Hybrid-DFT Modeling of Lattice and Surface Vacancies in MnO

Andrew J. Logsdail, Christopher A. Downing, Thomas W. Keal, Paul Sherwood, Alexey A. Sokol, and C. Richard A. Catlow

ABSTRACT: We have investigated the properties of defects in MnO bulk and at (100) surfaces, as used in catalytic applications, using hybrid-level density functional theory (i.e., inclusion of exact exchange within the exchange-correlation evaluation) in a hybrid QM/MM embedded-cluster approach. Initially, we calculate the formation energy for bulk Mn and O vacancies, comparing charged-defect compensation with charge carriers at the Fermi Level ($\epsilon_f$) and through Schottky defect formation. Oxygen vacancies were also investigated at the (100) surface, where the vacancy formation energy is very similar to the bulk. Defect levels associated with the most stable vacancies are calculated using the ∆SCF method: all are positioned mid band gap, with surface environments failing to alter strongly the overall nature of the defect relative to bulk. Chemical activity of the (100) MnO surface was considered through the adsorption of a probe CO$_2$ molecule, which is considered the initial step in the transformation of CO$_2$ into hydrocarbons. CO$_2$ adsorption was strongest over a neutral oxygen vacancy, where the associated trapped electrons of the defect transfer to the adsorbate and thus activate it. However, we have shown with our embedded-cluster approach that the neutral oxygen vacancy is not necessarily the dominant species, which has implications when interpreting results for future catalytic applications.

1. INTRODUCTION

Understanding the properties and stability of material defects is key in order to maximize the impact of materials engineering. Rock salt metal oxides are ideal test systems when studying how defects alter fundamental physico-chemical behavior, with their stability also making them particularly appealing for applications that do not involve ambient conditions. In particular, rock salt oxides formed from early group 2 alkaline earth metals (Mg, Ca) have been extensively studied, both with respect to their defect structure and properties and with the focus on potential applications in the catalytic transformation of CO$_2$. In these investigations, computational methods lend themselves to investigating both the viability of reaction mechanisms as well as the large compositional search space for novel materials, with systematic investigations establishing a new standard when looking for beneficial material characteristics. Computational screening of this nature was recently used to examine CO$_2$ adsorption on a variety of rock salt (100) surfaces, showing strong surface–adsorbate interactions and charge transfer for a range of materials. One material that plays an important role in these studies is MnO, which is typically considered as a supporting material or a reaction promoter, rather than an active part of catalytic processes, due to its variable oxidation state; in this highlighted work, however, MnO was identified as being reactive toward CO and CO$_2$. This is perhaps unsurprising given that MnO can be prepared by thermal decomposition of MnCO$_3$, which is done in an anaerobic environment to prevent the formation of higher oxidation state compounds. Observations of chemical reactivity are not without precedent for MnO: concerted effort has been applied toward characterizing the electronic properties of MnO-ZnO mixtures due to hypothesized photocatalytic capabilities. It has also recently been of interest to tune catalytic properties by introducing individual defects, intrinsic or extrinsic, to the oxide framework; such materials...
have the stability of the host material while displaying novel catalytic chemistry. An example application of this nature is CO₂ adsorption over the (100) surface of Mn-doped MgO, with and without neighboring anion vacancies. The results showed that the multivalent Mn caused chemical bonding between the support and the adsorbate that was not present for undoped-MgO, which complements experimental observations for mixed Fe-MnO/MgO catalysts as being reactive for the formation of light olefins from syngas.

In this work, we have performed calculations to investigate the defect energetics in the bulk and at the (100) surface of MnO, and also investigated the surface reactivity. Defects, both intrinsic and extrinsic, are key for controlling the chemical properties of catalytic materials and optimizing surface reactivity. However, one of the complexities when investigating magnetic materials like MnO is the effect of temperature on the spin configuration. At room temperature, MnO is paramagnetic (PM) and adopts a cubic structure with lattice constant (a) of 4.445 Å; but MnO has a Neél temperature (TN) of 118 K, below which it adopts an antiferromagnetic spin-configuration (AFM) with the δ valence electrons alternating in their spin alignment between nearest neighbor Mn²⁺ cations. In the case of MnO, the lowest energy AFM arrangement is with the spin aligned in a principal diagonal plane through the crystal (AFM(II)), rather than in the lowest-index (100), (010) or (001) plane (AFM(I)). Calculations have confirmed this energetic ordering of spin configurations, with the ferromagnetic (FM) arrangement shown as higher in energy than either AFM option.

