Bulk and surface exsolution produces a variety of Fe-rich and Fe-depleted ellipsoidal nanostructures in La0.6Sr0.4FeO3 thin films

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Bulk and surface exsolution produces a variety of Fe-rich and Fe-depleted ellipsoidal nanostructures in $La_{0.6}Sr_{0.4}FeO_3$ thin films

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Abstract

The past several years have seen a resurgence in the popularity of metal exsolution as an approach to synthesize advanced materials proposed for novel catalytic, magnetic, optical, and electrochemical properties. Whereas most studies to-date have focused on surface exsolution (motivated by catalysis), we instead report on the diversity of nanostructures formed in $La_{0.6}Sr_{0.4}FeO_3$ thin films during sub-surface or so-called 'bulk' exsolution, in addition to surface exsolution. Bulk exsolution is a promising approach to tuning the functionality of materials, yet there is little understanding of the nanostructures exsolved within the bulk and how they compare to those exsolved at gas-solid interfaces. This work combines atomic- and nano-scale imaging and spectroscopy techniques applied using a state-of-the-art aberration-corrected scanning transmission electron microscope (STEM). In doing so, we present a detailed atomic-resolution study of a range of Fe-rich and Fe-depleted nanostructures possible *via* exsolution, along with qualitative and quantitative chemical analysis of the exsolved nanostructures and oxide phases formed throughout the film. Local structural changes in the perovskite matrix, coinciding with nanostructure exsolution, are also characterized with atomic-resolution STEM imaging. Fe exsolution is shown to create local A-site rich domains of Ruddlesden-Popper phase, and some stages of this phase formation have been demonstrated in this work. In particular, phase boundaries are found to be the primary nucleation sites for bulk and surface exsolution, and the exsolved particles observed here tend to be ellipsoidal with shape factor of 1.4. We report a range of nanostructure types (core-shell, bulk core-shell, adjacent, and independent particles), revealing several possible avenues of future exploration aimed to understand the formation mechanism of each exsolution type and to develop their functionality. This work is thus relevant to materials scientists and engineers motivated to understand and utilize exsolution to synthesize materials with predictable nanostructures.

Introduction

Exsolution is an *in situ* synthesis method for creating highly-tunable catalytic nanostructures anchored in the surface of an oxide support [1–7]. This flexible nanocatalyst synthesis approach can produce active particles which are more durable than impregnated nanostructures, and which can be strained as a result of being partially submerged in the oxide – offering enhanced activity. Our current understanding of exsolution results from nearly two decades of catalysis research dating back to pioneering work on "self-regenerative" perovskite-supported precious metal (Pd, Rh and Pt) catalysts for automotive emissions control by researchers at Daihatsu [1,8–12]. Since then, a range of nanostructures have been developed for several catalytic applications, such as solid oxide fuel/electrolysis cells [4,4,5,13–29], hydrogen and syngas production [30–34,18,35–37], oxygen evolution [38], CO_2 reduction [39,40], and photocatalysis [41]. Very recently, focus has turned to engineering the bulk properties of functional oxides using the bulk exsolution design strategy, which, for instance have effectively tailored the electronic transport properties of a lanthanum ferrite perovskite [42].

Exsolution, also called "self-regeneration" or "solid-state crystallization" is a phase decomposition reaction that involves metal cation coalescence at/near individual atomic-scale point defects of opposite charge, such as oxygen vacancies. This is often observed in the vicinity of extended defects like surfaces and grain boundaries due to the relatively lower vacancy formation energy. Li, Liu *et al.* reviewed progress in the field of exsolution-derived electrocatalysts through 2017, see Table 3 of [2]. Many compositions have been investigated, mainly with the goal of efficient use of precious metals and various transition metals (though mostly Ni or its alloys) selected for particular reactions. Transition metals (MnO [3,26,43], Fe [3,13,28,30,43–45], Fe-Co [29], FeNi3 [40], Fe-Ni [17,30], Co [24,25,43,46], Co-Ni [25,47], Ni [4,5,13–16,18,22,24,25,27,31–37,43,48–50], Ni-Cu [32], Cu [3,24,32], Re/Ni/Fe [51], Ce [3], Ag [19], Rare-earth vanadates doped with transition metal [24]), noble metals (Pd [1,10–12,23], Pt [9,10,12,52], Ir [38], Rh [9,10,12,39,52,53], Ru [20–22]), and transition metal dichalcogenides (WS_2 [41]) have been widely studied for exsolution.

In terms of nanostructure morphology, the main focus has been surface-anchored nanostructures for catalysis; however, several researchers have also reported bulk (i.e., sub-surface) exsolution of embedded nanostructures[54–56]. Dai, Pan, *et al.* concluded based on *in situ* TEM studies of Rh exsolution from $CaTiO_{3-\delta}$ that initial-stage exsolution created nanostructures embedded in the perovskite matrix, making them inaccessible during heterogeneous catalysis [53]. Kim, Choi, *et al.* reported STEM dark field images clearly showing bulk exsolution of Ni particles throughout the thickness of epitaxially grown $La_{0.2}Sr_{0.7}Ti_{0.9}Ni_{0.1}O_{3-\delta}$ thin films, but did not interrogate these nanostructures beyond nanometer-scale energy dispersive spectroscopy (EDS) line profiles confirming the presence of Ni [50]. Han, Irvine, *et al.* showed as well that Ni exsolves in the bulk of epitaxial $La_{0.2}Sr_{0.7}Ni_{0.1}Ti_{0.9}O_{3-\delta}$ thin films, and while they did not explore this, there is indication that various core-shell nanostructures formed in their films during/after Ni exsolution [37]. In the context of bulk exsolution, Kousi, Metcalfe, *et al.* demonstrated that submerged Ni nanoparticles exsolved from $La_{0.8}Ce_{0.1}Ni_{0.4}Ti_{0.6}O_{3-\delta}$ strain the perovskite lattice upon oxidation to crystallographically-aligned NiO, yielding favorable oxygen exchange properties for methane conversion to syngas[34].

