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Band-gap variation and self-redox effect induced by compositional deviation in \( \text{Zn}_x\text{Ga}_2\text{O}_{3+x}:\text{Cr}^{3+} \) persistent phosphors

Yixi Zhuang,\textsuperscript{a} Jumpei Ueda,\textsuperscript{b} Setsuhisa Tanabe\textsuperscript{c} and Pieter Dorenbos\textsuperscript{d}

We made a systematic investigation on the optical properties of non-doped and Cr-doped \( \text{Zn}_x\text{Ga}_2\text{O}_{3+x} \) \((0.98 \leq x \leq 1.02)\) spinel crystals. Absorption, photoluminescence excitation, and persistent luminescence excitation spectra indicated that the bottom of conduction band (CB) was affected (i) by compositional deviation from stoichiometry or (ii) by atmosphere control in a similar way. By constructing an energy level diagram, a broad distribution of trap depth in a Cr\(^{3+}\)-doped \( \text{Zn}_x\text{Ga}_2\text{O}_{3+x} \) sample with composition of Zn excess or in one prepared in O\(_2\) atmosphere was attributed to lowering of the CB bottom, which resulted in faster decay of persistent luminescence in these two samples. Defect chemical reactions revealed that the anti-sites species of \( \text{Zn}^{'}\text{Ga} \) may be the reason for lowering of the CB bottom. A self-redox effect of Cr ions induced by the compositional deviation was found and explained by the defect reactions. The investigation indicated that the composition with slight Zn deficiency is important to obtain a narrow distribution of trap depths and longer persistent luminescence, however the self-reduction effect of Cr ions in the composition of Zn deficiency also need to be considered.

1 Introduction

Persistent phosphors, which continue light emission after removal of the excitation source, have long been of scientific interests due to their important applications as luminous paints, identification markers, or optical memories.\(^1\)\(^-\)\(^5\) Recently, another attractive application in \textit{in vivo} medical imaging system by using nano-sized persistent particles as optical probe was proposed by a French group, which focused more attention on the persistent phosphors.\(^6\)\(^-\)\(^7\) One of the key requirements for \textit{in vivo} imaging application is that the luminescence wavelength of persistent particles should be inside the biological window from 650 to 1300 nm where the optical signal shows rather high transmittance through the biological tissues.\(^8\) This spectral requirement rules out most Eu\(^{2+}\)-activated phosphors showing excellent persistent performance in the blue or green regions.\(^9\)\(^-\)\(^13\) Until now, only few Eu\(^{2+}\)-activated compounds were known to exhibit persistent luminescence beyond 650 nm. Most of them are hygroscopic sulfides and unsuitable for biological application.\(^14\)\(^-\)\(^15\) Alternatively, several Cr\(^{3+}\)-activated gallate phosphors, especially \( \text{ZnGa}_2\text{O}_4 \)-based spinel crystals, were reported showing long-lasting persistent luminescence from 680 to 1000 nm, which were considered as promising optical materials for \textit{in vivo} imaging application.\(^16\)\(^-\)\(^26\) Bessière \textit{et al.} firstly reported red persistent luminescence at around 695 nm in \( \text{Cr}^{3+}\)-doped \( \text{ZnGa}_2\text{O}_4 \) crystals in 2011.\(^16\) Pan and Allix \textit{et al.} greatly improved the red persistent luminescence by partially substituting Ge to form \( \text{Zn}_{1+x}\text{Ga}_{2-2x}\text{Ge}_x\text{O}_4 \) solid solution. The persistent luminescence was detectable by night vision equipment up to 360 h after the excitation source was removed.\(^17\)\(^-\)\(^18\) We also found that by codoping Bi\(_2\)O\(_3\) into \( \text{Zn}_{0.98}\text{Ga}_2\text{O}_{3.98} \) crystals, the intensity of red persistent luminescence can be improved by 10 times. The absolute intensity (radiance) of persistent luminescence in \( \text{Zn}_{0.98}\text{Ga}_2\text{O}_{3.98}:\text{Cr,Bi} \) phosphors was comparable to that in \( \text{SrAl}_2\text{O}_4: \text{Eu,Dy} \) phosphors.\(^19\) Recently, Abdakayum \textit{et al.} succeeded to demonstrate \textit{in vivo} imaging experiments by using \( \text{Zn}_{1+x}\text{Ga}_{2-2x}\text{Ge}_x\text{O}_4:\text{Cr,Pr} \) nano-particles. The feasibility, including Cr\(^{3+}\) persistent luminescence duration, particle biocompatibility, water solubility \textit{etc.} was proved in this study.\(^20\) Although notable progress on intensity improvement or nanoparticulation has been made, the Cr\(^{3+}\)-doped \( \text{ZnGa}_2\text{O}_4 \) crystals need further investigation in order to understand the energy levels governing the red persistent luminescence. As mentioned by Bessière \textit{et al.} in their milestone paper, a sample with Zn deficiency in the composition showed more intense persistent luminescence than other samples.\(^16\) The composition of Zn deficiency was adopted by many researchers. However, the reason for the composition dependency has not been given.
A systematic study on persistent luminescence induced by compositional deviation may give much information on structure variation, and offer theoretical guidance for the design of better persistent phosphors.

