Bilayered ceramic dental composites with adhesive or reactive bonded interfaces

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By close inspection of the well polished cross-sections, two categories of interfaces were classified, namely, adhesive bonding between veneering porcelains and zirconia or alumina cores versus reactive bonding between veneering porcelains and cores of glass infiltrated alumina or lithium disilicate based glass ceramics. Argon ion beam cross-section polishing technique was applied to achieve gentle and fine polishing required for high resolution interfacial characterisation by scanning electron microscopy. The observations suggest that it is desirable to enhance the interfacial reactive bonding in order to avoid delamination in alumina and zirconia based composites.

Keywords: Bioceramics, Fracture, Interface, Microscopy

Introduction

Owing to its excellent aesthetics and biocompatibility, ceramic dental restoration concept is widely spreading in dentistry. So far, the materials solution established for implementing this concept is to manually construct a bilayered composite of dome-like structure. An aesthetic 'porcelain' layer, known as veneering, is covering an individually manufactured inner load bearing ceramic core of high strength. The term 'porcelain' commonly quoted in dentistry bears a different meaning than that established in the ceramic science community. In dentistry, it refers to a pure solid silicate glass instead of a polycrystalline body containing additional phases of pores and glasses. The veneering technique is also essentially different from the common glazing process as the porcelain layer is usually thicker than the ceramic core, which results in much more complex stress distribution in the prepared dental bilayered ceramic composites. Four kinds of core ceramics are in wide use in clinical practice nowadays, namely, zirconia, alumina, glass infiltrated alumina and glass ceramics. It has been observed that entire crown cracking more frequently occurs when glass–ceramic, glass infiltrated alumina or alumina is used as core ceramic, whereas veneer flaking occurs more often when zirconia is used as a core ceramic.[1–4] The interfacial characteristic of the core ceramic and veneering porcelain is one of the critical parameters that determine the fracture behaviours of the bilayered composites.[1–3]

In practice, the veneer porcelain layer is built up by a multistep firing procedure, which may yield thermal degradation of the involved materials. All bilayered composites investigated in this study were, therefore, prepared by strictly following the established dental laboratory multistep firing procedures to reflect such possible influence. After mechanical bending tests, the broken specimens were carefully grinded and polished to reveal the interfaces between hard ceramic cores and the relatively softer glasses by a newly developed argon ion beam cross-section polishing technique. This method is unique as it allows achieving a very fine and gentle surface polishing without introducing secondary mechanical damaging. It allowed a close inspection of the interfacial microstructures of bilayered dental ceramic composites of four widely used types. These studies enabled to define and judge the interfacial characteristic and its influence on interfacial delamination, to achieve a better understanding of the fracture mechanism(s) and to guide further development of materials and clinical restoration criteria.

Experimental

Materials preparation

The compositions and manufacturers of the core ceramics and veneer porcelain materials used in the present study are listed in Table 1. Dense ceramic blocks made of high purity alumina and 3 mol.-% yttria stabilised tetragonal zirconia respectively are referred simply as alumina and zirconia in the text hereafter to follow the empirical definition in dentistry. The ceramics aimed for the cores were prepared by cold isostatic pressing followed with pressure less sintering in air. Rectangular testing bars were then prepared by cutting into a proper size that will give the final dimensions after

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Table 1 Core and veneer ceramics used for preparing bilayered composites in this study

Bilayered specimen	Core		Porcelain			
	Trading mark	Main composition*	Trading mark	Main composition*	Bending strength*/MPa	Manufacturer
Zirconia/porcelain		ZrO_2, Y_2O_3	Rondo Zirconia	SiO_2, Al_2O_3, CaO, K_2O, B_2O_3, Na_2O, Y_2O_3, ZrO_2	120	Nobel Biocare
Alumina/porcelain		Al_2O_3	Rondo Alumina	SiO_2, Al_2O_3, CaO, K_2O, B_2O_3, Na_2O, Y_2O_3, ZrO_2	120	Nobel Biocare
Glass infiltrated alumina/porcelain	In-Ceram Alumina	Al_2O_3, SiO_2, B_2O_3, TiO_2, La_2O_3, CeO_2, CaO	Vitadur Alpha	Al_2O_3, SiO_2, Na_2O, K_2O, CaO, B_2O_3	84	Vita Zahnfabrik,
Glass ceramic/porcelain	Empress 2	SiO_2, Al_2O_3, La_2O_3, MgO, ZnO, K_2O, Li_2O, P_2O_5	IPS Empress 2 Veneer	SiO_2, Al_2O_3, P_2O_5, K_2O, Na_2O, CaO	90	Ivoclar Vivadent

