Tunable SrAl$_2$Si$_2$O$_8$: Eu phosphor prepared in air via valence state-controlled means

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**Abstract**

SrAl$_2$Si$_2$O$_8$:xEu (x = 0.5–8%) phosphors were prepared by the high-temperature solid-state reaction in air atmosphere. The phenomenon of Eu$^{2+}$/Eu$^{3+}$ coexistence was observed and the color of the SrAl$_2$Si$_2$O$_8$:xEu phosphor could shift from light pink to blue by controllable and reproducible means. Photoluminescence (PL), excitation (PLE) spectra, X-ray photoelectron spectroscopy (XPS), and the fluorescence decay curves were employed to detect the presence of Eu$^{2+}$ and Eu$^{3+}$ ions in the compound. Under ultraviolet excitation, the broad band emission peaked at 410 nm originated from the transition of 4f$_5$5d$^1$–4f$_5^0$ from Eu$^{2+}$ and narrow band emissions peaked at 591 nm, 614 nm, 655 nm, and 703 nm are derived from the 4f–4f transition of Eu$^{3+}$ ions, although the Eu$^{3+}$ precursors were employed. The reduction mechanism from Eu$^{3+}$ to Eu$^{2+}$ in this compound was discussed in detail and verified by photoluminescence properties through changing the addition amount of Eu, temperature rise and holding time.

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1. Introduction

Lanthanide-doped oxides in inorganic phosphor materials have attracted wide attention for their potential applications in solid state lighting, display technology, and other fields [1-2]. The divalent or trivalent europium ions have been widely used as the activator in phosphors. The electronic configuration of Eu is [Xe]4f$^7$5d$^1$6s$^2$ [3]. It is well known that the emission of Eu$^{2+}$ ion in a solid state compound generally originates from the transition of 4f$^7$5d $\rightarrow$ 4f$^6$ with a broad band. Therefore, the position and width of the emission band strongly rely on the host lattice because the active electronic level is not shielded against the surrounding ligands [4-5]. However, the emission of Eu$^{3+}$ ion shows a series of narrow bands in the red spectral region corresponding to 5D$_{0,2}$–7F$_{J}$ (J = 0–4) transitions [6]. Accordingly, their positions are nearly independent of ligand field strength. Nevertheless, the intensity of Eu$^{3+}$ emission is partly connected with ligand symmetry [7]. Therefore, this clue may provide a route to prepare tunable phosphors by making full use of coexisting emissions of Eu$^{2+}$ and Eu$^{3+}$ [3].

Generally, the reduction from Eu$^{3+}$ to Eu$^{2+}$ should be carried out under reducing atmospheres, such as H$_2$, H$_2$/N$_2$, or CO, because the raw material of the europium source is usually Eu$_2$O$_3$ [8,9]. Nevertheless, recently, the reduction from Eu$^{2+}$ to Eu$^{3+}$ in air has been reported [6,10-18]. Since the aliovalent substitution method was first reported to reduce trivalent rare earth ions Eu$^{3+}$ into divalent ions Eu$^{2+}$ [10] in 1993, significant progresses had been made in phosphors. Until now, the reduction phenomena from Eu$^{3+}$ into Eu$^{2+}$ in air have been found in borates (SrBaO$_7$: Eu, SrBaO$_{10}$: Eu, BaB$_6$O$_7$: Eu and CaBPO$_4$: Eu) [10-14], phosphates (Ba$_5$(PO$_4$)$_2$; Eu) [15], sulfates (BaSO$_4$; Eu) [16], aluminates (SrAl$_4$O$_{15}$: Eu) [17], silicates (BaMgSiO$_4$: Eu) [6], and ZnO–B$_2$O$_3$–P$_2$O$_5$ glasses [18]. In the present work, possible reduction mechanisms from Eu$^{3+}$ into Eu$^{2+}$ in air have been investigated for a long time. Mingying Peng explained the reduction mechanism with the model of the charge compensation [6]. Liu et al. [19] demonstrated that the valence stability and the change of Eu$^{3+}$ in oxides were related to the energy coefficient and Dorenbos [20] pointed out that the valence stability of lanthanide ions in inorganic compounds was decided by the energy difference ($\Delta E$) between the 4f$^0$ ground state of lanthanide and the Fermi energy. However, tunable phosphors prepared in air via valence state-controlled means and luminescence properties associated with the local surrounding of phosphor were seldom reported.

