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Optical and structural properties of Mn-doped magnesium titanates fabricated with excess MgO

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Optical and structural properties of ceramics based on Mn-doped magnesium titanates synthesized by sintering in air at 1200 °C of MgO and TiO₂ powders of different molar ratios ranging from MgTiO₃ to Mg₂TiO₄ stoichiometric compositions were studied. The influence of excess MgO on Mn incorporation in crystal lattice of MgTiO₃ was also investigated. The Mn⁴⁺ ions substituted Ti⁴⁺ sites were controlled by the photoluminescence (PL) and diffuse reflectance spectroscopy, and the Mn²⁺ ions on Mg²⁺ sites were monitored by electron paramagnetic resonance (EPR). The ceramics produced using equimolar ratio of MgO and TiO₂ composed of a major MgTiO₃ and a minor MgTi₂O₅ crystal phases, and those made with excess MgO contained MgTiO₃ and Mg₂TiO₄ phases in different proportions. The EPR study showed that Mn incorporated in MgTiO₃ synthesized under 1:1 M ratio as Mn²⁺ ion mainly. This agreed with low intensity of Mn⁴⁺ red PL ascribed to low concentration of Mn⁴⁺ centers and partial absorption of the UV excitation light by the MgTi₂O₅ phase. The Mn-doped MgTiO₃ synthesized with excess MgO of 19 and 50 mol.% showed increased Mn⁴⁺ red PL by a factor 30–50, enhanced Mn⁴⁺ optical absorption and more than ten times decreased Mn²⁺ EPR signal. The Mg₂TiO₄ phase was found to be under compressive strains attributed to the presence of Mg vacancies and demonstrated Mn⁴⁺ red PL with modified spectrum shape and decay behavior. It is concluded that in MgTiO₃ the excess MgO facilitates the incorporation of Mn onto Ti⁴⁺ site and can be used for the increasing of Mn⁴⁺ PL intensity.

1. Introduction

For decades the interest to magnesium metatitanate, MgTiO₃, was motivated by its application in low loss microwave dielectric resonators as a material with high dielectric constant ($\epsilon_r \sim 17$) and high-quality factor ($Q_f \sim 160,000$ GHz) [1,2]. To improve the performance of MgTiO₃-based microwave dielectric ceramics the effect of doping with metal ions like Ni, Zn, Co, Mn and others on microstructure and dielectric properties of MgTiO₃ has been studied [3–5]. In particular, the doping with Mn⁴⁺ and W⁶⁺ was shown to be efficient in producing MgTiO₃ ceramics of high insulation resistance when sintering is carried out in reducing atmosphere [6]. It was found that the effect of dopants on intrinsic dielectric losses of MgTiO₃ ceramics depended on whether they form a distinct second phase or a solid solution with MgTiO₃, and in the last case, intrinsic losses were strongly affected [3].

Recently, MgTiO₃ has attracted attention as a suitable matrix for

Mn⁴⁺ ions [7–10]. Mn⁴⁺ activated titanates are known as the promising red phosphors for application in phosphor converted light emitting diodes, LEDs, which use blue LED chip for excitation [11]. In fact, titanate compounds are known for their excellent thermal stability and environmental safety, as well as the substitution of Mn⁴⁺ for Ti⁴⁺ in octahedral coordination is highly feasible due to the similar ionic radius and the same oxidation state [11]. The Mn⁴⁺ ion substituted Ti⁴⁺ cation in crystal lattice of MgTiO₃ shows deep red photoluminescence (PL) at about 701 nm under UV and blue-green excitation [7,10,12]. In the compounds of ATiO₃ (A = Sr, Ba, Ca, Mg), the energy of the emitting ²E → ⁴A₂ transition is the largest in MgTiO₃ due to the smallest fraction of covalence in the Mn⁴⁺-O²⁻ bonding [7]. Because of the thermal quenching of Mn⁴⁺ PL at room temperature the MgTiO₃:Mn⁴⁺ red phosphor does not seem to be attractive for the use in lighting. At the same time, the temperature dependent characteristics of Mn⁴⁺ red emission in MgTiO₃ enable the application of this phosphor in

luminescence temperature probes [8].

The Mn-doped MgTiO_3 can be easily synthesized via sintering in the air of equimolar mixture of MgO and TiO_2 powders [4,6,13]. However, high temperatures (1000–1300 °C) required for sintering of MgTiO_3 result in low intensity of Mn^{4+} PL [14]. This last result promoted the development of alternative approaches for low temperature synthesis of Mn^{4+} activated MgTiO_3 such as firing in the air of a magnesium-titanium complex tartrate [14] or the combination of sol-gel method with annealing in the melt of salts [7].

To enhance Mn^{4+} luminescence, it is necessary to increase the number of Mn^{4+} centers in a matrix, preventing energy migration between the ions, and to decrease concentration of Mn dopants in another charge states such as Mn^{3+} and Mn^{2+} . Above mentioned states are supposed to behave as the competitive centers of radiative [15] or nonradiative recombination [16,17]. In particular, the adding a small amount of Mg in the form of MgO [18,19], MgCO_3 [20] or MgF_2 [21] to raw oxide mixture was found to be a feasible way to increase the intensity of Mn^{4+} red luminescence in oxide matrixes. Several mechanisms were proposed for explanation of the effect. In Mn^{4+} activated $\text{CaAl}_{12}\text{O}_{19}$, the synergetic effect of flux and charge compensation by MgF_2 resulted in the increment of the fraction of ions in the +4 state was assumed [21]. Alternatively, in Mn^{4+} activated $\text{CaAl}_{12}\text{O}_{19}$ [22] and $\text{Y}_3\text{Al}_5\text{O}_{12}$ [20], the Mg^{2+} dopants were supposed to suppress adverse energy migration among neighboring Mn^{4+} activators and/or to behave as charge compensating ions, which reduce concentration of O^{2-} quenching centers. Recently [23], the Mn^{4+} -doped Mg_2TiO_4 phosphors fabricated with an excess MgO showed the increased Mn^{4+} red PL and the decreased Mn^{2+} electron paramagnetic resonance (EPR) signal. It has been concluded that an excess MgO stimulates substitution of manganese for Ti site in +4 charge state.

