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Surface Properties of α -MnO₂: Relevance to Catalytic and Supercapacitor Behaviour[†]

David A. Tompsett,^{*a} Stephen C. Parker,^a and M. Saiful Islam^{*a}

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Hollandite (α -)MnO₂ gives superior performance compared to other MnO₂ polymorphs in surface sensitive applications in supercapacitors and catalysis. However, a thorough understanding of its atomic-scale surface properties is lacking, which we address here using density functional theory (DFT). A Wulff construction based upon relaxed surface energies demonstrates that the equilibrium morphology expresses the low index (100), (110) and (111) surfaces as well as the high index (211) and (112) surfaces. The predicted morphology exhibits clear elongation along the *c*-axis which is consistent with the large number of nanorod type structures that are obtainable experimentally. The surface structures expressed in the morphology are discussed in detail and it is found that α -MnO₂ gives rise to larger surface relaxations than are observed for the less open rutile structured MnO₂. Enhanced magnetic moments at surface sites are rationalised by a crystal field argument. Experimental studies consistently find that α -MnO₂ has higher catalytic activity than other polymorphs of MnO₂. In this work, calculated formation energies for oxygen vacancy defects at the expressed surfaces are demonstrably lower, by ~ 1 eV, than for rutile MnO₂ surfaces [Tompsett *et al.*, *JACS*, **2014**, *136*, 1418]. The lowest vacancy formation energy occurs at the (112) surface, which despite its relative high Miller index constitutes 17 % of the surface area of the calculated morphology. This may play a key role in the favourable catalytic performance observed for α -MnO₂ in a broad range of applications.

1 Introduction

Energy storage technologies for electric vehicles and grid storage of renewable energy are major research activities for which the Li-O₂ battery and supercapacitors are key candidate systems. Rising demand is driving research into new materials to provide the energy storage capabilities for a low-carbon future¹.

While the conventional Li-ion rechargeable battery based upon a layered LiCoO₂ intercalation cathode has been a stalwart of the portable electronics market, future applications will require even higher energy storage densities². Consequently, the high capacity Li-O₂ battery is the subject of intensive research for which α -MnO₂ is a promising catalyst³. Furthermore, α -MnO₂ has been demonstrated to deliver high capacitance as a supercapacitor electrode, particularly in nanostructured form^{4,5}. In both technologies the good performance is directly connected to the surface properties of α -MnO₂, but despite this the structures and defect formation remain poorly characterised. With a clear motivation to better understand the performance of this material in catalytic and supercapacitor

applications, in this work we apply density functional theory (DFT) simulations to understand surface phenomena in this promising material.

Experimental studies have demonstrated that α -MnO₂ is capable of being synthesised in a diverse array of nanostructures^{3,6–10}. Since the material surface area is critical to electrochemical capacitance, investigators have exploited nanostructuring to obtain dramatically improved storage in supercapacitors based on α -MnO₂. Recent studies^{4,5,11} of nanostructured α -MnO₂ have demonstrated capacitances as high as 328 Fg⁻¹.

α -MnO₂ has also shown promise as a catalyst for the oxygen reduction reaction (ORR)^{3,12,13}. Historically, materials containing precious metals such as platinum, palladium and gold have been the most successful catalysts for the ORR. However, the high cost and limited availability of these materials is prohibitive in important commercial applications. In the search for a low cost alternative, manganese oxides have drawn particular attention due to their elemental abundance, high activity in alkaline media and non-toxicity^{14–19}. While there are many polymorphs of MnO₂, α -MnO₂ in particular has demonstrated high catalytic activity with respect to other polymorphs^{20–22} beyond the effects of high surface area¹².

The Li-O₂ battery is an alternative to Li-ion batteries that is the subject of intense interest due to its high theoretical energy density^{1,3,23–27}. However, experimental studies have shown

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^a Department of Chemistry, University of Bath, Bath BA2 7AY, UK. Fax: +44(0)1225 386231; Tel: +44(0)1225 384938; E-mail: dt331@bath.ac.uk, m.s.islam@bath.ac.uk

that present implementations of the Li-O₂ system may suffer from poor electrical efficiency, in particular due to overpotentials, and power density limitations^{26,28}. To address these issues the use of transition metal oxide catalysts has been suggested^{29,30}, and nanorod α -MnO₂ demonstrated effective catalytic activity for this purpose^{3,31–36}. However, recently the detection of electrolyte decomposition^{8,12,37,38} in LiO₂ cells has highlighted the need to explore the performance of catalysts in stable electrolyte conditions. With recent demonstrations of good electrolyte stability during cycling of redesigned Li-O₂ cells^{27,39,40} it is likely that the utility of catalysts will be important to future studies.

In other work α -MnO₂ has exhibited good catalytic activity in applications beyond Li-O₂ cells, including the reduction of NO_x⁴¹, microbial fuel cells⁴², oxidation of chlorobenzene⁴³, phenol degradation^{44,45}, toluene oxidation^{46,47} as well as oxidation of CO and benzene⁴⁸.