In this paper, we reported the coexistence of Eu$^{2+}$ and Eu$^{3+}$ by incomplete reduction in air and tunable single-doped single-host phosphors via valence state-controlled means in SrAl$_2$Si$_2$O$_8$ host. Moreover, we consolidated the charge compensation mechanism through changing local surrounding. Therefore, it is of great
significance for phosphor development, energy conservation, and even the development of a new strategy of WLEDs and tunable blue–red emission indicator.

2. Experimental

Powder samples of SrAl$_2$Si$_2$O$_8$: Eu were prepared through a conventional solid-state method with the stoichiometric quantities of SrCO$_3$ (A.R.), Al$_2$O$_3$ (A.R.), SiO$_2$ (A.R.), Eu$_2$O$_3$ (A.R.), H$_3$BO$_3$ (A.R.) (An excess of 6 mol% of H$_3$BO$_3$ was added as flux). The mixtures were homogenized in an agate mortar and divided into 6 samples. Then 3 samples were fired for 6 h in air respectively at 1300 °C, 1400 °C, and 1500 °C. The other 3 samples were fired at 1300 °C in air respectively for 6 h, 8 h, and 10 h. The crystalline phases of synthesized products were examined by X-ray diffraction (XRD; D8 Advance diffractometer, Germany) using Cu Kα1 radiation (λ = 1.5406 Å) with a step of 0.02° (2θ) and a scanning rate of 4° min$^{-1}$. The emission and excitation spectra were recorded on a Hitachi F-4600 fluorescence spectrophotometer at room temperature. The decay curves were characterized with a spectrophotometer (Horiba, Jobin Yvon TBPX), and 365-nm and 465-nm pulse laser radiation were used as the excitation source. The quantum efficiency was detected by a fluoromax-4 spectrophotometer (Horiba, Jobin Yvon) with an integral sphere at room temperature. Thermoluminescence (TL) glow curves were performed on a Thermaloluminescent Dosimeter (RGD-3A, China) in the temperature range between 25 °C and 380 °C. X-ray photoelectron spectroscopy (XPS) measurements were performed in a PHI 5300 ESCA system with an Al Kα X-ray source with constant pass energy of 55.00 eV. The charge effect referred to the C 1s signal (284.6 eV).

3. Results and discussion

The phase purity was checked by XRD measurements. Fig. 1 shows the comparison results between the five XRD patterns of SrAl$_2$Si$_2$O$_8$: Eu containing 0.5 mol%, 2 mol%, and 8 mol% of Eu synthesized at 1300 °C, 1400 °C, and 1500 °C for 6 h in air and the reference pattern of SrAl$_2$Si$_2$O$_8$ (JCPDS 38-1454) from PDF database [21]. It is clear that all the diffraction peaks of compounds can be readily indexed to corresponding standard data of the monoclinic phase of SrAl$_2$Si$_2$O$_8$ except SrAl$_2$Si$_2$O$_8$: 8% Eu at 1300 °C. The observation results suggest that the introduction of Eu ion into the SrAl$_2$Si$_2$O$_8$ lattice does not cause any significant change to the crystal structure of the host matrix, while the extra europium in the phosphors produces another impurity phase, SrEu$_4$(SiO$_3$)$_5$. In the SrAl$_2$Si$_2$O$_8$: Eu$^{3+}$ phosphor system, it is assumed that Eu$^{3+}$ (r = 0.120 nm when the coordinate number (CN) = 7) and Eu$^{3+}$ (r = 0.101 when CN = 7) ions occupy the Sr (r = 0.121 when CN = 7) sites because both Al$^{3+}$ (r = 0.039 nm) and Si$^{4+}$ (r = 0.026 nm) sites are too small for Eu$^{3+}$ ions [22]. For the purpose of better understanding the phase purity, the Rietveld structural refinement of SrAl$_2$Si$_2$O$_8$ and SrAl$_2$Si$_2$O$_8$: Eu$^{3+}$ prepared at 1300 °C was analyzed by the GSAS program, as illustrated in Fig. 1S. The SrAl$_2$Si$_2$O$_8$: Eu crystallizes as a monoclinic structure with space group I2(15). As for SrAl$_2$Si$_2$O$_8$: Eu crystal, the lattice parameters were determined to as follows: a = 8.375 Å, b = 12.960 Å, c = 14.250 Å, and V = 1398.374 Å$^3$. The weighted profile R-factor (Rwp) and the expected R factor (Rwp) converged to Rwp = 9.19% and R = 6.84%, respectively. When the Eu ions doped in SrAl$_2$Si$_2$O$_8$ host, the lattice parameters of SrAl$_2$Si$_2$O$_8$: 0.05Eu were respectively a = 8.376 Å, b = 12.958 Å, c = 14.249 Å, V = 1397.847 Å$^3$ and the Rwp as well as R were 9.42% and 6.89%, respectively. These Rietveld refinement results demonstrated that neither the host nor the doped 0.05 mol Eu ions caused any impurity or secondary phase in SrAl$_2$Si$_2$O$_8$ host structure. The refined crystallographic data and residual factors are listed in Table 1S. Additionally, the Rietveld refinement of SrAl$_2$Si$_2$O$_8$: 0.01Eu prepared at 1500 °C in Fig. 1S and Table 1S indicated that higher temperature did not cause the structural variation.

