



Cite this: *Nanoscale*, 2016, 8, 2832

Characterization of multifunctional β -NaEuF₄/NaGdF₄ core–shell nanoparticles with narrow size distribution†

Lilli Schneider,^{*a} Thorben Rinkel,^b Benjamin Voß,^b Artur Chrobak,^c Johann P. Klare,^a Jan Neethling,^d Jaco Olivier,^d Dominik Schaniel,^e El-Eulmi Bendeif,^e Federica Bondino,^f Elena Magnano,^f Igor Piš,^{f,g} Kamil Balinski,^a Joachim Wollschläger,^a Heinz-Jürgen Steinhoff,^a Markus Haase^b and Karsten Kuepper^{*a}

The properties of β -NaEuF₄/NaGdF₄ core–shell nanocrystals have been thoroughly investigated. Nanoparticles with narrow size distribution and an overall diameter of ~22 nm have been produced with either small β -NaEuF₄ cores (~3 nm diameter) or large β -NaEuF₄ cores (~18 nm diameter). The structural properties and core–shell formation are investigated by X-ray diffraction, transmission electron microscopy and electron paramagnetic resonance, respectively. Optical luminescence measurements and X-ray photoelectron spectroscopy are employed to gain information about the optical emission bands and valence states of the rare earth constituents. Magnetic characterization is performed by SQUID and X-ray magnetic circular dichroism measurements at the rare earth M_{4,5} edges. The characterization of the core–shell nanoparticles by means of these complementary techniques demonstrates that partial intermixing of core and shell materials takes place, and a significant fraction of europium is present in the divalent state which has significant influence on the magnetic properties. Hence, we obtained a combination of red emitting Eu³⁺ ions and paramagnetic Gd³⁺ ions, which may be highly valuable for potential future applications.

Received 6th October 2015,
Accepted 27th December 2015

DOI: 10.1039/c5nr06915g

www.rsc.org/nanoscale

1 Introduction

NaREF₄ nanoparticles (RE: rare earth) have attracted increasing interest in recent years for applications in biological labeling and imaging,^{1–20} photodynamic therapy and drug delivery,^{21–26} photovoltaics,^{27–30} photonics^{31–33} and security labeling.^{34–36} Frequently, nanocrystals with core–shell structure are synthesized, since a shell of inert material increases

the quantum yield of a luminescent particle core by decreasing energy losses on the particle surface.^{37–39} Particles containing one or more shells of the doped material furthermore allow us to spatially separate the dopant ions with different functionalities and to control the migration of energy within such structures.^{40–49} For lanthanide ions emitting in the red and near infrared spectral regions, fluoride host materials are of particular interest because the very low phonon frequencies of fluoride lattices decrease the probabilities of multiphonon relaxation processes, resulting in high quantum yields. One example is the Eu³⁺ ion, displaying pure red emission in many host materials, because all transitions except those from the ⁵D₀ level are quenched due to multiphonon relaxation. In the fluoride host, however, also transitions from the higher ⁵D₁, ⁵D₂ and ⁵D₃ levels are observed provided that the concentration of the Eu³⁺ dopant is low. At higher concentrations, however, cross-relaxation between adjacent Eu³⁺ in the lattice leads to quenching of the ⁵D₁, ⁵D₂ and ⁵D₃ transitions in these materials, too. Trivalent gadolinium, on the other hand, is characterized by a relatively large energy gap between the ground state (⁸S_{7/2}) and the lowest excited level (⁶P_{7/2}).²¹ Importantly, this excited level is situated in the ultraviolet spectral

^aDepartment of Physics, Center of Physics and Chemistry of New Materials, Universität Osnabrück, Barbarastr. 7, 49076 Osnabrück, Germany.
E-mail: lischroe@uos.de, kkuepper@uos.de

^bInstitute of Chemistry of New Materials, Center of Physics and Chemistry of New Materials, Universität Osnabrück, Barbarastr. 7, 49076 Osnabrück, Germany

^cInstitute of Physics, Silesian University, Uniwersytecka 4, 40-007 Katowice, Poland

^dCentre for HRTEM, Nelson Mandela Metropolitan University, P.O. Box 77000, Port Elizabeth, South Africa

^eUniversité de Lorraine, CRM2, UMR 7036, Vandoeuvre-les-Nancy, F-54506, France

^fIOM CNR, Laboratorio TASC, S.S. 14 km 163.5, 34149 Basovizza, Trieste, Italy

^gElettra-Sincrotrone Trieste S.C.p.A., S.S. 14 km 163.5, 34149 Basovizza, Trieste, Italy

†Electronic supplementary information (ESI) available. See DOI: 10.1039/C5NR06915G



region, *i.e.* well above the emitting 5D_0 , 5D_1 , 5D_2 and 5D_3 levels of Eu^{3+} . A NaGdF_4 shell around a NaEuF_4 core can therefore improve the luminescence efficiency of the core. Moreover, the lattice mismatch between the core and the shell is very low due to the very similar lattice constants of NaGdF_4 and NaEuF_4 . In addition to their interesting optical properties, europium and gadolinium have remarkable magnetic properties. Owing to their seven unpaired electrons, Gd^{3+} ions in the ground state can, for instance, efficiently alter the relaxation times of the surrounding water protons and are therefore the basis of the most commonly used contrast agents in magneto resonance imaging.⁵⁰ While gadolinium ions usually form the oxidation state III+, europium compounds can contain either Eu^{3+} or Eu^{2+} ions, or both. Trivalent europium in the ground state (7F_0 , $S = 3$, $L = 3$, $J = 0$) has no magnetic moment while divalent europium has seven unpaired electrons in the ground state ($^8S_{7/2}$, $S = 8$, $L = 0$, $J = 7/2$). Interestingly in some trivalent compounds formation of divalent europium states on the surface was observed.^{51,52} Moreover, Kachkanov *et al.* detected an optically induced magnetic moment of Eu^{3+} ions in GdN .⁵³

In this work, we synthesized core-shell systems composed of a $\beta\text{-NaEuF}_4$ core and a $\beta\text{-NaGdF}_4$ shell. This structure has two advantages. Firstly, the influence of different surrounding media on the $\beta\text{-NaEuF}_4$ core can be minimized or entirely removed. Secondly, the combination of red emitting Eu^{3+} ions and paramagnetic Gd^{3+} ions is highly valuable.

The nominal size is the expected size based on the used materials respectively precursor particles. We use a number of complementary methods to evaluate the structure of $\beta\text{-NaEuF}_4/\text{NaGdF}_4$ core-shell particles and to study their electronic and magnetic properties. Detailed knowledge of these properties is indispensable to compare the results with those of Eu^{3+} doped $\beta\text{-NaGdF}_4$ nanoparticles and to further optimize the core-shell system in question. Structural properties and core-shell formation are determined by means of X-ray powder diffraction (XRD) and high resolution transmission electron microscopy (HR-TEM). Core-shell formation is further investigated by electron paramagnetic resonance (EPR) and optical luminescence spectroscopy. Chemical and electronic properties with special emphasis on the Eu and Gd valence states are determined employing X-ray photoelectron spectroscopy (XPS), whereas magnetic characterization is performed with a superconducting quantum interference device (SQUID) magnetometer. The measured magnetization curves are analyzed by means of Monte Carlo based Heisenberg simulations. Finally we performed element specific X-ray magnetic circular dichroism (XMCD) at the Eu $M_{4,5}$ and Gd $M_{4,5}$ edges to gain a complete picture of the internal magnetic structure of the multifunctional $\beta\text{-NaEuF}_4/\text{NaGdF}_4$ core-shell nanoparticles.

2 Results and discussion

2.1 Total scattering measurement and PDF analysis

Fig. 1 shows the pair distribution functions (PDFs) for both core-shell samples as well as the 22 nm $\beta\text{-NaGdF}_4$ nanocrystals

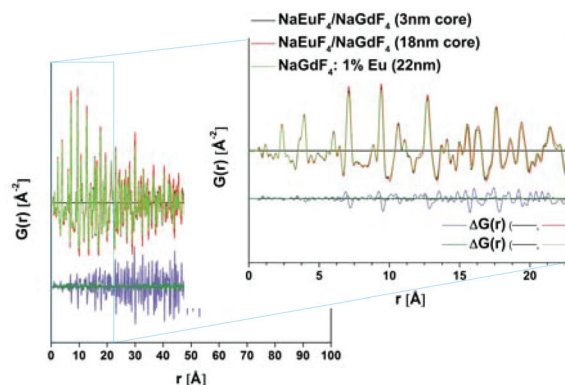


Fig. 1 Comparison of atomic PDFs in real space, $G(r)$ for 20 nm $\beta\text{-NaEuF}_4/\text{NaGdF}_4$ core-shell nanoparticles with a 3 nm $\beta\text{-NaEuF}_4$ core and ~ 8.5 nm $\beta\text{-NaGdF}_4$ shell, 22 nm $\beta\text{-NaEuF}_4/\text{NaGdF}_4$ core-shell nanoparticles (NPs) with an 18 nm $\beta\text{-NaEuF}_4$ core and 22 nm $\beta\text{-NaGdF}_4$ core-shell nanoparticle (NP) doped with 1% Eu and their differential PDFs.

