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Sintered and glazed glass-ceramics from natural and waste raw materials

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Abstract

Monolithic sintered glass-ceramics and porous glass-ceramic components were produced from natural raw materials together with industrial waste. The waste, including fly ash from thermal power plants and metallurgical slags, either was mixed with natural raw materials (e.g., clay and silica sand) and vitrified or was directly sintered, after dry pressing at 40 MPa. Fine powders of waste-derived glasses were also converted into sinter-crystallized glass-ceramics, after dry pressing at 40 MPa and subjection of the pressed powders to controlled heating cycles, producing dense components with bending strength as high as ~80 MPa.

The two types of waste-derived materials, namely, ceramic tiles from direct sintering and sinter-crystallized glass-ceramics, were combined to obtain double-layered glass-ceramics. These featured a dense wear-resistant coating on a porous substrate and possessed mechanical properties that make them suitable, for example, as structural lightweight panels in building facades.

Keywords: D. Glass-ceramic; Waste; Sinter-crystallization

1. Introduction

Vitrification has proved to be the safest approach for the treatment and disposal of non-combustible hazardous waste [1]. The process involves the thermal destruction of waste and, when properly formulated, the resulting glass exhibits a high degree of chemical inertness and can therefore be disposed of in landfills without any particular concerns arising. Despite the soundness of vitrification technology, confirmed by numerous scientific studies and experimental tests [2], this approach has encountered difficulties in establishing itself, being particularly cost- and capital-intensive. In the case of inorganic waste, for which (in contrast to radioactive waste), environmental safety is not an absolute priority over cost, vitrification may be justifiable only if the glass obtained can be reused in high-value applications.

The most significant application of waste-derived glasses is undoubtedly in glass-ceramics, starting with the well-known Russian "slag sitals" (developed nearly 50 years ago), based on the vitrification of metallurgical slags. Slag sitals may be considered as prototypical glass-ceramics derived from glasses belonging to the $CaO–Al_2O_3–SiO_2$ (CAS) system [3], since they combine good mechanical properties (bending strength and abrasion resistance) and chemical stability, making them suitable for their main application in the construction industry, namely, as facades for buildings.

The production of high-value glass-ceramics from inorganic wastes incurs additional costs, however: for example, the production of slag sitals involves rolling molten waste-derived glass into sheets, together with a secondary ceramization treatment. Recent investigations have looked into ways of modifying the production process of waste-derived CAS glasses in order to reduce costs. In particular, sintering of glass powders, leading to "sinter-crystallized" glass-ceramics, has attracted much attention [4]. Glass powders can be obtained easily by grinding glass fragments possessing high internal stresses, such as those produced by pouring a glass melt into water to produce

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a frit. With this approach, energy savings are associated with both the vitrification and ceramization steps, since waste glasses may be poured just after homogenization, thus avoiding the expensive refining step, and crystallization may occur very rapidly (in 1 h or less, at a sufficiently high sintering temperature) as a result of surface nucleation, even for glasses with a very low content of oxides that can act as nucleating agents (i.e., oxides with poor solubility in glass, such as TiO_2 and ZrO_2) [5].

Direct sintering of mixtures of inorganic waste including recycled glasses, which act as fluxes, is an important alternative. The products cannot strictly be considered as glass-ceramics, since no vitrification, i.e., a stage in which the starting raw materials are mixed and converted into a glass, is involved. However, there is much evidence in the literature supporting the classification of such products as "sintered glass-ceramics", owing to their observed phase evolution [6,7]. In fact, the recycled glass component, besides promoting an increase in density as a result of viscous flow sintering, reacts with the waste, leading to the formation of silicate and aluminosilicate crystals similar to those produced by devitrification of waste glasses. The process, clearly resembling that occurring in traditional ceramics, offers remarkable energy savings, due to the absence of a high-temperature (> 1350–1400 °C) melting stage and its simplicity. The impact on the chemical stabilization of waste, however, is somewhat controversial: on the one hand, treatment at moderate temperatures prevents the formation and escape in gaseous form of toxic compounds (e.g., hydrogen fluoride from waste containing fluorine compounds [8]); on the other hand, the chemical homogeneity of a sintered solid is much lower than that of a glass obtained from melting (pollutants may selectively accumulate in regions not completely dissolved in the liquid phase provided by the flux).

The production process described in this paper has the aim of combining direct sintering of waste mixtures and sintering of waste-derived glasses, with the creation of layered "hybrid glass-ceramics" (see Fig. 1 below). The good mechanical properties and homogeneous microstructure of sinter-crystallized glass-ceramics were also exploited in a glaze, white or colored, deposited on a porous base body obtained from direct sintering. Vitrification of waste is sustainable, since it is applied only to a limited amount of the starting materials and, in addition, the single firing reduces the costs associated with glaze deposition. Owing to their negligible water absorption (on the glazed side) and their low density, these layered glass-ceramics could be a viable alternative to the lightweight tiles currently used in building facades.

