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# Tb<sup>3+</sup> and Ce<sup>3+</sup> as a Functional Couple for Enhanced Luminescence in YAG Ceramics for X-Ray Imaging and High-Power White LEDs and Laser Diodes

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## Abstract

The exposure of luminescent materials to X-ray radiation and high-power laser excitation results in significant heating, leading to thermal quenching and a corresponding reduction in phosphor efficiency. This study aimed to address this limitation by enhancing material efficiency while maintaining high thermal stability. YAG:Ce,Tb transparent ceramics were successfully fabricated via vacuum reactive sintering method, with moderate concentrations of Tb<sup>3+</sup> ions (1, 5 and 10 at.%) codopants. These ceramics exhibited a dense microstructure without micropores, achieving high transparency of 77–80% at 900 nm. The incorporation of Tb<sup>3+</sup> ions, is evidenced by the systematic increase in lattice constants from 12.608 Å (YAG:Ce) to 12.615 Å (YAG:Ce,Tb10%) in accordance with Vegard's law. The phonon-assisted Ce<sup>3+</sup>↔Tb<sup>3+</sup> bidirectional resonance mechanism, which facilitated energy transfer between Ce<sup>3+</sup> and Tb<sup>3+</sup> ions, was observed. The optimal transfer rate was observed at approximately 480 K. Beyond this temperature, the rate progressively accelerated, leading to accelerated decay times. Notably, YAG:Ce,Tb10% ceramics demonstrated a two-fold increase in radioluminescence intensity compared to uncoded YAG:Ce ceramics. This substantial improvement in luminescence performance highlights the potential of YAG:Ce,Tb ceramics as highly efficient phosphors for X-ray imaging applications, offering enhanced brightness and energy transfer efficiency under high-temperature operating conditions. The Tb<sup>3+</sup> codoping also enhanced Ce<sup>3+</sup> emission, allowing for tuning of the correlated color temperature to a maximum of 6013 K, while maintaining a stable color rendering index of 69–73 and luminous efficacy of ~123 lm/W. It was



shown that the luminous efficacy was not a constant parameter, but depended on the excitation laser power, initially increasing with power up to 0.5 W and then saturating. These results establish YAG:Ce,Tb transparent ceramics as highly promising materials for white LEDs due to their good thermal stability, tunable optical properties, and enhanced luminescence performance.

**Keywords:** YAG,  $\text{Ce}^{3+}$ ,  $\text{Tb}^{3+}$ , energy transfer, light yield, photoconversion, white LEDs, Laser diode, X-ray imaging

## 1. Introduction

Inorganic scintillators effectively convert high-energy radiation into ultraviolet-visible or visible light. The scintillators are integral to numerous applications, such as their use in electromagnetic calorimeters for high-energy physics, non-destructive industrial inspections, X-ray imaging systems, and time-of-flight positron emission tomography (TOF-PET) in medical diagnostics [1–4]. Advancing the application of these scintillators requires the development of materials with enhanced efficiency, reduced afterglow and low production costs. Increased efficiency allows for a greater photon yield per radiation dose, producing higher-quality images without necessitating additional exposure. Similarly, a reduced afterglow minimizes the temporal overlap of high-energy events and improves the signal-to-noise ratio, thereby enabling a quicker imaging process. These performance improvements are essential for optimizing both medical imaging and high-energy physics instrumentation [5]. High-resolution X-ray imaging demands a balance among spatial resolution, fine detail detection, sensitivity, and a strong signal-to-noise ratio [6]. The thickness of the scintillating material largely determines this balance, while thicker scintillators improve sensitivity by increasing X-ray attenuation and light yield, they may reduce spatial resolution due to light propagation and scattering. The scattering or diffusion of emitted light is a primary contributor to blurred X-ray images [7,8]. As a result, single crystal scintillators with low light scattering coefficients have become promising candidates for high-resolution applications. Their good light propagation characteristics reduce internal re-absorption and scattering, resulting in sharper, more detailed images.

The single crystals of  $\text{Ce}^{3+}$  doped  $\text{Y}_3\text{Al}_5\text{O}_{12}$  (YAG) exhibit excellent spatial resolution, approximately 34.7 line pairs per millimeter (lp/mm) [9]. The YAG crystalline structure exhibits excellent mechanical strength, allowing scintillation screens to be fabricated with thicknesses as low as 100  $\mu\text{m}$ . This characteristic is especially advantageous in medical imaging and optoelectronic devices, where space and weight are critical factors [10]. Although single crystals offer advantages, polycrystalline ceramics are often chosen as an alternative [11]. Because of their faster and significantly cheaper production, easier fabrication of complex shapes, and more uniform dopant distribution [11,12]. The energy transfer efficiency from the host matrix to  $\text{Ce}^{3+}$  ions can be significantly improved by incorporating  $\text{Gd}^{3+}$  and  $\text{Tb}^{3+}$  ions as sensitizers.  $\text{Tb}^{3+}$  is considered an effective co-activator for facilitating energy transfer due to its absorption levels, which lie between the absorption of the host lattice and the emission of  $\text{Ce}^{3+}$ . Additionally, the  $^5\text{D}_4 \rightarrow ^7\text{F}_6$  transition of



Tb<sup>3+</sup> overlaps with the 4f→5d<sub>1</sub> absorption transition of the Ce<sup>3+</sup> ion [13–15].

The Tb<sup>3+</sup> codoping in YAG:Ce is also relevant in the field of high-power lighting. YAG:Ce phosphor converters (pc) have demonstrated significant potential in various lighting applications, particularly in high-power lighting sources that require high thermal stability [16–18]. Phosphor in glass (PiG), single-crystal phosphors and ceramic phosphors are currently the most common phosphor conversion materials. Their thermal conductivity typically ranges between 0.1 and 30 W m<sup>-1</sup>K<sup>-1</sup>. Phosphor ceramics are preferable among them in terms of their overall photothermal characteristics [19]. The limitations of using such white LEDs/LDs for indoor lighting include a low color rendering index (CRI < 80) and a high correlated color temperature (CCT > 4000 K), which result from a lack of red components in the emission spectrum of YAG:Ce<sup>3+</sup>. For instance, it has been widely observed that the emission wavelength of Ce<sup>3+</sup> shifts to a longer wavelength, improving both the color CRI and CCT, when larger rare earth ions like Gd<sup>3+</sup> and Tb<sup>3+</sup> are used to substitute the dodecahedral site in YAG lattice [20,21]. In high-power LDs, maintaining the thermal stability of phosphor materials during operation remains a serious challenge. This instability arises from material heating under high energy excitation, which induces a thermal quenching process. To mitigate this, extensive research focuses on enhancing the phosphor's thermal stability by engineering its electronic structure. One effective strategy involves increasing the energy barrier between the conduction band minimum (CBM) and the lowest-energy 5d excited state (5d<sub>1</sub>) of Ce<sup>3+</sup> ions. In YAG:Ce, the partial substitution of Y<sup>3+</sup> ions by larger Tb<sup>3+</sup> ions enhances the crystal field strength, which increases the splitting of the 5d states of Ce<sup>3+</sup>. This enlarged splitting elevates the energy barrier between the CBM and the Ce<sup>3+</sup> 5d<sub>1</sub> excited state, thereby reducing the probability of the thermal ionization process and improving the phosphor thermal stability under high-power energy excitation. However, the pronounced splitting of the Ce<sup>3+</sup> 5d<sub>1</sub> excited state substantially diminishes the Stokes shift. This in turn facilitates a reabsorption process leading to the quenching of the emission. The compositional engineering strategy is well-established and widely understood. However, there is a notable lack of systematic research on Tb<sup>3+</sup>-codoped YAG:Ce systems with low Tb<sup>3+</sup> ions concentrations (≤10%). One of the key advantages of incorporating both Tb<sup>3+</sup> and Ce<sup>3+</sup> in a garnet host is the enhancement of luminescence intensity. This improvement arises from a synergistic energy transfer mechanism, which has the potential to significantly improve luminous efficacy (LE) in high-power LED applications [22]. Nevertheless, at high Tb<sup>3+</sup> concentrations, the substantial overlap between the excitation and emission bands of Ce<sup>3+</sup> ions leads to luminescence quenching, particularly at elevated temperatures. Consequently, a balanced substitution of Y<sup>3+</sup> by Tb<sup>3+</sup> is essential for optimizing the performance of garnet phosphors. Addressing this knowledge gap and aiming to optimize the Ce<sup>3+</sup>/Tb<sup>3+</sup> concentration ratio in YAG transparent ceramics, a series of transparent ceramics were prepared, including undoped and Ce<sup>3+</sup> (0.5 at.%) doped YAG as reference samples as well as Tb<sup>3+</sup>-codoped YAG:Ce 0.5 at.%, with Tb<sup>3+</sup> concentrations 1, 5 and 10 at.%. A systematic study was conducted on the influence of the Ce<sup>3+</sup>/Tb<sup>3+</sup> concentration ratio on microstructure, phase purity, absorption, transmittance, luminescence, and scintillation characteristics, along with the underlying bidirectional energy transfer mechanism between Ce<sup>3+</sup> and Tb<sup>3+</sup> ions. Moreover, the photoconversion

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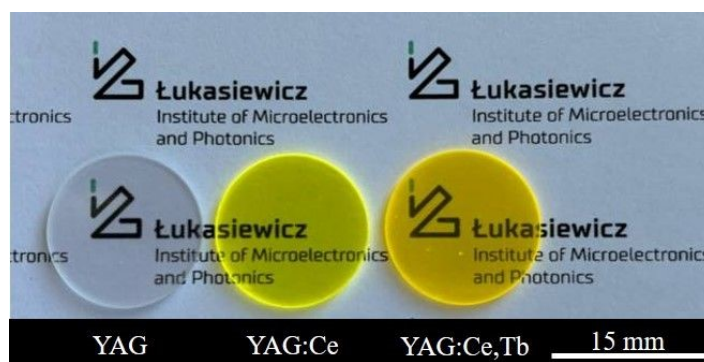


parameters of transparent ceramics were evaluated. Additionally, the thermal stability of  $\text{Ce}^{3+}$  emission and the efficiency of the bidirectional  $\text{Tb}^{3+} \leftrightarrow \text{Ce}^{3+}$  energy transfer process were analyzed over a temperature range from 77 to 767 K.

## 2. Experiment

### 2.1 Sample preparation

A set of YAG:Ce 0.5 at.%, Tb (0, 1, 5, 10 at.%) transparent ceramics was prepared. Initially, aqueous slurries of  $\text{Y}_2\text{O}_3$  (4N),  $\text{Al}_2\text{O}_3$  (4N),  $\text{CeO}_2$  (4N) and  $\text{Tb}_4\text{O}_7$  (4N) were prepared by blending the powders in a planetary mill with DolapixCE 64 dispersing agent, DuramaxB-1000 binder, and a laboratory-made octakis(tetramethylammonium)-T8-silsesquioxane hydrate (Octaanion) sintering aid. These homogeneous slurries were then sprayed into liquid nitrogen to form granulates, which were subsequently lyophilized for approximately 10-12 hours. The obtained granulates were pressed uniaxially in molds into disk shapes measuring 20 mm in diameter and approximately 4 mm in thickness. The samples underwent calcination in air at 1220 K to remove organic additives, followed by vacuum sintering at 1988 K for 6 hours. Subsequently, the samples were annealed in air at 1870 K for 2 hours to minimize oxygen-vacancy concentration. Finally, ceramics were cut and polished to dimensions of 4x4x1 mm<sup>3</sup>. Fig. 1 is the photograph of the as-sintered YAG, YAG:Ce, and YAG:Ce,Tb ceramics.