In this paper, we introduce methods to discuss the persistent luminescence mechanism by building an energy level diagram (Sec. 2.1) and to discuss structural variation by using a defect formation equation (Sec. 2.2), which are applied in the following sections. In Sec. 3, synthesis conditions (nominal composition and applied atmosphere) of prepared samples and important details about optical measurements are described. The results are summarized and discussed in Sec. 4. It starts from a discussion on band-gap variation in non-doped \( \text{Zn}_x\text{Ga}_2\text{O}_{3-x} \) samples prepared in different conditions (either composition or atmosphere) (Sec. 4.1), and moves to optical characterizations in various \( \text{Cr}^{3+} \)-doped \( \text{Zn}_x\text{Ga}_2\text{O}_{3-x} \) samples by discussing the change of valence state in Cr ions (Sec. 4.2) and the variation of energy level diagram as well as \( \text{Cr}^{3+} \) persistent luminescence performance (Sec. 4.3). Some conclusions are made in Sec. 5.

2 Methodology

2.1 Building an energy level diagram to discuss the persistent luminescence mechanism

It is indispensable to understand the mechanism of persistent luminescence for successful material synthesis. However, it was a bigger challenge to give a clear explanation for the observed persistent phenomenon, rather than to discovered an excellent material (sometimes accidently) in the past.\(^9\,2^7\) Several models, generally by using a simple energy level diagram to describe the charge carrier transfer process between active centers and trap centers, were proposed by different researchers.\(^4\) Two major controversial points in these models are (i) the species of charge carrier from the luminescent center (whether a negative electron or a positive hole) captured by the trap center and (ii) the path of the trapping process (whether through the conduction band or through the valence band or by a quantum tunnelling effect in the band-gap). The persistent luminescence mechanism can be correctly determined only when the relative energy locations of (i) the valence band (VB) and the conduction band (CB) of the host, (ii) the ground state (GS) and excited states (ES) of the luminescent center as well as, (iii) the trap energy levels are clear. It is significant to determine the relative locations of these energy bands/levels in the research of persistent phosphors, by using various optical characterizations.

In recent publications, most experimental results confirm that an electron from a luminescent center captured by a trap center through the CB is the most efficient path for charge carrier trapping process, although the contribution of quantum tunnelling cannot be completely excluded. Possible trapping processes are shown in Fig. 1. The electron can be excited to the CB in three ways: (i) from GS to ES above the bottom of CB; (ii) directly from GS to CB; (iii) from GS to ES below the bottom of CB. In the third case, the electron can “jump” to CB with thermal energy, which was named thermally assisted photoionization process. The specific trapping process is determined by the relative locations of the energy bands/levels.