Recent publications by co-authors of this study have been focused on Fe exsolution in $La_{0.6}Sr_{0.4}FeO_3$ (LSF) thin films[6,42]. Most recently, Wang et al. reported that Fe^0 exsolution in LSF results in major chemical and structural reorganization at the surface and also in the bulk, which can be used effectively to tailor electrical, magnetic and chemical properties of the resulting nanocomposite[42]. It was demonstrated that exsolution in LSF results in not only metallic iron nanoparticles whose magnetization could be modulated, but also Ruddlesden-Popper domains and Fe-deficient percolating channels which facilitated an increase in electronic conductivity by over 100 times[42]. Particularly, it was shown that the Fe-deficient percolating channels can enhance electronic conductivity by several orders of magnitude[42]. Such findings have direct implications for fabricating multifunctional nanocomposites for wide range of applications, making this LSF system valuable for exploring bulk and surface nanostructure morphology.

Here, we report a detailed-study of various Fe-based embedded nanostructures possible in LSF thin films *via* so-called bulk exsolution. LSF was chosen as a model perovskite system due to well-understood defect chemistry and the technological promise provided by its bulk properties in chemical looping, solid oxide cells, and facilitating magnetic nanostructures, as described in our recent study[42] which also includes discussion of the formation mechanism of the exsolved nanostructures in some detail. The aim of this work is to illustrate the variation in nanostructures that one can make in this material, which should help guide the design and synthesis of exsolved nanostructures in a broad range of chemical compositions, microstructures, and applications. We quantify the structural and chemical aspects of bulk exsolution using imaging and spectroscopy down to the atomic- and nanoscale using a state-of-the-art aberration-corrected STEM. There has not yet been a detailed study on the various types of nanostructures possible *via* bulk exsolution.

Methods

LSF thin films were synthesized by PLD. LSF powders were prepared by mixing and grinding La_2O_3 (Sigma Aldrich, 99.999%), $SrCO_3$ (Sigma Aldrich, 99.995%), and Fe_2O_3 (Alfa Aesar, 99.998%) powders in appropriate ratios, and then being calcined at 1000 °C for 5 hours. The powders are then pressed into a pellet with a hydrostatic press and sintered at 1350 °C for 20 hours in stagnant air, thus yielding the target.

MgO (001) single crystal substrates (MTI Corporation, 1-side polished) were employed in this work. During pulsed laser deposition (PLD), the substrate temperature was kept at 850 °C in an oxygen pressure of 20 mTorr. By applying 30000 laser pulses (laser energy: 400 mJ) to the LSF target, an LSF thin film of ~120 nm thickness was grown on the substrate (substrate to target distance is 85 mm). This LSF film before exsolution will be referred to as 'as-prepared LSF' for rest of the article. After deposition, the sample was cooled in the deposition atmosphere at a cooling rate of 5 °C/min.

Fe exsolution was induced by exposure to H_2. LSF sample is reduced in 0.5 Torr flushing H_2 at 400 °C for 0.5 hours to trigger exsolution.

Atomic-scale imaging and chemical analyses were performed by aberration-corrected STEM - energy dispersive spectroscopy (EDS). Exsolution in LSF was characterized using state-of-the-art aberration corrected JEOL Grand ARM300CF STEM operated at 300 keV. This instrument is equipped with dual silicon drift detectors for EDS with total detector area of 200 mm^2. The probe size was ~0.1 nm with a probe current of about 35pA. Low-background, high count analytical TEM holder was used to minimize any stray x-rays caused by holder. The film thickness of the exsolved LSF film (on MgO substrate) used in this study was measured to be ~165 nm.

Gatan Microscopy Suite (GMS version 3) software was utilized for STEM-EDS mapping and analysis using the following EDS peaks: O K_α (0.52 keV), Fe K_α and K_β (6.40 keV and 7.06 keV respectively), La L_α and L_β (4.65 keV and 5.04 keV and respectively), and Sr K_α and K_β (14.16 keV and 15.84 keV respectively). Net counts for EDS elemental maps were obtained after doing background subtraction using Kramers filter (with scaling option available in Gatan to better align the measured background to the original data). Due to sample drift observations during some EDS scans, aspect ratio for the EDS maps was accordingly adjusted to match the original STEM image. To interpret the chemistry of exsolved nanostructures in LSF, Cliff-Lorimer method [57–59] was used to calculate concentrations of cations, Fe, La and Sr. Due to the widely known limitations of quantification of light elements by EDS, we excluded O from quantitative analysis in this study. Experimental k-factors were calculated using the as-prepared LSF film as reference, since the stoichiometric composition of homogeneous LSF film is known.

Additional physical, chemical, and functional characterization methods (atomic force microscopy, X-ray diffraction, X-ray photoelectron spectroscopy, X-ray absorption spectroscopy, electrochemical impedance spectroscopy, finite element analysis, and vibrating-sample magnetometry) are detailed elsewhere [42].

Results and Discussion

STEM-EDS characterization of bulk and surface nanostructures. The variety of nanostructures formed in LSF via exsolution are comprised of Fe-rich and Fe-depleted regions, accompanied by variation in La and Sr content (Fig. 1) [42]. Using STEM high-angle annular dark field (HAADF) and bright field (BF) imaging, we characterized LSF film on (001) MgO substrate before (Fig. 1a-c) and after exsolution (Fig. 1 d-f). After the treatment to trigger exsolution, LSF film shows distinct variety of nanostructures and oxide phases formed throughout the film (Figs. 1d-f). We also observe the formation of Ruddlesden-Popper (RP) phase after exsolution; RP is a perovskite-like derivative with A-site rich domains and is quantified by a local translation of the $ABO_{3-\delta}$ structure by sqrt(½) of a unit cell. Using EDS elemental mapping, we show that most Fe-rich exsolved particles correspond to La and Sr depletion (Fig. 1g). There is also variation observed in La and Sr signal, compared to the LSF film matrix, for regions which are depleted in Fe.

