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Phase selective synthesis of quantum cutting nanophosphors and the observation of a spontaneous room temperature phase transition†

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Oxygen-free Eu^{3+} -doped NaGdF_4 nanocrystals with high quantum cutting efficiency are accessible at low temperatures (room temperature to 80 °C) using task-specific ionic liquids (ILs) as structure directing agents and only water as solvent. Selective tuning of the shape, morphology and, most importantly, the crystal phase of the host lattice is achieved by changing the alkyl side length, the H-bonding capabilities and the concentration of 1-alkyl-3-methylimidazolium bromide ILs, $[\text{C}_n\text{mim}]\text{Br}$. When using $[\text{C}_2\text{mim}]\text{Br}$, hexagonal NaGdF_4 nanoparticles are obtained. In the case of methylimidazolium bromides with longer pendant alkyl chains such as butyl (C_4), octyl (C_8) or decyl (C_{10}), extremely small nanoparticles of the cubic polymorph form, which then convert even at room temperature (RT) to the thermodynamically favored hexagonal modification. To the best of our knowledge, this kind of spontaneous phase transition is not yet reported. The hexagonal nanomaterial shows a substantial quantum cutting efficiency (154%) whilst in the cubic material, the effect is negligible (107%). The easy yet highly phase selective green synthesis of the materials promises large scale industrial application in environmentally benign energy efficient lighting.

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1. Introduction

There is growing societal demand for producing energy efficient lighting and developed countries have started to ban classic incandescent lamps. By far the most popular replacements are compact fluorescent lamps (CFLs), which consume less energy and have a longer rated life. However, their use of Hg as a discharge medium complicates their manufacturing and disposal, adds substantial health risks and causes environmental issues. Environmentally benign noble gases such as Xe present an alternative to Hg, due to immediate start-up, essential for applications like lamps in facsimile, photocopiers, brake lights *etc.*^{1,2} As greater energy loss occurs during the conversion of shorter wavelengths to visible light, and the discharge efficiency of Xe is lower than that of Hg, phosphors exhibiting quantum yields greater than 100% are needed to make Xe-based CFLs competitive. This can be

achieved by a two-photon down-conversion process based on a combination of two different rare earth ions, *i.e.* Gd^{3+} and Eu^{3+} in $\text{NaGdF}_4\text{:Eu}^{3+}$, where the excitation energy is transferred *via* a two-step process from the quantum cutter (Gd^{3+}) to the emitting ion (Eu^{3+}), resulting in the emission of two visible photons.^{1,3} Indeed, ternary lanthanide fluorides like NaGdF_4 are amongst the most effective optical materials and they are used widely in photonics and biophotonics, because of their high refractive index, low phonon energy and good stability.^{3–6} Various synthesis routes to nanofluorides are already explored, but, in the context of quantum cutting, special care must be taken to obtain absolutely oxide-free materials, as oxygen impurities can lead to non-radiative relaxation *via* europium–oxygen charge transfer states, followed by emission from these levels.^{4–30} In this context the application of ILs has been shown to be beneficial.^{31–37} Ionic liquids are often described as green and designer solvents, and draw currently considerable attention both in academia and industries.^{31–39} As ILs are composed of distinct cations and anions, they are a widely tunable class of compounds with interesting properties. Useful properties like negligible vapor pressure, wide liquidus range, good thermal stabilities, considerable electrical conductivities and a wide electrochemical window make them suitable for use in organic catalysis, electrochemistry, f-element separation and many other applications.^{31–40} Despite their versatility, their use in inorganic materials synthesis has just begun but holds great promises.^{41–56} ILs offer an advantage in the

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synthesis of inorganic nanomaterials as they can efficiently stabilize nanoparticles and shield them against agglomeration on the basis of their size and their charge.^{32,56} In addition, their ability to interact with the nanoparticle surface allows for morphology and also phase control.⁵⁶ ILs can not only be used as a solvent but also as a reaction partner which enabled us to synthesize the first highly efficient quantum cutting, small-sized nanofluorides, $\text{GdF}_3\text{:Eu}^{3+}$.⁵⁷

Here, we present an easy yet highly phase selective synthesis of oxygen-free Eu^{3+} -doped NaGdF_4 nanocrystals (particle size below 10 nm) with high quantum cutting efficiency. The synthesis only requires stirring an aqueous solution of the starting materials and a task-specific IL such as $[\text{C}_2\text{mim}]\text{Br}$ (1-ethyl-3-methylimidazolium bromide) for one hour at room temperature. This synthesis protocol not only meets with the requirements of green chemistry as proposed by Anastas and Warner⁵⁸ but also offers the possibility of easy scale-up for industrial manufacturing. Tuning of the crystal phase of NaGdF_4 can be achieved even at room temperature by judiciously choosing the IL. The extremely small, less thermodynamically stable nanoparticles of NaGdF_4 convert at room temperature without applying external stimuli like heating to the thermodynamically more stable hexagonal phase which also has a higher quantum efficiency. Normally phase transition of nanomaterials needs at least one of the following changes: heating at high calcination temperatures, change of the reactant ratio and other external influences *etc.*^{59–62} But in the present case, spontaneous phase transition over time without using any external force is not only new and novel for sodium ternary fluoride but also very rare for other systems.

2. Experimental

2.1 Synthesis of ionic liquids

Details of the synthesis of ionic liquids used in this work are given in the ESI (Scheme S1†).

2.2 Synthesis of $\text{NaGdF}_4\text{:Eu}^{3+}$ nanocrystals

In a novel approach, $\text{NaGdF}_4\text{:Eu}$ nanocrystals are prepared using imidazoilium-based ILs with tunable alkyl chain length, reaction temperature *etc.* The synthesis only requires water as solvent and the ILs act as a nanosynthetic template. When 1-ethyl-3-methylimidazolium bromide $[\text{C}_2\text{mim}]\text{Br}$ is used at room temperature, hexagonal modification appears but at higher reaction temperature (80 °C), cubic nanorods are obtained. But on using ionic liquid of a higher chain length such as $[\text{C}_n\text{mim}]\text{Br}$ ($n = 4, 8, 10$) at room temperature, very small sized (~1.5 nm) cubic nanomaterials were obtained which undergo spontaneous phase transition over time to hexagonal (general scheme in Fig. 1a).

