Compositional and microstructural design of highly bioactive
P_2O_5–Na_2O–CaO–SiO_2 glass-ceramics

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1. Introduction

One of the great challenges in the field of biomaterials is developing one that matches the biomechanical properties of bone and also has sufficient bioactivity to bond to living bone and soft tissues [1,2]. Biocomposites come close to achieving this goal but are usually limited in clinical applications due to unstable interfaces. Another approach has been to improve the mechanical properties of bioactive ceramics by formation of a glass-ceramic, such as the A/W glass ceramic Cerabone®, pioneered by Yamamuro and Kokubo [3,4]. This A/W glass-ceramic has been used successfully in about 60,000 cases, including vertebral replacement and iliac crest repair [4]. The high strength and high toughness of this A/W glass-ceramic makes it a good load-bearing replacement for cortical bone. However, its elastic modulus is about one order of magnitude higher than that of cortical bone, giving rise to the possibility of long-term stress shielding when the material is used as a bone replacement [5]. Also, its level of bioactivity is insufficient for bonding to soft connective tissues [5,6], as is needed for some clinical applications.

The effects of crystallization of monolithic and powdered glass ceramics of the system P_2O_5–Na_2O–CaO–SiO_2 on the activity of osteoblast cultures in vitro, in simulated body fluid (SBF) and in animals and humans have been studied previously [1,2,6–8]. These glass-ceramics show similar levels of bioactivity to the “golden standard” 4555 Bioglass®. In a previous work with glasses within the composition range 1Na_2O–2CaO–3SiO_2 and 1.5Na_2O–1.5CaO–3SiO_2, we showed that crystallization slows down, but does not inhibit the development of a crystalline hydroxyl carbonate apatite (HCA) layer, even on fully crystallized glass-ceramics [1]. The range of onset times for crystallization of HCA in tests in vitro using SBF-K9 varied from 8 h for a 1.5Na_2O–1.5CaO–3SiO_2 glass containing 6% P_2O_5 to 32 h for a fully crystallized 1.07Na_2O–2CaO–3SiO_2
glass-ceramic without phosphorous [1]. Nonetheless, in vitro HCA layer formation on these glass-ceramics, even for the least bioactive, is much faster than on partially or fully crystallized commercial bioceramics, such as hydroxyapatite and Cerabone, which usually takes at least 7 days. Peitl et al. [1] concluded that two simultaneously acting factors are responsible for the high bioactivity level of these glass-ceramics: (a) a highly soluble crystal phase (1N2C3S) and (b) the existence of phosphorus ions in solid solution, which can be readily released from the crystal structure. Both factors contribute to faster HCA layer formation on glass-ceramics of this system by the same reaction mechanisms observed for 45S5 Bioglass®.

The reaction mechanism of HCA layer formation in vitro in SBF K-9 on the glass and corresponding glass-ceramics studied by Peitl et al. [1,2] followed the same five-stages proposed by Hench [5] for 45S5 Bioglass®, summarized as:

a. Stage (I). Rapid exchange of Na⁺ with H⁺ or H₂O⁻ from the solution:
   \[ \text{Si-O-Na}^+ + \text{H}^+ + \text{OH}^- \rightarrow \text{Si-OH} + \text{Na}_\text{solution}^+ + \text{OH}^- \]

b. Stage (II). Loss of soluble silica as Si(OH)₄ to the solution resulting from breakage of Si-O-Si bonds and formation of Si-OH (silanols) at the glass/solution interface:
   \[ 2(\text{Si-O-Si}) + 2(\text{OH-Si}) \rightarrow \text{Si-OH} + \text{OH-Si} \]

c. Stage (III). Condensation and repolymerization of a SiO₂-rich layer on the glass surface that is depleted in alkali and alkaline earth cations:
   \[ 2(\text{Si-OH}) + 2(\text{OH-Si}) \rightarrow \text{Si-O-Si-O-Si-O-Si-O-Si-O-} \]

d. Stage (IV). Migration of Ca²⁺ and PO₄³⁻ to the glass surface through the SiO₂-rich layer forming a CaO-P₂O₅-rich film on top of the SiO₂-rich layer followed by growth of an amorphous CaO-P₂O₅-rich film by incorporation of soluble calcium and phosphate from solution.

e. Stage (V). Crystallization of the amorphous CaO-P₂O₅ film by incorporation of OH, CO₃²⁻ or F anions from solution to form a mixed HCA or hydroxy fluorapatite (HCFA) layer.

These five stages go to completion within 2–5 h for glasses and glass-ceramics with the highest bioactivity [1,2,5].

Despite the vast literature describing the mechanical properties of different types of glass-ceramics [9–16], to the best of our knowledge the effects of the crystallized volume fraction at constant grain size and of varying grain size with a constant crystallized volume fraction on the mechanical properties have not yet been established. There are only a few studies of the mechanical properties of bioactive glass-ceramics in general and in particular for this family of highly bioactive glass-ceramics. Several papers have compared the mechanical behavior of the parent glasses with (almost) fully crystallized glass-ceramics [9–11], but without consideration of partially crystallized glass-ceramics of varied microstructure. Some studies have compared the mechanical behavior of glass-ceramics with varying crystal size for a fixed volume percentage of crystal phase whereas others have described the effect of the volume fraction of the crystal phase, but neglected the crystal size effect.

In this work we independently focused on these two interdependent microstructural effects, plus the possible effect of residual stress on the mechanical behavior of glass-ceramics in general and bioactive glass-ceramics in particular. Our goal was to obtain an optimized bioactive glass-ceramic that combines improved mechanical properties, such as a relatively low elastic modulus, high fracture strength and relatively high toughness, with a bioactivity level comparable with that of 45S5 Bioglass®. Our development resulted in a patent filed in the USA and in the European Community [17], but we only recently completed an effort to understand the mechanical behavior of this glass-ceramic system, as reported here.