**Fig. 1.** Photograph of mirror-polished YAG, YAG:Ce, and YAG:Ce,Tb transparent ceramics with 1 mm thickness under daylight.

### 2.2 Methods

X-ray diffraction (XRD) was utilized for qualitative phase analysis of the ceramics. The measurements were conducted using a Rigaku SmartLab 3-kW X-ray diffractometer, equipped with a copper X-ray tube operating at 40 kV and 30 mA, along with a D/tex Ultra 250 solid-state detector. Phase analysis was conducted on bulk samples in continuous mode, employing Bragg-Brentano geometry ( $\theta/2\theta$  scan) across an angular range of  $5^\circ$ – $120^\circ$  ( $2\theta$ ) with a scan step of  $0.01^\circ$  and a speed of  $1.2^\circ/\text{min}$ . XRD data analysis was performed using Rigaku PDXL 2 software. To perform a microscopic examination, mirror-polished samples were thermally etched in air at 1870 K for 30 minutes to expose the grain boundaries. The ceramic microstructures were then analyzed using a



scanning electron microscope (SEM) (Carl Zeiss CrossBeam Workstation AURIGA, Oberkochen, Germany). Transmittance spectra were recorded with a Jasco V-730 UV-vis spectrometer in the 200–2000 nm spectral range at 300 K.

Photoluminescence excitation (PLE) and photoluminescence emission (PL) spectra were measured by a custom-made spectrofluorometer 5000M (Horiba Jobin Yvon, Wildwood, MA, USA) using a steady-state laser-driven xenon lamp (Energetic, a Hamamatsu Company) as the excitation source. The detection part of the setup involved a single-grating monochromator and a photon-counting detector TBX-04 (Hamamatsu). The PL and PLE spectra were corrected for the experimental distortion.

PL decay curves were measured by a custom-made spectrofluorometer 5000M (Horiba Jobin Yvon, Wildwood, MA, USA) under excitation with ns-pulsed nanoLED light source and operated in time-correlated single-photon counting mode (Horiba Scientific). The detection part of the setup involved a single-grating monochromator and a photon-counting detector TBX-04 (Hamamatsu). The convolution procedure was applied to the PL decay curves to determine true decay times (SpectraSolve software package, Ames Photonics). Variable temperatures were accomplished by liquid nitrogen bath cryostats (LakeShore Cryotronics, environment by Janis).

Light yield (LY) measurements were performed using a photomultiplier tube (PMT) under excitation with 662 keV  $\gamma$ -rays ( $^{137}\text{Cs}$  source). Each ceramic sample was coupled to a Hamamatsu R6231 PMT window with silicone grease and covered with several layers of Teflon tape to increase light collection. The pulse height spectra of  $\gamma$  rays were recorded with a PC-based multichannel analyzer (MCA) Tukan8k after pulse processing by a preamplifier and a spectroscopy amplifier set at 2  $\mu\text{s}$  shaping time constant. Scintillation decays were measured under excitation with 662 keV  $\gamma$  rays by coupling each ceramic sample to a Hamamatsu R7600U-200 PMT and recording the scintillation time profiles with a TDS3054B digital oscilloscope.

Radioluminescence (RL) spectra were measured by a custom-made spectrofluorometer 5000M (Horiba Jobin Yvon, Wildwood, MA, USA) using the Mo X-ray tube (40 kV, 15 mA, Seifert) as the excitation source. The setup's detection part involved a single-grating monochromator and a photon-counting detector TBX-04 (Hamamatsu). Measured RL spectra were corrected for the spectral dependence of detection sensitivity.

In the photoconversion experiments, ceramic samples with dimensions of  $4 \times 4 \times 1 \text{ mm}^3$  were placed on a collimator lens to ensure that the beam spot from the diode or laser illuminated the maximum area of the square sample. Measurements were performed in transmission mode. The 455 nm LED (FWHM = 21.6 nm) was used for excitation. The blue excitation optical power was maintained at 0.6 W. For the power-dependent experiment, a 445 nm laser diode (LD) from CNI Lasers was used as the excitation source.

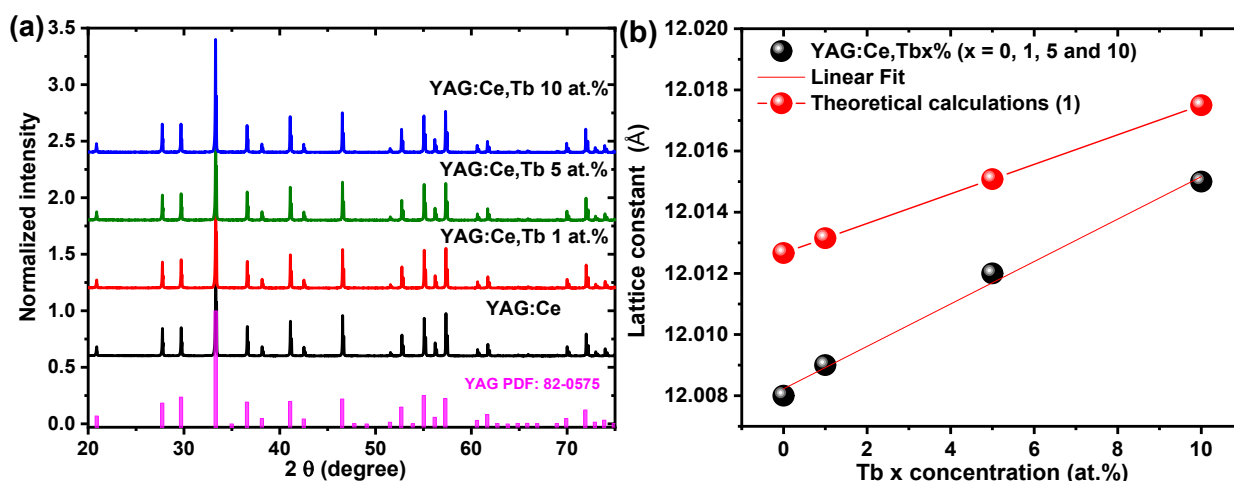
Measurements were conducted using a Gigahertz BTS-256LED spectrometer equipped with an integrating sphere for chromaticity parameters. The CIE 1931 coordinates ( $x$ ,  $y$ ), CRI and CCT were computed using the S-BTS256 software. All experiments were carried out at room temperature (RT).



### 3. Results and discussion

#### 3.1 Structural analysis and optical transmittance of YAG:Ce, Tb ceramics

The XRD patterns of YAG:Ce,Tbx% ( $x = 0, 1, 5$  and  $10$ ) transparent ceramics are shown in Fig. 2a. The crystal structure remains unchanged, validated through comparison with reference data: PDF No. 82-575. Analysis of the Rietveld refinement of XRD patterns indicates that all samples have been sintered in the pure cubic space group  $Ia-3d$  (No. 230), with no presence of secondary phases (Table 1).



**Fig. 2.** (a) XRD patterns of YAG:Ce, Tb $x$ % ( $x = 0, 1, 5$  and  $10$ ) transparent ceramics; (b) dependence of lattice constant on Tb concentration in YAG:Ce, Tb $x$ % ( $x = 0, 1, 5$  and  $10$ ) ceramics.

Table 1. Rietveld refinement of lattice parameters and volume of YAG:Ce, Tb $x$ % ( $x = 0, 1, 5$  and  $10$ ) transparent ceramics

Sample	Composition	a=b=c (Å)	V(Å <sup>3</sup> )	a (Å) calculated
YAG:Ce	Y <sub>2.985</sub> Ce <sub>0.015</sub> Al <sub>5</sub> O <sub>12</sub>	12.008	1731.4(1)	12.0127
YAG:Ce,Tb1%	Y <sub>2.955</sub> Ce <sub>0.015</sub> Tb <sub>0.03</sub> Al <sub>5</sub> O <sub>12</sub>	12.009	1731.7(1)	12.0131
YAG:Ce,Tb5%	Y <sub>2.835</sub> Ce <sub>0.015</sub> Tb <sub>0.15</sub> Al <sub>5</sub> O <sub>12</sub>	12.012	1732.98(1)	12.0151
YAG:Ce,Tb10%	Y <sub>2.685</sub> Ce <sub>0.015</sub> Tb <sub>0.3</sub> Al <sub>5</sub> O <sub>12</sub>	12.015	1734.6(1)	12.0175

As the concentration of Tb<sup>3+</sup> dopant increases in YAG:Ce, Tb $x$ % ( $x = 0, 1, 5$  and  $10$ ) ceramics, the lattice constant increases due to the well-known effect of displacement of smaller Y<sup>3+</sup> ions (ionic



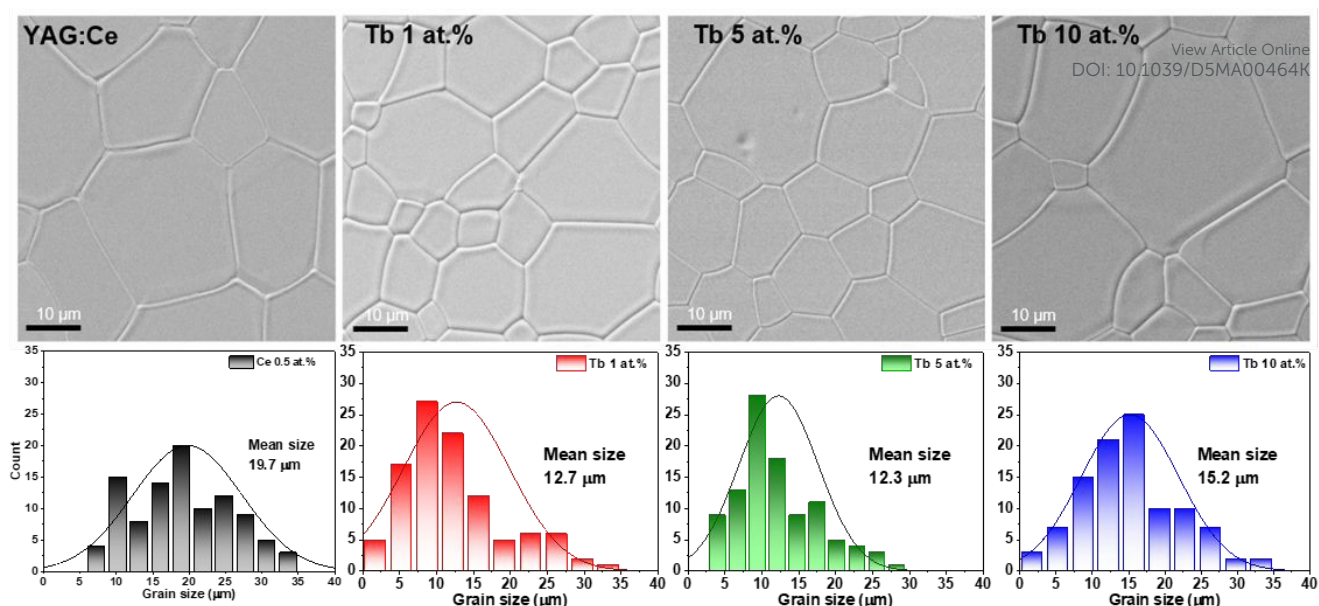
radius of  $Y_{VIII} = 1.019 \text{ \AA}$ ) by larger  $Tb^{3+}$  ions (ionic radius of  $Tb_{VIII} = 1.04 \text{ \AA}$ ) into the dodecahedral site in the garnet lattice [23]. Fig. 2b illustrates the experimental dependence of the lattice constant  $a_0$  on the increasing concentration of larger  $Tb^{3+}$  ions, in comparison with theoretical predictions. The unit cell parameters of garnets can be determined using analytical expressions derived from lattice geometry. These expressions account for the effective ionic radii of the constituent ions [24]

