Pr³⁺ doped new TZYN glasses and glass-ceramics containing NaYF₄ nanocrystals: Luminescence analysis for visible and NIR applications

Dagupati Rajesh^{a,b,*}

^a Centre for Functional and Surface Functionalized Glass, Alexander Dubček University of Trenčín, Študentská 2, 911 50, Trenčín, Slovakia

^b São Carlos Institute of Physics, University of São Paulo, PO Box 369, 13560-970, São Carlos, SP, Brazil

ARTICLE INFO

Keywords:

Glass
Glass-ceramics
NaYF₄ nano-crystals

ABSTRACT

In this work, I have been prepared a series of different concentration of Pr³⁺ doped TeO₂-ZnO-YF₃-NaF (TZYN) new oxyfluoro tellurite glasses by conventional melt-quenching technique. The oxyfluoro glass-ceramics were obtained by controlled heat treatment of 0.2 mol% Pr³⁺ doped glass at 390 °C, equal to the crystallization temperature (T_{c1}) obtained from DSC analysis, for different time intervals 5, 10 and 15h. The precipitation of α-NaYF₄ phase was confirmed by comparison of measured X-ray diffraction patterns with α-NaYF₄ JCPDS patterns. In glass ceramics, the intensity of the transitions, ³P₀→³F₂ in VIS, and ¹D₂→³F₄ & ¹D₂→¹G₄ in NIR regions was enhanced by 24.03, 9.4 and 4.54 times, respectively. Moreover, with heat treatment time, an experimental lifetime of the ³P₀ excited state of Pr³⁺ ion was also enhanced from 24 μs in glasses to 62 μs in glass-ceramics due to by the enhancement of crystallite size of precipitation of NaYF₄ nanocrystals.

1. Introduction

The synthesis of nano-scaled fluoride-crystals in the glass-ceramics has attracted a great deal of interest owing to rich potential applications invisible, near infrared and mid-infrared luminescence applications because it has been the combines the merits of high chemical and mechanical stability of oxide and low phonon energy of fluoride [1–4]. Since glass-ceramic has been obtained usually either by a one-step heat treatment (nucleation and growth at the same time) or via a two-step heat treatment (nucleation followed by the growth of nuclei), leading to different microstructures. The main advantage of glass–ceramic is that which could offer over either glass or ceramic material with outstanding mechanical properties. Based on the transmittance property, glass-ceramics could be classified broadly into two, (i) transparent and (ii) opaque [5]. Indeed, several authors have been reported rare earth doped transparent glass–ceramics having the micro/nano-crystals components of LaF₃, BaF₂, PbF₂, BaYF₅, NaYF₄, etc. [6–10]. It is reported that in fluoride nanocrystals, the dopant concentration enhances by an order or more than that in the precursor glass, results in the reduction of effective dopant separation [11].

As compared to various kinds of oxide glasses, tellurite glasses are the highly suitable host for lasing ions because these glasses present high refractive index, low phonon energy (≈800 cm^{−1}), wide transparency region (0.3–5 μm), and low melting temperature [12]. When

fluorine compounds are added into the tellurite glasses, they can offer important advantages over the pure fluoride or pure oxide compositions [13,14]. Moreover, the addition of Yb³⁺ can reduce the rare earth ions clustering in bulk glasses and the average distance between the rare earth ions is increased. As a consequence, non-radiative losses owing to cross relaxation become less probable, increasing photoluminescence (PL) intensity [15].