doped with 1% Eu. The two samples containing low Eu and high Gd concentration (20 nm $\beta\text{-NaEuF}_4/\text{NaGdF}_4$ core-shell nanoparticles (NPs) with a 3 nm $\beta\text{-NaEuF}_4$ core, 22 nm $\beta\text{-NaGdF}_4$ nanoparticle (NP) doped with 1% Eu) exhibit very similar PDFs that differ significantly from the core-shell sample with the 18 nm $\beta\text{-NaEuF}_4$ core. Comparing the two core-shell samples (20 nm $\beta\text{-NaEuF}_4/\text{NaGdF}_4$ core-shell nanoparticles with a 3 nm $\beta\text{-NaEuF}_4$ core, 22 nm $\beta\text{-NaEuF}_4/\text{NaGdF}_4$ core-shell nanoparticles with an 18 nm $\beta\text{-NaEuF}_4$ core) we can clearly identify shifts of the peak positions in the PDF (corresponding to atom-atom distances). Note that the structures of bulk $\beta\text{-NaEuF}_4$ and $\beta\text{-NaGdF}_4$ are quite similar and the differences in bond distances between the different atoms are thus expected to be small. In order to enhance the visibility of the structural differences we calculated the differences between the corresponding PDFs, given as $\Delta G(r)$ in Fig. 1. One can clearly see that the difference between the PDF of the 20 nm $\beta\text{-NaEuF}_4/\text{NaGdF}_4$ core-shell nanoparticles with a 3 nm $\beta\text{-NaEuF}_4$ core and the 22 nm $\beta\text{-NaGdF}_4$ NP doped with 1% Eu is very small, resulting in very weak features in the $\Delta G(r)$. A core-shell nanoparticle with a small $\beta\text{-NaEuF}_4$ core thus exhibits almost the same structure as a nanoparticle doped with 1% Eu, indicating that the structural relaxation is similar for these two cases. In other words, from the PDF analysis alone it is impossible to distinguish between the structure of the core-shell NP and a homogeneously doped NP. In contrast, the difference of the two PDFs ($\Delta G(r)$) of the two core-shell samples is much more pronounced. Almost all peak positions are affected and the shifts are more pronounced for larger atom-atom distances r . In order to investigate this result in more detail we performed simulations of the PDFs based on pure crystalline, spherical 22 nm $\beta\text{-NaGdF}_4$ and $\beta\text{-NaEuF}_4$ nanocrystals. The results are shown in Fig. 2. The distances are slightly shorter in the $\beta\text{-NaGdF}_4$ nanoparticle as expected from the smaller lattice parameters. We observe a good agreement between the simulated peak positions in the PDF for the



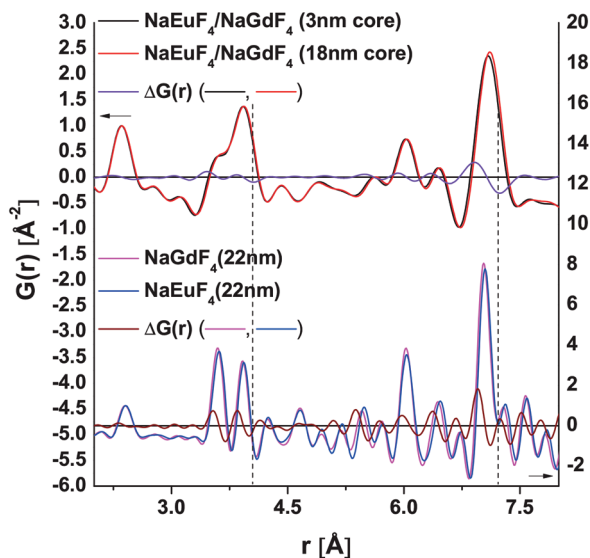


Fig. 2 Comparison of observed PDFs for 20 nm β -NaEuF₄/NaGdF₄ core-shell nanoparticles with a 3 nm β -NaEuF₄ core and \sim 8.5 nm β -NaGdF₄ shell and 22 nm β -NaEuF₄/NaGdF₄ core-shell NPs with an 18 nm β -NaEuF₄ core and calculated PDFs of 22 nm β -NaGdF₄ and β -NaEuF₄ spheres with the corresponding differential PDFs (spectra are normalized to the first peak).

two pure nanoparticles and the experimentally determined PDF. The peaks are broader in the experimental PDF, which is due to a lower resolution of the experiment, but also due to the fact that in the core-shell nanoparticles we expect some strain and defects as well as a not perfectly spherical shape. We don't expect a significant impact of the particle-size distribution on the PDF since the size distributions are rather narrow, *e.g.*, the β -NaEuF₄/NaGdF₄ (3 nm core) nanoparticles exhibit a size distribution of 16.8 nm \pm 5% (see Table 1), corresponding to particles ranging from 16 nm to 17.7 nm. If the structural difference (*e.g.* distances between first neighbours) between two such nanoparticles is well below the resolution of 0.138 Å, there will be no visible effect on the experimental PDF. For illustration one can compare the two simulated PDFs for the 22 nm β -NaGdF₄ and β -NaEuF₄ particles given in Fig. 2, where the difference in the first neighbour

Table 1 Sample label, nominal size and size distribution determined by transmission electron microscopy (TEM) analysis (Fig. S2 and S4)

Samples	Nominal size	Size (from TEM)
NaEuF ₄ /NaGdF ₄ (3 nm core)	20 nm	16.8 nm \pm 5%
NaEuF ₄ /NaGdF ₄ (18 nm core)	22 nm	23 nm \pm 3.8%
NaGdF ₄ :1% Eu	22 nm	25.4 nm \pm 4%
NaEuF ₄ (3 nm)	3 nm	Not evaluable
NaEuF ₄ (18 nm)	18 nm	19.7 nm \pm 4.2%
NaGdF ₄ /NaEuF ₄ (3 nm core)	20 nm	20.2 nm \pm 3.9%
NaEuF ₄ :1% Gd	22 nm	24.5 nm \pm 5.4%
NaGdF ₄ (3 nm)	3 nm	Not evaluable
NaGdF ₄ (22 nm)	22 nm	25.6 nm \pm 5.8%

distances between the pure Gd and the pure Eu sample is of the order of 0.02 Å, which is at the limit of visibility in the $G(r)$. Structural differences between two identical nanoparticles differing only by 5% in size would therefore not be detected in the experiment. As a first conclusion we can state that a β -NaEuF₄/NaGdF₄ core-shell nanoparticle with a 3 nm β -NaEuF₄ core exhibits approximately the same structure as a slightly Eu-doped β -NaGdF₄ nanoparticle and its main structural parameters are those of the β -NaGdF₄ structure. β -NaEuF₄/NaGdF₄ core-shell nanoparticles with an 18 nm β -NaEuF₄ core on the other hand resemble the corresponding pure β -NaEuF₄ structure. Due to the limited resolution of the experiment a more detailed modelling, *e.g.* to investigate the core-shell interface, is currently not possible. High-resolution electron microscopy investigations were carried out for further characterisation of the nanoparticles.

2.2 TEM-HRTEM

The morphology of the nanoparticles was characterized by transmission electron microscopy (TEM). It can be seen that the particles are nearly spherical or slightly elongated and retain the initial morphology of the precursor core particles (Fig. 3). Inter-planar spacings were measured using fast Fourier transform (FFT) analysis of the HRSTEM images. Inter-planar spacings with distances 0.29 nm, 0.18 nm and 0.16 nm were measured. These match very closely to the theoretical values for the (1,1,0), (0,0,2) and (1,1,2) planes respectively. Because the lattice constants between the core and the shell are similar, no significant differences are observed. Fig. 4(a)

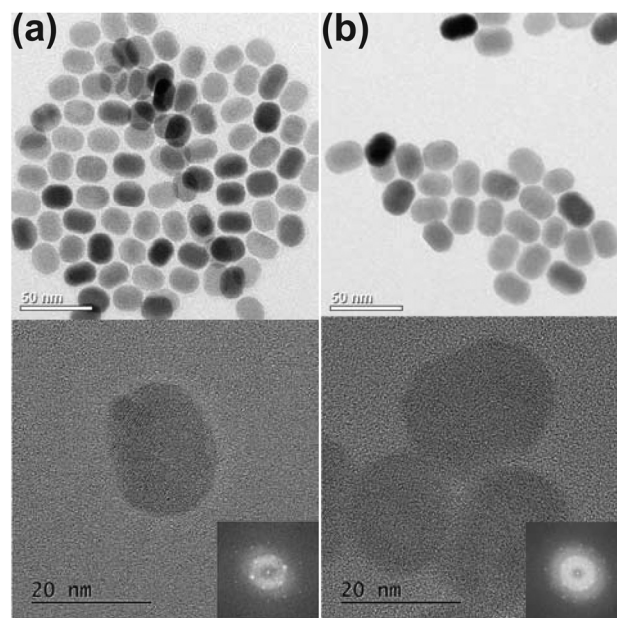


Fig. 3 TEM images showing (top) general morphology and (bottom) high-resolution images of the (a) 18 nm β -NaEuF₄ NP and (b) 22 nm β -NaEuF₄/NaGdF₄ NP with a 2 nm β -NaGdF₄ shell around the 18 nm β -NaEuF₄ core. The insets show the fast Fourier transforms (FFTs) of HR-TEM images.