Figure 1 shows the unit cell of α -MnO₂ where the approximate MnO₆ octahedra are indicated by polyhedra. In-plane there are both corner and edge sharing octahedra forming a 2×2 tunnel structure along the c-axis. Along the c-axis all of the octahedra are edge sharing forming a densely packed wall to the tunnel. The 2×2 tunnel structure has been described as “open” compared to the comparatively denser rutile (β) polymorph of MnO₂ and this has been implicated in its excellent catalytic performance²¹. The rationalisation is based upon the reduced structural distortion and bond-breaking that is required in important catalytic processes such as oxygen defect formation and adsorption. The large 2×2 tunnels also provide insertion channels for adsorbed ions such as Li-ions when α -MnO₂ is used as a battery intercalation cathode⁴⁹. Indeed, a previous modelling study of α -MnO₂ by Trahey *et al.*⁵⁰ has treated the thermodynamics of Li-ion and Li-oxide insertion at the surface of α -MnO₂. Based upon the obtained insertion potentials they suggested that Li-oxide insertion and extraction may be an active mechanism in the performance of α -MnO₂ as a catalyst in the Li-O₂ battery system.

A recent DFT study by Ling and Mizuno⁵¹ has considered Li-ion and Li-oxide insertion into bulk α -MnO₂. That work considered structural distortions upon lithiation and the thermodynamics of insertion. They also considered the effect of Li₂O insertion upon the electronic structure. Li and Na insertion into bulk α -MnO₂ has been considered by some of the present authors using DFT⁵². Cockayne and Li⁵³ have studied the magnetic structure and the adsorption of dopants for bulk α -MnO₂ and Crespo *et al.* also treated the magnetic structure using DFT⁵⁴. Liu *et al.* recently considered the surface energy and magnetic order of the (110) surface of α -MnO₂ by similar methods⁵⁵. Cheng *et al.* used DFT to consider adsorption of oxides at rutile MnO₂ surfaces⁵⁶. They found that oxygen vacancies enhance the interaction between oxide species and the MnO₂ surface. Furthermore, the presence of oxygen va-

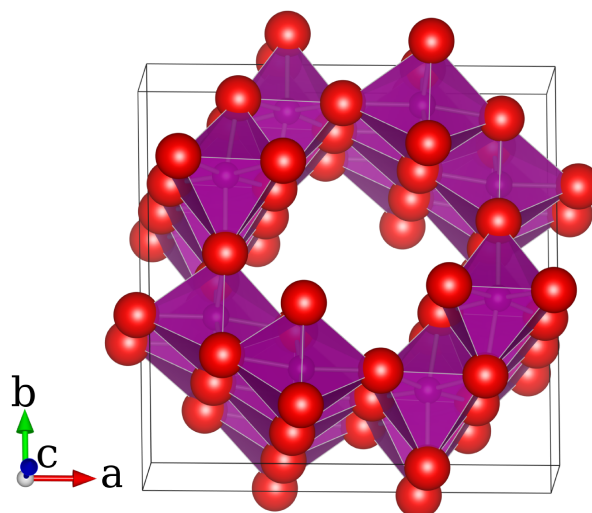


Fig. 1 Crystal structure of α -MnO₂. Small (red) spheres are oxygen and large (purple) manganese lie inside the indicated approximate MnO₆ octahedra. Note the large 2×2 tunnel structures along the c-axis.

cancies reduced the enthalpy of the oxygen reduction process at MnO₂ surfaces by more than 2 eV. A recent first-principles study of manganese oxides catalysts for the ORR and oxygen evolution reaction (OER) constructed Pourbaix diagrams of stable surface configurations of MnO_x stoichiometries as a function of applied potential and pH⁵⁷. That work found that the oxygen adsorbed (110) surface of rutile MnO₂ was active at experimental conditions for the OER. However, as highlighted by the authors they did not consider the presence of α -MnO₂, nor the energetics of oxygen vacancy formation.

Due to the renewed interest in α -MnO₂, in this work we perform a comprehensive *ab initio* study of its surface properties. The study of α -MnO₂ surfaces in the present work shows a tendency for this material to form rod-like morphologies that are elongated along the c-axis, which is likely to be important to the formation of nanostructures with improved capacitance. First we outline our computational methods. Then we present results on the structures and energies of stoichiometric surfaces, as well as the equilibrium crystal morphology. Following this the calculated energetics of oxygen vacancy formation at surfaces are discussed with reference to catalytic activity.

2 Computational Methods

Ab initio simulations of surface energies and morphologies, particularly in complex and lower symmetry structures, are computationally expensive. In this study, we aim take advantage of efficient interatomic potentials as a screening tool for candidate low energy surfaces. These low energy candidates

are then treated by *ab initio* DFT which more fully describes the electronic effects in the system.