Recently, due to the promising luminescent properties of AREF₄ (A, alkali metal, and RE, rare earth) phase, different oxyfluoride transparent glass-ceramics having this phase have been studied extensively. For instance, de Pablos-Martín et al. [16] were synthesized NaLaF₄ crystal phase in a Tm³⁺ doped K₂O–Na₂O–Al₂O₃–SiO₂–LaF₃ glass by heat treatments slightly above the glass transition temperature, T_g. The corresponding glass-ceramic shows a more intense blue up-conversion emission with respect to the parent glass. Herrmann et al. [17] studied crystallization of NaGdF₄ in an oxyfluoride glass, in which the doping Sm³⁺ ions are partially incorporated in this phase and reported that NaYF₄ nanocrystals phase containing glass-ceramics are also considering one of the most efficient hosts for up-conversion application [18–20], biology molecular labeling [21] and solar cell [22]. Usually, NaYF₄ possess two polymorphic forms, namely, α-NaYF₄ (cubic) and β-NaYF₄ (hexagonal) [23]; the efficiency of the up-conversion luminescence in hexagonal NaYF₄ is at least one order of magnitude higher than in its cubic counterpart [24]. Indeed, due to the phase transition, the

* Centre for Functional and Surface Functionalized Glass, Alexander Dubček University of Trenčín, Študentská 2, 911 50, Trenčín, Slovakia.

E-mail addresses: drssvu@gmail.com, rajesh.dagupati@tnuni.sk.

NaYF₄ bulk crystal is difficult to grow [25], which limits its application as bulk material in some fields such as luminescence device and a solid-state laser. Therefore, development of NaYF₄ nanocrystals in transparent glasses is of great scientific and technical significance. Several authors were succeeded for precipitation of NaYF₄ nanocrystals by heat treatment of silica-based glass matrices [16,20,26–28] but no reports, to the best of our knowledge, on the present oxyfluoro tellurite glasses.

In the present paper, to the best of my knowledge, reporting first time the results on the preparation of different concentrations of Pr³⁺ doped new TeO₂-ZnO-YF₃-NaF glasses and glass-ceramics containing a cubic phase of NaYF₄ nanocrystals by melt quenching and followed by heat treatment method. Stokes luminescence spectra and the fluorescence lifetimes of the steady-state excited states were measured in both glasses and glass-ceramics. The effect of rare earth concentration and NaYF₄ nanocrystals precipitation on stokes luminescence of glasses and glass-ceramics, respectively, were analyzed.

2. Experimental

The glass samples with composition (mol%) of (50-x) TeO₂-30ZnO-10YF₃-10NaF-xPr₂O₃ (where x = 0.2, 0.5, 1.0, and 1.5 mol%) were prepared by conventional melt-quenching method. For our convenience hereafter labeled as G-Pr02, G-Pr05, G-Pr10, and G-Pr15, for 0.2, 0.5, 1.0, and 1.5 mol% doped glasses, respectively. TeO₂ was obtained by heating Te metal in a furnace at 100 °C lower than the melting temperature. For each batch, about 10 g of TeO₂, ZnO, YF₃, NaF and Pr₂O₃ with 99.99% purity of raw materials were mixing homogeneously in agate mortar and melted in a platinum crucible at 1050 °C for 45 min in air atmosphere, and then quenched by pouring the melts into a pre-heated (200 °C) brass mould. The glasses were annealed at 250 °C for 4 h to relinquish the internal thermal stress and then cooled to room temperature naturally. The obtained samples were polished to optical quality. To obtain glass ceramics, G-Pr02 transparent glass was heat-treated at crystallization temperature 390 °C with a rate of 10 °C/min, for 5 h, 10 h, and 15 h labeled as hereafter GC-5h, GC-10h, and GC-15h respectively. The glasses and glass-ceramics with chemical composition and their labels are tabulated in the following Table 1.

For the glass and glass-ceramics, the X-ray diffraction (XRD) profiles were measured in an X-ray diffractometer (ULTIMA IV, Rigaku) with Cu K α radiation (1.54 Å) in the range 10–80° with 0.5s per step. The amorphous nature of the glass sample was confirmed by the lack of sharp crystalline peaks in profile not shown here. But, since, the broad bump has been observed at shorter angle side due to short range order of arrangement of an atom in glass at an atomic scale level. To investigate the size (micro/nano) of the NYF₄ particles in the glass ceramics, a transmission electron microscope (TEM) (TECNAI G² F20) was used operating at an accelerating voltage of 200 kV. For TEM measurement, the specimens were prepared by dispersing the powder

Table 1

Glasses, glass-ceramic composition and their labels.