*According to the manufacturer's specification.

gentle double side fine grinding. Test bars of glass infiltrated alumina were prepared by cutting porous aluminium blanks to right dimensions followed by infiltrating with a lanthanum aluminosilicate glass at 1140°C for 40 min. Finally, the test bars of a glass ceramic with lithium disilicate as major crystalline phase (referred simply as glass ceramic) were hot pressed at 920°C under 5 bar compressive air pressure according to the manufacturer's recommendation. The final dimension of the prepared core specimens is 20 mm in length, 4 mm in width and 1·2 mm in height.

Rectangular bilayered specimens with the four kinds of core materials were fabricated by veneering the recommended porcelains for each material onto the prepared core testing bars to achieve a core to veneer thickness ratio of 1 ($R=1$) (Table 1). The final dimension of bilayer test bars is therefore 20 mm in length, 4 mm in width and 2·4 mm in height. Before veneering, the surfaces of the core ceramic bars were finished by sandblasting with 25 μm aluminium oxide abrasive at 0·4 MPa of pressure at a distance of ~10 cm. The samples were faced to the sand beam with an angle of 45° with sandblasting time of ~1 min. It is worthwhile to mention that sandblasting may influence the mechanical performances of core ceramics. Based on previous studies, sandblasting reduced the strengths of sandblasted specimens in general due to the formation of surface sandblast flaws.[5] In the case of zirconia, sandblasting induced tetragonal to monoclinic phase transformation and a compressive stress on the surface due to the volume expansion associated with such a phase transformation.[6] In this study, sandblasting was adopted to reflect the possible influence on mechanical properties of each processing step involved in the clinical routine veneering process. We prepared all individual bilayered specimens strictly according to the dental laboratory procedures currently used worldwide, which were specially developed for making all ceramic crowns with multistep firing, interposed by rapid heating and cooling. During the first firing cycle, a veneer layer with approximate target thickness was built up and fired. In order to achieve the right dimensions, at least one more correction firing cycle was then conducted to compensate for the sintering shrinkage of veneer layer.

Three-point bending test

The specimens for bending test were fine ground on both surfaces with a two-step procedure using diamond discs composed of 30–40 and 15–20 μm diamond particles respectively. The cross-sectional dimension of each beam was carefully measured by a micrometer. Ten specimens of each group of composites were tested according ISO 6872:1995(E) for dental ceramics by three-point bending test using universal test machine (CMT5105, Sansi. Co., China) with a testing bar span of 15 mm. The load was applied at the centre of the veneer surface with a crosshead speed of 0·5 mm min^{-1}.

Microstructure characterisation and energy dispersive X-ray spectroscopy analyses

After the bending test, broken pieces of bilayer specimens were collected and studied with an optical microscope (Motic k400; Preiser Scientific, Louisville, KY, USA) to examine the features of the fracture surface. The high resolution fractography investigations were carried out using a field emission scanning electron

microscope (SEM; JEOL JSM-7000F; Jeol Ltd, Tokyo, Japan). For further investigation of the interfacial characteristic and the compositional variations across the interface between core and veneer layer by SEM/ energy dispersive X-ray spectroscopy (EDS), one test bar of each of group of bilayer specimens was selected for cross-section polishing using a newly developed argon ion beam polishing apparatus (SM-09010 cross-section polisher; Jeol Ltd, Tokyo, Japan).[7,8] A precut sample imbedded in resin with mechanically polished surfaces is covered with a shield plate, which stops half of the Ar ion beam. Only an ~75 μm wide part of the sample protrudes from the cover. This part is slowly milled by the ion beam, leaving behind a well polished surface at the position of the edge of the shielding plate. At this position, the incidence of Ar ions is parallel to the surface. The ion milling operated at 5 kV/90 μA, and the polishing time was set to 15 h. By this way of polishing, the possible surface damages often introduced by conventional mechanical polishing are minimised. It also enables to achieve a finely polished flat section of several tens of thousands square micrometres even when layered composites with two not well coherent layers having very different properties like toughness, hardness and stiffness are polished. For investigating the possible phase change occurring during the veneering process, one core and bilayer beam each in the groups Empress 2 and In-Ceram alumina were analysed after veneering by X-ray diffraction (XRD) phase analysis (X'Pert PRO; PANalytical B.V., The Netherlands). Specimen surfaces were scanned with Cu $K_{\alpha 1}$ X-ray from 20–80° 2θ degrees with a step size of 0·026°.