This study employs well-established interatomic potentials methods, which are reviewed in detail elsewhere^{58–60}. The model due to Parker and co-workers^{61,62} for rutile β -MnO₂, that has been utilised successfully in several studies^{63,64}, is augmented by the inclusion of a core-shell model for the oxygen. The model reproduces the experimental lattice parameters of α -MnO₂ within 1.5 % and full details are presented in the supporting information. The interactions between ions are represented in terms of a long-range Coulmbic term with the addition of an analytic term representing short-range interactions. The surface energies from interatomic potentials were then calculated using METADISE⁶⁵ up to high indexes.

Candidate low energy surfaces obtained from interatomic potentials were then treated by DFT as implemented in the VASP code⁶⁶ with PAW potentials. The electronic structure was calculated using the Generalized Gradient Approximation (GGA) of Perdew, Burke and Ernzerhof⁶⁷ with Hubbard *U* corrections (PBE+U). The PBE+U exchange-correlation functional has been demonstrated to give a good description of defect properties in other oxides including CeO₂⁶⁸ and TiO₂⁶⁹. All calculations were performed in a ferromagnetic spin polarised configuration, while effects of more complex magnetic order, due to their low energy scale, are left for future work. To obtain the equilibrium lattice parameters by relaxation of the bulk cell a cutoff for the planewave basis set of 520 eV was used to avoid Pulay stress. All subsequent calculations were performed based upon the obtained equilibrium lattice constants without cell relaxation using a cutoff of 400 eV. This included the recalculation of the energy for the bulk unit cell so that all comparative energies were obtained at a constant cutoff of 400 eV. A minimum of $4 \times 4 \times 4$ *k*-points was used in the Brillouin zone of the conventional cell and scaled appropriately for supercells.

The value of the *U* parameter for our PBE+U calculations was determined *ab initio* using Wien2k^{70,71}. Previous work conducted by some of the present authors^{72,73} demonstrated a good description of lithium intercalation, band gaps and magnetic interactions when PBE+U is applied in the fully-localized limit⁷⁴, which we use in this work as well. We employ $(U - J) = 5.2$ eV, for the spherical part of the interaction, and $J = 1.0$ eV. Further details are described in the supporting information. These parameters have been shown to give accurate results for the thermodynamics of Li-ion insertion and electronic structure of α -MnO₂ in previous work by some of the present authors⁵².

The α -MnO₂ crystal occurs in the tetragonal space group *I4/m* (# 87) with lattice parameters⁷⁵ $a = b = 9.750$ Å and $c = 2.861$ Å. In Table 1 we show the calculated lattice parameters for α -MnO₂ from PBE+U. These results agree with the experimental parameters within 2.5 %, but the common

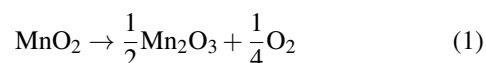
tendency for PBE+U to overestimate the unit cell volume is evident⁷⁶.

Table 1 Predicted PBE+U and experimental⁷⁵ lattice parameters for α -MnO₂.

α -MnO ₂	a (Å)	b (Å)	c (Å)
PBE+U	9.907	9.907	2.927
Experimental	9.750	9.750	2.861

Surface energies are calculated based upon slabs with a minimum thickness of 20 Å, where for each surface the slab thickness is increased until convergence of the surface energy is obtained. A vacuum thickness greater than 20 Å was used throughout to separate the slabs from their periodic image. All dipole free surface terminations were considered with Miller indices up to and including 2. Full geometrical relaxation of the ionic positions until the forces fell below 0.02 eV Å⁻² was performed to incorporate surface relaxation effects.

To obtain the defect energies with respect to the stoichiometric surface, a single *k*-point PBE calculation of an isolated spin triplet oxygen molecule in a 15 Å side cubic cell was performed using the same energy cutoff and convergence criterion as for the slab calculations. From this we obtain a reference energy of -9.85 eV for the oxygen molecule. However, the PBE+U functional is known to give rise to errors in formation energies of oxide materials due to both an inadequate description of the oxygen molecule (including overbinding) and of the solid oxides. Hence as in previous work⁷⁷, to obtain an estimate of the error as relevant to the process of oxygen vacancy formation with accompanying Mn reduction at α -MnO₂ surfaces we have calculated the enthalpy of the following reaction:



This reaction incorporates the removal of oxygen and the reduction of Mn. Comparison of the reaction enthalpy from our PBE+U calculation with the experimental enthalpy (40.5 kJ/mol)⁷⁸ indicates a correction of 1.07 eV per oxygen. The correction is applied to all oxygen vacancy formation energies quoted in this work.

3 Surface Energies and Crystal Morphology

Experimentally, α -MnO₂ has been synthesised in a range of morphologies. Commonly, the synthesised material resembles a network of small nanorods at the nanoscale^{10,79}. However, computational studies of the surfaces expressed in these crystals are lacking. Before a good understanding of the catalytic and surface properties in this system can be obtained it is essential that the characteristics of the exposed surfaces are understood at the atomic-scale. In this section we address this

Table 2 Predicted PBE+U surface energies for α -MnO₂.