Glass Name	Glass composition (mol%)				
	TeO ₂	ZnO	YF ₃	NaF	Pr ₂ O ₃
G-Pr02	49.8	30	10	10	0.2
G-Pr05	49.5	30	10	10	0.5
G-Pr10	49	30	10	10	1.0
G-Pr15	48.5	30	10	10	1.5

Glass-ceramic name	Glass-ceramic composition (mol%)					Heat treatment time
	TeO ₂	ZnO	YF ₃	NaF	Pr ₂ O ₃	
GC-5h	49.8	30	10	10	0.2	5h
GC-10h	49.8	30	10	10	0.2	10h
GC-15h	49.8	30	10	10	0.2	15h

Table 2

Physical properties of G-Pr10 glass.

S. No	Physical parameter	G-Pr10 glass
1	Refractive index n (at 543 nm)	1.914
2	Density (d) (g/cm ³)	4.21
3	Thickness (t) (cm)	0.194
4	Average molecular weight (M) (g)	124.71
5	Molar volume (V _m) (cm ³)	29.62
6	Ion concentration (N) (x10 ²⁰ ions/cm ³)	2.03
7	Dielectric constant (ϵ)	3.66
8	Reflection losses (R) (%) (at 543 nm)	9.83

Fig. 1. DSC profile of G-Pr02 and G-Pr15 glasses.

Fig. 2. XRD patterns of glass ceramics and also standard cubic NaYF₄ phase and ZnTeO₃ phase.

sample in acetone using an ultrasonic bath. Then after the solution was placed onto a copper grid and then allowed to dry before it was used for characterization. For investigated glasses and glass-ceramics, visible and infrared excitation, emission, and excited-state lifetime measurements were carried out using a Horiba Fluorolog spectrofluorimeter, equipped with CW and pulsed Xe lamps source. The visible and NIR

Fig. 3. FE-SEM image of the GC-10h glass ceramic.

signals were collected by a visible photodiode detector model PPD-850 and by an infrared Hamamatsu photomultiplier, respectively.

3. Results and discussion

3.1. Physical parameters

Different physical properties such as refractive index (n) at 543 nm, density (d), thickness (t), average molecular weight (M), molar volume (V_m), ion concentration (N), dielectric constant (ϵ), reflection loss (R) were calculated for all investigated glasses doped with various Pr^{3+} concentrations. There are no significant changes in the physical parameters so I have presented in Table 2 for G-Pr10 glass.

3.2. DSC analysis

The characteristic temperatures of the G-Pr02 and G-Pr15 glasses were obtained from differential scanning calorimetry (DSC) thermograph, which is shown in Fig. 1. For the G-Pr02 glass, the glass transition temperature (T_g) is obtained by taking the minimum inflection point of the endotherm, located at 330 °C, and three crystallization temperatures, T_{C1} , T_{C2} & T_{C3} , are the maximum of the three exothermic

peaks, which are 392 °C, 426 °C & 519 °C, respectively. Whereas, for the G-Pr15 glass, glass transition temperature, T_g and crystallization temperatures, T_{C2} & T_{C3} located at 358 °C & 480 °C and 520 °C, respectively. However, T_{C1} is disappeared in the G-Pr15 glass. The first crystallization peak T_{C1} maybe belonging to a NaYF_4 crystalline phase which conforms further from XRD profile and remaining crystallization peaks T_{C2} & T_{C3} are related to the combination of ZnO and TeO_2 phases.