Results

Flexure strength and fractography

The three-point bending test results for each group of bilayered composite specimens are summarised in Table 2. The mean flexural strength of the four core materials are very different, with the bending strength reducing in the order of zirconia>>alumina>glass infiltrated alumina>>lass ceramics. Thus, zirconia demonstrates a much higher mean flexural strength than the other three core materials. The mean strength of all composites is significantly lower than the pure core ceramics, but the same tendency for the strength order remains for the bilayered composites. It may be noted that the zirconia/porcelain composite is still even stronger than the other three pure core materials. However, no significant strength differences were found among the other three weaker composites, i.e. bilayered composites

with glass ceramic, glass infiltrated alumina or alumina as core material.

The specimens gave two major fracture pieces after the three-point bending test and with the fracture origin always located at the surface with the most tensile stress concentration. This point is located at the test bar lower surface opposite to the load piston and in the middle between the two support cylinders. Two types of fracture features with distinctly different crack propagation trajectories were observed. Namely, either an initial interfacial delamination between the core and veneering ceramic or a direct bar fracture accompanied without or only with very limited interfacial delamination. It is notable that interfacial delamination occurred in all zirconia composite specimens, whereas it never appears in the glass ceramic or the glass infiltrated alumina based composites. Finally, a very limited interfacial delamination occurs in alumina composites, and the major fracture mechanism is still a crack path straight through the test bar.

The fractography analyses were carried out on fracture surfaces of the four core ceramics by SEM observations of fractured bars. The following general features were revealed, as also illustrated in Fig. 1:

(i) zirconia core ceramic consists of densely sintered fine grains of 0·2–0·5 μm size with some occasionally occurring very small voids in nanometre scale. Both intergranular and transgranular fractures take place during crack propagation. This is seen on the fracture surface where parts of the grains reveal straight edges and shiny surfaces, whereas uneven grain surfaces appear on the other part of grains (Fig. 1a)

(ii) alumina core ceramic consists of larger grains, but still of size 2–10 μm, which are densely sintered. Submicrometre sized voids are observed by SEM to be entrapped inside the grains. Intergranular fracture is clearly dominating in this case, as seen in Fig. 1b

(iii) the glass infiltrated alumina core demonstrated an apparent composite microstructure. Image (SEM) taken by backscattered electron mode illustrates a microstructure comprising darker alumina grains with bimodal size distribution (2–10 μm) and an intergranular lanthanum enriched phase with a lighter grey contrast (*see* Fig. 1c). Well faceted alumina grains are clearly distinguishable, indicating that intergranular fracture is dominating

(iv) glass ceramic core based on lithium disilicate reveals a fracture surface dominating by transgranular cracks propagating through elongated

Table 2 Summary of three-point bending test results of core ceramics and bilayered composites

| Group | Mean*/MPa | SD | SE | 95% confidence interval | | Min. | Max. |
				Lower	Upper		
Zirconia	1102	45·7	18·6	1053·9	1149·8	1056	1171
Alumina	514	20·4	7·7	494·7	532·5	490	543
Glass infiltrated alumina	444	17·8	8·0	422·1	466·4	413	458
Glass ceramic	191[a]	19·5	8·0	170·1	211·0	168	210
Zirconia/porcelain	605	31·0	12·6	572·5	637·5	570	645
Alumina/porcelain	171[a,b]	14·2	5·3	158·1	184·3	149	182
Glass infiltrated alumina/porcelain	155[b]	8·1	3·3	146·7	163·7	145	164
Glass ceramic/porcelain	148[b]	21·0	8·6	125·6	169·8	122	173