Miller Index	E_{surf} (J m ⁻²)
(100)	0.64
(001)	1.76
(110)	0.75
(111)	1.32
(101)	1.91

(102)	1.55
(112)	1.40
(122)	1.54
(201)	1.51
(211)	1.08
(221)	1.59

(310)	1.41

question by the calculation of surface energies from which we determine a Wulff constructed crystal morphology.

First-principles surface studies deliver high accuracy, but are computationally costly. Consequently, practitioners are generally forced to treat only a limited set of surfaces, while experimentally the number of possible combinations of Miller indexes is notionally infinite. Typically, low index surfaces, i.e. those with only 1's and 0's like (110), are chosen for this set due to their tendency to have favourable surface coordination. In this work favourable coordinations are those close to the 6-fold Mn-O coordination found in the bulk of MnO₂. However, our recent study⁷⁷ of rutile MnO₂ surfaces has demonstrated that relatively high coordination and low surface energy does occur for high Miller indexes. To screen high Miller index surfaces we have calculated the surface energies of 34 surfaces with our interatomic potentials model. The resulting surface energies are shown in supporting information Table S2 and the resulting Wulff constructed morphology in Fig. S1. We have then treated all low index surfaces, and high index surfaces that are low in energy, by first-principles PBE+U.

In Table 2 we show the obtained PBE+U surface energies for α -MnO₂. The surface energy is calculated by taking the difference between the energy of the constructed slab and the same number of α -MnO₂ formula units in the bulk:

$$\gamma = \frac{E_s - nE_b}{2A} \quad (2)$$

Here, E_s is the energy of a slab containing n formula units and E_b is the total energy per formula unit of bulk α -MnO₂. A is the area of the slab surface and the factor of 2 reflects the fact that there are two surfaces for each slab. The (100) and (110) surfaces, with low energies of 0.64 and 0.75 J m⁻² respectively, are the most stable. All other surface energies are greater than 1 J m⁻².

We will return to discuss the surface terminations and electronic properties in more detail, but first we present the mor-

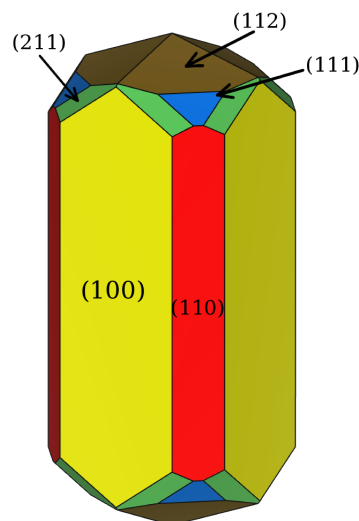


Fig. 2 The predicted equilibrium crystal morphology for α -MnO₂ based upon the surface energies in Table 2.

phology of α -MnO₂ from a Wulff construction using these energies. The resulting morphology is shown in Fig. 2 and has an elongated shape, which is consistent with the large number of experimental synthesis studies that observe nanorod crystals^{3,8,79-83}. The (100) surface dominates and represents 54 % of the crystal surface area followed by the (110) surface that is 20 % of the surface area. However, the high index (112) surface represents a substantial 17 % of the area.

The presence of the (100) surface has been observed by Transmission Electron Microscopy (TEM) experiments of Zhang *et al.*⁷⁹ based upon measured fringe spacings of α -MnO₂ nanorods. Furthermore, TEM studies have indicated the presence of the (110) surface^{8,80,81} in synthesized nanostructured α -MnO₂. These observations are consistent with the calculated low energy for these surfaces and their expression in the Wulff morphology of Fig. 2. These results indicate that the calculated surface energies are consistent with observed experimental morphologies. To our knowledge the present work is the first comprehensive study of the morphology of α -MnO₂ and the good correspondence with experiment shows the ability of computational simulations to enhance our understanding of this system.

Finally, some experimental studies also suggest the presence of the (310) surface^{82,83}, where we note the reaction of permanganate MnO₄⁻ with acidic reagents, H₂SO₄ and HNO₃, in both of those studies. However, our calculated surface energy of 1.41 J m⁻² for the (310) surface indicates that its expression is likely to be a result of non-equilibrium synthesis conditions. A treatment of off-stoichiometry at α -MnO₂, such as was recently performed by Oxford *et al.*⁸⁴ for low index surfaces of rutile MnO₂, may allow further insight regarding

Table 3 Mn-O coordinations and magnetic moments for low energy surfaces of α -MnO₂. Surface Mn-ions are labeled as in Figs. 3-7.

