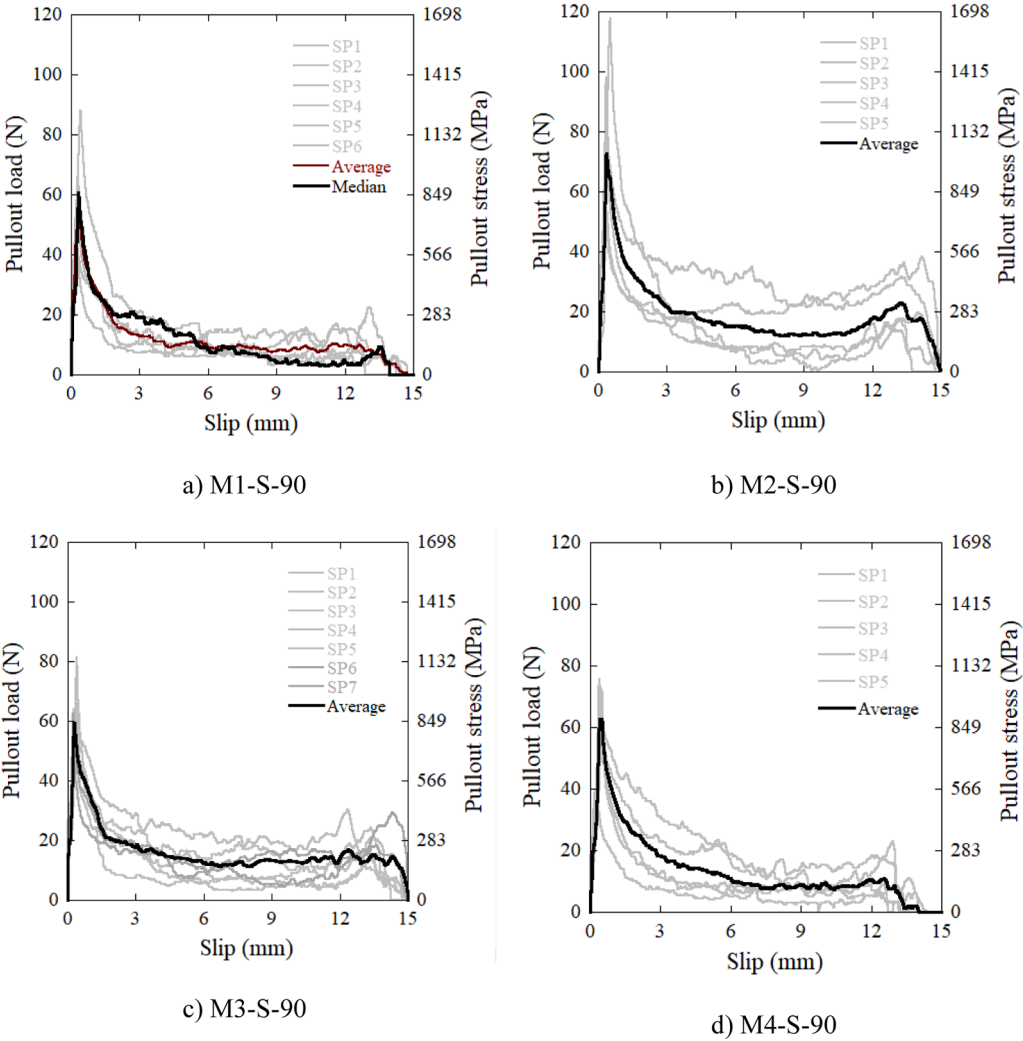


Figure A5. Pullout behavior of single-S fibers at 90 days

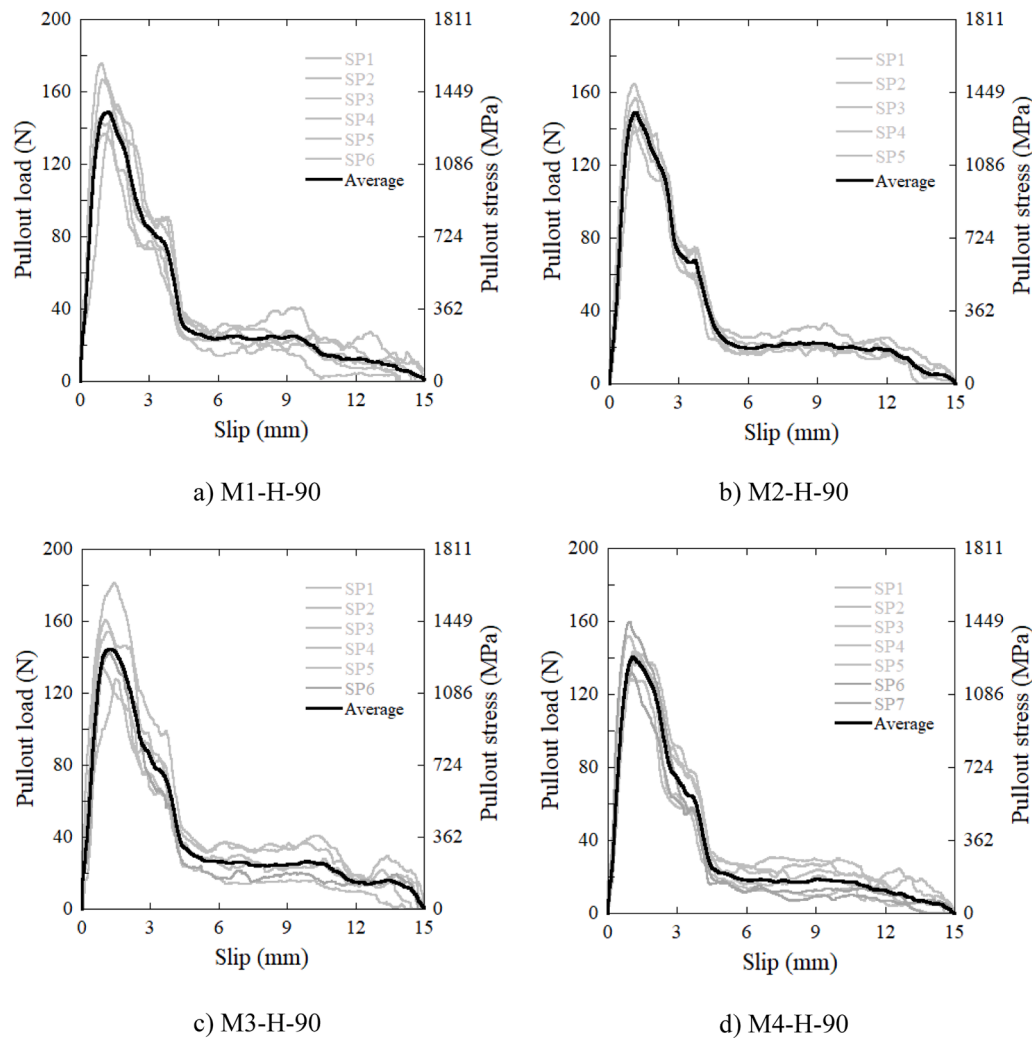


Figure A6. Pullout behavior of single-H fibers at 90 days

Data availability

Data will be made available on request.

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