Solid oxide proton conductors beyond perovskites

Sacha Fop

Solid oxide proton conductors are crucially emerging as key materials for enabling hydrogen-based energy conversion, storage, and electrochemical technologies. Oxides crystallising in the ideal ABO$_3$ perovskite structure, such as barium cerates and zirconates, are widely investigated thanks to their excellent proton conducting properties. Nevertheless, alternative structure-type solid oxide systems (hexagonal perovskite derivatives, brownmillerite, scheelite, etc.) can efficiently incorporate and enable the transport of protonic defects, with recent reports of materials exhibiting high ionic conductivity comparable to the conventional perovskite conductors. This perspective provides an overview of these alternative and less established proton conducting materials, with particular attention to the relationship between the structural and ionic conduction features and the mechanistic aspects. The goals are to highlight the differences between these materials and the traditional perovskites and to point out new potential crystal routes for the discovery of innovative solid oxide proton conductors.

1. Introduction

Carbon emissions must be reduced to mitigate the effects of climate change and keep the global average temperature increase below 2 °C above pre-industrial levels. To achieve this target, it is essential to transition away from fossil fuels and progress towards cleaner and renewable means of energy generation. Low carbon solutions provided by hydrogen-based energy conversion, storage and electrochemical technologies will have a fundamental role in contributing to broad decarbonisation.$^1$–$^4$ Conventional solid-state hydrogen-based electrochemical devices such as ceramic fuel cells and electrolyzers employ oxide ion conducting materials. These systems however generally show adequate ionic conductivities only at high working temperatures (>700 °C), thus posing technical challenges in terms of materials selection and durability and limiting the widespread application of these technologies.$^5$–$^7$

Systems based on solid oxide proton conductors offer promise of operations at reduced temperatures, thanks to the generally lower activation energy for proton diffusion ($\sim$0.3–0.6 eV) compared to oxide ion conduction in oxides ($\sim$0.6–1.2 eV).$^8$–$^9$

Iwahara et al. were the first to demonstrate the potential of perovskite-type solid oxide proton conductors in energy-related applications in the early 1980s and 1990s.$^8$–$^9$ Since then, the most intensively studied solid oxide proton conductors became doped barium cerate- and zirconate-based perovskite-type oxides, due to their high proton conductivities. The properties of Ba( Ce, Zr)O$_3$ perovskite proton conductors are well established and have been extensively reviewed.$^8$–$^9$, $^{15}$–$^{22}$

The crystal structure of BaCeO$_3$ and BaZrO$_3$ materials is based on the perovskite ABO$_3$ structure, which is composed by a network of corner-sharing BO$_6$ octahedra in which the larger A cations occupy the interstitial voids (Fig. 1a). Depending on the composition and the level of hydration, the ideal cubic perovskite structure (space group Pm3m) is usually stable at high temperatures, with rhombohedral, orthorhombic, and monoclinic distortions at lower temperatures due to tilting of the octahedral units.$^{21}$–$^{26}$ Proton conduction in these systems is enabled by the introduction of protonic defects via the dissociative absorption of water according to the hydration reaction,
Cerate and zirconate perovskites represent one of the most promising group of solid oxide proton conductors, as demonstrated by reports of excellent performances at low temperatures (300–600 °C) in proton ceramic fuel cells (PCFCs), electrolysis cells (PCECs), and reactor cells. For example, a peak power density of 548 mW cm\(^{-2}\) at 500 °C has been reported for a PCFC with an optimized BaZr\(_{0.4}\)Ce\(_{0.6}\)Y\(_{0.1}\)Yb\(_{0.1}\)O\(_{3}\) electrolyte,\(^{48}\) a value that greatly exceeds the performance of commercially available solid oxide fuel cells with yttria stabilized zirconia electrolytes.\(^{3,48}\) However, barium cerate- and zirconate-based electrolytes also have some distinct disadvantages. BarCeO\(_3\)-based compounds are chemically unstable under CO\(_2\) and H\(_2\)O, usually decomposing into BaCO\(_3\) or Ba(OH)\(_2\) and the corresponding oxide.\(^{25–37}\) BaZrO\(_3\)-based materials show greater stability,\(^{28,29}\) but their refractory nature makes grain growth difficult and results in a large grain boundary resistance which is detrimental for the overall conductivity.\(^{16,20,60,61}\)

In the last two decades, research into alternative solid proton conductors possessing good ionic conductivity and stability has led to the discovery of proton conduction in several structural families. These include perovskite derivative oxides, systems containing tetrahedral moieties and fluorite-related oxides, with recent reports of materials exhibiting proton conductivities comparable to benchmark Y-doped BaCeO\(_3\) and BaZrO\(_3\) perovskite conductors (Fig. 2). These systems have promising properties, and in some cases show structural and proton conducting characteristics which are considerably different from the ones of conventional perovskite proton conductors (Table 1). Although citing numerous and state-of-the-art studies, it is beyond the scope of this perspective to give an exhaustive summary of all the properties of these alternative proton conducting oxide systems. Rather, this perspective

\[
\text{H}_2\text{O}(\text{g}) + V_0^* + O_0^* \leftrightarrow 2\text{OH}_0^*.
\]

Water dissociates into a proton and a hydroxide ion by filling of an oxygen vacancy.\(^{8,27}\) Pure BaCeO\(_3\) and BaZrO\(_3\) show limited proton incorporation due to the absence of available oxygen vacancies. Water absorption, and hence the creation of protonic vacancies, is promoted by the introduction of oxygen vacancies via trivalent acceptor doping on the Ce/Zr-site. Y is the most commonly used acceptor dopant for effective proton conduction,\(^{16,20}\) although numerous dopants have been tested.\(^{28–32}\) The highest ionic conductivities are reported for barium cerate–zirconate solid solutions co-doped with Y and Yb.\(^{33,34}\) The dissociative absorption of water results in the incorporation of a hydroxyl ion and in the formation of a covalent bond between a proton and a lattice oxygen atom. Protons are generally located in proximity of the O–O octahedron edge (Fig. 1b)\(^{24,25,37,36}\) and have the tendency of forming hydrogen bonds with neighbouring oxygen atoms.\(^{37}\) Proton transport follows a Grotthuss mechanism characterised by the fast rotational diffusion of the protonic defect around an oxygen atom (with activation energy of \(\sim0.08–0.2\) eV), followed by intra-octahedral hopping of a single proton towards a neighbouring oxide ion (\(\sim0.4–0.5\) eV) and successive migration to an oxygen of an adjacent octahedron (Fig. 1c).\(^{38–42}\) Under humid reducing atmospheres these systems show pure proton conduction, with a proton transport number \(t_H\) (which is the ratio of proton conductivity to total conductivity) close to unity.\(^{13,43–45}\) However, at high temperatures and under higher partial pressure of oxygen, both barium cerates and zirconate present mixed oxide ion and electronic p-type conductivity, with \(t_H\) generally reducing to \(\sim0.6–0.7\) at 500 °C, due to an increase in the mobility of the oxygen vacancies and of the concentration of hole defects.\(^{13,46,47}\)

Cerate and zirconate perovskites have been shown to be superior to other oxide electrolytes, and their superior physical properties make them the preferred oxide electrolyte for PCFCs.\(^{21,36,37}\) The bulk conductivity of BaCe\(_{0.9}\)Y\(_{0.1}\)O\(_{3}\) (BCY)\(^{8}\), the extrapolated bulk conductivity data of BaZr\(_{0.8}\)Ce\(_{0.2}\)O\(_{3}\) (BZO)\(^{8}\) and the conductivity of the solid oxide ion electrolyte Zr\(_{0.92}\)Y\(_{0.08}\)O\(_{1.96}\) (YSZ)\(^{8}\) are shown for comparison.
Table 1  Summary of key properties of selected proton conducting oxide structures. Conductivity values are for measurements taken under typical conditions, i.e. air saturated with water (water partial pressure $p_{H_2O} \sim 0.023$ atm). Where not reported, proton transference numbers were estimated from the values of conductivity in dry and humidified atmospheres, calculated according to the relationship $\sigma_{\text{wet}} - \sigma_{\text{dry}}/\sigma_{\text{wet}}$ (see ref. 114). The obtained values are indicated with an asterisk and represent an overestimate, as they are calculated under the assumption that any electronic conductivity is negligible. The concentration of water uptake (which is directly correlated to the proton concentration, $[\text{OH}]$) observed under typical conditions at ca. room temperature is expressed as the number of water molecules per formula unit, $n_{H_2O}$. When reported, the theoretical maximum number of water molecules that can be incorporated by the structure is also presented between brackets; this usually corresponds with the available vacant oxygen sites for water incorporation. Hydration enthalpy values ($\Delta H_{\text{hyd}}$) and experimental activation energy values of proton conductivity ($E^\ddagger$) are also reported.