*The values indicated with same letter (a or b) have no statistical difference ($p > 0·05$).
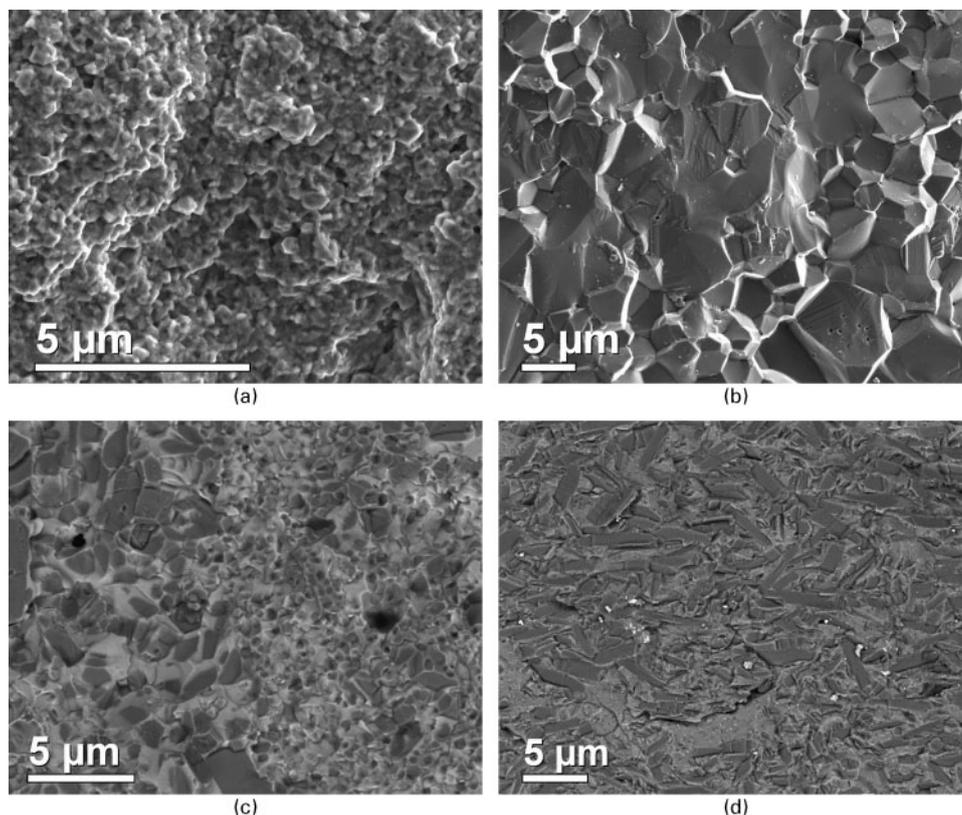
1 Scanning electron microscopy fractography of pure core ceramics of *a* zirconia, *b* alumina, *c* glass infiltrated alumina composite (In-Ceram Al) and *d* glass ceramics (Empress 2)

lithium disilicate grains of 5–6 μm, as seen in Fig. 1*d*.

Interfacial characteristics

Owing to severe interfacial delamination of the zirconia based composites during the bending test, it is difficult to observe the true interface microstructure from fragments of the flaking and fractured test bars. Therefore, the technique using the gentle argon ion beam section polishing was crucial for this purpose. On such well polished cross-sections, the presence of large voids and other types of micrometre sized microstructural defects on the porcelain side of the interface of zirconia composites were disclosed by the SEM investigation (*see* Fig. 2*a*). Similar defects were not found on the interface of alumina based composites where alumina is coherently bonded to the porcelain except for the observation of small microcracks in the alumina ceramic. The microcrack zone has a thickness of about one grain immediately below the interface, and the microcracks are mainly oriented parallel to the interface direction (*see* Fig. 2*b*). When the glass infiltrated alumina is used as core material, backscattered electron SEM images of well polished cross-sections revealed the segregation of a lanthanum enriched crystalline phase into the core material in a thin zone of about 1–2 μm thickness at the interface (*see* Fig. 2*c*). In addition, stress fringes are observed inside the veneer porcelain perpendicular and close to the interface on a fractured surface, indicating a high level of thermal mismatch between core ceramic and veneer porcelain. Finally, in case of bilayered composites with glass ceramic as the core material, no obvious stress fringes are noticed inside the veneer porcelain on the fracture surfaces, revealing the absence or only a very small thermal mismatch between the core material and veneer porcelain. As shown in Fig. 2*d*, a diffusion zone ~5 μm thick into the porcelain at the interface is observed on well polished cross-sections, in which tiny crystals formed during the veneering process.