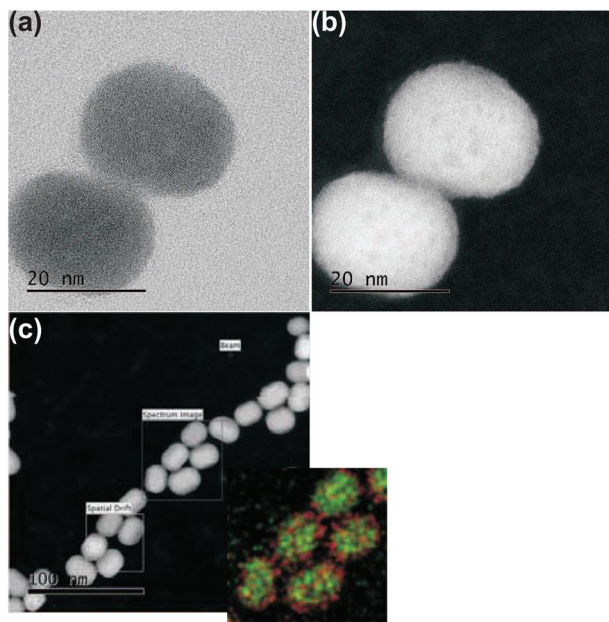


Fig. 4 High resolution scanning transmission electron microscopy bright field (HRSTEM-BF) image of (a) 22 nm β -NaEuF₄/NaGdF₄ core-shell NPs with an 18 nm β -NaEuF₄ core and 2 nm β -NaGdF₄ shell and (b) the corresponding high-angle annular dark field (HAADF) high resolution scanning transmission electron microscope (HRSTEM) image. And (c) scanning transmission electron microscope-electron energy loss spectroscopy (STEM-EELS) survey image and colour map of Eu (green) and Gd (red).

shows the high-resolution bright field scanning transmission electron microscopy (BF STEM) images of the core-shell nanoparticles with an 18 nm β -NaEuF₄ core and a 2 nm β -NaGdF₄ shell and (b) the corresponding high-angle annular dark field (HAADF) image.

The HAADF HRSTEM image of the 22 nm β -NaEuF₄/NaGdF₄ core-shell NPs with an 18 nm β -NaEuF₄ core and 2 nm β -NaGdF₄ shell in Fig. 4(b) does not show significant variation within the nanoparticle.

The STEM-EELS mapping of the 22 nm β -NaEuF₄/NaGdF₄ core-shell NPs with an 18 nm β -NaEuF₄ core and 2 nm β -NaGdF₄ shell (Fig. 4(c)) shows according to the colour map (inset) evidence of a core shell structure with Eu (green) in the center and Gd (red) on the surface.

2.3 Luminescence

The optical properties of the β -NaEuF₄ core and β -NaEuF₄/NaGdF₄ core-shell nanocrystals were studied by photoluminescence spectroscopy. As expected, excitation at the ${}^7F_{0,1} \rightarrow {}^5L_6$ transition of Eu³⁺ at ~ 394 nm results in Eu³⁺ emission for all nanoparticles containing Eu³⁺ ions (Fig. 5). In all cases the highest emission intensities are recorded at ~ 591 , 614 and 696 nm corresponding to the ${}^5D_0 \rightarrow {}^7F_1$, ${}^5D_0 \rightarrow {}^7F_2$ and ${}^5D_0 \rightarrow {}^7F_4$ intra-4f shell transitions of Eu³⁺ respectively.⁵⁴ These transitions are characteristic for luminescent materials containing Eu³⁺ ions and are, for instance, also observed for

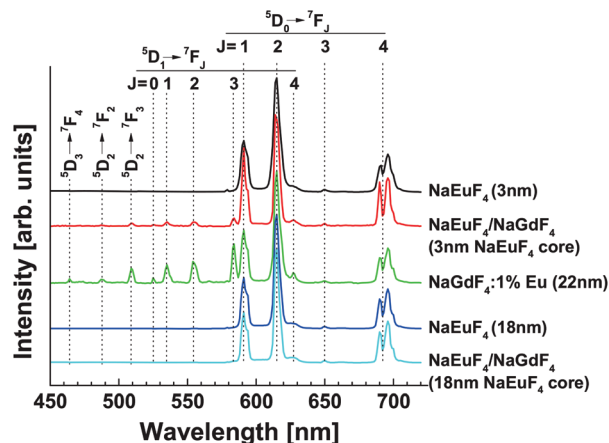


Fig. 5 Normalized emission spectra of 3 nm β -NaEuF₄, 20 nm β -NaEuF₄/NaGdF₄ NPs with a 3 nm core, 22 nm β -NaGdF₄ NP doped with 1% Eu, 18 nm β -NaEuF₄ and 22 nm β -NaEuF₄/NaGdF₄ NPs with an 18 nm β -NaEuF₄ core. The spectra were recorded under 394 nm excitation of the Eu³⁺ ions (${}^7F_{0,1} \rightarrow {}^5L_6$). All particles consist of hexagonal phase (β -phase).

the β -NaGdF₄ nanocrystals doped with 1 mol% of Eu³⁺. Fig. 5 shows, however, that the emission spectrum of such weakly doped particles displays additional lines between 450 nm and 580 nm, caused by transitions from the higher excited 5D_3 , 5D_2 and 5D_1 levels of Eu³⁺ to its 7F_J ground state multiplet. These emission lines are very weak in materials containing Eu³⁺ in high concentrations, since cross-relaxation between adjacent Eu³⁺ ions is known to quench the luminescence from these states. The 5D_3 , 5D_2 and 5D_1 emission lines are therefore characteristic for Eu³⁺ ions having no Eu³⁺ neighbors in the crystal lattice and display negligible intensity, for instance, in the case of β -NaEuF₄ core particles (Fig. 5).^{55,56} Fig. 5 shows, however, that after the formation of the NaGdF₄ shell, the 5D_3 , 5D_2 and 5D_1 emission lines appear in the spectrum of the core-shell particles. This indicates that some Eu³⁺ ions are released from the β -NaEuF₄ core and incorporate into NaGdF₄ during formation of the shell. The figure also shows that the transitions from the higher 5D states are much less intense for core-shell particles with 18 nm cores indicating that the number of Eu³⁺ ions released decreases with increasing core size. This observation is in accord with the lower surface-to-bulk ratio of 18 nm core particles compared to 3 nm core particles, since the release of Eu³⁺ ions should be proportional to the total surface area of all core particles. To further substantiate this result, we have investigated the core-shell particles containing Gd³⁺ by EPR spectroscopy.

2.4 Electron paramagnetic resonance spectroscopy

In general, the EPR spectrum of Gd³⁺ ions ($S = 7/2$) is characterized by the Zeeman interaction with a g -value ≈ 2 and exhibits significant contributions from zero-field splitting. The mean distance between two adjacent Gd³⁺ ions in the crystal lattice is reflected in the broadness of the EPR spectrum that



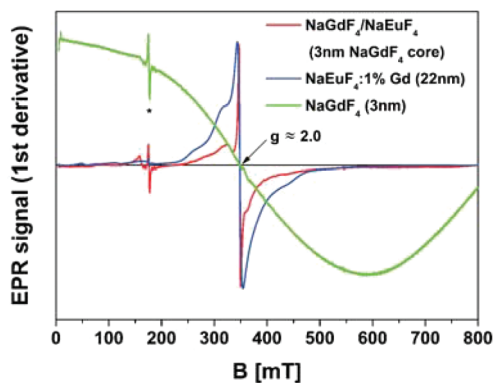


Fig. 6 Normalized EPR spectra of 20 nm core-shell particles composed of a 3 nm β -NaGdF₄ core and an \sim 8.5 nm β -NaEuF₄ shell (red), 22 nm β -NaEuF₄ NPs doped with 1% Gd (blue) and 3 nm β -NaGdF₄ precursor particles (green). Due to the large width of the EPR spectrum of the precursor particles it has a significantly lower intensity compared to the other spectra and consequently displays more noise when the spectra are normalized to the same signal height (scaling factors: blue, x1; red, x10; green, x11). The asterisk denotes an impurity signal of the microwave cavity present in all spectra at 175 mT.

increases with increasing Gd concentration due to dipolar and isotropic exchange interactions between the ions, as observed in Fig. 6 for 3 nm β -NaGdF₄ particles (green spectrum).⁵⁷ A similar broadening due to Gd³⁺–Gd³⁺ interactions has been observed for NaYF₄/NaGdF₄ core-shell nanocrystals, even if the shell of NaGdF₄ is as thin as 1 nm.⁵⁸ In the study presented here we investigated core-shell particles with significantly thicker shells, consisting of a small 3 nm β -NaEuF₄ core and an \sim 8.5 nm thick β -NaGdF₄ shell. Since the EPR signal of such particles would be fully dominated by the broad spectrum of the undiluted Gd³⁺ ions in the shell, we used core-shell nanocrystals with inverted composition in the EPR measurements, *i.e.*, 3 nm β -NaGdF₄ core particles with an \sim 8.5 nm β -NaEuF₄ shell. In Fig. 6, the EPR spectrum of these core-shell particles (red line) is compared with the EPR spectrum of β -NaEuF₄ nanocrystals doped with 1% Gd³⁺ (blue line) and the EPR spectrum of 3 nm β -NaGdF₄ core particles (green line). Note that all spectra are normalized at maximum signal height as indicated by the scaling factors given in the figure. The highest scaling factor was used for the NaGdF₄ core particles as the strong interaction of the Gd³⁺ ions in these particles leads to strong broadening of their EPR signal (green line). The lowest scaling factor was used in the spectrum of the weakly doped β -NaEuF₄:1% Gd particles. This spectrum exhibits a dominating narrow EPR signal (peak-to-peak width, $\Delta B_{pp} \approx 10$ mT) in the $g = 2$ region and additional contributions at higher g -values (lower magnetic fields), the latter indicating the presence of Gd³⁺ ions at sites with different symmetries and/or crystal fields compared to the bulk sites. Since we used rather large NaEuF₄:1% Gd particles with a diameter of 22 nm, the contribution from surface Gd³⁺ sites, however, is almost negligible in this case. The additional contributions therefore indicate that the Gd³⁺ dopant ions are not homogeneously dis-

tributed inside the particles as observed before.⁵⁸ Finally, the high scaling factor used for the β -NaGdF₄/NaEuF₄ core-shell nanocrystals shows that their EPR signal is much weaker than the signal of the doped β -NaEuF₄:1% Gd particles, although both samples contain europium and gadolinium in the same molar ratio of 99 to 1. The much lower intensity is in accord with the core shell structure, since the signal of the Gd³⁺ ions in the core is expected to be strongly broadened. Similar to the doped β -NaEuF₄:1% Gd particles, however, the normalized EPR spectrum of the core-shell particles also displays a narrow signal at $g \approx 2.0$ characteristic for isolated or weakly interacting Gd³⁺ ions. Since this signal is even narrower ($\Delta B_{pp} \approx 3$ mT) and of higher relative intensity compared to the doped β -NaEuF₄:1% Gd particles, it indicates the presence of highly diluted ($\ll 1\%$) Gd³⁺ ions in the β -NaEuF₄ shell. Kombar *et al.* concluded from the EPR data for NaYF₄:Gd core-shell particles that the shell structure grows from dissolved precursor particles and that most likely also Y³⁺ ions are released from the surface of the core-precursors during the reaction.⁵⁸ The EPR data for the β -NaGdF₄:Eu core-shell particles shown in Fig. 6 indicates that these nanocrystals are formed in a similar way and therefore support the conclusions drawn above from luminescence spectroscopy.