We observe four main categories of nanostructures formed via surface and bulk exsolution: (i) surface core-shell, (ii) bulk core-shell, (iii) adjacent particles, and (iv) independent particles (Fig. 2). We demonstrated in our recent paper using FFT analysis that the bulk exsolved particles are in fact metallic Fe, while the surface exsolved particles are Fe oxide (either Fe_3O_4 or γ-Fe_2O_3 [42]) due to exposure to air after exsolution [42]. We observe majority of nanostructures to be exsolved at interfaces in the LSF film indicating the activation energy of Fe nucleation is likely lower at the interfaces. The core-shell structures of surface and bulk particles, both exsolved at phase boundaries of perovskite matrix and RP phase, can be seen in STEM HAADF and BF images (Figs. 2a-f). STEM HAADF images display Z-contrast, with the image intensity being proportional to the mean atomic number Z^n (where $n \sim 1.7$) [60]. STEM BF images, on the other hand, show phase and diffraction contrast [61]. To highlight the complementary information provided by both methods, the difference between perovskite matrix and RP phase domains is more apparent in STEM BF images (phase and diffraction contrast), while the STEM HAADF images display Z-contrast making it easier to differentiate between Fe-rich and Fe-depleted particles.

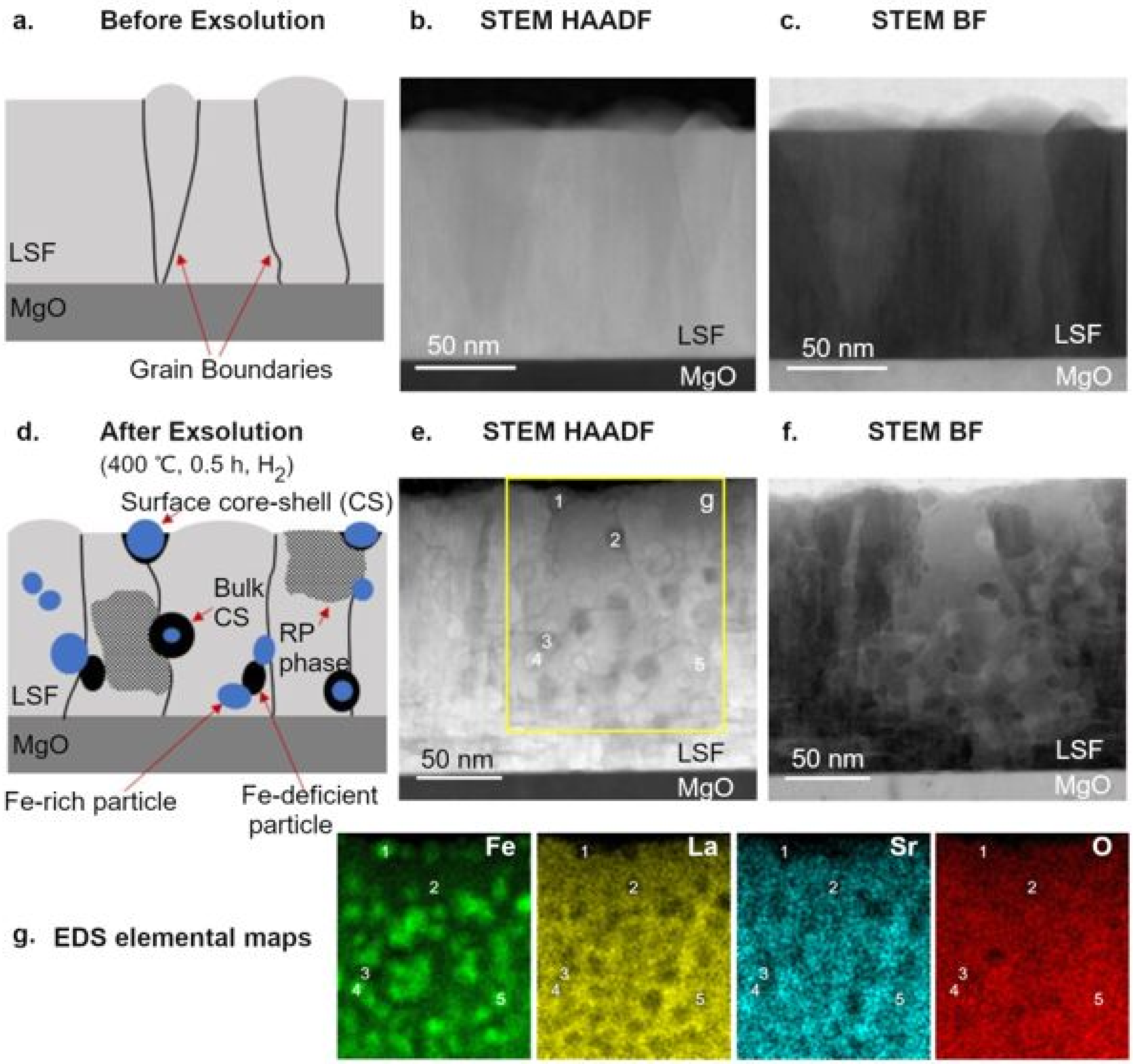

Fig. 1: $La_{0.6}Sr_{0.4}FeO_3$ (LSF) thin films were deposited by pulsed laser deposition (PLD) on 001-oriented MgO single crystals (a) and analyzed using scanning TEM (STEM) high-angle annular dark field (HAADF) images (b) and STEM bright-field (BF) images (c). Treatment of the films in 0.5 torr H_2 at 400^{o}C for 0.5 h induced Fe exsolution, yielding a variety of Fe-rich and Fe-depleted nanostructures and oxide phases through the film depth (d-f). EDS elemental maps (net counts) of Fe K, La L, Sr Ka and O K showing Fe exsolution throughout the LSF film (g).