Chemical composition changes across interfaces

The EDS analyses results revealing the chemical composition changes across the interfaces are summarised in Fig. 3. The following general tendencies are observed:

(i) a distinct biphasic interface was found in bilayered composites when zirconia and alumina are used as the core ceramics; neither any transdiffusion of the constitutional elements across the interfaces nor any additional compound formed at the interface zone was observed

(ii) for the bilayered composites based on glass infiltrated alumina, a diffusion of lanthanum is detected in a thin zone, about 1–2 μm thick, into the core material at the interface. This is also confirmed by the segregated lanthanum enriched crystals observed by backscattered electron SEM at the interface as reported above. A gradual decrease in silicon content or a gradual increase in the aluminium content is detected in a ~5 μm thick zone in the porcelain side of the interface towards the core

(iii) when glass ceramic is used as the core material, a gradual decrease in K, Na and Al was detected in a zone of ~5 μm thick in the porcelain close to the interface towards the core in which the enrichment of P and the deficiency of Si were noticed.
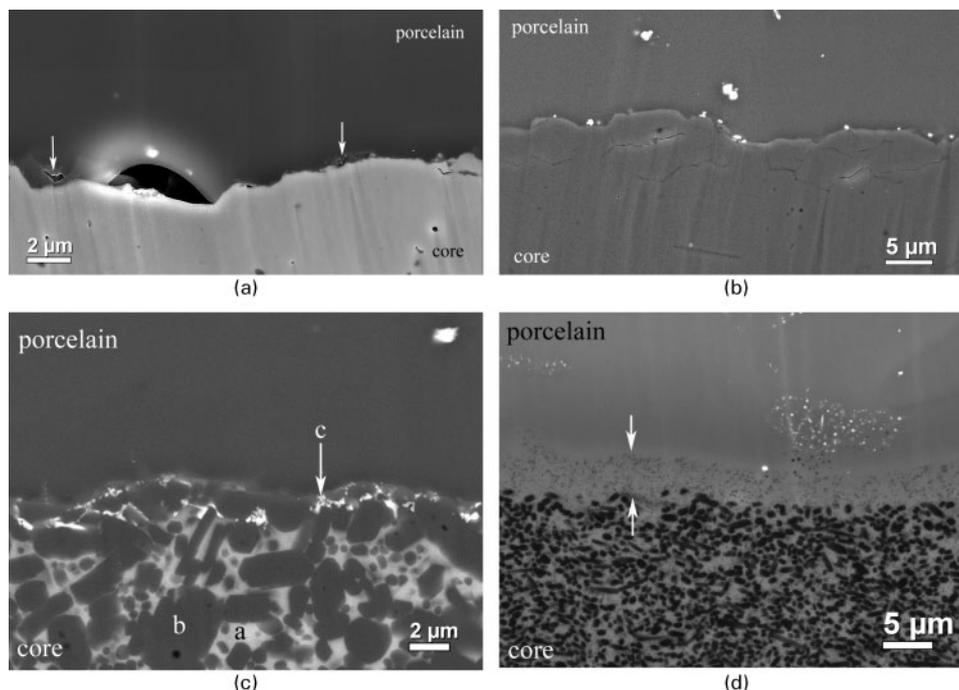
2 Images (SEM) of interface areas of polished cross-sections of bilayered composite samples with core made of *a* zirconia, *b* alumina, *c* glass infiltrated alumina composite and *d* glass ceramics

X-ray diffraction analyses confirm that the glass infiltrated alumina and glass ceramic cores consist of alumina and lithium disilicate as the major crystalline phase respectively. By closely comparing the XRD diagrams, in both cases, no phase change was observed to take place during the veneering process.

Discussion

Adhesive versus reactive bonding

It is the overall impression that the interfacial coherence between the core ceramics and veneer porcelain changes in the order of zirconia<alumina<glass infiltrated alumina<glass ceramics. Accordingly, the interfacial coherence can be classified into two categories, namely, adhesive interface and reactive interface. The fact that no transdiffusion of the constitutional elements occurs across the interfaces with bilayered composites having zirconia or alumina as the core material shows that in these two cases, the veneering porcelain is adhesively bonded to the core ceramics. On the other hand, when glass infiltrated alumina and glass ceramic are used as core materials, the observation of a compositional diffusion zone at the interface is a confident proof of the reactive bonding achieved through elemental diffusion and chemical reactions taking place during the veneering.