The effect of Mg-doping on crystal phase formation and dielectric characteristics of MgTiO_3 ceramics was also investigated [5,24]. In particular, a dramatic enhancement of Q_f value up to 262,900 GHz was demonstrated in 3 wt% MgF_2 -doped MgTiO_3 ceramic sintered at 1300 °C [5]. This effect was explained by the influence of MgF_2 on crystal phase formation (the Mg-rich compound inhibited the formation of MgTi_2O_5 and promoted the formation of Mg_2TiO_4), the production of the liquid phase, which optimizes the pore structure of ceramics and improves the sintering process, as well as the decrease of the leakage current owing to the grain boundary barrier effect. Similarly, the MgTiO_3 ceramics obtained by sintering at 1300 °C of MgO nanopowders and TiO_2 nanowires with excess MgO (in 1.23:1 M ratio) demonstrated strongly enhanced quality factor $Q_f \sim 300,000$ GHz explained by more uniformly distributed grain size and dense compact [24].

The adding of excess MgO over those necessary to obtain stoichiometric composition of MgTiO_3 is known to result in formation of Mg_2TiO_4 crystal phase side-by-side with the MgTiO_3 [10,25]. It has been shown [10] that Mn-doped magnesium titanate phosphors synthesized under different MgO to TiO_2 molar ratio in the range from 1:1 to 2:1 shows two PL bands caused by radiative transitions of Mn^{4+} ions substituted Ti^{4+} site ions in the MgTiO_3 and Mg_2TiO_4 , and relative intensity of each band depends on the content of corresponding crystal phase. However, the influence of excess MgO on Mn charge state in the MgTiO_3 was not investigated.

In this work, the effect of excess MgO on formation of crystal phases and the peculiarities of manganese incorporation (specifically its charge state) into crystal lattice of MgTiO_3 synthesized by conventional solid state reaction was investigated. We show that Mn charge state in the MgTiO_3 as well as Mn^{4+} emission spectra in the Mg_2TiO_4 critically depend on the ratio of Mg_2TiO_4 and MgTiO_3 phases in the final product. We hope that obtained results can be useful for enhancement of Mn acceptor concentration in MgTiO_3 -based ceramics, as well as Mn^{4+} PL in MgTiO_3 red phosphors.

2. Materials and methods

The Mn-doped magnesium titanate ceramics were synthesized via sintering of MgO (98.8 %) and TiO_2 (99.73 %, a mixture of rutile and anatase phases) powders taken in different molar ratios (1:1, 1.19:1, 1.5:1 and 2:1) with adding of an aqueous solution of MnSO_4 . The latter was taken in quantity necessary to provide 0.1 mol.% Mn concentration. Each powder mixture was thoroughly grinded in a ceramic mortar for 1 h. The propylene glycol used as a binder and aqueous solution of MnSO_4 were added to powder mixture and mixed until a homogeneous thick paste formed. The rectangular pellets of average size about $10 \times 10 \times 5 \text{ mm}^3$ were formed from the paste, dried at 150 °C for 3 h and sintered in the air at 1200 °C for 3 h in a muffle furnace. Sintered ceramic pellets were slowly cooled to room temperature as the furnace was naturally cooled down. The dense ceramic pellets were milled to obtain powders for further investigations.

Structural analysis was performed by X-ray diffraction (XRD) method in the θ - 2θ geometry using Philips X'Pert PRO MRD diffractometer with $\text{CuK}\alpha$ radiation. The topology and elemental analysis of sintered ceramic samples were studied using High-resolution scanning electron microscope Tescan Mira 3 LMU and Oxford Instruments energy dispersive spectrometer Oxford Instruments X-Max 80, respectively. The PL spectra were detected using a spectrometer equipped with a photomultiplier tube and excited by a 405 nm and 532 nm continuous diode lasers (50 mW) as well as by a light of Xe lamp (150 W) passed through a monochromator. The PL spectra were recorded at room and liquid nitrogen temperatures. Emission decay study was carried out at room temperature with 480 nm excitation wavelength and a 4 ns pulse duration generated by a "Continuum Horizon" OPO pumped with a Surelite frequency tripled Nd:YAG laser. The energy per pulse was approximately 20 mJ at the wavelength used. The repetition frequency was 10 Hz with 4 ns pulses. After dispersion by a monochromator ORIEL 1/4 m the luminescent transient signals were measured with a Hamamatsu 3896 PMT, sent to a digital oscilloscope, and averaged out to improve the signal-to-noise ratio. The diffuse reflectance spectra were recorded with a SILVER-Nova-TEC-X2 spectrometer (StellarNet, USA). The optical absorption spectra were recalculated from the diffuse reflectance spectra using the Kubelka-Munk theory. Electron paramagnetic resonance (EPR) study was carried out using X-band EPR spectrometer Varian E12 (~ 9.5 GHz) equipped by paired cavity that allows recording EPR signals from a standard (Mn-doped MgO sample) and a ceramic sample separately by modulation switching. The EPR spectra were normalized with respect to the intensity of the signal of a standard, as well as on the mass of each sample studied.

3. Results

In Fig. 1, the XRD patterns of the samples synthesized under various ratios of MgO and TiO_2 are shown. The ceramics fabricated under 1:1 ratio contains a major rhombohedral MgTiO_3 (ICDD PDF 010–75-3957) and a minor orthorhombic MgTi_2O_5 (ICDD PDF 010–89-6943) crystalline phases. This composition is typical for MgTiO_3 produced by sintering of MgO and TiO_2 powders [13]. The MgTi_2O_5 phase is generally formed in a minor amount as a second phase when sintering MgO and TiO_2 in equimolar proportions [13,25]. However, the phase composition changes as the excess MgO is added to the charge. The XRD patterns of the samples obtained with 1.19:1, 1.5:1 and 2:1 ratios demonstrate the peaks of both MgTiO_3 crystal phase and cubic Mg_2TiO_4 phase (ICDD PDF 010–73-1723). No traces of MgO phase are found. The increase of excess MgO content results in the increase of Mg_2TiO_4 phase and the decrease of MgTiO_3 phase concentrations. The ratios of Mg_2TiO_4 and MgTiO_3 phases estimated by Reference Intensity Ratio (RIR) method for the samples produced under 1.19:1 and 1.5:1 M ratios agree with charge composition (MgO and TiO_2 amounts). In the ceramics synthesized under 2:1 M ratio, the Mg_2TiO_4 phase dominates. This is to be expected since such ratio corresponds to stoichiometric composition of Mg_2TiO_4 .