The photoluminescence (PL) and PLE spectra of SrAl$_2$Si$_2$O$_8$: Eu shown in Fig. 2. Although the precursor of europium was the trivalent product from Eu$_2$O$_3$ and the samples were prepared through annealing at 1300 °C in air, it seems that the divalent europium coexists with trivalent europium in the compound. As shown in Fig. 2a, the two excitation spectra are respectively typical spectra for the two ions [23–25]. Hereinto, the broad band near 230 nm for SrAl$_2$Si$_2$O$_8$: Eu was attributed to the 4f–5d transition of Eu$^{3+}$. Thus, Eu$^{2+}$ and Eu$^{3+}$ can be excited at the same time or individually in a controllable way. The excitation spectra indicated that under useful excitation wavelengths the presented phosphor matrix can generate the luminescence of Eu$^{3+}$ and/or Eu$^{2+}$. The diffuse reflection spectrum of SrAl$_2$Si$_2$O$_8$: Eu (x = 0, 0.005, 0.01, 0.02, 0.05) is presented in Fig. 3. The SrAl$_2$Si$_2$O$_8$ host showed energy absorption in the short-wavelength UV region and the band gap were estimated to be about 4.76 eV (261 nm) based on the fitting line at the absorption edge [26]. Since the Eu$^{3+}$ ions were doped into the SrAl$_2$Si$_2$O$_8$ host, two broad absorption bands appeared in the 250–300 nm (peaked at 275 nm) and 320–380 nm in the near-UV range, which were attributed to the 4f–5d transition of Eu$^{3+}$. Other narrow absorption bands, which peaked at 361, 381, 391 and 453 nm, were ascribed to the transition of Eu$^{2+}$. These results match well with the excitation spectra of Eu$^{3+}$ and Eu$^{2+}$. It is observed that both the intensity of characteristic absorption band of Eu$^{3+}$ and Eu$^{2+}$ increases with the increase in the concentration of Eu ions. Emission spectra of different excitation features in the range of 234–393 nm are shown in Fig. 2b. It is reasonable to ascribe the band emission peaked at 410 nm to 4f–5d–4f transition of Eu$^{2+}$, while line emissions peaked at 591 nm, 614 nm, 655 nm, and 703 nm are assigned to 4f–4f transitions of 4f$^n$2f$^m$ transitions (n = 1, 2, 3 and 4) transitions of Eu$^{3+}$. Moreover, each emission spectrum consisting of luminescence characteristics of both Eu$^{2+}$ and Eu$^{3+}$ ions is complex due to different excitations. Therefore, the final emitting color
as-prepared phosphor is obviously controllable and largely varies with excitation wavelength.

**Fig. 4** presents the decay traces of the Eu$^{3+}$ and Eu$^{2+}$ luminescence in SrAl$_2$Si$_2$O$_8$: Eu with different Eu contents (0.5–5 mol%) prepared at 1300 °C upon excitation at 365 nm and 465 nm, respectively. All the samples can be fitted to a single-exponential function [27]:

$$I(t) = A \exp(-t/\tau),$$

where $I$ is the luminescence intensity; $A$ is a constant; $t$ is time; and $\tau$ is the lifetimes for the exponential components. As shown in Fig. 4a, the obtained lifetimes monitored at 365 nm were calculated to be 540.15, 536.00, 533.79, and 528.15 ns under the Eu contents of 0.5, 1, 2 and 5 mol%, respectively. All the values continuously become shorter with the increase in the Eu content, demonstrating that the concentration of Eu$^{2+}$ increases and the concentration quenching effect occurs. The minor variation of sample lifetime might indicate that only a small fraction of Eu exists in the divalent state. It is found from **Fig. 4b** that the calculated lifetimes monitored at 465 nm for different Eu$^{3+}$ concentrations are 1.80, 1.73, 1.04, and 0.74 ms, respectively. The linearity of decay curves is gradually lost with the increase in Eu$^{3+}$ concentration, indicating the existence of energy transfer and even concentration quenching effect [28]. It is obvious that the calculated lifetimes of Eu$^{3+}$ 4f–4f emission decrease with the increase in Eu$^{3+}$ concentration. The calculated lifetime is also related to the total relaxation rate according to Eq. (2) [29,30]:

$$\frac{1}{\tau} = \frac{1}{\tau_0} - A_{nr} + P_t,$$

where $\tau_0$ is the radiative lifetime; $A_{nr}$ is the non-radiative rate due to multiphonon relaxation; $P_t$ is the energy transfer rate between Eu$^{3+}$ ions. The distance among Eu$^{3+}$ ions decreases with the increase in Eu$^{3+}$ concentration. Therefore, the energy transfer rate among Eu$^{3+}$ and the probability of energy transfer to luminescent killer sites increase. In consequence, the lifetimes are shortened with the increase in Eu$^{3+}$ concentration. Furthermore, the luminescence decay behaviors and the lifetimes monitored at 365 nm and 465 nm of the samples are the characteristics of Eu$^{3+}$ and Eu$^{2+}$ ions, respectively [27,31], which are in accordance with the photoluminescence results. The reduction of Eu$^{3+}$ ions and the coexistence ions of Eu$^{3+}$/Eu$^{2+}$ in SrAl$_2$Si$_2$O$_8$ during the annealing process in air condition can be further proved.

The reduction of Eu$^{3+}$ to Eu$^{2+}$ in SrAl$_2$Si$_2$O$_8$: Eu in air can be explained with the model of the charge compensation mechanism [10,18]. When trivalent Eu$^{3+}$ ions were doped into SrAl$_2$Si$_2$O$_8$, they would non-equivalently substitute Sr$^{2+}$ ions. For the purpose of maintaining charge balance, three Sr$^{2+}$ ions must be replaced by two Eu$^{3+}$ ions. Therefore, each substitution of every two Eu$^{3+}$ ions would create one vacancy defect $V_{Eu}$, with two negative charges, and two positive defects of $E_{Eu}$ in the structure. Then the vacancy of $V_{Eu}$ and the two $E_{Eu}$ defects would act as the donor of electrons and the acceptor of electrons, respectively. Consequently, the negative charges in vacancy defects of $V_{Eu}$ would be transferred into Eu$^{3+}$ sites to reduce Eu$^{3+}$ to Eu$^{2+}$ via thermal stimulation. The whole process of the charge compensation mechanism could be expressed as:

$$3Sr^{2+} + 2Eu^{3+} \rightarrow V_{Eu}^0 + 2Eu_{Sr}^{3+},$$

$$V_{Eu}^0 \rightarrow V_{Sr}^0 + 2e,$$

$$2Eu_{Sr}^{3+} + 2e \rightarrow 2Eu_{Sr}^{2+}.$$  

Furthermore, tetrahedral anion group aluminosilicate also played a role of shield for Eu$^{3+}$ against the oxidation during the annealing process. The crystal structures of monoclinic SrAl$_2$Si$_2$O$_8$ are shown in **Fig. 5**. The tetrahedral framework structure consists...
of six-membered rings of corner-shared tetrahedra of AlO$_4$ and SiO$_4$ with Eu charge-compensating cations Sr$^{2+}$ occupying the large cavities within the structure [32]. Because these reduced Eu$^{2+}$ ions are in the cavities of the 3D network structure, it could effectively resist the attack of oxygen to Eu$^{2+}$ ions [6,10,13,14,17,33]. Thus the SrAl$_2$Si$_2$O$_8$ can stabilize Eu$^{2+}$ ions. Consequently, the coexistence of Eu$^{2+}$ and Eu$^{3+}$ offers us an opportunity to adjust the optical properties through controlling the amount of valence-varied activator ions in the host. We define a ratio of amount of Eu$^{2+}$/Eu$^{3+}$ as $\eta$ to investigate the variation trend of the ratio of Eu$^{2+}$/Eu$^{3+}$ with the change in the local surrounding of phosphor. The amount ratio of Eu$^{2+}$/Eu$^{3+}$ is approximately equal to the ratio of emission intensity peaked at 410 nm/614 nm in a specific form of $\eta = I_{410\text{nm}}/I_{614\text{nm}}$ [25].

