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High ionic conductivity in fluorite δ -bismuth oxide-based vertically aligned nanocomposite thin films†

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δ -Bi₂O₃ has long been touted as a potential material for use in solid oxide fuel cells (SOFC) due to its intrinsically high ionic conductivity. However, its limited operational temperature has led to stabilising the phase from >725 °C to room temperature either by doping, albeit with a compromise in conductivity, or by growing the phase confined within superlattice thin films. Superlattice architectures are challenging to implement in functional μ SOFC devices owing to their ionic conducting channels being in the plane of the film. Vertically aligned nanocomposites (VANs) have the potential to overcome these limitations, as their nanocolumnar structures are perpendicular to the plane of the film, hence connecting the electrodes at top and bottom. Here, we demonstrate for the first time the growth of epitaxially stabilised δ -Bi₂O₃ in VAN films, stabilised independently of substrate strain. The phase is doped with Dy and is formed in a VAN film which incorporates DyMnO₃ as a vertically epitaxially stabilising matrix phase. Our VAN films exhibit very high ionic conductivity, reaching 10⁻³ S cm⁻¹ at 500 °C. This work opens up the possibility to incorporate thin film δ -Bi₂O₃ based VANs into functional μ SOFC devices, either as cathodes (by pairing δ -Bi₂O₃ with a catalytically active electronic conductor) and/or electrolytes (by incorporating δ -Bi₂O₃ with an insulator).

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Preface

We are delighted to write this article to honour the long and distinguished career of Prof. John Kilner (JK). Judith MacManus-Driscoll (JLM-D), the last author of this article, first met JK in 1985 when she was an undergraduate at Imperial College London. JK was, at the time, working very closely with Brian Steele who connected JLM-D with the ionics field, *via* 2 summer internships with the ionic groups at the Max Planck Institute FKF. JLM-D then undertook her final year undergraduate research project working on studying the ionic conductivity and diffusivity of oxygen in δ -Bi₂O₃. Thanks to JK, within a few short weeks excellent results came out. At the same time, JK's outstanding lectures on transition metal oxides (JLM-D recalls references to the pioneering work of John Goodenough on the electronic structure of transition metal oxides) and on the excitement of the discovery of high temperature superconductivity in cuprates spurred JLM-D to do a PhD in that area. Upon receiving this invitation to write a paper to honour JK's

distinguished career, JLM-D, her team, and collaborators had the perfect opportunity to revisit δ -Bi₂O₃, after a nearly 35 year break. In presenting this work, we thank JK for his great dedication and pioneering work in the ionics field and beyond. We aim to convince the reader that δ -Bi₂O₃ can be stabilised from >725 °C down to room temperature in thin films using the vertically aligned nanocomposite (VAN) approach.

Introduction

To this day, cubic fluorite δ -Bi₂O₃ remains the highest ionic conducting material of all known oxygen ion conductors, achieving 1 S cm⁻¹ at 730 °C, up to two orders of magnitude higher conductivity than the industrial standard yttria stabilized zirconia oxide (YSZ).^{1,2} This high ionic conductivity is attributed to the inherent defect structures with 25% oxygen vacancy concentration, the weak Bi-O bond and high polarizability of the Bi³⁺ ion.³ However, δ -Bi₂O₃ has a limited operational temperature range in SOFC and ionic devices, melting above 823 °C, and undergoing a phase transition to the monoclinic α -phase below 730 °C, resulting in a greatly reduced conductivity.² Many studies have demonstrated the fluorite phase can be stabilised to room temperature with carefully selected dopants, predominantly trivalent lanthanides, but this comes at the expense of a reduction in the ionic conductivity, by up to 4 orders of magnitude.^{2,4-6} Extensive research effort has been focused on finding the optimum dopant and

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profilometer. EIS data was modelled using a Randles circuit. All EIS measurements were carried out on films grown on electronically conducting Nb-STO.

Results and discussion

Clear phase separation between high crystalline and epitaxial DSB-DMO phases in a VAN structure is evident in scanning transmission electron microscopy (STEM) images taken in the high-angle annular dark-field (HAADF) mode, both in cross-section (Fig. 1a), high resolution TEM (Fig. 1b), selected area electron diffraction (SAED) patterns (Fig. 1c) and the corresponding energy dispersive X-ray spectroscopy (EDS) maps (Fig. 1d and e).