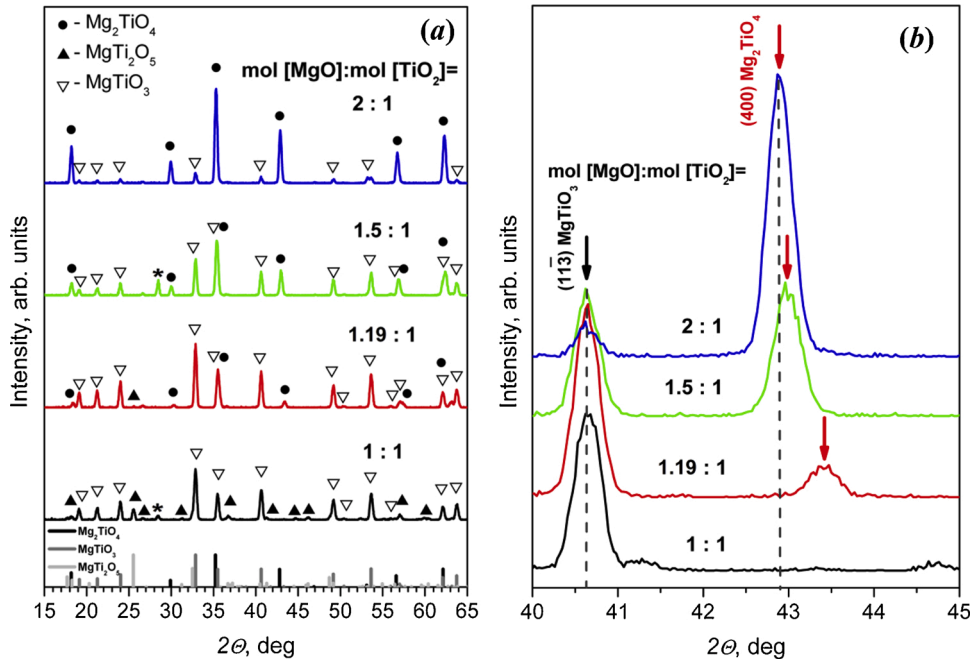


Fig. 1. XRD patterns of the powdered Mn-doped magnesium titanates sintered from MgO and TiO₂ of different molar ratios. The patterns are shifted along y-axis for convenience. The PDF-cards of Mg₂TiO₄ and MgTiO₃ are shown by linear spectra. The peak at about 28° marked by the symbol * is due to Si wafer used as a holder.

However, the Mg₂TiO₄ is known to be unstable at temperatures lower than 1400 °C and suffers from decomposition into MgTiO₃ and MgO [26]. As a result, the secondary MgTiO₃ phase is usually observed when Mg₂TiO₄ is produced by solid state reaction in stoichiometric proportions of 2:1 [27,28]. Since the most intense XRD peaks of cubic MgO overlap with the XRD peaks of cubic Mg₂TiO₄, it is difficult to confirm the presence of MgO crystal phase as well as to estimate reliably the phase ratios in this ceramic sample.

The reactions of MgTiO₃ and MgTi₂O₅ phase formation can be described as follows:



It has been shown [5,25] that adding of excess MgO as low as 2 mol. % accelerates MgTiO₃ formation and inhibits formation of the MgTi₂O₅ in accordance with the reaction:



However, excess MgO also promotes the appearance of Mg₂TiO₄ crystal phase [25]. The reactions of Mg₂TiO₄ formation and decomposition can be described as follows:



The adding of excess MgO to the charge does not affect the peak positions in XRD patterns for MgTiO₃ phase, while the peak positions of the Mg₂TiO₄ phase are found to be shifted towards larger angles 2θ as compared with the peak positions for unstrained Mg₂TiO₄ (Fig. 1b). The magnitude of the shift is the largest for the sample fabricated under 1.19:1 ratio. In particular, the shift of (400) diffraction peak of the Mg₂TiO₄ phase at 2θ = 42.8 degrees is about 0.5 degree (Fig. 1b). The shift magnitude decreases as the amount of excess MgO increases. The evaluated lattice parameter *a* of the cubic Mg₂TiO₄ phase was found to be less than the lattice parameter of unstrained Mg₂TiO₄ (*a* = 8.4420 Å) and decreased from 8.4314 Å for 2:1 ratio, to 8.4089 Å for 1.5:1 ratio

and to 8.3312 Å for 1.19:1 ratio. The Mg₂TiO₄ lattice shrinkage is apparently caused by the presence of excess point defects such as vacancies. Since the ceramics are sintered in the air and cooled slowly to room temperature, high concentration of oxygen vacancies is not expected. Therefore, these point defects are most likely the cation vacancies. This agrees with the fact that the lattice parameter depends on MgO to TiO₂ ratio, i.e. lattice shrinkage (compressive strains) decreases with the increase of excess MgO.

SEM images of cleaved ceramic's surfaces (Fig. 2, a–d) demonstrate similar dense structure with low porosity. The main ceramic body of all samples consists of well-sintered micron-sized grains. In the pores, the grains of different shapes are observed (Fig. 2, e–h). Some of them are faceted and contain the growth steps. The EDX analysis of the regions of about 10 × 10 μm² proved that Mg, Ti and O contents vary according to the MgO : TiO₂ molar ratio. In particular, the content of elements changes from [Mg] = 19 at%, [Ti] = 34 at%, [O] = 47 at% for 1:1 ratio to [Mg] = 20 at%, [Ti] = 26 at%, [O] = 54 at% for 1.19:1 ratio, and is [Mg] = 36 at%, [Ti] = 18 at%, [O] = 46 at% for 2:1 ratio. However, the EDX analysis of smaller micron-sized regions revealed noticeable variations in elemental composition over the certain ceramic body. In particular, in the ceramics of 1:1 and 1.19:1 ratios the regions with the reduced Mg content [Mg] = 7 at%, [Ti] = 53 at%, and [O] = 40 at% are found. In the ceramics of 1.19:1 ratio, the regions with the increased Mg and decreased Ti contents ([Mg] = 22 at%, [Ti] = 19 at%, [O] = 59 at%) are also observed. Such a large spread of composition reflects the presence of different magnesium titanate phases in the ceramic bodies.