The core of the surface particle is rich in Fe, while the shell has primarily Fe and La signals suggesting presence of La/Fe based oxide, which is confirmed from quantitative analysis in the next section (Fig. 3b; see supplementary information Figs. S1-S3 for individual EDS elemental maps). Because the STEM data are two-dimensional projections of a three-dimensional object, it is ambiguous at times to distinguish the nanostructure from the surrounding matrix. Similarly, the bulk core-shell particle also has an Fe-rich core (Fig. 3d).

'Adjacent particles' are defined as nanostructures that are adjacent to each other after exsolution, with distinct high and low Z contrast in STEM HAADF imaging as well as phase/diffraction contrast in STEM BF imaging (Fig. 2h-i). These adjacent particles correspond to Fe-rich particles in the bulk resulting in Fe-depleted regions in the immediate vicinity (Fig. 3e-f).

Such adjacent particles are observed throughout the LSF film after exsolution, and more examples can be seen in Fig. 4.

'Independent particles' are Fe-rich particles with no considerable Fe-depleted region nearby (Fig. 2 j-l). These particles are also formed throughout the film via bulk exsolution (Fig. 4). These independent particles have a very thin shell (~1-2 nm) which appears dark in Z-contrast images (Fig. 4a), suggesting chemistry like the shell of bulk core-shell particles characterized earlier. However, we categorized these particles as independent because clearly the Fe-rich 'core' region comprises of majority of the particle, unlike the bulk core-shell particle where both core and shell are prominently observed.

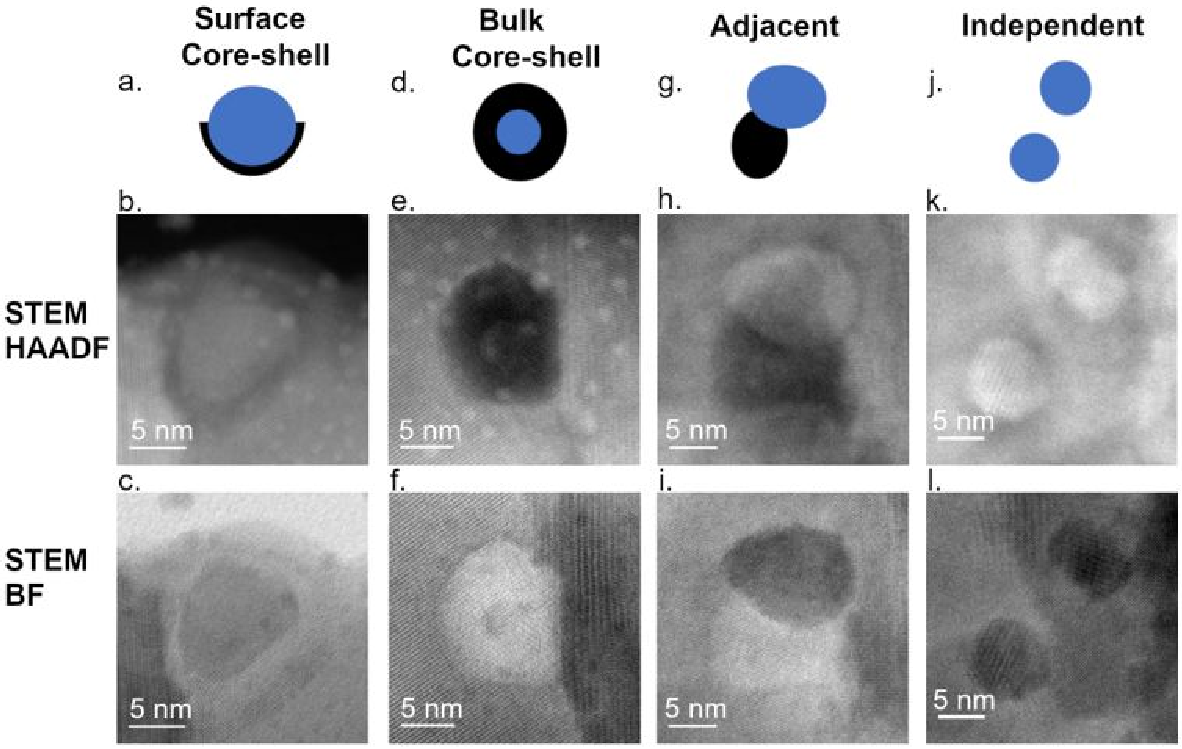

Fig. 2: Different types of Fe-based nanostructures observed in LSF thin film via exsolution shown using schematics, STEM HAADF images and STEM BF images. Surface core-shell (a-c) and bulk core-shell (d-f) structures have Fe-rich core and Fe-depleted shell. Another observed nanostructure is adjacent Fe-rich and Fe-depleted particles (g-i). Independent Fe-rich nanostructures (without adjacent Fe-depleted regions) are also observed throughout the LSF film (j-l).

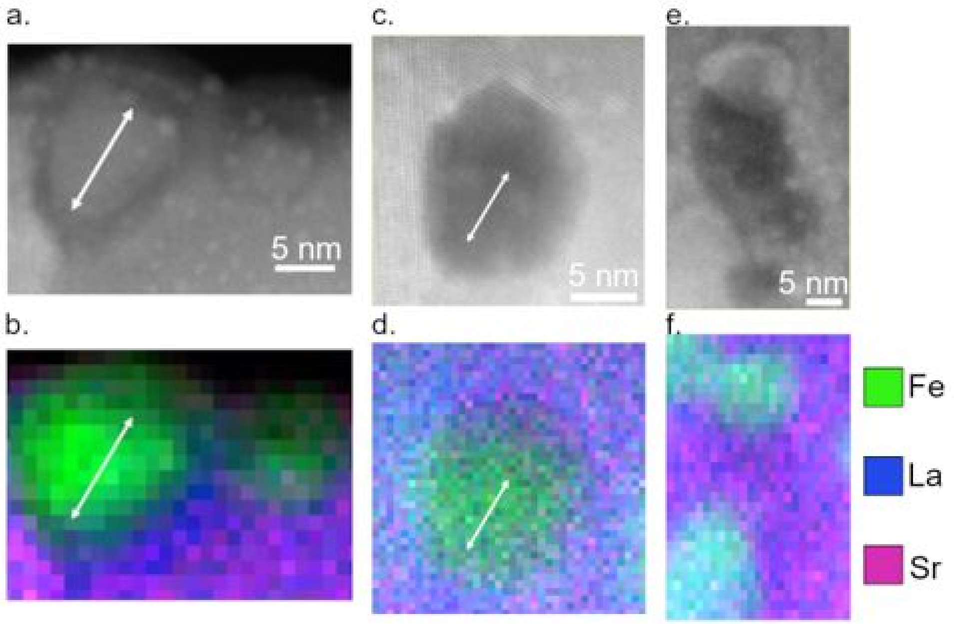

Fig. 3: STEM-HAADF images and overlay of EDS elemental maps (net counts) of surface core-shell particle (a-b), bulk core-shell (c-d) and adjacent particles (e-f).