The AFM(II) spin-configuration is associated with a rhombohedral distortion to the unit cell: the experimental lattice constant, a, is 4.43015–4.4316 Å, and the unit cell angle, α, is 90.60–90.62°. The AFM(II) arrangement mimics the attractive exchange interactions between the nearest and next-nearest neighbor Mn²⁺−Mn²⁺−Mn²⁺−Mn²⁺ Coulomb (\( E_{\text{Coul}} \)) and Buckingham (\( E_{\text{Buck}} \)) terms, defined as:

\[
E_{\text{Coul}} = \frac{q_i q_j}{r_{ij}}
\]

\[
E_{\text{Buck}} = A e^{-\rho/r_{ij}} - \frac{C}{r_{ij}^6}
\]

between two centers with indices \( i \) and \( j \), separated by a distance \( r_{ij} \). The FF parameters are the charge (\( q_i \), in e), the spring constant (\( k_s \), in eV Å⁻²), and the Buckingham potential parameters \( A \) (eV), \( \rho \) (Å) and \( C \) (Å⁶ eV); and \( k_s \) is Coloumb’s constant. Our work builds on the FF parametrization of Lewis and Catlow (Table 1), which gives an MnO lattice parameter (a) of 4.44 Å that matches experiment (4.45 Å) for the room temperature, cubic, PM MnO. However, this FF does not reproduce the low-temperature AFM(II) rhombohedral unit cell, which is required to prevent unphysical strain occurring from a structural match between the QM and MM regions at the QM/MM interface. As a remedy, a Lennard-Jones potential (\( E_{\text{LJ}}^{\text{MM}} \)), taking the form of:

\[
E_{\text{LJ}}^{\text{MM}} = -\frac{\sigma}{r_{ij}^{\sigma}}
\]

was introduced between nearest neighbor Mn²⁺ cations with opposing spin by using a short-range cutoff of 5 Å to include nearest and next-nearest neighbor Mn–Mn interactions. This implementation mimics the attractive exchange interactions between antiparallel spins, with the Buckingham potential continuing to provide short-range repulsion. The \( C_6 \) parameter,
Table 1. Force Field Parameters for the MnO Calculations.64

<table>
<thead>
<tr>
<th></th>
<th>( \eta_{\text{core}} )</th>
<th>( \eta_{\text{shell}} )</th>
<th>( k_2 )</th>
</tr>
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<tbody>
<tr>
<td>Mn</td>
<td>-1.420</td>
<td>3.420</td>
<td>95.00</td>
</tr>
<tr>
<td>O</td>
<td>1.000</td>
<td>-3.000</td>
<td>54.76</td>
</tr>
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</table>

Buck.     \( A \) \( \rho \) \( C \) \( r_c \)
Mn\text{shell} - O\text{shell} \quad 1007.4 \quad 0.3262 \quad 0.0 \quad 12.0
O\text{shell} - O\text{shell} \quad 22764.3 \quad 0.1490 \quad 23.0 \quad 12.0

LJ   \( \alpha \) \( C_6 \) \( r_c \)
Mn\text{core} - O\text{shell} \quad 52.5 \quad 5.0

\( \text{Coul.} \), \( \text{Buck.} \), and \( \text{LJ} \) are abbreviations/acronyms for Coulomb, Buckingham and Lennard-Jones, respectively. The parameters are the core charge \( (\eta_{\text{core}} \text{ in e}) \), the shell charge \( (\eta_{\text{shell}} \text{ in e}) \), the spring constants \( (k_2 \text{ in eV Å}^{-2}) \), the Buckingham potential parameters \( A \) (in eV), \( \rho \) (in Å), \( C \), the Lennard-Jones \( C_6 \) parameter, in units of \( \text{Å}^6 \text{eV} \), and the cutoff for the different potentials \( (r_c) \) in Å.

which has units of \( \text{Å}^6 \text{eV} \), was fitted in order to reproduce the low-temperature structure of the AFM(II) MnO phase \((a = 4.43 \, \text{Å}, \alpha = 90.6^\circ)\).45-47

For the QM calculations, hybrid density functional theory (hybrid-DFT) was used with the B3LYP65-67 and B97-368 exchange-correlation (XC) functionals. Results obtained with B3LYP provide a direct comparison with previous work on MgO, while B97-3 is a more advanced XC functional that has been fitted to a more extensive thermochemical data set. A Def2 triple-\( \zeta \) valence plus polarization (TZVP) basis set was used for oxygen and carbon,69 and a Stuttgart/Dresden relativistic small-core effective core potential (ECP), with modified basis sets, was used for Mn:30 the most diffuse functions were removed to prevent artificial spread of electron density outside the QM region (edited basis set provided in the Supporting Information).35,55 At the boundary of the QM region, it is typical to have cation-centered large-core ECPs to contain the electron density; however, no such ECPs have been derived for Mn.59 Given the similar ionic radii of Mg2+ and Mn2+, it was decided that large-core Mg2+ ECPs would be placed at the QM/MM boundary, with such an approach proving successful in a previous investigation of optical transitions for cubic-structured, FM- and AFM-ordered, MnO.25