As MgTiO₃ is produced via sintering of mixed MgO and TiO₂ powders, solid state chemical reaction between the components occurs during thermal treatment. The main transport mechanism is diffusion. Similarly to ZnO-Al₂O₃ pair, in MgO + TiO₂ powder mixture, diffusion of only one species into the other is considered, namely, faster diffusion of magnesium into titania grains is supposed [13,29]. The diffusion occurs in such a way that the product of reaction, MgTiO₃, forms as layers within an angle of certain aperture on the TiO₂ grains forming the neck growth regions between the MgO and TiO₂ grains [13]. Then, coalescence of neighbor grains takes place. This stage of sintering is captured in SEM images (Fig. 2). The faceted grains and the growth steps found on the grains in the pores (Fig. 2e–h) testify to vapor transport

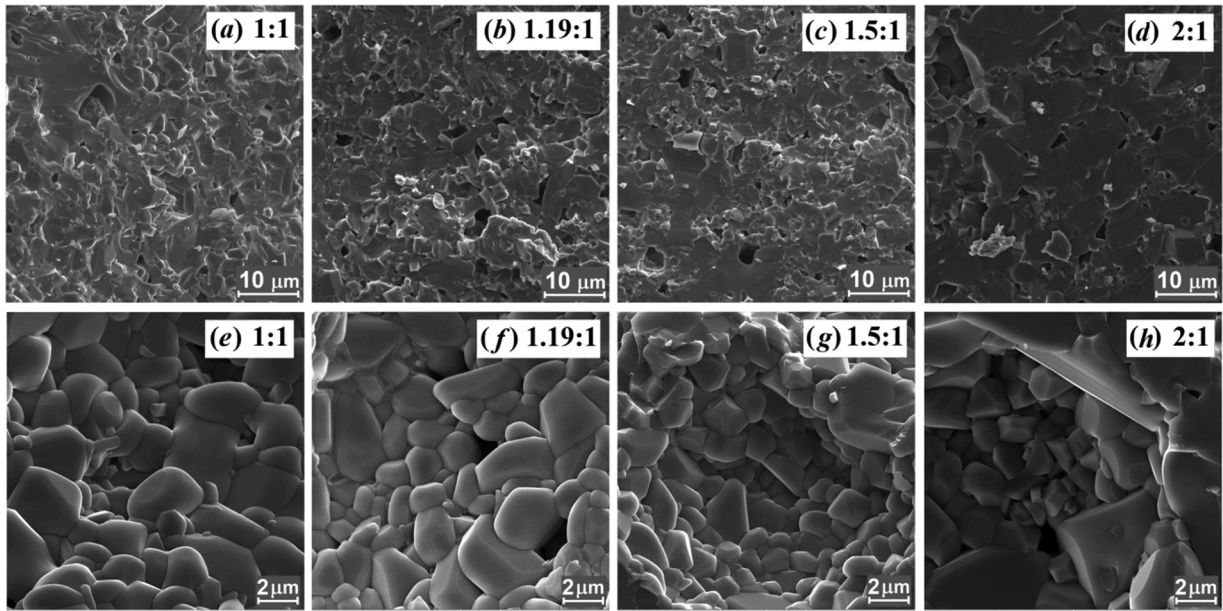


Fig. 2. SEM images of cleaved surface (a-d) and pore region (e-h) of Mn-doped magnesium titanate ceramics sintered from MgO and TiO₂ of different molar ratios.

accompanied by sublimation and condensation as one more transport mechanism contributing to sintering of ceramics studied.

In [29], analysis of magnesium titanate phases formed during

sintering of MgO and rutile single crystals had shown that the reacted layer consisted mainly of thin Mg₂TiO₄ layer adjacent to MgO piece, thick MgTiO₃ layer and relatively thick MgTiO₅ layer adjacent to TiO₂.