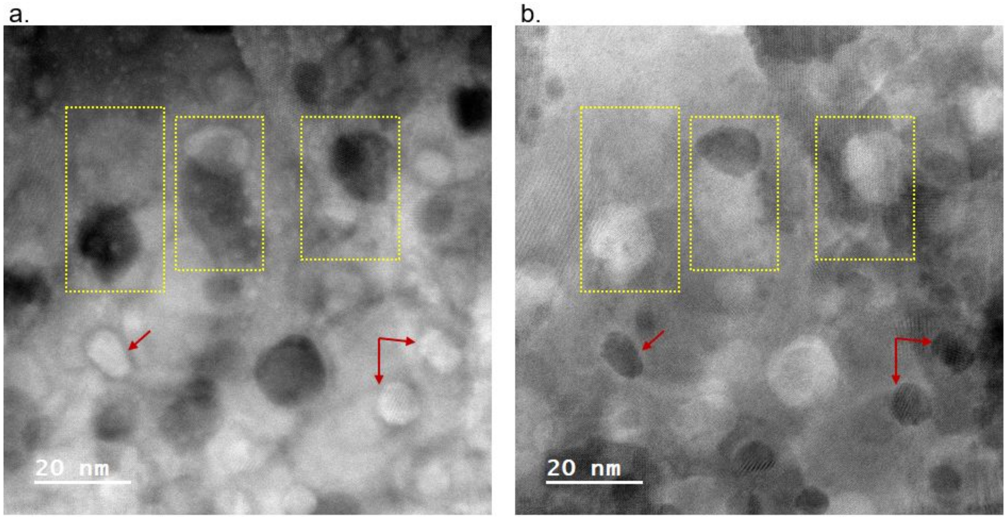

Fig. 4: STEM HAADF (a) and BF (b) images showing variety of ‘adjacent’ Fe-rich and Fe-depleted particles observed, some marked with yellow outline. ‘Independent’ Fe-rich particles are marked with red arrows.

Geometry of exsolved nanostructures in LSF. The Fe-rich nanostructures formed throughout the LSF film, both at surface and in bulk, are mostly observed to have ellipsoidal geometry. Most research studies to-date performed in H_2 (e.g. H_2/Ar) reducing environments, but focused only on surface exsolution, have also shown similar ellipsoidal or quasi-spherical geometry for exsolved particles [16,36,62]. Our results are thus consistent with the previous observations of surface particle geometry produced by H_2 environments, but we further demonstrate that the bulk exsolved particles also have similar ellipsoidal geometry. Neagu *et al.* [36] showed that the shape of the exsolved nanostructures can be tuned for various applications by changing the atmosphere for exsolution treatment as they explored with faceted particle growth in a vacuum environment. Our results show that both surface and bulk exsolved particles will likely maintain similar geometry depending on the exsolution treatment. For the exsolved particles in this study, the shape factor (ratio of major axis length to minor axis, *a/b*) was calculated to be 1.4 ± 0.4 using ~30 particles (see Fig. S4 in supplementary information). On average, the size of Fe-rich nanostructures is 12.1 ± 4.8 nm (Fig. 5). We don't observe any significant variation in particle size with respect to the depth of the film.

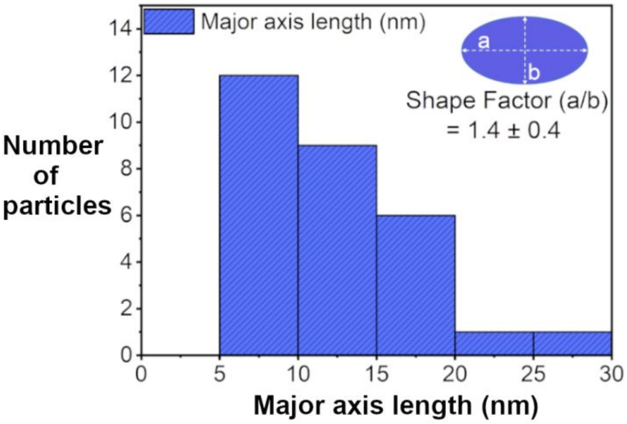

Fig. 5: Particle size distribution of Fe-rich nanostructures (based on major length 'a') formed throughout the LSF film via bulk exsolution.

The fact that the exsolved particles can be ellipsoidal points to the possibility of further designing this property through spatially orienting the particles. One might wish to orient the major axis of the ellipsoids to be parallel, perpendicular, or random, and this work would be foundational to developing such a future hypothesis. An example of functionality related to spatial orientation and vicinity of the produced nanostructures is electronic transport, wherein orienting the major axes of the particles in a parallel fashion could provide a more direct percolating pathway from particle-to-particle, thus enhancing conductivity through the particles. Alternatively, if the particles were designed for light absorption and/or scattering, it could be beneficial to orient their major axis parallel to incident illumination for optimal photon coupling.