3. BULK

3.1. Pristine. A bulk model of MnO was cut from the FF-optimized AFM(II) crystal structure. The QM/MM cluster was centered on the atomic species of interest: the O-centered model contains 14 Mn and 19 O atoms (Figure 1), whereas the Mn-centered model contains 14 O and 19 Mn atoms. Previous work has shown a model of this size is optimal as it ensures the QM region is large enough not to exhibit quantum confinement effects, which may occur due to the localized nature of the occupied states, while also not being excessively large so as to make calculations prohibitively expensive.33,55,56 We also note here that stoichiometry is not a requirement for the QM region in the QM/MM embedded-cluster approach.55-60 After geometry optimization, the AFM(II) electronic structure was preserved in both models, with Mulliken analysis showing an average spin density of 4.8 e on each Mn2+ cation when using either XC functional, in agreement with experiment;11 however, the geometric structure is noted as being slightly distorted. For both the B3LYP and B97-3 O-centered models of pristine MnO, the mean bond distances in the QM region stay constant at 2.21 Å but in a range of 2.16−2.24 Å, with bond lengths longer in the center of the QM region. Mean bond distances across the QM/MM boundary contracted (expanded) slightly for O\text{QM}−Mn\text{MM} (Mn\text{QM}−O\text{MM}) interactions, with averages of 2.18 and 2.24 ± 0.01 Å, respectively. Central bond angles were unaltered (\( \alpha = 90.6^\circ \)). These results were compared with geometry optimization of a range of QM/MM models with 7, 19, 27, 33, 45, 57, 63, 75, and 81 atoms in the QM region. It is noted that average bond length within the QM region and across the QM/MM boundary remains relatively constant (±0.01 Å), with the exception of the Mn\text{QM}−O\text{MM} interaction for which 2.24 Å is large compared to the more common observation of 2.21−2.22 Å. Interestingly, the range of bond lengths for those intersecting the QM/MM boundary (and in the MM region) remains stable, whereas within the QM region the range increases up to 2.24 − 2.27 Å for the 81-atom QM region, which also results in significant distortion of the bond angles. For the Mn-centered clusters, the distances between central and nearest neighbor atoms also increased, up to 2.25 Å; however, bond lengths in the QM region toward the boundary contracted significantly (2.15 Å). This contraction

![Figure 1. Structures of the QM region in the QM/MM models of the B3LYP-optimized (a) bulk and (b) surface models, as viewed with a cutplane through the \( x \)- and \( z \)-axes. Purple and gray spheres represent Mn atoms with spin-up and spin-down configurations, respectively, and red spheres represent oxygen atoms. Bond lengths are given in Å, and the denoted angles \( \alpha \), \( \beta \), and \( \gamma \) are 90.6, 90.6, and 90.4°, respectively.](image-url)
led to inversion of the bond angle in some cases, with variation between 89.0 and 90.7°. While these observations are indicative of a slight force mismatch between the QM and MM regions, the structural changes do not adversely affect energetics, as discussed subsequently in our analysis.

The energetic position of the valence band maximum (VBM) can be calculated by simulating the ionization of an electron from the bulk environment in to vacuum, as

$$\text{MnO}^0 \rightarrow \text{MnO}^+ + e^-$$  \hspace{1cm} (5)

The energy of an electron far removed from the bulk (i.e., at the vacuum level) is 0 eV, and so the vertical ionization energy ($I_v$) is

$$I_v = E(\text{MnO}^+) - E(\text{MnO}^0)$$  \hspace{1cm} (6)

As well as $I_v$, which represents the instantaneous process of removing an electron, one can subsequently calculate the stabilization of a hole state as the adiabatic ionization ($I_a$). For both $I_v$ and $I_a$, the starting structure is that of optimized MnO$^0$, the nuclear coordinates are frozen for $I_v$, allowing only the electronic configuration (and MM shells) to optimize, making this the appropriate method to place the VBM, whereas the nuclear centers are free to move for the $I_a$, the energy of which could be compared to emission spectra. We note that for $E(\text{MnO}^+)$, and any other charged QM/MM cluster, the *a posteriori* long-range dielectric correction of Jost is included in the total energy to compensate for the finite size of the polarizable region in our QM/MM model.$^{33,72,73}$