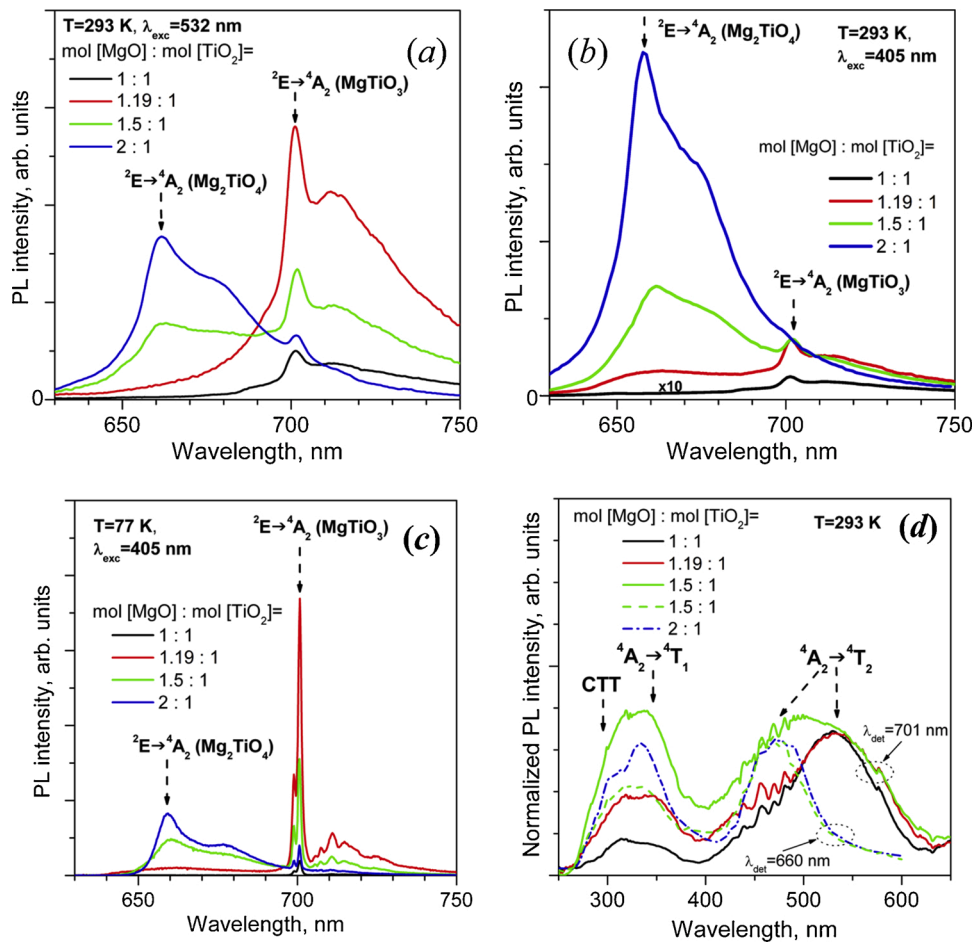


Fig. 3. PL (a, b, c) and normalized PL excitation spectra (d) of the Mn-doped ceramics sintered from MgO and TiO₂ of different molar ratios and detected at T = 293 K (a, b, d) and 77 K (c) under 532 nm (a) and 405 nm (b, c) excitation lines and at 701 nm and 660 nm detection wavelengths (d).

Therefore, it is naturally to suppose that in the ceramics synthesized with 19 and 50 mol.% excess MgO formation of Mg_2TiO_4 and MgTiO_3 crystal phases occurs in the same order along the path of magnesium diffusion from MgO into TiO_2 grains. The one-way diffusion, which proceeds via vacancy mechanism is accompanied by opposite movement of the lattice vacancies, which gather in the material of faster diffusing species. Therefore, accumulation of Mg vacancies (V_{Mg}) can be expected in the Mg_2TiO_4 phase, and concentration of vacancies will be the higher, the smaller the Mg_2TiO_4 phase content. Thus, magnesium vacancies can be considered as the main reason of compressive strains in the Mg_2TiO_4 phase.

The room temperature PL spectra excited by 532 nm laser line (Fig. 3a) show red band with relatively narrow peak at 701 nm (R-line) caused by spin forbidden ${}^2\text{E} \rightarrow {}^4\text{A}_2$ transitions of Mn^{4+} ions in MgTiO_3 . The lowest intensity of this PL is found in the ceramics synthesized under 1:1 M ratio. As 19 mol.% excess MgO is added, the emission increases by about 6 times. However, the increase of molar ratio to 1.5:1 and 2:1 produces gradual decrease of PL intensity.

The PL spectra recorded under 405 nm excitation show the similar tendencies, but with different variations of PL intensity (Fig. 3b). In particular, adding of 19 mol.% excess MgO causes the increase of PL at 701 nm in 30 times with respect to the PL of the ceramics synthesized under 1:1 M ratio. At the same time, the PL intensity remains the same for 1.5:1 and 2:1 M ratio. This is apparently due to the contribution of the emission caused by Mn^{4+} ions in Mg_2TiO_4 phase, which arises starting from 19 mol.% excess MgO. The PL band due to Mn^{4+} in Mg_2TiO_4 is located in the red and is peaked at about 660 nm (R-line emission). It is more pronounced under UV excitation (Fig. 3b). The intensity of this emission increases as an amount of Mg_2TiO_4 phase increases according to the XRD data. In the ceramics synthesized with 19 mol.% excess MgO, the PL band is fully unstructured, and in the ceramics synthesized with 50 mol.% excess MgO the peak at about 660 nm becomes noticeable. It should be noted that as the excess MgO content increases, the PL peak position slightly shifts towards shorter wavelengths (on about 2 nm), and a relative intensity of R-line decreases as well as the band broadens at least in the short wavelength region.

It can be supposed that low intensity of R-line is related to the presence of V_{Mg} and corresponding deformations of oxygen octahedron in which Mn^{4+} is incorporated. It has been proposed [30] that if the Mn^{4+} ion is located in non-centrosymmetric site (absence of inversion symmetry), the parity selection rule for ${}^2\text{E} \rightarrow {}^4\text{A}_2$ transition relaxes, the electric dipole intensity increases and the R-line intensity increases. Therefore, it can be supposed that compressive stress results in more centrosymmetric site for octahedral coordinated Mn^{4+} ion in Mg_2TiO_4 . Deformation of oxygen octahedron can also affect splitting of Mn^{4+} emitting levels, and nonuniform stresses can cause lost of the structure of the emission band.

The low temperature PL spectra of the ceramics excited by 405 nm laser line show the same trends for Mn^{4+} emission spectra via excess MgO content as the room temperature spectra (Fig. 3c). The difference is that the intensity of Mn^{4+} PL in the MgTiO_3 phase increases stronger than in the Mg_2TiO_4 phase upon cooling to liquid nitrogen temperature. This is apparently due to more intense thermal quenching of Mn^{4+} PL in MgTiO_3 as compared with Mg_2TiO_4 [8,31].

The excitation spectra of Mn^{4+} PL in MgTiO_3 detected at 701 nm show two broad bands centered at 325 and 535 nm (Fig. 3d). The UV band is due to both charge transfer transition (CTT) and ${}^4\text{A}_2 \rightarrow {}^4\text{T}_1$ transition, and the green band corresponds to ${}^4\text{A}_2 \rightarrow {}^4\text{T}_2$ spin allowed transition of Mn^{4+} ions in MgTiO_3 . In the ceramics obtained under equimolar ratio of MgO and TiO_2 , the intensity of green excitation band significantly exceeds the intensity of the UV band. When the excess MgO is added, relative intensity of the UV excitation band becomes higher. In addition, the green excitation band expands to the blue. The broadening of the excitation band is apparently caused by contribution of Mn^{4+} PL of the Mg_2TiO_4 phase. In fact, in the ceramics produced under 1.19:1 and 2:1 ratio, the PL excitation spectra detected at 660 nm (Fig. 3d) are

found to be similar.