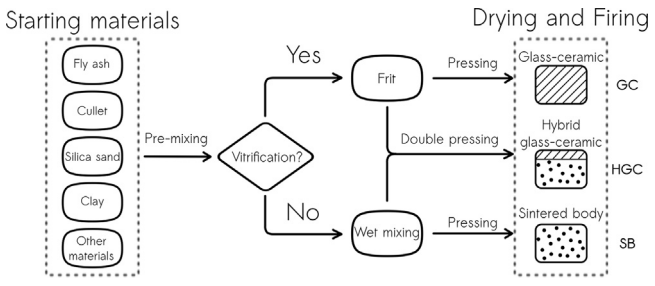

Fig. 1. Schematic representation of three routes for the production of ceramic materials employing selected natural and waste raw materials.

2. Experimental

The starting waste raw materials for the production of the sintered body and the glass-ceramics in this study were pozzolanic waste and fly ash residues from a cement plant in the Riyadh region of Saudi Arabia, soda lime cullet (the fraction of recycled material that is difficult to reuse in conventional glass production owing to its impurities), and panel glass from dismantled cathode ray tubes (CRTs). The other starting natural raw materials were ball clay, limestone, and silica sand from the Riyadh region.

Table 1 gives the data from X-ray fluorescence analysis (Philips XRF Sequential Spectrometer PW2400, Eindhoven, The Netherlands) of the starting raw materials and the percentages of raw materials employed in the preparation of the different types of samples: sintered body (SB) and glass-ceramic 1 (GC1), with the same composition, and glass-ceramic 2 (GC2).

The raw materials for SB were first ground separately into fine powders using a ball mill and then sieved; only particles with a diameter below 90 μm were kept. About 300 g in total of the ground raw materials were mixed thoroughly in a food mixer, adding 30 wt% of water. After mixing, the homogeneous mixture was dried in an oven at 80 °C for 24 h and then ground again, keeping only the particles with a diameter below 90 μm. Sintering experiments were performed on samples of SB prepared by pressing 15 g of the dry powder mixture in a rectangular die (cross-section 50×34 mm^2) at a pressure of 30 MPa. The samples were further dried at 120 °C for 30 min and then fired at 1000, 1100, 1150, and 1250 °C for 60 min at a heating rate of 10 °C/min.

Glasses with compositions GC1 and GC2 were obtained by melting at 1300 °C for 2 h and then pouring into cold water. The frits were collected, dried at 80 °C, and then ball-milled in an agate jar and sieved to obtain particles with a size below 90 μm.

Dilatometric analysis and differential thermal analysis (DTA) (DSC 404, Netzsch Gerätebau GmbH, Selb, Germany, 10 °C/min heating rate) were performed on both fine powders (< 90 μm) and coarse powders (> 90 μm) of GC1 glass in order to determine the optimum conditions for processing. Preliminary sintering experiments were performed on disks of 30 mm diameter obtained by uniaxially pressing 2 g of powder at 40 MPa. The disks were fired at different temperatures (900, 950, and 1000 °C) for different periods (30 and 60 min). After selection of the preferred firing temperature and time, two samples were obtained by uniaxially pressing 15 g of dry powder GC1 in a rectangular die (with a cross-section of 50×34 mm^2) at a pressure of 30 MPa without adding binders. The samples were fired at 950 °C for 30 min at a heating rate of 40 °C/min.

A hybrid glass-ceramic (HGC1) was obtained by first lightly pressing (at 10 MPa) 15 g of the dry powder SB mixture in a rectangular die (again with a cross-section of 50×34 mm^2)

Table 1
Chemical composition of raw materials and formulations of glass-ceramics.

Oxide	Silica sand	Limestone	Clay	Fly ash 1	Fly ash 2	Pozzolanic waste	Cullet	CRT panel glass	SB/GC1	GC2
Chemical composition (wt%)										
SiO_2	98.6	2.3	41.6	14.2	12.8	47.4	71.9	57.3	55.7	60.3
Al_2O_3	0.6	0.9	39.1	4.8	4.5	15.9	0.7	3.8	12.8	11.3
P_2O_5			0.1			0.5				0.1
Na_2O	0.1		0.1	0.1	1.2	3.4	13.3	12.8	2.6	2.2
K_2O		0.1	0.1	0.5	14.5	1.0	0.1	7.2	4.1	0.4
MgO		0.6	0.2	0.7	0.6	9.6	3.3		0.8	2.6
CaO		52.7	0.5	58.3	50.4	7.7	10.1		22.0	18.6
BaO								7.9		
SrO								8.4		
Cr_2O_3						0.1				
Fe_2O_3		0.3	1.3	2.8	2.2	11.4	0.1	0.2	1.0	3.4
TiO_2	0.4		2.9	0.3	0.2	1.9		0.4	1.0	1.1
MnO_2						0.2				
ZnO								0.6		
ZrO_2								1.4		
LOI	0.3	43.1	14.1	18.3	13.6	0.9	0.5			
Formulation (wt%)										
SB/GC1	25	10	25		25		15			
GC2	30		15	25		20	10			

and then depositing 3 g of a glass frit on this substrate at a surface density of 0.176 g/cm^2. The layered sample was then uniaxially pressed at 30 MPa. In contrast to the successful results of this procedure, it was found that pressing the substrate at 30 MPa, followed by frit deposition and a second pressing at 30 MPa led to limited adhesion between frit and substrate, probably because of limited interpenetration between the glaze and the substrate during pressing.