2. Experimental procedure

High purity silica and reagent grade calcium carbonate, sodium carbonate and sodium phosphate were used to obtain glasses of approximate compositions 1.07N2C3S and 1.5Na₂O–1.5CaO–3SiO₂ with 0, 1, 2, 4 and 6 wt.% P₂O₅. The exact nominal compositions are given in Table 1. The appropriate chemicals were weighed and mixed for 2 h in a polyethylene bottle. Premixed batches were melted in a platinum crucible in the temperature range 1320–1400 °C for 4 h and the melts were cast into cylindrical 8 × 30 mm graphite molds. The resulting glass cylinders were used to measure nucleation and growth rates and the elastic modulus.

The strategy used to study and control crystallization follows that proposed by Kalinin and Filipovitch [18], who employed a two-step heat treatment to measure nucleation rates. The procedure consists of growing the nuclei formed in the first step by a second heat treatment at a higher temperature. The number of crystals per unit volume (Nᵥ) of glass produced at different temperatures was calculated from Nᵥ, the number of crystals per unit area measured on optical micrographs using a stereological equation developed by De Hoff and Rhines [19]:

\[ Nᵥ = \frac{N₂F}{z} \]

where F is a geometrical factor equal to 1 for cubic crystals and 2/π for spherical crystals and z is:

\[ z = \frac{\sum 1/d}{n} \]

where d is the crystal diameter and n is the number of crystals measured.

The crystal growth rates were determined by optical microscopy considering the size of the biggest crystal (the primogenitor) as a function of time and temperature for the second treatment. Measurements were taken on polished and HF etched surfaces. Table 2 shows the range of temperatures and times used to develop the desired microstructures. The glass transition temperatures (Tᵥ) were determined using differential scanning calorimetry (DSC) to study the effect of P₂O₅ content on the nucleation rates. From the DSC curves the Hruby parameter Kᵥ was calculated to estimate the glass stability against crystallization during heating;

<table>
<thead>
<tr>
<th>Glass composition (wt.%)</th>
<th>Component (wt.%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>SiO₂ Na₂O CaO P₂O₅</td>
<td></td>
</tr>
<tr>
<td>1.07N2C3S</td>
<td></td>
</tr>
<tr>
<td>50.3 18.5 31.3 –</td>
<td></td>
</tr>
<tr>
<td>1.5N1.5C3S</td>
<td></td>
</tr>
<tr>
<td>50.5 24.8 24.8 –</td>
<td></td>
</tr>
<tr>
<td>1.5N1.5C3S + 1P</td>
<td></td>
</tr>
<tr>
<td>50.0 24.5 24.5 1.0</td>
<td></td>
</tr>
<tr>
<td>1.5N1.5C3S + 2P</td>
<td></td>
</tr>
<tr>
<td>49.5 24.2 24.2 2.0</td>
<td></td>
</tr>
<tr>
<td>1.5N1.5C3S + 4P</td>
<td></td>
</tr>
<tr>
<td>48.5 23.8 23.8 4.0</td>
<td></td>
</tr>
<tr>
<td>1.5N1.5C3S + 6P</td>
<td></td>
</tr>
<tr>
<td>47.5 23.2 23.2 6.0</td>
<td></td>
</tr>
</tbody>
</table>

Table 2
Thermal treatment ranges used to produce different microstructures for the compositions studied.

<table>
<thead>
<tr>
<th>Composition</th>
<th>Nucleation</th>
<th>Growth</th>
<th>Crystallized volume (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>T (°C)</td>
<td>t (min)</td>
<td>T (°C)</td>
</tr>
<tr>
<td>1.07N2CS</td>
<td>600</td>
<td>960</td>
<td>690</td>
</tr>
<tr>
<td>1.5N1.5CS</td>
<td>520–560</td>
<td>3–180</td>
<td>620–640</td>
</tr>
<tr>
<td>1.5N1.5CS + 4P</td>
<td>540–590</td>
<td>30–6000</td>
<td>650–700</td>
</tr>
<tr>
<td>1.5N1.5CS + 6P</td>
<td>540–590</td>
<td>60–9000</td>
<td>650–700</td>
</tr>
</tbody>
</table>

\[
K_{GL} = \frac{T_C - T_G}{T_C - T_M} \quad (3)
\]

where \(T_C\) and \(T_G\) are the glass transition and crystallization temperatures, respectively, and \(T_M\) is the melting point of the crystalline phase. According to Cahalal et al. [20] and Nascimento et al. [21] \(K_{GL}\) is one of the best parameters to estimate glass stability, i.e. its resistance to crystallization on the heating path from room temperature to above \(T_C\).

Some mechanical properties (flexural strength, elastic modulus, microhardness and indentation fracture toughness) were measured using rectangular bars with the dimensions 5 × 3.5 × 35 mm. The effect of the crystalline volume fraction at a constant grain size was evaluated for five different percentages of crystals: 0%, 15%, 34%, 60% and 100% with an average grain size of 13 μm (with a very narrow grain size distribution). For the crystalline volume fractions that presented the best flexural strength and indentation fracture toughness we developed thermal treatments to produce microstructures having the widest possible range of crystal sizes without adding any nucleating agent. In this way the mechanical properties of microstructures with crystal sizes in the range 5–21 μm and a constant crystalline volume fraction were investigated.