In a typical synthesis, 0.113 g NaCl (J. T. Baker), 0.88 g $\text{Gd}(\text{NO}_3)_3 \cdot 6\text{H}_2\text{O}$ (Alfa Aesar) and the required amount of $\text{Eu}(\text{NO}_3)_3 \cdot 6\text{H}_2\text{O}$ (Alfa Aesar) was added at RT to 12.5 mL of an aqueous solution of the respective ionic liquid (IL) with the

desired concentration. To the well-stirred solution, 12.5 mL aqueous NH_4F (Sigma Aldrich) solution was added keeping the $\text{Gd}^{3+}/\text{F}^-$ ratio at 1 : 8 in all cases. The reaction mixture was stirred at RT for 1 hour and the obtained nanocrystals were collected by centrifugation, washed several times with methanol, ethanol and acetone and dried in an oven at 80 °C. In another experiment, 0.113 g NaCl, 0.88 g $\text{Gd}(\text{NO}_3)_3 \cdot 6\text{H}_2\text{O}$ and the required amount of $\text{Eu}(\text{NO}_3)_3 \cdot 6\text{H}_2\text{O}$ was added to 12.5 mL (0.1 mol L^{-1}) aqueous solution of $[\text{C}_2\text{mim}]\text{Br}$ at 80 °C followed by addition of 12.5 mL aqueous NH_4F solution. After 1 hour of constant stirring of the reaction mixture at 80 °C, the obtained nanocrystals are centrifuged, washed and collected as described.

2.3 Characterization

PXRD (powder X-ray diffraction) measurements were carried out on a Huber G70 diffractometer (Rimsting, Germany) using Mo $\text{K}\alpha$ radiation ($\lambda = 0.07107 \text{ nm}$). The crystallite size was calculated using the Scherrer equation, $D = K\lambda/\beta \cos \theta$, where $K = 0.9$, D represents the crystallite size (\AA), λ is the wavelength of Mo $\text{K}\alpha$ radiation, and β is the corrected half width of the diffraction peak. TEM (transmission electron microscopy; FEI Tecnai F20 field emission gun TEM, Philips Electron Optics, Holland) was used to map the shape, size and lattice structure of the nanocrystals dispersed on a carbon-coated copper grid from acetone solution. Morphological characterization was also carried out by a JEOL, JSM-6700F FESEM (Japan). The excitation and emission spectra and decay time of all samples were recorded on a Fluorolog 3 (HORIBA JOBIN YVON, Germany) luminescence spectrometer equipped with steady and pulsed Xe lamps for sample excitation and a photomultiplier for signal detection. All measurements were performed at room and liquid nitrogen temperature. Excitation and emission spectra using high energy synchrotron radiation were taken at beamline I (SUPERLUMI) of the Doris III storage ring at HASYLAB, Deutsches Elektronen Synchrotron (DESY) in Hamburg, Germany.⁶³ ^1H NMR was measured on a Bruker DPX-200 (200 MHz) (Karlsruhe, Deutschland).

3. Results and discussion

3.1 Structural characterizations by powder X-ray diffraction and phase evolution

The phase evolution using various task-specific ILs and transformation is depicted schematically by Fig. 1a.

3.1.1 Effect of the aromatic π -system of the IL on phase evolution. Pure hexagonal NaGdF_4 is obtained when using $[\text{C}_2\text{mim}]\text{Br}$ (NP1, NP2 and NP3 of Table 1). The crystallinity is enhanced when the concentration of $[\text{C}_2\text{mim}]\text{Br}$ is increased from 0.1% to 1 mol% (Fig. 1b). By maintaining the same reaction conditions, if no IL is used, cubic NaGdF_4 of an extremely low degree of crystallinity is obtained instead of hexagonal (ESI, Fig. S1†), indicating the significant role played by the IL on phase evolution. The peak related to the (110) plane of the hexagonal phase is the most intense and is distinctly broader



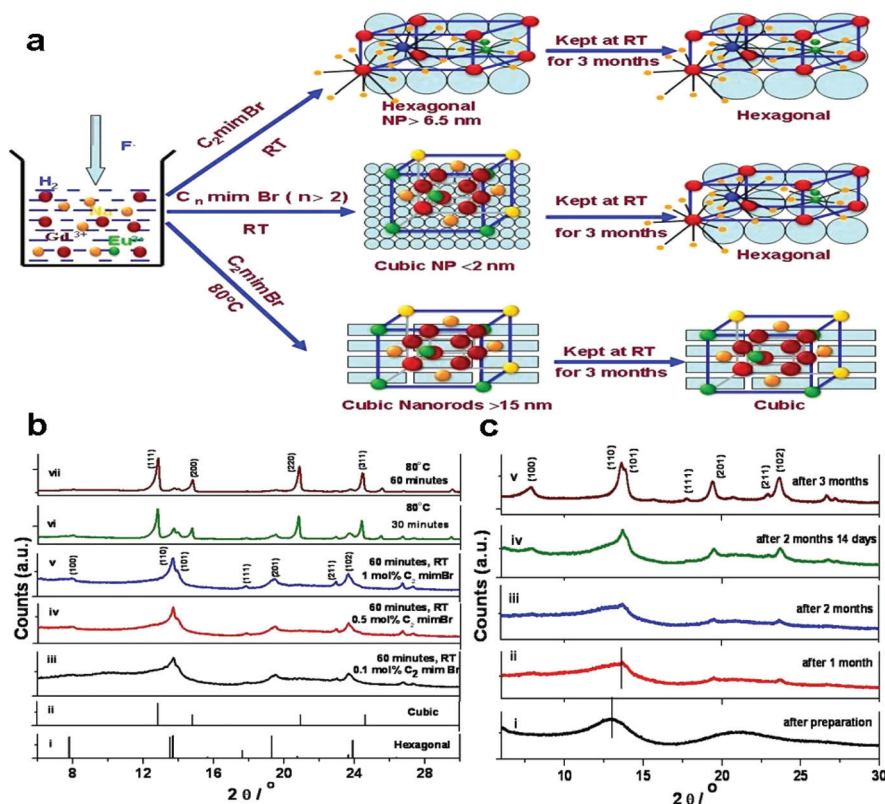


Fig. 1 (a) A general scheme illustrating the phase selective synthesis of Eu³⁺-doped NaGdF₄ nanocrystals employing ILs and their subsequent crystal phase behaviour. (b) PXRD pattern of NaGdF₄:Eu³⁺ nanocrystals prepared at different temperatures with different concentrations of IL; (i) hexagonal database pattern (JCPDS card no-27-699); (ii) cubic database pattern (JCPDS card no-27-697); (iii)–(v) doped NaGdF₄:Eu³⁺ prepared by 0.1, 0.5 and 1.0 mol% [C₂mim]Br respectively at RT and with 1 hour stirring; (vi) and (vii) NaGdF₄:Eu³⁺ samples prepared at 80 °C and with 0.1 mol% [C₂mim]Br for 30 minutes and NaGdF₄:Eu³⁺ samples prepared at a reaction time of 1 hour. (c) PXRD pattern of NaGdF₄:Eu³⁺ nanocrystals prepared with 1 mol% [C₄mim]Br at RT and 1 hour stirring time. PXRD is taken after different time intervals of preparation.