2.5 Chemical properties, core level XPS

We employed core level XPS to investigate the chemical states of the Eu and Gd ions of the nanoparticles and core-shell systems in question. Fig. 7(a) displays the Gd 3d X-ray photoelectron spectra of the core-shell and pure β -NaGdF₄ (3 nm diameter) nanoparticles. The corresponding spectrum of a GdF₃ single crystal⁵² is also shown for comparison. For all samples the Gd 3d_{5/2} and 3d_{3/2} core level binding energies are located at 1187.5 eV and 1219.5 eV, respectively. All the spectra show no distinct satellite features between the 3d_{5/2} and 3d_{3/2} main peaks indicating the absence of any metallic Gd contributions.⁵⁹

Fig. 7(b)–(d) show the Eu 4d (and Gd 4d (Fig. 7(d))) spectra of pure β -NaEuF₄ (3 nm and 18 nm (Fig. 7(b) and (c))) and the β -NaEuF₄/NaGdF₄ core-shell nanoparticles with an 18 nm β -NaEuF₄ core (Fig. 7(d)). The Eu 4d XPS of the 3 nm β -NaEuF₄ nanoparticles comprises three distinct features located at 129.9 eV (A), 137.4 eV (B), and 143.4 eV (C). Features B and C can be attributed to the Eu³⁺ 4d_{5/2} and Eu³⁺ 4d_{3/2} states, whereas the lower binding energy peak A represents the Eu²⁺ 4d_{5/2} states,⁵² whereas the Eu²⁺ 4d_{3/2} states are overlapping with the Eu³⁺ 4d_{5/2} states and not visible as distinct features in the corresponding Eu 4d spectra. We de-convolute these spectra using Gaussian/Lorentzian line profiles. In the case of the β -NaEuF₄/NaGdF₄ core-shell nanoparticles (18 nm β -NaEuF₄ core) this de-convolution approach is extended to the Gd³⁺ 4d_{5/2} and Gd³⁺ 4d_{3/2} states, assuming a pure trivalent Gd valence state in accordance with the results presented in Fig. 7(a). With the fitting approach described we are able to determine the Eu cation distribution of the nanoparticles investigated here. One can already see by the naked eye that the 3 nm β -NaEuF₄ nanoparticles comprise the highest



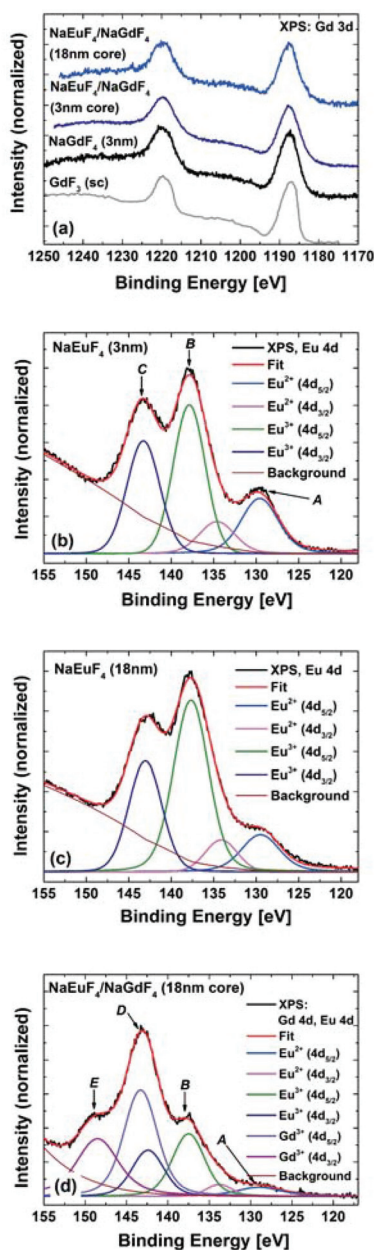


Fig. 7 Normalized XPS spectra of: (a) Gd 3d-core level of β -NaEuF₄/NaGdF₄ core-shell nanoparticles with 18 nm and 3 nm β -NaEuF₄ cores, 3 nm pure β -NaGdF₄ nanoparticles, and a GdF₃ single crystal⁵² as a reference. (b) Eu 4d-core level of pure 3 nm β -NaEuF₄ nanoparticles, de-convoluted into Eu²⁺ and Eu³⁺ fractions. (c) Eu 4d-core level of pure 18 nm β -NaEuF₄ nanoparticles, de-convoluted into Eu²⁺ and Eu³⁺ fractions. (d) Eu 4d-core level of β -NaEuF₄/NaGdF₄ core-shell nanoparticles with an 18 nm β -NaEuF₄ core. Eu and Gd spectra are partly overlapping, and are de-convoluted into Eu²⁺, Eu³⁺ and Gd³⁺ contributions.

amount of Eu²⁺ ions, we find the Eu valence state in this sample to be composed out of 30% Eu²⁺ and 70% Eu³⁺ ions. The Eu²⁺ fraction is significantly reduced for the other two samples, we find fractions of 20% Eu²⁺ for the 18 nm β -NaEuF₄ nanoparticles and 18% Eu²⁺ for the β -NaEuF₄/NaGdF₄ core-shell nanoparticles, respectively. We assume that

this result can be likely due to the different surface to bulk ratios of the samples as the formation of divalent Eu should predominantly occur at the surface (interface) of the sample, since a similar effect has been reported by Burian *et al.* on different EuF₃ surfaces and interfaces.⁵²

2.6 Electronic and magnetic properties

In order to elucidate further the magnetic properties all over the nanocrystals superconducting quantum interference device (SQUID) magnetometry measurements were performed.

Fig. 8(a) and (b) show thermomagnetic curves $M(T)$ of the precursor particles of pure β -NaGdF₄ and β -NaEuF₄ respectively. In the case of β -NaGdF₄ particles, the $M(T)$ dependences fulfill the Curie law but with small deviation that is clearly demonstrated in the $M \times T(T)$ curves. For pure Curie law (*i.e.* $M \sim 1/T$), the $M \times T(T)$ dependence should be a flat line, while the observed linear deviation indicates temperature independent component of magnetization which, in our case, can be attributed to some blocked magnetic moments of Gd. In the case of β -NaEuF₄ particles, the thermomagnetic curves are typical as for Eu³⁺ ions, showing the so-called van Vleck paramagnetism which is relatively weak and temperature independent. At low temperatures, below 20 K, traces of classical paramagnetic behavior are observed, *i.e.* sharp increase in magnetization with decreasing temperature. This effect is caused by an expected contribution of Eu²⁺ ions that pose magnetic moment and follow the Curie law. The possible origin of the Eu²⁺ presence is a surface valence transition, typical for nanoparticles containing Eu.

Fig. 8(c) shows hysteresis loops, determined at 2 K, for (i) pure β -NaEuF₄ and β -NaGdF₄ precursor particles, (ii) 22 nm β -NaGdF₄ nanocrystals doped with 1% of Eu and (iii) β -NaEuF₄/NaGdF₄ core-shell particles. Generally, the β -NaEuF₄ samples (3 nm and 18 nm) are paramagnetic. However, for the β -NaEuF₄ 3 nm nanocrystals some saturated component is detected which confirms the presence of magnetic Eu²⁺ ions. This answers the expectation of Eu²⁺ formation on the surface, consequently due to the large bulk to surface ratio in the 18 nm nanocrystals no signal is detectable. The pure β -NaGdF₄ precursor particles, β -NaEuF₄/NaGdF₄ core-shell particles with the 18 nm β -NaEuF₄ core and 22 nm β -NaGdF₄ nanocrystals doped with 1% Eu are paramagnetic with saturation magnetization as expected from the literature.⁶⁰ Surprisingly the β -NaEuF₄/NaGdF₄ core-shell nanocrystals with the 3 nm core show a butterfly shaped hysteresis loop. A similar behavior was recently found in rare earth based single molecular magnets.⁶¹ The explanation of the hysteresis loop for the β -NaEuF₄/NaGdF₄ core-shell particles with the 3 nm β -NaEuF₄ core can be some magnetic anisotropy. Accounting the fact that for 22 nm β -NaGdF₄ and 3 nm β -NaGdF₄ samples the effect of hysteresis broadening was not observed one may propose a model in which the anisotropy exists at the interface between a magnetic shell and a non-magnetic core.