<table>
<thead>
<tr>
<th>Compound</th>
<th>$\eta$, 300 °C (S cm$^{-1}$)</th>
<th>$\sigma$, 300 °C (S cm$^{-1}$)</th>
<th>$t_{\text{on}}$, 300 °C</th>
<th>$t_{\text{off}}$, 500 °C</th>
<th>$n_{H_2O}$</th>
<th>$\Delta H_{\text{hyd}}$, (kJ mol$^{-1}$)</th>
<th>$E^\ddagger$, (eV)</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td>BaCe$<em>{0.3}$Y$</em>{0.7}$O$_{3-\delta}$</td>
<td>$5.8 \times 10^{-4}$</td>
<td>$7.1 \times 10^{-3}$</td>
<td>$\sim 1$</td>
<td>$0.8-0.9$</td>
<td>$\sim 0.09 \ (0.1)$</td>
<td>$-163$</td>
<td>$0.53$</td>
<td>8 and 46</td>
</tr>
<tr>
<td>BaZr$<em>{0.6}$Y$</em>{0.4}$O$_{3-\delta}$</td>
<td>$3.5 \times 10^{-3}$</td>
<td>$1.8 \times 10^{-2}$</td>
<td>$\sim 1$</td>
<td>$0.98$</td>
<td>$0.178 \ (0.2)$</td>
<td>$-22 \ (T &lt; 500 °C)$</td>
<td>$0.47$</td>
<td>$-93 \ (T &gt; 500 °C)$</td>
</tr>
<tr>
<td>Ba$_2$Nb$<em>2$MoO$</em>{20}$</td>
<td>$7.0 \times 10^{-4}$</td>
<td>$4.0 \times 10^{-3}$</td>
<td>$0.35$</td>
<td>$\sim 0.80$</td>
<td>$\sim 0.80 \ (1)$</td>
<td>$-18 \ (T &lt; 300 °C)$</td>
<td>$0.57$</td>
<td>$-76 \ (T &gt; 300 °C)$</td>
</tr>
<tr>
<td>Ba$_3$Er$_2$Al$<em>2$ZrO$</em>{13}$</td>
<td>$9.8 \times 10^{-4}$</td>
<td>$3.0 \times 10^{-3}$</td>
<td>$\sim 1$</td>
<td>$\sim 1$</td>
<td>$\sim 0.27 \ (2)$</td>
<td>$-13 \ (T &lt; 400 °C)$</td>
<td>$0.40$</td>
<td>$-192 \ (T &gt; 400 °C)$</td>
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<tr>
<td>Ba$_2$Nb$_2$O$_5$</td>
<td>—</td>
<td>$6.0 \times 10^{-6}$</td>
<td>$\sim 0.2$ (at 700 °C)</td>
<td>$0.333 \ (0.333)$</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>73</td>
</tr>
<tr>
<td>Ba$_2$Ta$_2$O$_5$</td>
<td>—</td>
<td>$7.8 \times 10^{-7}$</td>
<td>$\sim 0.15$ (at 700 °C)</td>
<td>$\sim 0.375 \ (1)$</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>74</td>
</tr>
<tr>
<td>BaLaInO$_4$</td>
<td>$4.0 \times 10^{-8}$</td>
<td>$9.5 \times 10^{-7}$</td>
<td>$\sim 0.9$</td>
<td>$\sim 0.6$</td>
<td>$0.62$</td>
<td>$0.87$</td>
<td>82 and 83</td>
<td></td>
</tr>
<tr>
<td>BaLa$<em>{0.3}$Al$</em>{0.3}$In$_{0.3}$,5; A = Ca, Sr, Ba</td>
<td>$0.6-1.8 \times 10^{-6}$</td>
<td>$3.1-5.1 \times 10^{-5}$</td>
<td>$\sim 0.95$</td>
<td>$0.65-0.75$</td>
<td>$0.70-1.05$</td>
<td>$\sim 0.80$</td>
<td>73, 82 and 83</td>
<td></td>
</tr>
<tr>
<td>La$<em>2$NiO$</em>{4-\delta}$</td>
<td>$4.0 \times 10^3$</td>
<td>$3.6 \times 10^3$</td>
<td>—</td>
<td>—</td>
<td>$0.03$</td>
<td>—</td>
<td>—</td>
<td>89</td>
</tr>
<tr>
<td>Pr$<em>2$NiO$</em>{4-\delta}$</td>
<td>$1.0 \times 10^2$</td>
<td>$1.0 \times 10^2$</td>
<td>—</td>
<td>—</td>
<td>$\sim 0.03$</td>
<td>—</td>
<td>—</td>
<td>89 and 90</td>
</tr>
<tr>
<td>BaNd$<em>{0.8}$Ca$</em>{0.2}$InO$_{3.9}$</td>
<td>$3.3 \times 10^{-6}$</td>
<td>$1.3 \times 10^{-4}$</td>
<td>$0.47$</td>
<td>$0.53$</td>
<td>$0.011$</td>
<td>$0.67$</td>
<td>106</td>
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<td>BaNd$<em>{0.8}$Ca$</em>{0.2}$ScO$_{3.9}$</td>
<td>$8.3 \times 10^{-5}$ (at 600 °C)</td>
<td>$0.62$ (at 700 °C)</td>
<td>$0.316$</td>
<td>—</td>
<td>—</td>
<td>$0.23-0.37$</td>
<td>107</td>
<td></td>
</tr>
<tr>
<td>Ba$_2$In$_2$O$_5$</td>
<td>$1.3 \times 10^{-6}$</td>
<td>$3.2 \times 10^{-5}$</td>
<td>$0.84$</td>
<td>$0.35$</td>
<td>1 (1)</td>
<td>$-74$ to $-63$</td>
<td>$\sim 0.60$</td>
<td>114-117 and 124</td>
</tr>
<tr>
<td>RE$<em>{0.99}$Ca$</em>{0.01}$NbO$_{4-\delta}$; RE = La, Nd, Tb, Er</td>
<td>$2.4-6.3 \times 10^{-5}$</td>
<td>—</td>
<td>$\sim 1$</td>
<td>—</td>
<td>$-165$ to $-115$</td>
<td>$0.52-0.59$</td>
<td>138</td>
<td></td>
</tr>
<tr>
<td>RE$<em>{0.99}$Ca$</em>{0.01}$TaO$_{4-\delta}$; RE = La, Nd, Gd, Er</td>
<td>$2.8-4.7 \times 10^{-5}$</td>
<td>—</td>
<td>$\sim 1$</td>
<td>—</td>
<td>$-170$ to $-100$</td>
<td>$0.52-0.67$</td>
<td>140</td>
<td></td>
</tr>
<tr>
<td>La$<em>{0.8}$Ba$</em>{1.2}$Ga$<em>3$O$</em>{12}$</td>
<td>$7.4 \times 10^{-6}$</td>
<td>$1.4 \times 10^{-4}$</td>
<td>$\sim 0.90^*$</td>
<td>$\sim 0.80^*$</td>
<td>$\sim 0.08 \ (0.1)$</td>
<td>—</td>
<td>$0.44-0.75$</td>
<td>167, 169 and 172</td>
</tr>
<tr>
<td>La$<em>{1.9}$Ga$</em>{0.8}$ZrO$_{14}$</td>
<td>$5.0 \times 10^{-7}$ (at 600 °C)</td>
<td>—</td>
<td>—</td>
<td>$\sim 0$</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>200 and 201</td>
</tr>
<tr>
<td>La$<em>{0.95}$Ca$</em>{0.05}$ZrO$_{17-\delta}$</td>
<td>$2.4 \times 10^{-4}$</td>
<td>—</td>
<td>$\sim 1$</td>
<td>$\sim 0.03$</td>
<td>$-190$</td>
<td>0.68</td>
<td>216-219</td>
<td></td>
</tr>
<tr>
<td>Sm$<em>{1.92}$Ca$</em>{0.08}$Ti$<em>2$O$</em>{7-\delta}$</td>
<td>$1.6 \times 10^{-5}$</td>
<td>$5.0 \times 10^{-4}$</td>
<td>$\sim 0.77^*$</td>
<td>—</td>
<td>$\sim 0.04 \ (0.08)$</td>
<td>—</td>
<td>$0.52-0.59$</td>
<td>202</td>
</tr>
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</table>
Table 1  (Contd.)