The DSB forms nanopillars (red, Fig. 1d) of ~ 30 nm diameter and they are uniformly distributed throughout a DMO matrix (blue, Fig. 1d). The DSB columns protrude the surface of a very dense film (Fig. 1b), with pyramidal faceting indicating (111) termination. The phase boundaries between the two phases are very clean (Fig. 1b), *i.e.* no secondary phases are present, which is expected since VAN structures form by thermodynamic self-assembly. STEM and EDS confirm our films contain Dy-rich DSB, with a stoichiometry of $(\text{Dy}_2\text{O}_3)_{0.6\pm 0.1}(\text{Bi}_2\text{O}_3)_{0.4\pm 0.1}$ (determined from 4 regions), with a minor Mn content ($2 \pm 0.7\%$ atomic fraction). The high Dy content in the Bi_2O_3 is expected due to the high Dy content in the target. The DMO has a stoichiometry of $\text{Dy}_{0.9\pm 0.1}\text{Mn}_{1.00\pm 0.03}\text{Bi}_{0.16\pm 0.02}\text{O}_{3\pm\delta}$.

The film orientation and lattice parameters were studied by high resolution XRD (Fig. 2). Symmetric $2\theta-\omega$ scans (Fig. 2a) are indexed to a (00 l) oriented cubic phase with $a = 5.418(2)$ Å, which is well matched to the Dy stabilised $\delta\text{-Bi}_2\text{O}_3$ phase.¹ A weak reflection at $28.51^\circ \pm 0.05^\circ$ is indexed to a very minor fraction of (111) DSB. From φ -scans (Fig. 2b), we confirm that the DSB phase is highly aligned and in-plane rotated 45° with an epitaxial relationship of $[100]\text{DSB}/[110]\text{STO}$. This is expected as $a_{\text{DSB}} \approx \sqrt{2}a_{\text{STO}}$. By extension, DMO epitaxially aligns with DSB with the same 45° relationship, as seen in other REMnO_3 phases.²¹⁻²³ This is also expected as DMO has the GdFeO_3 perovskite structure.

To further confirm the presence of the cubic δ -fluorite phase, detailed high-resolution asymmetric X-ray reciprocal space maps (RSMs) were collected around the (103) STO reflection (Fig. 2c). This reveals a single broad reflection which is indexed to cubic DSB (204) with $a = 5.38(1)$ Å. Additional evidence that the DSB is indeed the δ fluorite phase of Bi_2O_3 comes from indexing the SAED pattern in TEM (Fig. 1c). Taken from the STO [100] zone axis, the SAED pattern shows a $(2\bar{2}0)\text{DSB}/(020)\text{STO}$ and $[110]\text{DSB}/[100]\text{STO}$ orientation relationship, consistent with XRD φ -scans (Fig. 2b), and is indexed with $a = 5.452(5)$ Å, congruent with stabilised $\delta\text{-Bi}_2\text{O}_3$.¹ Non-indexed SAED spots arise from DMO. The cannot be indexed to the other known phases of Bi_2O_3 (see Fig. S1† for discussion). Hence, these observations taken together are conclusive proof that cubic DSB can be grown epitaxially in VAN films grown on STO substrates.