than that of the (101) reflection, indicating an anisotropic, oriented growth of the hexagonal structure. It may be concluded that [C₂mim]Br acts as a capping agent and interacts strongly with the (110) plane inhibiting its growth. Potential interactions could originate from π -stacking between aromatic rings and hydrogen bonding between the initial nuclei facets and the H atom in the C(2) position of [C₂mim]⁺. To understand the effect of hydrogen bonding, 1-ethyl-dimethyl-imidazolium bromide [C₂dmim]Br, in which C(2)–H is replaced by methyl, is used in a similar synthesis. Here also pure hexagonal material is obtained (Fig. S2a† and NP21 of Table 1) suggesting that the evolution of the hexagonal phase is not affected by H-bonding, but rather [C₂mim]⁺ ions are anchored on to the (110) plane *via* the π -system. To prove this, a quaternary ammonium IL, tetramethylammonium bromide ([Me₄N]Br) is used which leads to cubic NaGdF₄ (NP19 of Table 1). If [C₂mim]Cl is used instead of [C₂mim]Br, again the hexagonal phase (NP20 of Table 1) appears (Fig. S2b†) suggesting that indeed the π -system of the [C₂mim]⁺ cation is the key factor.

3.1.2 Effect of reaction temperature on phase evolution. A drastic change in phase evolution appears when the reaction temperature is changed from RT to 80 °C, with otherwise iden-

tical parameters. Here the formation of quite well crystallized particles of the cubic phase with only small impurities of the hexagonal phase already appears after 30 minutes. When the reaction time is prolonged, the cubic phase becomes even more dominant (Fig. 1b and S3†). If the reaction is conducted at 80 °C without the presence of a [C₂mim] IL, the cubic phase is also formed preferentially (Fig. S4†). This indicates that in the case of a fast reaction rate the less thermodynamically favoured NaGdF₄ polymorph is formed, which is in agreement with Ostwald's step rule. At room temperature the [C₂mim]⁺ cation apparently is able to interact with the NaGdF₄ nuclei, slowing down the reaction rate and favouring the formation of the thermodynamically stable hexagonal modification. At higher reaction temperatures the [C₂mim]⁺ cation can bind less efficiently to the NaGdF₄ surfaces and thus a mix of the two phases (cubic as the major one) appears. If the as-prepared hexagonal and cubic materials (NP1 and NP15 of Table 1) are heated to a higher temperature (400 °C, for 5 hours), the hexagonal material retains its phase, but the cubic converts to hexagonal. No phase transitions are observed on cooling. This undoubtedly proves the thermodynamic stability of the hexagonal over the cubic phase (for details see the ESI, Fig. S5 and 6, and Scheme S2†).



Table 1 Phase composition, crystallite size and reaction conditions to obtain nanocrystals using ILs. The crystallite size is estimated from the (110) diffraction peak of the hexagonal and (111) plane of the cubic material

Name of the sample		IL concentration (mol%)	Reaction temperature and time	Crystal phase	Crystallite size (nm)
NP1	NaGdF ₄ :Eu(2)	0.1 [C ₂ mim]Br	RT (1 hour)	Hexagonal	6.6 (±0.2) ^a
NP2	NaGdF ₄ :Eu(2)	0.5 [C ₂ mim]Br	RT (1 hour)	Hexagonal	9.8 (±0.2)
NP3	NaGdF ₄ :Eu(2)	1.0 [C ₂ mim]Br	RT (1 hour)	Hexagonal	11.1 (±0.3)
NP4	NaGdF ₄ :Eu(2) (measured after 3 months)	1.0 [C ₂ mim]Br	RT (1 hour)	Hexagonal	11.5 (±0.2)
NP5	NaGdF ₄ :Eu(2) (measured after preparation)	1.0 [C ₄ mim]Br	RT (1 hour)	Cubic	~1.5 (±0.2)
NP6	NaGdF ₄ :Eu(2) (measured after 1 month)	1.0 [C ₄ mim]Br	RT (1 hour)	Hexagonal	4.8 (±0.2)
NP7	NaGdF ₄ :Eu(2) (measured after 2 months)	1.0 [C ₄ mim]Br	RT (1 hour)	Hexagonal	7.6 (±0.2)
NP8	NaGdF ₄ :Eu(2) (measured after 2.5 months)	1.0 [C ₄ mim]Br	RT (1 hour)	Hexagonal	7.8 (±0.2)
NP9	NaGdF ₄ :Eu(2) (measured after 3 months)	1.0 [C ₄ mim]Br	RT (1 hour)	Hexagonal	8.2 (±0.2)
NP10	NaGdF ₄ :Eu(2) (measured after preparation)	1.0 [C ₈ mim]Br	RT (1 hour)	Cubic	~1.5 (±0.2)
NP11	NaGdF ₄ :Eu(2) (measured after 3 months)	1.0 [C ₈ mim]Br	RT (1 hour)	Hexagonal	12.6 (±0.2)
NP12	NaGdF ₄ :Eu(2) (measured after preparation)	1.0 [C ₁₀ mim]Br	RT (1 hour)	Cubic	~1.5 (±0.2)
NP13	NaGdF ₄ :Eu(2) (measured after 3 months)	1.0 [C ₁₀ mim]Br	RT (1 hour)	Hexagonal	7.1 (±0.2)
NP14	NaGdF ₄ :Eu(2)	0.1 [C ₂ mim]Br	80 °C (30 minutes)	Cubic	14.0 (±0.2)
NP15	NaGdF ₄ :Eu(2)	0.1 [C ₂ mim]Br	80 °C (1 hour)	Cubic	18.1 (±0.2)
NP16	NaGdF ₄ :Eu(2)	—	80 °C (1 hour)	Cubic	19 (±0.2)
NP17	NaGdF ₄ :Eu(2)	—	RT (1 hour)	Cubic	~1.5 (±0.2)
NP18	NaGdF ₄ :Eu(2) (measured after 3 months)	0.1 [C ₂ mim]Br	80 °C (1 hour)	Cubic	19.2 (±0.2)
NP19	NaGdF ₄ :Eu(2)	0.1 [Me ₄ N]Br	RT (1 hour)	Cubic	~1.5 (±0.2)
NP20	NaGdF ₄ :Eu(2)	1.0 [C ₂ mim]Cl	RT (1 hour)	Hexagonal	12.0 (±0.3)
NP21	NaGdF ₄ :Eu(2)	1.0 [C ₂ dmim]Br	RT (1 hour)	Hexagonal	10.0 (±0.3)

^a Estimated deviation, RT = room temperature, NaGdF₄:Eu(2) = NaGdF₄ doped with 2 mol% Eu³⁺.

