The Influence of Chemical Activators on the Hydration Behavior and Technical Properties of Calcium Sulfoaluminate Cements Blended with Ground Granulated Blast Furnace Slags

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Abstract: The manufacture of Ordinary Portland cement (OPC) generates around 8% of the global CO₂ emissions related to human activities. The last 20 years have seen considerable efforts in the research and development of methods to lower the carbon footprint associated with cement production. Specific focus has been on limiting the use of OPC and employing alternative binders, such as calcium sulfoaluminate (CSA) cements, namely special hydraulic binders obtained from non-Portland clinkers. CSA cements could be considered a valuable OPC alternative thanks to their distinctive composition and technical performance and the reduced environmental impact of their manufacturing process. To additionally reduce CO₂ emissions, CSA cements can also be blended with supplemental cementitious materials. This paper investigates the influence of two separately added chemical activators (NaOH or Na₂CO₃) on the technical properties and hydration behavior of four CSA blended cements obtained by adding to a plain CSA cement two different ground granulated blast furnace slags. Differential thermal-thermogravimetric, X-ray diffraction and mercury intrusion porosimetry analyses were done, along with shrinkage/expansion and compressive strength measurements.

Keywords: low-CO₂ binders; calcium sulfoaluminate cements; supplementary cementitious materials; ground granulated blast furnace slag; hydration; expansion/shrinkage; mechanical strength

1. Introduction

Cement production requires a great number of raw materials and a large amount of energy and represents one of the leading causes of climate change due to greenhouse gas emissions, particularly CO₂. In 2019, the cement production all over the world was equal to around 4.10 billion tons and its contribution to the anthropogenic carbon dioxide emissions was estimated to be as high as 8% [1–3]. With a growing focus on energy efficiency and sustainable production [4], technological advancements in cement production have revealed methods to reduce CO₂ emissions in the following ways: (a) more efficient manufacturing processes; (b) targeted use of non-traditional fuels and raw materials (e.g., non-carbonated CaO sources instead of limestone); (c) carbon capture and storage technologies [5]; (d) increased utilization of blended Portland cements; (e) further use of special hydraulic binders (such as alkali-activated and magnesium-based cements as well as calcium sulfoaluminate (CSA), belite and belite–calcium sulfoaluminate (BCSA) binders) [6–23]).

CSA cements are obtained from non-Portland clinkers; ye’elimite (3CaO·3Al₂O₃·CaSO₄) is their principal component and, on the basis of the synthesis temperature as well as the proportioning and the type of raw materials (typically limestone, natural gypsum and bauxite), different amounts of calcium sulfates, belite (2CaO·SiO₂),
brownmillerite (4CaO·Al₂O₃·Fe₂O₃), ternerite (4CaO·SiO₂·CaSO₄) and various calcium aluminates are also present [24].

The technical behavior of CSA cements is mainly controlled by the fast ettringite generation (3CaO·Al₂O₃·3CaSO₄·32H₂O), produced upon hydration of ye’elimite and lime and/or calcium sulfates; in addition, the amount of CaSO₄ strongly influences the hydration rate as well as the phases generated upon the hydration of CSA cements [25–30]. During CSA hydration, aluminum hydroxide (Al₆O₁₃·3H₂O), monosulfoaluminate (3CaO·Al₂O₃·CaSO₄·12H₂O), strätlingite (2CaO·Al₂O₃·SiO₂·8H₂O) and calcium silicate hydrate (C-S-H, where C, S and H stand for CaO, SiO₂ and H₂O, respectively) can be formed. CSA binders are also characterized by high chemical and frost resistance. As a result, the molar ratio of CaSO₄ to CaO·3Al₂O₃·CaSO₄ (M ratio) is a crucial parameter influencing the behavior of these cements. An M-value lower than 1.5 is characteristic for CSA binders with rapid setting and early hardening; when M is in the range of 1.5–2.5, self-stressing characteristics appear. Finally, CSA binders display expansive properties if M is higher than 2.5 [31–33]. CSA binders are also characterized by high chemical and frost resistance. Furthermore, their manufacturing process, in comparison with ordinary Portland cement (OPC), exhibits very interesting environmentally friendly features, namely: (I) reduced limestone concentration in the clinker-generating raw meal (around 60% less); (II) lower synthesis temperature (about 1300 °C); (III) minor grinding energy required for cement milling. All these aspects imply energy saving and reduced CO₂ emissions [34]. Despite all these advantages, the high cost of CSA cements (depending on the elevated cost of bauxite) strongly limits their widespread use. In order to both reduce the cost and further lower the amount of CO₂ associated with their production process, CSA cements can be blended with supplementary cementitious materials (SCMs), such as natural materials (e.g., limestone, natural pozzolan and calcined clays), energy production residues (e.g., coal fly ash, FA) and industrial byproducts (e.g., ground granulated blast furnace slag, GGBFS).