The luminescent properties of the investigated phosphors prepared at 1300 °C for 6 h were investigated within the Eu concentration range of 0.5–11 mol%, as illustrated in Fig. 6(a). All investigated phosphors were monitored at an excitation wavelength of 254 nm. It is found that the changes in the PL intensities of red and blue emissions are very sensitive to the concentration variation of the Eu dopant. When the concentration of Eu is less than 2 mol%, Eu$^{3+}$ emission of red light is the dominant effect. Eu$^{2+}$ emission of blue light becomes the dominant effect when the concentration of Eu increases above 2 mol%. Nevertheless, when the Eu concentration is increased to 5 mol%, the $\eta$ increases to its maximum value and then declines sharply. This observation indicates that with the increase in Eu concentration, the $V_0$ and electrons are increased, thus promoting Eu$^{3+}$ to accept electron and be reduced. This decrease of $\eta$ with the increase in the doping concentration can be attributed to the formation of the phase: SrEu$_4$(SiO$_4$)$_3$O.

The dependence of $\eta$ on the annealing temperature (1300 °C, 1400 °C, and 1500 °C) of SrAl$_2$Si$_2$O$_8$: 0.01Eu for 6 h in air was measured experimentally (Fig. 6(b)). It is found that the PL spectra of those phosphors could be tuned efficiently through changing the temperature. The $\eta$ is increased with the temperature rise and Eu$^{3+}$ was nearly completely reduced at 1500 °C. The inset of Fig. 6(b) presents the influence of the annealing temperature on the emission intensity. Along with the temperature rise, the blue emission of Eu$^{2+}$ was increased and the red emission of Eu$^{3+}$ was reduced. It demonstrates that temperature rise can improve the reduction from Eu$^{3+}$ to Eu$^{2+}$. Fig. 6(c) displays the influence of

![Fig. 4. The decay curves of SrAl$_2$Si$_2$O$_8$: xEu ($x = 0.5$–5%) monitored at 410 nm (a) and 614 nm (b).](image)

![Fig. 5. The crystal structures of SrAl$_2$Si$_2$O$_8$ (a) and the coordination environment of the Sr (b). The related bond lengths are in the units of Å.](image)
holding time (6 h, 8 h, and 10 h) of SrAl₂Si₂O₈: 0.01Eu on luminescent properties at 1300 °C in air. The η was increased with the increase in holding time and clearly varied with holding time. These results suggest that thermal stimulation is a key factor for transferring negative charges from electron traps (vacancies which have stored electrons) into the Eu³⁺ sites and reducing Eu³⁺ into Eu²⁺.

Electrons in the sample which is heated after the irradiation phase are thermally excited out of their traps and migrate through the conduction band to recombine with the hole on Eu, thus yielding a Eu²⁺ 5d→4f thermoluminescence (TL) glow [34]. To further investigate the relation between the thermal stimulation and electron traps, the dependence of thermoluminescence spectrum of SrAl₂Si₂O₈: 0.01Eu on annealing temperature (1300 °C, 1400 °C, and 1500 °C) for 6 h monitored at integrated spectrum is plotted on Fig. 7. These phosphor samples were irradiated by UV lights at ~200 nm (150 W Xe lamp) for 5 min, and the heating rate was controlled at 15 °C/s. As shown in Fig. 7, all TL glow curves have one low-temperature peak within the temperature range between 217 °C and 227 °C and one high-temperature peak at the temperature about 372 °C, indicating that all the samples have two different depths of electron traps. When the annealing temperature increases, the TL intensity of samples decreased because the electron transfer from negative vacancies to Eu²⁺ caused by thermal stimulation was increased [35–37] and then the vacancy defects of Vₓ⁺ decreased. Thus, the capability of storing and releasing energy was reduced. Additionally, the low-temperature peak shifted from 227 °C to 217 °C with the rise in the annealing temperature. The activation energy E_T of a hole from a trap level to the valence band of the host crystal was proportional to kT_m [38], where T_m is the peak of the TL glow curves and k is a Boltzmann’s constant. The values of TL peaks implied that the depth of traps was the lowest in the samples prepared at 1500 °C. The result was in agreement with the results shown in Fig. 6(b). All the above results demonstrate that thermal stimulation can promote the reduction process of Eu³⁺ to Eu²⁺ in air.