The glass frit for HGC1 was produced by mixing a GC1 frit (40 wt%) with additives in order to obtain a white color and to provide a perfect match in terms of physical properties (e.g., coefficient of thermal expansion) between top layer and substrate. The additives were a whitening agent (zircon, $ZrSiO_4$, 0.8 μm mean particle size, Industrie Bitossi SpA, Vinci, Italy, 20 wt%) and a flux (CRT panel glass, 40 wt%). Another hybrid glass-ceramic (HGC2), with a green color, was produced by using a GC2 frit without additives. Both HGC samples were fired at 1150 °C for 60 min, at a heating rate of 10 °C/min.

Young's moduli of the glass-ceramic samples were determined using the resonant frequency in the flexural mode of vibration (GrindoSonic Mk5, Leuven, Belgium). Four-point bending tests (30 mm outer span and 8 mm inner span) were performed using an Instron 1121 UTS instrument (Instron, Danvers, MA) on at least 15 specimens for each sample type with dimensions of $3.8 \times 2.7 \times 46$ mm^3. In order to remove surface flaws, all samples were carefully polished to a 6 μm finish before testing, using abrasive papers and diamond paste. The edges of the bars were beveled using fine abrasive papers and diamond paste. The cross-head speed was 1 mm/min until fracture. The HGC samples were tested by positioning the SB layer on the compression (upper) side and the GC layer on the tensile (lower) side.

The morphological features of sintered samples were characterized by scanning electron microscopy (SEM-ESEM Quanta 200, FEI Company, Eindhoven, The Netherlands). The crystalline phase assemblage was investigated on powdered samples by X-ray diffraction (Bruker D8 Advance, Karlsruhe, Germany), employing Cu Kα radiation (0.15418 nm), with data being collected in the range $2\theta = 10$–60° (in 0.05° steps, with a 3 s counting time). The identification was performed by means of a semi-automatic software package (Match!, Crystal Impact GbR, Bonn, Germany), supported by data from the PDF-2 database (International Centre for Diffraction Data (ICDD), Newtown Square, PA, USA).

3. Results and discussion

3.1. Sintering of waste-derived glass and direct sintering of waste

Fig. 1 illustrates the three different routes that were followed for the production of the various samples using the selected raw materials:

1 The first route (at the top in Fig. 1) involves the formation of a dense glass-ceramic body by sinter-crystallization of a waste-derived glass frit.
2 The second route (at the bottom in Fig. 1) involves the direct firing of a uniaxially pressed green body.
3 The third route (in the middle in Fig. 1) corresponds to the production of a hybrid glass-ceramic glazed body based on the previous two mixtures. In this case, the glass-ceramic frit is employed as a glaze to provide a dense, esthetically pleasing layer on the surface of the brown-colored fired body derived from natural and waste raw materials.

As can be seen from Fig. 2, the composition of the SB/GC1 samples is similar to that of slag sitals, although the content of iron oxide was greatly decreased in our samples in order to

achieve a light coloration. The content of TiO_2,which can act as a nucleating agent, was also very low, to help prevent crystallization of the GC1 glass by surface nucleation.

The DTA results for the GC1 frit (Fig. 3) clearly show that the glass is prone to surface crystallization; in fact, the main crystallization exothermic peak is more intense for fine powders (< 90 μm) than for coarser ones and, more importantly, the crystallization temperature is also lower for fine powders, at about 1000 °C. A secondary crystallization peak, located at about 900 °C for coarse powders, is also shifted to a lower temperature with decreasing particle size.

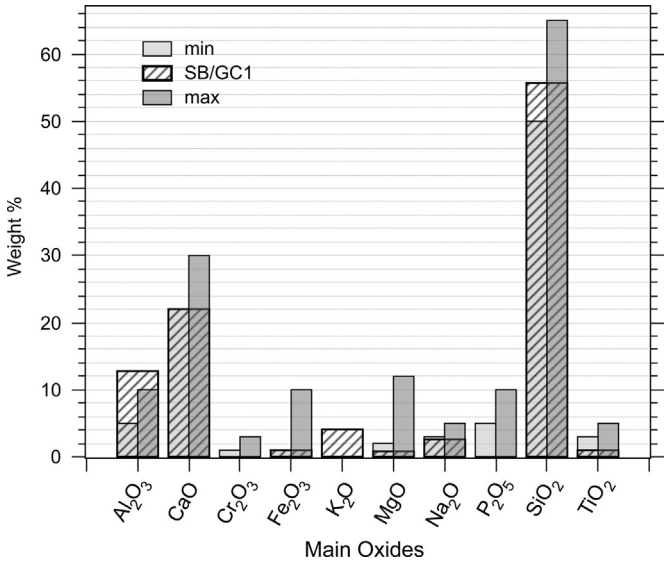

Fig. 2. Comparison between the composition of the investigated glass-ceramics and that of slag sitals ("min" and "max" indicate the minimum and maximum percentages of each component in the latter).