Surface	Site	n_{coord}	Moment (μ_B)
(100)	Mn(1)	5	3.70
	Mn(2)	6	3.12
	Mn(3)	6	3.12
	Mn(4)	6	3.15
(110)	Mn(1)	5	3.55
	Mn(2)	6	3.17
	Mn(3)	6	3.12
(111)	Mn(1)	3	4.31
	Mn(2)	4	3.88
	Mn(3)	4	3.43
	Mn(4)	5	3.35
	Mn(5)	5	3.28
(211)	Mn(1)	5	2.99
	Mn(2)	5	2.91
	Mn(3)	5	3.14
	Mn(4)	4	1.83
(112)	Mn(1)	3	3.87
	Mn(2)	4	3.80
	Mn(3)	4	3.97
	Mn(4)	5	3.43

the morphological dependence upon experimental synthesis conditions. Such investigations, which like the present study require a large computational cost, will be the subject of future work.

4 Surface Structures and Magnetic Moments

In order to fully understand the impact of the surfaces of α -MnO₂ on its catalytic and electrochemical performance it is essential to probe these surfaces beyond the calculation of their energies. Having determined the surface energies and resulting Wulff crystal morphology, in this section we inspect the atomic structure at specific cleaved surfaces in detail. The goal here is to understand how the fundamental characteristics of the cleaved crystal drive the creation of the obtained elongated morphology and furthermore to gain insight into the phenomenology that occurs at these surfaces when α -MnO₂ is used in electrochemical and catalytic applications.

In Table 3 we summarize the coordination numbers and magnetic moments for the surface Mn ions of the low energy surfaces. We begin by considering the (100) surface that dominates the Wulff morphology. The (100) surface Mn ions are clearly well coordinated with three of the four surface sites possessing 6-fold Mn-O coordination. Indeed 6-fold coordination is the same as that which occurs in the octahedra that are the building block of the bulk α -MnO₂ crystal. The fact that the cleavage plane that forms the (100) surface, shown in

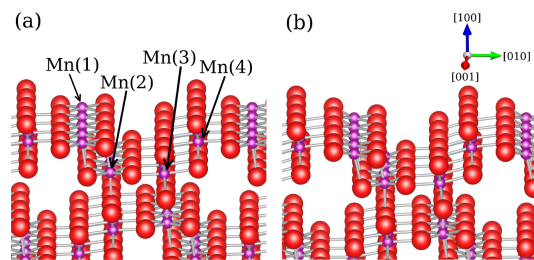


Fig. 3 Structure of the (100) surface of α -MnO₂: (a) Before relaxation and (b) after relaxation.

Fig. 3, maintains the 6-fold coordination so well is very likely to be important to the low calculated surface energy of 0.64 J m⁻². The stability of the cleaved surface is also evident in a comparison of the surface before relaxation (Fig. 3(a)) and after relaxation (Fig. 3(b)). The primary difference between the structures is a small outwards movement of two oxygen ions that coordinate Mn(1). This outward movement acts to improve the coordination sphere around Mn(1) which is the only 5-fold site produced by the (100) cleavage plane. Oxygen terminated surfaces may present favoured sites for ion adsorption, e.g. H⁺, when α -MnO₂ is employed as a supercapacitor electrode. Finally, as shown in Table 3 there is a clear enhancement of the magnetic moment for Mn(1) to 3.70 μ_B due to its lower 5-fold coordination, while the 6-fold coordinate Mn(2-4) possess bulk-like moments of $\sim 3.12 \mu_B$. This indicates that structural surface distortions do have a marked effect on the electronic properties and we will return to discuss the causes of this enhancement in detail.

At the (110) surface the coordination of Mn ions, listed in Table 3, is high, two Mn ions being 6-fold and one 5-fold coordinate. This is a significant contributor to the low surface energy of 0.75 J m⁻² obtained at this surface. As shown in Fig. 4 this surface has the structure of a series of repeating ledges formed by rows of oxygen ions that run along the [001] direction. Prominent at the surface is a bridging oxygen ion attached to Mn(1) ions. Indeed it is the relaxation of this oxygen site that dominates the structural relaxation at the cleaved surface. However, due to the good coordination at this surface, average coordination number $5\frac{2}{3}$, the crystal structure undergoes only a small amount of relaxation when cleaved as indicated by a comparison of Fig. 4(a) and Fig. 4(b). Furthermore, since the bridging oxygen site is only attached to two Mn ions it is likely to be a site of facile oxygen vacancy formation, which is relevant to the catalytic activity of α -MnO₂. The (110) surface was also found to be expressed in a Wulff morphology of rutile MnO₂⁶² and together with the (100) dominates the make up of the in-plane components of the calculated equilibrium morphology in α -MnO₂.

The crystal morphology must be capped by surfaces with a

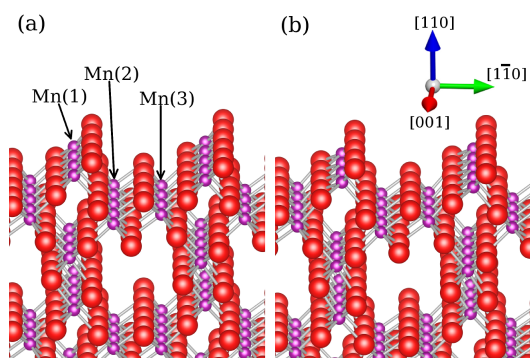


Fig. 4 Structure of the (110) surface of α - MnO_2 : (a) Before relaxation and (b) after relaxation.