Fig. 6. Dependence of η and PL spectra on the amount of Eu in SrAl₂Si₂O₈: xEu (x = 0.5–11%) prepared at 1300 °C for 6 h (a); the annealing temperature (1300 °C, 1400 °C and 1500 °C) of SrAl₂Si₂O₈: 0.01Eu for 6 h (b); holding time (6 h, 8 h and 10 h) of SrAl₂Si₂O₈: 0.01Eu at 1300 °C (c). All emissions were observed at an excitation wavelength of 254 nm.

Fig. 7. The thermoluminescence glow curves of SrAl₂Si₂O₈: 0.01Eu annealed at different temperatures (1300 °C, 1400 °C and 1500 °C) for 6 h in air.
The temperature-dependent PL spectra of the SrAl$_2$Si$_2$O$_8$: 0.1Eu phosphor is depicted in Fig. 8. The intensities of the characteristic peaks of Eu$^{2+}$ at 150 °C are respectively reduced to 85.75%, 83.49% and 71.78% of the initial intensities at room temperature. When the temperature rises to 300 °C, the intensities of the characteristic peaks of Eu$^{2+}$ respectively decline to 39.28%, 32.65%, and 30.26% of the initial intensities. These results proved the high quenching temperature of the samples. The emission intensity decrease with temperature rise is attributed to the interaction of electron–phonon in both the ground state and the excited state of luminescence center. This non-radiative transition probability depends on the temperature, leading to the decrease in emission intensity [39].

The $x$ and $y$ values of the CIE chromaticity coordinates of SrAl$_2$Si$_2$O$_8$: Eu phosphors prepared under different dopant contents and annealing temperatures were measured (Fig. 25). The color tone can be tuned from light pink (0.35,0.23) to blue (0.22,0.16) by doping different concentrations of Eu or from light pink (0.28,0.19) to blue (0.18,0.10) by changing the annealing temperature. The photographs of emitting phosphors are given in the inset and the as-observed emitting color is very obvious, indicating that the tunable luminescence can be realized in the SrAl$_2$Si$_2$O$_8$: Eu phosphors. The coexistence of Eu$^{2+}$ and Eu$^{3+}$ in SrAl$_2$Si$_2$O$_8$: Eu was also proved by the XPS spectra, as shown in Fig. 9. Only one peak at 1134 eV in the samples annealed at 1300 °C and 1400 °C was observed. The result was well consistent with the signal of Eu$^{3+}$ 3d$_{5/2}$ [40]. The results can be interpreted below. The amount of Eu$^{2+}$ in the powders prepared at 1300 °C and 1400 °C is limited. Moreover, the XPS technique can only measure the element content in the sample surface, where Eu$^{2+}$ is easily to be oxidized in an ambient atmosphere. When the annealing temperature rises to 1500 °C, two peaks around 1126 and 1134 eV are observed in the sample. The result was well consistent with the signals of Eu$^{2+}$ 3d$_{5/2}$ and Eu$^{3+}$ 3d$_{5/2}$, respectively. Furthermore, we measured the QE value of the blue (370–462 nm) and red emission (580–630 nm) of the selected product, SrAl$_2$Si$_2$O$_8$: 0.1Eu phosphor. The internal QE for them are 42.88% and 10.35% under 319 nm and 391 nm excitation, respectively. Although the values are not very high, the QE can be improved by controlling the processing conditions and compositions [41,42].

4. Conclusions

In summary, a controllable and reproducible tunable single-host multi-color-emitting SrAl$_2$Si$_2$O$_8$: Eu phosphor prepared in air is realized through the coexistence strategy of Eu$^{2+}$/Eu$^{3+}$. The coexistence of Eu$^{2+}$/Eu$^{3+}$ in SrAl$_2$Si$_2$O$_8$: Eu phosphor is verified by photoluminescence (PL), excitation (PLE) spectra, X-ray photoelectron spectroscopy (XPS), and the fluorescence decay curves. The multi-color-emitting phosphors can be easily tuned from light pink to blue by simply adjusting the valence state of Eu through changing the addition amount of Eu$^{2+}$, temperature rise, and holding time. The prepared SrAl$_2$Si$_2$O$_8$: Eu phosphor can be regarded as the potential candidate as tunable blue–red emission indicator through adjusting the wavelength of UV irradiation.

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Appendix A. Supplementary material

Supplementary data associated with this article can be found, in the online version, at [http://dx.doi.org/10.1016/j.optmat.2014.12.023](http://dx.doi.org/10.1016/j.optmat.2014.12.023).

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