The lattice parameter of garnet structures can be described by the following analytical expression:

$$a = b_1 + b_2 r_c + b_3 r_a + b_5 r_c r_a + b_6 r_c r_d + b_4 r_d \quad (1)$$

where the coefficients are given as:  $b_1 = 7.02954$ ,  $b_2 = 3.31277$ ,  $b_3 = 2.49398$ ,  $b_4 = 3.34124$ ,  $b_5 = -0.087758$ , and  $b_6 = -1.38777$ . The variables  $r_c$ ,  $r_a$ , and  $r_d$  represent the effective ionic radii of cations occupying the respective sites in the garnet lattice ( $r_c(Tb^{3+}) = 1.04 \text{ \AA}$ ,  $r_a(Al^{3+}) = 0.55 \text{ \AA}$ , and  $r_d(Al^{3+}) = 0.39 \text{ \AA}$ ). The calculated values are summarized in Table 1. Fig. 3 shows SEM micrographs of thermally etched surfaces of YAG:Ce,Tbx% ( $x = 0, 1, 5$  and  $10$ ) ceramics. The ceramics synthesized at  $1988 \text{ K}$  exhibit a homogeneous microstructure, characterized by uniform and well-defined grain boundaries. Notably, no evidence of secondary phases or amorphous (glassy) regions is observed, indicating high phase purity and effective densification of the material. EDS analysis confirms a homogeneous elemental distribution both within the grains and at the grain boundaries. The grain size distribution and average grain size of the constituent phases in YAG:Ce,Tbx% ( $x = 0, 1, 5$  and  $10$ ) ceramics were evaluated using the linear intercept method applied to SEM images of the thermally etched surface. Substituting larger rare-earth ions (such as  $Nd^{3+}$ ,  $Tb^{3+}$ ,  $Yb^{3+}$ , or  $Gd^{3+}$ ) for  $Y^{3+}$  in the YAG lattice introduces lattice strain due to the size mismatch. This strain promotes the segregation of dopant ions to grain boundaries during sintering. Their presence at the boundaries impedes grain boundary motion through the solute drag effect, a mechanism where migrating boundaries must overcome the energy associated with solute-boundary interactions. As a result, grain growth is suppressed, leading to finer and more uniform grains, especially at higher dopant concentrations. However, at the highest  $Tb^{3+}$  concentration in YAG ceramics, the grain growth mechanism transitions from solute drag to more complex processes, which results in an increase of the grain size [27–29].





**Fig. 3.** SEM images and grain size distribution in YAG:Ce,Tbx% (x = 0, 1, 5 and 10) ceramics obtained by vacuum sintering at T = 1988 K.

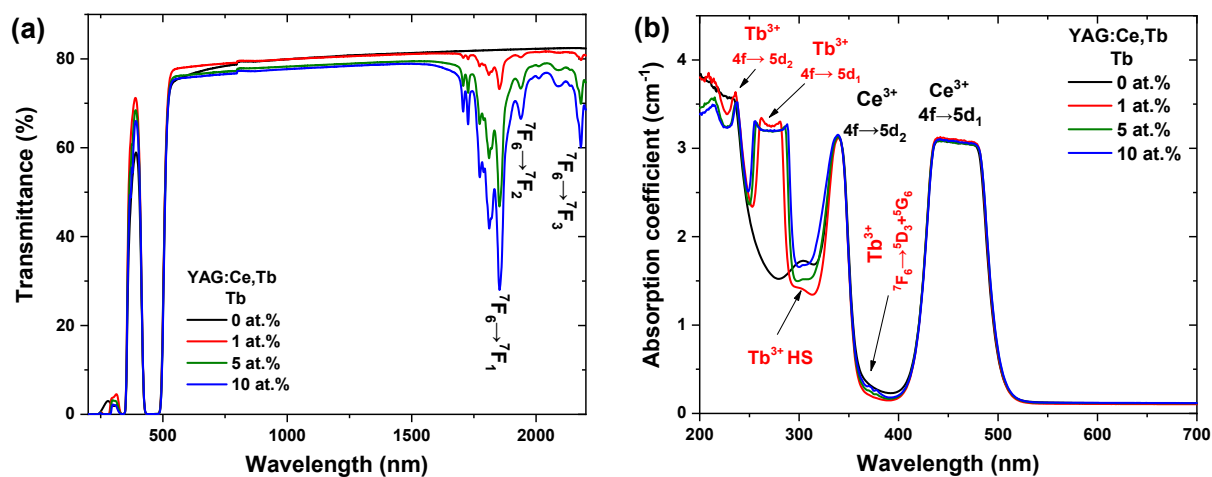
### 3.2 Transmittance properties

Fig. 4a presents the RT transmittance spectra of the synthesized ceramics. The Tb<sup>3+</sup>-doped YAG:Ce ceramics exhibit high optical transparency, with transmittance values ranging from 77% to 80% at 900 nm. In the infrared region, each transmission spectrum exhibits three prominent peaks at approximately 1853 nm ( ${}^7F_6 \rightarrow {}^7F_1$ ), 1953 nm ( ${}^7F_6 \rightarrow {}^7F_2$ ), and 2179 nm ( ${}^7F_6 \rightarrow {}^7F_3$ ). The intensity of these peaks increases with the rising concentration of Tb<sup>3+</sup> ions [30]. Detailed assignment for the absorption lines is listed in Table 2. Fig. 4b displays the RT absorption spectra of YAG:Ce,Tbx% (x = 0, 1, 5 and 10) transparent ceramics in the 200–700 nm range. The characteristic absorption bands of Ce<sup>3+</sup> ions in the YAG lattice are observed at 340 nm and 450 nm, corresponding to the  $4f \rightarrow 5d_2$  and  $4f \rightarrow 5d_1$  electronic transitions, respectively. The absorption observed in the ultraviolet region between 200 and 300 nm for YAG:Ce, can be primarily attributed to the charge-transfer transition from O<sup>2-</sup> to Ce<sup>4+</sup> [31]. This involves the transfer of electrons from the oxygen 2p to the cerium 4f energy levels. The presence of Ce<sup>4+</sup> in YAG:Ce ceramics sintered under the same conditions was previously confirmed by EPR measurements, and the Ce<sup>4+</sup>/Ce<sup>3+</sup> ratio was estimated to be 2.6 [31]. It is worth noting that the stabilization of Ce<sup>4+</sup> ions within the YAG host lattice is promoted by the formation of oxygen vacancies, which serve as a mechanism to maintain charge neutrality and structural integrity.

The higher-energy  $4f \rightarrow 5d_{3,4}$  transitions of Ce<sup>3+</sup> ions as well as structural defects further increase the absorption intensity in the UV spectral range [31]. In addition, the absorption spectra contain sets of bands belonging to the transitions of Tb<sup>3+</sup> ions. Namely, the bands centered around ~224 and ~270 nm belong to the low spin-allowed (LS)  $4f \rightarrow 5d_2$  and  $4f \rightarrow 5d_1$  absorption transitions of Tb<sup>3+</sup> ions. The spin-forbidden (HS)  $4f \rightarrow 5d_1$  transition within Tb<sup>3+</sup> ions is centered at 325 nm. The lines centered at



373 nm are attributed to the  ${}^7F_6 \rightarrow {}^5D_3 + {}^5G_6$  absorption transitions of  $Tb^{3+}$  cations [14].



**Fig. 4.** (a) Optical transmittances and (b) absorption spectra of YAG:Ce,Tbx% (x = 0, 1, 5 and 10) transparent ceramics.

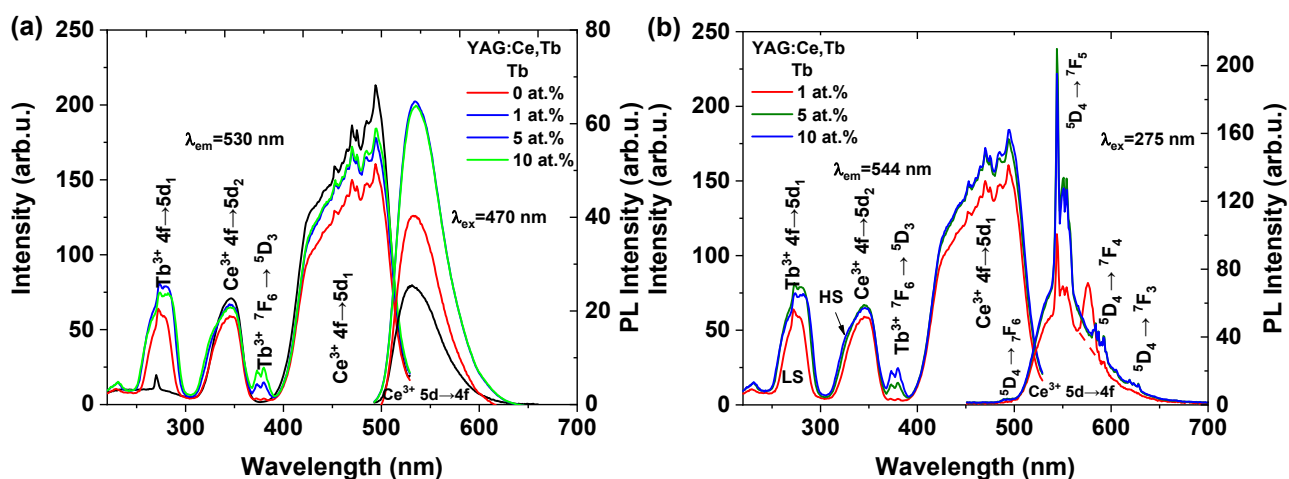
**Table 2.** Summary of the absorption transitions in the YAG:Ce,Tbx% (x = 0, 1, 5 and 10) ceramics.