For the O-centered clusters, which we will focus on throughout our work, $I_v$ and $I_a$ are 4.79 (5.14) eV and 3.46 (3.68) eV, respectively, when using the B3LYP (B97-3) XC functional; for the Mn-centered clusters, $I_v$ is 4.76 (5.07) eV for the same XC functionals. The similarity in $I_v$ between our two model configurations gives us significant confidence in the ability to compare the results of our calculations. To further clarify the error bar in our approach, we also calculated $I_v$ for the O-centered cluster when varying the size of the QM region between 7, 19, 27, 33, 45, 57, and 63 atoms. The B3LYP calculated values, which were respectively 5.43, 4.63, 4.85, 4.79, 4.76, 4.75, and 4.77 eV, vary significantly at small sizes, due to quantum confinement effects, but converge to within 0.04 eV for sizes of 33 atoms and greater. This similarity is particularly noteworthy given the structural differences of the respective QM regions, as highlighted above, when changing the QM region size. Overall, we can combine the highlighted differences to present a conservative compound error bar of ±0.1 eV on our results. While previous calculations of $I_v$ for bulk MnO are unavailable, calculated position of the VBM at surfaces has been presented in the range of 4.5–5.5 eV (discussed further in section 4).$^{12,13}$ However, we also note that the only available experimental data, taken from photoelectrochemical measurements of PM MnO, shows $I_v$ to be considerably greater as 7.09 eV.$^{7,34}$ This discrepancy between experiment and theory could be related to the differences in magnetic and structural configuration, though neither effect results typically in such significant deviations; therefore, the difference is perhaps more likely due to experimental error and/or DFT methodological choices. Tests show that the ionization potential of gas-phase Mn atoms match experiment (7.43 eV in experiment, 7.64 eV with B3LYP), but ligand-field interactions result in splitting of degenerate levels within the bulk materials: an underestimation of $I_v$ with DFT XC functionals has been previously observed for bulk materials and should be remembered when considering the position of both conduction band (CB) and valence band (VB) edges.$^{35}$

Previous literature tells us the VBM is of mixed nature in AFM(II) MnO, with PBE0 giving a 25/75% split between O 2p- and Mn 3d-states, respectively$^{36}$ in our calculations, Mulliken analysis on both the O-centered and Mn-centered models shows that the electron is removed from the Mn valence shells when forming Mn$^{+}$, hence the good agreement between the two models, and the remaining hole localizes on an Mn atom during subsequent geometry relaxation.

The discrete nature of the QM region in the QM/MM cluster, and the resulting quantum confinement, results in a poor representation of the CB and an overestimation of the band gap, $E_{g}$; therefore, bulk $E_{g}$ has been taken from literature to position the conduction band minimum (CBM). Experimental reports of $E_{g}$ are for the room temperature, PM ordering, and range between 3.6 and 4.9 eV;$^{76,77}$ calculated band gaps of 2.9–4.02 eV and 3.5–4.2 eV have been presented for AFM(II) MnO from hybrid-DFT$^{40–42}$ and GW$^{12,79,80}$ calculations. In this work, we have taken the result of $E_{g}$ for the B3LYP XC functional (3.92 eV) as this exactly matches one of our XC functionals; thus, we position the CBM of AFM(II) MnO using the $I_v$ from our O-centered clusters, obtaining −0.87 (−1.22) eV relative to the vacuum level when using the B3LYP (B97-3) XC functionals.

### 3.2. Intrinsic Defects

In order to illustrate the capabilities of the embedded-cluster QM/MM approach, we have calculated the defect formation energy for neutral and charged vacancies in MnO, as well as the energy levels of trapped electrons/holes at the remaining vacancies. A Kröger–Vink defect notation is used. A neutral cation vacancy may be considered as forming thus

$$\text{Mn}^{x+}_{\text{Mn}} \rightarrow V^{x-}_{\text{Mn}} + \text{Mn}^{x+}_{\text{ref}}$$  \hspace{1cm} (7)

with a formation energy of

$$\Delta H_f(V^{x-}_{\text{Mn}}) = E(V^{x-}_{\text{Mn}}) + E(\text{Mn}^{x+}_{\text{ref}}) - E(\text{Mn}^{x+}_{\text{Mn}})$$  \hspace{1cm} (8)