In order to confirm this hypothesis Monte Carlo simulations were performed. The simulations are based on the Heisenberg model and the so-called simulated annealing, as



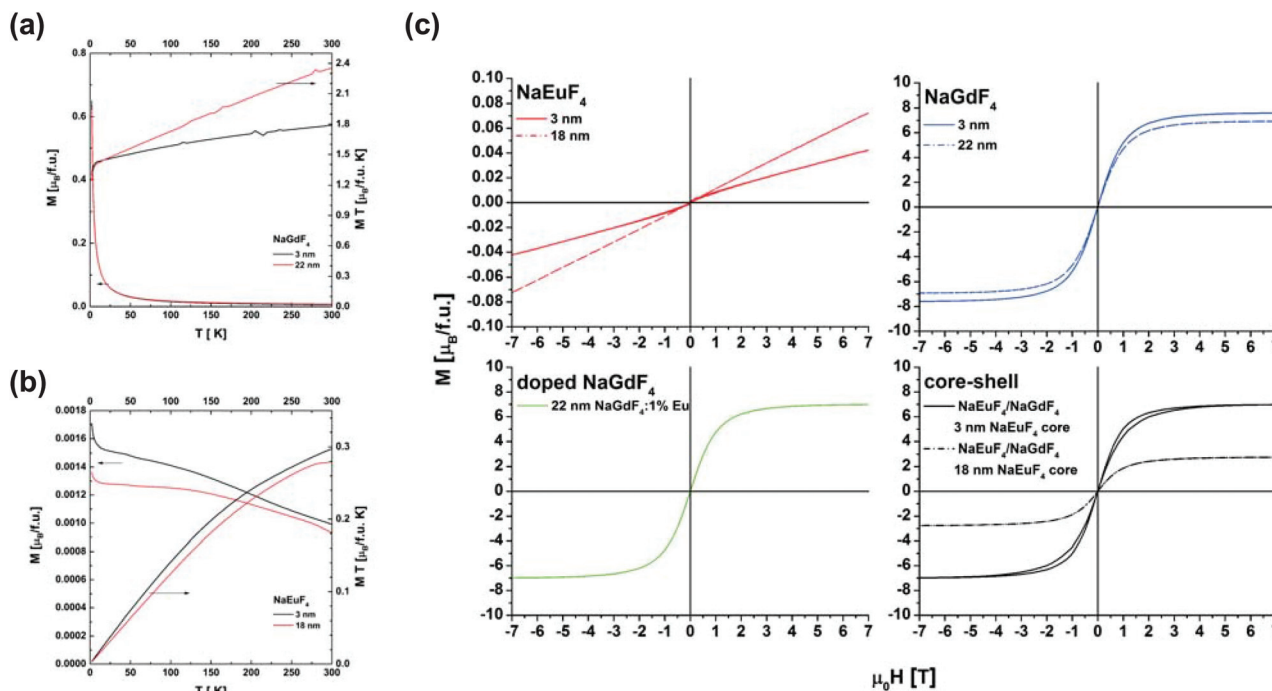


Fig. 8 Thermomagnetic $M(T)$ curves measured in $\mu_0 H = 0.1$ T of (a) β - NaGdF_4 and (b) β - NaEuF_4 particles and (c) hysteresis loops of pure β - NaEuF_4 and β - NaGdF_4 precursor particles, 22 nm β - NaGdF_4 nanocrystals doped with 1% Eu and β - $\text{NaEuF}_4/\text{NaGdF}_4$ core-shell particles at 2 K.

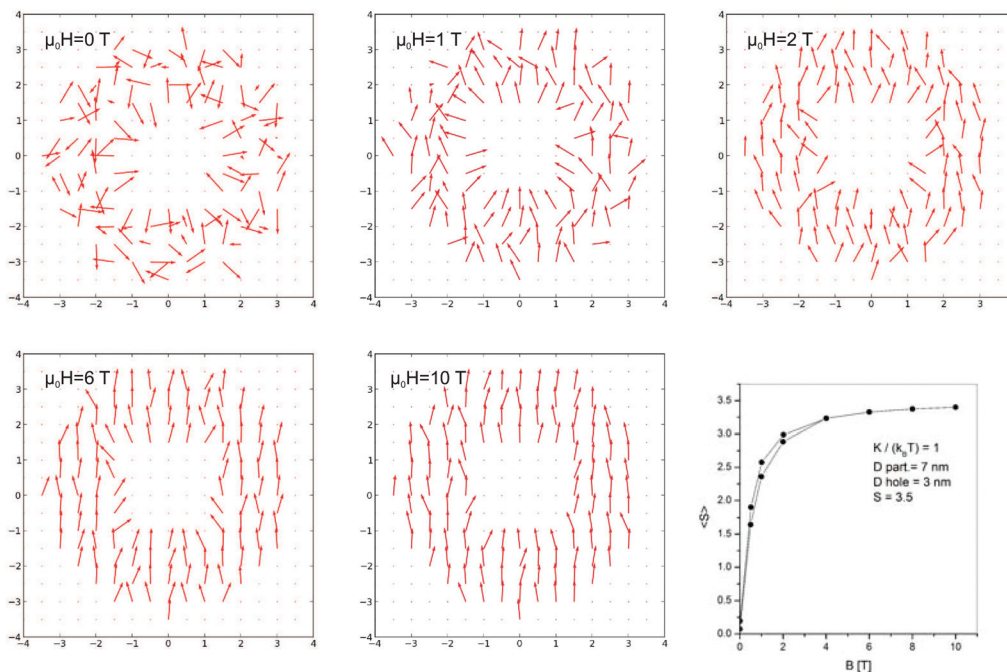


Fig. 9 Spin configurations (central section) for the simulated particles in the fields of 0 T, 1 T, 2 T, 6 T and 10 T (scales in nm) and the simulated hysteresis loops (the first quadrant) for the particle.

described by Ziółkowski and Chrobak.⁶² The analyzed particle consists of a spherical magnetic shell (7 nm in diameter) and a nonmagnetic core (3 nm). In the magnetic net, spin $S = 3.5$,

exchange integral parameter $J = 0$ and radial magnetic anisotropy coefficient $K = k_B T$. Fig. 9 depicts the spin configurations for different applied magnetic fields from 0 T to 10 T.



The resulting average spin in the direction of magnetic field (*i.e.* magnetization of the particle) as a function of increasing and decreasing field is presented in Fig. 9. As shown, this part of simulated hysteresis loop also reveals the butterfly shape. It should be underlined that good qualitative agreement with experimental data could be obtained only with the assumption of the surface anisotropy. The performed simulations including an influence of dipolar interactions, magnetic frustrations, weak ferro and antiferro couplings do not bring satisfactory results. Finally, one can conclude that the magnetization processes of β -NaGdF₄ nanoparticles can be affected by anisotropy that leads to the appearing of butterfly shaped hysteresis loops and deviation from the Curie law (see Fig. 8(a)).

In order to tackle the internal magnetic structure of selected β -NaEuF₄ and the β -NaEuF₄/NaGdF₄ core-shell nanocrystals with an 18 nm β -NaEuF₄ core in detail we performed X-ray magnetic circular dichroism (XMCD) at the Gd and Eu M_{4,5}-edges. XMCD is a very powerful tool to determine the magnetic properties of a compound in question valence and element specific.

Fig. 10 presents the Eu M_{4,5}-edge X-ray absorption spectroscopy (XAS) and XMCD spectra of the β -NaEuF₄/NaGdF₄ core-shell nanocrystals with an 18 nm β -NaEuF₄ core along with the Gd M₅-edge of the 2 nm β -NaGdF₄ shell. The Gd M₅-edge exhibits a large dichroic signal ($\sim 20\%$) of negative sign and a typical Gd³⁺ multiplet structure.⁶³

In contrast the Eu M_{4,5}-edge XMCD shows a rather weak negative dichroism of $\sim 2\%$ at the M₅ edge. However, this measured dichroic signal is significantly stronger than expected according to the SQUID data of the pure β -NaEuF₄ and β -NaGdF₄ nanoparticles. Here the maximum recorded magnetization (see Fig. 8(c)) of the β -NaEuF₄ nanocrystals is less than 1% compared to the β -NaGdF₄ nanocrystals of same size. Since both, Eu and Gd M₅ exhibit negative signs the net moments are parallelly aligned to each other. Fig. 10(b) displays the Eu M_{4,5}-edge XA- and XMCD spectra of the core-shell nanoparticles with those of pure 3 nm β -NaEuF₄ nanoparticles. The XPS spectra (see section 2.5) reveal an increased amount of divalent Eu ions which is also reflected in the X-ray absorption- and XMCD spectra of this sample. The M₅ edge XMCD of the 3 nm β -NaEuF₄ nanoparticles is dominated by two distinct features located at 1128 eV and 1131 eV, which can be attributed to Eu²⁺ and Eu³⁺ contributions to the XMCD signal.⁶⁴ In the case of the core-shell nanocrystals the intensity of the XMCD at 1128 eV is much less pronounced, which is in accordance with the XPS results indicating a significantly lower Eu²⁺ fraction compared to the 3 nm β -NaEuF₄ nanoparticles. The XMCD peak at 1132 eV can be presumed as being due to Eu³⁺ ions with a total magnetic quantum number $J \neq 0$, since Eu³⁺ in its magnetic ground state with $J = 0$ would produce no dichroic signal at all. On the other hand, Eu²⁺ ions are in a $J = 7/2$ ground state, leading to a “spin only” moment of $7\mu_B$ per Eu ion. This means that already quite a small fraction of uncompensated moments originating from Eu²⁺ ions may lead to a significant contribution to the overall Eu M_{4,5}-edge XCMD signal. As we have already pointed out the