Quantitative analysis of bulk and surface nanostructures. The elemental compositional analysis for the various nanostructures demonstrates that the exsolution of Fe directly affects the nearby regions in the film. We observe two main phenomena with quantitative analysis: (a) formation of lanthanum ferrite or $La_xFe_yO_{3-\delta}$ (LFO) around Fe particles because of reaction with the LSF matrix, and (b) formation of Fe-depleted LSF adjacent to many Fe particles. Extracting the absolute composition of the various Fe-based nanostructures is challenging due to unavoidable contribution from the LSF matrix through the thickness of TEM lamellae specimen. However, useful results were obtained on the relative compositional changes between the exsolved nanostructures and the matrix. We quantified the concentration of cations using the Cliff-Lorimer method [57–59].

The surface core-shell particle is rich in Fe and depleted in both La and Sr, compared to the LSF matrix (Fig. 6a). There is a decrease in HAADF intensity at the 'shell' region of the particle (points 1 and 4). Points 2 and 3 correspond to the particle core region, while point 5 corresponds to the LSF matrix. Point 2 corresponds to about 97 at% Fe, demonstrating that the core of the particle is Fe-based only [42]. The gradual increase of La content and negligible change in Sr content, from the core to shell region, indicate that the Fe core is surrounded by LFO. Considering the sum of all cation concentrations to be 100 at%, the LSF matrix (as prepared) correspond to 50 at% Fe, 30 at% La and 20 at% Sr. Here, the LSF matrix region is 36.5 at% Fe, 42.4 at% La and 21.1 at% Sr (Fig. 6a, point 5). This indicates that exsolution of the surface Fe particle resulted in depletion of Fe in the surrounding perovskite matrix.

Although the contribution of the LSF matrix is more prominent in the bulk core-shell particle, our analysis suggests that the core is Fe-rich and the shell is made of LFO like the surface particle, Fig. 6b. We observe local increase in HAADF intensity corresponding to the core of bulk core-shell particle (point1, Fig. 6b). Like the surface particle, the core is rich in Fe and depleted in La and Sr. The increasing concentration of La in the shell region (between points 2 and 3, Fig. 6b), while the Sr concentrations remains similar, indicates that the shell is made of LFO. This is like the observation made for the surface particle.

The quantitative results from adjacent particles show that exsolution of Fe in the LSF film can result in Fe-depleted LSF in the immediate vicinity of the Fe particles, Fig. 6c. The increase in HAADF intensity (point 2, Fig. 6c) corresponds to the position of the Fe particle, while the decrease in HAADF intensity (point 3, Fig. 6c) corresponds to the Fe-depleted particle. Going from the bright to dark particle, the Fe concentration gradually decreases while La and Sr concentrations gradually increase. However, the increase in Sr concentration is less prominent than La. We don't observe an increase in Fe concentration at the onset of the bright particle (between points 1 and 2, Fig. 6c) due to contribution from another superpositioned Fe-exsolved particle; the particle is slightly out of focus in the HAADF image but within the interaction volume. Furthermore, consistent La and Sr concentrations along the bright particle suggests it is the contribution from LSF matrix.

The quantitative analysis of independent particles confirms our hypothesis in the previous section that the composition of these particles is like bulk core-shell particles, with a primarily Fe 'core' region and thin LFO 'shell' (Fig. 6d). The Fe core location corresponds to a local maximum in HAADF intensity (point 4, Fig. 6d), and a plateau in the Fe content profile (5 nm – 15 nm). The relatively consistent La and Sr content within the particle (between points 3 and 5, Fig. 6d) is attributed to the LSF matrix concentration within the interaction volume. The elemental concentrations in the LSF matrix region are 52.7 at% Fe, 30.4 at% La and 16.9 at% Sr (point 1, Fig. 6d). Like the observation of the bulk core-shell particle, the La content increases going from the core to shell region (point 3 to 2, Fig. 6d), whereas the Sr signal remains unchanged, showing the presence of LFO shell around the independent particles. This is also evident by the dark contrast around the Fe-rich particles in the HAADF image of Fig. 6d.

The dark HAADF contrast of the LFO shell formed around the Fe particles (Fig. 6) suggests lower mean Z compared to the LSF film matrix, but can also be due to structural distortion at phase boundaries affecting the HAADF signal. Mean Z for a compound can be calculated using different methods; we used the following to estimate mean Z values [63]:

$$Z = \sqrt{\sum_{i=1}^{n} a_i Z_i^2}$$

where, a_i is the atomic fraction and Z_i is the atomic number of the constituent elements. Mean Z for $LaFeO_3$ and LSF are calculated to be 28.7 and 26.1 respectively. Since $LaFeO_3$ has higher mean Z than LSF, the shell would have a higher HAADF signal than the matrix if it was composed of $LaFeO_3$, which is not the case here. Thus, we can assume that La:Fe ratio is not 1:1 in the shell. Based on our quantitative analysis of shell regions as well as mean Z calculations, we believe the shell around the Fe particles to be a La-depleted and Fe-rich (perovskite-like) oxide, compared to the LSF matrix. The core-shell nature of some nanostructures could, for example, be leveraged in a situation where one wishes to tune a film's magnetic properties. In this case a variable to optimize is the thickness of the oxide shell and thus the volumetric ratio of metal core to metal oxide shell.