GGBFSs are generated in blast furnaces during the reduction process (at about 1800 °C) of iron ores (IOs) using siderurgical coke (SK) and flux limestone. Molten iron represents the principal product of the blast smelting furnace; all the other IOs and SK components form a liquid slag (LS), which, due to its lower density, floats on the melt pig iron (MPI); LS is then easily separated from the MPI and water-quenched to produce a GGBFS (usually containing more than 95% of glassy phase). This byproduct is mainly used as mineral addition in composite cements; when stimulated with the addition of a proper activator (e.g., calcium and sodium sulfates, calcium hydroxide, sodium hydroxide and carbonate) [35–37], GGBFS exhibits a cementitious behavior (latent hydraulic activity) and/or pozzolanic features (ability to react with calcium hydroxide, Ca(OH)₂) [38–40]. In both cases, hydration products, such as C-S-H, C-A-H (calcium aluminate hydrate, where A stands for Al₂O₃), C-(A)-S-H (calcium (alumino)-silicate hydrate) and Mg₆Al₂(CO₃)(OH)₁₀·4H₂O (hydrotalcite, H₄), can be formed. While blended-Portland cements were extensively investigated in the past [41–46], far fewer studies have focused on blended CSA cements [47–57].

This paper was devoted to the exploration of the influence of two separately added chemical activators (NaOH or Na₂CO₃) on the hydration properties and technical behavior of four blended CSA cements containing two different GGBFSs. The hydration properties were evaluated using differential thermal-thermogravimetric (DT-TG), X-ray diffraction (XRD) and mercury intrusion porosimetry (MIP) analyses; the technical behavior was estimated by means of setting times and mechanical compressive strength (MCS) measurements as well as dimensional stability (DMS) tests. To obtain further insight into the influence of the two activators on both slags, the hydration behavior of four binary systems, each based on a different GGBFS and Na₂CO₃ or NaOH, was also evaluated.
2. Experimental Set-Up

2.1. Materials

A CSA industrial cement (R_C) was kindly supplied by the BUZZI UNICEM cement Company. Two GGBFSs (A and B) were used as cement substitutes, each generated in a pig iron plant, in South and Central Italy. The slags were finely pulverized in a laboratory planetary mill (Fritsch Pulverisette 6) in order to pass through a 90 μm sieve.

The chemical and mineralogical (only for R_C) composition together with the loss on ignition (l.o.i.) for R_C (containing 18% mass of natural anhydrite); A and B are indicated in Table 1.

Table 1. Chemical and mineralogical (only for R_C) composition for the binder components, wt%.

| Chemical Composition | Mineralogical Phase Composition | ICDD Ref. Number |
|----------------------|---------------------------------|------------------|
| R_C | A | B | R_C | Ye’elimite | 30-0256 | 43.0 |
| CaO | 44.58 | 41.02 | 37.32 | 43.0 |
| SiO₂ | 8.95 | 35.27 | 35.28 | β-belite | 33-0302 | 21.7 |
| Al₂O₃ | 22.42 | 11.85 | 11.17 | Celite | 38-1429 | 3.8 |
| FeO₃ | 1.86 | 0.88 | 1.03 | Anhydrite | 37-1496 | 19.1 |
| TiO₂ | 1.10 | 0.52 | 0.53 | Calcite | 05-0586 | 1.1 |
| K₂O | 0.30 | 0.58 | 0.34 | Brownmillerite | 30-0256 | 4.5 |
| MnO | 0.08 | 0.33 | 0.75 | Gehlenite | 73-2041 | 1.6 |
| Na₂O | 0.08 | 0.31 | 0.42 | Others | - | 5.2 |
| MgO | 0.94 | 7.15 | 10.55 | - | - | - |
| Cl⁻ | 0.07 | - | - | - | - | - |
| SO₃ | 16.85 | 0.33 | 1.80 | - | - | - |
| P₂O₅ | 0.05 | - | - | - | - | - |
| l.o.i * | 2.16 | 1.10 | 0.70 | 9.44 | 99.59 | 99.89 |
| Total | 9.44 | 99.59 | 99.89 | Total | 100.0 |

* Loss on ignition measured at 950 °C ± 25 °C.