3.3. XRD, FE-SEM and TEM analysis

The as-synthesized glasses are completely amorphous with no diffraction peaks not shown here. After heat-treatment at 390 °C, intense diffraction peaks are observed and are attributed to cubic NaYF_4 nanocrystals, JCPDS: 06-0342, shown in Fig. 2 [29]. According to Ostwald's step rule the least stable polymorph, $\alpha\text{-NaYF}_4$, is formed first in the glassy system far from thermodynamic equilibrium. This phenomenon is observed in several oxyfluoride glass ceramics along with preset glass system [26–28]. Moreover, impurity diffraction peaks are also observed for GC-15 h along with $\alpha\text{-NaYF}_4$ diffraction peaks owing to ZnTeO_3 (COD: 1528556) [30]. The mean crystallite size is calculated by using following Scherrer formula based on the major intensity peak widths is about 20 nm.

$$d = \frac{0.9\lambda}{\beta \cos \theta} \quad (1)$$

where d is the mean size of the crystalline, λ is the X-ray wavelength, θ is the Bragg angle, and β is the line broadening at half the maximum intensity.

The field emission scanning electron microscope (FE-SEM) image of the GC-10h glass-ceramic, shown in Fig. 3, demonstrated that NaYF_4 particles are spherical in nature and average size around 18 nm distributed homogeneously across the glass matrix.

The TEM micrograph and corresponding selected area electron diffraction (SAED) patterns of the GC-10h sample displays in Fig. 4a and b, respectively. The TEM allows the direct imaging of nanoparticles and provides some information on the quality of individual particles including their sizes and distributions. As it can be seen, the dark and nearly spherical particles correspond to NaYF_4 nanocrystals, as identified by the XRD results, are observed, which are random and homogeneously dispersed in the glass host. The average size of the NaYF_4 nanocrystal is found to be around 20 nm, which is consistent with the value calculated from SEM and XRD. In addition, a high-resolution

Fig. 4. a) TEM micrograph b) SAED patterns and c) HRTEM micrograph of the GC-10h glass ceramic.

Fig. 5. PLE spectra for investigated glasses with an emission wavelength a) 950 nm and b) 1050 nm, respectively.

Fig. 6. PL spectra for as prepared glasses in a) Visible and b) NIR region with Pr^{3+} ions of various concentration.

Fig. 7. Energy level diagram of Pr^{3+} ion in TZYn glass.

transmission electron microscope (HR-TEM) image shown in Fig. 4c reveals that high crystalline nature and the resolved lattice fringes of the NaYF_4 nanocrystals.

3.4. Luminescence analysis in glasses and glass ceramics

The photoluminescence excitation spectra (PLE) were measured for all studied glasses upon two different emission wavelengths 950 and 1050 nm, corresponds to $^3\text{P}_0 \rightarrow ^1\text{G}_4$ and $^1\text{D}_2 \rightarrow ^3\text{F}_4$ transitions, are depicted in Fig. 5a and b, respectively. As it can be seen, intensity variation of excitation transitions with Pr^{3+} concentration depends on the measured emission wavelengths. Based on the intensity variation of excitation transitions associated with $^3\text{P}_J$ ($J = 0, 1$ and 2) manifolds and $^1\text{D}_2$ state reveals that emission transition attributed from $^3\text{P}_J$ ($J = 0, 1$ and 2) manifolds having a higher intensity in the G-Pr05 glass, whereas emission transitions originate from the $^1\text{D}_2$ state having a higher intensity in the G-Pr02 glass. This would also confirm further from the emission spectra shown below.

The Vis and NIR photoluminescence (PL) spectra of Pr^{3+} doped TZYn glasses are shown in Fig. 6a and b, respectively. Visible spectra, consist of nine characteristic bands located around at 484, 527, 540, 594, 612, 643, 684, 707 and 728 nm, which are assigned to the electronic transitions $^3\text{P}_0 \rightarrow ^3\text{H}_4$, $^3\text{P}_1 \rightarrow ^3\text{H}_5$, $^3\text{P}_0 \rightarrow ^3\text{H}_5$, $^3\text{P}_1 \rightarrow ^3\text{H}_6$, $^3\text{P}_0 \rightarrow ^3\text{H}_6$, $^3\text{P}_0 \rightarrow ^3\text{F}_2$, $^3\text{P}_1 \rightarrow ^3\text{F}_3$, $^3\text{P}_1 \rightarrow ^3\text{F}_4$ and $^3\text{P}_0 \rightarrow ^3\text{F}_4$, respectively. NIR spectra consist of three emission transitions located around at 945, 1041 and

Fig. 8. PLE spectra for as prepared glass-ceramics with an emission wavelength a) 950 nm and b) 1475 nm, respectively.

