Simultaneous mechanical property and biodegradation improvement of wollastonite bioceramic through magnesium dilute doping

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ABSTRACT

The large-area bone defects in head (including calvarial, orbital, and maxillofacial bone) and segmental bone are attracting increased attention in a wide range of clinical departments. A key requirement for the clinical success of the bioactive ceramics is the match of the mechanical behavior of the implants with the specific bone tissue to be filled. This raises the question as to what design strategy might be the best indicators for the balance between mechanical properties and biological performances. Here we go beyond the traditional approaches that use phase conversion or biphasic hybrid; instead, we achieved a simultaneous enhancement of several mechanical parameters and optimization of biodegradability by using a dilute doping of Mg in a single-phase wollastonite bioceramic. We show that the wollastonite ceramic can be rationally tuned in phase (α or β), mechanical strength (in compression and bending mode), elastic modulus (18–23 GPa), and fracture toughness (4.3 MPa m^{1/2}) through the usage of Mg dopant introduced at precisely defined dilute concentrations (Mg/Ca molar ratio: 1.2–2.1%). Meanwhile, the dilute Mg-doped wollastonite ceramics are shown to exhibit good bioactivity in vitro in SBF but biodegradation in Tris is inversely proportional to Mg content. Consequently, such new highly bioactive ceramics with appreciable strength and toughness are promising for making specific porous scaffolds for enhancing large segmental bone defect and thin-wall bone defect repair.

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1. Introduction

Trauma, infection, and congenital deformity can result in over critical-size bone defects or lost that cause a series of challenges for reconstructive surgeons. In cases where thin-wall calvarial, orbital, and maxillofacial bone defects, as well as load-bearing long bone defects are caused by high-energy trauma or debridement for infected tissues or tumors, the local blood circulation in bone bed is poor, and the specific requirement of mechanical properties and biological performances of the grafts lead to fail to survive (Kilinc and Aytekin, 2014; Kinnunen et al., 2010; Leu et al., 2009; Takumi and Akimoto, 2009; Zhou et al., 2011). Therefore, the materials should be able to fabricate complex shapes with desirable patient-tailored applications, meanwhile the promotion of bone tissue regeneration and in situ repair are the optimal directions in orthopedic research.

Since Hench et al. firstly discovered Bioglass® which was proved to possess excellent bioactivity and osteoconductivity in the early 1970s (Hench et al., 1971), Ca-silicate bioactive glasses, glass-ceramics, and bioceramics have been widely studied for their potential applications as hard tissue repair materials (De Aza et al., 2000; Gerhardt and Boccaccini, 2010; Kokubo et al., 1986; Sprio et al., 2009), special implants or prosthesis (Baino and Vitale-Brovarone, 2014, 2015; Huhtinen et al., 2013), as well as drug delivery (Soundrapandian et al., 2014). For bone regeneration, the artificial materials need to meet the requirements of good osteoconduction, osteoinduction, as well as controllable bioactivity and degradation (Hench and Polak, 2002). Besides, the bone implants should have considerable fracture strength and toughness which match the bone biomechanical requirement in thin-wall bone trauma.

In the last decade, β phase of wollastonite (CSI) has attracted significant attention among the most promising candidates due to its outstanding bioactivity (Siriphanon et al., 2002). De Aza PN and colleagues found that the CSI ceramic exhibited faster apatite formation than other CaP in human parotid saliva (De Aza et al., 2000). Some studies have also shown that Ca and Si ions play important roles in the formation of the apatite layer, and affect the biological metabolism of osteoblastic cells which indirectly influenced the mineralization process and bone-bonding mechanism (Mohammadi et al., 2014). However, the slightly faster degradation rate which matches with the bone tissue regeneration during the new bone ingrowth (Xu et al., 2008), especially, the relatively low fracture strength and toughness which fails to meet the specific mechanical requirement in thin-wall bone defects constrains its clinical applications (Bratton and Durairaj, 2011). To reinforce the structure and to enhance the mechanical strength, biopolymer (Shirazi et al., 2014b), metallic oxide (Shirazi et al., 2014a) and many other materials are used to modify the CSI based materials scaffolds (Mehrali et al., 2013).