3.1.3 Effect of alkyl chain length of IL and spontaneous phase transition. When the alkyl chain length at the C(1) position of the imidazolium ring is extended to four or more carbon atoms as in [C₄mim]Br, [C₈mim]Br or [C₁₀mim]Br, instead of getting hexagonal NaGdF₄ (as for [C₂mim]Br), cubic phased material of a comparatively low degree of crystallinity is obtained (Fig. 1c and S7†). The PXRD patterns are quite similar to those where no IL was used (Fig. S1†). This result clearly indicates that unlike [C₂mim]Br, imidazolium cations bearing longer alkyl chains have difficulty in interacting with the primary nuclei facets due to steric hindrance. Interestingly, the as-prepared samples (NP5) of the cubic phase convert completely to the hexagonal phase if kept at room temperature. Only one month after its preparation, reflections of the hexagonal phase start to appear (NP6), becoming more prominent after 2 months and 14 days (NP8) and completely crystalline hexagonal phase material is present after 3 months (Fig. 1c and NP9). Here the PXRD patterns are checked in regular intervals of 7 days.

Similar phenomena are observed for [C_nmim]Br, (*n* = 4, 8, 10) (Fig. S7†). This kind of spontaneous conversion from the cubic to pure hexagonal phase with time is new and previously undocumented to the best of our knowledge. However, the [C₂mim]Br mediated hexagonal (NP1–3 of Table 1) and the cubic phase material obtained from the reaction at elevated temperature (80 °C, NP15 of Table 1) did not undergo this phase change (Fig. S8†). When analyzing the crystallite size (Table 1), it is seen that there exists a critical size for the cubic to hexagonal phase transition at room temperature: prepared under similar conditions but with [C₄mim]Br, sample NP5 has an average crystallite size of ~1.5 nm. After 3 months, when conversion to the hexagonal phase has occurred, the crystal size becomes 8.2 nm. However, the cubic material obtained

(NP15) at a higher reaction temperature (80 °C) is much larger (18.1 nm) and does not change its phase with time. It is known that phase transformation in nanoparticles is governed by interface nucleation and surface energies.^{59–62} When the particle size is small, the surface energy is high which facilitates the transformation to the thermodynamically more stable phase. When the particles are larger the surface energy is less and additional activation energy is required for the phase transition. In summary, we observe that whenever the IL cation is able to attach to the surface, the reaction is slowed down to such an extent that the thermodynamically stable (hexagonal) polymorph forms, as seen for [C₂mim]Br.

In the case of high reaction rates, when the IL cannot interact *via* the π -system with the NaGdF₄ nuclei due to steric hindrance or elevated reaction temperatures, the thermodynamically less stable cubic polymorph forms, in accordance with Ostwald's step rule. When the particles are small, the surface energy is high and the particles easily undergo a phase transition from the less thermodynamically stable to the more stable form (see the scheme in Fig. 1a and 1c). However, when the particles are larger, activation energy must be provided to provoke the phase transition.

3.2 Lattice strain and phase evolution

Generally, the broadening of the diffraction peaks depends upon strain and crystallite size. It is possible to calculate the lattice strain according to the function derived by Williamson and Hall:⁶⁴

$$\beta \cos \theta / \lambda = 1/D + \eta \sin \theta / \lambda \quad (1)$$

where β is the full width at half maximum (fwhm), θ is the diffraction angle, λ is the X-ray wavelength, D is the effective



crystallite size, and η is the effective strain. When plotting $\beta \cos \theta / \lambda$ against $\sin \theta / \lambda$ the strain can be derived from the slope and the crystallite size (D) is given by the intercept of the plot (Fig. 2).⁶⁰ An extremely high compressive strain of -14.3% is observed for cubic $\text{NaGdF}_4\text{:Eu(2)}$ obtained from 1.0 mol\% $[\text{C}_4\text{mim}]\text{Br}$ at room temperature (sample NP5, Tables 1 and 2). According to both the Williamson–Hall and the Scherrer equation a crystallite size of $1\text{--}1.5 \text{ nm}$ is derived (Tables 1 and 2). When the sample is kept for 1 and 2 months at room temperature after its preparation (NP6 and NP7), the lattice strain decreases significantly to -5.42% and finally to -0.796% (Table 2) and a phase transition from cubic to hexagonal is observed. At the same time the crystallite size grows from 1.1 nm (for the sample measured after preparation) to 4.69 and 8.2 nm respectively (for the sample measured after one and two months respectively). After 2.5 months from the preparation, the lattice strain and crystallite size stops changing significantly (Table 2). A reversed lattice strain *i.e.* tensile strain ($+0.21\%$) is observed for the cubic material (NP15) obtained from 0.1 mol\% $[\text{C}_2\text{mim}]\text{Br}$ at 80°C (Fig. 2a and b). For 2 mol\% Eu^{3+} -doped hexagonal NaGdF_4 (NP1) prepared by 0.1 mol\% $[\text{C}_2\text{mim}]\text{Br}$ at room temperature, a small compressive strain of -1.26% is observed and the crystallite size is calculated as 8.44 nm , which is in good agreement with the value obtained from the Scherrer equation (Table 1).