Fig. 9. PL spectra for as prepared glass ceramics in a) VIS b) & c) NIR regions upon excitation wavelength 440 nm.

1487 nm assigned to the $^3P_0 \rightarrow ^1G_4$, $^1D_2 \rightarrow ^3F_4$ and $^1D_2 \rightarrow ^1G_4$, respectively. In the visible PL spectra of all glasses, red emission at ~ 643 nm is dominated and intensity of all bands intensified with concentration and higher in the G-Pr05 glass. Then after intensity is diminished with

concentration. However, in NIR spectra, the intensity of emission transitions originates from 1D_2 state is diminished with concentration, as I expected from excitation spectra. The quenching in the intensity of 3P_J manifolds with concentration is due to energy transfer from Pr^{3+} ions

Fig. 10. Decay profiles for the 3P_0 excited state of Pr^{3+} ions in a) glasses b) glass ceramics.

so-called concentration quenching. Moreover, intensity diminish of emission transitions originate from 1D_2 state is owing to cross relaxation energy transfer through the channels, namely $^1D_2 + ^3H_4 \rightarrow ^1G_4 + ^3F_4$ and $^1D_2 + ^3H_4 \rightarrow ^3F_4 + ^1G_4$ [31]. Thus, a full-width half maximum of transition, $^1D_2 \rightarrow ^1G_4$ is enhanced with concentration, as shown in inset Fig. 6b. Fig. 7. illustrates the energy level diagram includes both excitation and emission transitions of Pr^{3+} ions. Fig. 8a and b depict the PLE spectra for G-Pr02 glass and glass-ceramics, obtained after heat treatment, were measured upon an emission wavelength 950 and 1475 nm, respectively. As it can be seen, excitation bands intensity is intensified with heat treatment time compared to the precursor G-Pr02 glass.

The photoluminescence emission spectra in VIS and NIR regions were measured for glass ceramics upon excitation wavelength 440 nm and are shown in Fig. 9a, b, and 9c, respectively. It is noted that, compared with precursor G-Pr02 glass, the intensity of emission transitions originates from 3P_J ($J = 0, 1$ and 2) manifolds and 1D_2 state both in VIS and NIR spectra is enhanced by heat treatment time, as we expected from excitation spectra (see Fig. 8). Among all the transitions, $^3P_0 \rightarrow ^3F_2$, $^1D_2 \rightarrow ^3F_4$ and $^1D_2 \rightarrow ^1G_4$ transitions intensity are intensified by 24.03, 9.4 and 4.54 times, respectively. This is due to, perhaps, precipitation of $NaYF_4$ crystalline phase, which causes changing of crystal field around Pr^{3+} ion. Pablos-Martín [32], and Zhao [33] reported that, in Er^{3+} and Eu^{3+} doped silicate glasses, during the crystallization process dopant ions are incorporated into the $NaYF_4$ nanocrystals and sharper stark split in emission transitions leads efficient emission intensities and longer lifetimes of excited states due to minimization of the non-radiative losses and multiphonon relaxations. However, in the present case, sharper stark splitting is not being observed in emission bands. Therefore, the investigation of the local environment around the Pr^{3+} ions is necessary and is the future scope of work.