The three materials were also characterized in terms of specific mass and surface (determined according to the EN 196-6) using a pycnometer and the Blaine apparatus, respectively. The particle size distribution for R_C, A and B was also determined by means of the Malvern Mastersizer 2000 (Malvern-Panalytical Italy, Milan, Italy) laser particle analyzer. Both NaOH (I) and Na₂CO₃ (C) were analytical reagents grade (Carlo Erba ≥99% purity).

Eight systems, namely the four binary and four ternary mixtures, were laboratory prepared. The binary mixtures contained the sole slag (A or B) and I or C. The ternary mixtures included R_C and A or B plus I or C. The composition of the binary and ternary mixtures is indicated in Table 2. R_C was employed as a control term.

Table 2. Composition of raw mixtures, wt%.

| AI | AC | BI | BC | R_AI | R_AC | R_BI | R_BC |
|----|----|----|----|------|------|------|------|
| R_C | - | - | - | 80.0 | 80.0 | 80.0 | 80.0 |
| A | 92.0 | 92.0 | - | - | 18.4 | 18.4 | - | - |
| B | - | - | 92.0 | 92.0 | - | - | 18.4 | 18.4 |
| I | 8.0 | - | 8.0 | - | 1.6 | - | 1.6 | - |
| C | - | 8.0 | - | 8.0 | - | 1.6 | - | 1.6 |
2.2. Hydration Procedures, Setting Times and Expansion/Shrinkage Measurements

R_C, the four binary and the four ternary mixtures were hydrated using a water/solid (w/s) mass ratio equal to 0.50; the pastes were then poured into cylindrical molds (30 mm in diameter and 15 mm high), which were put in a thermostatic bath at 20 °C. At each established aging period, the hardened binary pastes were finely grounded for DT-TG measurements; the other cylinders were broken in half: one part was submitted to MIP analyses, the other grounded for DT-TG and XRD measurements. Acetone (to stop hydration) and diethyl ether (to remove water + acetone) were employed for the treatment of all the specimens, which were subsequently left in a desiccator containing soda lime and silica gel to ensure protection against atmospheric CO2 and H2O. The DT-TG and XRD analyses were conducted on both binary and ternary mixtures cured from 2 to 180 days. The MIP studies were carried out on samples, aged from 2 up to 180 days. Setting times were determined on CSA-based cement pastes using the Vicat apparatus according to the EN 196-3. For the DMS measurements [17], CSA-based pastes were shaped as small prisms (15 × 15 × 78 mm); the prisms were left at 20 °C in the air for 8 h and then demolded. One set of samples was kept in still water at 20 °C; the remaining samples were stored in a climate chamber at 20 °C and 50% R.H. The length variations were evaluated as the average values of four measurements taken with a caliper (accurate to ± 1 μm); the reference length was measured after demolding [17].

2.3. Mechanical Tests

The MCS tests were carried out on CSA cement-based mortars prepared in accordance with the EN 196-1 and aged for periods ranging from 2 to 180 days. Three prisms (six determinations) were tested for each CSA-based mortar at each aging period.

2.4. Characterization Techniques

2.4.1. X-ray Fluorescence (XRF) Analysis

XRF analysis was employed to evaluate the chemical composition of the investigated materials (Table 1) and was performed using an energy dispersive BRUKER Explorer 54 apparatus (Bruker Italy, Milan, Italy; maximum power = 1 kW).

2.4.2. Simultaneous DT-TG Analysis

Simultaneous DT-TG analysis was carried out with a NetzchTasc (Netzch Italy, Verona, Italy) 414/3 apparatus operating in the air in the 20–1000 °C temperature interval (at a heating rate of 10 °C/min). To quantify the chemically bound water (CBW), TG analysis was also used; CBW was calculated from the mass loss from 30 °C to 600 °C and normalized to 100 g of dry binder.

2.4.3. XRD Analysis

XRD analysis was conducted using a BRUKER D2 PHASER diffractometer (Bruker Italy, Milan, Italy; CuKα radiation and 0.02°2θ s−1 scanning rate); it was equipped with EVA software for the estimation of the mineralogical composition and the amorphous content of the two GGBFSs and the phase composition of cements hydration products. The quantitative mineralogical composition for R_C- and CSA-based hydrated cements (of which the results were normalized to 100% of the accounted crystalline phases) was determined using the Rietveld method. Table 1 also lists the phases used for the Rietveld refinement and their ICDD (International Centre for Diffraction Data) codes. For the Rietveld refinement, the TOPAS software was utilized.