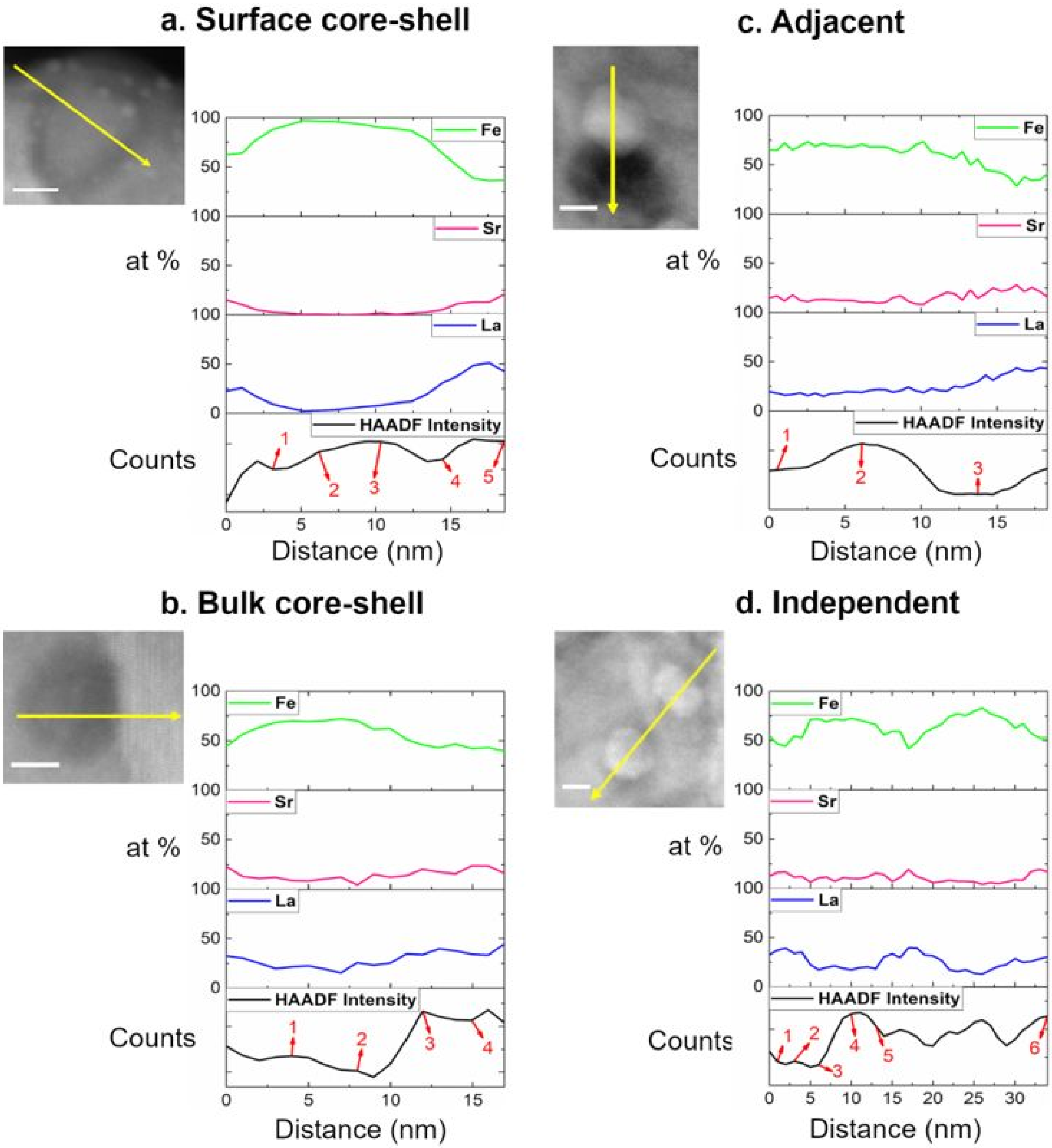

Fig. 6: STEM-HAADF images and line profiles showing HAADF image intensity (counts) and elemental concentration (at%) of cations Fe, La, and Sr across surface core-shell (a), bulk core-shell (b), adjacent (c), and independent particles (d). Scale bar in HAADF images is 5 nm.

Formation of Ruddlesden-Popper phase (LSF_{214}) *via* bulk exsolution. STEM characterization and FFT analysis confirm that Fe exsolution in the LSF film creates local A-site rich domains of RP phase. The general formula of RP phase can be given by $A_{n+1}B_nO_{3n+1}$; i.e. A_2BO_4 for n=1 [64]. We found an inhomogeneous distribution of RP phases throughout the LSF film. XRD results, detailed elsewhere[42], showed that perovskite LSF continues to be the dominant phase after exsolution with chemical expansion in the LSF lattice due to formation of oxygen vacancies.

Many Fe nanostructures are observed to form at the boundaries of the LSF perovskite matrix and RP phases. Fig. 7a shows a bulk core-shell particle formed at the phase boundary between the original perovskite matrix (LSF_{113}) and RP phase (LSF_{214}) (Fig. 7).

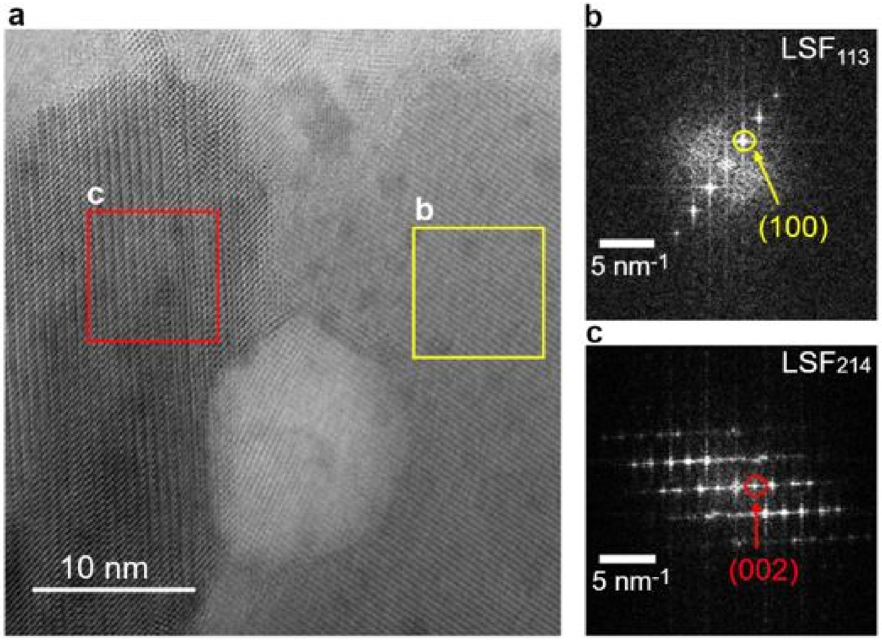

Fig. 7: STEM-BF image (a) and FFT patterns of LSF_{113} and LSF_{214} respectively (b-c).