Flexural strength \(S_f\) of the rectangular specimens was measured in four-point bending tests following the ASTM procedure [22] and is given by:

\[
S_f = \frac{3PL}{2bd^2} \quad (4)
\]

where \(P\) is the load required to break the specimen, \(L\) is the distance between the adjacent support and loading edges and \(b\) and \(d\) are the width and thickness of the specimen, respectively. For each condition a minimum of six specimens were tested. The glass bar surfaces were polished prior to testing with cerium oxide with an average grain size of 6 μm.

The elastic modulus \(E\) was measured by an echo pulse technique with a frequency of 30 MHz. The glass density was 2.57 g cm\(^{-3}\) and density of the fully crystallized material was 2.67 g cm\(^{-3}\). For glass-ceramics with intermediate volume fractions of crystals the densities were between these two values. The overall uncertainty in \(E\) was estimated at 5%.

The Vickers hardness \(H_V\) was measured inside the crystals and in the residual glassy phase. The measurements were performed with loads of 100 and 250 mN in an Olympus instrument with a dwell time of 15 s. A minimum of 10 indentations were performed for each condition.

The indentation fracture toughness \(K_C\) was estimated by two indentation methods [23,24] using a Vickers indenter applying enough load to produce radial cracks as shown in Fig. 1. The first method, proposed by Baikova et al. [23], estimates \(K_C\) according to the equation:

\[
K_C = 0.244x^{-1} \left(\frac{P}{H^2}\right)^{0.4} \left(\frac{P}{c^2}\right)^{0.6} \left(\frac{C}{c}\right)^{0.6} \quad (5)
\]

where \(P\) is the applied load, \(c\) is the radial crack length measured from the center of the impression, \(a\) is the indentation half-diagonal and \(x\) is given by:

\[
x = 14 \left(1 - 8\left(\frac{\nu - 0.5}{1 + \nu}\right)^4\right) \quad (6)
\]

where \(\nu\) is the Poisson ratio. After indentation some samples were broken in bending for fracture surface analysis. This method was applied for samples with different crystalline volume fractions and at least 10 indentations were performed for each sample condition. Poisson ratios of 0.229 and 0.217 were assumed for the glass and precipitate, respectively [25].

Some samples were tested using a second method to measure \(K_C\) proposed by Chantikul et al. [24]. In this method indentation is performed at the surface and the sample is tested under bending until fracture. \(K_C\) in this method is estimated as:

\[
K_C = \frac{E}{H^2} \left(\sigma F^{\nu/3}\right)^{3/4} \quad (7)
\]

where \(\eta\) is a constant equal to 0.6 ± 0.1, \(P\) is the indentation load and \(\sigma\) is the stress at fracture. The advantage of this method is that no measurement of crack length is required. Loads from 0.1 to 3 N were used to introduce the indentation pre-cracks and samples were tested under four-point bending tests. At least five tests were performed for each sample and load condition.

In order to estimate the residual stresses the thermal expansion coefficients of the 1.5N1.5CS + 4P glass and its fully crystallized sample were measured by dilatometry using a Netzsch DIL 402 PC dilatometer in air. The temperature range considered for the glass was 50–450 °C, approximately 80 °C below \(T_G\).

3. Results

3.1. Crystal nucleation and growth rates

Fig. 2 shows the nucleation rates for glasses 1.07N2CS, 1.5N1.5CS, 1.5N1.5CS + 4P and 1.5N1.5CS + 6P. P_2O_5 has a dramatic effect on the nucleation rate of this system; each 1% of added...
P₂O₅ reduces the nucleation rate by about one order of magnitude. Therefore, a glass-ceramic with a fine microstructure with a high P₂O₅ content cannot be produced by homogeneous nucleation. For example, to produce a fully crystallized glass-ceramic with a grain size of 13 μm and 6 wt.% P₂O₅, the nucleation time at the temperature of maximum nucleation rate was 150 h.

The temperature of maximum nucleation rate Tₘₐₓ (indicated by arrows in Fig. 2) did not change on increasing the P₂O₅ content, but strongly depended on the Na₂O content. For instance, Tₘₐₓ is reduced by 50 K for a Na₂O increase from 1.07 to 1.50 mol. Na₂O content also strongly affects the nucleation rate. For the compositional range studied, an increase in Na₂O from 1.07 to 1.5 mol with a simultaneous reduction in CaO from 2 to 1.5 mol raises the nucleation rate by about two orders of magnitude, as shown in Fig. 2.

Table 3 shows the Tₕ as determined by DSC. These measurements show the effect of viscosity on the nucleation rate dependence with P₂O₅ content shown in Fig. 2. For compositions without P₂O₅, Tₕ decreases with increasing Na₂O content. For compositions containing P₂O₅ we cannot attribute its effect on the nucleation rate purely to viscosity changes. In the system 1.5N1.5C3S + P with 6 wt.% P₂O₅ (Table 3), the Tₕ increment was only 11 K, but the reduction in nucleation rate was five orders of magnitude. The Hruby parameter values KGL calculated according to Eq. (3) are also displayed in Table 3 for the different glass compositions. The higher the concentration of P₂O₅, the higher is KGL.

Fig. 3 shows the crystal growth rates for the 1.5N1.5C3S, 1.5N1.5C3S + 4P and 1.5N1.5C3S + 6P glass compositions. P₂O₅ decreases the growth rate in the same manner as observed for the nucleation rate. For a given crystal growth rate (in the range 1–2 μm min⁻¹) the temperature increases by about 7 K for each 1% P₂O₅ added in glasses with up to 4 wt.% P₂O₅, and 12 K in glasses with 4–6 wt.% P₂O₅.

