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Comparison of optical properties in Er_xSc_{2-x}Si₂O₇ and Er_{2-x}Y_xSi₂O₇ silicate and their applications

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Abstract

A review of the current status of research on Er-Sc silicates ($Er_xSc_{2-x}Si_2O_7$) and Er-Y silicates ($Er_xY_{2-x}Si_2O_7$) $_{x}$ Si₂O₇) is presented. Rare earth silicates have been fabricated by RF-sputtering followed by thermal annealing at high temperature. The structural properties were characterized using RBS, TEM and Xray measurements. The Er_xSc_{2-x}Si₂O₇ show an emission peak at 1537 nm for annealing temperature at 1250° C, and at 1535 nm for Er_xY_{2-x}Si₂O₇ annealed at 1200° C. From PLE and PL spectra of Er_xSc₂₋ $_{x}$ Si₂O₇ thin film, Er³⁺ Stark energy levels of ⁴ $I_{13/2}$ to ⁴ $I_{15/2}$ manifolds are due to the crystal field strength effect of Sc³⁺, however, a large peak for Er_xSc_{2-x}Si₂O₇ was determined. The full-width at half-maximum (FWHM) of Er_xSc_{2-x}Si₂O₇ and Er_xY_{2-x}Si₂O₇ silicates are 2.3 nm and 50 nm, respectively. The narrow FWHM is due to the small ionic radii of Sc^{3+} , which enhance the crystal field strength affecting the optical properties of Er3+ ions located at the well-defined lattice sites of Sc silicate compared to Y ions. Temperature-dependent PL of the Er_xSc_{2-x}Si₂O₇ phase exhibits a variation of FWHM from 1.1 to 2.3 nm. The excitation cross-section (σ_{ex}) is in the same order for Er-Sc silicates and Er-Y silicates equal to 1.4 x 10^{-21} cm², and 2.0 x 10^{-21} cm² at λ_{ex} = 980 nm, respectively. Thus, show that Sc ions play the same role than Y ions in the structure as an inactive elements and acts as a disperser of Er ions. Finally, Er-Y silicate layer (ErxY2-xSi2O7) on the top of PhC cavity was used to demonstrate a silicon light source operating at room temperature.

Introduction

Integrated photonics has enabled much progress towards silicon micro-photonics. This requires silicon compatibility of all the integrated optical components. A significant work has already been published and different approaches to the realization of different photonic devices have been attempted examples: micro-detectors based on III-V structures coupled to a silicon waveguide [1], Ge detectors deposited on Si substrate [2, 3], and Ge implanted in silicon substrate [4] have been demonstrated; modulators device based on the free carrier plasma dispersion effect [5] and electro-absorption effect [6-8] have been realized; hybrid amplifiers [9], and switchers [10] have been shown. Erbium doped materials are attracting an increasing interest for the possible development of Si-compatible micro-photonic devices such as high-efficiency light sources [11, 12] and erbium doped porous silicon [13, 16], owing to the fact that the Er ions emission wavelength of 1.5 µm corresponds to a minimum in the loss spectrum of silica optical fibers. However, in the case of Er incorporation in crystalline Si (c-Si), although the co-implantation of Er and O allowed to limit Er segregation and precipitation due to the formation of Er–O complexes [17], and the room temperature luminescence efficiency remains low because of the strong non-radiative processes competing with the radiative Er de-excitation. An alternative approach is represented by Er compounds, such as oxides or silicates, in which Er is a bulk constituent [18, 19, 20]. It has been demonstrated that all Er ions about 10²² at/cm³ are optically active [19], and this fact could allow to obtain a high gain coefficient.Furthermore, it has been demonstrated that the influence of upconversion phenomena on the 1.54 μ m emission can be strongly limited by using a low pump power [19]. However, the high Er content necessarily involves a very small mean Er–Er distance; this leads to a strong Er-Er interaction and to the travelling of the excitation inside the material, hence making optical properties of the material itself very sensitive to the presence of defects [21]. Therefore, an optimization of the structural properties is fundamental to limit the detrimental effects of nonradiative phenomena. Annealing treatments at high temperatures are effective for

defect density reduction and for the improvement of the emission efficiency [22, 23]. An effective strategy for reducing up-conversion is to incorporate yttrium (Y) cations into the structure, where they substitute Er ions in the silicate lattice and prevent neighboring Er ions from causing up-conversion due to the similar ionic radius between Y and Er [24, 25]. However, scandium ions (Sc³⁺) radius is smaller (ionic radius = 0.75 Å) than erbium (Er³⁺) (ionic radius = 0.881 Å). Generally, this can result in enhancing the crystal field strength for Er doped silicates and oxides [26, 27]. In fact, Fornasiero et al. synthesized single crystal of Er doped Sc silicates using the Czochralski technique with the idea that Sc³⁺ ions would increase the Stark-splitting of the thermally populated erbium ground state as well as that of other electronic energy levels of the silicates and thereby reduce reabsorption losses [27].