Both the DTA and dilatometric plots establish the glass transition temperature T_G at about 700 °C. The remarkable difference between T_G and the crystallization temperature T_{cryst} (about 300 °C) is useful for promoting the sinter-crystallization process; i.e., it is possible to successfully increase the density of glass powders via viscous flow while simultaneously achieving a high degree of crystallization. The sintering temperature was selected as 1000 °C to enhance crystallization, and this temperature was also high enough to achieve a remarkable increase in density, since viscous flow starts above T_G; more precisely, viscous flow sintering is generally optimized at 50–100 °C above the dilatometric softening point T_D [9]. In our case, the dilatometric softening occurs at about 770 °C (see Fig. 3), which is far below 1000 °C.

The effectiveness of surface nucleation for the GC1 frit is confirmed by Fig. 4(a) and (b), which shows that crystals had already started to develop after 30 min at 900 °C, and that after 1 h at 1000 °C, little evidence remained of the presence of a residual amorphous phase. The first crystalline phase to appear, after the heat treatment at 900 °C, was wollastonite (β-$CaSiO_3$, PDF#27-0088), followed by crystallization of Ca–Na feldspar (sodium-exchanged anorthite or bytownite, $Ca_{0.85}Na_{0.14}Al_{1.83}Si_{2.16}O_8$, PDF#76-0832). Both phases are typical of waste-derived CAS glass-ceramics [3]. In fact, anorthite crystallization was optimized at 950 °C, considering that its peak is clearly visible at a holding time of 30 min (Fig. 4(a)); for a longer treatment (60 min), crystallization of this phase is evident even at 900 °C and 1000 °C (Fig. 4(b)).

As can be seen from the data in Table 2, the differences in density and shrinkage were not particularly significant in samples sintered under different conditions; however, we observed that the samples that were heat-treated at 1000 °C displayed an excessive viscous flow, resulting in loss of the original disk shape. Taking into account the optimized crystallization of both wollastonite and anorthite and the retention of

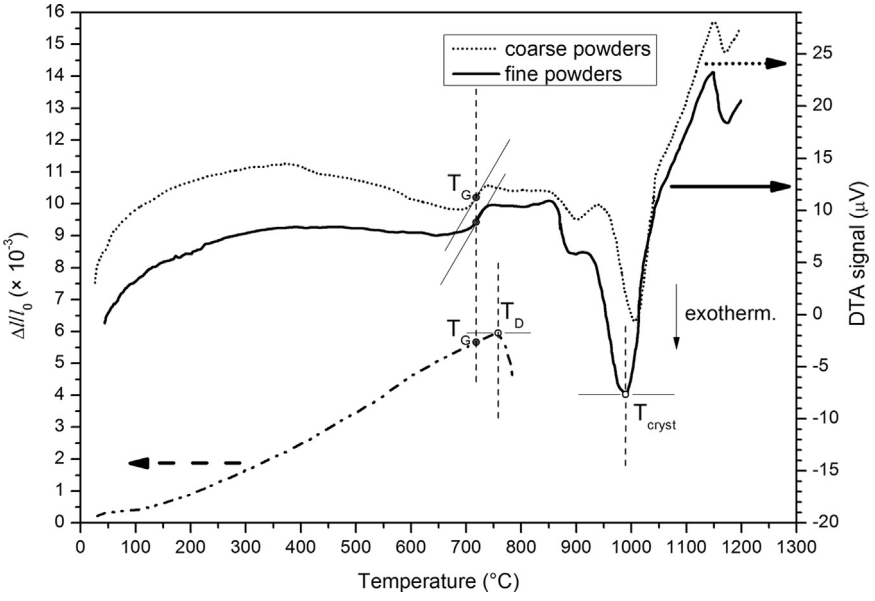

Fig. 3. DTA and dilatometric plots for GC1 glass.