### Table 3

<table>
<thead>
<tr>
<th>Composition</th>
<th>Tₕ (onset)</th>
<th>KGL</th>
</tr>
</thead>
<tbody>
<tr>
<td>1.07N2C3S</td>
<td>576</td>
<td>0.28</td>
</tr>
<tr>
<td>1.5N1.5C3S</td>
<td>519</td>
<td>0.25</td>
</tr>
<tr>
<td>1.5N1.5C3S + 1P</td>
<td>515</td>
<td>0.29</td>
</tr>
<tr>
<td>1.5N1.5C3S + 2P</td>
<td>517</td>
<td>0.34</td>
</tr>
<tr>
<td>1.5N1.5C3S + 4P</td>
<td>520</td>
<td>0.50</td>
</tr>
<tr>
<td>1.5N1.5C3S + 6P</td>
<td>531</td>
<td>0.81</td>
</tr>
</tbody>
</table>

### 3.2. Mechanical properties – different crystallized volume fractions with constant crystal size

#### 3.2.1. Flexural strength

Fig. 4a shows the flexural strength Sₜ of glass-ceramic 1.5N1.5C3S + 4P as a function of the crystallized volume fraction with constant grain size (13 μm). The average Sₜ increases 2.7 times, from 80 (glass) to 214 MPa for the 34% crystallized glass ceramic. In this range of crystallinity (0–34%) the increase in flexural strength is practically linearly dependent on the volume percentage of crystals. Higher percentages of crystalline phase cause some decrease in strength, but the vast majority of glass-ceramics are stronger than the parent glass by at least 100%. Fig. 4b shows the corresponding optical micrograph for each crystallized fraction. These designed microstructures were obtained by controlling the nucleation and growth kinetics (time and temperature of the two thermal treatments (Table 2)) without adding any nucleating agent. The highest flexural strength is observed for a crystallized volume fraction of 34%, and is thus not exhibited for the fully crystallized glass-ceramic, although the values are equal within the statistical error.

#### 3.2.2. Elastic modulus

Fig. 5 shows the elastic modulus as a function of crystallized volume fraction for three different compositions. Crystallization
always increases $E$. For instance, there is a linear increase in $E$ for glass 1.5N1.5C3S + 4P as a function of crystallization. The Voigt model ($E_{GC} = \frac{1 - \alpha}{C_0} E_G + \alpha E_C$, where $E_{GC}$, $E_G$ and $E_C$ are the elastic modulus of the glass-ceramic, glass and fully crystallized sample, respectively) can be applied and predicts a linear dependence of $E$ on $\alpha$. However, this model assumes that the strain in each phase is the same, and it only gives an upper bound for the ideal case of the "law of mixtures". Fig. 5 shows that the solid straight line calculated using the Voigt model for 1.5N1.5C3S + 4P does not accurately describe the experimental results but gives the correct tendency.

3.2.3. Microhardness

Fig. 6 shows the hardness variation with different crystalline volume fractions of samples with the compositions 1.5N1.5C3S + 4P and 1.5N1.5C3S + 6P. $H_v$ of both the crystal and residual glass phase were measured. The thermal treatment schedule was the same used to prepare the samples for flexural strength measurements, so the crystal size was kept constant at 13 μm in all samples measured.

The hardness of the glass phase increases with the percentage of crystalline phase up to about 40%. It does not change significantly between 40 and 60%, and finally decreases for higher volume percentages of crystallinity. The hardness of the crystal phase does not change until 60% crystalline volume percentage and then exhibits a slight decrease. The hardness of the crystalline phase is always higher than that of the glass phase.
3.2.4. Indentation fracture toughness

The indentation fracture toughness ($K_C$) increases with crystalline volume fraction for both compositions studied, as shown in Fig. 7. Despite the fact that we did not observe a significant effect on $K_C$ of changing the P2O5 content from 4 to 6%, the variation in crystalline volume fraction shows a strong effect on $K_C$. Indentation fracture toughness thus exhibits the same behavior as observed for flexural strength and microhardness, i.e. it linearly increases with crystalline volume percentage, reaches a maximum, and then decreases for higher amounts of crystalline phase.

Fig. 8 shows fracture surfaces after flexural tests of the 1.5N1.5C3S + 4P glass-ceramic with different volume fractions of crystal phase. On the microscopic scale the parent glass (Fig. 8a) shows a smooth fracture surface, indicating that there is no mechanism to promote flaw deviation. Fig. 8b shows the fracture surface of a 34% crystalline volume fraction sample. The crack path is generally inside the residual glass phase. The crack surface shows considerable deflection from the crystalline precipitates, with tilting and twisting. The fully crystalline sample presents a rough fracture surface due to the change in direction when the crack front passes from one crystal to another, as observed in Fig. 8c.

3.3. Mechanical properties – different grain size with constant crystalline volume fraction

3.3.1. Flexural strength and indentation fracture toughness

Fig. 9 shows the flexural strength as a function of grain size for samples with 34% and 60% crystallization. The average flexural strength for samples having a crystalline volume percentage of 60% is close to but slightly lower than for a crystalline volume fraction of 34% for all crystal sizes measured. Furthermore, the values of $S_F$ for both crystalline volume percentages follow the same pattern. $S_F$ apparently increases with crystal size, reaches a maximum for 13 μm, and decreases for 21 μm. Hence, the maximum flexural strength is exhibited by samples having a crystalline volume percentage of 34% and average crystal size of about 13 μm (80 (glass)-214 (glass-ceramic) MPa). The dependence on grain size is not so strong and $S_F$ varies only 17% from the lowest value to that for the optimum grain size.

Fig. 10 summarizes $K_C$ variation with grain size for glasses 1.07N2C3S, 1.5N1.5C3S + 4P and 1.5N1.5C3 + 6P with various degrees of crystallinity. There is no dependence of $K_C$ on crystal size for a constant crystallized volume fraction. The indentation fracture toughness of the fully crystallized samples does not vary significantly with crystal size and composition. In all situations the $K_C$ values of the crystallized glass-ceramics are higher than those of the parent glasses.