<table>
<thead>
<tr>
<th>Compound</th>
<th>(\sigma_{200 , ^\circ\mathrm{C}}) (S cm (^{-1}))</th>
<th>(\sigma_{500 , ^\circ\mathrm{C}}) (S cm (^{-1}))</th>
<th>(t_{\text{on}}) 300 (^\circ\mathrm{C})</th>
<th>(t_{\text{on}}) 500 (^\circ\mathrm{C})</th>
<th>(n_{\text{H}_2\text{O}})</th>
<th>(\Delta H_{\text{hydr}}) (kJ mol (^{-1}))</th>
<th>(E^*_\text{F}) (eV)</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Sm(<em>{1.92})Ca(</em>{0.08})Sn(_2)O(<em>7)(</em>{1-\delta})</td>
<td>(2.7 \times 10^{-6})</td>
<td>(5.8 \times 10^{-2})</td>
<td>(\sim 0.98^*)</td>
<td>—</td>
<td>(\sim 0.05) (0.08)</td>
<td>(100)</td>
<td>0.83</td>
<td>203 and 204</td>
</tr>
<tr>
<td>La(_2)Ce(_2)O(_7)</td>
<td>(2.1 \times 10^{-6})</td>
<td>(2.4 \times 10^{-4})</td>
<td>(\sim 1)</td>
<td>(\sim 0.20)</td>
<td>(\sim 0.20) (1)</td>
<td>(-90) to (-77)</td>
<td>0.44–0.84</td>
<td>205–207</td>
</tr>
<tr>
<td>La(<em>{2.95})Ca(</em>{0.05})Ce(_2)O(<em>7)(</em>{1-\delta})</td>
<td>—</td>
<td>(1.0 \times 10^{-3})</td>
<td>—</td>
<td>(\sim 0.70^*)</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>222</td>
</tr>
<tr>
<td>Nd(_2)Ce(_2)O(_7)</td>
<td>(1.6 \times 10^{-7})</td>
<td>(2.7 \times 10^{-3})</td>
<td>(\sim 0.40^*)</td>
<td>—</td>
<td>(\sim 0.03)</td>
<td>(-72)</td>
<td>—</td>
<td>207</td>
</tr>
<tr>
<td>La(<em>{2.60})Sr(</em>{0.40})Nb(_2)O(_7)</td>
<td>—</td>
<td>(4.3 \times 10^{-9})</td>
<td>—</td>
<td>(\sim 0.18^*)</td>
<td>(2.5 \times 10^{-3}) (1)</td>
<td>—</td>
<td>—</td>
<td>233</td>
</tr>
<tr>
<td>La(<em>{2.97})Ca(</em>{0.03})Ta(_2)O(_7)</td>
<td>(5.4 \times 10^{-7})</td>
<td>(2.0 \times 10^{-5})</td>
<td>(\sim 0.54^*)</td>
<td>(\sim 0.40^*)</td>
<td>—</td>
<td>(-117)</td>
<td>0.77</td>
<td>231</td>
</tr>
<tr>
<td>La(<em>{2.85})Sr(</em>{0.15})Nb(_2)O(_7)</td>
<td>(1.7 \times 10^{-6})</td>
<td>(6.6 \times 10^{-5})</td>
<td>(\sim 0.50^*)</td>
<td>(\sim 0.42^*)</td>
<td>—</td>
<td>(-116)</td>
<td>0.77</td>
<td>231</td>
</tr>
<tr>
<td>La(<em>{2.60})Sr(</em>{0.40})Ta(_2)O(_7)</td>
<td>(1.2 \times 10^{-3}) (at 600 (^\circ\mathrm{C}))</td>
<td>(\sim 0.40) (at 600 (^\circ\mathrm{C}))</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>230</td>
</tr>
<tr>
<td>La(_7)TaO(_7)</td>
<td>—</td>
<td>(1.3 \times 10^{-7})</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>—</td>
<td>232</td>
</tr>
<tr>
<td>La(<em>{2.97})Ca(</em>{0.03})Ta(_2)O(_7)</td>
<td>(1.7 \times 10^{-7})</td>
<td>(5.4 \times 10^{-6})</td>
<td>(\sim 0.49^*)</td>
<td>(\sim 0.35^*)</td>
<td>—</td>
<td>(-122)</td>
<td>0.74</td>
<td>231</td>
</tr>
<tr>
<td>La(<em>{2.85})Ca(</em>{0.15})Ta(_2)O(_7)</td>
<td>(4.7 \times 10^{-7})</td>
<td>(2.4 \times 10^{-3})</td>
<td>(\sim 0.42^*)</td>
<td>(\sim 0.57^*)</td>
<td>—</td>
<td>(-133)</td>
<td>0.71</td>
<td>231</td>
</tr>
<tr>
<td>La(<em>{2.60})W(</em>{0.5})O(<em>{5.5}) (La(</em>{4})WO(_{12}))</td>
<td>(1.0 \times 10^{-4})</td>
<td>(1.0 \times 10^{-3})</td>
<td>(\sim 1)</td>
<td>(\sim 1)</td>
<td>(\sim 0.40) (0.5)</td>
<td>(-90) to (-87)</td>
<td>0.65</td>
<td>242, 244 and 247</td>
</tr>
</tbody>
</table>

provides an overview of the most relevant alternative and less established proton conducting materials, with a concise and up-to-date account of the structure–property relationships and the mechanistic aspects of hydration and ionic conduction. The aims are to highlight crucial and diverse structural features enabling proton incorporation and transport, and to invite further fundamental research into the proton conducting properties of other promising and unexplored oxide systems for the discovery of novel solid proton conductors.

2. Perovskite derivatives

The intrinsic flexibility and versatility of the perovskite structure allows the formation of a large number of derivatives via the introduction of defects and structural disorder. In addition, layered or intergrowth derivatives can be formed when the ABO\(_3\) layers are separated by different structural motifs. Some of these derivative systems are able to incorporate and enable fast diffusion of protonic defects thanks to the presence of intrinsic structural oxygen vacancies and flexible metal frameworks.

2.1 Hexagonal perovskites

Hexagonal perovskites form from mixed stacking sequences of hexagonal (h) and cubic (c) close-pack of [AO\(_3\)] layers, which result in the formation of face-sharing and corner-sharing BO\(_6\) octahedra.\(^{62}\) Mixed combinations of corner-sharing and face-sharing octahedra can give rise to a variety of hexagonal perovskite derivatives.\(^{63}\) Different hexagonal polytypic structures are usually designated by specifying the total number of layers contained in the unit cell followed by a letter denoting the lattice type (H for hexagonal, C for cubic and R for rhombohedral).\(^{64}\) These hexagonal structures are highly versatile and able to accommodate intrinsic cationic and anionic vacancies, with the formation of mixed metal coordination environments and different stacking arrangements.\(^{63,65–67}\)

High proton conductivity in a hexagonal perovskite was first discovered in the cation deficient 7H derivative Ba\(_3\)Nb\(_2\)MoO\(_{20}\) by Fop et al.\(^{68}\) In dry conditions, the conductivity of Ba\(_3\)Nb\(_2\)MoO\(_{20}\) is purely oxide ionic, with an oxide ion transport number of >0.99. Proton conduction is enabled under a humidified atmosphere, with a proton transport number of \(\sim 0.8\) at 500 \(^\circ\mathrm{C}\). The proton conductivity is \(4.0 \times 10^{-3}\) S cm \(^{-1}\) at 500 \(^\circ\mathrm{C}\), comparable to doped barium cerate and zirconate perovskites. This system also presents good stability in reducing and CO\(_2\)-rich environments and chemical compatibility with typical electrode materials such as NiO and Ba\(_{0.5}\)Sc\(_{0.5}\)Co\(_{0.8}\)Fe\(_{0.2}\)O\(_{3}\) (BSCF).\(^{68}\)

In dry Ba\(_3\)Nb\(_2\)MoO\(_{20}\) the cation vacancies are ordered, and the average structure is composed by an intergrowth of 12R hexagonal perovskite units and oxygen deficient ([BaO\(_2\)])\(_{1-\delta}\) palmierite-like layers formed by MoO\(_x\) polyhedra with mixed local 4-, 5-, and 6-fold coordination due to partial occupation of two different average crystallographic tetrahedral and octahedral oxygen sites (O\(_{\text{tet}}\) and O\(_{\text{oc}}\)) (Fig. 3a).\(^{68}\) Water is absorbed on the intrinsic oxygen vacancies present on the
palmierite-like layer and predominantly distributed on the average O\text{oct} sites.\textsuperscript{68,69} The water absorption leads to disordering of the metal-stacking vacancy distribution due to a shift of the cation in the palmierite layer towards an adjacent mutually exclusive vacant site, as result of the repulsion between a proton pointing towards the nearest MO\text{6} unit and a metal cation (Fig. 3b). Partial occupation of the two metal sites results in the formation of disordered local stacking configurations creating a complex average network of isolated, face-sharing and corner-sharing polyhedral units in the hydrated material (Fig. 3c).

The significant structural flexibility of Ba\textsubscript{7}Nb\textsubscript{4}MoO\textsubscript{20} allows the accommodation of the local cation and anion disorder introduced by hydration and enables a high concentration of water uptake, $\geq0.8$ molecules of H\textsubscript{2}O per formula unit.\textsuperscript{68,69}

Neutron diffraction experiments and atomistic calculations demonstrate that protons are in proximity of the average O\text{oct} sites.\textsuperscript{68} The positional oxide ion disorder generated by the close proximity of available oxygen sites due to the particular topology of the palmierite-like layers leads to delocalisation of the protonic defects over a variety of low energy configurations around their equilibrium site. Such delocalisation results in the creation of two-dimensional low energy transport pathways along the palmierite-like layers, where the proton defects can migrate via rotation and hopping motion with calculated migration energies in the range 0.18–0.61 eV depending on the local configuration (Fig. 3d). Importantly, the proton diffusion is assisted by the high flexibility and rotational mobility of the isolated variable coordination MO\text{6} units. Proton diffusion is usually slower in lower symmetry perovskite oxides such as orthorhombic SrCeO\textsubscript{3} and CaZrO\textsubscript{3}, due to an additional enthalpy contribution needed for tilting of adjacent octahedra closer enough to allow proton hopping.\textsuperscript{67,68}
protonic defects also generally leads to reduced proton conductivity, as in the case of the hexagonal 6H phase of Sc-doped BaTiO$_3$. In hydrated Ba$_2$Nb$_5$MoO$_{20}$, frustration of the proton sub-lattice and the high dynamic flexibility of the variable coordination MO$_x$ moieties enable fast proton transport.