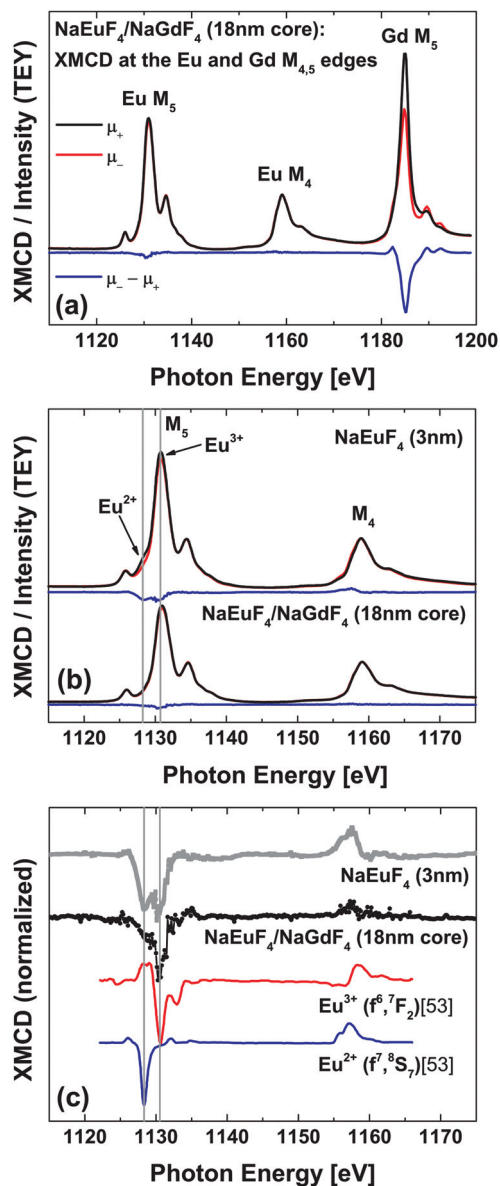


Fig. 10 (a) Gd M₅ and Eu M_{4,5}-edge XAS and XMCD of the β -NaEuF₄/NaGdF₄ core-shell nanocrystals with an 18 nm β -NaEuF₄ core. (b) Eu M_{4,5}-edge XAS and XMCD of the β -NaEuF₄/NaGdF₄ core-shell nanocrystals with an 18 nm β -NaEuF₄ core and the pure 3 nm β -NaEuF₄ nanoparticles. (c) Eu M_{4,5}-edge XMCD spectra in comparison with the corresponding atomic multiplet calculations of the 7F_2 Eu³⁺ and the 8S_7 Eu²⁺ final states. The simulated spectra have been extracted from Kachkanov *et al.*⁵³ The experiments have been performed under an external magnetic field of $B = 6$ T and with the samples at a temperature of 6 K. Spectra are offset vertically for clarity.

overall dichroic Eu signal is much higher than expected from the SQUID magnetometry data. One potential reason is that the Eu³⁺-signal stems from ions in the 7F_2 ($J = 1$) state. This is likely populated *via* the excitation of the probing X-ray beam during the XMCD measurement. Note that 4f intrashell transitions into the 7F_2 also dominate the luminescence spectra presented in Fig. 5. A similar effect has been reported by Kachkanov *et al.*⁵³ for Eu³⁺ doped GaN layers. It seems that



β -NaEuF₄ based nanocrystals and core-shell systems could be another example of a material in which the magnetic state can be altered by light excitation. In order to investigate this highly interesting effect in more detail future experiments by means of magneto optical SQUID and/or XMCD measurements in luminescence yield would be desirable.

3 Conclusions

We synthesized β -NaEuF₄/NaGdF₄ core-shell nanoparticles with narrow size distribution and an overall size of around 22 nm diameter. The particles contained either a small β -NaEuF₄ core with a diameter of only \sim 3 nm or a large β -NaEuF₄ core with a diameter of \sim 18 nm. Structural PDF analysis of the XRD measurements shows differences in peak positions between the both core-shell samples, although the differences in bond distance between the different atoms of bulk β -NaEuF₄ and β -NaGdF₄ are small. Comparison with simulation based on pure crystalline, spherical β -NaEuF₄ and β -NaGdF₄ nanoparticles point out that the β -NaEuF₄/NaGdF₄ core-shell nanoparticle with a 3 nm β -NaEuF₄ core exhibits structural parameters of β -NaGdF₄ structure, whereas β -NaEuF₄/NaGdF₄ core-shell nanoparticles with an 18 nm β -NaEuF₄ core resemble the β -NaEuF₄ structure. HR-TEM images reveal core-shell formation for the β -NaEuF₄/NaGdF₄ core-shell nanoparticles with an 18 nm β -NaEuF₄ core. The Eu³⁺ photoluminescence and Gd³⁺ ESR spectra both indicate that rare earth ions of the small 3 nm particle cores were released during shell growth and incorporate into the shell material. In contrast the system with the large β -NaEuF₄ core does not exhibit additional peaks in the Eu³⁺ luminescence spectra indicating that only a small number of Eu³⁺ ions are present in the NaGdF₄ shell. As to the Eu and Gd valance states, X-ray photoelectron-core level spectroscopy reveals significant fractions of Eu²⁺ ions in the β -NaEuF₄ precursors as well as the β -NaEuF₄/NaGdF₄ core-shell nanoparticles with an 18 nm β -NaEuF₄ core. Larger β -NaEuF₄ nanoparticles and surrounded by the β -NaGdF₄ shell reduce the Eu²⁺ fraction in agreement with results reported for EuF₃ thin film surfaces and interfaces.⁵² The presence of Eu²⁺ in the β -NaEuF₄ nanoparticles as well as in the core-shell systems is confirmed by SQUID magnetometry. $M(T)$ curves of the β -NaEuF₄ precursors exhibit a sharp increase in magnetization below 20 K stemming from Eu²⁺ ions since only the Eu³⁺ ions contribute a weak van Vleck paramagnetism. In the case of the core-shell system with 3 nm β -NaEuF₄ core we find a butterfly-shaped hysteresis loop which is known for single molecular magnets, for instance. This result could be reproduced by Heisenberg model simulations of a magnetic sphere comprising a non-magnetic core. XMCD at the Eu and M_{4,5} edges demonstrates the parallel alignment of Gd and Eu moments in the core-shell nanoparticles with a large β -NaEuF₄ core. The magnetic properties are dictated by the paramagnetic Gd³⁺ ions in the surrounding β -NaGdF₄ shell. However, both the Eu²⁺ and Eu³⁺ ions contribute to the overall magnetic properties of the

system. In particular the Eu³⁺ contribution is much higher than expected from the weak van Vleck paramagnetism. This might be due to the fact that the exciting X-rays populate the $J = 1$ state of Eu³⁺ via 4f transitions which can be also seen in the optical spectra. In future studies the core-shell formation with small β -NaEuF₄ cores will be further optimized, also to investigate the interesting butterfly shaped magnetic hysteresis in more detail. The fact that the magnetic state of Eu³⁺ might be altered by external light excitation is another interesting aspect which may be investigated by SQUID experiments under optical light excitation or XMCD using luminescence yield as the detection mode. Finally the multifunctionality of the β -NaEuF₄/NaGdF₄ core-shell nanoparticles could be further improved, e.g. by transition metal doping into the β -NaEuF₄ core and/or the β -NaGdF₄ shell. In summary we have successfully synthesized and carefully characterized multifunctional β -NaEuF₄/NaGdF₄ core-shell nanoparticles combining intense red optical emission from the Eu³⁺ ions with the paramagnetic properties stemming mainly from the Gd³⁺ ions.

4 Experimental

4.1 Materials

Sodium oleate (82%, Sigma-Aldrich), sodium fluoride (99%, Sigma-Aldrich), ammonium fluoride (98%, Sigma-Aldrich), oleic acid (90%, Alfa Aesar), 1-octadecene (90%, Alfa Aesar) and hydrated rare-earth chlorides of EuCl₃ and GdCl₃ (99.9%, Treibacher Industrie AG) were used as received.

Rare-earth oleates. The rare-earth oleates were prepared as given in the literature.⁶⁵ In a typical synthesis of europium oleate 60 mmol europium chloride and 180 mmol sodium oleate were dispersed in 120 ml ethanol, 80 ml water and 210 ml hexane. The resulting turbid solution was heated for 14 h under reflux (at about 60 °C). The resulting transparent organic phase containing the europium oleate was separated at room temperature and the hexane was removed with a rotary evaporator, yielding europium oleate. Gadolinium oleate was prepared analogously.

Synthesis of cubic phase precursor nanocrystals. The synthesis of the sub-10 nm particles of α -NaREF₄ (RE = Eu, Gd) is based on the reaction of rare-earth oleates and NaF in a mixture of oleic acid and 1-octadecene, similar to the procedure given by Liu *et al.*⁶⁶ In our case, rare-earth oleate and sodium fluoride were used in a molar ratio of 1 to 8. Oleic acid and 1-octadecene were added to the 1 to 8 mixture of rare-earth oleate and NaF (10 ml of each solvent per 1 mmol of rare-earth oleate) and the mixture was degassed in a vacuum Schlenk-line (100 Pa) for 1 h at 100 °C under stirring. The setup was three times subjected to vacuum and then refilled with nitrogen before the reaction mixture was heated at 200 °C. After 60 min at 200 °C and slow nitrogen flow the stirred solution was allowed to cool to room temperature. The excess of sodium fluoride was removed by centrifugation. The clear supernatant was mixed with an equal volume of ethanol leading to precipitation of the nanoparticles, which were



separated by centrifugation. The particles were purified by redispersing the precipitate in hexane (3 ml hexane per 1 mmol of rare-earth oleate), followed by precipitation with ethanol and separation of the particles by centrifugation.