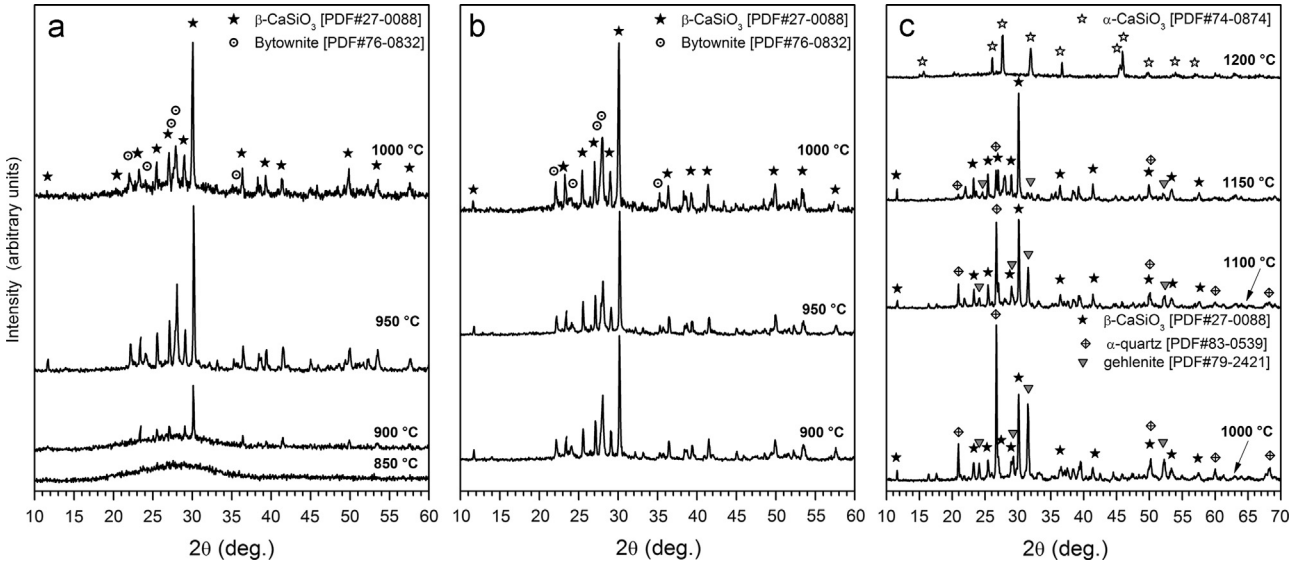

Fig. 4. X-ray diffraction patterns of GC1 and SB samples: (a) GC1 glass, sintered for 30 min; (b) GC1 glass, sintered for 60 min; (c) SB, sintered for 60 min (heating rate 10 °C/min).

Table 2
Summary of density and shrinkage values for sintered samples.

Type	Sintering temperature (°C)	Time (min)	Density (g/cm^3)	Linear shrinkage (%)
GC1	900	30	2.63 ± 0.02	21
	900	60	2.65 ± 0.02	21
	950	30	2.65 ± 0.02	25
	950	60	2.66 ± 0.02	25
	1000	30	2.65 ± 0.02	ND[a]
	1000	60	2.68 ± 0.02	ND[a]
SB	1000	60	1.94 ± 0.04	2
	1100	60	1.87 ± 0.04	3
	1150	60	1.83 ± 0.04	7
	1200	60	ND[b]	ND[b]

ND, not determined.
[a]Samples had no regular shape.
[b]Samples were glued to the refractory substrate after firing.

shape, we chose 950 °C as the reference temperature for the manufacture of larger samples. Rectangular tile samples (with a cross-section of 50 × 34 mm^2) of composition GC1 were successfully fired at 950 °C for 30 min, at a heating rate of 40 °C/min, which are conditions suitable for an industrial process.

The mechanical properties of bars cut from the tiles, as reported in Table 3, were of the same order of magnitude as those of similar CAS waste-glass sinter-crystallized glass-ceramics, and compare favorably with those typical of traditional ceramics [10]. In particular, the high strength is associated with negligible water absorption, in turn due to the smooth and dense surface, as shown by Fig. 5(a). It should be noted that a similar smooth and impermeable surface can be obtained in conventional stoneware ceramics only if they are sintered at temperatures much higher than 1000 °C [11]. The dense surface is accompanied by a homogeneous distribution of microcrystals, as shown by Fig. 5(b) (from a polished sample). The total residual porosity, evaluated by gas pycnometry, is of the order of 5 vol%.

The phase evolution of SB samples heat-treated at different temperatures is shown in Fig. 4(c). At 1000 °C, the sintering had just started to occur (see the limited shrinkage value in Table 2) and the interactions between the constituents are evident. Intense quartz peaks, associated with silica sand, were accompanied by peaks related to a calcium silicate, namely, wollastonite (β-$CaSiO_3$, PDF#27-0088), and a calcium aluminosilicate, namely, gehlenite ($Ca_2Al_2SiO_7$, PDF#79-2421), another phase that is typical of waste-derived CAS glass-ceramics [3].

The phase assemblage changed with increasing sintering temperature, owing to the progressive dissolution and reaction of some components. At 1150 °C, the peaks associated with gehlenite practically disappeared, whereas those related to quartz were much smaller. The liquid phase promoted by softening of the soda-lime glass gradually incorporated silica from sand, and partially crystallized into a species with a relatively high silica content (while alumina-rich gehlenite disappeared, wollastonite exhibited a slight increase in intensity of the associated peaks). The significant softening of the glass component was also testified to by an enhanced shrinkage (7%). For the highest sintering temperature, 1200 °C, the only phase detected was pseudo-wollastonite (α-$CaSiO_3$, PDF#74-0874, a high-temperature variant of calcium silicate), indicating that the other crystalline species dissolved in the residual glass. In other words, the assemblage of crystal phases resembled that of conventional glass-ceramics (only "newly formed" crystals), even without preliminary vitrification of the components. However, the loss of shape at this temperature (in fact, the sample remained glued to the refractory substrate upon firing) forced us to consider treatments below 1200 °C.