Fig. 10a and b demonstrate the time-dependent luminescence decay of 3P_0 excited state for all glasses and glass-ceramics, respectively. Form Fig. 10a, for the G-Pr02, G-Pr05, and G-Pr10 glasses, decay profile is single exponential in nature and is non-exponential for the G-Pr15 glass. The decay lifetime from decay curves was obtained, since, by fitting the suitable exponential function [29]. The experimental lifetime of the 3P_0 excited state is 24, 13, 9 and 6 μs for the G-Pr02, G-Pr05, G-Pr10, and G-Pr15 glasses, respectively. Whereas in glass-ceramics, in Fig. 10 b, the decay lifetime of the 3P_0 excited state is enhanced from 24 μs to 62 μs with heat treatment time.

4. Conclusions

Pr^{3+} doped new oxyfluoro tellurite glasses and glass-ceramics containing cubic $NaYF_4$ nanocrystals are prepared by conventional melt quenching and subsequent heat treatment method, respectively. The average size of precipitation of nanocrystals is 18 nm obtained from FE-SEM and TEM analysis and is consistent with the grain size 20 nm obtained from XRD spectra. In the visible region, both in glasses and glass-ceramics, emission bands centered at 484, 527, 540, 594, 612, 643, 684, 707 and 728 nm and of them 643 nm emission band fall in the red region showing high intensity. In glasses, with concentration, the intensity of emission transitions originate from 3P_J manifolds is intensified and higher in the G-Pr05 glass. However, emission transitions originate from 1D_2 state, diminishes with concentration could be attributed to cross relaxation energy transfer through the channels, namely $^1D_2 + ^3H_4 \rightarrow ^1G_4 + ^3F_4$ and $^1D_2 + ^3H_4 \rightarrow ^3F_4 + ^1G_4$. In glass-ceramics, the intensity of all emission transitions is enhanced with heat treatment time due to the precipitations of $NaYF_4$ nanocrystals, which causes changing the crystal field around the Pr^{3+} ions. Among all transitions, the intensity of transitions $^3P_0 \rightarrow ^3F_2$, in VIS, and $^1D_2 \rightarrow ^3F_4$ & $^1D_2 \rightarrow ^1G_4$ in NIR regions is enhanced by 24.03, 9.4 and 4.54 times, respectively. Overall results suggesting that the G-Pr05 glass and the GC-Pr10 glass-ceramic, in terms of transparent ceramic, are the potential candidates for VIS and NIR applications.

Appendix A. Supplementary data

Supplementary data to this article can be found online at <https://doi.org/10.1016/j.optmat.2018.10.004>.

References

- [1] Y. Wang, J. Ohwaki, Appl. Phys. Lett. 63 (1993) 3268.
- [2] P.A. Tick, N.F. Borrelli, L.K. Cornelius, M.A. Newhouse, J. Appl. Phys. 78 (1995) 93.
- [3] M. Abril, J. Méndez-Ramos, I.R. Martín, U.R. Rodríguez-Mendoza, V. Lavín, A. Delgado-Torres, V.D. Rodríguez, J. Appl. Phys. 95 (2004) 5271.
- [4] G. Dantelle, M. Mortier, D. Vivien, G. Patriarche, Chem. Mater. 17 (2005) 2216.
- [5] A. Edgar, G.V.M. Williams, M. Secu, S. Schweizer, J.M. Spaeth, Radiat. Meas. 38 (2004) 413.
- [6] A. de Pablos-Martín, N. Hémono, G.C. Mather, S. Bhattacharyya, T. Höche, H. Bornhöft, J. Deubener, F. Muñoz, A. Duran, M.J. Pascual, J. Am. Ceram. Soc. 94 (2011) 2420.
- [7] K. Biswas, A.D. Sontakke, J. Ghosh, K. Annapurna, J. Am. Ceram. Soc. 93 (2010) 1010.
- [8] F. Lahoz, I.R. Martín, J. Méndez-Ramos, P. Núñez, J. Chem. Phys. 120 (2004) 6180.
- [9] F. Liu, Y. Wang, D. Chen, Y. Yu, E. Ma, L. Zhou, P. Huang, Mater. Lett. 61 (2007) 5022.
- [10] F. Liu, E. Ma, D. Chen, Y. Yu, Y. Wang, J. Phys. Chem. B 110 (2006) 20843.
- [11] H. Zheng, X.-i. Wang, S.-X. Qu, M.J. Dejneka, R.S. Meltzer, J. Lumin. 119–120