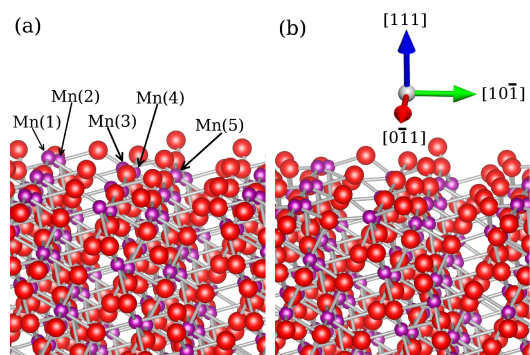


Fig. 5 Structure of the (111) surface of α - MnO_2 : (a) Before relaxation and (b) after relaxation.

Miller index that includes a non-zero c -axis component. The (111) facet with a surface energy of 1.32 J m^{-2} forms part of the end cap for the calculated morphology in Fig. 2, but is significantly higher in energy than the (110) and (100) surfaces. The end cap of the morphology refers to facets that have some component of their surface normal along the c -axis i.e. they form part of the ends of the rod. It is the disparity in energy between the lower energy in-plane (110) and (100) surfaces, and the higher energy surfaces that cap the morphology that leads to its elongated structure. The qualitative reason for the disparity in surface energies may be ascertained from Table 3 where it is clear that the Mn ions at the (110) and (100) surfaces possess higher coordination than at any other expressed surface. Inspection of Fig. 5 shows that the (111) surface has a complex structure with five symmetry inequivalent surface Mn ion sites. As listed Table 3 the undercoordinated 3-fold Mn(1) site possesses a strongly enhanced magnetic moment of $4.31 \mu_B$, a topic that we will return to discuss in detail.

The (211) surface, which has a surface energy of 1.08 J m^{-2} , is also expressed in the end caps of the morphology. The average coordination number of the Mn ions at this surface, $4\frac{3}{4}$, is considerably lower than for the in-plane (100) and

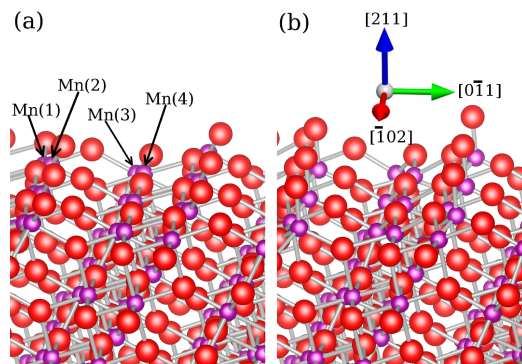


Fig. 6 Structure of the (211) surface of α - MnO_2 : (a) Before relaxation and (b) after relaxation.

(110) surfaces. However, if we compare this value to the other expressed surfaces that cap the morphology their average coordination numbers are 4.2 for (111) and 4 for (112), which emphasises why the (211) is lower in energy than these surfaces. The primary relaxation at this surface, shown in Fig. 6, is the outward movement of a dangling oxygen ion that is singly bonded to the Mn(2) site. Finally, an interesting feature of the low coordinated Mn(4) ion is its low magnetic moment of just $1.83 \mu_B$.

The remaining surface that is expressed in the morphology is the (112) facet, which dominates the end cap surface area and represents 17 % of the crystal surface area as a whole. Comparison of Fig. 7(a) and (b) indicates that the primary relaxation at the cleaved surface is the outward movement of oxygen sites to improve the coordination spheres around Mn(1) and Mn(2), which are only 3-fold and 4-fold coordinated to bonded oxygen sites. This is accompanied by a drawing in of the Mn(1) and Mn(2) sites towards the surface. We note that this indicates, along with the other expressed surfaces, a tendency to favour oxygen termination in the outermost surface layer. The presence of these oxygens may provide key sites for surface reduction during catalytic activity and cation adsorption during electrochemical application.

Finally, we return to discuss the interesting enhanced magnetic moments observed at some surface Mn sites. The presence of enhanced moments accompanies undercoordination of the surface Mn site. For instance, the 5-fold coordinated Mn(1) site of the (100) surface possesses a magnetic moment of $3.70 \mu_B$ compared to $3.10 \mu_B$ for the bulk material. Inspection of Table 3 indicates the prevalence of these enhanced moments, which occur at all of the expressed surfaces in the crystal morphology. The enhancement of these moments strongly suggests a tendency for the presence of Mn^{3+} (d^4) at the surfaces, as demonstrated by a recent X-ray photoelectron spectroscopy study of α - MnO_2 nanowires³⁴. This is consistent with the absence of a full coordination sphere of electroneg-

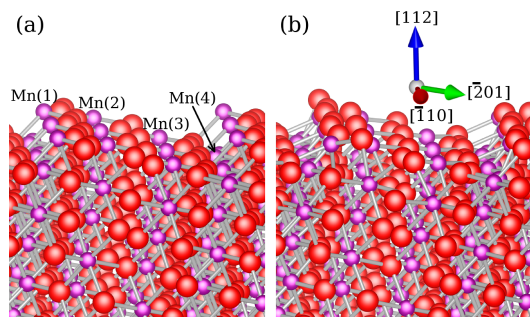


Fig. 7 Structure of the (112) surface of α -MnO₂: (a) Before relaxation and (b) after relaxation.