Fig. 11 shows the variation in $K_C$ for the parent glass and 34% crystallized volume fraction and fully crystallized glass-ceramics measured by the bending method. $K_C$ values are lower than that measured by the indentation method but the normalized $K_C$ in relation to the glass shows the same trend as observed before: it is maximum for the 34% crystallized sample with a 13 μm crystal size.

The thermal expansion coefficients for the 1.5N1.5C3S + 4P glass and the fully crystallized sample were 15.3 × 10⁻⁶ and 13.8 × 10⁻⁶ °C⁻¹, respectively. Therefore, the glass matrix contracts more than the crystal on cooling from $T_G$. In this case we expect the crystal precipitates to be under compression and the matrix to be under radial tensile stress. $T_G$ measured by dilatometry was 537 °C for the glass, being slightly higher than that measured by DSC (520 °C), as reported in Table 3. The phase
transition from high to low combeite at 260 °C, which is normally observed on cooling the fully crystallized stoichiometric 1N2C3S composition [25], was not observed for our material. This indicates that excess Na stabilizes the low combeite phase, and this has been confirmed previously by X-ray diffraction [1]. It has been shown by Fokin and Zanotto [26] that a metastable solid solution with the composition Na\(_{4+x}\)Ca\(_{4-x}\)[Si\(_6\)O\(_{18}\)] (0 < x < 1) crystallizes in this glass family. During crystallization, even in the stoichiometric glass, the average composition continuously changes and reaches the exact 1N2C3S composition only when the material is fully crystallized. We expect the crystallization of this same solid solution for the off-stoichiometric glasses studied here.

The change in stoichiometry also affects the thermal expansion of the glass and crystals. Due to the excess Na in the crystals and its depletion in the glass the thermal expansion of the crystalline phase is decreased and that of the glass is increased from 13.7 \(\times\) 10\(^{-6}\) to 15.3 \(\times\) 10\(^{-6}\) °C\(^{-1}\) compared with the 1N2C3S stoichiometric glass [25]. The main consequence of the change in thermal expansion coefficients is to change the magnitude and sign of the residual stresses. In the stoichiometric glass-ceramic of this family the glass is under radial compression and the crystalline phase is under tension [25]. In the present off-stoichiometric glass-ceramics the opposite is observed.

These changes in the thermal expansion coefficients of the glass and the crystal phase affect the type (tensile × compressive) and intensity of the residual stresses. As we will discuss in the next section, residual stresses do not significantly affect the toughness of these glass-ceramics and the changes in the chemical composition do not significantly change other mechanical properties, such as the elastic modulus and hardness.

4. Discussion

Gonzalez-Oliver et al. [27] previously showed that the maximum nucleation rate of these soda lime–silica glasses strongly
The change in viscosity is confirmed by our DSC measurements, shown in Table 3. For compositions without P2O5, Tc decreases with increasing Na2O content, as predicted by Gonzalez-Oliver et al. [27]. For compositions containing P2O5 we cannot attribute its effect on the nucleation rate purely to viscosity changes. In the system 1.5N1.5C3S + P with 6 wt.% P2O5 the Tc increment was only 11 °C, but the reduction in nucleation rate was five orders of magnitude. The reason why P2O5 is such a strong anti-nucleant should be further explored.

Table 3 displays the values of the Hruby parameter for different glass compositions. It is known that the higher the value of KGL, the higher is the glass stability against crystallization during heating. Therefore, from Fig. 2 we expected the lowest stability for compositions containing P2O5. For compositions containing P2O5 we cannot attribute its effect on the nucleation rate purely to viscosity changes. In the system 1.5N1.5C3S + P with 6 wt.% P2O5, the Tc increment was only 11 °C, but the reduction in nucleation rate was five orders of magnitude. The reason why P2O5 is such a strong anti-nucleant should be further explored.

We observed a three-fold increase in flexural strength and a 40% increase in indentation fracture toughness in a sample with 34% crystallized volume fraction over the values of the parent glass. While the flexural strength was constant for higher volume fractions, the indentation fracture toughness decreases slightly. We did not observe any variation in these properties with grain size in the range of crystal sizes used in this study.

Three main mechanisms are usually considered to explain the toughness increase in glass-ceramics: residual stresses, crack deflection processes and elastic modulus [28–30].

Residual stresses arise in glass-ceramics on cooling the sample from high temperature due to thermal expansion differences between the precipitates and the glass matrix. If the thermal expansion of the crystalline phase xθp is lower than the thermal expansion of the matrix xθm, homogeneous compressive stresses will develop inside the precipitates, given as [31]:

$$\sigma_p = \frac{\Delta T}{\frac{1}{E_m} + \frac{1}{E_p} + \frac{1}{E_0}}$$

where E is the shear modulus, K is the bulk modulus and f is the volume fraction of precipitates. The subscripts m and p refer to matrix and particle, respectively. ΔT is the difference between the linear thermal expansion of the precipitate and the matrix and ΔT is the difference between Tc (when the glass stops to flow on cooling) and room temperature.

In the matrix, the stresses decay with increasing distance from the precipitate. The radial component is compressive whereas the tangential components are tensile in our samples. The average stress in the matrix is the opposite to that in the precipitate (they are compressive in the precipitates) and is given by the equilibrium condition:

$$f \sigma_p + (1 - f) \sigma_m = 0$$

Fig. 12 shows the variation in the residual stresses in the precipitate and in the glass as a function of the crystallized volume fraction as predicted by Eqs. (8) and (9). The stresses varied almost linearly and while the stress is compressive in the crystal and maximum at low volume fractions, the average stress in the glass is tensile and increases with f.