Ion	Absorption transition	Band position (nm)
$Tb^{3+}$	$4f \rightarrow 5d_2$	224
	$4f \rightarrow 5d_1$	270
	${}^7F_6 \rightarrow {}^5D_3$	373
	${}^7F_6 \rightarrow {}^5G_6$	372, 378
	${}^7F_6 \rightarrow {}^7F_1$	1853
	${}^7F_6 \rightarrow {}^7F_2$	1939
	${}^7F_6 \rightarrow {}^7F_3$	2179
$Ce^{3+}$	$4f \rightarrow 5d_2$	340
	$4f \rightarrow 5d_1$	450

3.3 Photoluminescence properties

Fig. 5 displays the PLE and PL spectra of YAG:Ce,Tbx% (x = 0, 1, 5 and 10) ceramics. The PLE spectra recorded for  $Ce^{3+}$  emission at 530 nm (Fig. 5a) and  $Tb^{3+}$  emission at 544 nm (Fig. 5b) display distinct excitation bands attributable to both  $Ce^{3+}$  and  $Tb^{3+}$  ions, which are in agreement with the absorption spectra shown in Fig. 4b. The appearance of  $Tb^{3+}$  absorption bands within the  $Ce^{3+}$  excitation spectra and  $Ce^{3+}$  absorption bands within the  $Tb^{3+}$  excitation spectra provide evidence for the bidirectional energy transfer process between  $Tb^{3+}$  to  $Ce^{3+}$  ions. This observation underscores the role of  $Tb^{3+}$  ions in enhancing  $Ce^{3+}$  emission via a nonradiative transfer mechanism optimizing the luminescent efficiency of these materials [14,32–36]. The PL spectra of YAG:Ce,Tbx% (x = 0, 1, 5

and 10) excited into  $\text{Ce}^{3+}$  ions at 470 nm show a broad band peaking around 530 nm related to the  $5d_1 \rightarrow 4f$  ( $^2F_{5/2}$ ,  $^2F_{7/2}$ ) emission transition in the  $\text{Ce}^{3+}$ . The absence of  $\text{Tb}^{3+}$  emission lines in the spectra suggests that the energy states of  $\text{Ce}^{3+}$  ( $5d_1$ ) and  $\text{Tb}^{3+}$  ( $^5D_4$ ) are in optimal resonance. This resonance facilitates an efficient, bidirectional energy exchange mechanism between  $\text{Tb}^{3+}$  and  $\text{Ce}^{3+}$  ions. The process is ultimately dominated by  $\text{Ce}^{3+}$  emission, attributed to the allowed  $5d_1 \rightarrow 4f$  emission transition within  $\text{Ce}^{3+}$  ions [19,20,36].



**Fig. 5.** (a) PLE spectra for  $\text{Ce}^{3+}$  emission at 530 nm and PL spectra excited into  $\text{Ce}^{3+}$  ions at 470 nm in YAG:Ce,Tbx% ( $x = 0, 1, 5$  and  $10$ ) ceramics; (b) PLE spectra for  $\text{Tb}^{3+}$  emission at 544 nm and PL spectra excited into  $\text{Tb}^{3+}$  ions at 275 nm. The PL and PLE spectra of the singly doped  $\text{Ce}^{3+}$  and  $\text{Tb}^{3+}$  ceramics are presented in Fig. S1 in the Supplementary Information.

The PL spectra excited into  $\text{Tb}^{3+}$  ions at 275 nm show  $\text{Tb}^{3+}$  ions  $4f \rightarrow 4f$  emission lines at  $\sim 490$ ,  $\sim 544$ ,  $\sim 588$  and  $\sim 624$  nm corresponding to  $^5D_4 \rightarrow ^7F_{(6,5,4,3)}$  transitions and  $\text{Ce}^{3+}$  ions emission bands, which intensity increase with increasing of  $\text{Tb}^{3+}$  ions concentration. The relatively high  $\text{Tb}^{3+}$  ions concentration causes strong quenching emission from  $^5D_3$  energy state due to  $^5D_3$ – $^5D_4$  cross relaxation interaction [37]. The PL spectra excited at 275 nm, corresponding to the  $\text{Tb}^{3+}$   $4f \rightarrow 5d_1$  LS transition show characteristic  $\text{Tb}^{3+}$  emission lines associated with the  $^5D_4 \rightarrow ^7F_6$  ( $\sim 480$  nm),  $^5D_4 \rightarrow ^7F_5$  ( $\sim 550$  nm),  $^5D_4 \rightarrow ^7F_4$  ( $\sim 590$  nm),  $^5D_4 \rightarrow ^7F_3$  ( $\sim 625$  nm) and  $^5D_4 \rightarrow ^7F_2$  ( $\sim 680$  nm) transitions. Additionally, a broad emission band spanning from 500 to 700 nm is observed, originating from the  $5d_1 \rightarrow 4f$  transition of  $\text{Ce}^{3+}$  ions.

The variation in emission profiles depending on the excitation wavelength (i.e.,  $\text{Ce}^{3+}$  at 470 nm and  $\text{Tb}^{3+}$  at 275 nm) provides an insight into the efficiency of the energy transfer process between  $\text{Tb}^{3+}$  and  $\text{Ce}^{3+}$  ions. This finding highlights that the energy transfer efficiency is significantly influenced by the concentration of  $\text{Tb}^{3+}$  ions. Specifically, the bidirectional energy transfer between  $\text{Tb}^{3+}$  and  $\text{Ce}^{3+}$  is restricted to ions that are in close spatial proximity. In such cases, the excitation energy from  $\text{Tb}^{3+}$  is efficiently transferred to  $\text{Ce}^{3+}$ , where the emission is ultimately observed. Conversely, for  $\text{Tb}^{3+}$  ions that are not spatially coupled and thus not in resonance with  $\text{Ce}^{3+}$  ions, the excitation energy remains localized within the  $\text{Tb}^{3+}$  ions and terminates with  $\text{Tb}^{3+}$  emission. As a result, the restricted



spatial interaction ensures that excitation energy is either confined to  $\text{Tb}^{3+}$  ions (leading to  $\text{Tb}^{3+}$  emission) or partially transferred to  $\text{Ce}^{3+}$  ions that are in resonance, resulting in a combination of emissions from both  $\text{Tb}^{3+}$  and  $\text{Ce}^{3+}$  ions under excitation into  $\text{Tb}^{3+}$  excitation bands [19,20].

Fig. 6a presents the PL decay curves of YAG:Ce, Tb $x\%$  (Tb = 0, 1, 5, 10) ceramics ( $\lambda_{\text{ex}} = 450\text{nm}$ ,  $\lambda_{\text{em}} = 544\text{nm}$ ) measured at RT. For the samples without  $\text{Tb}^{3+}$  ions and with low  $x=1\%$  concentration a single exponential decay is observed with a decay time of  $\sim 70$  ns. With an increase in  $\text{Tb}^{3+}$  ions concentration up to 5%, the initial part of the decay curve exhibits a non-exponential profile, thus, the double-exponential fit is used for the qualitative description of two components:  $\tau_1 = 41$  ns and  $\tau_2 = 71$  ns.

$$I(t) = \sum I_i \exp[-t/\tau_i] + B, i = 1 \text{ to } 2, \quad (2)$$

the parameters  $I_i$  and  $\tau_i$  stand for the pre-exponential factor and decay time, respectively, in the  $i^{\text{th}}$  decay component.

Due to the double-exponential fitting of the decay curves, effective decay time,  $\tau_{\text{effective}}$ , is introduced for consideration, defined by the equation:

$$\tau_{\text{effective}} = \frac{\sum A_i \tau_i^2}{\sum A_i \tau_i} \quad (3)$$

where  $A_i$  is the amplitude and  $\tau_i$  is the decay time value of the  $i$ -th component from the fit.

The non-exponential character of the decay curves provides further evidence of the bidirectional energy transfer process between  $\text{Ce}^{3+}$  and  $\text{Tb}^{3+}$  ions, as depicted in the simplified energy transfer scheme (Fig. 7). The energy transfer dynamics between  $\text{Ce}^{3+}$  and  $\text{Tb}^{3+}$  ions exhibit a complex interplay, wherein the initial excitation energy absorbed by  $\text{Ce}^{3+}$  undergoes a bidirectional transfer mechanism. Specifically, while energy is efficiently transferred from  $\text{Ce}^{3+}$  to  $\text{Tb}^{3+}$  ions through resonant energy transfer, a significant back-transfer process to  $\text{Ce}^{3+}$  occurs due to the energy level match between both ions. This energy transfer cycle influences the decay kinetics of  $\text{Ce}^{3+}$ , significantly accelerating the fast decay components while decelerating the slow components. Consequently, this dynamic interaction significantly suppresses the emission intensity of  $\text{Tb}^{3+}$  ions at higher concentrations. Such behavior underscores the intricate balance of energy transfer processes between  $\text{Ce}^{3+}$  and  $\text{Tb}^{3+}$ , highlighting the cooperative effects that govern their luminescent properties in  $\text{Tb}^{3+}$  ion concentrated systems. For further analysis of the  $\text{Tb}^{3+} \rightarrow \text{Ce}^{3+}$  energy transfer process, the PL decay measurements were performed for  $\text{Tb}^{3+}$  ( $^5\text{D}_4 \rightarrow ^7\text{F}_5$ ) emission measured at  $\lambda_{\text{em}} = 544$  nm under excitation into  $4\text{f} \rightarrow 5\text{d}_1$  transition of  $\text{Tb}^{3+}$  at 281 nm (Fig. 6b). The decay time of  $\text{Tb}^{3+}$  ions in YAG:Tb5% demonstrate single-exponential decay kinetics with decay time around  $3\mu\text{s}$  typical for  $\text{Tb}^{3+}$  ions in YAG host [38]. The PL decay characteristics of  $\text{Tb}^{3+}$  ions emission at 544 nm in  $\text{Tb}^{3+}, \text{Ce}^{3+}$  co-doped ceramics exhibit significant acceleration, demonstrating a multi-exponential behavior. Analysis of YAG:Ce0.5%, Tb $x\%$  ( $x = 0, 1, 5, 10$ ) ceramics reveals that the



decay time systematically decreases with increasing  $\text{Tb}^{3+}$  ion concentration, providing evidence for concentration-dependent energy transfer mechanisms between  $\text{Tb}^{3+}$  and  $\text{Ce}^{3+}$  ions. The higher  $\text{Tb}^{3+}$  ions concentration enhances the probability of resonant energy transfer as the spatial proximity between  $\text{Tb}^{3+}$  and  $\text{Ce}^{3+}$  ions decreases at higher  $\text{Tb}^{3+}$  concentrations. The observed correlation between dopant concentration and energy transfer efficiency suggests a direct relationship between the local ion density and the quantum mechanical coupling strength facilitating the transfer process. Quantitative analysis of the decay time characteristics is comprehensively presented in Table 3.