The importance of magnesium (Mg) in human body, its key role in mineralization of calcined bone tissues and indirectly influences mineral metabolism (Althoff et al., 1982), in addition to its application to improve and modify physical, thermal, and mechanical properties of bioactive glasses and ceramics, make Mg a very interesting element as a component of bioactive materials for medical applications (Diba et al., 2014, 2012). In vivo studies have also proved that appropriate concentration of Mg ion is beneficial to vascularization (Witte et al., 2005). According to the above consideration, intensive research has been devoted to developing Ca–Mg–silicate ceramics to improve the osteogenesis (Hoppe et al., 2011), strength, and compositions of the bone implants. Chang’s group (Wu and Chang, 2006; Wu et al., 2005; Wu and Chang, 2013) and other researchers (Chen et al., 2008, 2010; Nonami and Tsutsumi, 1999) have investigated a series of studies involving stoichiometric compounds with Mg content increased from 3.61% for bredigite (Ca2Mg(SiO4)2), 7.39% for merwinite (CaMg(SiO4)), 8.92% for akermanite (CaMg2Si2O7), 11.22% for diopside (CaMgSi2O6), to 15.53% for monticellite (CaMgSiO4). These ceramics showed acceptable bioactivity, whereas the dissatisfaction mechanical strength and fracture toughness mismatch limited their applications.

In this work, we aimed to develop the Mg dilute doping CSI ceramics with varied Mg/Ca molar ratio of 3–10%, and studied their sintering behavior and mechanical properties. As expected, as Mg dilute doping may retain the crystalline phase of CSI when below 10% of Ca is substituted by Mg. The new bioceramics with appropriate Young’s modulus, high flexural strength and fracture toughness were obtained by using pressureless sintering technique. This study suggests that the Mg dilute doping is perfect for making new bio ceramic products with exceptionally improved mechanical properties for some challengeable bone defect repair.

2. Materials and methods

2.1. Preparation of ceramic powders

The Mg-doped CSI (CSI-Mgx) powders with different Mg/CaO ratio, x (x=3, 6, 10, 14 mol%), were synthesized through a chemical precipitation method. Briefly, Ca(NO3)2·4 H2O and Na2SiO3·9H2O were separately dissolved in de-ionized water in the concentration of 0.6 mol l−1 with certain proportion of Ca(NO3)2 replaced by Mg(NO3)2. Then the Na2SiO3 solution was dropped into the Ca(NO3)2/Mg(NO3)2 solution mixture under continuous stirring and the pH value maintained at 10.0–10.5. The powder precipitate was filtered, washed four times with de-ionized water, and finally washed by ethanol. The powder was dried at 80 °C, and then calcined at 950 °C for 150 min. The pure CSI powder was also synthesized as control while the other conditions remain the same. The ceramic powders were ground in a planetary ball miller (MP-2L; Chishun Sci&Tech Co., China) with 320 rpm, using 3.5 mm diameter Zirconia ball media in ethanol to obtain superfine powders (below 10 μm).

2.2. Preparation of CSI and CSI-Mgx ceramics

The CSI-Mgx green compacts with cylindrical (Ø6 × 2 mm, Ø8 × 10 mm, Ø25 × 4 mm) and cuboid (45 × 5 × 8 mm3) shapes were prepared with a pressure of 8 MPa. Specifically, the polyvinyl alcohol (PVA; ~6 kDa) solution (3.0% v/v) was prepared under magnetic stirring at 45 °C. Then the PVA
solution was dispersed to the ceramic powders with a solution-to-powder mass ratio of 0.8:1. The paste was stirred for 10 min and then kept at 60 °C overnight. Subsequently, the ceramic green compacts were prepared using a uniaxial pressure in stainless steel moulds and followed by isostatic pressing at 200 MPa. Finally, these samples were underwent a one-step sintering in a bench top muffle furnace (Hefei Kejing Co., China) for 3 h at the temperature of 1100 °C, 1150 °C, 1200 °C, and 1250 °C, and followed by cooling naturally. The heating rate was determined to be 2 °C/min for all samples. Similarly, the pure CSI ceramics were prepared as control at the same conditions.