In this chapter, we will compare the structural characteristic, the optical properties of Er-Sc disilicates to Er-Y disilicates, and application of their materials will be presented.

Experimental Section

Er-Sc multilayer thin films were grown by RF-sputtering by alternating 15 nm thick layers of Er_2O_3 , and Sc_2O_3 separated by 15 nm thick SiO_2 layer. These layers were deposited on 50 nm thick Er_2O_3 on SiO₂ (1.3 μ m)/Si (100) substrate at room temperature. After deposition, the samples were annealed in O₂ at 900 °C, 1000 °C, 1100 °C, or 1250 °C for 1 h. The concentration of different chemical elements in the films was determined by Rutherford backscattering spectroscopy (RBS). He⁺⁺ ions delivered by a 2.275 MeV Van de Graf accelerator with the normal detector angle of 160° was used. Spectra are fitted by applying a theoretical layer model and iteratively adjusted elemental concentrations and thickness until good agreement is found between the theoretical and the experimental spectra. The samples were analyzed from transmission electron microscopy (TEM) images obtained at 200 KeV, and by synchrotron grazing incidence X-ray diffraction (GIXD) experiments performed on the as-grown and annealed samples at the BL24 in SPring-8 using an Xray wavelength of 1.24 Å and an incidence angle of 1.0°. Photoluminescence (PL) and photoluminescence excitation (PLE) spectra were obtained by using two excitation sources at 980 and 1527.6 nm. The excitation laser was focused to a spot with a diameter of about 15 μ m and an incident angle of 45 degrees through an objective lens. The luminescence from the sample was collected perpendicularly by a different objective lens with a numerical aperture of 0.40. The PL spectra were detected using a 0.5-m spectrometer and cooled InGaAs detector. Time-resolved PL was measured by using a pulsed laser formed by an acousto-optic modulator and a streak camera (Hamamatsu C11293S). The spectral and temporal resolutions of this system are 500 µeV and 20 ps (for a 1 ns time range), respectively.

Results and Discussion

Structural properties

To determine the concentration distribution of chemical elements in the samples, Najar *et al.* performed RBS measurements [28]. RBS spectra in Figure 8.1(a) show the profiles of Er, Sc, Si, and O atom distributions after deposition and annealing at 1100 and 1250 °C. RBS signal intensities changed with increasing annealing temperature, suggesting that the concentration changes for Er^{3+}

and Sc³⁺. Note that for the as-grown sample, the Er distribution is formed by peaks corresponding to Er₂O₃ layers. Whereas, the intensity decrease for sample annealed at 1100 °C and erbium diffuse in-depth. The concentration of Si and O has an atomic ratio of Si to O of about 1:4, indicating that the films are mainly formed by two phases Er_xSc_{2-x}SiO₅ and Er_xSc_{2-x}Si₂O₇. In contrast, at 1250 °C erbium starts to penetrate the silicate matrix and becomes more homogeneous. From the RBS analysis, the concentration of Si and O, has an atomic ratio of 2:7, indicating that the films are mainly composed of the Er_xSc_{2-x}Si₂O₇ phase. The RBS analysis indicates erbium and scandium doping concentrations close to $[Er] = 12.7 \times 10^{20}$, and $[Sc] = 4.1 \times 10^{20}$ at/cm³ at 1100 °C and [Er] =6.7x10²¹, and [Sc] = $1.4x10^{21}$ at/cm³ at 1250 °C. Also, (Y_{1-x}Erx)₂Si₂O₇ has been prepared using cosputtering from Er₂O₃, Y₂O₃ and SiO₂ targets by Miritello et al. [29]. Figure 8.1(b) reports the RBS spectra of $(Y_{1-x}Erx)_2Si_2O_7$ to estimate their chemical composition. The elemental signals are constant, indicating a uniform chemical composition along all the film thickness. The green line and open triangles in Figure 8.1 (b) shows the disilicate stoichiometry (Y + Er):Si:O = 2:2:7. The atomic concentration of the single RE is also evaluated, and it varies in the range between 0.2 at.% and 18.0 at.%. In contrast in (Yb-Er) compounds it is not possible to distinguish the Yb and Er contributions to the RBS signal because the energy difference between their surface edges is lower than the energy resolution of the detector.