Synthesis of hexagonal phase precursor nanocrystals. Hexagonal phase precursor particles of NaREF_4 were prepared by dispersing the corresponding rare-earth oleate and sodium oleate in a molar ratio of 1 : 2.5 in oleic acid and 1-octadecene (10 ml of each solvent per 1 mmol of rare-earth oleate) followed by degassing at 100 °C for 1 h under vacuum (100 Pa) in a Schlenk-line. Subsequently, 4 mmol ammonium fluoride per 1 mmol of rare-earth oleate was added at 100 °C to the clear yellowish solution under nitrogen flow. The vessel was three times subjected to vacuum and then refilled with nitrogen (vacuum applied for 5 seconds only) and the stirred solution was subsequently heated at 200 °C for 60 min under slow nitrogen flow. After cooling to room temperature an equal volume of ethanol was added to the solution, resulting in precipitation of the precursor particles. The particles were separated by centrifugation and purified as given above for the cubic phase precursor nanocrystals.

Growth of the final product particles. We used a method published earlier by the group of one of the present authors.⁶⁷ Product particles with different final sizes were obtained by varying the molar ratio of hexagonal and cubic phase precursor particles. In a typical synthesis, 3.76 mg $\beta\text{-NaEuF}_4$ and 1.28 g $\alpha\text{-NaGdF}_4$ precursor particles were re-dispersed in a solvent mixture consisting of 10 ml of oleic acid and 10 ml of 1-octadecene. The mixture was then heated at 100 °C for 60 min under vacuum and stirring. After the vessel was three times subjected to vacuum and then refilled with nitrogen, the solution was heated to 320 °C under nitrogen flow. After the desired reaction time, in this case 40 min, the reaction was stopped and the solution was cooled down to room temperature. After adding 60 ml of a 1 : 2 mixture of hexane and ethanol, the precipitate was separated from the solution by centrifugation. The precipitate was purified two times as given above for the precursor particles. The synthesis procedure for all NaREF_4 product particles was the same.

4.2 Experimental methods

The total scattering X-ray measurements were performed on the high energy scattering beamline P02.1 at PETRA III (DESY, Hamburg, Germany). The data sets were collected using a high energy monochromatic beam (59.875 keV, $\lambda = 0.20707 \text{ \AA}$) and a Perkin Elmer flat panel detector placed at 303 mm from the samples. The wavelength and the sample to detector distance have been determined by using a LaB_6 NIST standard. The used experimental setup allows to collect total scattering data over a sufficiently high momentum transfer ($Q \sim 22.72 \text{ \AA}^{-1}$) corresponding to a d spacing of 0.138 Å. We note here that the total scattering measurements have been conducted under the same experimental conditions for all the studied samples. The collected data were then corrected for experimental effects (absorption, multiple scattering, polarization, Compton scattering and Laue diffuse scattering) and the scattering signals

from the air and the experimental set up were measured independently under exactly the same conditions as the samples and subtracted as a background in the data reduction procedure. For obtaining the experimental atomic pair distribution function $G(r)$ by a direct sine Fourier transformation of the resulting total scattering structure function $S(Q)$, the data were truncated at a finite maximum value of $Q_{\text{max}} = 18.13 \text{ \AA}^{-1}$ beyond which the signal-to-noise ratio became unfavourable. All data processing was done using Fit-2D⁶⁸ and PDFgetX3⁶⁹ software.

A double Cs corrected JEOL ARM 200F transmission electron microscope operated at 200 kV was used for the analysis in the study. High angle annular dark field (HAADF) scanning transmission electron microscopy (STEM) was used along with electron energy loss spectroscopy (EELS) Spectrum Imaging (SI) using the DualEELSTM mode on the Gatan GIF Quantum 965 ERSTM spectrometer fitted on the microscope as well as EDS using an Oxford XMax 80 SDD.

Fluorescence spectra of the nanocrystals were recorded with a Fluorolog 3-22 spectrometer (Jobin Yvon) equipped with double monochromators for both channels.

The EPR spectra of the sample were recorded with a home made X-band EPR spectrometer (9 GHz) equipped with a dielectric resonator (DM5) (Bruker Biospin, Rheinstetten, Germany). The microwave power was set to 1.0 mW, the B-field modulation amplitude was 0.3 mT.

The XPS measurements were performed using a PHI 5600CI multitechnique spectrometer with monochromatic $\text{Al K}\alpha = 1486.6 \text{ eV}$ radiation of 0.3 eV at full width at half-maximum. The overall resolution of the spectrometer is 1.5% of the pass energy of the analyzer, 0.88 eV in the present case. The XPS measurements were performed at room temperature.

The XAS and XMCD measurements were performed at the Beamline for Advanced Dichroism (BACH) at the Elettra Synchrotron Facility.⁷⁰ The samples were mounted into a cryostat equipped with a 6.5 T superconducting magnet, the sample stage was connected to a pumped helium cryostat, reaching a base temperature of around 6.0 K during the experiments presented here. The measurements at the Gd and Eu $M_{4,5}$ edges have been recorded under external magnetic fields of 6 T in the total electron yield mode.

The FC-ZFC and hysteresis loop measurements were performed at the A. Chełkowski Institute of Physics, University of Silesia, Katowice, Poland, with the use of SQUID magnetometer (MPMS XL7 Quantum Design). All measurements have been performed in the 2 K–300 K temperature range up to 7 T magnetic field.

Acknowledgements

Parts of this research were carried out at the light source PETRA III at DESY, a member of the Helmholtz Association (HGF). We would like to thank J. Gamcová for assistance in using beamline P02.1. We thank all Beamline scientists for their excellent and generous technical support.