- (2006) 153.
- [12] A.N. Moiseev, V.V. Dorofeev, A.V. Chilyasov, et al., *Opt. Mater.* 33 (2011) 1858.
- [13] D. Chen, Y. Wang, Y. Yu, P. Huang, F. Wen, *Op. Lett.* 33 (2008) 1884.
- [14] J.J. Velázquez, A.C. Yanes, J. del-Castillo, J. Méndez-Ramos, V.D. Rodríguez, *J. Non-Crys. Solids* 356 (2010) 1349.
- [15] T. Kalpana, M.G. Brik, V. Sudarsan, P. Naresh, V. Ravi Kumar, I.V. Kityk, N. Veeraiah, *J. Non-Cryst. Solids* 419 (2015) 75.
- [16] A. de Pablos-Martín, M. O Ramírez, A. Durán, L.E. Bausá, M.J. Pascual, *Opt. Mater.* 33 (2010) 180.
- [17] A. Herrmann, M. Tylkowski, C. Bocker, C. Rüssel, *Chem. Mater.* 25 (2013) 2878.
- [18] A.C. Yanes, A. Santana-Alonso, J. Méndez-Ramos, J. del-Castillo, V.D. Rodríguez, *J. Alloy. Comp.* 480 (2009) 706.
- [19] X. Shang, P. Chen, W. Cheng, K. Zhou, J. Ma, D. Feng, S. Zhang, Z. Sun, J. Qiu, T. Jia, *J. Appl. Phys.* 116 (2014) 101.
- [20] A. Sarakovskisa, G. Krieke, *J. Eur. Ceram. Soc.* 35 (2015) 3665.
- [21] G. Yi, H. Lu, S. Zhao, *Nano Lett.* 4 (2004) 2191.
- [22] A. Shalav, B.S. Richards, T. Trupke, *Appl. Phys. Lett.* 86 (2005) 013505.
- [23] R.E. Thoma, G.M. Hebert, H. Insley, C.F. Weaver, *Inorg. Chem.* 2 (1963) 1005.
- [24] Z. Gu, L. Yan, G. Tian, S. Li, Z. Chai, Y. Zhao, *Adv. Mater.* 25 (2013) 3758.
- [25] H. Page Ralph, K.I. Schaffers, P.A. Waide, J.B. Tassano, *J. Opt. Soc. Am. B* 15 (1998) 996.
- [26] G. Dominiak-Dzik, R. Lisiecki, W. Ryba-Romanowski, L. Krajczyk, *J. Alloy. Comp.* 511 (2012) 189.
- [27] S. Zhao, S. Xu, G. Jia, D. Deng, L. Huang, H. Wang, *Mater. Lett.* 65 (2011) 2407.
- [28] S. Jiang, P. Zeng, L. Liao, S. Tian, H. Guo, Y. Chen, C. Duan, M. Yin, *J. Alloy. Comp.* 617 (2014) 538.
- [29] Z. Hund, *Anorg Chem* 261 (1950) 109.
- [30] Lin-Gun Liu, *J. Phys. Chem. Solid.* 48 (1987) 719.
- [31] M.V.V. Kumar, K.R. Gopal, R.R. Reddy, G.V.L. Reddy, N.S. Hussain, B.C. Jamalaiah, *J. Non-Cryst. Solids* 364 (2013) 20.
- [32] A. de Pablos-Martín, J. Méndez-Ramos, J. del-Castillo, A. Durán, V.D. Rodríguez, M.J. Pascual, *J. Eur. Ceram. Soc.* 35 (2015) 1831–1840.
- [33] Shilong Zhao, Xiao Sun, Xiuli Wang, Lihui Huang, Yong Fei, J. Shiqing Xua, *Europ. Ceram. Soc.* 35 (2015) 4225.