3.3 Structural characterization by transmission electron microscopy

To understand the phase evolution from its atomistic origin, TEM images are analyzed in detail. Fig. 3a, b and S10† show

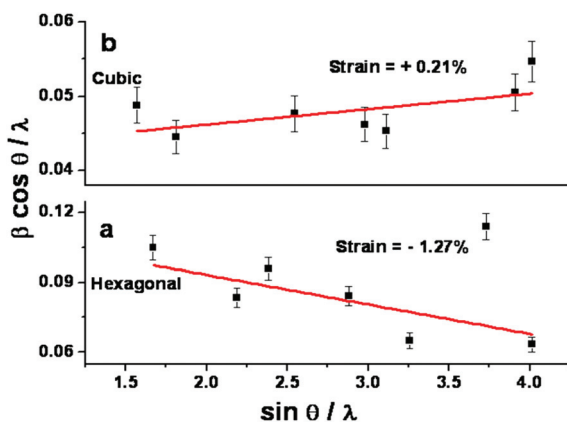


Fig. 2 Plot of $\beta \cos \theta / \lambda$ against $\sin \theta / \lambda$ for $\text{NaGdF}_4\text{:Eu(2)}$ prepared by 0.1 mol\% $[\text{C}_2\text{mim}]\text{Br}$ at room temperature (a) and at 80°C (b).

the low magnification images of NaGdF_4 nanocrystals prepared using various concentrations of $[\text{C}_2\text{mim}]\text{Br}$ (NP1–NP3 of Table 1). The nanostructures obtained appear to be built up by intergrown nanocrystals of slightly elongated shape. The average aspect ratio (particle length/particle diameter) increases slightly with increasing IL concentration. It appears that the IL evokes a preferential growth of NaGdF_4 crystal nuclei in certain directions by selectively attaching to growth facets, thus allowing the nanocrystals to undergo self-organization into the desired nanostructure, just as a soft template.⁴⁵

HRTEM images have shown that for rod-shaped NaGdF_4 nanocrystals, an imperfect oriented attachment *via* the low energy plane (110) among the multiple nanorods separated by a twin boundary occurs (Fig. 3d).⁶⁵ Normally twins are formed when coherence is achieved in the interfacial plane. A series of dislocations is observed due to small misorientation in the interface as indicated by arrow heads (Fig. 3d). This lattice mismatch arises when the surfaces of adjacent oriented particles are not atomically flat.

Importantly, the dominant exposed plane of hexagonal NaGdF_4 is consistently (110) (Fig. 3d and S11†). So it can be inferred that the $[\text{C}_2\text{mim}]^+$ ion serves as a capping agent based on the strong interaction with the (110) facets which favour the directional growth of the hexagonal phase. As the most exposed facets of the NaGdF_4 nanorods are the (110) planes or symmetrical equivalents, the energy difference between them is minimal, facilitating the oriented attachment. Though the crystal phase remains constant, the morphology changes if 1-ethyl-2,3-dimethylimidazolium bromide ($[\text{C}_2\text{dmim}]\text{Br}$) is used instead of $[\text{C}_2\text{mim}]\text{Br}$ (Fig. 3c). Instead of nanorods, spherical particles with an average diameter of 10 nm are obtained. It is assumed that in the case of $[\text{C}_2\text{mim}]\text{Br}$, a stronger directional interaction of the imidazole ring with the growth facets of hexagonal NaGdF_4 occurs, promoted through the acidic hydrogen atom in the 2-position of the imidazolium ring. When the latter is replaced by a methyl group, hydrogen bonding becomes weaker, the interaction between $[\text{C}_2\text{dmim}]$ cations and NaGdF_4 nuclei facets is less and NaGdF_4 can grow more isotropic, thus producing nanoparticles. The measured lattice spacings (Fig. 3e) of the $[\text{C}_2\text{dmim}]\text{Br}$ mediated sample belong to the (110) plane of the hexagonal phase, as further confirmed by the FFT (fast Fourier transform) pattern (Fig. 3f). However, a significant change in morphology along with the crystal phase is observed if the reaction temperature is raised from RT to 80°C , keeping the concentration of $[\text{C}_2\text{mim}]\text{Br}$ the same (Fig. S12e and f†). Flower-like nanoaggregates with nano-

Table 2 Lattice strain of $\text{NaGdF}_4\text{:Eu(2)}$ prepared from 1.0 mol\% $[\text{C}_4\text{mim}]\text{Br}$ at room temperature

$\text{NaGdF}_4\text{:Eu(2)}$ ($0.1 [\text{C}_4\text{mim}]\text{Br}$, RT)	Freshly prepared sample (NP5)	After 1 month (NP6)	After 2 months (NP7)	After 2.5 months (NP8)	After 3 months (NP9)
Crystal phase	Cubic	Hexagonal	Hexagonal	Hexagonal	Hexagonal
Lattice strain	-14.32% (compressive)	-5.42% (compressive)	-0.796% (compressive)	-1.01% (compressive)	-1.13% (compressive)
Crystallite size	1.09 nm	4.69 nm	8.24 nm	8.60 nm	8.62 nm