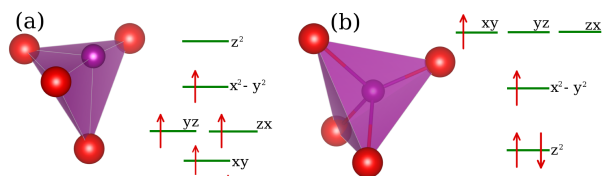


Fig. 8 Structure of the crystal field environment for (a) the high spin 4-fold Mn(2) sites at the (111) surface and (b) the low spin 4-fold Mn(4) site at the (211) surface.

ative O ions around these sites. However, we emphasise that our calculations indicate that α -MnO₂ should not be thought of as purely ionic but possesses intermediate ionic-covalent character.

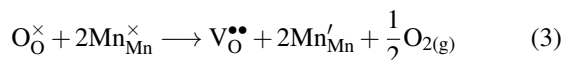
At the (111) surface the low 4-fold coordination of the Mn(2) site results in an enhancement of the magnetic moment to $3.88 \mu_B$. However, there is no clear mathematical relationship between coordination number and magnetic moment. Rather, the resulting magnetic moment depends not only on the coordination number, but also the specific geometry of the coordination. For the Mn(2) 4-fold coordinate sites at the (111) and (112) surfaces the Mn environment can best be thought of as severed octahedral. Compared to an octahedral environment the Mn coordination is missing a bond along the x and y directions, which lowers the energy of the d_{xy} and $d_{x^2-y^2}$ orbitals as shown in Fig. 8(a). It is clear that the moment of $3.88 \mu_B$ at the (111) surface corresponds to a high spin state. Enhanced moments in this environment have recently been observed in a combined experimental (NMR and magnetic susceptibility) and theoretical investigation of Co sites in LiCoO₂⁸⁵. For (211) the 4-fold coordinate environment is more like a strongly distorted tetrahedral environment as depicted in Fig. 8(b), which results in a low spin moment of $1.80 \mu_B$. In the perfect tetrahedral environment the e_g orbital lies lower than the t_{2g} . However, the distorted tetrahedon possesses one bond angle that is strongly decreased from the ideal 109.5° to 87° . Furthermore, one of the Mn-O bond lengths is

contracted to 1.59 \AA . These distortions split the e_g orbital by increasing the $d_{x^2-y^2}$ energy. This results in the 4-fold Mn site of the (211) surface possessing a double occupation of the d_{z^2} orbital and a low spin moment. Further analysis of the unusual surface electronic structure is warranted, but since our prime motivation is to explore surface morphologies and oxygen vacancy formation this will be pursued in future work.

5 Oxygen Vacancy Formation and Catalytic Activity

α -MnO₂ has been demonstrated to have high catalytic activity. A recent study by Giordani *et al.* showed that α -MnO₂ delivered the highest catalytic activity for the decomposition of H₂O₂ when tested alongside a set of other transition metal oxides⁸⁶. Furthermore, comparative investigations^{3,21,87} consistently indicate that the hollandite crystal structure of α -MnO₂ delivers substantially better catalytic performance than other polymorphs of MnO₂. Consequently, a key question for this section is to answer why studies persistently find the α polymorph to have the highest catalytic activity.

The process of oxygen defect formation at MnO₂ surfaces may play a key role in its catalytic activity. For instance, the Mars-van Krevelen mechanism⁸⁸ requires a redox process in which oxygen is consumed from the surface and then replenished from the gas phase. Recent work has shown that the presence of oxygen vacancies at rutile MnO₂ surfaces enhances its catalytic activity for the ORR^{41,56,89}, which makes this investigation of surface properties timely. Indeed an *in situ* X-ray photoelectron study of the catalytic activity of α -MnO₂ for reduction of NO and N₂O correlated high catalytic performance with the presence of Mn³⁺ ions and oxygen vacancies⁴¹. Vacancy formation energies have been calculated according to the following reduction process:



where Kroger-Vink notation has been used. Here, the lattice site that a species occupies is indicated by the subscript, and the superscript shows the electric charge of the species relative to that of the charge of that site in a stoichiometric crystal: \times indicates no change to the charge, a single $'$ represents a negative charge and a \bullet represents a positive charge. We note that this process incorporates oxygen vacancy formation as well as the reduction of Mn⁴⁺ to Mn³⁺, which corresponds to the process observed by X-ray photoelectron experiments⁴¹.