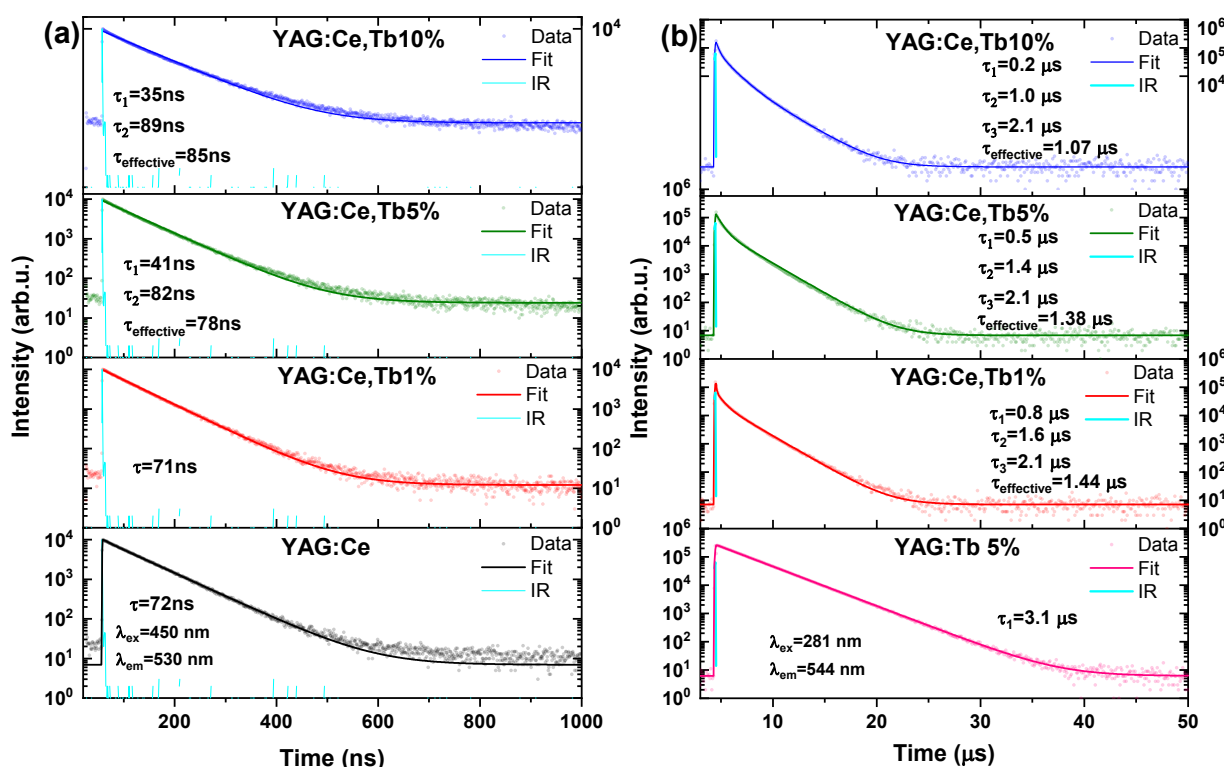
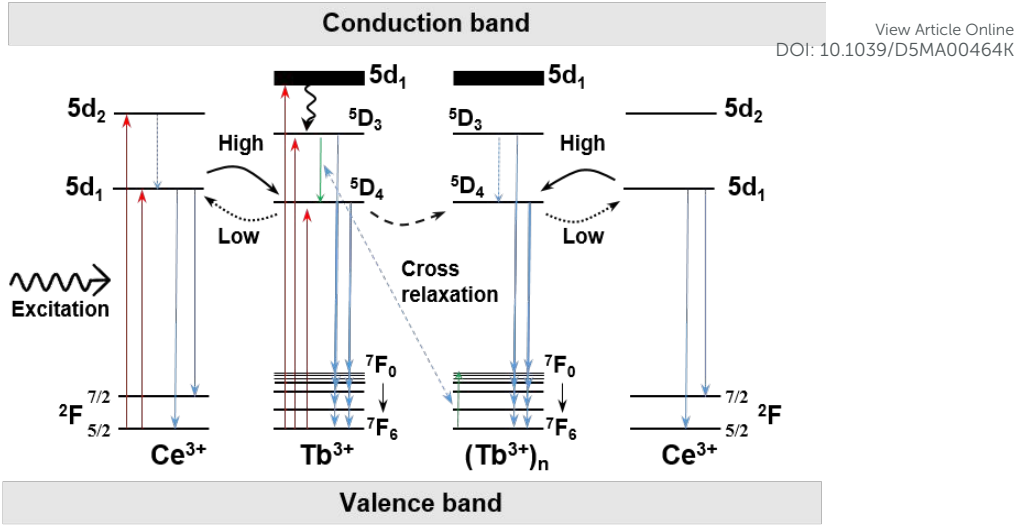


Fig.

6. (a) Prompt  $5d_1 \rightarrow 4f$  decay curves of the  $\text{Ce}^{3+}$  luminescence measured at  $\lambda_{\text{em}}=530\text{ nm}$  and  $\lambda_{\text{ex}}=450\text{ nm}$ ; (b) decay curves of  $\text{Tb}^{3+}$  ( $^5D_4 \rightarrow ^7F_5$ ) emission measured at  $\lambda_{\text{em}}=544\text{ nm}$  and  $\lambda_{\text{ex}}=281\text{ nm}$  for YAG:Ce,Tbx% ( $x = 0, 1, 5$  and  $10$ ) ceramics samples.





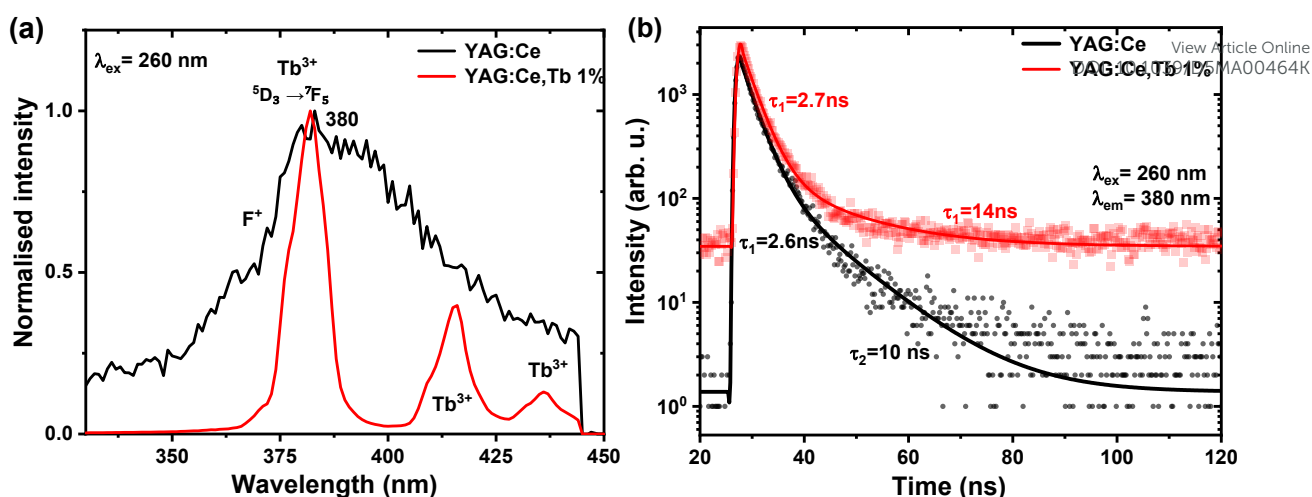
**Fig. 7.** Simplified energy diagram for Tb<sup>3+</sup> and Ce<sup>3+</sup> ions in Y<sub>3</sub>Al<sub>5</sub>O<sub>12</sub>. The red arrows indicate the absorption of photons, leading to excitation to higher energy levels. The black wavy arrow illustrates the relaxation process, while the black curved dotted arrows depict energy transfer processes. The blue dash arrows represent the cross-relaxation between neighboring Tb<sup>3+</sup> ions, and the blue arrows signify radiative de-excitation mechanisms (emission of photons).

**Table 3.** PL decay time values for Ce<sup>3+</sup> and Tb<sup>3+</sup> emissions.

Sample	Ce <sup>3+</sup> ions decay		Tb <sup>3+</sup> ions decay		
	(λ <sub>ex</sub> =450nm, λ <sub>em</sub> =530nm)		(λ <sub>ex</sub> =281nm, λ <sub>em</sub> =544nm)		
	τ <sub>1</sub> ns (I%)	τ <sub>2</sub> ns (I%)	τ <sub>1</sub> μs	τ <sub>2</sub> μs	τ <sub>3</sub> μs
YAG: Ce	72 (100)	n/a	n/a	n/a	n/a
YAG: Tb 5%	n/a	n/a	3.1	n/a	n/a
YAG: Ce,Tb 1%	71 (100)	n/a	0.8	1.6	2.1
YAG: Ce,Tb 5%	41 (17)	82 (83)	0.5	1.4	2.1
YAG: Ce,Tb 10%	35 (16)	89 (86)	0.2	1	2.1

Oxygen vacancies play a key role in stabilizing Ce<sup>4+</sup> ions as previously discussed in Section 3.2. The trapped electron in an F<sup>+</sup> center creates localized electronic states within the bandgap of the YAG structure [39]. Fig. 8a shows the normalized emission spectra of the F<sup>+</sup> center with a maximum at around 380 nm to demonstrate the low intensity of its emission, compared to the Tb-codoped sample, where F<sup>+</sup> emission is strongly overlapped by a dominant peak centered at 382 nm, attributed to the <sup>5</sup>D<sub>3</sub> → <sup>7</sup>F<sub>5</sub> transition of Tb<sup>3+</sup> ions [40,41].





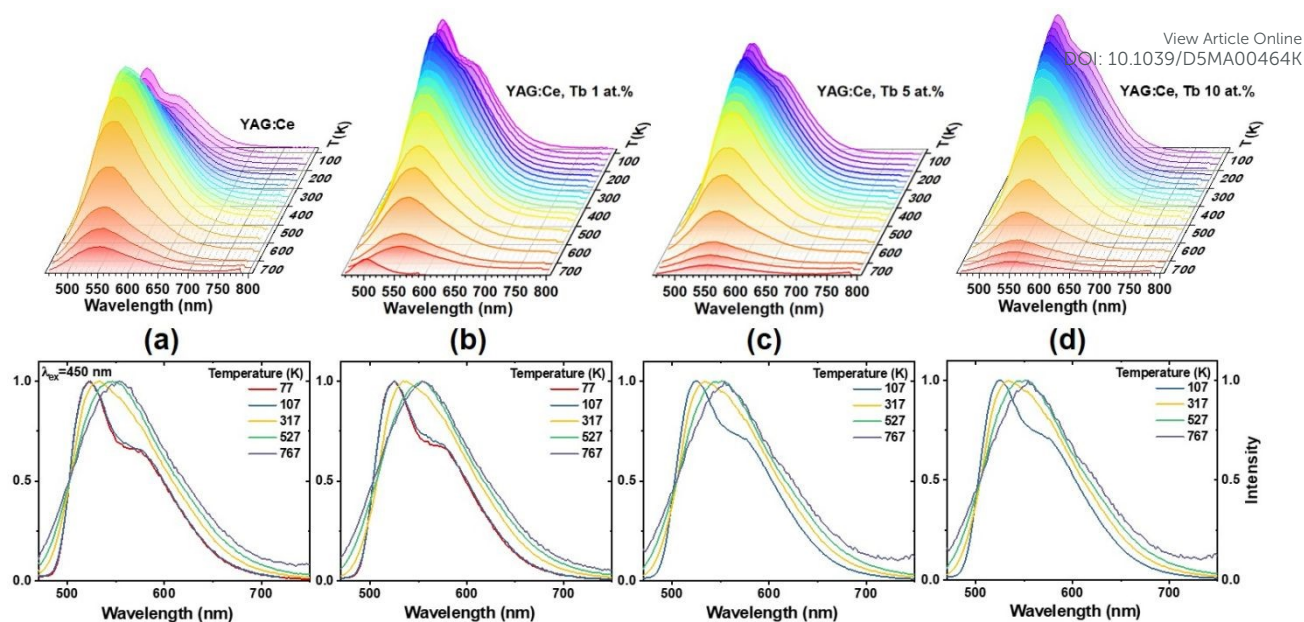
**Fig. 8.** (a) RT photoluminescence spectra of  $F^+$  center excited at 260 nm, and (b) photoluminescence decay kinetics of  $F^+$  center at 380 nm excited at 260 nm.