Dorenbos has developed an empirical model to determine the GS and ES (f and d orbitals) of \( \text{R}^{2+} \) and \( \text{R}^{3+} \) (rare earth ions in +2 and +3 valence states) relative to CB and VB, which was a very useful tool to discuss the persistent properties in many rare-earth-activated phosphors.\(^2^8\,3^1\) We consider that the same research approach can be applied to transition-metal-activated materials. Firstly, a host referred binding energy (HRBE) scheme is adopted, where the energy of the VB top is defined as zero. The HRBE of the CB bottom can be determined by the band-gap energy from absorption spectra of a non-doped sample. In the second step, we need to know the relative locations of energy levels of luminescent center (GS and ES) relative to the energy bands of the host (CB and VB). Photoconductivity excitation (PCE) spectra, persistent luminescence excitation spectra (PersLE), or photoluminescence excitation (PLE) spectra measurements are useful to evaluate the energy of electronic transition from the GS to the bottom of CB (arrow ii in Fig. 1). Also, the energy values for 3d-3d intra-transitions from GS to various ES can be read from absorption spectra or PLE spectra in a transition-metal-doped sample. Finally, the depth of trap levels (\( \Delta \epsilon \), thermal activation energy) can be determined in the measurement of thermoluminescence (TL) glow curves.
2.2 Using Kröger-Vink notation to discuss structural variation

In many persistent phosphors, synthesis condition (compositional deviation from stoichiometry or atmosphere control) show either positive effect or negative effect on the persistent luminescence performance. It is helpful if we know what kind of structural change occurs, i.e., which species of defects are created in a real synthesis process. Since direct detection of the defects is generally difficult or unavailable, a theoretical method by writing defect chemical reactions in Kröger-Vink notation can be used. The key defects responsible for the optical properties could be inferred by connecting the applied synthesis condition with the obtained optical properties, although experimental proof is usually required to verify this. For example, the great improvement of persistent luminescence in SrAl₂O₄:Eu²⁺ phosphor by Dy³⁺ codoping was as attributed to the formation of DySr⁺ defect as following:

\[
\text{Dy}_2\text{O}_3 + 2\text{Al}_2\text{O}_3 \xrightarrow{\text{SrAl}_2\text{O}_4} 2\text{Dy}^\text{Sr} + 4\text{Al}^\text{Al} + 8\text{O}_0^\text{Zn} + 2e^- + \frac{1}{2}\text{O}_2 \uparrow \quad (1)
\]

In Sec. 4, we will show what kinds of anti-sites may be created under certain conditions and how these defects affect the optical properties in ZnₓGa₂O₃₊ₓ:Cr³⁺ phosphors by using the Kröger-Vink notation to describe defect chemical reactions.

3 Experimental Procedures

3.1 Sample preparation

Poly-crystalline ceramics were prepared by using a solid-state reaction method. The chemical reagents, ZnO (4N), Ga₂O₃ (4N), and Cr₂O₃ (3N), were used as starting materials. The starting materials were ground in an alumina mortar to form homogeneous and fine powder mixtures. Pellets (0.8 g and φ15 mm) were shaped from the powder mixtures with a uniaxial loading of 50 MPa in a stainless steel mold. The compacts were carefully moved to an electric furnace and fired at 1300 °C for 5 h.

Nominal composition (the molar ratio of Zn to Ga) and applied atmosphere (O₂ partial pressure) were used as two variables to investigate their effects on optical properties. Various synthesis conditions for non-doped ZnₓGa₂O₃₊ₓ samples are shown in Table 1. Firstly the atmosphere was fixed as Air and the composition was varied from 2% Zn deficiency (Zn-d), stoichiometric (Zn-s), to 2% Zn excess (Zn-e) to obtain samples non-d-Air, non-s-Air, and non-e-Air, respectively. Then the composition was kept as stoichiometric and the gas flow was changed from pure N₂ (PO₂ = 0%), Air (PO₂ = 21%), to pure O₂ (PO₂ = 100%) to obtain samples non-s-N₂, non-s-Air, and non-s-O₂. In Table 2, Cr-doped ZnₓGa₂O₃₊ₓ samples were prepared in the same synthesis conditions except for 0.5 mol% substitution of Cr₂O₃ for Ga₂O₃.