Table 3
Physical and mechanical properties of the different samples.

Sample	Sintering temperature (°C)	Residual porosity (%)	Water absorption (%)	Density ρ (g/cm^3)	Elastic modulus E (GPa)	Four-point bending strength σ (MPa)	$\sigma^{1/2}/\rho$ (MPa$^{0.5}$ cm^3/g)
GC1	950	5	<0.5	2.63 ± 0.02	84 ± 6	73 ± 8	3.3
Sintered body	1150	35	8	1.83 ± 0.04	41 ± 3	27 ± 5	2.8
HGC1[a]	1150	32	<0.5, on glazed surface[b]	1.89 ± 0.05	48 ± 5	26 ± 4	2.7
HGC2	1150	30	<5, on glazed surface[b]	1.99 ± 0.03	60 ± 4	40 ± 10	3.2

[a]20% zircon, 40% panel glass, and 40% GC1.
[b]Data inferred from measurements on frits sintered under the same conditions.

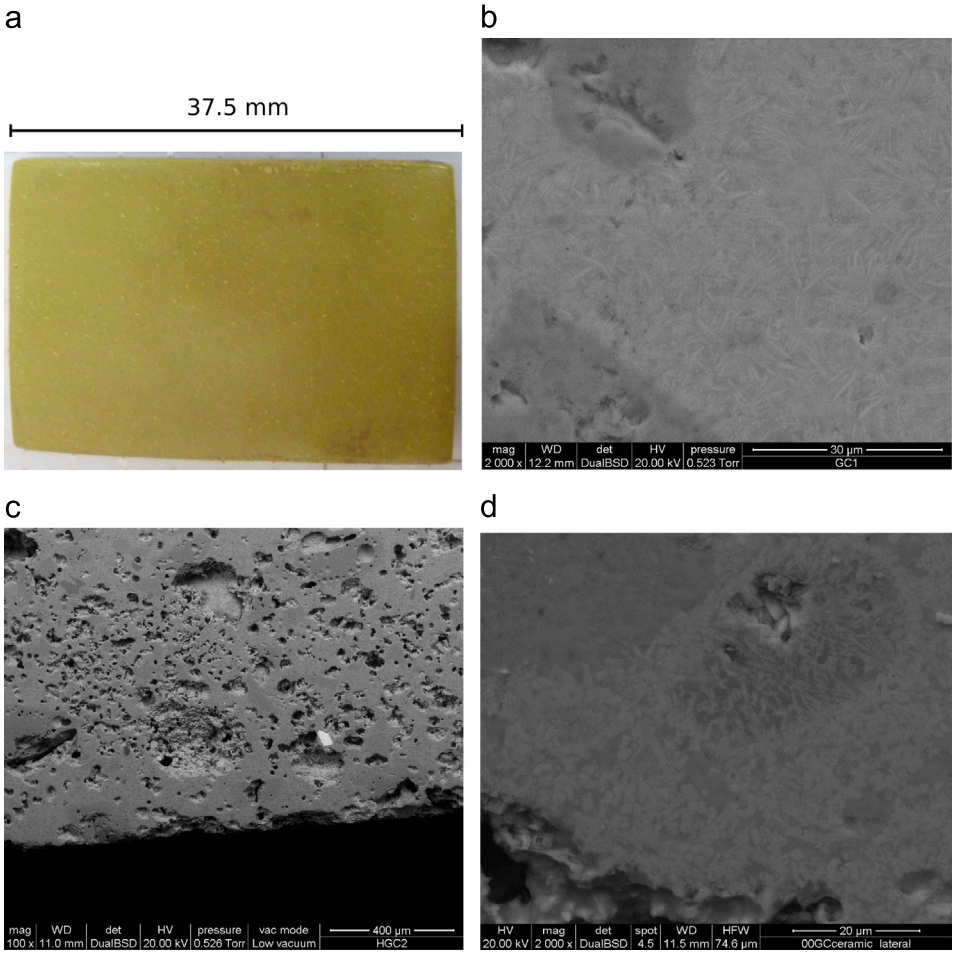

Fig. 5. Microstructural details of sintered materials: (a) and (b) sintered glass-ceramic (GC1); (c) and (d) ceramic from direct sintering (SB).

The mechanical properties of the SB sample sintered at 1150 °C were considerably poorer than those of the GC1 glass-ceramic, which can be explained primarily by the presence of a significant amount of residual porosity, of the order of 35 vol %. Part of the porosity is open, as testified to by the remarkable water absorption of 8% and illustrated by Fig. 5(c). Fig. 5(d) confirms the substantial microstructural similarity between directly sintered bodies and conventional glass-ceramics.