In the same specimen, we observe different stages of RP phase formation near the surface of the LSF film (Fig. 8). For the partially transformed phase (Fig. 8a-b), (200) planes with non-uniform linear contrast appear to be connected by zig-zag patterns, which is attributed to disordered planar defects [65]. These planar-like defects are manifested as horizontal streaking on the FFT pattern (Fig. 8c). In the fully transformed RP phase (Fig. 8d-e), we see well-ordered arrangement of planar defects, shifted by ½a[100] as can be seen clearly in the inset of the HAADF image (Fig. 8d). Such ordered planar defects are commonly observed in A-site rich oxides, and have been discussed previously by Suzuki *et al.*[65] This results in sharp FFT spots, along with superlattice spots, as shown in Fig. 8f.

RP phase formation (simultaneous to Fe-exsolution) was also observed in a model epitaxial thin-film LSF (~20nm) [see details in supplementary section, Fig. S5]. The characterization of surface exsolution for this thin-film LSF has been previously published[6]. Our results on both the thick LSF film (~165 nm) and thin LSF film (~20 nm) confirm that with modifications in exsolution treatments, the formation of RP phase can be controlled to optimize exsolution of catalytic particles and thus achieve improved catalytic efficiency.

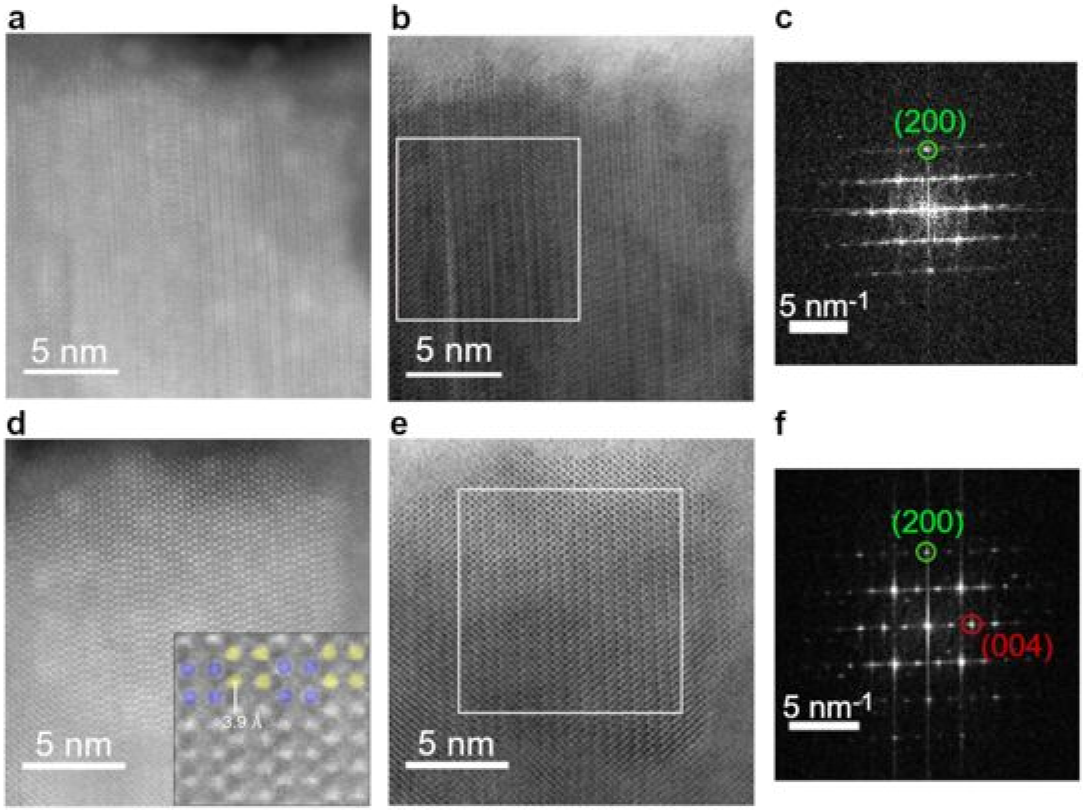

Fig. 8: Observation of two different stages of the RP phase formation via exsolution process shown with STEM HAADF images (a,d) and STEM BF images (b,e). Corresponding FFT patterns (c,f) are also shown. The yellow and purple circles in Fig. 8d highlight the shift of atomic columns in the fully transformed RP phase.

Conclusions

We report four distinct types of Fe-based nanostructures formed in LSF thin films (on MgO substrate) via bulk and surface exsolution: surface core-shell, bulk core-shell, adjacent and independent nanostructures. The core-shell particles are determined to have a metallic Fe core and LFO shell. The adjacent nanostructures are Fe exsolved particles with Fe-depleted LSF in their immediate vicinity. The independent nanostructures are found to have the same structure and chemistry like the bulk core-shell particles, but with a prominent Fe core region and very thin ~ 1-2 nm LFO shell. The exsolved bulk and surface particles have a shape factor of 1.4. We show that Fe exsolution creates local A-site rich domains of RP phase and two different stages of this phase formation via exsolution process is captured and analyzed using STEM imaging in this work. The phase boundaries, such as between RP phase and perovskite matrix, are attributed to be the primary sites for surface and bulk exsolution, suggesting the activation energy of Fe nucleation is lower at these boundaries. This work presents an in-depth nanostructure-based analysis of bulk and surface exsolution in thin films, offering a beneficial guide for scientists and engineers interested in exploring exsolution to optimize advanced materials. In the future it would be valuable to systematically investigate the impact of film synthesis and processing conditions on the amount of each nanostructure type, to establish more predictive synthesis protocols for each type of nanostructure. This should point the way forward towards designing and realizing tunable functional materials containing a specific subset of the nanostructures presented here.

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Conflict of Interest

None.

Author Contributions

WJB and KS designed the study. JW prepared the thin films. KS conducted the STEM characterization and data analysis. KS wrote the manuscript and all authors contributed to editing/revision. WJB supervised the project.

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