The energy of the reference states, $E(\text{Mn}^{x+}_{\text{ref}})$, is defined as $E(\text{Mn}^{x+}_{\text{ref}}) + \mu(\text{Mn})$, where $E(\text{Mn}^{x+}_{\text{ref}})$ is the calculated energy of the gas-phase atom and $\mu(\text{Mn})$ is the chemical potential of Mn. Here, $E(\text{Mn}^{x+}_{\text{ref}})$ was calculated as $E(\text{Mn}^{x+}_{\text{ref}})$ minus the first and second experimental ionization energies, as successfully applied previously for embedded-cluster calculations with modified basis sets.$^{3,34}$ Ideally, $\mu(\text{Mn})$ would also be calculated at the same level of theory as our QM/MM calculations of MnO, which in turn requires calculating $\Delta H_f(\text{Mn})$, i.e., the energy of bulk Mn; however, due to the weakness in hybrid-DFTs at representing delocalized states in metallic systems, we found $\Delta H_f(\text{Mn})$ to be underestimated by >1 eV when using hybrid-DFT (see the Supporting Information). This inability to accurately model transition metals is not isolated to our particular choices of hybrid DFT exchange-correlation functionals, but a more general problem for the field as highlighted by Janthon et al.$^{2,24}$ Thus, in order to minimize the impact of these inaccurate hybrid-DFT models for bulk Mn, the experimental $\Delta H_f$ was used for the reference states of Mn and MnO, and only $\Delta H_f(\text{O}_2)$ was calculated to acquire $\mu(\text{Mn})$; from this, the respective products in high- and low-partial pressures of O$_2$, $p(\text{O}_2)$, are MnO($\Delta H_f(\text{MnO}) - \Delta H_f(\text{O}_2)) = −126.3$ kJ mol$^{-1}$ and Mn($\Delta H_f(\text{Mn}) = −279.58$ kJ mol$^{-1}$).$^{57,66}$ For the equivalent calculation of O$_2$ vacancies in high- and low-$p(\text{O}_2)$, the reference states are O$_2$
\( \frac{1}{2} \Delta H_f (O_2) = -258.90 \ \text{kJ mol}^{-1} \) using B3LYP and \(-261.67 \ \text{kJ mol}^{-1} \) using B97-3) and MnO (s) \( [\Delta H_f (\text{MnO}) - \Delta H_f (\text{Mn}) = -105.62 \ \text{kJ mol}^{-1} ]^{\text{ref}} \) respectively. If a given vacancy forms in a charged state, the energy associated with adding/removing an electron must be included. The Fermi Level (\( \epsilon_f \)) is a common source of electrons and holes in traditional semiconductors, depending on its positioning, and its inclusion in our calculations makes \( \Delta H_f \) equal to:

\[
\Delta H_f (V_{\text{Mn}}^{\text{m}}) = E(V_{\text{Mn}}^{\text{m}}) + E(\text{Mn}_{\text{ref}}^x) - E(\text{Mn}_{\text{ref}}^{x+}) - q \epsilon_f
\]

(9)

Both positive and negative values of the charge, \( q \), are feasible: in the case of \( q < 0 \), electrons are taken from the Fermi Level and thus \(-q \epsilon_f\) is positive to reflect the energy required for this abstraction; in the case of \( q > 0 \), electrons are transferred to the Fermi level; thus, there is an energy gain. For an oxygen vacancy, the enthalpy of formation is similarly:

\[
\Delta H_f = E(V_{\text{O}}^{\text{m}}) + E(\text{O}_{\text{ref}}^x) - E(\text{O}_{\text{ref}}^{x+}) - q \epsilon_f
\]

(10)

An alternative charge compensation is the formation of Schottky defect pairs, which are prevalent in wide band gap materials, whereby the charge on a cation vacancy is compensated by an opposing charge on an anion vacancy. For a pair of double-charged vacancies, this is:

\[
\text{Mn}_{\text{Mn}}^x + \text{O}_O^* \rightarrow V_{\text{Mn}}^{\ast} + \text{Mn}_{\text{Mn}}^{2+} + V_{\text{O}}^{\ast} + \text{O}_O^* \]

(11)

\[
V_{\text{Mn}}^{\ast} + V_{\text{O}}^{\ast} \rightarrow V_{\text{Mn}}^{\ast} + V_{\text{O}}^{\ast} + \text{MnO}(s)
\]

(12)

with the recombination of the charged ions in eq 11 leading to the reformation of MnO(s) in eq 12. In this case, \( \Delta H_f \) for the Schottky defect is:

\[
\Delta H_f (V_{\text{Mn}}^{\ast} + V_{\text{O}}^{\ast}) = E(V_{\text{Mn}}^{\ast}) + E(V_{\text{O}}^{\ast}) + E[\text{MnO}(s)] \\
- E(\text{Mn}_{\text{Mn}}^{x+}) - E(\text{O}_O^*)
\]