3.2. Layered hybrid glass-ceramics

The main aim of the fabrication of hybrid glass-ceramics was the preparation of a low-cost waste-derived tile by direct sintering, with limited water absorption (at least at the surface exposed to the environment when the tile is mounted) provided by a glass-ceramic top layer, preferably white. Since glazes from the pure GC1 glass had an unattractive yellow tint,

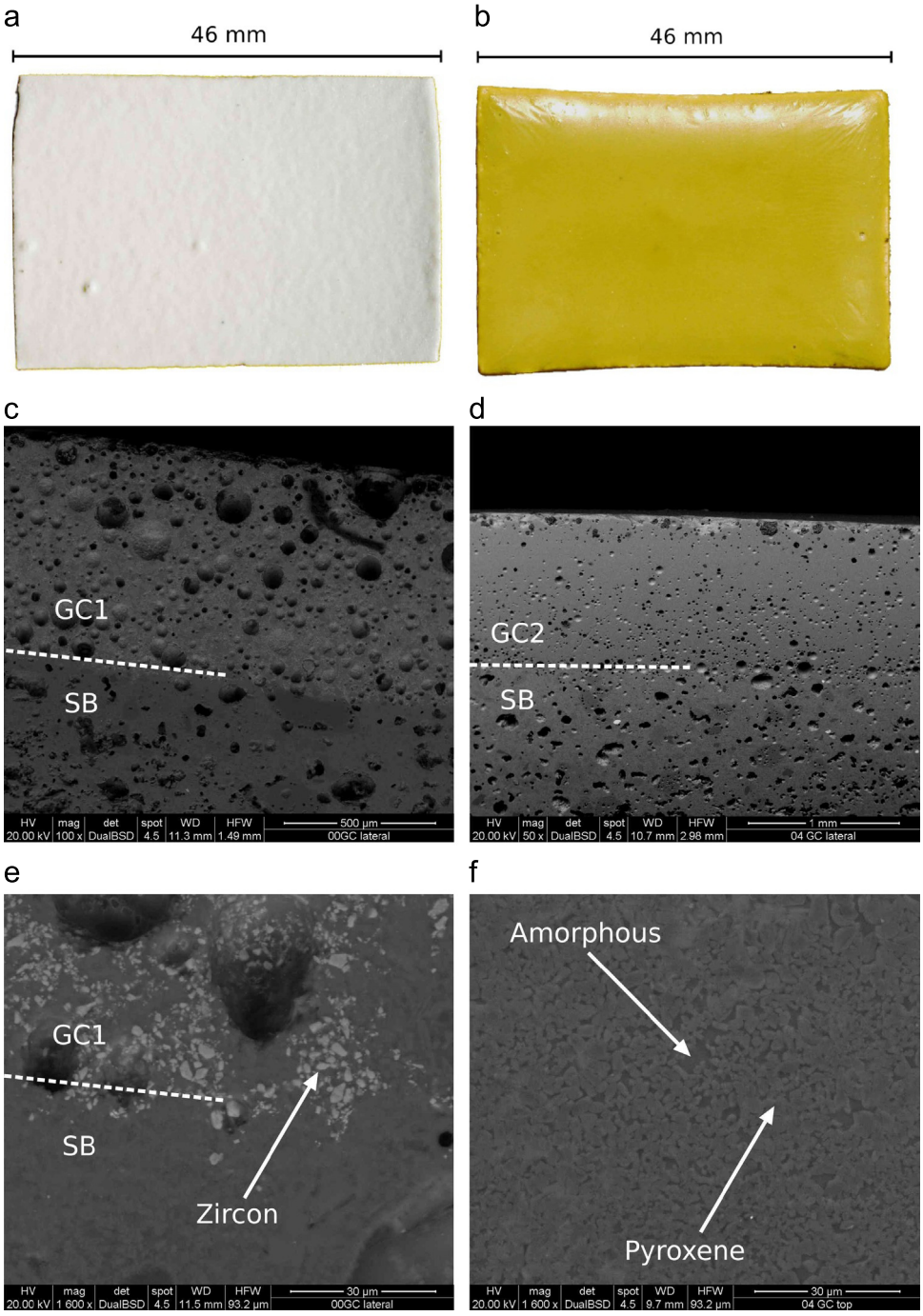

Fig. 6. Microstructural details of layered hybrid glass-ceramics: (a) sintered glass-ceramic HGC1; (b) sintered glass-ceramic HGC2; (c) interface between GC1 and SB layers; (d) interface between GC2 and SB layers; (e) magnification of the interface between GC1 and SB layers; (f) pyroxene crystals embedded in a glass matrix at the surface of the GC2 layer.

despite their low Fe_2O_3 content, zircon was added as a whitening agent, in the form of a micro-powder mixed with the glass frit, as is commonly done in commercial glass-ceramic glazes [12]. While GC1 glass, after firing at the same temperature (1150 °C) used for the substrate, had a very similar coefficient of thermal expansion ($7.6 \times 10^{-6}\,°C^{-1}$ for the coating, compared with $7.7 \times 10^{-6}\,°C^{-1}$ for the substrate), there was a significant mismatch with the zircon ($4.1 \times 10^{-6}\,°C^{-1}$ [13]); in order to compensate for this, CRT panel glass ($9.9 \times 10^{-6}\,°C^{-1}$ [14]) was also added. Moreover, since the panel glass is itself a waste product, its use was consistent with the overall approach of obtaining low-cost waste-derived materials.