Another member of the hexagonal perovskite family displaying high proton conductivity is Ba$_5$Er$_2$Al$_2$ZrO$_{13}$. This system is a 10H hexagonal perovskite derivative with a structure composed by the stacking of cubic [BaO$_3$] and oxygen deficient hexagonal [BaO] layers. The intrinsic oxygen deficient layers result in the creation of tetrahedral sites which are occupied by the aluminium atoms to form Al$_2$O$_7$ units [Fig. 4a].

While in dry air the conductivity is electronic p-type, Ba$_5$Er$_2$Al$_2$ZrO$_{13}$ presents a large protonic component in humidified air, showing a proton transport number of $\sim$1 from 300 to 600 °C and conductivity of $3.0 \times 10^{-3}$ S cm$^{-1}$ at 500 °C.\(^{79}\) Water is absorbed on the intrinsic oxygen vacancies of the hexagonal [BaO] layer, with the proton sites located on the apical oxygen atoms of the Al$_2$O$_7$ units [Fig. 4b].\(^{79}\) Both Ba$_2$Nb$_{5}$MoO$_{20}$ and Ba$_5$Er$_2$Al$_2$ZrO$_{13}$ share the existence of oxygen deficient Ba–O layers with intrinsic oxygen vacancies enabling water absorption and fast proton transport.

Compounds of the series Ba$_n$M$_2$O$_9$ (M = Nb, Sb, Ta) are hexagonal derivatives exhibiting mixed oxide ion, proton and electronic conduction.\(^{73,75}\) While the ionic conductivities are far below those of representative proton conductors (<10$^{-7}$ S cm$^{-1}$ at 400 °C), these systems present complex polymorphic behaviour and interesting hydration and conduction mechanisms. The Nb and Ta compounds have the same low temperature x polymorph, which is composed by partially disordered isolated face-sharing octahedral (Nb/Ta)O$_6$ dimers separated by large voids. The high temperature phases of Ba$_2$Ta$_2$O$_9$ and Ba$_5$Sb$_2$O$_9$ both exhibit a 6H structure formed by face sharing octahedral units connected to BaO$_6$ octahedra through corner-sharing [Fig. 5a].\(^{74,78}\) On the other hand, the niobate presents a high temperature orthorhombic $\gamma$-phase composed by layers of isolated face-sharing NbO$_6$ dimers separated by ordered rows of isolated NbO$_6$ and NbO$_5$ units [Fig. 5b].\(^{77}\)

The high temperature 6H–Ba$_2$Ta$_2$O$_9$ and $\gamma$-Ba$_5$Nb$_2$O$_9$ phases can be quenched to room temperature and are able to incorporate a substantial amount of water. In 6H–Ba$_2$Ta$_2$O$_9$, discrete water molecules are intercalated in the void between the TaO$_6$ dimers. Ab initio molecular dynamics (AIMD) simulations show that the H$_2$O molecule unusually does not dissociate into H$^+$ and OH$^-$, migrating intact through the voids [Fig. 5c].\(^{78}\) In the $\gamma$-Ba$_5$Nb$_2$O$_9$ phase, water is absorbed via hydroxylation of the NbO$_6$ tetrahedra and hydrogenation of the NbO$_5$ trigonal bipyramids to give two NbO$_5$OH$_4$ units. AIMD simulations evidence that proton conduction mainly occurs within the ab plane of the NbO$_5$ polyhedra by hopping between oxygen atoms of adjacent NbO$_5$OH$_4$ units [Fig. 5d].\(^{78}\)

Other hexagonal perovskite derivatives which exhibit proton conductivity are cation deficient 12R Sr$_n$RE$_m$Nb$_3$O$_{12}$ (RE = La, Nd) and 5H Ba$_n$Nb$_m$O$_{15}$.\(^{79-81}\) The structures of both compounds are composed by trimers (12R) or tetramers (5H) of corner-sharing NbO$_6$ octahedra spaced by cationic vacancies. These systems present mixed ionic and electronic conduction, with proton transport dominating under wet oxidising conditions. The ionic conductivities are modest ($\sim$10$^{-8}$ S cm$^{-1}$ at 550 °C), but their values can be substantially increased by acceptor doping on the Nb-site.\(^{79,80}\)

These accounts highlight the inherent structural versatility of hexagonal perovskite derivatives of accommodating cation and anion defects/disorder and their ability in incorporating and enabling transport of protonic defects through different mechanisms. Even though reports of proton conduction in these systems are quite scarce, the most recent results on Ba$_7$-Nb$_4$MoO$_{20}$ and Ba$_5$Er$_2$Al$_2$ZrO$_{13}$ are particularly promising and encourage further exploration of proton conduction in the vast family of hexagonal perovskite derivatives.

2.2 Ruddlesden–Popper structures

Layered perovskite derivatives form when perovskite-like slabs are spaced by some different structural motif. Ruddlesden–Popper (RP) phases are a particular class of layered perovskites with general formula $A_{n+1}B_nO_{3n+1}$. The structures of RP phases are composed by $n$ perovskite layers alternated with rock-salt slabs, thus forming a framework in which $n$ layers of corner-sharing BO$_6$ octahedra are separated by AO layers [Fig. 6a]. BaLaInO$_4$-based oxides have K$_3$NiF$_3$-type structures ($n = 1$) and exhibit mixed oxide ion and proton conduction.\(^{82-85}\) These
systems have generally large protonic components with proton transport numbers \( \mu > 0.80 \) under wet conditions at temperatures below \( 450 \, ^\circ\text{C} \), although their conductivities are rather low \( 10^{-6} \) to \( 10^{-5} \, \text{S cm}^{-1} \). Several ferrite and nickelate RP oxides with general compositions \( A_xB_2O_4 \) \( (n = 1) \) or \( A_xB_2O_7 \) \( (n = 2) \) \( (A = \text{Sr, La, Nd, Pr; B = Fe, Ni}) \) show triple mixed conductivity \( (i.e. \) protonic, oxide ion and electronic). These systems present good chemical compatibility with barium cerate and zirconate \( \text{electron charge carriers} \) \( \text{proton charge carriers} \) \( \), oxide ion, and electronic). These systems present good chemical compatibility with barium cerate and zirconate.

Fig. 7 (a) Structure of BaNdScO4. (b) Proton locations and representation of the intra-octahedral (1) and inter-octahedral (2) proton diffusion pathways. Ba atoms are omitted for clarity. (c) Schematic of the intra- and inter-octahedral O–O distances in cubic ABO3 perovskites and BaNdScO4. In the cubic ABO3 perovskite structure the inter-octahedral distance is approximately \( \sqrt{2} \) times the O–O intra-octahedral separation. In contrast, the intra- and inter-octahedral distances are similar in BaNdScO4.

Fig. 8 (a) Brownmillerite \( A_xB_2O_5 \) structure composed by alternating layers of \( \text{BO}_6 \) octahedra \( (\text{BO}_6) \) and \( \text{BO}_4 \) tetrahedra \( (\text{BO}_4) \). (b) Representation of the transition from dry orthorhombic \( \text{BaIn}_2\text{O}_5 \) to the fully hydrated tetragonal phase \( \text{BaIn}_2\text{O}_5 \). Water is absorbed onto the intrinsic oxygen vacancies \( \text{O}_{\text{in}} \) distributed along the tetrahedral layer. Protons in the tetragonal \( \text{BaIn}_2\text{O}_5 \) structure occupy two distinct average positions: \( \text{H}^+ \), which is partially occupied, and \( \text{H}^+ \). (c) Local protons configuration.

pathway has a very large energy barrier \( (\sim 3.5 \, \text{eV}) \) and probably does not contribute to the proton diffusion. The inner-layer pathway presents considerably lower energy barrier \( (1.29–1.66 \, \text{eV}) \) and involves a Grotthuss mechanism with rotation and hopping of the protons between the apical and equatorial oxygen positions of the metal octahedra \( (\text{similar to the case of cubic perovskites}) \).