### Table 1 Nominal compositions and applied atmospheres of non-doped ZnₓGa₂O₃₊ₓ poly-crystalline ceramics.

<table>
<thead>
<tr>
<th>Comp. (Zn/Ga ratio)</th>
<th>Atmosphere (PO₂)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Zn₀.98Ga₂O₄.₀₂: Zn-d</td>
<td>N₂</td>
</tr>
<tr>
<td>Zn₁.₀₀Ga₂O₄.₀₀: Zn-s</td>
<td>non-s-O₂</td>
</tr>
<tr>
<td>Zn₁.₀₂Ga₂O₄.₀₂: Zn-e</td>
<td>non-s-N₂</td>
</tr>
</tbody>
</table>

### Table 2 Nominal compositions and applied atmospheres of Cr-doped ZnₓGa₂O₃₊ₓ poly-crystalline ceramics. All the synthesis conditions were the same with the non-doped samples except for using 0.5 mol% Cr₂O₃ as dopant.

<table>
<thead>
<tr>
<th>Comp. (Zn/Ga ratio)</th>
<th>Atmosphere (PO₂)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Zn₀.₉₈(Ga₀.₉₉Cr₀.₀₁)O₄.₀₂: Zn-d</td>
<td>Cr-d-Air</td>
</tr>
<tr>
<td>Zn₁.₀₀(Ga₀.₉₉Cr₀.₀₁)O₄.₀₀: Zn-s</td>
<td>Cr-s-Air</td>
</tr>
<tr>
<td>Zn₁.₀₂(Ga₀.₉₉Cr₀.₀₁)O₄.₀₂: Zn-e</td>
<td>Cr-e-Air</td>
</tr>
</tbody>
</table>

3.2 Optical Measurements

Diffuse reflectance spectra of the samples were measured by using a scanning-type spectrophotometer (Shimadzu, UV3600) with a BaSO₄-based integrating sphere. The scanning wavelength region was from 850 to 200 nm. Photoluminescence (PL), photoluminescence excitation (PLE), and persistent luminescence excitation (PersLE) spectra were measured in a fluorescence spectrophotometer (Shimadzu, RF5000). Persistent luminescence excitation spectra were obtained by plotting Cr³⁺ persistent luminescence intensity in 60 s after excitation for 60 s at each excitation wavelength. Excitation wavelength was changed from 500 to 200 nm with a step of 5 nm. Before moving to the next excitation wavelength, a waiting time of 10 min was set to reduce the effect of the last excitation. Thermoluminescence (TL) glow curves of Cr³⁺ emission (around 695 nm) were recorded using a silicon photodiode (Electro-Optical System Inc., S-025-H). A band-pass filter (350-750 nm) and a long-pass filter (> 580 nm) were fixed before the silicon photodiode to cut off the noise. The samples were first cooled to 100 K (-173 °C) and exposed to a 300 W xenon lamp (Asahi Spectra, MAX 302) in an ultra-violet (UV) mirror module (250 to 380 nm) for 5 min. After another 5 min of waiting time, the samples were heated to 600 K (327 °C) in a heating rate of 10 K/min. The persistent luminescence decay curves at room temperature (RT, 20 °C) were recorded by using the same silicon photodiode after the samples were excited by the same xenon lamp in the UV module for 5 min. Then, the afterglow intensity was converted to an absolute intensity (radiance, in unit of mW·sr⁻¹·m⁻²) by using a CCD spectrometer (B&W Tek, Glacier X) calibrated by Konica-Minolta company.
4 Results and Discussion