Fig. 4a shows the PL decay curves in the ceramics studied. In the sample produced under 1:1 ratio, the decay curve of 701 nm emission shows a single exponential behavior with a PL decay time of 17 μs . The relaxation of PL at 660 nm occurs with decay time of 14 μs (Fig. 4b). These magnitudes agree well with decay time of 25 μs reported for Mn^{4+} emission in the MgTiO_3 [8] and can be explained by intense PL thermal quenching at room temperature. At the same time, the PL decay curves of the ceramics with 19 and 50 mol.% excess MgO contain the regions of faster and slower decline. The PL decay times of slower decline region are in the range of 0.40–0.45 ms (Table 1). In these samples, the relaxation of the PL at 660 nm occurs with similar decay times of 0.44–0.50 ms (Table 1). These values are close to PL decay times reported for Mn^{4+} PL in Mg_2TiO_4 [23]. Therefore, the region of slower decline is apparently caused by contribution of Mn^{4+} emission in the Mg_2TiO_4 phase [10]. The decay time of this PL is the smallest in the sample with the largest compressive strains in the Mg_2TiO_4 grains (1.19:1 ratio). This implies that deformations in Mg_2TiO_4 affect not only the shape of Mn^{4+} PL spectrum, but also the PL decay. At room temperature the relaxation times of Mn^{4+} PL in Mg_2TiO_4 and MgTiO_3 are influenced by contribution of non radiative recombination caused by PL thermal quenching [8,12]. The latter is usually ascribed to thermally activated crossing of the $\text{Mn}^{4+} {}^4\text{A}_2$ ground state and some excited state (${}^4\text{T}_2$ state or CTT). Thus, the shortening of PL relaxation time is probably due to the increased contribution of PL thermal quenching. In the samples with excess MgO, a noticeable slowing down of PL relaxation is found for the region of faster decline (Fig. 4b). However, because of changing slope it is difficult to say whether the decay time of Mn^{4+} PL in the MgTiO_3 increases or not with adding of excess MgO.

The optical absorption of the ceramics also changes with adding of excess MgO. The sample produced with stoichiometric ratio 1:1 has no pronounced color, and those with 19 mol.% excess MgO is pink. As the excess MgO content increases, the color acquires a brownish yellow tint, apparently due to contribution of optical absorption in Mn^{4+} activated Mg_2TiO_4 phase. In fact, the absorption spectrum of the ceramics with 19 mol.% excess MgO (Fig. 5) indicates that a pink color is due to absorption band peaked at about 540 nm and stretched to the red. This band is apparently caused by ${}^4\text{A}_2 \rightarrow {}^4\text{T}_2$ transition of Mn^{4+} ions in the MgTiO_3 phase. The same band of smaller intensity is found in the absorption spectrum of the ceramics produced with 1:1 ratio. The rise of this absorption with adding of excess MgO indicates the increase of concentration of Mn^{4+} centers in the MgTiO_3 host. In the ceramics with 50 mol.% excess MgO, intensity of this band decreases and another one at about 480 nm arises. This absorption band is obviously due to ${}^4\text{A}_2 \rightarrow {}^4\text{T}_2$ transition of Mn^{4+} ions in the Mg_2TiO_4 phase and its intensity agrees with content of Mg_2TiO_4 phase in the ceramics.

The EPR spectra of the Mn-doped ceramics presented in Fig. 6 show the same sextets of lines with the g-factor about 2.00. This signal is apparently originated from Mn ions in MgTiO_3 phase. The experimental spectra were simulated using the program Powder included in the software package Visual-EPR [32]. The program is based on an algorithm that uses an exact solution of the spin-Hamiltonian including electron Zeeman interaction, the fine and hyperfine interactions. The spin-Hamiltonian parameters of the simulated signal are derived as follows: g-factor value is 2.004, fine-structure parameter $b_2 = 0.0165 \text{ cm}^{-1}$ and ultrafine-structure parameter $A = 0.0079 \text{ cm}^{-1}$. These magnitudes agree well with the parameters of Mn^{2+} ions ($S = 5/2$) in crystal lattice of MgTiO_3 reported by other authors [33,34]. The adding of excess MgO promotes the decrease of both the intensity and halfwidth of the EPR signal (Fig. 6) that indicates the decrease of concentration of the Mn^{2+} centers. The concentration of Mn^{2+} ions in MgTiO_3 phase was evaluated taking into account the weight content of this phase in the ceramics estimated by RIR method from the XRD patterns. The concentration decreases from $9 \cdot 10^{19} \text{ spin g}^{-1}$ for 1:1 ratio (72 wt.% of MgTiO_3) to $1.7 \cdot 10^{19} \text{ spin g}^{-1}$ for 1.19:1 ratio (77 wt.% of MgTiO_3) and to $4.5 \cdot 10^{18} \text{ spin g}^{-1}$ for 1.5:1 ratio (44 wt.% of MgTiO_3).

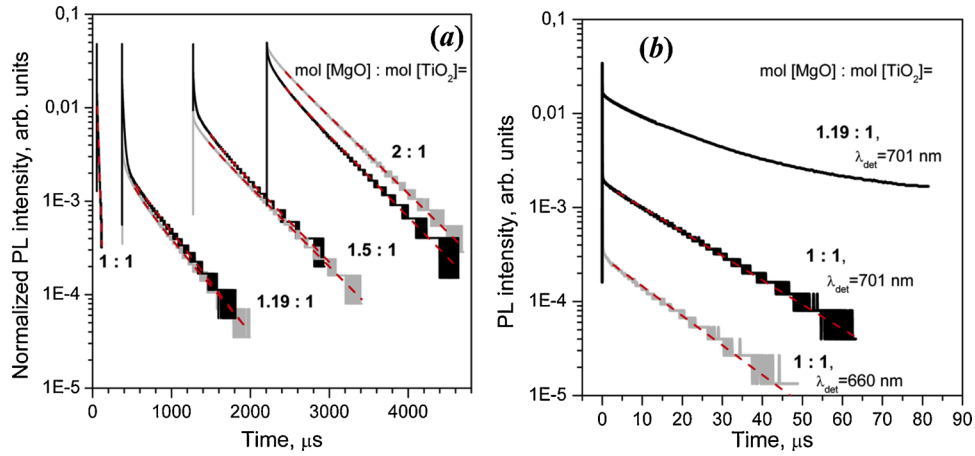


Fig. 4. PL decay curves detected at 701 nm (black) and 660 nm (gray) for the ceramics produced from MgO and TiO₂ of different molar ratios and their approximation by exponential dependence (dotted lines). The curves on (a) are shifted on time axis for convenience.