(13)

where we calculate \( E[\text{MnO}(s)] \) as:

\[
E[\text{MnO}(s)] = E(\text{Mn}_{\text{Mn}}^x) + \Delta H_f (\text{Mn}) + E(\text{O}_O^*)
\]

\[
+ \frac{1}{2} \Delta H_f (\text{O}_O) + \Delta H_f [\text{MnO}(s)]
\]

(14)

B3LYP-calculated defect formation energies are presented in Figure 2 for the atomic vacancies. For O vacancies, calculated with the O-centered model, \( V_{\text{O}}^{\ast} \) is the most stable at low \( \epsilon_f \) with the neutral \( V_{\text{O}}^{\ast} \) defect preferred as \( \epsilon_f \) tends toward \( \epsilon_{\text{CBM}} \); however, the \( V_{\text{O}} \) defects are only more favorable than \( V_{\text{Mn}}^{\ast} \) under high-\( p(\text{O}_2) \) when \( \epsilon_f \) is in the lower half of the band gap. The neutral and negatively charged \( V_{\text{O}}^{\ast} \) defects show little disruption of the perfect bulk structure, with minimal atomic displacements. Mulliken analysis shows the two electrons (\( e^\ast \)) freed when forming the \( V_{\text{O}}^{\ast} \) defect would be trapped in the vacancy, forming an F-center, with an F"-center also stable when one of the defect electrons has been removed i.e., a \( V_{\text{O}}^{\ast} \) defect. Unsurprisingly, negatively charged \( V_{\text{O}}^{\ast} \) sites (\( V_{\text{O}}^{\ast} \) and \( V_{\text{O}}^{\ast} \)) are unstable at all values of \( \epsilon_f \).

For Mn vacancies, calculated with the Mn-centered model, \( \Delta H_f \) decreases with \( q \) when \( \epsilon_f = \epsilon_{\text{CBM}} \) and increases with \( q \) when \( \epsilon_f = \epsilon_{\text{VBM}} \); this shows that the defects would donate electrons when \( \epsilon_f \) is low (large energy gain) and accept electrons when \( \epsilon_f \) is high (small energy cost). For the B3LYP XC functional, we calculate that \( \Delta H_f (V_{\text{Mn}}^{\ast}) \) is exothermic when \( \epsilon_f = \epsilon_{\text{CBM}} \) at both low- and high-\( p(\text{O}_2) \); we note that \( \epsilon \)-type conditions of \( \Delta H_f \) would require an excess of electrons generated from e.g. oxygen vacancies. In previous literature, \( \epsilon \)-type conditions have more commonly been considered: Aschauer et al. presented results from periodic PBESol+U calculations with \( \epsilon_f \) positioned 2.8 eV above the VBM, i.e. toward the CBM, and found that \( \Delta H_f (V_{\text{Mn}}^{\ast}) \) was 1.5 eV lower than \( \Delta H_f (V_{\text{Mn}}^{\ast}) \), which in turn was 1.7 eV more preferable than \( \Delta H_f (V_{\text{Mn}}^{\ast}) \); they also reported \( \Delta H_f (V_{\text{Mn}}^{\ast}) \) as 4.5 eV higher in energy than \( \Delta H_f (V_{\text{Mn}}^{\ast}) \). These observations are in agreement with our results, with the same \( \epsilon_f \) positioning giving respective differences between \( \Delta H_f (V_{\text{Mn}}^{\ast}) \), \( \Delta H_f (V_{\text{Mn}}^{\ast}) \), and \( \Delta H_f (V_{\text{Mn}}^{\ast}) \) of 1.27 and 1.13 eV (Figure 1, Supporting Information), and between \( \Delta H_f (V_{\text{Mn}}^{\ast}) \) and \( \Delta H_f (V_{\text{Mn}}^{\ast}) \) of 4.26 (7.45) eV for high (low)-\( p(\text{O}_2) \). Previous experimental work focuses on the dominant Mn vacancy at varying high temperatures \((T > 1000 \, \text{K})\), the modeling of which is beyond the scope of this work, but generally notes the prevalence of \( \Delta H_f \).

As the VBM is lower at \( \epsilon_f \) and \( \epsilon_{\text{VBM}} \), the modeling of which is beyond the scope of this work, but generally notes the prevalence of \( \Delta H_f \) when \( \epsilon_f = \epsilon_{\text{VBM}} \).
in the exothermic regime, the material properties would change due to the propensity for defect formation. Thermodynamic stability regimes for the singly charged \( V_{\text{Mn}}^\bullet \) and \( V_{\text{Mn}}^\bullet \) defects are not observed for either B3LYP or B97-3.