2.4.4. MIP Analysis

The porosity measurements were executed with a Hg-porosimeter (Thermo-Finnigan Pascal 240 Series, Thermo Electron S.p.A., Milan, Italy, EVA software v 4.0, maximum
pressure 200 MPa); it was equipped with a low-pressure unit (140 Series, operating between 100 and 400 kPa) able to generate a high vacuum level (10 Pa).

3. Results and Discussion

From the chemical analysis (Table 1), it can be argued that CaO, Al₂O₃, SO₃ and SiO₂ are, in the order, the main oxides present in R_C. QXRD values clearly show that ye’elimite, belite, anhydrite, brownmillerite, celite (3CaO·Al₂O₃) and gehlenite (2CaO·Al₂O₃·SiO₂) are, in sequential order, the main mineralogical phases for R_C. Table 1 also indicates that CaO and SiO₂ are the main oxides for both the GGBFS samples, while Al₂O₃ and MgO can be considered as secondary components. Furthermore, the quality factor ($Q_f = (\text{CaO} + \text{MgO} + \text{Al}_2\text{O}_3)/(\text{SiO}_2 + \text{TiO}_2)$), the hydration modulus ($H_m = (\text{CaO} + \text{MgO} + \text{Al}_2\text{O}_3)/(\text{SiO}_2)$) and the coefficient of basicity ($K_b = (\text{CaO} + \text{MgO})/(\text{SiO}_2 + \text{Al}_2\text{O}_3)$) values for A and B are equal to 1.67 and 1.65, 1.70 and 1.67 and 1.02 and 1.03, respectively. These results allow for the classification of both the slags as hydraulic (as their $H_m$ is higher than 1.4) and neutral (as $0.9 < K_b < 1.1$) [58]. The mineralogical composition for the two slags is reported in Figure 1.

![Figure 1](image_url)

**Figure 1.** XRD patterns for A (a) and B (b). Symbols: $\$ = calcite (CaCO₃); # = akermanite (2CaO·MgO·2SiO₂); @ = anhydrite (CaSO₄).

The XRD patterns for A (Figure 1a) and B (Figure 1b) are quite similar; their glassy nature, responsible for their reactivity, is highlighted by a broad hump located in the 2θ range 14°–37°. The glassy content (evaluated by adding 10% mass of corundum as internal standard in each sample) was equal to 96% and 98% mass for A and B, respectively. The presence of crystalline phases (calcite, anhydrite and akermanite) is clearly evident in the diffraction profiles.

The specific mass and Blaine fineness for R_C, A and B were equal to 2.75, 2.78 and 2.83 g/cm³ and 4500 ± 50, 4200 ± 50 and 4300 ± 50 cm²/g, respectively. Figure 2 reports the particle size distribution for R_C, A and B in terms of the cumulative distribution percentage.
Figure 2. Particle size (cumulative) distribution for R_C, A and B.

The DT outcomes for the four binary mixtures, hydrated between 2 to 180 days, are displayed in Figure 3.

Figure 3. DT for Al (a), AC (b), BI (c) and BC (d) hydrated pastes (w/s = 0.50) for 2, 28, 90 and 180 days. Symbols: # = calcium silicate hydrates/calcium silico (aluminate) hydrates (C-S-H/C-(A)-S-H); * = hydrotalcite; & = calcite (CaCO₃); % = crystallization slag effect.

Five endothermic peaks were observed for all the investigated curing periods [59]. The first peak (T = 135 °C ± 4 °C) was assigned to the dehydration of C-S-H/C-(A)-S-H, while the second (T = 247 °C ± 4 °C), the third (T = 382 °C ± 8 °C) and the fourth (T = 551 °C ± 5 °C) correspond to the dehydration, dehydroxylation and decarbonation of H₄. The
last endothermal peak \((T = 735 \, ^\circ C \pm 17 \, ^\circ C)\) is related to the decarbonation of calcite. In addition, an exothermic effect was recorded at \(894 \, ^\circ C \pm 20 \, ^\circ C\) due to the crystallization of the slags. The growth of the endothermal peaks is clear when the curing period is increased. The TG curves (not reported here) revealed a mass loss increase due to ongoing hydration.