FIGURE 8.1

(a) RBS signal of as-grown and after annealing at 1100°C and 1250°C for 1h in O₂, (b) RBS spectra of (Y_{1-x}Er_x)₂Si₂O₇, Yb₂Si₂O₇ and (Yb_{1-x}Er_x)₂Si₂O₇ thin films [28, 29]

To understand the microscopic structures of these materials and determine the dominant phase at etch annealing temperature, a TEM analysis were performed. After deposition, the TEM analysis on the cross-section coupled with selected area electron diffraction (SAED) images and EDS measurements of the samples prepared by Najar *et al.* were studied after deposition, annealing at 1100 °C, and 1250 °C [28]. Figure 8.2(a) shows a schematic cross-section of the different layers. The cross-sectional image in Figure 8.2(b) obtained after deposition shows different layers of Er_2O_3 , Sc_2O_3 , and SiO_2 with a total deposition thickness of around 109 nm. The inset SAED image from the Er_2O_3 layer at the bottom shows multicrystalline rings. The interplanar spacings d are about 1.29, 1.32 and 1.52 Å, corresponding respectively to the (203), (440), and (20-3) planes for $Er_2Si_2O_7$, and with d of 1.32 and 1.52 Å corresponding to the (800) and (444) planes for Er_2O_3 . For the sample annealed at 1100 °C, the TEM image in Figure 8.2(c) shows two layers separated by some voids. The appearance of these voids may be due to the diffusion of SiO_2 layer. The interplanar spacings are about 1.7, 2.83 and 1.88 Å, corresponding respectively to the (233), (-402), and (-523) planes for Er_2SiO_5 , and the (132), (130), and (22-2) planes for $Er_2Si_2O_7$.



FIGURE 8.2

(a) The fabricated structure. (b)-(c) and (d) cross-sectional TEM images after deposition and after annealing at 1100 °C and 1250 °C, respectively, with SAED images in insets [28]

These results show the mixture formation of $Er_xSc_{2-x}SiO_5$, $Er_xSc_{2-x}Si2O_7$ phases at 1100 °C. After thermal annealing at 1250 °C, we formed a unique layer with an average thickness of 102 nm as shown in Figure 8.2(d) and the size of the voids becomes very small compared to 1100 °C. The SAED images show a single crystal compound. The interplanar spacings are 1.30, 1.54 and 2.61 Å, nearly corresponding respectively to the (203), (33-2), and (220) planes for $Er_2Si_2O_7$, and Er_2SiO_5 phase with (-442) plane for 1.54 Å spacing. EDS measurements show that Er and Sc concentrations are 6.7×10^{21} and 1.4×10^{21} at/cm³, respectively, with the Er/Sc ratio of 4.5. TEM images show for both annealing temperatures, the existence of Er_xSc_{2-x}SiO₅ and Er_xSc_{2-x}Si2O₇ phases in the structures. Er diffusion at 1250°C was analyzed by measuring the Er concentration profiles before and after heat treatment in Figure 8.3. After deposition, the atomic weight of Er is estimated to be 35% to 40%, and these values decrease from 11% to 14% after annealing at 1250 °C due to the homogeneous redistribution of Er atoms in the annealing layers. Er diffuses in the depth with a diffusion length of around 39 nm in the bottom layer of SiO₂ compared to the as-grown sample, but we suppose that Er diffuses with the same thickness in the other layers. The diffusion length is given by L= 2(DT)^{1/2}, where D is the diffusion coefficient and t is the duration of the thermal treatment. For the annealing temperature of 1250 °C, the diffusion coefficient D is 1 × 10⁻¹⁵ cm² /s. This value is fairly consistent with the value of 0.63×10^{-15} cm² /s for Er in the SiO₂ layer prepared by magnetron sputtering and annealed at 1100 °C for 1.5 h in N₂ [30] and about two orders of magnitude higher than the diffusion coefficient of silicon-rich silicon oxide (SRSO) of 1.2×10^{-17} cm² /s at 1100 °C [31].