In order to understand the increase in flexural strength of the glass-ceramics, we estimated the critical flaw size that caused catastrophic failure in four situations.

The first assumes that residual stresses play no role in the measured fracture strength, and the critical flaw size is calculated for comparison. For simplicity, we assumed a semicircular flaw with radius c at the specimen surface equal to $K_{IC}/\sigma_y f$. $K_{IC}$ is the fracture toughness of the glass-ceramic, Y is a geometric constant equal to 1.28 and $\sigma_y$ is the fracture stress. Because there is no available data for $K_{IC}$ for different volume fractions for these glass-ceramic, we assumed the indentation fracture toughness $K_C$ as a reasonable approximation for $K_{IC}$.

The second and third cases consider the influence of residual stresses (and its variation with crystallized volume fraction) for a crack nucleated in the glass matrix or in the crystallized region. We simply deduced or added the calculated residual stress from Fig. 12 to the applied stress, depending on whether the crack is in the glass or in the precipitate, and use the appropriate fracture toughness for each volume fraction.
The fourth situation is for a crack nucleated in the glass around a precipitate under the effect of a remote (tensile) stress. Green [32] considered the effect of stress magnification around the precipitate when an external stress is applied. By considering the conditions for the nucleation of an annular crack around the inclusion he was able to calculate the critical radius of the precipitate for crack nucleation. Once this crack is nucleated, it may cause failure of the specimen. The critical radius $R_c$ is:

$$R_c = \frac{\pi K_c^2 (x + 1)^2}{2(\sigma_\infty^2 (x + 1)^2 - \sigma_p)^2 (x + 2)}$$

where $x$ is a geometrical parameter equals to $\pi a/2$. The experimental values for $\sigma_p$ and $K_c$ for each volume fraction were used.

The calculated values are shown in Table 4 for each situation. The estimated defect sizes in the glass are considerably smaller than in the precipitate due to the tensile stresses in the glass. Also, the residual stresses reduce the critical defect size. The critical defect sizes calculated considering residual stresses in the glass and under the influence of a remote stress are approximately the same as the crystallized volume fraction increases. They are in the range 4–6 $\mu$m and may be defects produced during handling the sample.

Based on these calculations, the critical defects from which fracture originated were (with high probability) located in the glass matrix due to its smaller critical size. The mean free path between the crystals is also shown and is given as $2d(1 - f)/3f$, where $d$ is the precipitate diameter.

The effect of the residual stresses on flexural strength is displayed in Fig. 13a and b, where the calculated tensile residual stresses in the glass according to Eq. (9) were added to the flexural strength of the glass-ceramic. The calculated strength not only underestimates the experimental strength but actually predicts an opposite trend: a slight decrease in strength with increasing $f$. This indicates that another mechanism is responsible for the increase in flexural strength.

It is well known that glass-ceramics are harder and tougher than their parent glasses. Therefore, it is expected that crystallization is able to reduce the flaw size compared with the glass. This may be due to several factors: defect size reduction produced by the crystallization thermal treatment, a reduction in the surface area of the glass (with lower toughness and hardness than the crystal) and an increase in the area of crystalline precipitates (with higher toughness and hardness than the crystal). We do not believe the increase in strength is related to fracture processes that occurred within the precipitates. This is because we varied the crystal size below and above all calculated defect sizes (4–22 $\mu$m for the 34% volume fraction) and no variation in flexural strength was observed. We also observed that the flexural strength does not vary with the mean free path between the precipitates in the range studied.

It is not clear yet why the flexural strength increases up to about 30% crystallized volume fraction and remains more or less constant until about 100% crystallization. It may be because the weakest phase is the glass, and it controls the fracture behavior at low volume fractions. As $f$ increases, toughening mechanisms become important to the point where the crystalline phase dominates. It is interesting to note that at a volume fraction of ~26% the crystals percolate and become a rigid network. At higher volume fractions, the crystals dominate the fracture process and toughening mechanisms in the glass are less effective.

One important point is that residual stresses cannot explain the experimental increase in strength with crystallized volume fraction in our system. As $f$ increases the residual tensile stresses in the glass increase and a reduction in strength would be expected. One factor that might contribute to the observed increase in toughness is a reduction in the glass volume as $f$ increases. This would be accompanied by an increase in the volume of the crystallized phase with reduced (although compressive) residual stress, which also contributes as a toughening mechanism. Precise calculation of this contribution requires a knowledge of the Weibull exponents for these samples and this is outside the scope of this article. Another factor that may contribute to the observed increase in strength is crack deflection, which is discussed below.

There is considerable controversy in estimating $K_{IC}$ for glass-ceramics using indentation methods: (a) the calculated toughness depends on the equation used [34,35]; (b) the estimated toughness of glass-ceramics usually differs from those measured by standard methods such as SENB [36,37]; (c) densification may occur in the indentation zone, which decreases the indentation induced residual stress field [38,39]; (d) changes in crack geometry from semicircular to Palmqvist and lateral cracks affect the stress distribution around the indenter [40]; (e) slow crack growth [41]; and (f) R-curve behavior [37]. On the other hand, the indentation method has been widely used because it is simple, easy and non-destructive. It also shows trends when comparing samples under different conditions, such as those used in this study. We will analyse our results based upon existing fracture toughness models, keeping all these limitations in mind.