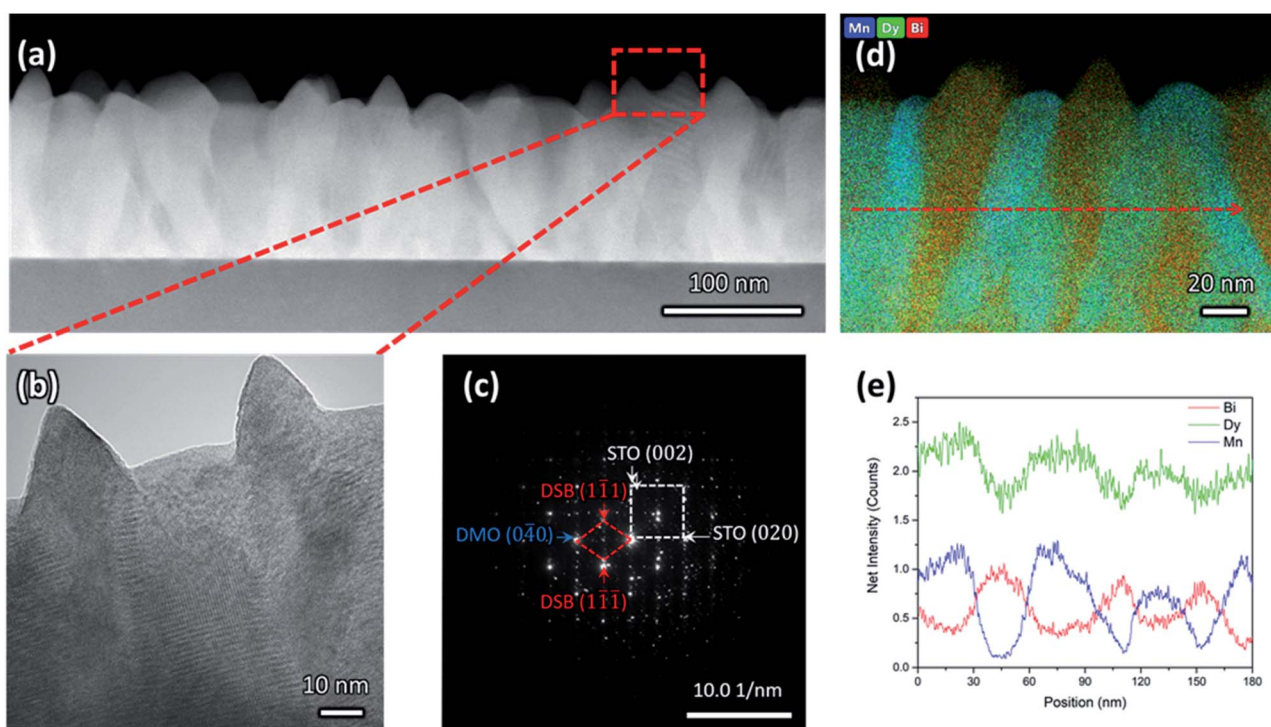


Fig. 1 (a) Low magnification cross-sectional HAADF-STEM shows clear columnar features that are uniformly distributed (b) high resolution TEM displaying DSB nanopillars which protrude the film surface with pyramidal (111) termination. (c) SAED pattern in STO [001] zone axis indexed to $(2\bar{2}0)\text{DSB}/(020)\text{STO}$ and $[110]\text{DSB}/[100]\text{STO}$. (d) EDS map and (e) the corresponding line scan showing clear separation between DSB and DMO phases. Bi (red) and Mn (blue) are localised, whereas Dy (green) is distributed throughout the structure.