FIGURE 8.4

EDS concentration profiles of Er after deposition and annealing at 1250 °C. ³²

We confirm the dominant phase after annealing; we performed GIXD experiments on the as-grown and annealed samples [28]. GIXD profiles of the crystalline structure after deposition and annealing are shown in Figure 8.4. The GIXD profile of the sample after deposition shows the presence of Er_2O_3 , Er_2SiO_5 , Er_2SiO_5 , Sc_2SiO_5 , and $Er_2Si_2O_7$ in the films. After annealing at 1100 °C, the main intense peaks are close to Er_2SiO_5 and Sc_2SiO_5 with two peaks also close to $Er_2Si_2O_7$. This result confirms the TEM and RBS measurements showing the presence of both $Er_xSc_2.xSiO_5$ and $Er_xSc_2.xSi_2O_7$ phases. The size of the polycrystalline $Er_xSc_2.xSiO_5$ is around 39 nm at 1100 °C according to the analysis of the diffraction peak width, whereas is equal to 50 nm for $Er_xSc_2.xSi_2O_7$ polycrystalline. After annealing at 1200 °C (data not shown here) and 1250 °C, we have only $Er_2Si_2O_7$ because of diffusion of Er and Sc into different layers and the formation of new polycrystalline mixed compounds assigned to $Er_xSc_{2-x}Si_2O_7$, which is the dominant phase. The size of the polycrystalline $Er_xSc_{2-x}Si_2O_7$ is around 35 nm at 1250 °C according to the analysis of the diffraction peak width.



FIGURE 8.4

GIXD profiles obtained from samples after deposition and after annealing at 1100° C and 1250° C for 1h in O₂ with the Joint Commission of Powder diffraction standard (JCPDS) corresponding to the different compound [28]

Optical properties: Er_xSc_{2-x}Si₂O₇ and Er_xY_{2-x}Si₂O₇

To demonstrate a relation between composition stoichiometry and optical properties, we analyzed the annealed samples by PL [28]. Figure 8.5 (a) shows PL spectra of samples annealed between 900 and 1250 °C using λ_{ex} = 980 nm at room temperature. For annealing temperature equal to 900 and 1100 °C, the spectra exhibit several emission peaks, corresponding to the ⁴ I_{13/2} to ⁴ I_{15/2} transition of Er³⁺ ions. The main intense peak is observed at 1528 nm and the second intense peak is at 1537 nm at 1100 °C. The PL spectrum with a single main intense peak at 1528 nm is similar to the PL spectrum of the Er silicates [26]. In this case, the peak at 1528 nm corresponds to the Er_xSc_{2-x}SiO₅ phase and confirms the previous observations at 1100 °C.

However, the samples annealed at 1200°C and 1250 °C exhibit PL spectra with a different shape compared to 900 and 1100 °C, with the main intense peak at 1537 nm with narrower line width, indicating that a relatively homogeneous crystalline environment for Er^{3+} ions, and with sub-peaks at 1546.2 and 1551 nm.

The PL peak intensity of annealed samples reaches the maximum at 1250 °C. The peak at 1537 nm appears as the main peak for the films at 1200 and 1250 °C, while the 1528 nm main PL peak typical of $\text{Er}_x\text{Sc}_{2-x}\text{SiO}_5$ is observed for the films with low intensity. It is considered that the 1537 nm peak come from another Er silicate structure formed by the reaction of the Er–O and Si–O precursors with the SiO₂ interlayers. The high temperature annealing at 1200 and 1250 °C for 1 h enhances the reaction of the Er–O and Si–O precursors with the SiO₂, converting the most of $\text{Er}_x\text{Sc}_{2-x}\text{SiO}_5$ to $\text{Er}_x\text{Sc}_{2-x}\text{SiO}_5$ to $\text{Er}_x\text{Sc}_{2-x}\text{SiO}_5$. The surface morphology of the films has observed by AFM, which revealed that the surface is still flat, smooth, and noncracking after annealing up to 1250 °C [32].

To distinguish the role of Sc^{3+} compared to Y^{3+} , we compared the spectra from $\alpha-({}^{Y}_{1-x}Er_x)_2Si_2O_7$ prepared by Miritello *et al.* studied under 980 nm excitation [29] to $Er_xSc_{2-x}Si_2O_7$ phase. The PL is shown is Figure 8.5(b). The recorded PL shows a large peak at 1535 nm with FWHM = 50 nm. We conclude that Sc^{3+} affect the FWHM of the emission peak.