References

- S. Gai, C. Li, P. Yang and J. Lin, *Chem. Rev.*, 2014, **114**, 2343–2389.
- C. Liu, Y. Hou and M. Gao, *Adv. Mater.*, 2014, **26**, 6922–6932.
- D. K. Chatterjee, M. K. Gnanasammandhan and Y. Zhang, *Small*, 2010, **6**, 2781–2795.
- F. Wang, D. Banerjee, Y. Liu, X. Chen and X. Liu, *Analyst*, 2010, **135**, 1839–1854.
- D. Vennerberg and Z. Lin, *Sci. Adv. Mater.*, 2011, **3**, 26–40.
- O. S. Wolfbeis, *Chem. Soc. Rev.*, 2015, **44**, 4743–4768.
- Z. Li, Y. Zhang and S. Jiang, *Adv. Mater.*, 2008, **20**, 4765–4769.
- G. Chen, T. Y. Ohulchanskyy, S. Liu, W.-C. Law, F. Wu, M. T. Swihart, H. Hans Ågren and P. N. Prasad, *ACS Nano*, 2012, **6**, 2969–2977.
- Q. Liu, Y. Sun, T. Yang, W. Feng, C. Li and F. Li, *J. Am. Chem. Soc.*, 2011, **133**, 17122–17125.
- J. Liu, Y. Liu, W. Bu, J. Bu, Y. Sun, J. Du and J. Shi, *J. Am. Chem. Soc.*, 2014, **136**, 9701–9709.
- Y. Sun, X. Zhu, J. Peng and F. Li, *ACS Nano*, 2013, **7**, 11290–11300.
- Y.-F. Wang, G.-Y. Liu, L.-D. Sun, J.-W. Xiao, J.-C. Zhou and C.-H. Yan, *ACS Nano*, 2013, **7**, 7200–7206.
- R. Naccache, P. Chevallier, J. Lagueux, Y. Gossuin, S. Laurent, L. Vander Elst, C. Chilian, J. A. Capobianco and M.-A. Fortin, *Adv. Healthcare Mater.*, 2013, **2**, 1478–1488.
- R. Qiao, C. Liu, M. Liu, H. Hu, C. Liu, Y. Hou, K. Wu, Y. Lin, J. Liang and M. Gao, *ACS Nano*, 2015, **9**, 2120–2129.
- Y. Deng, H. Wang, W. Gu, S. Li, N. Xiao, C. Shao, Q. Xu and L. Ye, *J. Mater. Chem. B*, 2014, **2**, 1521–1529.
- H. Groult, J. Ruiz-Cabello, J. Pellico, A. V. Lechuga-Vieco, R. Bhavesh, M. Zamai, E. Almarza, I. Martin-Padura, E. Cantelar, M. P. Martinez-Alcazar and F. Herranz, *Bioconjugate Chem.*, 2015, **26**, 153–160.
- C. Liu, W. Ma, Z. Gao, J. Huang, Y. Hou, C. Xu, W. Yang and M. Gao, *J. Mater. Chem. C*, 2014, **2**, 9637–9642.
- R. Wang, X. Li, L. Zhou and F. Zhang, *Angew. Chem., Int. Ed.*, 2014, **126**, 12282–12286.
- H. Xing, S. Zhang, W. Bu, X. Zheng, L. Wang, Q. Xiao, D. Ni, J. Zhang, L. Zhou, W. Peng, K. Zhao, Y. Hua and J. Shi, *Adv. Mater.*, 2014, **26**, 3867–3872.
- Q. Zhan, J. Qian, H. Liang, G. Somesfalean, D. Wang, S. He, Z. Zhang and S. Andersson-Engels, *ACS Nano*, 2011, **5**, 3744–3757.
- D. Yang, P. Ma, Z. Hou, Z. Cheng, C. Li and J. Lin, *Chem. Soc. Rev.*, 2015, **44**, 1416–1448.
- Y. Dai, H. Xiao, J. Liu, Q. Yuan, P. Ma, D. Yang, C. Li, Z. Cheng, Z. Hou, P. Yang and J. Lin, *J. Am. Chem. Soc.*, 2013, **135**, 18920–18929.
- Y. I. Park, H. M. Kim, J. H. Kim, K. C. Moon, B. Yoo, K. T. Lee, N. Lee, Y. Choi, W. Park, D. Ling, K. Na, W. K. Moon, S. H. Choi, H. S. Park, S.-Y. Yoon, Y. D. Suh, S. H. Lee and T. Hyeon, *Adv. Mater.*, 2012, **24**, 5755–5761.
- G. Tian, Z. Gu, L. Zhou, W. Yin, X. Liu, L. Yan, S. Jin, W. Ren, G. Xing, S. Li and Y. Zhao, *Adv. Mater.*, 2012, **24**, 1226–1231.
- D. Yang, X. Kang, P. Ma, Y. Dai, Z. Hou, Z. Cheng, C. Li and J. Lin, *Biomaterials*, 2013, **34**, 1601–1612.
- L. Wang, J. Liu, Y. Dai, Q. Yang, Y. Zhang, P. Yang, Z. Cheng, H. Lian, C. Li, Z. Hou, P. Ma and J. Lin, *Langmuir*, 2014, **30**, 13042–13051.
- R. Naccache, F. Vetrone and J. A. Capobianco, *ChemSusChem*, 2013, **6**, 1308–1311.
- H. Lian, Z. Hou, M. Shang, D. Geng, Y. Zhang and J. Lin, *Energy*, 2013, **57**, 270–283.
- J. Shen, Z. Li, R. Cheng, Q. Luo, Y. Luo, Y. Chen, X. Chen, Z. Sun and S. Huang, *ACS Appl. Mater. Interfaces*, 2014, **6**, 17454–17462.
- G.-B. Shan, H. Assaoui and G. P. Demopoulos, *ACS Appl. Mater. Interfaces*, 2011, **3**, 3239–3243.
- H. Zhu, X. Chen, L. M. Jin, Q. J. Wang, F. Wang and S. F. Yu, *ACS Nano*, 2013, **7**, 11420–11426.
- Y. Zhou, S.-T. Han, X. Chen, F. Wang, Y.-B. Tang and V. Roy, *Nat. Commun.*, 2014, **5**, 1–8.
- D. Peng, Q. Ju, X. Chen, R. Ma, B. Chen, G. Bai, J. Hao, X. Qiao, X. Fan and F. Wang, *Chem. Mater.*, 2015, **27**, 3115–3120.
- W. J. Kim, M. Nyk and P. N. Prasad, *Nanotechnology*, 2009, **20**, 185301.
- Y. Lu, J. Zhao, R. Zhang, Y. Liu, D. Liu, E. M. Goldys, X. Yang, P. Xi, A. Sunna, J. Lu, Y. Shi, R. C. Leif, Y. Huo, J. Shen, J. A. Piper, J. P. Robinson and D. Jin, *Nat. Photonics*, 2014, **8**, 32–36.
- J. M. Meruga, A. Baride, W. Cross, J. J. Kellar and P. S. May, *J. Mater. Chem. C*, 2014, **2**, 2221–2227.
- G.-S. Yi and G.-M. Chow, *Chem. Mater.*, 2007, **19**, 341–343.
- Y.-F. Wang, L.-D. Sun, J.-W. Xiao, W. Feng, J.-C. Zhou, J. Shen and C.-H. Yan, *Chemistry*, 2012, **18**, 5558–5564.
- D. Chen and P. Huang, *Dalton Trans.*, 2014, **43**, 11299–11304.
- G. Chen, H. Agren, T. Y. Ohulchanskyy and P. N. Prasad, *Chem. Soc. Rev.*, 2015, **44**, 1680–1713.
- M.-K. Tsang, G. Bai and J. Hao, *Chem. Soc. Rev.*, 2015, **44**, 1585–1607.
- F. Vetrone, R. Naccache, V. Mahalingam, C. G. Morgan and J. A. Capobianco, *Adv. Funct. Mater.*, 2009, **19**, 2924–2929.
- Y. Liu, D. Tu, H. Zhu, R. Li, W. Luo and X. Chen, *Adv. Mater.*, 2010, **22**, 3266–3271.
- F. Wang, R. Deng, J. Wang, Q. Wang, Y. Han, H. Zhu, X. Chen and X. Liu, *Nat. Mater.*, 2011, **10**, 968–973.
- M. Quintanilla, F. Ren, D. Ma and F. Vetrone, *ACS Photonics*, 2014, **1**, 662–669.
- C. Dong, A. Korinek, B. Blasiak, B. Tomanek and F. C. J. M. van Veggel, *Chem. Mater.*, 2012, **24**, 1297–1305.
- Q. Su, S. Han, X. Xie, H. Zhu, H. Chen, C.-K. Chen, R.-S. Liu, X. Chen, F. Wang and X. Liu, *J. Am. Chem. Soc.*, 2012, **134**, 20849–20857.



- 48 H. Wen, H. Zhu, X. Chen, T. F. Hung, B. Wang, G. Zhu, S. F. Yu and F. Wang, *Angew. Chem., Int. Ed.*, 2013, **125**, 13661–13665.
- 49 X. Li, R. Wang, F. Zhang and D. Zhao, *Nano Lett.*, 2014, **14**, 3634–3639.
- 50 M. Bottrill, L. Kwok and N. J. Long, *Chem. Soc. Rev.*, 2006, **35**, 557–571.
- 51 E.-J. Cho and S.-J. Oh, *Phys. Rev. B: Condens. Matter*, 1999, **59**, R15613–R15616.
- 52 W. Burian, J. Szade, T. O’Keevan and Z. Celinski, *Phys. Status Solidi B*, 2004, **241**, R15–R18.
- 53 V. Kachkanov, M. J. Wallace, G. van der Laan, S. S. Dhessi, S. A. Cavill, Y. Fujiwara and K. P. O’Donnell, *Sci. Rep.*, 2012, **2**, 1–5.
- 54 G. H. Dieke, *Spectra and energy levels of rare-earth ions in crystals*, Interscience Publishers, New York, 1968.
- 55 P. Ptacek, H. Schaefer, K. Koempe and M. Haase, *Adv. Funct. Mater.*, 2007, **17**, 3843–3848.
- 56 C. Li, Z. Quan, J. Yang, P. Yang and J. Lin, *Inorg. Chem.*, 2007, **46**, 6329–6337.
- 57 C. M. Brodbeck and L. E. Iton, *J. Chem. Phys.*, 1985, **83**, 4285–4299.
- 58 R. Kompan, J. P. Klare, B. Voss, J. Nordmann, H.-J. Steinhoff and M. Haase, *Angew. Chem., Int. Ed.*, 2012, **51**, 6506–6510.
- 59 J. Szade and M. Neumann, *J. Phys.: Condens. Matter*, 2001, **13**, 2717.
- 60 F. Li, C. Li, X. Liu, Y. Chen, T. Bai, L. Wang, Z. Shi and S. Feng, *Chem. – Eur. J.*, 2012, **18**, 11641–11646.
- 61 L. Margheriti, D. Chiappe, M. Mannini, P.-E. Car, P. Saintavit, M.-A. Arrio, F. B. de Mongeot, J. C. Cezar, F. M. Piras, A. Magnani, E. Otero, A. Caneschi and R. Sessoli, *Adv. Mater.*, 2010, **22**, 5488–5493.
- 62 G. Ziółkowski and A. Chrobak, *Acta Phys. Pol., A*, 2015, **127**, 597–598.
- 63 G. Champion, N. Lalioti, V. Tangoulis, M.-A. Arrio, P. Saintavit, F. Villain, A. Caneschi, D. Gatteschi, C. Giorgetti, F. Baudelet, M. Verdaguer and C. Cartier dit Moulin, *J. Am. Chem. Soc.*, 2003, **125**, 8371–8376.
- 64 B. J. Ruck, H. J. Trodahl, J. H. Richter, J. C. Cezar, F. Wilhelm, A. Rogalev, V. N. Antonov, B. D. Le and C. Meyer, *Phys. Rev. B: Condens. Matter*, 2011, **83**, 174404.
- 65 J. Park, K. An, Y. Hwang, J.-G. Park, H.-J. Noh, J.-Y. Kim, J.-H. Park, N.-M. Hwang and T. Hyeon, *Nat. Mater.*, 2004, **3**, 891–895.
- 66 C. Liu, H. Wang, X. Zhang and D. Chen, *J. Mater. Chem.*, 2009, **19**, 489–496.
- 67 B. Voss and M. Haase, *ACS Nano*, 2013, **7**, 11242–11254.
- 68 A. P. Hammersley, S. O. Svensson, M. Hanfland, A. N. Fitch and D. Hausermann, *High Pressure Res.*, 1996, **14**, 235–248.
- 69 P. Juhas, T. Davis, C. Farrow and S. Billinge, *J. Appl. Crystallogr.*, 2013, **46**, 560–566.
- 70 M. Zangrando, M. Finazzi, G. Paolucci, G. Comelli, B. Diviacco, R. P. Walker, D. Cocco and F. Parmigiani, *Rev. Sci. Instrum.*, 2001, **72**, 1313–1319.