2.3. Characterization of CSI and CSI–Mgx ceramics

Shrinkage of the sintered samples was measured with a horizontal-loading dilatometer, and the apparent density was measured by Archimedes’ method in distilled water at room temperature. The theoretical density of the wollastonite (triclinic, β phase) and pseudowollastonite (monoclinic, α phase) can be calculated based on unit crystal cell parameter according to the expressions valid for a three-point bending configuration of

\[ E_s = \frac{3FL}{8Bd^3} \]

where \( E_s \) is the stress (in this case \( E_s \)) and \( \varepsilon \) is the strain expressed in the percentage of compressive deformation. \( F_s \) was measured in a three-point bending strength configuration using the same machine with bars of ceramics at the dimensions of 45 × 8 × 6 mm³ (n=6). The outer span was 30 mm and the crosshead speed was 1.0 mm min⁻¹. The flexural strength (F₀) and Young’s modulus (E₀) according to the expressions valid for a three-point bending configuration of

\[ E_0 = \frac{PL^2}{4bd^3} \]

where \( P \) is the breakload and \( L \) the section area of the specimen. \( E_0 \) was calculated according to the stress–strain curves. To be specific, \( E_0 = \sigma/\varepsilon \), where \( \sigma \) is the stress (in this case \( E_0 \)) and \( \varepsilon \) is the strain expressed in the percentage of compressive deformation. The compressive strength (\( C_0 \), elastic modulus (\( E_m \), flexural strength (\( F_0 \)), and Young’s modulus (\( E_0 \)) were measured before and after soaking in the simulated body fluid (SBF) for 3 weeks. Specifically, \( C_0 \) and \( E_m \) of the ceramics (Ø8 × 10 mm²; n=6) were evaluated by computer-controlled Instron universal testing machine with a 20 KN load cell and a crosshead speed of 0.5 mm min⁻¹ according to the 5833 ISO standard with the following formulae of

\[ C_0 = \frac{F}{A} \]

\[ E_m = \frac{3FL}{2bd^2} \]

where \( F \) is the breakload and \( L \) the section area of the specimen. \( E_m \) was expressed in the percentage of compressive deformation. \( F_0 \) was measured in a three-point bending strength configuration using the same machine with bars of ceramics at the dimensions of 45 × 8 × 6 mm³ (n=6). The outer span was 30 mm and the crosshead speed was 1.0 mm min⁻¹. The flexural strength (\( F_0 \)) and Young’s modulus (\( E_0 \)) according to the expressions valid for a three-point bending configuration of

\[ F_0 = \frac{3FL}{2bd^2} \]

\[ E_0 = \frac{PL^2}{4bd^3} \]

where \( P \) is the breakload, \( L \) is the supporting span, \( b \) is the specimen width, \( d \) is the specimen thickness, and \( m \) the slope of the tangent to the initial line portion of the load deflection curve. The results are the average of at least six specimens. The fracture toughness, \( K_{foc} \), was measured by the procedure described in ASTM C 1421 at room temperature. The specimens were cut into 40 × 4 × 6 mm bar. A single-edge pre-cracked beam (SEPB) method was conducted and \( K_{foc} \) was calculated using the equation reported in ASTM C 1421-99 Standard (2001).

2.5. In vitro bioactivity of CSI–Mgx ceramics

The reagent-grade Ca(NO₃)₂·4H₂O, Mg(NO₃)₂·6H₂O, Na₂SiO₃·9H₂O, NaCl, KCl, NaHCO₃, MgCl₂·6H₂O, CaCl₂, KH₂PO₄ and Tris was used to prepare the SBF containing equal inorganic composition with human plasma (pH 7.40) and 0.05 M Tris buffer (pH ~7.25), respectively. The SBF (pH 7.40) with ion concentrations nearly equal to that of human blood plasma was prepared according to the process described by Kokubo and Takadama (2006).