FIGURE 8.5

(a) PL spectra at room temperature obtained from the sample annealed from 900 to 1250 °C using 980 nm excitation wavelength, (b) PL spectra from α -(Y_{1-x}Er_x)₂Si₂O₇ under λ_{ex} = 980 nm [28, 29]

To determine the Stark energy levels of the $Er_xSc_{2-x}Si_2O_7$ phase, the photoluminescence excitation (PLE) and PL tests were conducted at 4 K in Figure 8.6. Figure 8.6(a) shows a PLE color plot under ${}^{4}I_{13/2}$ manifold excitation. The dashed lines in this Figure indicate the assigned energy levels of the ${}^{4}I_{13/2}$ manifold measured for $Er_xSc_{2-x}Si_2O_7$ in the polycrystalline phase. The PLE spectrum of the sample annealed at 1250 °C in Figure 8.6(b) shows peaks at 1512, 1519, and 1529 nm corresponding to the direct absorption mechanism with an FWHM of 2, 2.5, and 2.25 nm, respectively. The spectral line widths of the PL emissions are 1.1, 1.76, and 1.37 nm, which correspond respectively to the emission peaks at 1537, 1546, and 1551 nm. By varying the excitation wavelength, λ_{exc} , between 750 nm and 1000 nm, the PLE spectrum recorded at 1535 nm

is characterized by two sharp peaks depicted in Figure 8.6(c). The peaks centered respectively at 800 nm and at 980 nm correspond to the ${}^{4}I_{15/2} - {}^{4}I_{9/2}$ and ${}^{4}I_{15/2} - {}^{4}I_{11/2}$ Er³⁺ transitions. Since Y³⁺ ions are optically inactive, Er³⁺ ions can be excited only by direct excitation.



FIGURE 8.6

(a) PLE color plot measured at 4K obtained from the sample annealed at 1250 °C, (b) PLE and PL spectra at 4K, and (c) PLE spectra recorded from α -(Y_{1-x}Er_x)₂Si₂O₇ [28, 29]

From the collected spectra, we obtained the energy level schematic in Figure 8.7 corresponding to transition between the ${}^{4}I_{15/2}$ and ${}^{4}I_{13/2}$ manifolds of Er in Er_xSc_{2-x}Si₂O₇ for the sample annealed at 1250 °C. The excited state ${}^{4}I_{13/2}$ level of Er³⁺ is composed of four levels instead of seven levels, and the ground state ${}^{4}I_{13/2}$ level is composed of two levels instead of eight levels. The six component peaks corresponding to Y₂-Z₁, Y₃-Z₃, Y₁-Z₁, Y₃-Z₁, Y₁-Z₂ and Y₁-Z₃ transitions are presented in Figure 8.7.



FIGURE 8.7

Schematic illustration of main optical transition of Er^{3+} in $Er_xSc_{2-x}Si_2O_7$. The observed absorption and emission transitions are indicated in red and blue, respectively [28]

These level energies of $Er_xSc_{2-x}Si_2O_7$ are different from $Er:Sc_2O_3$ and Er in scandium silicate [33, 34]. In addition, M. Miritello *et al.* obtained a peak emission at 1535 nm for α - $(Y_{1-x}Er_x)_2Si_2O_7$ with a 37 nm FWHM using 532 nm excitation wavelength after annealing at 1200 °C [29]. Furthermore, the energy levels were different from Er_2O_3 with strong luminescence from Y_2 - Z_1 and Y_1 - Z_1 transitions in C_{3i} site observed at wavelengths of 1544.7 and 1548.2 nm [35]. The split of Stark levels are due to the smaller Sc³⁺ radius's enhancing the crystal field strength, which affects the luminescence properties with a smaller FWHM compared to the effect of Y ions. However, Er can be substituted with Y ions in the silicate phase, but Sc has additional effect due to it is smaller radius compared to Er and Y. The scalar crystal-field strength is defined by following Eq. (1) [32, 36]:

$$N_{v} = \left[\sum_{k,q} \frac{4\pi}{2k+1} (B_{q}^{k})^{2}\right]^{\frac{1}{2}}$$
(1)