Table 1

PL decay times obtained from approximation by exponential dependence of PL relaxation detected at 701 and 660 nm for the ceramics produced from MgO and TiO₂ of different molar ratios.

MgO to TiO ₂ molar ratio, mol [MgO] : mol [TiO ₂]	PL decay time, μs (detected at 701 nm)	PL decay time, μs (detected at 660 nm)
1 : 1	17	14
1.19 : 1	405	438
1.5 : 1	453	500
2 : 1	499	520

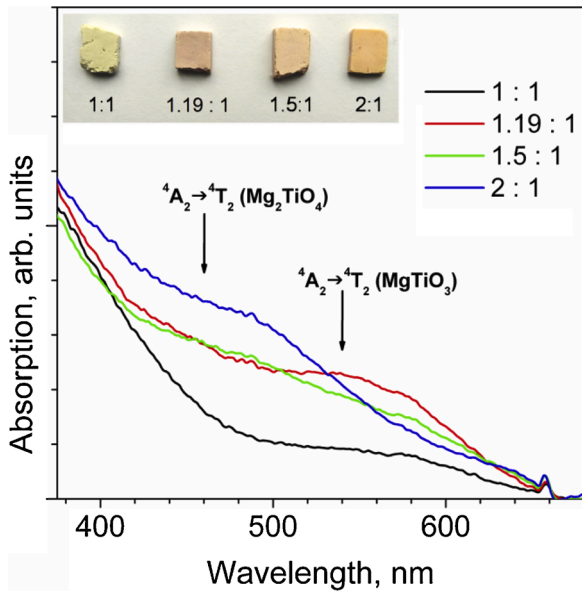


Fig. 5. Room temperature optical absorption spectra of the Mn-doped ceramics sintered from MgO and TiO₂ of different molar ratios.

It should be noted that no EPR sextets caused by Mn⁴⁺ ion in MgTiO₃ were detected even in the ceramics with excess MgO content that showed the pronounced Mn⁴⁺ absorption and PL. In fact, the Mn⁴⁺ ion (S = 3/2) is usually characterized by lower value of g-factor and much higher fine constant than the Mn²⁺ ion [35]. Due to the latter, the signal from Mn⁴⁺ ions can also be observed in the low field region. This is often used as a signature of Mn⁴⁺ [35,36]. No signal that can be ascribed to

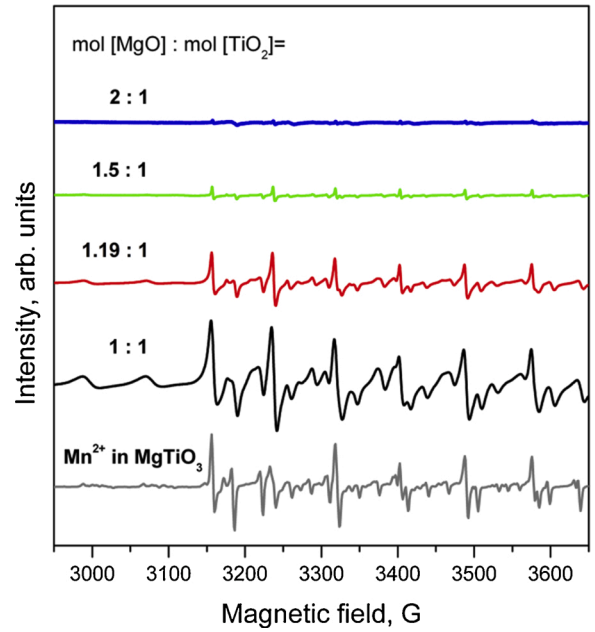


Fig. 6. Experimental EPR spectra of Mn-doped ceramics sintered from MgO and TiO₂ of different molar ratios and simulated spectrum of Mn²⁺ ion in MgTiO₃.

Mn⁴⁺ ions was observed in this region. The possible reason can be the adverse relaxation conditions for Mn⁴⁺ ions, i. e. too short or too long spin relaxation times.

4. Discussion

MgTiO₃ has the trigonal “ilmenite” structure, which consists of a close-packed-hexagonal O²⁻ anion sublattice that has 2/3 of its octahedral sites occupied alternately by Mg²⁺ and Ti⁴⁺ cations [37]. When Mn incorporates into crystal lattice of MgTiO₃, it can replace the sites of the host Mg²⁺ as Mn²⁺ and Ti⁴⁺ as Mn⁴⁺ without the need for charge compensation. Taking into account similarity of the ionic radii of corresponding Mn ions and the host cations (r(Mn⁴⁺, VI) = 0.53 Å, r(Ti⁴⁺, VI) = 0.605 Å, r(Mn²⁺, VI) = 0.67 Å (low spin), r(Mg²⁺, VI) = 0.72 Å), it is difficult to conclude which cation (Ti⁴⁺ or Mg²⁺) manganese prefers to substitute. The EPR spectra of the ceramics sintered in stoichiometric molar ratio of 1:1 reveal the Mn²⁺ ions in concentration close to nominal concentration of Mn dopant. This means that manganese tends to

incorporate into MgTiO₃ as Mn²⁺ on Mg site. This can be an important reason of low intensity of Mn⁴⁺ PL. It should be noted, that Mn²⁺ ions might also enter the Ti⁴⁺ sites being compensated by oxygen vacancy defects. Such situation has been observed in Mn-doped BaTiO₃ ceramics sintered in oxygen deficient atmosphere [38,39]. It has been shown that Mn on Ti site in BaTiO₃ gradually changes its valence from +4 to +3 and finally to +2 with decreasing oxygen partial pressure from 1 atm at elevated temperatures [39]. However, considering sintering conditions of the ceramics studied (the ambient atmosphere and slow cooling to room temperature), the observed intense EPR signal should be ascribed to Mn²⁺ ions on Mg²⁺ site.