Compounds of the series \( \text{BaNdMO}_4 \) \( (M = \text{In, Sc}) \) are RP derivatives.104,105 Their structures are composed by \( \text{BaNd}_{1/2} \text{Nd}_{1/2} \text{In}_{2/3-x} \text{MO}_{4-x/3} \) perovskite layers alternated with \( \text{NdO} \) rock-salt blocks. They differentiate from the common RP structure because the edge of the \( \text{MO}_6 \) octahedron faces the rock-salt unit \( (\text{Fig. 7a}) \). These systems are mixed oxide ionic and electronic conductors and recent reports have revealed significant proton conductivity in acceptor-doped materials of composition \( \text{BaNd}_{1/2} \text{Nd}_{1/2} \text{In}_{2/3-x} \text{MO}_{4-x/3} \) \( x \).106,107 The doped indate exhibits higher conductivity than the \( \text{Sc} \) compound. At 500 \( ^\circ\text{C} \), \( \text{BaNd}_{1/2} \text{Nd}_{1/2} \text{In}_{2/3-y} \text{Sc}_{y} \) exhibits conductivity of \( 1.3 \times 10^{-4} \, \text{S cm}^{-1} \) under humidified air and a proton transport number of \( \sim 0.53 \), although this system presents poor chemical stability in a humid atmosphere.106

Water is absorbed onto the oxygen deficient layer at the interface between the perovskite and the rock-salt layers, with the protons located on the equatorial oxygen atoms of the \( \text{MO}_6 \) octahedra.107 Protons diffuse via rotational and hopping
motions along the lowest energy intra- and inter-octahedral pathways, which have similar migration barriers of 0.63 eV and 0.69 eV respectively (Fig. 7b). Inter-octahedral proton hopping is generally unfavoured in cubic perovskite structures due to the large inter-octahedral O–O distance (Fig. 7c). Inter-octahedral diffusion predominantly occurs in lower symmetry perovskite-type structures, where the hopping distance is shortened by octahedral tilting.8,78 However, in the BaNdMO4 structure the inter-octahedral O–O separation is comparable to the intra-octahedral O–O distance, thus enabling both pathways (Fig. 7c).107

The results on RP materials suggest that similar layered perovskite structures could potentially support analogous mechanisms of proton incorporation and transport. Dion-Jacobson phases with general formula MA_{n–1}B_nO_{3n+1} (M = alkali metal) are closely related to the RP structure, having a layer of alkali metal as the separating motif between the perovskite blocks.108 Aurivillius phases are constituted by perovskite layers spaced by rock-salt BiO sheets in which Bi is in a square-based pyramidal coordination.109 Both these structure systems support oxide ion conduction,106,109 with signatures of proton conductivity in some Dion–Jacobson materials,110,111 and could therefore constitute an interesting avenue of research into novel layered perovskite proton conductors.

2.3 Brownmillerite materials

Another class of layered perovskite derivatives is the brownmillerite family. The brownmillerite structure, with general formula AB2O5, can be described as an oxygen deficient perovskite structure, having 1/6 of oxygen sites empty. The intrinsic oxygen vacancies are fully ordered in rows along the [110]_p direction (where the subscript p denotes the perovskite subcell) and form corner-sharing octahedral perovskite layers alternated with tetrahedral layers (Fig. 8a). The most studied brownmillerite material is the oxide ion conductor Ba2In2O5.112 At high temperature (T > 900 °C), disordering of the oxygen vacancies leads to a transition to a highly conductive cubic phase.114,115

Under humidified atmosphere, Ba2In2O5 exhibits significant proton conductivity, with a proton transport number >0.70 below 400 °C.114,115 The large number of intrinsic oxygen vacancies enables the water absorption. Ba2In2O5 can uptake a considerable concentration of water, forming the fully hydrated phase Ba2In2O5(H2O), or BaInO3H.116–118 Upon hydration, the dry brownmillerite structure transforms from orthorhombic to a tetragonal perovskite-like phase (at T < 300 °C), where the intrinsic oxygen vacancies in the tetrahedral layer are fully occupied by the water oxygen atoms (Fig. 8b).118 Protons are located on two different sites; H1 is on a partially occupied position bonded to an apical octahedral oxygen, while H2 is on a fully occupied site lying between two equatorial oxygen atoms corresponding with the tetrahedral layer in the dehydrated structure.118–121 Protons are ordered according to a defined local configuration, in which H1 is hydrogen bonded towards the equatorial oxygen of the octahedral layer of the dry structure (Fig. 8c).121 The hydrogen bond configuration affects the mobility of the protonic defects and restricts the number of sites available for hopping, thus resulting in anisotropic proton conduction.122,124

The conductivity of Ba2In2O5 under humidified conditions is ~10–5 S cm–1 at 400 °C.114 Doping on the In-site generally leads to stabilisation of the perovskite-like structure and to increased proton conductivity.118,125 Interestingly, the cubic phase can be stabilised also by oxyanion doping via the introduction of silicate, phosphate or sulfate groups, with the conductivity increasing to 2.4 × 10–4 S cm–1 at 400 °C in Ba2In1.8Si0.2–O5.1.126,127 Ba2In2O5 decomposes in hydrogen-containing atmospheres above 500 °C, with the formation of In metal and BaCO3 and a considerable reduction of the ionic conductivity over time.115 Due to the presence of electronic conductivity and the generally poor stability under reducing conditions, fuel cells with Ba2In2O5-based electrolytes exhibit generally low power densities (<10 mW cm–2 at 600 °C).128

Other brownmillerite compounds exhibiting proton conductivity are BaInGaO5 and Sr2ScGaO5.129,130 The ionic conductivity of BaInGaO5 is quite modest due to ordering of the In and Ga cations along the stacking axis, which further stabilises the orthorhombic brownmillerite structure.129 Sr2–ScGaO5 also exhibits a moderate protonic component.130 However, heavily Zn-doped Sr2ScGaO5 (Sr2Sc1–xZn0.5Ga0.5O5, 0.4 ≤ x ≤ 0.6) presents a highly disordered and highly oxygen deficient perovskite structure with a peculiar one-coordinate oxygen site and shows significant proton conductivity, with three orders of magnitude increase in the conductivity in comparison with the parent compound.130

Fig. 9 (a) Tetragonal LaNbO4 scheelite structure composed by isolated NbO4 tetrahedra. (b) Schematic of the oxygen vacancy stabilisation. Two NbO4 tetrahedral units relax and condensate to form a Nb2O7 unit in the neighbourhood of the vacant oxygen site (O vacancies). (c) Protons positions in scheelite LaNbO4. (d) Representation of the long-range H1–H2 proton migration showing rotation (1) and hopping onto an adjacent tetrahedron (2), assisted by motion of the isolated tetrahedral units (3).
A recent study has demonstrated that the brownmillerite Sr$_2$Co$_3$O$_5$ can absorb one H per formula unit, forming hydroxynitrate $\text{H}Sr\text{CoO}_2\text{Sr}$ [although this phase is only stable up to 160 °C].$^{131,132}$ Atomicistic calculations show that protons are distributed among apical and equatorial octahedral oxygen sites and that they adopt a wide range of configurations with similar energies.$^{133}$ The cooperative transition between the different configurations creates a frustrated proton sub-lattice enabling high proton mobility. Measurements on a thin film of $\text{H}Sr\text{CoO}_2\text{Sr}$ showed conductivity of 0.33 S cm$^{-1}$ at 140 °C.$^{132}$ However, further measurements on a bulk dense sample are required to confirm such high conductivity and rule out any other effects from proton transport along physiosorbed water at the inner surface of the porous film.$^{134}$ Brownmillerite-based materials with intrinsic oxygen vacancies offer potential of high proton conductivity, thanks to the ability of the structure of incorporating large concentrations of water. Other brownmillerite compounds showing oxygen intercalation ability may be able to incorporate water and enable proton transport,$^{135-137}$ thus constituting further candidate proton conductors.

3. Oxides with tetrahedral moieties

Proton conduction has been reported in various systems possessing tetrahedral moieties as their principal structural feature. These tetrahedral units are present in isolated or corner-sharing motifs and generally assist the protonic transport thanks to their high flexibility and rotational mobility.

3.1 REMO$_4$ phases

Haugsrud and Norby were the first to demonstrate high proton conductivity in REMO$_4$-based materials, namely in several acceptor doped ortho-niobates and ortho-tantalates with general formula $\text{RE}_1-x\text{A}_x\text{MO}_4$ ($\text{RE} = \text{La, Gd, Nd, Tb, Er, Y}; \text{A} = \text{Ca, Sr, Ba}; \text{M} = \text{Nb, Ta}$).$^{138}$ Compared to acceptor doped perovskite conductors, proton conductivity in these ortho-niobates and ortho-tantalates is attained for relatively low doping levels ($x = 0.01-0.05$). Proton conduction is dominant (proton transport number $\approx 1.0$) under humidified and reducing atmospheres below 800 °C, while the conductivity is p-type electronic under oxidizing conditions at high temperatures.$^{138-140}$

These rare earth ortho-niobates and ortho-tantalates are isostructural, crystallising in the monoclinic fergusonite-type structure at low temperatures and in the tetragonal scheelite phase at high temperatures (Fig. 9a). Both structure types are composed by isolated tetrahedral units sharing corners with $\text{REO}_3$ dodecahedral units; the low temperature fergusonite structure can be considered as a slight monoclinic distortion of the tetragonal high temperature phase.$^{141}$ The transition temperature depends on the composition and for LaNbO$_4$ is in the range 300–530 °C.$^{139,141}$ The fergusonite–scheelite transition in LaNbO$_4$ has originally been described as a second order transition between the ferroelastic monoclinic phase and the paraelastic tetragonal phase.$^{141}$ However, recent reports suggest a reconstructive first order transition induced by displacement of the Nb cation from the centre of the tetrahedron and change in NbO coordination from 4 in the scheelite phase to a 6-coordinated distorted octahedral arrangement in the fergusonite phase, with two long and four short Nb–O distances.$^{142,143}$ The fergusonite–scheelite transition is relevant to the proton conductivity, since it coincides with a reduction in the activation energy for proton transport (from 0.78 eV to 0.55 eV in La$_{0.99}\text{Ca}_{0.01}\text{NbO}_4$), which is associated to an increase in the mobility of the proton defects in the scheelite phase.$^{138,142}$ However, the presence of this transition may be detrimental for practical electrolyte applications due to the different thermal expansion coefficients of the two phases (around $15 \times 10^{-6}$ °C$^{-1}$ for the low temperature monoclinic phase and $8.6 \times 10^{-6}$ °C$^{-1}$ for the high temperature polymorph) and the tendency to microcracking.$^{146}$ The scheelite phase can be stabilised at room temperature via substitution of Nb with isovalent cations such as V, As, and Sb, generally resulting in an increase of the conductivity at lower temperatures.$^{147-150}$