Where B_q^k represents the non-zero crystal-filed parameters that contain the radially dependent parts (k and q) of the one-electron crystal-field. B_q^k characterizes the interaction between ligands and the central ions and includes the radial integral of the wavefunction. The values of the crystal field strength N_v given in Eq. (1) have been calculated for Er³⁺ doped into different matrices Sc₂O₃ (5300 cm⁻¹), Y₂O₃ (2700 cm⁻¹), and Er₂O₃ (2200 cm⁻¹) [37, 38], and they show clearly that the crystal-field strength increases regularly with decreasing ionic radius of the RE host cation, which generates the Stark-splitting energy levels. The presence of peaks at 1542, 1546, and 1551 nm is due to the presence of Sc ions, which increase the crystal field strength and thereby enhance the Stark-splitting of the thermally populated Er energy levels (${}^4I_{15/2}$ and ${}^4I_{13/2}$ levels) as well as that of the other electronic energy levels. Figure 8.8 shows the variation of FWHM at 1537 nm from 4 K to room temperature of Er_xSc_{2-x}Si₂O₇. FWHM increases from 1.1 to 2.3 nm with increasing of temperature. The PL spectra are plotted as a function of temperature in the inset in Figure 8.8. As can be seen, the Er^{3+} PL intensity becomes stronger at low temperature without shifting. The absence of shoulders peak emission for the main peak at 1537 nm like Er doped Si nanocluster that can be inducing larger inhomogeneous broadening of Er^{3+} : ${}^{4}I_{13/2} - {}^{4}I_{15/2}$ transition [39].



FIGURE 8.8

Variation of FWHM at 1537 nm as a function of temperature. Inset shows temperature dependent PL spectra of $Er_xSc_{2-x}Si_2O_7$ for λ_{ex} = 1527.6 nm [28]

The results of integrated PL intensity at 1537 nm as a function of excitation power are shown in Figure 8.9. The PL intensity increases linearly and then exhibits a sublinear increase associated with the upconversion phenomena where two Er^{3+} ions both excited at the ${}^{4}I_{13/2}$ level interact, with one being de-excited to the ${}^{4}I_{15/2}$ ground state and the other being resonantly excited to the ${}^{4}I_{9/2}$ level [40]. At low pump powers, however, a nonradiative concentration quenching process may occur. In this case the excitation is resonantly transferred from one excited Er^{3+} ion to a nearby Er^{3+} ion in the ground state and hence it travels along the sample. In order to determine the excitation cross-section, it can be shown that the pump intensity dependence of the PL from erbium is given by the following Eq. (2) [16]:

$$I_{PL} = I_{PL \max} / (1 + hc\tau^{-1} / (\sigma_{ex} p_{ex} \lambda_{ex}))$$
(2)

where I_{PLmax} is the maximum PL intensity, h is Plank's constant, c is the speed of light, pex is the pump power, and τ the lifetime at room temperature, which is equal to 12 µs fitted with single exponential for the excitation wavelength at $\lambda_{ex} = 1527.6$ nm. Fitting the experimental data of this equation, an estimate of the excitation cross-section σ_{ex} can be obtained. The value of σ_{ex} is 3.0×10^{-20} cm² at $\lambda_{ex} = 1527.6$ nm. This is one order of magnitude higher than the σ_{ex} of 1.4×10^{-21} cm² at $\lambda_{ex} = 980$ nm reported by us for the same structure [41]. The difference between both values is due to the excitation wavelength (resonant excitation at 1527.6 nm) and the life time depending on excitation wavelength. In addition, σ_{ex} at $\lambda_{ex} = 980$ nm is on the same order as the excitation cross-section in α -($_{Y1-x}Er_x$)₂Si₂O₇ of 2.0×10^{-21} cm². This is means that Sc plays the same role as Y in the structure - an inactive element and acts as a disperser of Er ions – in addition, it presents a higher

crystal field strength that enhances the Stark-splitting of Er energy levels. On the other hand, σ_{ex} in Er_xSc_{2-x}Si₂O₇ (1.4x10⁻²¹ cm²) is smaller than that in α -(Y_{1-x}Er_x)₂Si₂O₇, 2.0x10⁻²⁰ cm² at 980 nm, because of the energy transfer from Yb to Er ions [29]. The key advantage of Er_xSc_{2-x}Si₂O₇ is that present high excitation cross-section, small FWHM at 1537 nm, and high crystal field strength due to Sc ions that enhances the Stark splitting of Er. Er_xSc_{2-x}Si₂O₇ compound is promising material by combining with photonic crystal structures to build light sources or on-chip optical amplifiers.