For evaluation of in vitro bioactivity, the bioceramic discs (Ø6 × 2 mm²) were soaked in SBF at 37 °C for 96 h with a surface area-to-volume ratio of 0.1 cm⁻¹. At the present time point the pH value was measured using an electrolyte-type pH meter (PHS-2C; Jingke) and then 20% of supernatant was refreshed with fresh SBF for ion concentrations measurement by inductively coupled plasma atomic emission spectrometer (ICP-AES; Varian Co., USA). The specimens were gently rinsed with ethanol and dried at room temperature for SEM observation.

2.6. In vitro biodegradation of CSI–Mgx ceramics

The in vitro biodegradation of the specimens were respectively measured by soaking the ceramic samples (W₀, Ø25 × 4 mm²; n=4) in SBF and 0.05 M Tris buffer (pH 7.4) at 37 °C with a surface area-to-volume ratio of 0.1 cm⁻¹. 20% of the buffer was refreshed every 48 h. After the set soaking time, the samples were gently rinsed with ethanol, and dried at 80 °C to weight constancy (W₀) before weighing by using an electronic analytical balance (FA2104, Sartorius, Germany). The weight loss (degradation) was expressed as the following equation: weight loss = (W₀ − Wₜ)/W₀ × 100%. Meanwhile, the Ca, Mg, Si, P concentrations were measured by ICP-AES after certain period of soaking, and the pH value was also measured in the supernatant.

2.7. Statistical analysis

All the data were expressed as mean ± SD. Statistical analysis performed using the one-way ANOVA. In all cases the results were considered statistically significant with a p-value less than 0.05.

3. Results

3.1. Primary characterization of ceramic powders

CSI ceramic powders with and without dilute Mg doping (CSI–Mgx) were prepared with Mg substituting Ca by 0%, 3%, 6%, 10%, and 14% in molar percent. That is, Ca ions were partly substituted (0–14%) by Mg ions in the crystal lattice of pure wollastonite, and thus the new synthetic products such as CSI–Mg₃, CSI–Mg₆, CSI–Mg₁₀, and CSI–Mg₁₄ can be signed as Ca₀.97Mg₀.03SiO₃, Ca₀.94Mg₀.06SiO₃, Ca₀.90Mg₀.10SiO₃ and Ca₀.80Mg₀.20SiO₃, respectively. Theoretically, the Mg content in these crystalline products was 0%, 0.62%, 1.25%, 2.12%, and 2.95% in weight according to the stoichiometric compounds. According to ICP measurement, the actual Mg content in the
CSi-Mgx was 0.63%, 1.24%, 2.08%, and 2.94%, respectively. It should be mentioned that the sample CSi-Mg14 was only used for evaluation of phase composition, but not characterized on other properties due to existence of akermanite.

The processing techniques used for the preparation of the samples, the specimen codes, and sintering conditions were summarized in Table 1. The linear shrinkage of the as-sintered CSi–Mgx ceramics displayed a steady increase from 8.63% for pure CSi to 11.4–13.1% for CSi–Mg6 and CSi–Mg10, and the relative density increased from 89.5% to 91.7% for pure CSi to 89.0% to 93.8% for CSi–Mg6 and 93.4–98.3% for CSi–Mg10 when the sintering temperature varied from 1100 °C to 1200 °C.

Fig. 1 shows that the relationship between the crystalline phases and sintering temperature of the powder compacts. The diffraction peaks of CSi ceramic (1150 °C) were all belonged to the patterns of wollastonite-2M (β-CSi; PDF# 27-0088), indicating pure wollastonite nature (Fig. 1(A)). However, the ceramics (1250 °C) were all transformed into α-CSi (PDF# 31-0300; Fig. 1(B)). Moreover, the XRD patterns of the ceramics sintered at 1100 °C and 1200 °C were respectively identical to those sintered at 1150 °C (Not shown), suggesting a β-to-α phase transformation occurred between 1200 °C and 1250 °C.