Predicting the coefficient of thermal expansion of a composite material is not straightforward [15], but a GC1 frit/panel glass/zircon weight ratio of 40/40/20 was found to be the most appropriate choice (with an expected coefficient of thermal expansion close to $8 \times 10^{-6}\,°C^{-1}$). Samples fired at 1150 °C,

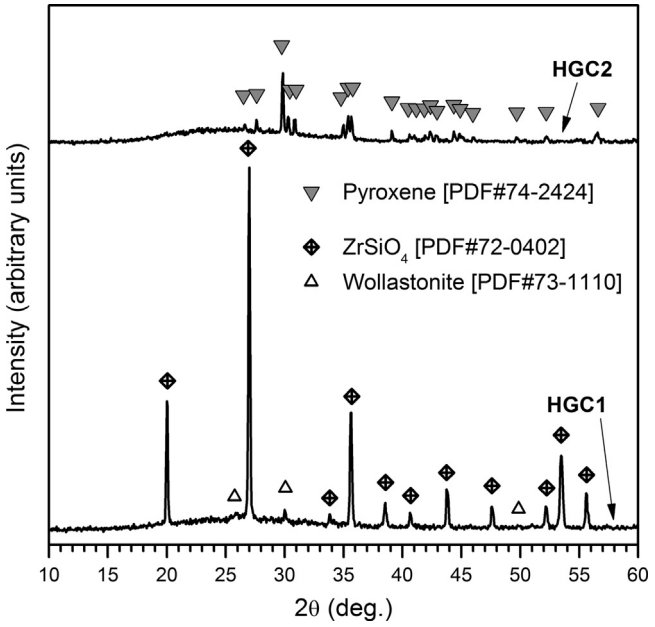

Fig. 7. X-ray diffraction patterns of glaze materials.

with a substrate coated with a GC1/panel glass/zircon (40/40/20 by weight) composite frit were effectively crack-free and possessed a continuous, dense, and brilliant white layer on the top surface, as can be seen in Fig. 6(a). While the water absorption became negligible, the mechanical properties remained virtually unchanged compared with the uncoated SB (see Table 3). This can be explained by the fact that the glaze, although perfectly interpenetrated with the substrate, contained some bubbles, as shown in Fig. 6(c). The development of the pores could be favored by the presence of the panel glass, a highly oxidized glass known to release dissolved oxygen upon sintering [14]. The limited mechanical properties could be explained by the almost complete absence of crystalline phases, as illustrated by Figs. 6(e) and 7; besides un-dissolved zircon, only weak traces of wollastonite are visible.

GC2 glass was prepared from a selected mixture of wastes, in the compositional range of slag sitals (as for GC1), but without the constraint of having a low amount of Fe_2O_3. This choice led to a brilliant olive green color when applied as a glaze without additives (Fig. 6(b)). As demonstrated by Figs. 6(f) and 7, the use of GC2 led to a crystallized layer, with a pyroxene (ferrous diopside, $(Ca_{0.92}Fe_{0.08})(Al_{0.14}Fe_{0.33}Mg_{0.53})Si_2O_6$, PDF#74-2424) as the main crystal phase. This specific phase is obviously consistent with the increased iron content in the glass-ceramic formulation. The partial crystallization, the absence of bubbles, and the close matching of the coefficient of thermal expansion ($7.2 \times 10^{-6}\ ^{\circ}C^{-1}$) had a positive effect on the mechanical properties of the component.

Because of their improved mechanical properties (see Table 3), the second series of layered glass-ceramics could find interesting applications as structural materials. In fact, although lighter than HGC1, HGC2 possessed an almost identical specific strength (which is the property determining the mechanical efficiency of panels, as proposed by Ashby [15]). In addition, this value was quite close to that of lightweight stoneware tiles ($2.9\ MPa^{0.5}\ cm^3/g$), previously developed for high-value applications such as so-called ventilated facades [16].

The first series of layered glass-ceramics (incorporating zircon and panel glass in the glaze) could be used for tiles subjected to moderate loads, although future investigations may lead to improvements in mechanical properties by altering the formulation (the type and content of additives).

4. Conclusions

From the results of the experiments described here, the following conclusions can be drawn:

- A selected mixture of waste and low-cost mineral starting materials can be converted into glass-ceramic components either by vitrification and subsequent crystallization or by direct sintering.
- The sinter-crystallization approach leads to strong glass-ceramics, under particularly simple conditions (pressing of fine powders and sintering at 950 °C for only 30 min).
- The surface porosity of a glass-ceramic body from direct sintering can be sealed by a glass-ceramic glaze, produced from the same starting mixture of waste and minerals.
- The characteristics of waste-derived glass-ceramic glazes can be tailored by the addition of secondary components (panel glass from dismantled CRTs and zircon).
- Owing to their characteristic specific strength, layered glass-ceramics could find applications in the building industry as lightweight tiles.

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