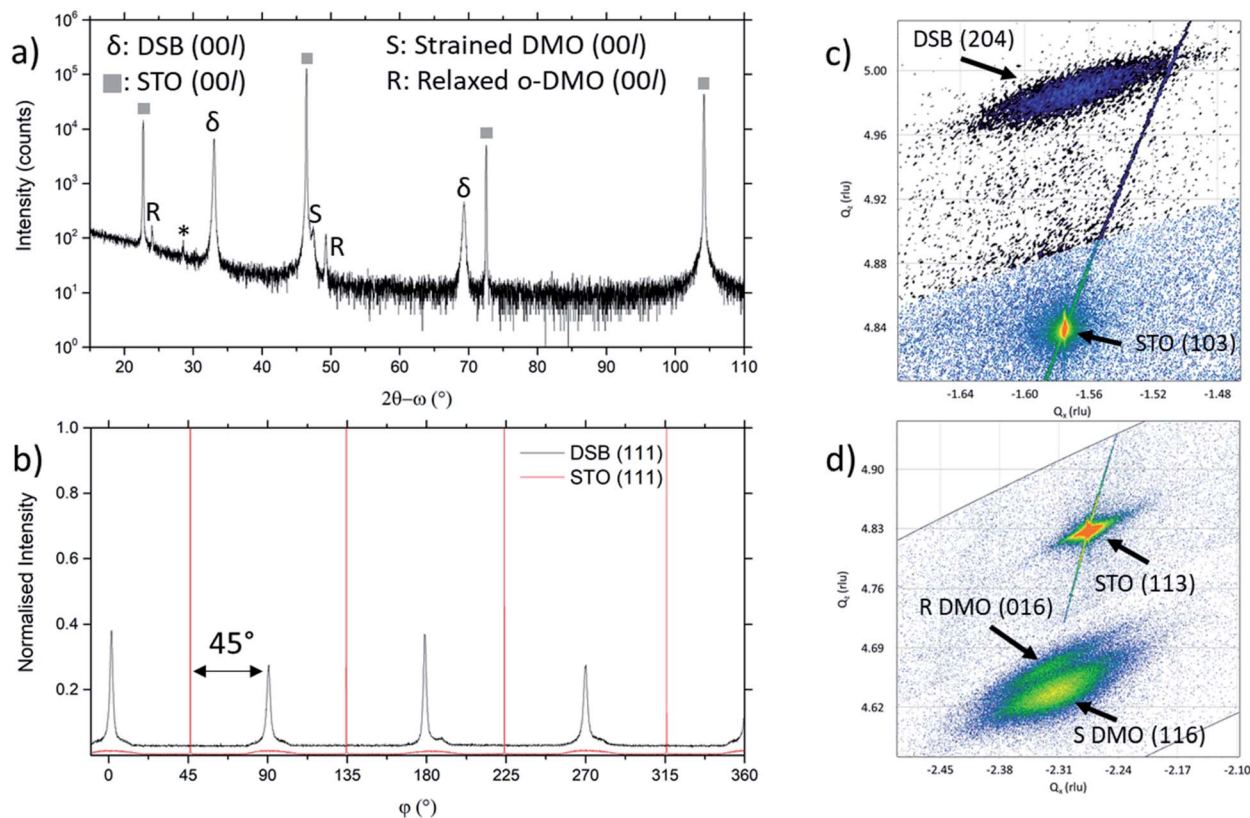


Fig. 2 (a) 2θ - ω scans of DSB-DMO VAN films show clear presence of (00l) orientated δ -fluorite DSB and DMO phases. A very small (111) DSB reflection is also present at $28.51^\circ \pm 0.05^\circ$, marked by * (b) ϕ scan of (111) STO and DSB reflections, confirming epitaxial relationship between DSB and STO of [100]DSB//[110]STO (c) asymmetric RSM around the STO (103) reflection. The DSB (204) is observed nearby, confirming the fluorite phase (d) asymmetric RSM around the (113) reflection of STO. The (116) reflection of the tetragonally strained DMO is clearly present, alongside the (016) reflection of relaxed DMO.

We now consider the role of the DMO. The DMO forms mainly as a highly strained phase in the VAN film. This is seen from 2θ - ω XRD scans (Fig. 2a), which shows two reflections to the right of the (002) STO, corresponding to the strained (S) and relaxed (R) DMO phase. The reflection labelled S at lower 2θ arises from DMO ($a = 5.280 \text{ \AA}$, $b = 5.832 \text{ \AA}$, $c = 7.381 \text{ \AA}$)²⁴ which is compressed in-plane by STO (bulk average pseudo-cubic lattice parameter of DMO is 3.932 \AA vs. STO 3.905 \AA) resulting in a 3.64% out-of-plane tensile strain with $c = 7.650(5) \text{ \AA}$, as indexed. The VAN structure enables the high strain levels, much higher than standard epitaxial thin films.^{25,26} An additional reflection observed at higher 2θ , labelled R, indexed to $c = 7.393(5) \text{ \AA}$ corresponds to relaxed DMO which adopts the bulk lattice parameters. This relaxation occurs at the upper part of the film away from the substrate and is minor contribution.

To understand more about the DMO phase, detailed high-resolution asymmetric X-ray reciprocal space maps (RSMs) around the (113) reflection of STO were performed (Fig. 2d). Two broad reflections are witnessed in close proximity to STO (113). We note that both reflections show greater in-plane (Q_x) broadening than out-of-plane (Q_z). This is related to thermal contraction effects particular to VAN films,²⁵ and is not relevant to the uniform out-of-plane straining, which is the key factor for controlling the stabilisation of the δ -phase of DSB. The larger