Internal residual stresses have been measured before in other types of glass-ceramics and it has been thought to be important in their overall mechanical properties and fracture [42–44]. The effect of residual stresses on fracture toughness has been investigated by Taya et al. [28]. A crack subjected to varying residual stress will tend to rest in regions of compressive stress. An increase in toughness $\Delta K_I$ is observed if the residual stresses are compressive and is given by:

$$\Delta K_I = 2\sigma_m \sqrt{\frac{2(\lambda - d)}{\pi}}$$

where $\sigma_m$ is the residual stress given in Fig. 12, $\lambda - d$ is the free mean path, $\lambda$ is the distance between the precipitates and $d$ is the precipitate grain size. However, if the residual stresses are tensile, the toughness will decrease by $\Delta K_I$.

In our case there are two limits: at low crystallized volume fractions the radial residual stress in the matrix (glass) is tensile and thus a decrease in toughness is expected; at high crystallized vol-

<table>
<thead>
<tr>
<th>$f$ (%)</th>
<th>Critical flaw (μm)</th>
<th>Critical flaw size considering residual stresses (μm)</th>
<th>Mean free path</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>No residual stresses</td>
<td>Glass Precipitate</td>
<td>Critical radius for crack under remote stress</td>
</tr>
<tr>
<td>0</td>
<td>49 ± 25</td>
<td>49 ± 25</td>
<td>-</td>
</tr>
<tr>
<td>15</td>
<td>16 ± 5</td>
<td>12 ± 4</td>
<td>30 ± 11</td>
</tr>
<tr>
<td>34</td>
<td>13 ± 5</td>
<td>5 ± 2</td>
<td>11 ± 4</td>
</tr>
<tr>
<td>62</td>
<td>14 ± 4</td>
<td>6 ± 2</td>
<td>11 ± 4</td>
</tr>
<tr>
<td>100</td>
<td>10 ± 3</td>
<td>-</td>
<td>10 ± 3</td>
</tr>
</tbody>
</table>

Table 4

Critical semi-circular flaw size at the surface calculated from flexural strength data as a function of the crystallized volume fraction for S5P4 with a grain size of 13 μm, with and without considering the residual stresses. The critical grain size for spontaneous cracking considering an annular crack around the precipitate under a remote applied stress [32] and the mean free path are also presented.
ume fractions the residual stress within the precipitates (the major phase now) are compressive and thus an increase in toughness is expected.

The variation in $K_C$ at low and high volume fractions according to Eq. (11) is shown in Fig. 14a and b. For the high volume fraction end we used $d = 13 \mu m$ for the glass and estimated $i$ for the precipitate phase. At low $f$, the residual stress model predicts a decrease in $K_C$ with increasing $f$ due to the average tensile stress in the glass matrix. However, the opposite is observed experimentally, indicating that other toughening mechanisms are operating. At high values of $f$, the residual stress model predicts an increase in $K_C$ with decreasing $f$ due to presence of residual compressive stresses within the major crystalline phase. The agreement with the experimental data is reasonably good, although it slightly underestimates the toughness at intermediate values of $f$. For constant $f$, the model underestimates $K_C$ and predicts a weak dependence on crystal size, as shown in Fig. 14b, which is contrary to what was observed experimentally, at least for the size range in this work.

As a crack propagates in a glass ceramic it experiences the residual stress field. As in the present case the thermal expansion coefficient of the glass is higher than that of the precipitate, there are compressive radial and tensile tangential stresses around the precipitates and the crack is thus directed towards the precipitate. This promotes toughening as the surface area increases due to deflection of its original plane, and different modes of loading are induced as the crack becomes non-planar.

The contribution of crack tilt and twist as the crack propagates in the glass-ceramic has been estimated by Faber and Evans [30]. As a crack approaches the precipitate, it will tilt and twist. The amount of crack deflection depends on the position and orientation of the precipitates relative to the crack front. Tilting induces modes I and II in the crack front, while twisting induces modes I and III. The increase in the strain energy release rate $G$ of the deflected crack is:

$$EG = K_1^2 (1 - v^2) + K_2^2 (1 - v^2) + K_3^2 (1 - v^2)$$

where $K_1$, $K_2$ and $K_3$ are the local stress intensity factors as a function of the deflected angle. The toughening increment $G_C$ is $(G^m/G)G^m$, where $G^m$ is the strain energy release rate for an undeformed crack, $(G)$ is the average strain energy release rate across

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Fig. 13. Influence of residual stresses on flexural strength compared with experimental data for 1.5N1.5C3S + 4P glass-ceramic with (a) different crystalline volume fractions and (b) different crystal sizes.

Fig. 14. Influence of residual stresses on indentation fracture toughness compared with experimental data for 1.5N1.5C3S + 4P glass-ceramic with (a) different crystalline volume fractions and (b) different crystal sizes.
Energy release rate for the experimental data was calculated as Eqs. (2) and (15) in Faber and Evans [30] for details, and the calculation, when applied to our results, are shown in Fig. 15. The strain energy release rate for the experimental data was calculated as $K_e^2(1 - v^2)/E$. The increase is not significant when compared with the experimental data. Faber and Evans [30] also calculated the increase in toughness for the case where the spacing between particles is assumed to vary. If the distance follows a normal distribution with a standard deviation $\sigma$ equal to $0.2(\lambda/d)$, a significant increase in toughness is observed (see eq. (29) and fig. 6 in Faber and Evans [30] for details). For values of $\sigma$ higher than $0.2(\lambda/d)$, the toughness is insensitive to $\sigma$. Fig. 15 shows the calculated increase in toughening when the distance between particles is distributed with a standard deviation $\sigma$ equal to $0.2(\lambda/d)$. A significant increase in toughness is observed which is close to the experimental data and to the residual stress contribution. It also overestimates the increase in toughness at low volume fractions. A simple statistical analysis of the distance between the precipitates in Fig. 4 reveals that $\sigma$ in our glass-ceramics varies between 0.8(\lambda/d) and 1.3(\lambda/d) and, therefore, we assume the curve with distributed spacing in Fig. 15 corresponds to our experimental condition. The contribution to the surface area increase due to crack deflection was also calculated for the case of spheres by Farber and Evans [30]. In their model the increase in area considered an array of spherical precipitates.