Structurally, \( V_{\text{Mn}}^\bullet \) formation results in expansion around the vacant site, with contraction of Mn–O bonds between neighboring atoms to compensate for the lost coordination. For the B3LYP optimized model, Mn–O bonds around the \( V_{\text{Mn}}^\bullet \) defect are 2.12–2.14 Å, in agreement with Aschauer et al., who report bond lengths of 2.11 Å,85 while we observe shorter bonds of ~2.01 Å for bonds between O atoms neighboring the defect and Mn atoms at the edge of the QM region. As the charge of QM region is reduced (i.e., electrons are removed), much shorter Mn–O bonds are observed: for \( V_{\text{Mn}}^\bullet \), there is a shorter bond of 1.98 Å; for \( V_{\text{Mn}}^\bullet \), two bonds of 1.81 and 1.88 Å are observed; for \( V_{\text{Mn}}^\bullet \), there is an additional bond of 1.98 Å; and for \( V_{\text{Mn}}^\bullet \), four bonds of between 1.81 and 1.88 Å are calculated. These shorter bonds are all indicative of Mn \( \text{Mn}^\bullet \) cations (\( \text{MnMn}^\bullet \)), with contraction of Mn\( ^\bullet \), with investigations having been considered cubic, paramagnetic MnO.35,88,89 O’Keeffe et al. stated that the majority defects in their samples were \( V_{\text{Mn}}^\bullet \) with charge compensation via two \( \text{MnMn}^\bullet \) centers and \( 1/2 \text{O}_2(g) \).85 In our calculations, Mulliken analysis shows that multiple \( \text{MnMn}^\bullet \) defects are the preferred stabilizer for all \( V_{\text{Mn}}^\bullet \) defects, but the formation enthalpies imply that \( V_{\text{Mn}}^\bullet \) would only form with \( \epsilon^p \) positioned midgap or higher (Figure 2). Our results for low-

\[
\Delta H_f (\text{eV})
\]

point. As shown in eq 13, Schottky defects are mutually cooperative as holes and electrons recombine. Our results show that a Schottky defect with Mn\( ^\bullet \)/O\( ^\bullet \)– pairings are energetically more favorable. By summing \( \Delta H_f \) for the charge-carrier-compensated vacancies and comparing to the formation energy of Schottky defects, we also see that, when \( \epsilon^p \) is positioned midgap or higher (i.e., n-type), this Schottky defect is more favorable than all possible combinations of charge-carrier-compensated defects; however, in general the individual \( \epsilon^p \)-compensated defects remain energetically more preferable.

<table>
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<th>defect reactions</th>
<th>( \Delta H_f (\text{eV}) )</th>
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<tbody>
<tr>
<td>( \text{O}<em>2^- \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>4.44 (3.81)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>11.25 (12.04)</td>
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<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>8.48 (8.70)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>5.88 (5.58)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>4.99 (4.54)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>4.44 (3.81)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>1.95 (2.76)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>2.38 (2.92)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>2.67 (2.74)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>3.35 (3.10)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>3.81 (3.11)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>3.54 (4.35)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>3.97 (4.51)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>4.26 (4.33)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>4.94 (4.69)</td>
</tr>
<tr>
<td>( \text{V}<em>{\text{Mn}}^\bullet \rightarrow \text{V}</em>{\text{Mn}}^\bullet + 1/2\text{O}_2(g) )</td>
<td>5.40 (4.69)</td>
</tr>
</tbody>
</table>

As an additional note, we see in Table 2 that the most stable \( V_{\text{Mn}}^\bullet \) defects differ between the B3LYP and B97-3 XC functionals at the midgap point. This difference occurs because of the larger \( I_c \) when using B97-3 (5.14 eV) compared to B3LYP (4.79 eV), which in turn lowers the energy of the midgap point and favors the loss of e\( ^p \) charge carriers more when using B97-3 than with B3LYP.

The energy levels of electrons trapped at defect sites are presented in Figure 3. Here, the \( \Delta\text{SCF} \) method (analogous to
eq 6) was used to calculate the energy levels when coupled with values of \( I_a \) and \( E_g \) one can determine the position of defects relative to the VBM and CBM. Starting with the Mn vacancies, the sites with two and one associated electrons (\( V_{Mn}^{2} \) and \( V_{Mn}^{+} \) respectively) are mid band gap, as may be expected from the behavior of cation defects in similar materials e.g. \( V_{Mg} \) in MgO and \( V_{Zn} \) in ZnO. However, unlike \( V_{Zn} \) neutral and positive \( V_{Mn} \) defects are positioned at the VBM. In the case of both \( V_{Mn}^{2} \) and \( V_{Mn}^{+} \), it is noted that the electron energy levels are higher than defects with an extra electron, further consolidating our observation that these defects are unstable; similar results are observed for the B97-3, as shown in the Supporting Information, in agreement with our observations from Table 2. Such results are not observed for \( V_{O} \), where \( V_{O} \) and \( V_{O}^{=} \) are positioned midgap, similar to the negatively charged Mn vacancies and the equivalent sites that have been well studied in MgO. Negatively charged \( V_{O} \) sites are positioned very close to the CBM and are not expected to be stable.