Both zirconia and alumina core ceramics are solid state sintered, implying that no liquid phase is involved during sintering; thus, no residual glassy phase exists in the grain boundaries of the dense ceramics. Therefore, at the applied veneering temperature, 910°C, both zirconia and alumina core materials are chemically inert to the used porcelain based on silicate glasses. The fully dense ceramics also prevent any penetration of veneer porcelain through open voids or grain boundaries.

The glass infiltrated alumina is a composite formed by infiltrating a lanthanum aluminosilicate melt into a porous alumina blank with pore content well above the percolation limit.[4,9] Based on the information supplied by the manufacturer, a bimodal grain distribution with a medium grain size of ~ 3 μm is formed at 1120°C. The lanthanum glass used possesses a low viscosity at this temperature, which favours completely filling of open pores, yielding a composite consisting of 75%Al_2O_3 and 25% glass. The infiltrated melt reacts with alumina during the infiltration process and may partially crystallise during cooling depending on the processing procedure. Guazzato *et al.* investigated the influence of heat treatment of In-Ceram alumina cores and observed some indication of crystallisation of the infiltrated glass during this heat treatment, but they were not able to determine the exact crystal phase formed.[9] The presence of a liquid/glass phase, even in small amounts, may otherwise act as pathways for enhancing atomic diffusion and chemical reactions. Therefore, during the veneering process up to a temperature of 960°C, the softened residual glass phase in the used core ceramic is able to react with the viscous veneer porcelain through enhanced transdiffusions over the interface.

The lithium disilicate glass ceramics, Empress 2, is also a composite with homogeneous distribution of crystalline grains of elongated lithium disilicate, $Li_2Si_2O_5$ ($\sim 70\%$), in a glass matrix, formed by controlled crystallisation of glass. The overall composition of this glass ceramic corresponds to stoichiometric $Li_2Si_2O_5$ with trace addition of Na, K, Al and P. The applied veneering temperature, 755°C, is very close to the reported crystallisation temperature of $Li_2O–Al_2O_3–SiO_2$ glass being 745°C.[10] Albakry *et al.* investigated the influence of heating process on the crystal phase formed in this glass ceramic by XRD analysis. They noticed that the XRD background intensity decrease and ascribed this observation to a reduced amount of glassy phase.[11] They also found that the size of the lithium disilicate crystals increased by prolonging the heating. This was considered to be a consequence of Ostwald ripening of the crystalline
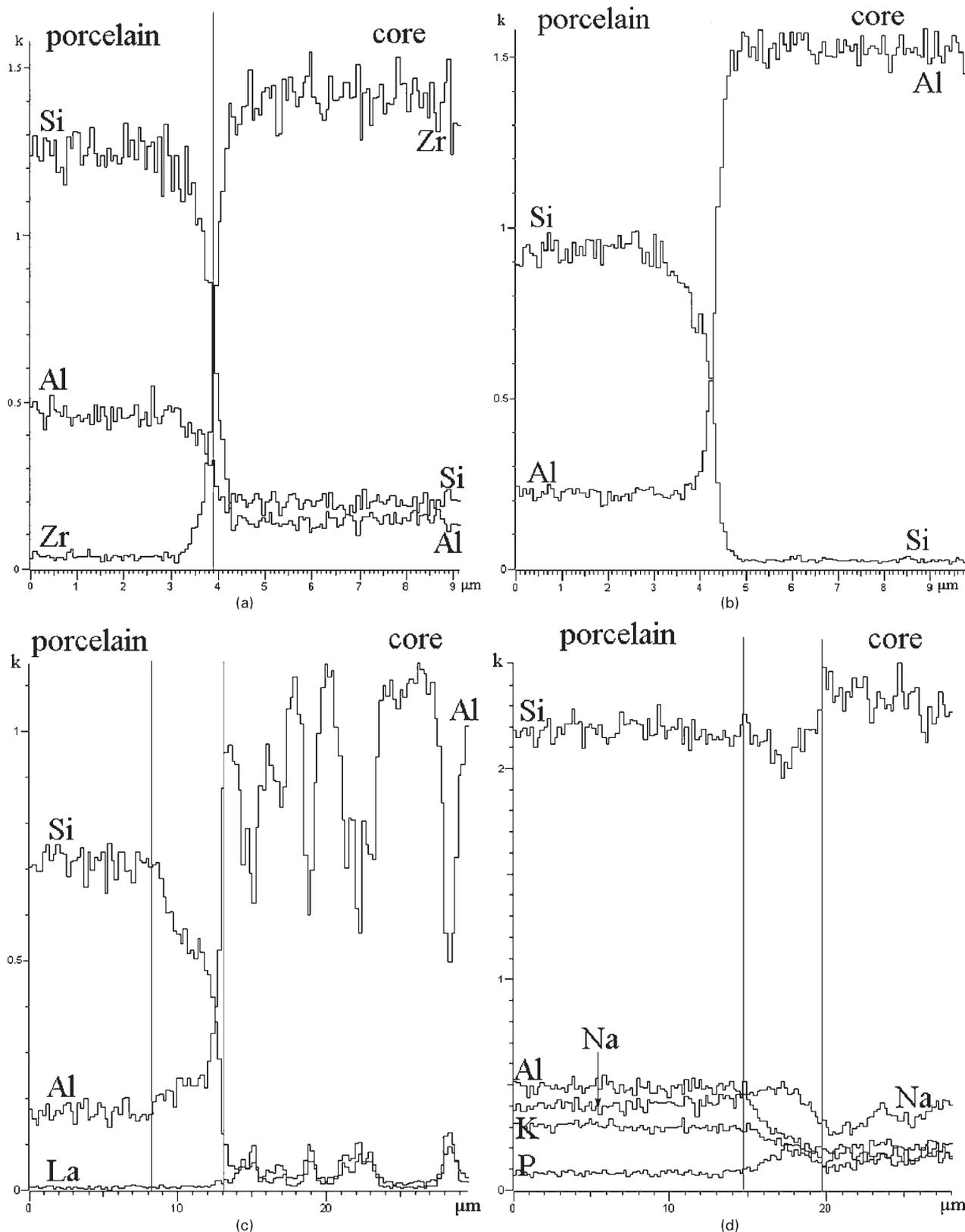
3 Energy dispersive X-ray spectra recorded during linear scanning across interface of bilayer composites with core made of *a* zirconia, *b* alumina, *c* glass infiltrated alumina composite and *d* glass ceramics