It should be noted that optical absorption data do not show specific features caused by optical transitions of Mn²⁺ ions in the MgTiO₃. Usually, absorption spectrum of Mn²⁺ ions in tetrahedral and octahedral sites composes of several narrow and wider bands lying in the UV and blue-green spectral ranges. However, all transitions from the ground to the excited levels of Mn²⁺ are spin-forbidden contrary to spin allowed ⁴A₂→⁴T₁ and ⁴A₂→⁴T₂ transitions of Mn⁴⁺ found in the same spectral regions. Therefore, even if the concentrations of Mn²⁺ and Mn⁴⁺ centers are comparable, optical absorption due to Mn²⁺ ions can be hidden by Mn⁴⁺ transitions. In MgTiO₃, the Mn²⁺ ions occupy the octahedrally coordinated lattice sites and can show a structureless PL band in the orange-red spectral region. However, because of weak excitation the Mn²⁺ PL will be of low intensity and hardly distinguishable on the background of Mn⁴⁺ red PL.

The results of the EPR and optical studies show that adding of excess MgO promotes the decrease of concentration of Mn²⁺ ions and the increase of Mn⁴⁺ ion concentration in the MgTiO₃ phase. It is naturally to suppose that excess MgO hinders substitution for Mg site ions with Mn²⁺ and promotes incorporation of Mn⁴⁺ onto Ti⁴⁺ site. This is similar to a long time ago discovered Ca acceptor behavior in BaTiO₃ ceramics sintered under BaO excess [40]. It has been shown that when there is an excess of alkaline earth oxide ((Ca + Ba)/Ti > 1), some Ca²⁺ ions, which are well known Ba site substituents in BaTiO₃, occupy Ti sites where they act as doubly charged acceptors and compensate oxygen vacancies acting as shallow donors. The Ca acceptor behavior was proved by the electrical conductivity measurements [41] and Atom Location Using Channeling-Enhanced Microanalysis measurements [40]. Similarly, in Mn-doped Mg₂TiO₄ phosphor, the enhancement of manganese incorporation onto the Ti site with adding of excess MgO has been found [23].

The results of EPR study show that concentration of Mn²⁺ in MgTiO₃ phase gradually decreases as the excess MgO increases, which suggest gradual increase of Mn⁴⁺ concentration. However, the intensity of PL at 701 nm shows non-monotonic dependence on excess MgO content (Fig.2a). The largest intensity of Mn⁴⁺ PL is found for 19 mol.% excess MgO. The decrease of PL intensity for 1.5:1 and 2:1 M ratio can be ascribed to the decrease of amount of MgTiO₃ phase in the ceramics (44 and 17 wt.%, respectively, as opposed to 77 wt.% for the ceramics synthesized with 1.19:1 ratio).

Therefore, the rise of Mn⁴⁺ concentration can be considered as the main reason of the PL intensity increase in MgTiO₃ ceramics with excess MgO. The increase is larger for UV excitation than for 532 nm excitation. The difference can be explained by the decrease of the excitation light absorption by the MgTi₂O₅ secondary phase. In fact, the MgTi₂O₅ shows smaller band gap of 3.9 eV [42] as compared with 4.2 eV of MgTiO₃ [43, 44] and 4.33 eV of Mg₂TiO₄ [23,45]. Therefore, direct band-to-band transitions in MgTi₂O₅ can cause strong absorption of UV excitation light in the ceramics produced under 1:1 ratio.

The obtained results show that excess MgO not only affects crystal phase formation, but also changes the manner of Mn incorporation in crystal lattice of MgTiO₃. This should be taken into account when using Mn-doped MgTiO₃ as a component of dielectric ceramics or as a red phosphor. In particular, the adding of excess MgO can be considered as an efficient tool for the increasing of Mn acceptor concentration as well as Mn⁴⁺ PL intensity in MgTiO₃.

5. Conclusions

The effect of excess MgO on Mn⁴⁺ emission and crystal phase formation in Mn-doped magnesium titanates produced by a solid state reaction has been investigated. The samples synthesized using equimolar ratio of MgO and TiO₂ contained the major MgTiO₃ phase and minor MgTi₂O₅ phase. The EPR study revealed intense EPR signal due to Mn²⁺ in MgTiO₃, the estimated concentration of Mn²⁺ centers is found to be close to nominal content of Mn dopant. It is concluded that in addition to PL thermal quenching, both preferential incorporation of Mn onto Mg²⁺ site and formation of secondary MgTi₂O₅ phase, that causes strong absorption of the UV excitation light, are responsible for low intensity of Mn⁴⁺ PL in MgTiO₃. The ceramics synthesized with excess MgO were composed of the MgTiO₃ and Mg₂TiO₄ crystal phases in different weight proportions and demonstrated red PL due to Mn⁴⁺ ions onto Ti⁴⁺ site in these phases. The Mg₂TiO₄ phase is found to be under compressive strains originated from intrinsic defects (presumably, magnesium vacancies). The lattice parameter of Mg₂TiO₄ crystal phase as well as spectral shape and decay times of red PL due to Mn⁴⁺ in the Mg₂TiO₄ are shown to be strongly affected by MgO to TiO₂ ratio. In the MgTiO₃ phase, the adding of excess MgO produces successive decrease of Mn²⁺ ion concentration and the increase of Mn⁴⁺ PL intensity. The increase of Mn⁴⁺ optical absorption testified the increased concentration of Mn⁴⁺ centers. It is supposed that excess MgO hinders substitution for Mg site by Mn²⁺ ions and promotes the incorporation of Mn⁴⁺ ion onto Ti⁴⁺ site in MgTiO₃. The adding of excess MgO can be considered as an efficient tool for the increase of Mn⁴⁺ PL intensity in MgTiO₃.

CRedit authorship contribution statement

L. Borkovska: Conceptualization, Writing - original draft, Supervision. **L. Khomenkova:** Investigation, Formal analysis, Writing - review & editing. **T. Stara:** Investigation, Formal analysis. **I. Vorona:** Investigation, Conceptualization, Writing - original draft. **V. Nosenko:** Investigation, Formal analysis. **O. Gudymenko:** Investigation, Formal analysis. **V. Kladko:** Conceptualization, Writing - review & editing. **K. Kozoriz:** Investigation. **C. Labbé:** Investigation, Writing - review & editing, Supervision. **J. Cardin:** Writing - review & editing. **J.-L. Doualan:** Investigation. **T. Kryshtab:** Conceptualization, Writing - original draft.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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