The present results give one relevant example of the great flexibility of microstructural design of glass-ceramics as recently reported in a review paper [46].

5. Conclusions

We developed P$_2$O$_5$–Na$_2$O–CaO–SiO$_2$ bioactive glass-ceramics with tailored microstructures through two-stage thermal treatment and independently varied the crystalline volume percentage from about 15% to 100%, and the average crystal size from 5 to 20 $\mu$m. This enabled us to separately study, for the first time, the effect of crystalline volume fraction and crystal size on the mechanical properties of a bioactive glass-ceramic.

The introduction of crystallinity in this bioactive glass significantly improves its mechanical properties. In the best case the fracture strength increased from 80 to 210 MPa and the indentation toughness from 0.60 to 0.95 MPA.m$^{0.5}$, while the elastic modulus underwent only a small increase from 60 to 70 GPa. These overall improvements in mechanical properties were attributed to crack deflection mechanisms within the material. For a wide range of crystalline volume percentages, keeping the crystal size constant, the best mechanical performance was shown by the partially crystallized glass-ceramic with 34% crystallized volume. Flexural strength and fracture toughness show similar behavior with an increase in crystallinity: they both increase from 0% to 34% crystallinity and then slowly decrease. For a constant crystalline volume percentage (34% or 60%) an increase in crystal size from 5 to 21 $\mu$m did not have a significant effect, and the strength vs. grain size values were the same within the statistical variation.

The present results contribute to a better understanding of the influence of crystallinity and crystal size on the mechanical behavior of glass-ceramics. As we have shown in former publications, the in vitro bioactivity of our partially crystallized glass-ceramic is comparable with that of 45S5 Bioglass®. Therefore, a very important technological goal was achieved. The flexural strength of our best material is significantly greater than that of cortical bone and comparable with that of A/W bioactive glass-ceramics, with the advantage that it has a lower modulus of elasticity and a higher level of bioactivity. These results have technological relevance for the micro-structural design of glass-ceramics with high levels of bioactivity.

The results are also plotted in Fig. 15 and the contribution is small compared with tilt and twist. This model predicts no variation in toughness with crystal size, which is in agreement with our data shown in Fig. 10.

In summary, the main contribution to toughening in these glass ceramics appears to be crack deflection. Residual stresses do not significantly contribute to toughening; in fact they cause a decrease in toughness at low $f$ due to the increased average tensile stresses in the glass matrix. The main finding of the present research is that the estimated fracture toughness (using the same indentation technique) of our 34% crystallized glass-ceramic is about half those of the commercial glass-ceramics Cerabone and Bioverit (Table 5), but is twice as large as that of 45S5 Bioglass® (and other bioactive glasses). On the other hand, the elastic modulus of our partially crystallized glass-ceramic is much closer to that of cortical bone: the elastic modulus of bioceramics and other bioactive glass-ceramics are significantly higher, their average fracture strength is among the highest, they are reasonably machinable (i.e. they can be easily cut or drilled using a hand tool, for instance by a physician, with no cracking or spalling), and their bioactivity level is far the highest of all glass-ceramics (demonstrated in Peitl et al. [2]). Thus our best glass-ceramic is a unique bioactive ceramic that combines excellent mechanical properties with a high level of bioactivity. Clinical trials of this biomaterial are now underway [45].

The present results give one relevant example of the great flexibility of microstructural design of glass-ceramics as recently reported in a review paper [46].

Table 5

<table>
<thead>
<tr>
<th>Bioceramics</th>
<th>Modulus of elasticity (GPa)</th>
<th>Four-bending strength (MPa)</th>
<th>$K_e/K_c$ (MPa.m$^{0.5}$)</th>
<th>Bioactivity$^a$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cerabone A/W$^b$</td>
<td>220</td>
<td>215</td>
<td>2.0</td>
<td>3.2</td>
</tr>
<tr>
<td>Bioverit$^b$</td>
<td>90</td>
<td>160</td>
<td>1.2–2.0</td>
<td>3.0</td>
</tr>
<tr>
<td>Hydroxyapatite$^b$</td>
<td>120</td>
<td>40–70</td>
<td>$#$</td>
<td>$#$</td>
</tr>
<tr>
<td>Ceravital$^b$</td>
<td>150</td>
<td>$#$</td>
<td>$#$</td>
<td>5.6</td>
</tr>
<tr>
<td>Bioglass$^b$</td>
<td>50</td>
<td>70</td>
<td>0.5</td>
<td>12.5</td>
</tr>
<tr>
<td>GC 34%</td>
<td>70</td>
<td>100</td>
<td>1.0</td>
<td>10.0</td>
</tr>
<tr>
<td>(this work)</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Cortical bone$^c$</td>
<td>7–30$^b$</td>
<td>50–150</td>
<td>2–12</td>
<td></td>
</tr>
</tbody>
</table>

$^a$ [1–4].

$^b$ Results from tensile tests.

$^c$ Results from flexural tests.

*Fig. 15. Calculated effect of crack tilt and twist around precipitates on the strain energy release rate and experimental data for 1.5N1.5C3S + 4P glass-ceramic with different crystallized volume fractions. The contribution of surface area increase and the effects of crystals with uniform and distributed spacings are also shown.*
Acknowledgements

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Appendix A. Figures with essential colour discrimination

Certain figures in this article, particularly Figures 1, 4 and 8, are difficult to interpret in black and white. The full colour images can be found in the on-line version, at doi: 10.1016/j.actbio.2011.10.014.

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