4.1 Band-gap absorption and self-activated emission in non-doped Zn$_x$Ga$_2$O$_{3+x}$

In order to evaluate the optical band-gap, the diffuse reflectance ($R_\infty$) of the non-doped Zn$_x$Ga$_2$O$_{3+x}$ pellets were converted to a Kubelka-Munk function $F(R_\infty)$ according to:\cite{38}

$$ F(R_\infty) = S \times (1 - R_\infty)^2/(2 \times R_\infty). \quad (2) $$

As an approximation, the diffusion coefficient $S$ was treated independent on wavelength. Then a $(\hbar v F(R_\infty))^2 - \hbar v$ plot was made according to an expression proposed by Tauc:\cite{39}

$$ (\hbar v \times F(R_\infty))^2 = A \times (\hbar v - E_g). \quad (3) $$

The values of the band-gap energy can be read from the intercepts of fitted straight lines.

![Fig. 2](image-url) (h$\nu$F($R_\infty$))$^2$-$\hbar v$ plot of the non-doped Zn$_x$Ga$_2$O$_{3+x}$ samples. Band-gap energy for each sample was estimated from the intercept of a fitted straight line. The optical band-gap energy decreased in the non-e-Air and non-s-O$_2$ samples.

![Fig. 3](image-url) PL and PLE spectra of the non-doped Zn$_x$Ga$_2$O$_{3+x}$ samples. Top abscissa and bottom abscissa in each figure are scaled in wavelength (nm) and energy (eV), respectively. Excitation wavelength for the PL spectra (left part) was 250 nm; monitoring wavelength for the PLE spectra (right part) was 360 nm.

The fitting results of the three samples with different compositions (molar ratio of Zn to Ga) are given in Fig. 2(a). The sample with Zn deficiency (blue) showed a band-gap of 4.90 eV. The band-gap energy decreased to 4.88 and 4.58 eV when the molar ratio of Zn to Ga was varied to stoichiometric (black) and Zn excess (red), respectively. The effect of atmosphere (P$_{O_2}$) on the band-gap energy is shown in Fig. 2(b). The band-gap decreased from 4.90, 4.88 to 4.74 eV when the applied atmosphere for the solid state reaction was varied from N$_2$ (blue), Air (black), to O$_2$ (red).

In Fig. 3(a) and (b), the PL and PLE spectra of the non-doped Zn$_x$Ga$_2$O$_{3+x}$ are given. The PLE spectra in the non-d-Air (right and top, blue) and non-s-N$_2$ samples (right and bottom, blue) showed excitation bands maximizing at approximately 5.00 eV (248 nm) due to optical band-gap transitions. The excitation bands red-shifted to 4.96 eV (250 nm) in the non-s-Air sample (right, black), and further moved to 4.43 eV (280nm) in the non-e-Air sample (right and top, red) and 4.59 eV (270 nm) in the non-s-O$_2$ sample (bottom and right, red). The variation of PLE bands is consistent with the results in the absorption spectra (Fig. 2). The PL spectra (left part of Fig. 3) of the non-doped Zn$_x$Ga$_2$O$_{3+x}$ samples showed self-activated emissions in the UV and visible regions. The emission bands shifted towards lower energy in the sample with Zn excess (left and top, red) or in the one prepared in O$_2$ atmosphere (left and bottom, red) compared with other samples.