Acceptor doping of LaNbO$_4$ introduces extrinsic oxygen vacancies, enabling water absorption. Atomicistic calculations on the scheelite LaNbO$_4$ structure have demonstrated that oxygen vacancies are stabilized by local relaxation and condensation of the NbO$_4$ tetrahedra around a vacant oxygen site to form Nb$_2$O$_4$ or Nb$_2$O$_5$ polyhedral linkages (Fig. 9b).$^{151,152}$

The thermodynamics of hydration and the proton mobility depend on the size of the rare-earth element. Hydration is more exothermic (i.e. more favourable) as the rare-earth cation becomes smaller, with the enthalpy of the water absorption reaction reducing from $-115 \text{kJ mol}^{-1}$ in the La niobate, to $-165 \text{kJ mol}^{-1}$ for the Er-containing composition.$^{138}$ This trend is in contrast with REPO$_4$ phosphates, of which crystal structures are composed by similar isolated tetrahedral units.$^{153}$ On the other hand, the proton mobility is lower when the RE cation is smaller due to lattice size effects and to the decreasing lattice polarizability of the smaller unit cell.$^{138}$ The LaNbO$_4$-based materials exhibit the highest conductivity, with a value of $\approx 10^{-3}$ S cm$^{-1}$ at 800 °C in La$_{0.99}\text{Ca}_{0.01}\text{NbO}_4$.$^{138,139}$

Structural optimization calculations on scheelite LaNbO$_4$ have demonstrated that protons are located on two stable positions in proximity of the only crystallographic oxygen site: along the edge of the NbO$_4$ tetrahedra (H1, lowest energy) and on a site between two lanthanum ions (H2) (Fig. 9c).$^{152,154}$ Long-range proton migration occurs along the pathway connecting the H1 and H2 sites, through rapid rotation of the proton around the oxide ion (with an energy barrier of 0.04–0.06 eV) followed by hopping between two adjacent tetrahedra (with a barrier of 0.41 eV) (Fig. 9d).$^{154}$

Despite exhibiting lower conductivities than perovskite cerate oxides, doped LaNbO$_4$-based materials shows greater stability in CO$_2$-containing atmospheres,$^{155}$ which together with a proton transport number close to unity make these materials highly interesting for fuel cell applications. Tests on different proton-conducting electrode materials have demonstrated generally good chemical compatibility of LaNbO$_4$ materials with LaMO$_4$ ($\text{M} = \text{Mn, Fe, Co}$) perovskite type cathodes and NiO,$^{154}$ while they show poor compatibility with proton-
conducting ceramic anodes because of undesirable reactions at high temperatures.\textsuperscript{155} Anode supported fuel cell with \( \text{La}_{0.99} \text{Ca}_{0.01} \text{NbO}_4 \) electrolytes have shown modest peak power densities (24 mW cm\(^{-2}\) at 750 °C) due to the large polarization resistance between the electrodes and the electrolyte (20–40 Ω cm\(^{-2}\)) at 800 °C.\textsuperscript{155,157,158}

Acceptor doped monazite-type systems \( \text{LaVO}_4 \), \( \text{LaAsO}_4 \), and \( \text{LaSbO}_4 \) also present significant proton conductivity under humidified atmosphere.\textsuperscript{159–161} The monoclinic monazite structure is similar to the scheelite structure and is composed of isolated distorted \( \text{MO}_4 \) tetrahedra sharing their edges with \( \text{LaO}_9 \) polyhedra.\textsuperscript{162} Doped \( \text{LaVO}_4 \) and \( \text{LaAsO}_4 \) exhibit the highest conductivities; the proton conductivity of \( \text{La}_{0.99} \text{Ca}_{0.01} \text{VO}_4 \) is \( \sim 3 \times 10^{-4} \) S cm\(^{-1}\) at 800 °C, although the ortho-vanadate is unstable under reducing conditions.\textsuperscript{159}

Atomistic calculations on \( \text{LaAsO}_4 \) have revealed that extrinsic oxygen vacancies are introduced via condensation of two adjacent tetrahedra to form a pyro-arsenate \( \text{As}_2\text{O}_7 \) unit, similar to the case of \( \text{LaNbO}_4 \).\textsuperscript{160} Protons are statistically distributed on two lowest energy positions located on two non-equivalent oxygen sites within the same tetrahedron. Interestingly, while incorporated protons in oxides generally reside around oxide ions forming OH groups, in \( \text{LaAsO}_4 \) the protonic defects are stabilised as covalent hydrogen arsenate ions, \( \text{HAsO}_4 \).\textsuperscript{2–159} During the inter-tetrahedral proton migration, the host lattice mainly exhibits rotational and translational relaxation with little distortion of the \( \text{AsO}_4 \) polyhedra, a feature that is common in proton conducting oxides with isolated tetrahedral moieties and in contrast with \( \text{ABO}_3 \) perovskites where proton transport is generally accompanied by large distortion of the corner-sharing \( \text{BO}_6 \) octahedra.\textsuperscript{161}

A variety of compositions crystallize with the scheelite, monazite or in closely related structures\textsuperscript{162,164–166} and could offer an interesting crystal chemistry space for the investigation of proton conduction in systems with isolated tetrahedral moieties.

### 3.2 Gallium-based oxides

High proton conductivity has been reported in lanthanum-barium gallates of general formula \( \text{La}_{1-x} \text{Ba}_x \text{GaO}_4 \cdot x/2 \).\textsuperscript{167,168} The parent compound \( \text{LaBaGaO}_4 \) possesses an orthorhombic \( \beta\)-\( \text{K}_2\text{SO}_4 \)-type structure composed by isolated distorted \( \text{GaO}_4 \) tetrahedra and ordered alternating layers of \( \text{Ba} \) and \( \text{La} \) atoms (Fig. 10a). \( \text{LaBaGaO}_4 \) shows high resistivity with limited proton conduction. Variation of the \( \text{La}/\text{Ba} \) ratio to produce the solid solution \( \text{La}_{1-x} \text{Ba}_{1+x} \text{GaO}_4 \cdot x/2 \) results in the introduction of oxygen vacancies enabling high ionic conduction. The member with \( x = 0.2 \), \( \text{La}_{0.8} \text{Ba}_{1.2} \text{GaO}_3 \), shows proton conductivity of \( \sim 10^{-4} \) S cm\(^{-1}\) at 500 °C under humidified conditions.\textsuperscript{167,168}

The oxygen vacancies in \( \text{La}_{0.8} \text{Ba}_{1.2} \text{GaO}_3 \) are accommodated through the condensation of two adjacent \( \text{GaO}_4 \) units to form a \( \text{Ga}_2\text{O}_7 \) group (in a way analogous to acceptor doped \( \text{LaNbO}_4 \)).\textsuperscript{169} On the local scale, the oxygen vacancies are pinned to the nearest-neighbour cation site by the \( \text{Ba} \) defects and there is no ordering of the \( \text{Ga}_2\text{O}_7 \) units.\textsuperscript{170} Water incorporation induces the break-up of the \( \text{Ga}_2\text{O}_7 \) units,\textsuperscript{169} according to the reaction

\[
\text{Ga}_2\text{O}_7 + \text{H}_2\text{O} \rightarrow 2\text{GaO}_2\text{H}.
\]

Protons in \( \text{La}_{0.8} \text{Ba}_{1.2} \text{GaO}_3 \) are located on three different sites, pointing almost perpendicular to the Ga–O bond direction (Fig. 10b).\textsuperscript{171} Protons migrate following a curved path between
two intra-tetrahedral oxide ions due to the repulsive interactions with nearby large Ba/La cations, followed by jump onto an oxygen of an adjacent GaO$_4$ unit (Fig. 10c).

The rate-determining step for proton diffusion is the intra-tetrahedral migration, which has a considerably higher energy barrier (≈0.4–0.7 eV) than the inter-tetrahedral hopping (≈0.07 eV). The latter is favoured by the inter-tetrahedral hydrogen bond O⋯H interactions, which are shorter (≈2.0 Å) than the intra-tetrahedral ones (≈2.3 Å), thus highlighting the significance of the H bonding configuration (Fig. 10b) in defining the mechanism of proton transport in La$_{0.8}$Ba$_{1.2}$GaO$_{3.9}$. Importantly, the inter-tetrahedral proton migration is also facilitated by the rotational mobility of the GaO$_4$ units, which is similar to the rapid phosphate group reorientation in the solid-acid protonic conductor CsH$_2$PO$_4$. It is worth noticing that while intra-polyhedral hoppings are common in perovskite systems with corner-sharing octahedral units, intra-tetrahedral proton hopping is particularly unfavoured in La$_{0.8}$Ba$_{1.2}$GaO$_{3.9}$ and in other oxide structures with isolated tetrahedral units such as LaNbO$_4$ and LaAsO$_4$.