Fig. 8b demonstrates the decay curves of the  $F^+$  emission ( $\lambda_{ex} = 260$  nm,  $\lambda_{em} = 380$  nm) measured at RT and fitted using a double-exponential function as described in Eq. (2). The primary decay component, corresponding to the allowed  $1B \rightarrow 1A$  singlet-singlet transition of  $F^+$  centers, has a decay time of approximately 2 ns in the YAG:Ce host [42]. The presence of a slower component ( $\tau_2 = 10$  ns) in the YAG:Ce sample indicates disturbances or the aggregation of  $F^+$  centers as well as the energy transfer process to the surrounding  $Ce^{3+}$  ions, which can absorb ( $4f \rightarrow 5d_2$  transition) the  $F^+$ -center-related emission near the short-wavelength side. The slower component in the  $Tb^{3+}$ -codoped sample decelerated up to 14 ns, which can show complex interactions between oxygen vacancy and  $5D_3$  energy state of  $Tb^{3+}$  ions.

### 3.4 Temperature-dependent characteristics

Fig. 9a–d provides a detailed analysis of the temperature-dependent  $Ce^{3+}$  emission spectra recorded under 450 nm excitation corresponding to the  $4f \rightarrow 5d_1$  absorption band. The spectra were measured over a temperature range (77–767 K) for YAG:Ce,Tbx% ( $x = 0, 1, 5$  and 10) ceramics. In YAG:Ce ceramics, a significant increase in emission intensity is observed over the temperature range of 77–400 K. However, in  $Tb^{3+}$ -codoped samples, this enhancement occurs only between 77 and 300 K and remains negligible, see Fig. 10. To eliminate potential interference from thermoluminescence (TL) effects and ensure that the observed intensity variations result solely from intrinsic temperature-dependent phenomena, the PL spectra were measured in a high-to-low temperature sequence (767 K to 77 K). This approach ensures the reliability of the thermal behavior analysis of  $Ce^{3+}$  emission in YAG:Ce,Tb ceramics [43].



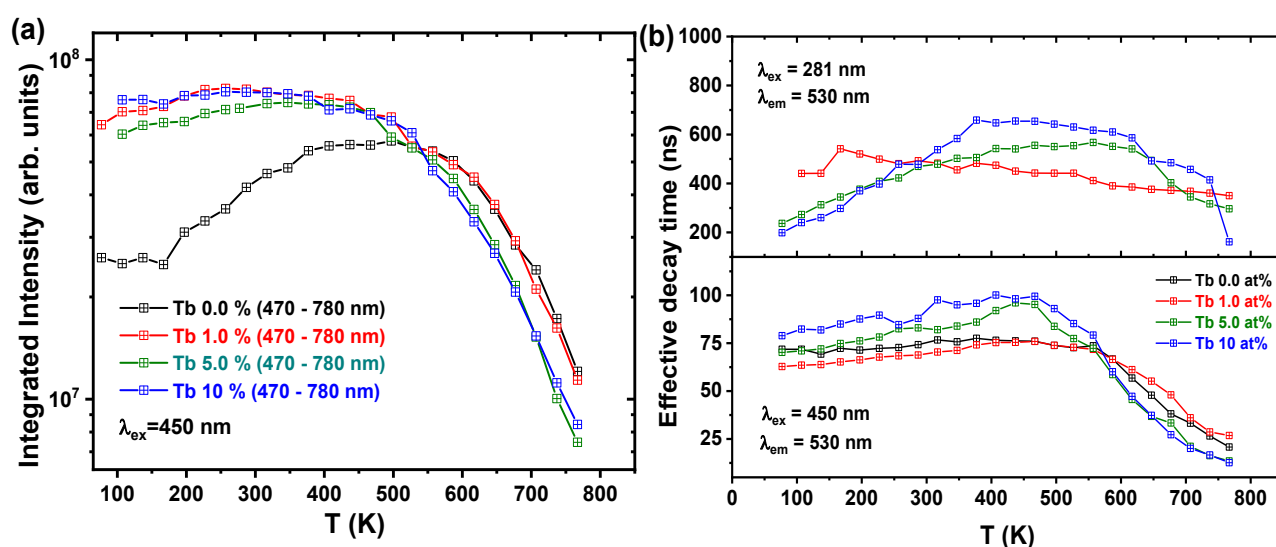


**Fig. 9.** PL spectra of  $\text{Ce}^{3+}$  ions in  $\text{YAG:Ce,Tbx}\%$  ( $x = 0, 1, 5$  and  $10$ ) ceramic samples (a-d) excited at  $450\text{ nm}$  and recorded in the  $77\text{--}767\text{ K}$  temperature range.

The mechanism underlying the progressive increase in the emission intensity with increasing temperature can be attributed to phonon-assisted nonradiative energy transfer (PA-NET). This process occurs between two distinct  $\text{Ce}^{3+}$  centers, i.e. the donor sites, comprising distorted  $\text{Ce}^{3+}$  centers (characterized by interactions with antisite defects, interstitial non-stoichiometric yttrium, lattice vacancies, and structural distortions at grain boundaries, resulting in lowered  $5d_1$  state energy), and the acceptor sites, consisting of regular  $\text{Ce}^{3+}$  centers (occupying dodecahedral c-sites with  $D_2$  symmetry in the crystal lattice) [31,44]. This phenomenon is particularly pronounced in transparent ceramics, where the material exhibits challenges associated with  $\text{Ce}^{3+}$  ion segregation and agglomeration at grain boundary interfaces. The inhomogeneous distribution of  $\text{Ce}^{3+}$  ions, primarily manifested through segregation processes during sintering, significantly impacts the optical and luminescent properties as it can lead to localized variations in activator concentration [31]. Therefore, in transparent ceramics, the degree of disorder among  $\text{Ce}^{3+}$  ions is significant, resulting in inhomogeneously broadened emission and excitation bands. This indicates that  $\text{Ce}^{3+}$  ions are not uniformly excited within the transparent ceramic matrix, which in turn leads to reduced emission intensity at lower temperatures. As temperature rises, the excitation and emission bands broaden (see Fig. 9a-b, lower panels) [45–47], allowing both regular and disordered  $\text{Ce}^{3+}$  ions to be more efficiently excited. Furthermore, the increased temperature enhances the energy of the host lattice phonons, facilitating the bridging of the energy mismatch between regular and disordered  $\text{Ce}^{3+}$  ions, thereby enabling the PA-NET process. A detailed explanation of this mechanism can be found in [44]. The occurrence of the PA-NET process is further supported by a minor increase in emission



intensity with elevated temperatures and the presence of  $\text{Tb}^{3+}$  ions. The incorporation of  $\text{Tb}^{3+}$  ions significantly enhances crystal field strength which induces a downward shift in the energetic position of the  $5d_1$  electronic state of  $\text{Ce}^{3+}$  ions in their regular lattice sites. This modification facilitates the energetic convergence between  $\text{Ce}^{3+}$  ions occupying both regular and disordered crystallographic positions. As a consequence, the excitation efficiency of  $\text{Ce}^{3+}$  ions exhibits enhanced performance at reduced temperatures, requiring lower-energy phonons to facilitate the PA-NET process. The enhanced emission intensity observed in  $\text{Tb}^{3+}$ -codoped YAG:Ce ceramics, relative to their undoped YAG:Ce counterpart, can be attributed to the bidirectional energy transfer processes between  $\text{Ce}^{3+}$  and  $\text{Tb}^{3+}$  ions (Fig. 10a). The excitation of  $\text{Ce}^{3+}$  ions initiates an energy transfer cascade wherein the excitation energy preferentially migrates to neighboring  $\text{Tb}^{3+}$  ions through resonant coupling mechanisms. This preferential pathway demonstrates higher efficiency compared to alternative routes involving  $\text{Ce}^{3+}$ - $\text{Ce}^{3+}$  pair interactions at ceramic grain boundaries or energy dissipation through structural defect sites, where non-radiative decay processes dominate. The energy transfer dynamics are governed by a bidirectional, non-radiative interaction between  $\text{Tb}^{3+}$  and  $\text{Ce}^{3+}$  ions, most plausibly mediated by dipole-dipole and higher-order multipolar coupling mechanisms. Upon excitation, energy can be transferred from the  $^5\text{D}_4$  excited state of  $\text{Tb}^{3+}$  to the  $5d$  levels of  $\text{Ce}^{3+}$ , resulting in enhanced  $\text{Ce}^{3+}$  emission intensity. Under specific excitation conditions and particularly at elevated  $\text{Tb}^{3+}$  concentrations, a reverse energy transfer process from excited  $\text{Ce}^{3+}$  back to  $\text{Tb}^{3+}$  is also observed. This bidirectional transfer behavior leads to a dynamic redistribution of excitation energy within the host lattice, contributing to improved excitation channel efficiency and enhanced overall luminescence performance. This interpretation is consistent with previous experimental studies and theoretical models describing  $\text{Ce}^{3+} \leftrightarrow \text{Tb}^{3+}$  energy transfer mechanism. [35,36].



**Fig. 10.** (a) temperature dependence of the integrated  $\text{Ce}^{3+}$  emission intensity under excitation into  $4f \rightarrow 5d_1$  absorption transition at 450 nm in the temperature range 77-767 K; (b) temperature



dependence of photoluminescence decay times for the  $\text{Ce}^{3+}$  emission at 530 nm under excitation into  $4f \rightarrow 5d_1$  absorption transition of  $\text{Tb}^{3+}$  at 281 nm (upper panel) and excitation into  $4f \rightarrow 5d_1$  absorption transition at 450 nm (lower panel).

The onset of thermal quenching progressively shifts toward lower temperatures, decreasing from 660 K in YAG:Ce ceramics to 600 K in YAG:Ce,Tb10% ceramics. Notably, even with 10%  $\text{Tb}^{3+}$  ions, the ceramics retain good thermal stability. This luminescence quenching is primarily attributed to a thermally induced crossover from the excited-state  $5d_1$  potential energy to the ground-state  $4f$  potential energy, which facilitates nonradiative relaxation processes [31,48].