### 3.2. Microstructure evaluation of ceramics

The fracture surfaces of ceramics are displayed in Fig. 2 revealing the influence of the Mg content on the internal microstructures of CSi ceramics. It indicates a transgranular type of fracture for the CSi and CSi–Mg3 (1100, 1150 °C), but both transgranular and intergranular types of fracture were observed for the CSi–Mg6 and CSi–Mg10 (1100, 1150 °C). At low temperature, i.e. 1100 °C, the shape of particles in CSi and CSi–Mg3 almost remained unchanged. At 1150 °C, the grains showed softening while beginning to bond to each other, and here the pores were connective open-pores. With increasing Mg/Ca up to 6–10%, the open-pores in the ceramics disappeared when sintering at 1100 and 1150 °C, and closed pores were noticeable. It was noticed that all the ceramics sintered at 1250 °C resulted in a significant grain growth, and there were few pores in ceramics.

### 3.3. Mechanical properties

The mechanical strength and modulus of the ceramics were showed in Fig. 3. Compressive strength (720–740 MPa) and
Fig. 2 – SEM images of the fracture microstructures of CSi and CSi-Mgx ceramics sintered at different temperatures. Arrow showing the connective pores. Bar: 10 μm.

Fig. 3 – Compressive strength (A), flexural strength (B), elastic modulus (C), and Young’s modulus (D) of the CSi and CSi-Mgx ceramics sintered at different temperatures.
flexural strength (120–146 MPa) of the CSi-Mg10 (1100, 1150 °C) were over 3-fold higher than those of pure CSi (1100, 1150 °C). A 100% increase for CSi-Mg6 (1150 °C) in comparison with the pure CSi (1150 °C) was obtained whereas the CSi-Mg3 composition did not show a significant increase in strength compared with the pure CSi except for sintering at 1150 °C. Moreover, the compressive and flexural strength of CSi-Mg10 (1200 °C) were much higher than other samples, but all of ceramics abruptly decreased the compressive strength (<200 MPa) and flexural strength (<45 MPa) when sintering at 1250 °C. Additionally, the elastic modulus and Young’s modulus of the CSi-Mg6 and CSi-Mg10 versus CSi (1100, 1150 °C) were significantly enhanced (p<0.05), and the CSi-Mg10 (1200 °C) showed significantly higher Young’s modulus (~15 GPa) than other samples sintered at the same temperature. Fracture toughness of the ceramics sintered at 1100 and 1150 °C were shown in Fig. 4. The average values (3.2–3.7 MPa m^{1/2}) of CSi-Mg6 and CSi-Mg10 (1100, 1150 °C) were over 3-fold greater than those of pure CSi (1100, 1150 °C). A 100% increase for the CSi-Mg3 (1150 °C) in comparison to the CSi (1150 °C) was obtained but it was only nearly 52% of the fracture toughness value for CSi-Mg6 (1150 °C).

3.4. Biodegradation evaluation in vitro
The biodegradation rate (i.e. solubility in vitro) of the CSi and CSi-Mgx ceramics sintered at 1100–250 °C was monitored by weight loss in two different types of buffer systems. As can be seen from Fig. 5(A–D), the pure CSi ceramic had the fastest degradation in Tris buffer and had lost almost 11–14% of the weight after 7 weeks (1176 h). The CSi-Mg3 and CSi-Mg6 samples had slightly slower degradation rates. And the CSi-Mg10 (1100, 1150 °C) showed a 100% increase for the CSi-Mg6 (1150 °C) in comparison to the CSi (1150 °C) was obtained but it was only nearly 52% of the fracture toughness value for CSi-Mg6 (1150 °C).
might be identified as the first apatite nuclei. After 96 h, all ceramic surfaces were covered with continuous CaP precipitates. Generally, the globular precipitates could be always existed in CSi–Mg6 and CSi–Mg10 (1150 °C), but almost no Mg was detected in the Tris buffer during the pure CSi immersing. In addition, it was indicated that the pH value in Tris buffer during the pure CSi immersing showed maximal variation, while the minimal variation existed in CSi–Mg10 (Not shown).