Schematic energy level diagrams of the non-doped Zn$_x$Ga$_2$O$_{3+x}$ samples are depicted in Fig. 4. The HRBE concept was used where the top of the VB was defined as zero of energy. According to the results of absorption, PLE and PL spectra, the energies of band-gap absorption and self-activated emission in the non-d-Air and non-s-N$_2$ samples were approximately 4.9 and 3.5 eV, respectively; the energies decreased and covered broader regions in 4.4~4.9 eV for band-gap absorption and 3.0~3.5 eV for self-activated emission. The decrease was attributed to lowering of the lowest component (bottom) of CB as shown in Fig. 4.
It is known that the ZnGa₂O₄ crystallizes in an ideal normal spinel under ambient pressure, i.e. Zn and Ga ions occupy tetrahedral and octahedral sites, respectively. However, real ZnGa₂O₄ crystals (by high-temperature sintering) should contain some degree of anti-site coordination, i.e., Zn²⁺ in octahedral sites (Zn′Ga) and Ga³⁺ in tetrahedral site (Ga₂Zn). These defects, as compared to other defects, require much lower formation energy (0.9 eV) according to Pandey et al.: The concentration of Cr³⁺ ions. As shown in Fig. 5 (a), when the absorption coefficient, can be used to quantify the Kubelka-Munk function, which is proportional to the abnormal state can be formed, as Peng’s model suggests, in a case when there is a driving force for the stabilization of the abnormal state and the total free energy is reduced. As the composition is varied from Zn excess to Zn deficiency, the absorption coefficient of Cr³⁺ decreases from the case when there is a driving force for the stabilization of the abnormal state and the total free energy is reduced. As the concentration of the two anti-site defects also should be considered that the formation of Zn⁺⁺ may be the reason of the CB lowering. Indeed, we cannot totally exclude the possibility connections between the observed properties and these defects have not been found.

4.2 Redox effects of Cr ions induced by compositional deviation from stoichiometry

The Cr-doped Zn₁Ga₂O₃₋ₓ crystals show broad absorption bands peaked at around 560, 420, and 280 nm due to ⁴A₂→⁴T₂, ⁴A₂→⁴T₁, and ⁴A₂→⁴T₁(⁴P) electronic transitions of Cr³⁺ ions. The Kubelka-Munk function, which is proportional to the absorption coefficient, can be used to quantify the concentration of Cr³⁺ ions. As shown in Fig. 5 (a), when the composition is varied from Zn excess to Zn deficiency, the absorption coefficient of Cr³⁺ decreases, indicating a reduction of Cr³⁺ concentrations in the sample with Zn deficiency. In Fig. 5 (b), the absorption coefficient of Cr³⁺ also decreases from the sample prepared in O₂ (Cr-s-O₂) to that prepared in N₂ (Cr-s-N₂).

Cr ions are expected to dominate in the Cr³⁺ valence state in the ZnGa₂O₄ spinel crystals as luminescent centers. However, abnormal state can be formed, as Peng’s model suggests, in a case when there is a driving force for the stabilization of the abnormal state and the total free energy is reduced. As discussed in Sec. 4.1, in the sample with composition of Zn deficiency or in the one prepared in N₂, the defect reaction of Eq. (6) is promoted. Notice that additional negative electrons (e⁻) are created in this reaction. These electrons can combine with Cr³⁺ ions incorporated in Ga³⁺ sites (Cr′Ga) and reduce them into Cr²⁺ (Cr′Ga), as shown in Eq. (7). The decrease of the Cr³⁺ absorption coefficient in the Cr-d-Air sample may be due to such “self-reduction effect” in the sample with Zn deficiency. For the sample with Zn excess or the one prepared in O₂, the defect reaction of Eq. (5) is promoted. Because of the resultant positive holes (h⁺), an oxidizing reaction in Eq. (8) can keep more Cr ions in the Cr³⁺ state.

\[ \text{Cr}^{3+} + e^- \rightarrow \text{Cr}^{2+} \]  

(7)

\[ \text{Cr}^{2+} + h^+ \rightarrow \text{Cr}^{3+} \]  

(8)

In order to support the argument of the self-redox effect, electron spin resonance (ESR) measurements were performed in the Cr-doped ZnₓGa₂O₃₋ₓ samples. The ESR spectra at X band frequency (9.10 GHz) showed a resonance absorption at approximate 0.32 mT at room temperature. This absorption was attributed to nearest-neighbour Cr³⁺ pairs in the spinel ZnₓGa₂O₃₋ₓ crystals. The ESR spectra in the Cr-e-Air and Cr-s-O₂ samples showed larger intensity of the resonance absorption, which confirmed higher concentration of Cr³⁺ ions in these two samples. It is note that the Cr²⁺ concentration should increase in the Cr-d-Air and Cr-s-N₂ samples when the Cr³⁺ concentration decreases due to the reduction effect. Unfortunately, quantitative evaluation of Cr²⁺ concentration in the Cr-doped ZnₓGa₂O₃₋ₓ samples was unavailable in the absorption, PL, PL spectra and ESR measurements.