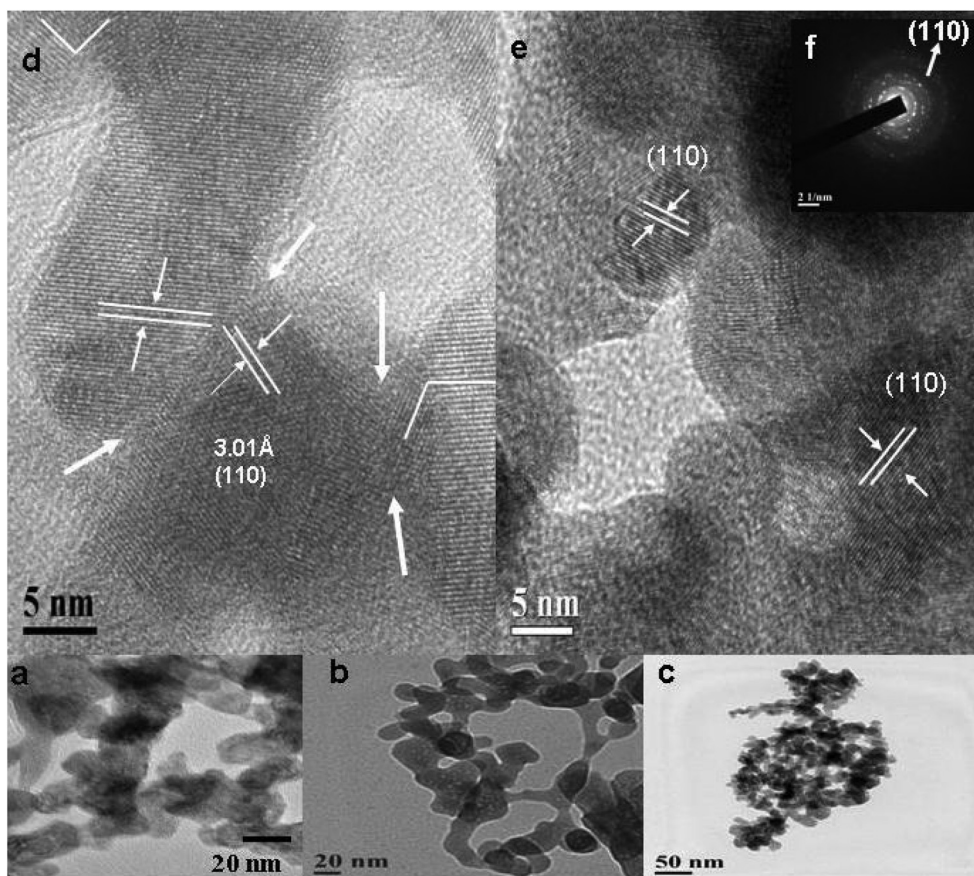


Fig. 3 (a) and (b) Low magnification TEM images of NaGdF₄:Eu³⁺ nanocrystals of the hexagonal phase prepared at room temperature (RT), in the presence of 0.1 and 1.0 mol% [C₂mim]Br, respectively; (d) HRTEM image of NaGdF₄:Eu³⁺ nanocrystals prepared at RT and in the presence of 1.0 mol% [C₂mim]Br; (c), (e) and (f) TEM, HRTEM and FFT pattern of hexagonal phase nanocrystals prepared at RT with 1 mol% [C₂dmim]Br.

discs and nanorods are observed, which on further heating agglomerate to sphere-like aggregates and complete spheres (Fig. S12g and h†) indicating that the IL no longer efficiently can shield against particle agglomeration.

Fig. 4a and b depict the low magnification images of the NaGdF₄ samples using [C₄mim]Br at room temperature (NP5 and NP9 of Table 1). Fig. 4a is measured immediately after preparation, whereas Fig. 4b is after 3 months. Here also spherical particles are obtained, despite the acidic 2-H proton of the imidazolium ring. It seems that the steric requirements of the butyl side chain lead to a reduced interaction of the imidazolium cation with the primary nuclei. When ILs with even longer alkyl chains such as [C₁₀mim]Br are used, the cation is no longer able to interact with the primary nuclei and the same results are obtained (Fig. S13†).

To understand the phase conversion over time from its atomistic origin, HRTEM images were analyzed carefully (Fig. 4c and d). The measured lattice spacing in Fig. 4c for the as-prepared samples (NP5) clearly belongs to the (200) plane of cubic NaGdF₄. The smallest particle size also is below 5 nm (considering the lattice fringes in the circled area of Fig. 4c). Interestingly, lattice planes in one region are not connected to others (e.g. area A and B or C and D in Fig. 4c).

However, HRTEM images measured after three months of the sample preparation (NP9) are entirely different (Fig. 4d). The average diameter of the nanocrystals has grown to ~15 nm and the measured lattice spacing belongs to the (110) plane of hexagonal NaGdF₄. Now the lattice planes are connected and defect structures can also be observed in some areas, as circled (Fig. 4d). The analysis clearly demonstrates that the particle undergoes Ostwald ripening with time, growing and transforming to the more thermodynamically stable form. The presence of proper elemental compositions was confirmed by energy dispersive X-ray analysis (EDXA) (Fig. S14†).

3.4 Optical characterizations

Fig. 5A, C and E show the room temperature photoluminescence (PL) spectra of NaGdF₄:Eu³⁺ nanocrystals in the hexagonal modification (NP1, NP2 and NP3 of Table 1) obtained with different concentrations of [C₂mim]Br under direct excitation of the ⁷F₀–⁵L₆ transition of the Eu³⁺ ion (λ_{ex} = 393 nm). (The details of the excitation spectra measured at room and liquid nitrogen temperature (77 K) are given in the ESI (Fig. S15 and 16)†). The most prominent emission bands appear at 615 and 592 nm which originate from the ⁵D₀–⁷F₂ (electric dipole, *e*) and ⁵D₀–⁷F₁ (magnetic dipole, *m*) transitions



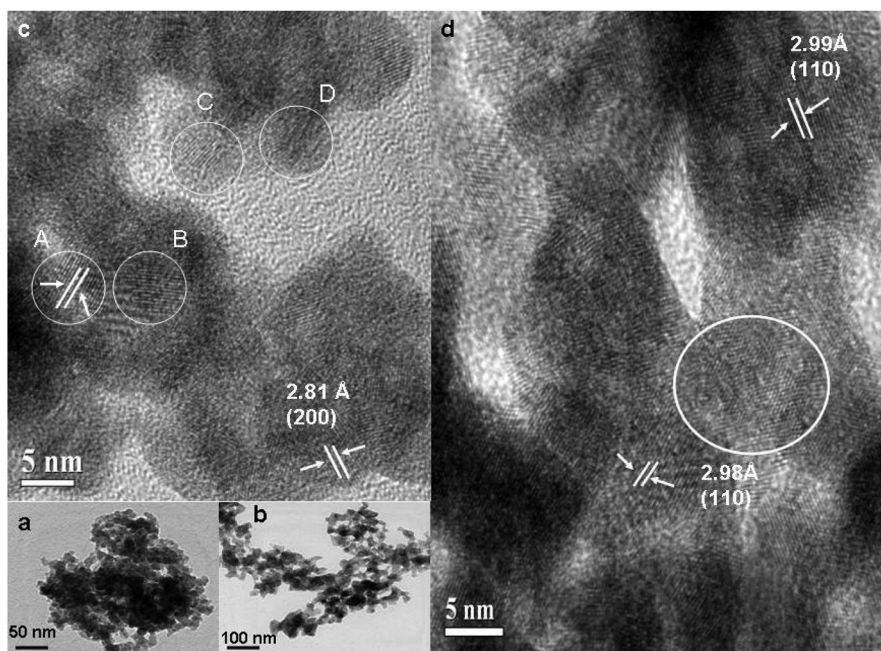


Fig. 4 (a) and (b) Low magnification TEM images of NaGdF₄:Eu³⁺ nanocrystals prepared at RT and in the presence of 1.0 mol% [C₄mim]Br (measured just after preparation and after 3 months); (c) and (d) HRTEM images of NaGdF₄:Eu³⁺ nanocrystals prepared at RT and in the presence of 1.0 mol% [C₄mim]Br (measured just after preparation and after 3 months).