FIGURE 8.9

Integrated PL intensity as a function of excitation power at 1527.6 nm [28], Inset for the same sample with excitation power at 980 nm [41]

Applications: silicon light source using Er_{2-x}Y_xSi₂O₇ silicates

Lo Savio et *al.* explored an alternative approach for Purcell enhancement. The idea is to deposit a thin layer of $Er_{2-x}Y_xSi_2O_7$ on top of a silicon photonic crystal cavity [25]. The schematic of the device and an SEM image of the L3 PhC cavity after deposition of 10 nm layer of Y-Er are shown in Figure 8.10(a). The photonic structures studied are L₃ cavities, formed in a PhC slab by removing three holes from the Γ -K direction of a triangular lattice. The lattice constant (a) was varied between 405 nm and 430 nm in steps of 1 nm to fine-tune the cavity resonance wavelength, while the hole radius (r) was kept at fixed ratio r/a = 0.275. This wide parameter-range was chosen to tune the fundamental cavity mode around 1.54 μ m and to allow for the modification of the resonance wavelength induced by the deposition of the silicate thin film. The two holes closest to the cavity in the Γ -K direction (circled in blue in Fig. 8.10(b) below) were shrunk and displaced in order to increase the cavity quality factor (Q-factor). The stoichiometry of [Y + Er]:[Si]:[O] = 2:2:7, with an Er concentration of 8 × 10²⁰Er/cm³.



FIGURE 8.10

(a) Schematic of the PhC cavity coated with Y-Er disilicate film, (b) SEM image (top view) of L3 PhC cavity with 10 nm Y-Er layer deposited on the top [25]

The PL spectra of the Y-Er disilicate film deposited on the un-patterned SOI and on the L_3 PhC cavity (on-resonance) are shown in Figure 8.11. The black dashed line corresponds to the PL spectrum of the film deposited on un-patterned SOI while the spectra with the sharp peaks are obtained when the emission is on resonance with the cavity. Only the fundamental mode is shown in Figure 8.11(a) for different lattice periods for a fixed far-field parameter of 6 nm. The peaks are superimposed on a weak background, which is the emission of erbium ions present in the Y-Er film. The emission intensity of the peaks follows the background emission level and shifts towards longer wavelengths with increasing lattice constant, as shown in Figure 8.11(a).



FIGURE 8.11

PL emission from top-coated L3 cavities by varying (a) the lattice parameter and (b) the far-field optimization parameter. The dashed black line in (a) is the shape of PL observed in the Y-Er disilicate film [25]

As expected, the emission intensity from the far-field optimized cavities is higher than from the unoptimized ones but at the expense of linewidth broadening, as shown in Figure 8.11(b). The authors measured the Q-factors by the RS method [42] for the three values of far-field optimization parameters ($\Delta r=0$, +6 nm and +21 nm) and were equal to 26000, 16200 and 4000, respectively. By taking the peak to background ratio, it is clear that the erbium emission level increases by two orders of magnitude by coupling it on resonance with the cavity. This enhancement is consistent with the values achieved with the cavity-enhanced band- edge and defect luminescence [43, 44]. As $Er_xSc_2*Si_2O_7$ present a smaller FWHM (2.3 nm) compared to $Er_{2*}Y_xSi_2O_7$, the Purcel enhancement will be much higher.

Conclusions

In summary, the study of Er doped silicates materials is rapidly increased in the last years. In this paper, we reviewed relevant results on the effect of Sc ions doped Er-silicates compared to Y ions for the same phase structure prepared by the same method. Polycrystalline $Er_xSc_{2-x}Si_2O_7$ and $Er_xY_{2-x}Si_2O_7$ show a sharp emission peak at room temperature centered at 1537 nm, however, $Er_xY_{2-x}Si_2O_7$ presents a larger emission peak. The narrow FWHM of $Er_xSc_{2-x}Si_2O_7$ is due to the small ionic radii of Sc^{3+} which enhance the crystal field strength which is not the case for $Er_xY_{2-x}Si_2O_7$ due to the same ionic radius of Er and Y. Both silicates compounds show a compared excitation cross-section. We show a recent application of Er-Y silicates compounds to demonstrate a silicon light source operating at room temperature. The narrow emission of $Er_xY_{2-x}Si_2O_7$ is very promising for photonic crystal light emitting devices because the extraction efficiency can be increased with a pronounced narrowing of the emission. Thus, would greatly broaden the range of applications for the present Si-based nanoemitters opening the route to the development of small-sized Er light source and amplifiers.

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