reflection can be indexed to a tetragonal phase with lattice parameters of $a = b = 5.528(5) \text{ \AA}$, $c = 7.650(5) \text{ \AA}$, and is assigned as the (116) reflection of the strained DMO phase as seen in the 2θ - ω scans (Fig. 2a). This confirms that the DMO strains from an orthorhombic to tetragonal unit cell with a 0.63% tensile strain in the [110] direction. We note the significance of 5.52 \AA , as it is equal to $\sqrt{2}a_{\text{STO}}$, indicating that DMO is in-plane straining to perfectly match the [110]STO direction. A second broad reflection is observed which corresponds to the (016) of the relaxed DMO, and can be indexed with $b = 5.837(5) \text{ \AA}$, $c = 7.393(5) \text{ \AA}$, thus consistent with the bulk DMO.²⁴ These observations are similar to previously reported $(\text{SmMnO}_3)_{0.5}(-\text{Bi,Sm})_2\text{O}_3)_{0.5}$ VAN films.²¹

In our film, the DMO matrix plays an important role in stabilising the cubic DSB phase; the strained DMO and DSB have a $\sim 2\%$ lattice mismatch. Thus, the strained DMO matrix acts as scaffold to stabilise and promote the formation of cubic DSB nanopillars. Here, vertical strain is induced by the DMO matrix on the DSB pillars in the VAN architecture, enabling stabilisation of the δ -fluorite phase throughout the thickness of the VAN film. Thus, strain relaxation is controlled when confined within VAN, unlike standard planar films where strain is localised to a $\sim 20 \text{ nm}$ region at the film/substrate interface resulting in relaxation as the film thickness.¹¹ In fact, vertical strain



stabilisation in VAN can enable stabilisation to micron thick films.²⁷ This should enable the growth of robust, thick films of the δ -phase, $>50\times$ thicker than previous superlattice films.^{12,13}

We now discuss the excellent ionic properties of our DSB-DMO VAN films grown on electronically conducting Nb-STO. The out-of-plane ionic conductivity of the DSB-DMO VAN films was studied with EIS, presented in Fig. 3. Nyquist plots, a typical example of which is presented in Fig. 3b, display two features that can be modelled with a Randles circuit (Fig. 3b inset); a small arc at high frequency corresponding to the ionic conductivity of the DSB, modelled by R_i and C_i ; and a finite length Warburg (FLW) response appearing as a 45° line at higher frequencies before transitioning into a semi-circle at low frequencies. This FLW response models the complex impedance arising from the diffusion of oxide ions at the electrode interface. From Fig. 3a, it is clearly observed that the ionic conductivity of the DSB-DMO VAN is thermally activated and obeys Arrhenius behaviour with an activation energy of 1.00 ± 0.02 eV, reaching 10^{-3} S cm⁻¹ at 500 °C. These values are well matched with bulk DSB (1.05–1.12 eV)⁵ and comparable with erbium stabilised δ -Bi₂O₃ superlattices (1.05 eV MgO/CGO/[ESB/CGO]_{N=1}, 1.18 eV MgO/CGO/[ESB/CGO]_{N=20}).¹² The high ionic conductivity again confirms the presence of the δ -fluorite structure in our VAN films. The ionic conductivity values are moderately lower ($3 - 8x$) at 500 °C (1.0×10^{-3} S cm⁻¹ our film) than previously reported for Er-doped superlattice films (3.3×10^{-3} S cm⁻¹ MgO/CGO/[ESB/CGO]_{N=1}, 8.1×10^{-3} S cm⁻¹ MgO/CGO/[ESB/CGO]_{N=20}).^{12,13} However, it is important to note that the magnitude of ionic conductivity in doped δ -Bi₂O₃ decreases with increasing dopant concentration,^{1,2,5} and our dopant concentrations are higher than reported superlattice films ($x = 0.6$ cf. nominally $x = 0.2$ (Er) in [ESB/Ce_{0.8}Gd_{0.2}O_{2- δ]_N,¹² $x = 0$ in [δ -Bi₂O₃/YSZ]_N but with possible yttrium intermixing¹³). As can}

be seen from Fig. 3a, the ionic conductivity of our films is at a value that is consistent with bulk (Dy₂O₃)_x(Bi₂O₃)_{1-x} with $x \approx 0.40$.⁵ Since, from EDS maps (Fig. 1e), our $x = 0.6 \pm 0.1$, this indicates the VAN structure has enhanced ionic conduction compared to bulk material of this same composition⁵ by a factor of $\sim 50\times$ at 500 °C. This could result from either the enhanced crystallinity of the DSB in the film, and/or possible nanostructuring effects.^{14,28}