The defect energies for removal of all symmetry inequivalent surface oxygens have been calculated and the lowest defect energy obtained for each surface is shown in Table 4. Since it is the dilute, low concentration vacancy defects that are likely to be important to catalysis the defect formation en-

Table 4 Predicted PBE+U oxygen vacancy formation energies at surfaces of α -MnO₂ based on equation 3.

Miller Index	$\Delta E_f(O-vac)$ (eV)
(100)	1.02
(110)	0.97
(111)	0.30
(211)	1.09
(112)	0.07
Bulk	2.57

energies were calculated for large surface supercells and converged with respect to slab thickness.

The key results firstly, reveal the low defect formation energies of 0.07 eV and 0.3 eV for the (112) and (111) surfaces respectively. Secondly, there are low defect formation energies at all expressed surfaces. In fact, the *highest* energy of 1.09 eV at the (211) surface is only slightly higher than the *lowest* defect formation energy of 0.96 eV obtained for rutile MnO₂ surfaces in a recent study by the present authors⁷⁷. This strongly suggests that the good catalytic performance of α -MnO₂ compared to other transition metal oxides and other polymorphs of MnO₂ results from the facile oxygen vacancy formation at its surfaces.

It is worth noting that surface coordination, which we have discussed for its impact on surface energies and magnetic moments, is also important to the energetics of oxygen vacancy formation. For example, to provide the low formation energy of 0.07 eV at the (112) surface an oxygen that is attached to the 5-fold coordinate Mn(4) site is removed. For the low formation energy of 0.3 eV at the (111) surface an oxygen is removed at a 6-fold site in the subsurface. However, we note that it is not only the coordination of manganese sites that is important, but also the Mn coordination around the removed oxygen. For example, at the (110) surface the removal of a bridging oxygen from the 5-fold coordinate Mn(1) is favoured over the removal of oxygens that are attached to 6-fold Mn(2) sites because such oxygens give coordination to three manganese sites, while removing the bridging oxygen only breaks the coordination of two Mn sites. Therefore, we suggest empirically that oxygen vacancy formation is most favourable when the removed oxygen has low coordination to Mn, but the attached Mn sites have high coordination to oxygen.

It is interesting to note that the bulk oxygen vacancy defect formation energy of 2.57 eV is actually slightly higher than that calculated for bulk rutile MnO₂ (2.26 eV). Consequently, it is clear that the surfaces produced by cleaving the α -MnO₂ crystal structure are the important factor in the facile production of oxygen vacancies, rather than its bulk properties.

The crystal structure of α -MnO₂ is more open than other polymorphs such as rutile and ramsdellite due to the presence of the large 2×2 tunnel structures. The importance of this

openness has been discussed with reference to the high catalytic activity of α -MnO₂ for oxidation of CO²¹. Upon cleaving, the more open crystal structure gives rise to a greater diversity of oxygen surface sites as well as better overall coordination of Mn surface sites compared to rutile MnO₂. While effects of the electronic structure will also be important, this larger phase space in the surface crystal structure is likely to be a key influence on the low oxygen vacancy formation energies.

The low defect formation energies obtained at α -MnO₂ surfaces demonstrate a mechanism for its good catalytic performance for the ORR. The results clearly indicate that the catalytic activity is driven by fundamental surface properties, beyond the large surface area of nanoparticles. These high performance properties make it an important, low cost alternative to precious metal catalysts for technological applications, including the topical Li-O₂ battery.

6 Conclusions

Advanced materials simulations have been employed to enhance our understanding of α -MnO₂ surfaces, an important material for technological applications. The key findings include:

- 1) The calculated equilibrium morphology is elongated along the c-axis, which is consistent with the large number of nanorod structures obtained from experimental synthesis. The (100) and (110) surfaces dominate the in-plane morphology and constitute 54 % and 20 % of the overall surface area respectively. Both surfaces have been observed by TEM experiments. The surfaces that form the ends of the rod-like morphology are substantially higher in energy, by more than 45 %, which is the driver of the elongation along the c-axis.
- 2) Due to their ability to provide favourable surface coordination, the higher Miller index (211) and (112) surfaces are expressed in the morphology. Coordination at surface sites is found to be important to obtaining low surface energies. The open crystal structure of α -MnO₂ with 2×2 tunnels provides a large number of surface configurations for satisfying surface coordinations.
- 3) We find evidence for the presence of both enhanced high spin and low spin Mn sites at cleaved surfaces. The magnetic moments are explained rationally by the crystal field energy scales in the cleaved surface environments.
- 4) Regarding the key question as to why α -MnO₂ provides higher catalytic activity than other related oxide materials, our results present compelling evidence that facile oxygen vacancy formation is at play. The low oxygen vacancy formation energies, all in the range 0.07 eV to 1.09 eV, indicate a favourable thermodynamic pathway for catalytic processes, for instance via the Mars-van Krevelin mechanism. The results demonstrate that the catalytic activity of α -MnO₂

is not merely a consequence of the large surface area of synthesised nanoparticles.

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