precipitates within the glass matrix.[12] Höland *et al.* analysed the microstructure of this glass ceramic after veneering and found that the lithium disilicate remained as the main phase of the glass ceramic, but a second crystalline phase in low concentration was determined as lithium orthophosphate, Li_3PO_4.[13] This crystal phase is approximately 0·1–0·3 μm in diameter, and it is located in the glassy matrix and on the surface of the lithium disilicate crystals. Other reports, like Cattell *et al.*, came to similar results by XRD that lithium disilicate is the major phase and lithium orthophosphate is found as a minor phase, and they found further that no significant

phase changes occurred following the recommended veneering heat cycles.[14,15] The veneering porcelain used for Empress 2 features also a composite microstructure consisting of glass and a small amount of very finely dispersed crystals precipitated in the matrix. The crystals were determined as fluoroapatite, $Ca_5(PO_4)_3F$, by XRD, which appear with needle-like shape and are $0.3–3$ μm in length.[13] The enrichment of P in the interfacial diffusion zone seems to be the consequence of the diffusion of P from the core towards the porcelain that probably yields the formation of tiny crystals of lithium orthophosphate, Li_3PO_4.

Relevance of weak reactive bonding to interface delamination

The stress leading to fracture failures of four types of bilayered composites is very different, being almost four times higher for zirconia composite in comparison with the rest. The factors leading to the different overall fracture resistances of the bilayered composites are a complex combination of several underlying parameters, and this will be analysed in more detail in our continuous studies. However, the often observed interfacial delamination in zirconia bilayered composite is a combination effect of a high failure stress resistance of the zirconia core and a low failure resistance of the veneer porcelain and, especially, a weak reactive bonding between the veneer porcelain and core ceramics. Zirconia does not react with silica below 1250°C even in nanopowder mixtures.[16] It is thus not surprising that zirconia lacks reactive bonding with silicate glasses at the veneering temperature, 910°C, applied in this study, which results in a week interface. The observed presence of large voids and microstructural defects in the porcelain at the interface further indicates that the wetting between zirconia and silicate glasses is poor. The concept of a zirconia/porcelain combination still has a great prospect because of its overall strength and the possibility to improve the interfacial bonding. Thus, defects like interfacial voids may be avoided by processing upgrading, e.g. using a pressure during the veneering process.[17] The wettability and reactive bonding behaviours of zirconia towards porcelain can be further improved through studying of the veneering process parameters and by adjustment of the porcelain materials.