It may be possible to further enhance the ionic conductivity of our films by fine tuning the excess Bi₂O₃ content such that the Bi:dopant ratio is smaller, hence accessing higher conductivities. Alternatively, different stabilising matrixes could be utilised, or different volume fractions of the matrix/column in the VAN, leading to different strain levels in the δ -Bi₂O₃, which may further enhance the ionic conductivity.^{29,30}

We note that bulk DMO displays electronic conductivity, with an E_a of 0.268 eV and conductivity greater than 10^{-1} S cm⁻¹ above 320 °C,³¹ distinctly different from the observed EIS feature (Fig. 3). To further affirm we are not measuring the EIS of DMO in Fig. 3, we present the EIS of a reference Ag/Nb-STO/DMO/Pt system in Fig. S2.† By comparison, it is clearly seen that the impedance of the DMO planar system is very different to the DSB-DMO VAN. Thus, we conclude the EIS of our VAN system is dominated by DSB. But, from the microstructure in Fig. 1a, there may be a very thin layer of DMO between the DSB column and Nb-STO substrate. In the context of our electrode setup, the DMO acts as an extension of the electrode, and its impedance contribution is contained within R_{serial} (which also contains contributions from the Nb-STO substrate and the internal resistances of the EIS setup). As the electronic conductivity of DMO would yield a resistance $\sim 100\times$ smaller than R_{serial} , it is expected that this feature cannot be deconvoluted from R_{serial} .

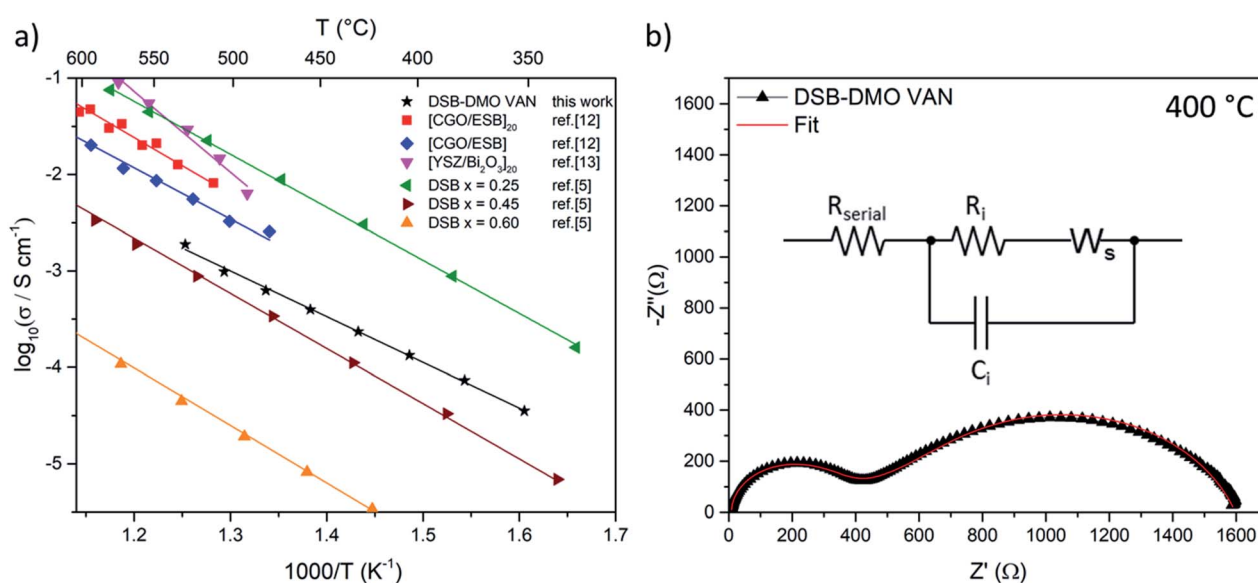


Fig. 3 (a) Arrhenius plot of DSB-DMO VAN film in air displaying high ionic conductivity as compared to the literature results for superlattices^{12,13} and bulk powder DSB⁵ (b) typical Nyquist plot at 400 °C. Inset: the Randles model used to fit the data, where R_i/C_i models the ionic conductivity of DSB, W_s the FLW response, and R_{serial} the internal resistances of the EIS setup.



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