![Fig. 5 Kubelka-Munk plot of the Cr-doped ZnₓGa₂O₃₋ₓ crystals. Electronic transitions from ground state (⁴A₂) to various excited states (⁴T₂, ⁴T₁, and ⁴T₁(⁴P)) of Cr³⁺ ion are notated.](image-url)
4.3 Variation of PL and persistent luminescence in the Cr-doped Zn$_x$Ga$_2$O$_{3+x}$ crystals

Under excitation by UV or visible light, the Cr-doped Zn$_x$Ga$_2$O$_{3+x}$ samples showed several emission peaks at around 700 nm due to $^2E \rightarrow ^4A_2$ transitions of Cr$^{3+}$. One single peak at 695 nm (N$_2$ line) was attributed to Cr$^{3+}$ in distorted octahedral sites; other peaks at 688 (R line), 679, 708 nm etc. (phonon side bands, PSB) were originated from Cr$^{3+}$ in ideal octahedral sites.$^{33-55}$

The PLE spectra of the Cr-doped Zn$_x$Ga$_2$O$_{3+x}$ samples are shown in Fig. 6. The bands at short wavelength (designated as band 0) were attributed to the band-gap absorption. This band shifts from 250 nm (4.96 eV) in the Cr-d-Air and Cr-s-N$_2$ samples to 280 nm (4.42 eV) in the Cr-e-Air and Cr-s-O$_2$ samples, which is consistent with those in the non-doped Zn$_x$Ga$_2$O$_{3+x}$ crystals (Fig. 3). Another band at around 340 nm (3.6 eV) (band 2) was attributed to an electronic transition from GS of Cr$^{3+}$, i.e. $^4A_2$ level to CB of the host. This band shifts to lower energy in the Cr-e-Air and Cr-s-O$_2$. Other bands at around 280 nm (4.43 eV) (band 1), 420 nm (2.95 eV) (band 3), and 560 nm (2.11 eV) (band 4) due to 3d-3d transitions from $^4A_2$ to $^4T_1(^4P)$, $^4T_1$, and $^4T_2$ of Cr$^{3+}$ were also detected in the PLE spectra. The energy of these 3d-3d transitions showing very weak dependence on the preparing condition (composition or atmosphere), are ignorable compared with the variation in $^4A_2 \rightarrow$ CB and VB $\rightarrow$ CB transition energies (band 0 and band 2).

The PersLE spectra of the Cr-doped Zn$_x$Ga$_2$O$_{3+x}$ samples in Fig. 7 showed broad bands from 400 nm (3.05 eV) to 270 nm (4.59 eV). These bands have two components: $^4A_2 \rightarrow$ CB (band 2) and $^4A_2 \rightarrow ^4T_1(^4P)$ transition (band 1). Different from the PLE spectra, the $^4A_2 \rightarrow$ CB transition (band 2) in PersLE spectra shows much higher intensity than the other bands. Also, a red-shift of the $^4A_2 \rightarrow$ CB transitions (band 2) in the Cr-e-Air and Cr-s-O$_2$ samples was observed as compared with the same transitions in the Cr-d-Air and Cr-s-N$_2$ samples. The bands 0, 3, and 4 present in the PLE spectra could be hardly detected in Fig. 7.