Fig. 10b shows the temperature dependence of the decay times of  $\text{Ce}^{3+}$  emission at 530 nm under excitation into the  $4f \rightarrow 5d_1$  transition of  $\text{Ce}^{3+}$  at 450 nm. For the undoped sample and the YAG:Ce,Tb1% sample, the effective decay time of  $\text{Ce}^{3+}$  emission remains relatively constant around 60 ns up to 480 K. Above this temperature, the decay rate accelerates significantly due to thermal quenching. As the temperature and  $\text{Tb}^{3+}$  ions concentration increase, the effective decay time of  $\text{Ce}^{3+}$  ions emission first increases, but after reaching approximately 380 K, a noticeable acceleration is observed. This temperature-dependent behavior varies between lightly and moderately  $\text{Tb}^{3+}$ -codoped samples and can be attributed to bidirectional energy transfer efficiency between coupled  $\text{Tb}^{3+}$  and  $\text{Ce}^{3+}$  ions, which becomes increasingly efficient with rising  $\text{Tb}^{3+}$  concentration. This effect likely influences the luminescence properties by facilitating nonradiative energy transfer processes within the material. As temperature increases, phonon interactions effectively bridge the energy gap between the  $\text{Ce}^{3+}$  ( $5d_1$ ) and  $\text{Tb}^{3+}$  ( $^5D_4$ ) states, bringing them into resonance and thereby facilitating a bidirectional energy transfer process. Consequently, higher temperatures promote the formation of  $\text{Ce}^{3+}$ - $\text{Tb}^{3+}$  pairs, which leads to an extended effective decay time. Notably, this phonon-assisted resonance achieves optimal efficiency at temperatures above 480 K, further enhancing the overall energy transfer rate. The temperature dependence of photoluminescence effective decay times of  $\text{Ce}^{3+}$  ( $5d_1 \rightarrow 4f$ ) emissions at 530 nm under excitation into  $4f \rightarrow 5d_1$  absorption band of  $\text{Tb}^{3+}$  at 281 nm in YAG:Ce,Tbx% ( $x = 5$  and 10) samples are shown in Fig. 10b. The observed differences in the effective decay time of  $\text{Ce}^{3+}$  emission at 530 nm under excitation at 450 nm ( $\text{Ce}^{3+}$  excitation) and 281 nm ( $\text{Tb}^{3+}$  excitation) result from variations in the rate and extent of energy migration between  $\text{Tb}^{3+}$  ions. In the first case, when  $\text{Ce}^{3+}$  ions are directly excited at 450 nm, the excitation energy is efficiently transferred between  $\text{Ce}^{3+}$  and  $\text{Tb}^{3+}$  ions, which are already in resonance. As a result, this energy transfer process is highly efficient and occurs rapidly. In contrast, when excitation occurs at 281 nm ( $\text{Tb}^{3+}$  excitation), the energy initially migrates among  $\text{Tb}^{3+}$  ions until it reaches a  $\text{Tb}^{3+}$  ion that is in resonance with a  $\text{Ce}^{3+}$  ion, allowing for the final energy transfer step. In other words, the  $\text{Tb}^{3+}$  sublattice actively seeks a suitable  $\text{Tb}^{3+}$  ion that is in resonance with  $\text{Ce}^{3+}$  ions, making the energy transfer process slower. Thus, this mechanism requires additional time due to the intermediate energy migration among  $\text{Tb}^{3+}$  ions before reaching a  $\text{Ce}^{3+}$  ion capable of accepting the excitation energy.

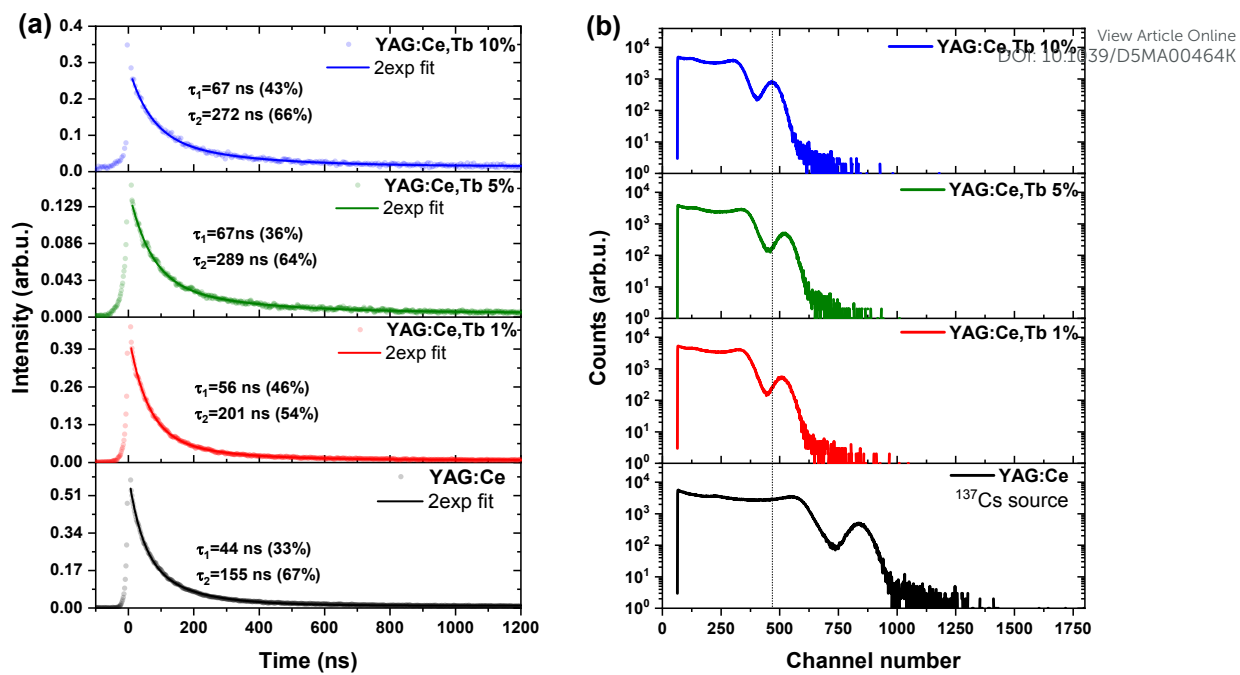


### 3.5 Scintillation properties

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Fig. 11a presents the scintillation decay time recorded under  $\gamma$ -ray excitation (662 keV,  $^{137}\text{Cs}$  radioisotope source). The decay times exhibit two exponential components: a fast component ( $\tau_1$ ) and a slow component ( $\tau_2$ ), both demonstrating strong dependence on  $\text{Tb}^{3+}$  ion concentration, see Table 3. In the  $\text{Tb}^{3+}$ -free YAG:Ce sample, the fast decay component ( $\tau_1 = 44$  ns) exhibits a slightly shorter decay time compared to the characteristic emission of isolated  $\text{Ce}^{3+}$  ions in garnet structures (typically 50-60 ns) [49]. This observation aligns with the photoluminescence data, providing further evidence for  $\text{Ce}^{3+}$  emission quenching in  $\text{Ce}^{3+}$ - $\text{Ce}^{3+}$  clusters and through interactions with structural defects (refer to Fig. 6 and associated discussion). The slow decay component ( $\tau_2 = 155$  ns) is attributed to the interaction of charge carriers (free holes and electrons) with intrinsic lattice defects, predominantly  $Y_{Al}$  antisite defects [50], which significantly delay the scintillation response time. The YAG:Ce,Tb1% sample exhibits a fast scintillation decay component of  $\tau_1 = 56$  ns, while YAG:Ce,Tbx% ( $x = 5$  and  $10$ ) samples demonstrate fast decay components of  $\tau_1 = 67$  ns. These decay time values are notably shorter compared to previously reported values of 78-80 ns in YAG:Ce 0.5% ceramics [12,51,52]. This improvement can be attributed to the more homogeneous spatial distribution of  $\text{Ce}^{3+}$  ions and reduced  $\text{Ce}^{3+}$ - $\text{Ce}^{3+}$  clustering at grain boundaries. Additionally, the acceleration of decay kinetics may be facilitated by bidirectional  $\text{Tb}^{3+}$ -to- $\text{Ce}^{3+}$  energy transfer process cycles. The dynamics of the slow scintillation decay component ( $\tau_2$ ) exhibit strong concentration dependence on  $\text{Tb}^{3+}$  content, as detailed in Table 4. This conclusion is further supported by pulse-height spectra analysis, see Fig. 11b and Table 4. The observed decrease in light yield (LY) with increasing  $\text{Tb}^{3+}$  concentration suggests that a portion of the excitation energy involved in the bidirectional  $\text{Tb}^{3+}$ -to- $\text{Ce}^{3+}$  energy transfer cycles undergoes radiative recombination via  $\text{Ce}^{3+}$  centers beyond the 2  $\mu\text{s}$  time gate used in LY measurements. Additionally, some of the charged carriers (electrons and holes) may undergo radiative recombination with  $\text{Tb}^{4+}$  ions beyond the 2  $\mu\text{s}$  time gate used for LY measurements. This process suggests that the recombination dynamics contribute to the overall scintillation response and may play a role in influencing the observed LY characteristics of the material.





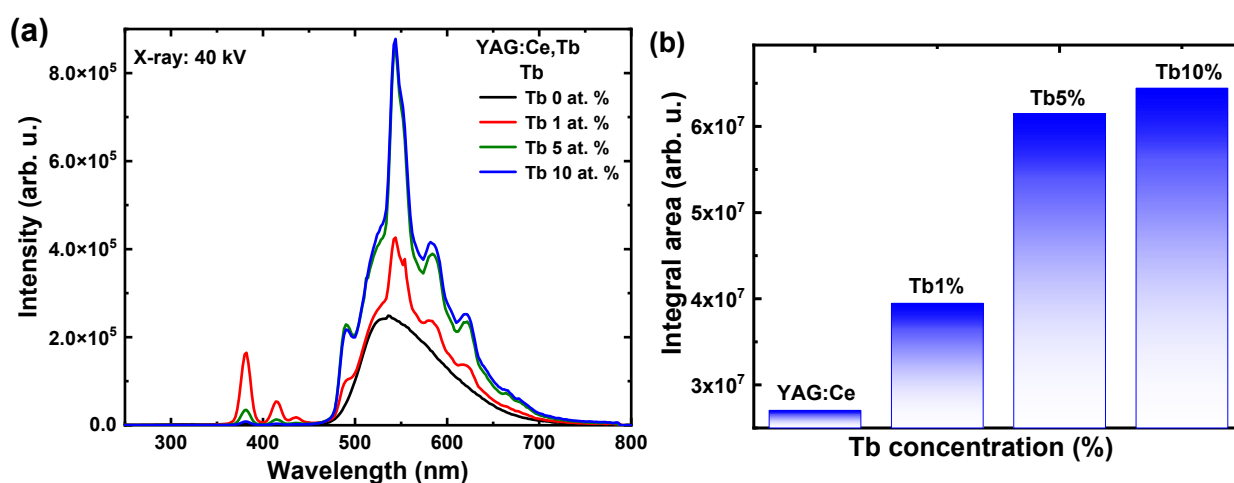
**Fig. 11.** (a) Scintillation decay times measured within 1  $\mu\text{s}$  range and (b) pulse-height spectra measured with 2  $\mu\text{s}$  shaping time under excitation with 662 keV  $\gamma$ -rays from a  $^{137}\text{Cs}$  source for YAG:Ce,Tbx% (x = 0, 1, 5 and 10) ceramics.

**Table 4.** Scintillation parameters of YAG:Ce,Tbx% (x = 0, 1, 5 and 10) ceramics (2  $\mu\text{s}$  shaping time). LY- light yield, R- energy resolution.