Proton conductivity has been reported in gallium-based oxides having general composition RE$_2$Ga$_3$MO$_{14}$ (RE = La, Nd; M = Si, Ti, Sn, Ta). These compounds belong to the langasite family and their structure is composed by distorted isolated GaO$_4$ tetrahedra, with water absorption leading to the introduction of distorted SiO$_5$ units. In the apatite structure, the interstitial sites are close to the centre of the La channel, and hydration leads to the formation of distorted SiO$_5$ units. The Ge-apatite shows similar structural features, and hydration leads to the formation of distorted SiO$_5$ units. In the Ge-apatite, the interstitial site is in proximity of the GeO$_4$ tetrahedra, with the water absorption resulting in the creation of Ge$_2$O$_6$ units. Protons are located on the oxide ion at the centre of the La ring. (c) Schematic of the possible proton transfer mechanism composed by rotation (1) and hopping (2) within the channel parallel to the c-axis.

![Fig. 12](image_url)  
(a) Crystal structure of apatite $A_{10-x}B_xO_{26-2x}$, where A is a rare-earth metal such as La or an alkaline earth metal such as Sr or Ba, and B is a p-block element such as P, Si or Ge. The apatite structure possesses hexagonal symmetry and is formed by a framework of isolated BO$_4$ tetrahedra constituting two channels running parallel to the c-axis; the inner channel contains rows of A cations in a ring formation with a row of oxide ions at the centre (Fig. 12a).

Si- and Ge-based lanthanum apatites are known oxide ion conductors thanks to the presence of interstitial oxygen defects which mediate the oxide ion transport. These systems are able to absorb significant concentrations of water for example, $\sim$0.75H$_2$O molecules per formula unit in La$_{0.8}$SiO$_{26.4}$ and $\sim$0.55H$_2$O molecules in La$_{0.67}$GeO$_{26.5}$. Similarly to the case of Ruddlesden–Popper oxides, water is incorporated on the interstitial oxygen sites.

The interstitial sites are close to the centre of the La channel in Si-apatites, forming distorted SiO$_5$ units, while they are in proximity of the GeO$_4$ tetrahedra forming Ge$_2$O$_6$ units in Ge-apatites (Fig. 12b). Water absorption leads to the introduction of protonic defects which are located on the oxide ions at the centre of the La ring, in an arrangement which agrees
transports, thus motivating further investigation of oxides crystallising in this structural family.

4. Fluorite-related oxides

Fluorite-based oxides are usually known for their oxide ion conducting properties. Fluorite materials such as yttria stabilised zirconia (YSZ) and doped ceria constitute the most conventional oxide ion conductors thanks to ability of the structure of supporting oxide ion conduction. Some oxide materials having fluorite or fluorite-related superstructures however exhibit significant proton conductivity.

4.1 A<sub>B</sub>2O<sub>7</sub> compounds

Rare-earth oxides with general formula A<sub>B</sub>2O<sub>7</sub> (A = La, Nd, Sm, Er; B = Ti, Zr, Sn, Ce) exhibit proton conductivity. These compounds are typically classified as having fully ordered pyrochlore structures or oxygen deficient disordered fluoride-type structures. The pyrochlore structure (space group Fd3m) is composed by chains of distorted BO<sub>6</sub> octahedra, with the A cations in a distorted 8-fold cubic coordination, while the fluorite structure (space group Fm3m) can be described by a face-centred cubic arrangement of A cations in a cubic 8-fold coordination (Fig. 13a). The pyrochlore structure is considered as a superstructure of the ideal cubic fluorite, having twice the unit cell parameter and 1/8 of the oxygen sites unoccupied.

The formation of the A<sub>B</sub>2O<sub>7</sub> pyrochlore phase is driven by the tendency for ordering of the cations on the A and B sites. This tendency depends on the relative sizes of the cations: as the difference in cationic sizes increases (i.e. the radius of the A-site cation becomes larger), stabilisation of the ordered pyrochlore unit cell is favoured. The ratio between the ionic radii of the A and B cations (r<sub>A</sub>/r<sub>B</sub>) can be used as a tolerance factor for describing the formation of one structure over the other. For example, La<sub>2</sub>Zr<sub>2</sub>O<sub>7</sub> adopts the pyrochlore structure, while La<sub>2</sub>Ce<sub>2</sub>O<sub>7</sub> is most often reported as a disordered fluorite. The oxygen vacancies can exhibit short-range ordering or disordering, thus leading to the formation of either pyrochlore-like (in fluorite systems) or fluorite-like (in pyrochlore structures) domains on the local scale.

Acceptor doping on the A site with alkaline earth cations promotes water incorporation through the introduction of oxygen vacancies. The hydration is almost independent of the nature of the B site cation but correlates with the size of the rare-earth element, becoming more favourable (enthalpy of hydration more exothermic) on increasing the ionic radius of the rare-earth element, becoming more favourable (enthalpy of hydration more exothermic).

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Fluorite-related superstructures or oxygen deficient disordered fluoride-type structures. The pyrochlore structure (space group Fd3m) is composed by chains of distorted BO<sub>6</sub> octahedra, with the A cations in a distorted 8-fold cubic coordination, while the fluorite structure (space group Fm3m) can be described by a face-centred cubic arrangement of A cations in a cubic 8-fold coordination (Fig. 13a). The pyrochlore structure is considered as a superstructure of the ideal cubic fluorite, having twice the unit cell parameter and 1/8 of the oxygen sites unoccupied.

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conductivity, with a significant protonic component under wet oxidising conditions. The compounds containing the larger lanthanides (RE = La, Gd) generally crystallise with a weberite structure, while the compositions with the smaller lanthanide cations have fluoride, distorted fluoride or pyrochlore structures. The weberite structure is an anion deficient fluoride-related superstructure exhibiting different polytype forms. The La₃MO₇ weberite structure (space group Cmcm or Pmma) can be described as composed by chains of corner-sharing MO₆ octahedra perpendicular to the b-axis and edge-sharing distorted LaO₈ cubes (Fig. 14a). Similarly to the case of A₂B₂O₇ oxides, the Ln₃MO₇ compositions having the fluoride structure exhibit shot-range ordering of the anion vacancies and contain pyrochlore microdomains on the local scale.

In the niobates, the enthalpy of hydration regularly decreases with the radius of the rare-earth cation. The water absorption is favoured for the compounds containing smaller RE ions: the water concentration is 2–3% water per formula unit for the compounds with RE = Y, Yb, Lu, while it is considerably lower for LaNbO₇. Extrinsic oxygen vacancies can be introduced in the weberite lanthanum niobate/tantalate structures via acceptor doping. The doped La₃₋xAₓMO₇₊x/2 (A = Ca, Sr) compositions are stable under oxidising and reducing conditions and exhibit mixed ionic–electronic conductivity with a wide electrolytic domain (∼10⁻¹⁰ to 10⁻³ atm in pO₂) and a proton transport number of 0.4 at 600 °C. The enthalpies of hydration are in the range −133 kJ mol⁻¹ to −117 kJ mol⁻¹ and their conductivities are ∼10⁻⁴ to 10⁻³ S cm⁻¹ at 600 °C.

Calculations on weberite La₃NbO₇ showed that the most stable proton sites are on the non-shared equatorial oxygen atom of the NbO₆ octahedra (Fig. 14), in contrast with La₃Zr₂O₇ where protons are located on the shared apical oxygen. Protons migrate via rotational and hopping motion through a long-range lowest energy percolation pathway along the non-shared equatorial oxygen atoms and parallel to the a-axis (Fig. 14b). The rotational paths around the non-shared equatorial oxygen atoms have energy barrier of 0.34–0.43 eV, while the intra- and inter-octahedral hoppings have barriers of 0.41–0.54 eV.

The anisotropicity of migration is related to the configuration of the octahedral chains and the preference of protons of residing on the non-shared equatorial oxide ions. The rate-determining step for diffusion is the inter-octahedral hopping between two adjacent non-shared oxygen atoms, which presents the highest energy barrier.

### 4.3 Lanthanum tungstate oxides

Several lanthanide tungstate oxides with a ~3 : 1 Ln₃O₂–WO₃ molar ratio (Ln = La, Nd, Gd and Er) have been reported to show predominant proton conduction under humidified atmosphere below 900 °C. Lanthanum tungstate oxides in the compositional region 25–30 mol% La₃O₉ have general formula La₂₈–xWₓO₅₄+3x/2Y₂–x/2z (v denotes the number of oxygen vacancies) and are also known as LaₓWₓO₅₅–₄₋ₓ, La₅₃WO₁₁₋₄₋ₓ.

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**Fig. 14** (a) Crystal structure of weberite La₃MO₇ composed by chains of corner sharing MO₆ octahedra and edge-sharing LaO₈ cubes. (b) Calculated most stable proton locations in La₃NbO₇ and representation of the lowest energy proton percolation pathway formed by hopping (1) and rotation (2) motion along the non-shared equatorial oxygen atoms of the NbO₆ octahedra.