3.5 Apatite-forming ability evaluation in vitro

The bioactivity in vitro was characterized by using SEM-EDX analysis (Fig. 7). Before soaking the grains in the surface layers of CSi–Mg6 and CSi–Mg10 (1150 °C) exhibited prismatic shapes with crystal boundaries. SEM observation showed some globular CaP precipitates on the ceramics at 24 h, which shapes with crystal boundaries. SEM observation showed that the Ca/P ratio was significantly decreased before and after soaking in SBF, but that was slightly lower for the CSI ceramic than that of CSI–Mg10.

As for the change of ion concentration in the SBF (Fig. 8), it is seen that the change of Mg concentration in SBF was of the similar pattern as in Tris buffer. The release rate of Ca and Si increased with Mg content elevated for CSi–Mg6 ceramics sintered at 1150 °C. The Ca and Si concentrations soared in SBF within 12 h to 130–220 and ~60 ppm, respectively, and then gradually increased with different rate. However, the CSi released appreciable amount of Ca and Si within 24 h. The concentration of Ca then meets a fluctuation at a level above 200 ppm within 96 h. The P concentration in SBF decreased from 31 ppm before soaking, to 7–16 ppm for the CSi–Mg6 (1150 °C) after soaking for 96 h.

3.6 Mechanical decay after immersion in SBF

To order to investigate the potential influence of degradation on mechanical decay of CSi–Mg6 ceramics, the main mechanical parameters were also evaluated after soaking in SBF for 3 weeks (Fig. 9). Compared with those before immersion in SBF (Fig. 3), the compressive strength of the CSi–Mg6 and CSi–Mg10 (1100, 1150 °C) showed a significant reduction, but still kept over 450 MPa. In contrast, the CSi and CSi–Mg3 (1100 °C) displayed different degree of increase in flexural strength, which is possibly attributed to the bonding effect of the deposited CHA in the open pores of the ceramics. Additionally, the elastic modulus and Young’s modulus of four CSi–Mg6 groups sintered at different temperatures showed a mild decrease after 3 weeks of immersion.

4 Discussion

Doping is a widely applied technological process in biomaterials science that involves incorporating foreign atoms or ions of appropriate elements into host lattices to yield hybrid materials with desirable properties and functions (Yang et al., 2012). For inorganic bio ceramics, doping is of fundamental importance in stabilizing a specific crystallographic phase, modifying mechanical properties, modulating sintering behavior as well as tuning biological performances. In this study, we go beyond the traditional approaches that use phase conversion or biphasic hybrid (Carrodeguas et al., 2007; de la Casa-Lillo et al., 2011; Ma et al., 2012; Sainz et al., 2010; Sprio et al., 2009), i.e., which rely on either enhancing or degrading mechanical properties or bioactivity. Instead, by using a dilute doping of Mg in a single-phase CSI, we described a Ca-silicate system in which Mg dilute doping influenced the sintering process to give simultaneous control over the mechanical properties, biodegradation, and microstructures of the single-phase CSI. It is shown that the CSI ceramic could be rationally tuned in phase purity control (< 2.12% Mg) and mechanical parameters through the usage of Mg dopant introduced at precisely defined concentrations.

Our previous studies have shown that the fracture toughness and elastic modulus of pressureless sintered γ-dicalcium silicate were about 1.48–1.80 MPa m 1/2 and 34–40 GPa (Gou et al., 2005). In this study, it is seen that the Mg dilute doping approach can significantly reinforce the single-phase
Fig. 7 – SEM images of the ceramics sintered at 1150 °C before and after soaking in SBF for 96 h. Bar: 5 μm.