Figure 4 shows the DT curves for R_C, R_AI, R_AC, R_BI and R_BC cured between 2 and 180 days. The DT plots display that the hydration behavior of the explored systems is mainly governed by the reaction of \(3\text{CaO} \cdot 3\text{Al}_2\text{O}_3 \cdot \text{CaSO}_4\) with \(\text{CaSO}_4\) to form \(3\text{CaO} \cdot \text{Al}_2\text{O}_3 \cdot 3\text{CaSO}_4 \cdot 32\text{H}_2\text{O}\).

**Figure 4.** DT for R_C (a) R_AI (b), R_AC (c), R_BI (d) and R_BC (e) hydrated pastes \((w/s = 0.50)\) for 2, 28, 90 and 180 days. Symbols: # = calcium silicate hydrates/calcium silico (aluminate) hydrates; ® = ettringite; § = aluminum hydroxide; * = hydrotalcite; $ = magnesite \((\text{MgCO}_3)\); & = calcite.

Ettringite, aluminum hydroxide (identified only in R_C), magnesite and calcite were identified by the following fourth dehydration endothermal peaks [59]: 155 \(^\circ C \pm 2 \, ^\circ C\), 274
Furthermore, three peaks, related to the H₃S, were observed in the CSA-blended cements. C-S-H and C-(A)-S-H, coming from the hydration of the activated slags, were not evident due to the overlapping of the ettringite DT signal.

The hydration rates for R_C, R_AI, R_AC, R_BI and R_BC were also estimated through the calculation of the amount of CBW (Figure 5).

**Figure 5.** CBW determined up to 180 days of curing (w/b = 0.5 and normalized to 100 g of dry binder) for R_C, R_AI, R_AC, R_BI and R_BC.

Figure 5 clearly demonstrates that the five binders follow an analogous trend during the first 7 days of hydration; in this period, the hydration rate was very fast thanks to the rapid formation of ettringite. After about one month of curing, the R_C curve displays an almost steady value, while those for the blended cements show a slight linear growth, owing to the continuous hydraulic reaction of the slags.

The hydration process was also estimated by means of QXRD analysis; it allowed for the evaluation of the reduction of reactants, the development of new products and the presence of inert phases between 2 and 180 days. Figure 6 shows the evolution of ettringite, hydrotalcite and belite concentrations, as a function of curing times. On the whole, it can be concluded that QXRD results strongly agree with the DT analyses outcomes. High amounts of ettringite were already present after two days of curing in all CSA-based cements; at 28 days, the ettringite concentration reaches its maximum value and remains almost constant until 180 days of curing. No ye’elimite was observed after since two days of curing; however, the presence of belite was found in the five systems at all the investigated curing periods, and its concentration value gradually decreases with the increase in curing times.
As expected, peaks related to the presence of H₄, formed thanks to the hydration reaction of A or B, were found in the XRD patterns for the four ternary systems; in particular, the concentration values for H₄ slightly rose with the increase in hydration age.

The setting times for the CSA-based systems are reported in Figure 7.

The initial and final setting time for R_C are equal to 10.0 and 18.0 min, respectively; the initial setting times for blended cement pastes increase to 15.0 min for R_AC and R_BC and to 15.5 and 16.0 min for R_AI and R_BI, respectively. Similarly, the final setting times for CSA-blended cements also increase, being equal to 19.0 min for R_AC, 20.0 min for R_AI and R_BI and 20.5 min for R_BC. Consequently, it can be argued that the delay of setting times for blended CSA cements only depends on the presence of the slags; however, the blended CSA binders can be still considered as rapid setting cements.

The MCS results for the five CSA-based mortars are reported in Figure 8; as expected, for all the systems, the mechanical strength values continuously increase with an increase in curing time.
At the shortest curing period (2 days), the blended cements exhibit compressive strength values about 22% lower than mortars only containing R_C; this is due to the dilution effect of the GGBFS + activator. Subsequently, as curing time increases, the MCS gap reduces thanks to the continuous hydration of the chemically activated slag. Finally, after 180 days of curing, the blended cements containing Na\textsubscript{2}CO\textsubscript{3} show almost the same MCS values as the reference system; on the contrary, the MCS values for the others mortars are almost 8% lower than those for R_C-based mortar for the same curing period.