In the alumina bilayered composite, no obvious pores and other microstructural defects were found at the interface, indicating a better wetting between alumina and silicate veneer porcelain. However, interfacial delamination was still observed in 30% of the test samples at a maximum critical stress of 171 MPa, clearly indicating that the reactive bonding at the interface is limited. In addition, this bending strength is much lower than what is expected, indicating that also a possible thermal damage of the core ceramic occurs during veneering process. Extensive microcracking was observed in the alumina microstructure close to the rapidly heated/cooled interface towards the veneering that supports this hypothesis. The possible formation of such extensive microcracks by sandblasting was excluded, as similar crack formation phenomena were observed in alumina bilayered specimens veneered with either Rondo or Vintage porcelains, regardless if the alumina bar surface is sandblasted or gently ground by sand paper before the veneering procedure.

An enhanced chemical reaction and/or transdiffusions of constitutional atoms at the interface seem to increase the bonding between the core ceramic and the veneer reducing veneer flaking. This is demonstrated in the two cases of bilayered composites when a glass infiltrated alumina and a lithium disilicate based glass ceramic are used as core ceramics.

The interfacial bonding can be classified into two major categories, namely, adhesive bonding and reactive bonding. Zirconia and alumina demonstrate adhesive bonding with silicate porcelains, whereas glass infiltrated alumina and lithium disilicate based glass ceramic have reactive bonding towards silicate porcelains. This study revealed the necessity of enhancing the reactive bonding between silicate porcelains and ceramic cores made of zirconia and alumina to reduce flaking tendency. The bonding is also one of several factors in the complex stress situation when a major fracture occurs, but better bonding will help to improve the composite's overall strength resistance. Finally, it is also of importance to use a careful thermal treatment during the veneering process when the core ceramic is a thermal shock sensitive material like pure alumina.

Implications

It appears to be an empirical preference in dental practice to apply sandblasting of dental cores as a common processing step. The bilayered all ceramic restoration structure is an updated version of the early prosthesis called porcelain fused on metal in which a porcelain layer is fused on an inner core made of metal alloys. The metal cores are usually produced by casting, which implies that sandblasting is a necessary process step for removing adhered investment material. For a ductile metal, sandblasting is also beneficial for increasing the adhesive bonding between the metal core and the veneering porcelain by increasing the surface roughness of the metal core, and it does not damage the material.[18] This is, however, not the case when brittle ceramics are used as dental cores, where cautions have to be taken to balance the positive enhancing of adhesive bonding and the negative surface damages both caused by sandblasting.[5] In the two cases of using glass infiltrated alumina or lithium disilicate based glass ceramics as dental cores, the formation of a strong reactive bonding between veneering porcelains and these cores makes it not necessary to increase adhesive bonding by sandblasting. However, a gentle sandblasting is still needed for removing adhered investment material as the cores are nowadays produced by hot pressing inside an investment material. Finally, in the cases of alumina and zirconia core ceramics, the bonding between veneering porcelains and cores is purely adhesive. For improving these adhesive bonding, sandblasting seems to become a preferable operation. To this end, it is worth recalling earlier findings that zirconia is exceptionally strengthened by sandblasting due to the compressive surface layer formed by sandblasting induced phase transformation of tetragonal to monoclinic zirconia.[6,19] However, a similar strengthening effect does not occur for alumina, restricting that the sandblasting operation shall not be too vigorous to induce larger surface flaws. For future development, it would be desirable to enhance the interfacial reactive bonding by surface chemical modification of dental cores in order to avoid delamination in alumina and zirconia based prosthesis.

Conclusion

By close inspection of the interfacial characteristics of four widely using bilayered ceramic dental composites on well polished cross-sections, it was revealed that the interfacial bonding can be classified into two major categories, namely, adhesive bonding and reactive bonding. Zirconia and alumina ceramic cores demonstrated adhesive bonding with silicate veneers, whereas lithium disilicate glass ceramic and glass infiltrated alumina are both reactively bonded to silicate porcelains. A newly developed argon ion beam cross-section polishing technique was proved to be a feasible technique to achieve the gentle and fine polishing required for such high resolution SEM studies. These results bear implications for dental practice and for further development of dental cores and porcelains.

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