Fig. 8 – Changes in ion concentration of the SBF during immersion of the CSi and CSi-Mgₓ ceramics sintered at 1150 °C.
The CSi-Mg6 and CSi-Mg10 ceramics show extremely high fracture toughness (3.2–3.6 MPa m$^{1/2}$; Fig. 4), superior to the lower limit reported for cortical bone (2–12 MPa m$^{1/2}$) (Nalla et al., 2005), and this value is over twice that of other stoichiometric Ca–Mg–silicate ceramics (1.5–1.8 MPa m$^{1/2}$) (Chen et al., 2010; Diba et al., 2014), calcium phosphates (CaPs; 0.8–1.5 MPa m$^{1/2}$) (Ryu et al., 2004), and pure CSi ($\leq 1.0$ MPa m$^{1/2}$) when sintering them at conventional temperature conditions. Meanwhile, the flexural strength (90–146 MPa), elastic modulus (9–13 GPa), and Young’s modulus (18–23 GPa) of the CSi–Mg$^x$ are all significantly optimized (Fig. 3), which better match with the bone tissue (flexible strength: 50–150 MPa; elastic modulus: 7–30 GPa) compared with other Ca–Mg-silicate ceramics. The Mg dilute doping CSi, furthermore, has certain inherent advantages over pure CSi and CaP ceramics, viz. sinterability (Ryu et al., 2004; Wu and Chang, 2013). Thus, this study provides a reliable strategy that the dilute doping of Mg ions is beneficial in improving mechanical properties with respect to the as-known phase conversion and biphasic hybrid routes (Carrodeguas et al., 2007; de la Casa-Lillo et al., 2011; Ma et al., 2012; Sainz et al., 2010; Sprio et al., 2009).

It has been proven that traces of Mg ion can reduce the overall rate of seeded CaP crystallization and markedly delay the transformation of amorphous CaP to more stable apatite phase (Tao et al., 2009). This effect is also in agreement with the CaO–MgO–SiO$_2$ system, in which the crystallization shifts to higher temperature with increasing MgO content (Diba et al., 2012). As we know, $\beta$-CSI and $\alpha$-CSI is respectively low and high temperature form of wollastonite. The polymorphic transition of $\beta$- to $\alpha$-CSI with high purity usually takes place at 1130±5°C. Meanwhile, it is known that Ca$^{2+}$ and Mg$^{2+}$, whose ion radius is respectively 0.99 Å and 0.65 Å, possess the similar coordinator polyhedron in crystal spatial structure. When entering into CSI structure, Mg ion will replace the position of Ca ion and form a substitutional solid solution. It is reported that, the presence of Mg in solid solution in $\beta$-CSI causes no obvious change in $\beta$-CSI structure (<5% of declination of a-, c-axis, and elongation of b-axis) (Ma et al., 2012), though this solid solution shifts to much high transition temperature (1370±20°C) (Sainz et al., 2010). Thus, when taking into account the solid solution of dilutely doped Mg in $\beta$-CSI (Ca$_{x}$Mg$^x$SiO$_3$; 0 $\leq x \leq 0.14$), it is reasonable to understand the retardation of phase transition of CSI–Mgx when sintering at 1150–1250°C (Fig. 1).

According to previous studies, with the decrease of MgO content, the mechanical strengths of stoichoimetric montellite, akermanite, and merwinitic ceramics decreased (Chen et al., 2010). It is indicated that Mg plays an important role in affecting the mechanical properties of bioceramics in the Ca–Mg–silicate system, and the mechanical properties of bio-ceramics in this system may be controlled by adjusting the Mg content. On the other hand, additions of MgO, alone and with other metal oxides in various combinations, have been used to promote the densification of structural ceramics. The presence of these additives in ceramic powder compacts is believed to provide a liquid phase at sintering temperatures by reaction with other components from the ceramic grain surfaces (Yoo et al., 1993). Also, MgO normally is added as a host grain growth inhibitor (Zhao et al., 2014). Presence of MgO causes obvious signs of discontinuous grain growth which accounts for the decrease of residual porosity and the relatively high strength. In this study, our investigations further confirm the dilute doping of Mg (e.g. 6% or 10% Ca...
substitution by Mg is much favorable for improving the mechanical properties of β-CSi ceramic at an appropriate sintering temperature range. It is known that the bond energy of Mg-O is higher than that of Ca-O according to their differences in ion radius (Diba et al., 2012). In view of this, it can be seen from the present results that when appropriate amount of Mg atoms occupy the position of Ca atoms in the crystal lattice in CSi, the crystal structure of CSi becomes more stable, and this may also benefit for the mechanical strength of β-CSi ceramic when sintering at appropriate temperatures.