Figure 9 reports the porosimetric plots for the CSA-based binder pastes cured at 2, 28 and 180 days; they display the cumulative curves for the intruded Hg volume vs. pore radius. The plots show that the total intruded volume reduces with the increase in curing time. For R_C (Figure 9a), the total cumulative volume (TCV) reduced from about 173 mm\textsuperscript{3}/g at 2 days of curing to 70 mm\textsuperscript{3}/g after 6 months. In the same period, the TCV, respectively, decreases of around 68% (from 243 mm\textsuperscript{3}/g to 79 mm\textsuperscript{3}/g) and 69% (from 218 mm\textsuperscript{3}/g to 67 mm\textsuperscript{3}/g) for the R_AI and R_AC systems (Figure 9b,c). As far as the two ternary B-based systems are concerned, the TCV decreases of about 63% (from 213 mm\textsuperscript{3}/g to 79 mm\textsuperscript{3}/g) for R_BI and 72% (from 195 mm\textsuperscript{3}/g to 55 mm\textsuperscript{3}/g) for R_BC.
Finally, Figure 10 shows the total porosity percentage (TPP) for the five systems as a function of curing time.

The curves highlight an analogous trend for all the cement pastes, whose TPP values significantly vary between 2 and 28 days. However, no relevant changes for TPP were observed at longer curing times. In particular, at 2 days of hydration, the lowest and the highest TPP values were shown by R_C (30.5%) and R_BI (36.2%), respectively. Finally, at the longest curing time, the total porosity for the ternary systems containing NaOH was...
higher (17.6% for R_AI and 20.8% for R_BI) than that for R_C (15.6%); on the other hand, when Na₂CO₃ was employed, the TPP values for R_AC (14.5%) and R_BC (14.7%) were slightly lower than that of the reference system.

Figure 11 displays the expansion (E)/shrinkage (S) curves (in terms of length changes) for R_C, R_AI, R_AC, R_BI and R_BC.

![Figure 11. DMS curves for CSA-based pastes (air- and water-cured).](image)

The E and S curves show that the five binders differ very little from each other, both when cured in the air and submerged under water. Similarly to R_C, when left in the air, the four blended cement pastes display continuous and low shrinkage for the first 20 or so days of curing; for all the systems, the maximum value of steady shrinkage is in the range of ~0.1% (for R_C) and ~0.2% (for R_BC). As far as water curing is concerned, the highest expansion values were achieved after about 40 days of hydration; they were in the range of 0.06% (for R_BC) to 0.14% (for R_C).

4. Conclusions

This paper aimed to investigate the influence of two ground granulated blast furnace slags (GGFBs, A and B), activated by NaOH or Na₂CO₃, on the performances of GGBFS-based calcium sulfoaluminate (CSA) cements. The use of GGBFS as a secondary cementitious component reduces the costs of CSA cements and enhances their environmentally friendly features, mainly in terms of decreased CO₂ emissions and high energy saving associated with their manufacturing process.

The hydration tests, carried out on four blended CSA cements (each containing 18.4% of the mass of A or B and 1.6% of the mass of I or C), indicated that the ettringite concentration was lower than that found in the plain CSA binder (R_C); however, this gap was outweighed by the formation of other products (e.g., hydrotalcite, calcium (alumino) silicate hydrate) from the hydration of the activated slags and forming at curing periods longer than 56 days. This aspect was particularly evident in the systems containing Na₂CO₃.

Furthermore, GGBFS-based CSA cement mortars demonstrated compressive mechanical strength values lower than the R_C mortars for curing times up to 56 days and almost equal for longer curing times. Specifically, at 90 and 180 days of curing, the best mechanical performances were obtained by mortars containing Na₂CO₃; these results were confirmed by the porosimetric investigations. Similarly to R_C pastes, the blended CSA cements exhibited negligible expansion and shrinkage values, both when submerged in water and exposed to the air, thus confirming their good dimensional stability. On the whole, the properties of the investigated blended cements systems are influenced by the
type of the added chemical activator; as a matter of fact, systems containing Na₂CO₃ displayed better technical performances than those with NaOH. Therefore, Na₂CO₃ can be recommended as a GGBFS activator in the CSA-blended cements, because it is cheaper, safer and easier to manage than NaOH.

Author Contributions: A.T.: Conceptualization; methodology; formal analysis; investigation; data curation; writing—original draft preparation; writing—review and editing. M.M.: Conceptualization; methodology; formal analysis; investigation; data curation; writing—original draft preparation; writing—review and editing. All authors have read and agreed to the published version of the manuscript.

Funding: This research received no external funding.

Institutional Review Board Statement: Not appropriate.

Informed Consent Statement: Not applicable.

Data Availability Statement: The data presented in this study are available on request from the corresponding author.

Conflicts of Interest: The authors declare no conflict of interest.

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