Sample	LY (ph/MeV)	%R	$\tau_1$ ns (I%)	$\tau_2$ ns (I%)	$\tau_{\text{effective}}$
YAG: Ce	10200	10.6%	44.6 (33%)	155.7 (67%)	119 ns
YAG: Tb 5%	6250	13.1%	56.2 (46%)	201.6 (54%)	135 ns
YAG: Ce,Tb 1%	6370	12.4%	67.0 (36%)	289.3 (64%)	209 ns
YAG: Ce,Tb 5%	5700	13.8%	67.6 (43%)	272.6 (57%)	203 ns

Fig. 12 shows the RL spectra of the YAG:Ce,Tbx% (x = 0, 1, 5 and 10) ceramics, exhibiting trends similar to those observed in the PL spectra (Fig. 5). In YAG crystals doped with 0.5%  $\text{Ce}^{3+}$ , a broad emission band centered at 530 nm is observed, corresponding to the  $\text{Ce}^{3+} 5d_1 \rightarrow 4f$  emission transition. Co-doping with  $\text{Ce}^{3+}$  and  $\text{Tb}^{3+}$  ions in YAG results in the spectral overlap of the  $\text{Ce}^{3+}$  emission and the characteristic  $\text{Tb}^{3+} 4f \rightarrow 4f$  emission lines. The YAG:Ce,Tb1% sample exhibits a prominent emission line centered at 385 nm, corresponding to the characteristic  $^5D_3 \rightarrow ^7F_5$  transition of  $\text{Tb}^{3+}$  ions. However, the emission intensity decreases notably with increasing  $\text{Tb}^{3+}$  concentration, primarily due to the cross-relaxation process occurring between  $\text{Tb}^{3+}$  pairs. This behavior is consistent with the

trends observed in the PL. In contrast, the emission intensity from the  $^5D_4$  energy state of  $Tb^{3+}$  ions increases significantly with rising  $Tb^{3+}$  concentration. This enhancement can be attributed to the more efficient energy migration and transfer processes between  $Tb^{3+}$  ions at higher doping levels, leading to stronger luminescence. Fig. 12b presents the integrated RL intensity (350 - 700 nm range) for the studied YAG:Ce,Tbx% ceramics. The integrated RL intensity gradually increases with the rising concentration of  $Tb^{3+}$  ions. The obtained result can be attributed to the increased density of emission centers within the garnet host lattice. As more  $Tb^{3+}$  ions are introduced into the garnet lattice, the number of emitting ions per unit volume rises, allowing for more efficient excitation and radiative decay processes. This results in a greater number of available radiative transitions, leading to a significant luminescence enhancement (Fig. 12b). Furthermore, the closer proximity of these ions facilitates energy transfer between them, contributing to the higher intensity of the emitted light.



**Fig. 12.** (a) RL spectra for YAG:Ce,Tbx% (x = 0, 1, 5 and 10) ceramic samples; (b) Integrated RL intensity in the spectral range 350 - 700 nm.

### 3.6 Photoconversion properties

Based on the absorption spectra analysis, a 455 nm LED was used as the excitation source for photoconversion measurements. This wavelength corresponds to the maximum absorption band of  $Ce^{3+}$  ions. The measurements were conducted in a transmission setup, with ceramic samples of 1 mm thickness. The photoconversion spectra are presented in Fig. 13. A broadening of the LED emission line with a local maximum at 470 nm is observed, remaining consistent across all ceramic samples. The conversion of blue radiation into broadband emission, peaking at approximately 550 nm, is evident, as discussed in Section 3.2. Furthermore, the presence of  $Tb^{3+}$  ions influences the efficiency of blue radiation conversion into broadband emission. Specifically, with increasing  $Tb^{3+}$  concentration, the intensity of  $Ce^{3+}$  emission is enhanced, indicating an interaction between  $Tb^{3+}$  and  $Ce^{3+}$  that affects the luminescence properties of the ceramics. This effect translates into the correlated

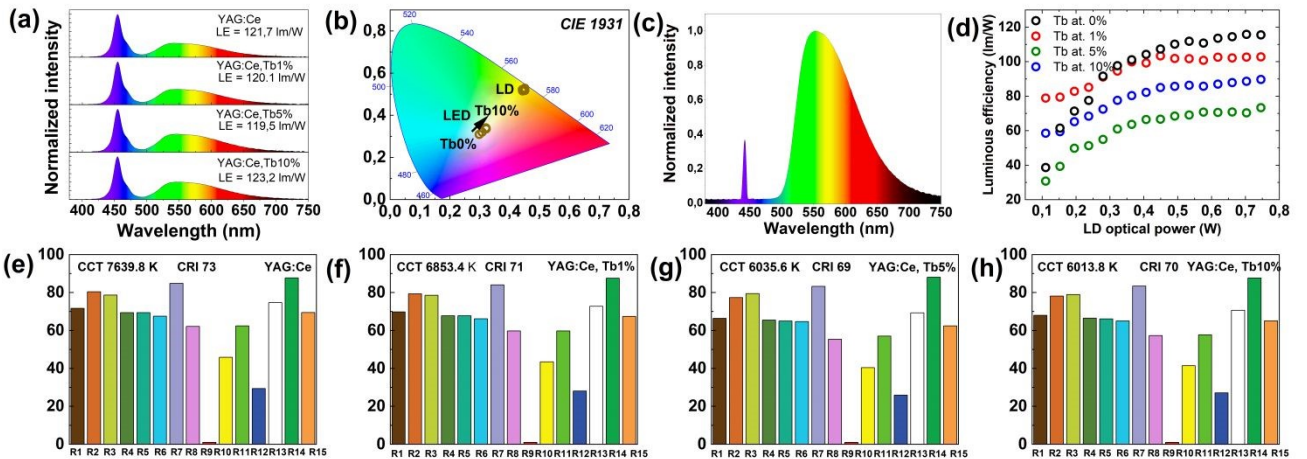


color temperature (CCT) of the emitted light, which reaches a maximum of 6013 K for YAG:Ce,Tb10% sample. The Tb<sup>3+</sup>-free sample exhibits a CCT of 7639 K. As demonstrated on the CIE chromaticity diagram, doping with Tb<sup>3+</sup> ions enables tuning of the light source's correlated color temperature. The color rendering index for all examined ceramics remains within a range of 69 - 73. Additionally, the luminous efficacy is similar for all samples, reaching approximately 120 lm/W.

**Table 5.** Photoconversion parameters of YAG:Ce,Tbx% (x = 0, 1, 5 and 10) ceramics under 455 nm LED excitation.

Composition	Photoconversion parameters			Year/reference
	CCT (K)	CRI	LE (lm/W)	
YAG:Ce ceramics	5528	62.8	122.4	2021/[53]
Al <sub>2</sub> O <sub>3</sub> /Al <sub>2</sub> O <sub>3</sub> –YAG:Ce/YAG composite	6615	69.9	126	2024/[54]
YAG:Ce	7639	73	121.7	This work
YAG:Ce,Tb1%	6853	71	120.1	This work
YAG:Ce,Tb5%	6035	69	119.5	This work
YAG:Ce,Tb10%	6013	70	123.2	This work

Fig. 13c shows a representative spectrum under 445 nm LD excitation for YAG:Ce. Due to the much narrower FWHM of the excitation source and consequently less coverage in the blue range – CRI values do not exceed 55 for any sample in the transmissive configuration, and the emission color is yellowish-white (Fig. 13b). The effect of excitation power on luminescence efficiency is shown in Fig. 13d. It is shown that the efficiency is not a constant parameter and depends on the excitation power. An increase in efficiency is observed up to a power of 0.5 W, followed by saturation. This effect agrees with the increase in sample temperature described in Section 3.4.



**Fig. 13** Photoconversion properties of YAG:Ce,Tbx% (Tb=0,1,5 and 10) ceramics: spectrum under LED excitation (a), CIE 1931 diagram of LED and LD excited samples (b) representative spectrum

of YAG:Ce under LD excitation (c) Influence of optical power on optical luminous efficiency (d) CRI of YAG:Ce,Tbx% (x = 0, 1, 5 and 10) ceramic samples (e-h)

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#### 4. Conclusions

The YAG:Ce,Tbx% (0.5, 1, 10 ) transparent ceramics were successfully fabricated using the vacuum reactive sintering technique. The progressive increase in lattice constants from 12.608 Å in YAG:Ce ceramics to 12.615 Å in YAG:Ce,Tb10%, in accordance with Vegard's law, confirmed the effective incorporation of Tb<sup>3+</sup> ions into dodecahedral crystallographic sites within the cubic garnet structure. SEM analysis revealed a fully densified microstructure without detectable micropores, either at triple grain boundaries or within grain interiors, which enabled high transparency. These ceramics exhibited an optical transmittance of 77 – 80% at 900 nm. Furthermore, their high transparency in the NIR-to-IR region positions them as promising candidates for applications as NIR-to-IR optical windows. Analysis of photoluminescence excitation and emission spectra revealed a strong dependence of the Ce<sup>3+</sup>↔Tb<sup>3+</sup> bidirectional energy transfer efficiency on Tb<sup>3+</sup> ions concentration. With increasing temperature, phonon interactions bridged the energy mismatch between Ce<sup>3+</sup> and Tb<sup>3+</sup> ions, enhancing the resonance condition and increasing the concentration of Tb<sup>3+</sup>–Ce<sup>3+</sup> pairs. This resonance led to improved energy transfer efficiency from Tb<sup>3+</sup> to Ce<sup>3+</sup> ions. The incorporation of Tb<sup>3+</sup> ions resulted in a modest reduction in the thermal stability of Ce<sup>3+</sup> emission, however, the overall stability remained at a high level. The bidirectional energy transfer significantly prolonged the effective decay time, reaching an optimal value at approximately 480 K. Above this temperature, the transfer rate accelerated further, demonstrating the temperature-driven enhancement of the energy conversion process. Scintillation decay times measured within the 1 μs range confirmed the acceleration of both fast and slow decay components due to the Tb<sup>3+</sup>–Ce<sup>3+</sup> energy transfer. However, the light yield measurements indicated a decline with increasing Tb<sup>3+</sup> concentration, attributed to radiative recombination via Ce<sup>3+</sup> centers as well as Tb<sup>3+</sup> and Tb<sup>3+</sup> ions occurring beyond the 2 μs time gate. Despite this, compared to undoped samples, the YAG:Ce,Tb10% ceramics demonstrated a remarkable two-fold increase in radioluminescence intensity measured within 1 s time gate, attributed to the higher density of emitting centers and collected light from all Tb<sup>3+</sup> ions with recombination time of hundreds ms. This enhancement underscores the potential of YAG:Ce,Tb transparent ceramics in X-ray imaging applications. The codoping of YAG:Ce ceramics with Tb<sup>3+</sup> ions also proved effective in enhancing photoconversion properties. Increasing Tb<sup>3+</sup> concentration elevated Ce<sup>3+</sup> emission and enabled tuning of the correlated color temperature (CCT). While the optimal CCT of 6013 K was observed for YAG:Ce,Tb10%, the Tb<sup>3+</sup>-free sample exhibited a higher CCT of 7639



K. The color rendering index (CRI) remained stable in the range of 69 - 73 across all samples, and luminous efficacy was consistently maintained at approximately 120 lm/W, regardless of composition. Under 455 nm laser excitation, luminescence efficiency rises with power up to 0.5 W and then saturates. The narrow FWHM of the laser limits blue light coverage, resulting in CRI values below 55 for all samples and a yellowish-white emission color. These results highlight the potential of YAG:Ce,Tb transparent ceramics as materials for X-ray imaging and white phosphors for LED applications.

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### Conflict of Interest

The authors declare no conflict of interest.

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## Data Availability Statement

Upon acceptance of the manuscript for publication, all raw data supporting the findings of this study will be made available in the ASEP repository, provided by the Institute of Physics of the Czech Academy of Sciences, Prague, Czech Republic. The data will be assigned a DOI and will be accessible after contacting either the corresponding author or the ASEP repository administration.

For further inquiries regarding data access, please contact the corresponding author, Karol Bartosiewicz, via email at [bartosiewicz@fzu.cz](mailto:bartosiewicz@fzu.cz). Alternatively, you can refer to the ASEP repository guidelines available at <https://asep-portal.lib.cas.cz/> or reach out to the ASEP repository support team via email at [arl@lib.cas.cz](mailto:arl@lib.cas.cz).