respectively, in addition to transitions from higher ⁵D_J (*J* = 1–3) levels to the ⁷F manifold. The hypersensitive forced electric dipole transition is more intense than the magnetic dipole and increases relative to the magnetic dipole with increasing IL concentration. For better understanding and quantification, the respective asymmetry parameter (Ω_2) using Judd–Ofelt theory was calculated.^{66,67} It is seen that the J–O parameter (Ω_2) steadily increases with increasing IL concentration and points to attainment of an increasingly less symmetric environment for the Eu³⁺ ion (the J–O parameter (Ω_2) is 6.69×10^{-20} cm², 9.88×10^{-20} cm², and 12.29×10^{-20} cm² for 0.1, 0.5 and 1.0 mol% [C₂mim]Br IL respectively) as expected for an increasingly crystalline hexagonal material (for details see the ESI†). When the reaction temperature is raised from RT to 80 °C, cubic nanocrystals (NP15) form (Fig. 1b) and the nature of the spectrum also changes (Fig. 5B). Here the magnetic dipole becomes more prominent than the electric dipole transition. The lower Ω_2 value (4.29×10^{-20} cm²) of the cubic material compared to that of the hexagonal (6.69×10^{-20} cm²) indicates a higher site symmetry of the Eu³⁺ ion in the cubic phase which is expected from the crystal structure.

Fig. 5D and F represent the PL spectra recorded at –196 °C for the sample obtained with 0.5 and 1.0 mol% [C₂mim]Br at RT under direct excitation. Interestingly, the transitions from higher levels such as ⁵D₁ increase significantly compared to RT measurements, due to a decreased multiphonon relaxation between the higher ⁵D levels at lower temperatures. Luminescence lifetime measurements show that the ⁵D₁ lifetime increases by a factor of ~2 (from 1.77 to 3.82 ms) upon lower-

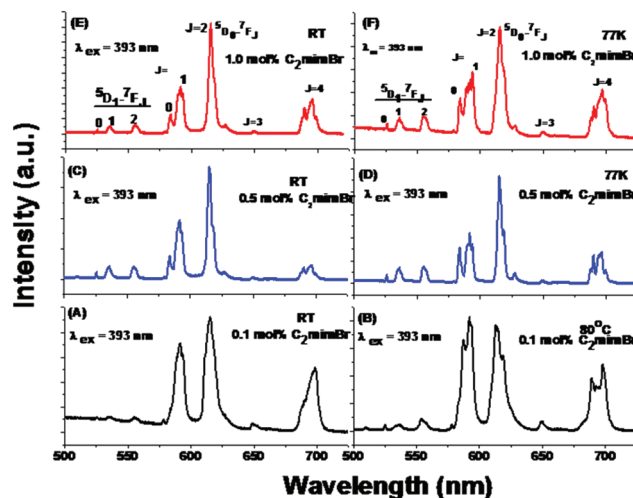


Fig. 5 (A) and (B) Emission spectra of NaGdF₄:Eu³⁺ nanocrystals measured at RT and prepared in the presence of 0.1 mol% [C₂mim]Br at RT and 80 °C. (C) and (E) Emission spectra for the 0.5 mol% and 1.0 mol% [C₂mim]Br mediated sample measured at RT. (D) and (F) Emission spectra of nanocrystals prepared in the presence of 0.5 mol% [C₂mim]Br at RT and 1.0 mol% [C₂mim]Br at RT (F), measured at –196 °C. (Excitation wavelength λ_{exc} = 393 nm).

ing the temperature from RT to –196 °C whereas the ⁵D₀ lifetime increases only by 3% (from 11.29 to 11.64 ms) under the same conditions (Fig. S17–S19†). The transitions from higher levels such as ⁵D₁ also decrease drastically when the



Eu^{3+} concentration is increased from 0.5 to 10 mol%, due to concentration quenching (Fig. S20†).

3.5 Understanding the spontaneous phase transition using Eu^{3+} ions as a spectroscopic probe

The photoluminescence properties of $\text{NaGdF}_4:\text{Eu}^{3+}$ synthesized in the presence of 1 mol% $[\text{C}_4\text{mim}]\text{Br}$ change as expected with time (Fig. 6). PL measurements immediately after sample preparation (NP5) show a slightly stronger magnetic dipole transition compared to the electric dipole suggesting high site symmetry of the Eu^{3+} dopant ion. When samples are measured one, two and three months after preparation, the electric dipole transition increases gradually (see the e/m ratio in Fig. 6), indicating the cubic to hexagonal phase transformation. Likewise, the calculated Judd–Ofelt (Ω_2) parameters gradually increase with time, reaching a maximum for samples measured 3 months after preparation. The Ω_2 values are $6.45 \times 10^{-20} \text{ cm}^2$, $6.97 \times 10^{-20} \text{ cm}^2$, and $9.79 \times 10^{-20} \text{ cm}^2$ for samples measured one, two and three months after preparation (ESI†). This confirms the attainment of highly asymmetric site symmetry for Eu^{3+} as expected for a hexagonal material. Similar trends are observed for the samples prepared in the presence of $[\text{C}_8\text{mim}]\text{Br}$ and $[\text{C}_{10}\text{mim}]\text{Br}$ (Fig. S21†). As expected, no change in PL with time is noticed for the $[\text{C}_2\text{mim}]\text{Br}$ mediated hexagonal samples as its phase remains unchanged over time (Fig. S22†). Thus, the Eu^{3+} ion has been successfully used as a spectroscopic probe to monitor the phase transformation.

3.6 Vacuum-UV spectroscopy using synchrotron radiation

In the vacuum-UV (VUV) excitation spectrum (Fig. S23†) the $\text{Eu}^{3+}\text{--F}^-$ charge transfer (CT) band below 180 nm is apparent but, most importantly, no $\text{Eu}^{3+}\text{--O}^{2-}$ CT band is observed.¹⁰

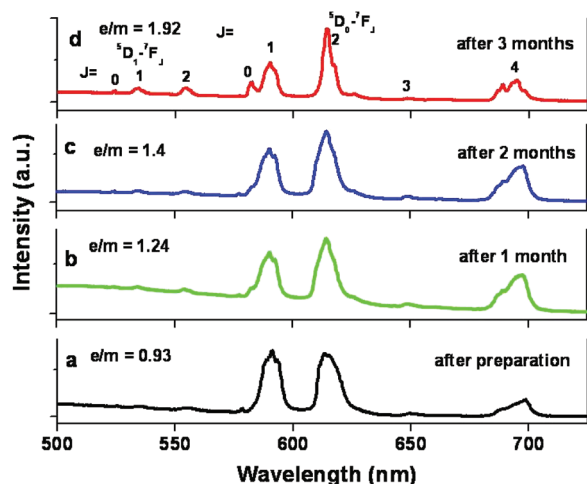


Fig. 6 Understanding of the spontaneous cubic-to-hexagonal phase transition using Eu^{3+} as a spectroscopic probe. (a)–(d) Emission spectra of $\text{NaGdF}_4:\text{Eu}^{3+}$ nanocrystals measured at RT and prepared with 1 mol% $[\text{C}_4\text{mim}]\text{Br}$ upon excitation at 393 nm (right after preparation, after one month, after 2 months and after 3 months respectively).