Based on the obtained results, the energy level diagram in Fig. 8 including CB, VB, and various energy levels of Cr$^{3+}$ is constructed. The $^4A_2 \rightarrow$ CB transition (band 2) provides the energy location of the $^4A_2$ level (GS of Cr$^{3+}$) below the bottom of the CB. The ES of Cr$^{3+}$ can then be located using the PLE spectra. For example, in the Cr-d-Air or Cr-s-N$_2$ samples (Fig. 8, left part), the energy difference between $^4A_2$ and the CB bottom is approximately 3.7 eV. The $^4T_1(^4P)$ level is 4.4 eV above the $^4A_2$ level and located inside the CB. On the other hand, the $^4T_1$ and $^4T_2$ levels are lower than CB. Consequently, electron transfer through the CB to the traps can be realized by excitation of $^4A_2 \rightarrow ^4T_1(^4P)$ (band 1) and $^4A_2 \rightarrow ^4T_2$ transitions (band 2), but fail by excitation of $^4A_2 \rightarrow ^4T_1$ (band 3) and $^4A_2 \rightarrow ^4T_2$ transitions (band 4). Also, direct excitation into CB ($^4A_2 \rightarrow$ CB $\rightarrow$ traps) was proved as a more effective route for electron trapping than $^4A_2 \rightarrow ^4T_1(^4P) \rightarrow$ CB $\rightarrow$ traps. A fast relaxation of $^4T_1(^4P)$ to $^4T_1$ may decrease the electron trapping efficiency in the latter case.
As shown in Fig. 4, when the molar ratio of Zn to Ga was varied from Zn deficiency to Zn excess or when the applied atmosphere was changed from N₂ to O₂, the CB bottom will be lowered by 0.3-0.4 eV. Based on the PersLE spectra the energy difference from ⁴A₂ to the CB bottom decreases by approximately 0.3 eV in the Cr-e-Air or Cr-s-O₂ samples. The energy of the ⁴A₂ level relative to VB top remains unchanged in different samples (Fig. 8).

TL glow curves of the Cr-doped ZnₓGa₂O₃₊ₓ samples are shown in Fig. 9. Again, the glow curve in the Cr-d-Air sample was similar to that in the Cr-s-N₂ sample. The glow curves show a band peaked at approximately 340 K with narrow full width at half maximum (FWHM = 70 K). The glow curves broaden in the Cr-e-Air and Cr-s-O₂ samples (FWHM = 180 K) and slightly shifted to lower temperature.

The peak temperature and FWHM of the TL glow curves reflect trap depth (thermal activation energy, ∆<i>e</i>) and the trap depth distribution. As shown in Fig. 8, the change in ∆<i>e</i> (decrease of the average value and broadening of the distribution) can be due to lowering of the CB bottom.

5 Conclusion

In the process of investigating optical properties in non-doped ZnxGa2O3+x (0.98 ≤ x ≤ 1.02) crystals, decrease of band-gap energy due to lowering of CB bottom was observed in the sample with 2 % Zn excess in composition and in the sample prepared in O₂ atmosphere. The similar lowering of CB bottom in these two samples may be due to significant formation of Zn⁺⁺⁺ anti-sites considering the most probable defect chemical reactions in composition of Zn excess or high partial oxygen pressure. The lowering of CB bottom resulted in decrease and broader distribution of trap depth. Slight Zn deficiency in the composition should be adopted to keep narrow distribution and trap depth for longer persistent luminescence decay. However, the sample with composition of Zn deficiency showed a self-

Fig. 9 TL glow curves of the Cr-doped ZnₓGa₂O₃₊ₓ samples with heating rate of 10 K/min. Excitation source (250-380 nm) is from a 10 W xenon lamp. Detecting wavelengths are fixed in red region by using a 600-715 nm band-pass filter. TL intensity of each sample was normalized to 1 at peak temperature.

Fig. 10 Persistent luminescence decay curves of the Cr⁻³⁺-doped ZnₓGa₂O₃₊ₓ crystals after excitation of UV light (250-380 nm) for 5 min. The intensity of persistent luminescence was converted to radiance (absolute value).
reduction effect on Cr ions into lower valence states. Effective measure should be taken to stabilize Cr ions in Cr\(^{3+}\) state.

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Notes and references