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**4.2 REₓBO₇ oxides**

Niobates and tantalates of composition REₓMO₇ (RE = Y, La–Lu; M = Nb, Ta) present mixed electronic, oxide ion and proton transport properties, with the rotational step having the lowest energy barrier. Contrary to BaCeO₃-based perovskites, doped-A₂Ce₂O₇ materials exhibit excellent chemical stability in H₂O- and CO₂-containing atmospheres, showing no signs of reaction after exposure at 700 °C for 100 h. These systems are investigated for ammonia synthesis, hydrogen separation membranes and fuel cell applications, with peak power densities >700 mW cm⁻² at 700 °C reported for PCFCs with doped A₂Ce₂O₇ electrolytes.

Atomistic calculations on pyrochlore La₃Zr₂O₇ demonstrated that protons are located on a lowest energy site on the shared apical oxygen of the corner-sharing ZrO₆ octahedral units (Fig. 13b). Long-range proton diffusion occurs along the ZrO₆ octahedral network, through an intra-polychedral migration pathway composed by two separated hopping paths. A proton hops on the edges of the equilateral triangle formed by three adjacent equivalent proton sites and can then migrate towards the next closed region (Fig. 13c). The two paths have very similar migration energies of 0.32 eV and 0.39 eV respectively. A third path corresponding to partial rotation around the oxygen atom has little contribution to the long-range conduction due to its considerable higher energy (0.54 eV). This is in contrast with conventional perovskite conductors, where migration occurs via a rotation and hopping mechanism with the rotational step having the lowest energy barrier.
or \( \text{La}_6\text{WO}_{12} \). They exhibit proton transport number of \( \sim 1 \) and conductivities of \( \sim 10^{-3} \text{ S cm}^{-1} \) at 600 °C.\(^{242,244,246}\) These systems present good stability under \( \text{CO}_2 \) atmosphere, with no signs of degradation after annealing at 700 °C and 800 °C.\(^{249}\) \( \text{La}_6\text{WO}_{12} \) is also chemically and mechanically compatible with conventional cathode materials such as \( \text{La}_{0.7}\text{Sr}_{0.3}\text{MnO}_3 \) (LSM) and \( \text{La}_{0.75}\text{Sr}_{0.25}\text{MnO}_3 \) (LSCM) and chromite anode perovskites such as \( \text{La}_{0.85}\text{Sr}_{0.15}\text{Cr}_0.9\text{Mn}_0.1\text{O}_3 \) (LSCN) and \( \text{La}_{0.75}\text{Ce}_0.2\text{Sr}_{0.1}\text{O}_3 \) (LSCCo), while it reacts with BSCF.\(^{250-253}\) Fuel cells with \( \text{La}_6\text{WO}_{12} \)-based electrolytes display open circuit voltage (OCV) values close to the theoretical Nernst voltage, although delivering only modest power densities (\( \sim 50 \text{ mW cm}^{-2} \) at 700 °C).\(^{244,251,254}\) Due to the presence of mixed protonic and electronic conduction at \( T > 700 \text{ °C} \), \( \text{La}_6\text{WO}_{12} \)-based materials have also been investigated for application as high temperature hydrogen separation membranes, which constitute an interesting solution for the recovering of \( \text{H}_2 \) from low-quality gases.\(^{255}\) High hydrogen permeability rates \( > 10^{-2} \text{ mL min}^{-1} \text{ cm}^{-2} \) have been obtained for Mo- and Re-substituted \( \text{La}_6\text{WO}_{12} \)-based membranes due to a substantial increase in the electronic conductivity,\(^{256,257}\) while permeation rates of 0.15 \( \text{ mL min}^{-1} \text{ cm}^{-2} \) have been reported for dual-phase composite membranes.\(^{258}\) These values are significantly higher than the typical hydrogen rates of state-of-the-art doped \( \text{SrCeO}_3 \) perovskite membranes (\( \sim 10^{-4} \text{ mL min}^{-1} \text{ cm}^{-2} \)).\(^{259,260}\)

\[ \text{La}_{28-x}\text{W}_{4x}\text{O}_{44+2x/2}\text{V}_{2-x/2} \] compounds crystallise in an oxygen deficient fluorite-related tetragonal superstructure composed by \( \text{WO}_6 \) octahedra oriented in alternating directions and with the lanthanum cations in a regular 8-fold cubic coordination (La1) or having a highly distorted 7-fold coordination (La2).\(^{261}\) Some of the La2 sites are partially occupied by the excess tungsten present in the structure (\( x \)), which effectively act as a self-dopant creating intrinsic \( \text{W}_{1/2} \) substitutional defects.\(^{261,262}\) The cation disorder is accompanied by large static disorder in the oxygen sub-lattice.\(^{263-265}\) Characterisation of the hydration and proton conduction mechanisms of these systems is still lacking, probably due to challenges associated in modelling the large anion and cation disorder present in the structure.

Overall, the presence of substantial proton conductivity in oxides having fluorite-related superstructures motivates the investigation of the proton conducting properties of similar structural systems. For example, several bismuth-based oxides having complex fluorite-related superstructures exhibit very high oxide ion conductivities.\(^{266-269}\) The high ionic conduction is due to the presence of anion disorder and of flexible metal polyhedral units with variable coordination; these structural features could potentially also allow for significant proton transport in these systems.

5. Concluding remarks

This perspective has presented an overview of the most significant types of solid oxide proton conductors beyond cerate and zirconate perovskite oxides. Such compounds belong to different structural families and exhibit a variety of distinct water absorption and proton transport mechanisms. The structural and mechanistic features of hydration and proton conduction were discussed, highlighting the salient differences between these alternative materials and the traditional perovskite oxide conductors. The diverse characteristics of these less established conductors point to new potential crystal routes and can be used to inform the discovery of novel solid oxide proton conductors.

A fundamental prerequisite for proton conduction is the incorporation of protonic defects, which in perovskite oxides (and several other proton conductors) usually occurs \( \text{via} \) water absorption on the extrinsic oxygen vacancies introduced by acceptor doping. However, the solubility and nature of the dopants limit the oxygen vacancy concentration and influence the degree of hydration.\(^{2,8,138,270}\) In this respect, systems with inherently defective lattices such as hexagonal perovskite derivatives with oxygen deficient layers and brownmillerite oxides are particularly interesting, as they can absorb large concentrations of water on the intrinsic oxygen vacancies already present in their structures. This also theoretically eliminates any proton trapping effects caused by donor doping, which are usually detrimental for the overall conductivity of conventional perovskite conductors.\(^{8,42,271}\)

Proton transport in perovskites and other oxides having extended octahedral networks usually occurs \( \text{via} \) rotation and hopping of protons along intra-polyhedral pathways. Intra-polyhedral hopping is instead unfavoured in systems containing tetrahedral units, where inter-tetrahedral migration is the common diffusion route. Notably, flexible isolated tetrahedral and variable coordination moieties with high dynamic and rotational mobility enable proton transport. An important aspect that emerges is that the mechanisms of proton transport strongly depend on the coordination environments and the particular topology of the crystal structure. This is in analogy with oxide ion conducting materials,\(^{169,268,272-274}\) and encourages the exploration of structure types with disordered sub-lattices and/or able to support variable coordination environments. Structure-type oxides with extended tetrahedral networks could also constitute an interesting avenue of research.

Recent reports demonstrating significant proton conductivity in semiconducting oxides,\(^{275}\) \( \text{Li}-\)intercalation materials\(^{276,277}\) and high entropy oxides\(^{278}\) further expand the explorable chemical and structural space and inspire innovative strategies for the design of novel proton conducting systems. Highly hydrated metastable oxide phases potentially showing high proton conductivity could also be stabilised \( \text{via} \) high temperature synthesis under humidified atmospheres or with low temperature fabrication processes.\(^{279}\) As in the case of \( \text{Li}-\)ion and oxide ion conducting materials,\(^{280,281}\) computational screening methods could be implemented for selecting candidate solid oxide proton conductors and guide the experimental discovery process.\(^{282}\) The latter should go hand in hand with a thorough determination of the crystal structure characteristics enabling the desired proton transport properties. Neutron and X-ray diffraction techniques will still have a central role in the characterisation of the average crystal structures and identification of the proton sites, essential pieces of information in order to understand the properties of proton conducting oxides.\(^{283-285}\) Analysis of the local structural details with total
scattering (pair distribution function, PDF) methods,286 X-ray absorption spectroscopy287 and solid-state NMR284,285 will be important for the investigation of increasingly complex systems or disordered oxides and to characterise the effects of local order/disorder on the functional properties. The employment of neutron spectroscopic techniques such as inelastic and quasi elastic neutron scattering (INS and QENS), which make use of the contrast given by the large incoherent neutron scattering cross section of the hydrogen atom, will allow to complement the structural studies with the characterisation of the proton dynamics and the diffusional phenomena.284,286 Further opportunities will arise with the development of advanced sample environments enabling in situ studies of materials under operating conditions and with multiple probes. In addition, it is clear how computational techniques and atomistic modelling methods will be complementary to the experimental approach for the comprehensive characterisation of the mechanisms and energetics of protonation and conduction.

Discovery of the next-generation solid oxide proton conductors crucially depends on exploring novel materials and structural types, and on the fundamental understanding of the structure and mechanistic relationships. The latter is of paramount importance for the development of solid proton conducting materials equating or having superior performances to perovskite-based oxides.

Conflicts of interest
There are no conflicts to declare.

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References