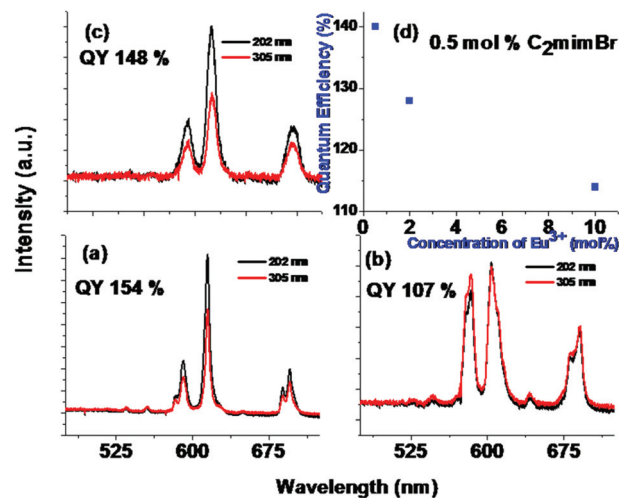


Fig. 7 Measurements of quantum cutting using synchrotron radiation. (a) and (b) Emission spectra of $\text{NaGdF}_4:\text{Eu}(2)$ nanocrystals prepared at RT in the presence of 1 mol% $[\text{C}_2\text{mim}]\text{Br}$ and at 80 °C in the presence of 0.1 mol% $[\text{C}_2\text{mim}]\text{Br}$. (c) At RT in the presence of 1 mol% $[\text{C}_{10}\text{mim}]\text{Br}$ (after 3 months of preparation). (d) The effects of the concentration of Eu^{3+} on the quantum efficiency for the hexagonal samples prepared in the presence of 0.5 mol% $[\text{C}_2\text{mim}]\text{Br}$. The samples are excited in the $^6\text{G}_J$ levels of Gd^{3+} at 202 nm (black) and in the $^6\text{P}_J$ levels at 305 nm (red).

This confirms the absence of oxygen impurities in the sample. The emission spectra of pure hexagonal and cubic $\text{NaGdF}_4:\text{Eu}^{3+}$ nanocrystals recorded with high intensity VUV synchrotron radiation are depicted in Fig. 7. Upon excitation into the $^6\text{G}_J$ level (202 nm), quantum cutting through a two-step energy transfer can occur. In the first step, cross relaxation $\text{Gd}^{3+}(^6\text{G}_J), \text{Eu}^{3+}(^7\text{F}_0) \rightarrow \text{Gd}^{3+}(^6\text{P}_J), \text{Eu}^{3+}(^5\text{D}_0)$ generates one photon and, in the second step, energy is transferred from the $^6\text{P}_J$ level of Gd^{3+} to a higher energy level of Eu^{3+} , yielding an emission with a normal branching ratio for the different $^5\text{D}_J$ emission lines.¹

As the first energy transfer step only gives emission from the $^5\text{D}_0$ level, a significant increase of its relative intensity upon excitation in the $^6\text{G}_J$ level is expected if quantum cutting happens. When excited by 305 nm ($^6\text{P}_J$), a single energy transfer step to Eu^{3+} occurs with a normal branching ratio (Fig. S24†). Comparing the PL spectra of different incident wavelengths, it is seen that the intensity of the $^5\text{D}_0$ emission line is significantly higher for excitation in the $^6\text{G}_J$ levels than the $^6\text{P}_J$ levels, confirming the occurrence of quantum cutting for the hexagonal samples (Fig. 7 and Table S1†). The intensity ratios of $^5\text{D}_0/^5\text{D}_{1,2,3}$ were calculated both for the $^6\text{G}_J$ and $^6\text{P}_J$ level excitation for the hexagonal (Fig. 7a and c and NP3 and NP13 of Table 1) and the efficiencies of the cross relaxation step are determined to be 54% and 48%, respectively.¹ Consequently, high visible quantum efficiencies of 154% and 148% are achieved. However, for the cubic sample (Fig. 7b, NP15 of Table 1), the $^5\text{D}_0$ emission line for both the $^6\text{G}_J$ and $^6\text{P}_J$ excitations is almost similar indicating little to no quantum cutting (107%). Detrimental effects of a high dopant concentration on quantum cutting are also witnessed although the crystal phase is the same (Fig. 7d). Such as for $\text{NaGdF}_4:\text{Eu}(10)$,

only a 114% efficiency is observed although the phase is still hexagonal.

4. Conclusions

In summary, we have succeeded in preparing efficient quantum cutting NaGdF₄:Eu nanocrystals useful for energy efficient lighting, using ILs as nanosynthetic templates at room temperature. To the best of our knowledge, it is the simplest method ever reported for the synthesis of rare earth ion-doped ternary fluorides which can be scaled up easily. Despite its simplicity in synthesis, very small nanomaterials below 10 nm in particle size (sometimes ~2 nm) are obtained. Analysis suggests that [C₂mim]⁺ ions anchor to the (110) plane of the primary nuclei, leading to reduced reaction rates and formation of the thermodynamically stable hexagonal polymorph through an imperfect oriented attachment mechanism. However, a less thermodynamically stable cubic polymorph is obtained for ILs with higher alkyl chain lengths such as [C₄mim]Br, [C₈mim]Br and [C₁₀mim]Br. As initially extremely small particles with an extraordinarily high surface energy form, the crystal phase completely changes to the hexagonal form spontaneously over an extended time at room temperature. The cubic material obtained is also characterized by a high lattice strain which gets lessened after the phase transition and particle growth. To the best of our knowledge, this kind of spontaneous phase transition at room temperature is not yet documented. Judicious choice of nucleation temperature aside from the ionic liquid is also mandatory, as it not only controls the morphology of the nanoparticles, but also tunes the crystal phase, which is essential for optical property tuning. Only at low temperatures (RT) IL can reveal its true power in directing the phase transition.

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