On the other hand, because grain growth and phase conversion of CSi are very sensitive to sintering temperature, sintering of β-CSi products usually occurs at temperatures below 1200 °C (Carrodeguas et al., 2007; de la Casa-Lillo et al., 2011). Abnormal grain growth in β-CSi would occur in the final stage of densification if sintering temperature is too high, which has a great influence on flexural strength and fracture toughness with increasing grain size. For pure β-CSi ceramic, it has poor sinterability and is difficult to obtain dense sintering body, which leads to very poor mechanical parameters and to tune the biodegradation rate (Hench, 1998). Hence, the slight decrease in solubility of CSi-Mgx ceramics with increasing Mg content affects the apatite mineralization rate. Conversely, this is substantially favorable for overcoming the relatively fast dissolution rate of pure CSi ceramic that could compromise its mechanical strength in vivo (Xu et al., 2008). On the other hand, the good apatite induction ability of such CSi-Mgx ceramics would be favorable for retarding the mechanical decay in physiological environment. We did observe some degree of decrease in compressive strength for the highly dense CSi-Mg6 and CSi-Mg10, but we also found an increase in flexural strength for the low-densification CSi ceramic after immersion in SBF. It is reasonable to assume that, accordingly, the biomimetic apatite deposition on the pore walls of porous formulation of CSi-Mgx ceramics would benefit their mechanical reinforcement.

As mentioned above, Mg–O bond energy is higher than that of Ca–O bond, and the higher Mg–O bond energy results in difficulty in Ca, Mg and Si ion release from crystal lattice. Thus, CSi-Mg10 should be less reactive in aqueous solution and the ion leaching rate should be lower (Fig. 6). Hence, the degradation rate decreased with the increase of the stronger Mg–O bonds from pure CSi to CSi-Mg10 ceramics, resulting in the slower Ca ion release (Fig. 8) and decrease of apatite-formation ability of ceramics. This is in agreement with the experimental results of weight loss of the CSi-Mgx when immersing in Tris buffer.

According to the above analysis, it seems that the degrada- tion of CSi ceramics correlates to Mg doping ratio and sintering temperature. It is known that the higher is the sintering temperature, the higher is the densification of the CSi-Mgx ceramics and the slower is the dissolution rate in Tris buffer. Thus, it is comprehensible that the role of Mg doping in altering the degradability of CSi ceramic is mainly achieved by the enhanced densification and stability of the ceramics. Indeed, innovative two-step sintered Ca–Mg-sili cate system has a slightly higher densification and grain growth resistance (Nadernezad et al., 2014), we believe that our Mg dilute doping approach is very promising since optimization of the sintering schedule is under investigation and promises to yield more intriguing properties.

5. Conclusion

In summary, we have found that a new strategy of Mg dilute doping is an effective way to improve the several important mechanical parameters and to tune the biodegradation rate...
in vitro of β-CSi ceramic. With a simple pressureless sintering process, the CSI-Mg6 and CSI-Mg10 ceramics exhibited significant enhancement in mechanical strength and fracture toughness, showed depressed degradation rate, and maintained good bioactivity. In this regard, the β-Csi with appropriate Mg dilute doping ceramic is promising for different bone defect repair areas, especially from the perspective of segmental bone and head bone tissue repair applications.

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