4. (100) SURFACE

4.1. Pristine. The (100) surface is the most stable for MnO, as with many other rock salt oxides, and it is considered here for reactivity investigations. A surface model was prepared at the MM level, using the AFM(II) configuration, with minor rumpling effects occurring such that the O atoms displaced 0.01 Å outward (0.5% change in bond lengths), relative to the Mn atoms, and the distance between the top-surface layer and subsurface layer contracted to 2.20 Å (0.5% change in interlayer distance). The results for rumpling are mixed in the previous literature, with MEIS and LEED giving inward and outward O displacement, respectively; likewise, computational results for the LDA XC functional give retraction of O from the surface, while PBE and PBE+U give outward O displacement by 1.3–1.6%. The results for interlayer contraction are more consistent within each method, yet contradictory between theory and experiment: MEIS and LEED indicate that the distance between surface and subsurface layers increases, by 0.1–1.1%, whereas DFT gives interlayer contraction by 0.3–1.6%. Our calculated displacements match previous computations well, and we note that the error bars on experiment are large compared to the values being considered.

A QM/MM model was constructed from our MM surface, centered on a surface oxygen atom; structural optimization resulted in atom displacements of only 0.01 Å (0.5%) compared to the MM model. The surface ionization potential was then calculated using eq 6, with both high- and low-spin configurations considered for the MnO surface: a high-spin configuration corresponds to ionization from the 2p-orbitals, whereas the low-spin configuration corresponds to removing an electron from the Mn 3d-orbitals. Our calculations show that the highest occupied states (VBM) belong to the Mn 3d-orbitals, for which \( I_a \) is 4.77 (5.09) eV and \( I_a \) is 3.32 (3.52) eV when using the B3LYP (B97-3) XC functional, whereas the O 2p-states are marginally lower in energy with \( I_a \) = 4.95 (5.29) eV, for the same XC functionals. We note that our results for \( I_a \) and \( I_a \) are similar to bulk: for the favorable low spin ionization of the surface, \( I_a \) differs by <0.1 eV and \( I_a \) by <0.2 eV. Mulliken analysis shows that this similarity is because surface ionization removes an electron from a subsurface, rather than surface, Mn cation. Further calculations using constrained DFT, whereby the ionized electron is removed specifically from a surface atom, resulted in \( I_a \) = 5.06 and 5.43 eV for B3LYP and B97-3 XC functionals, respectively, which indicates the surface valence-band bending downward.

Motivated by these results, we also calculated second and third ionization energies for the MnO surface. It was noted that the lowest energy configurations for the vertical ionization of a second electron is with two Mn\(^{3+}\) cations positioned subsurface, forming a [Mn\(^{3+}\)]\(_n\) complex, which is more stable than a Mn\(^{4+}\) positioned subsurface or on the surface; the respective energetic ordering is 5.65 (5.97) < 6.27 (6.72) < 6.45 (6.91) eV when using the B3LYP (B97-3) XC functional. The [Mn\(^{3+}\)]\(_n\) configuration was also more favorable for \( I_a \). Analogous to eq 9, we then looked at formation energies of the ionized surface defects, finding that Mn\(^{3+}\) through to Mn\(^{4+}\), and combinations therein, would be exothermic as \( \epsilon_f \) approaches the VBM; isolated occurrences of Mn\(^{3+}\) are most favorable when considered per hole created: \(-1.46 \pm 1.57\), \(-1.20 \pm 1.32\), and \(-0.97 \pm 1.22\) eV for single, double and triple Mn\(^{3+}\) defect coupling, when using the B3LYP (B97-3) XC functional. Again, we note that negative \( \Delta H \) values indicate a physically unstable \( \epsilon_p \) as the material would spontaneously lose electrons; however, we emphasize the trend in stability of different defects. Compensating Mn vacancies were also considered, but found to be higher in energy than compensation via \( \epsilon_p \).

Our positioning of the VBM at the (100) surface is in good agreement with previous computations by Toroker et al. using hybrid-DFT methods, where a periodic slab model was used with a vacuum region and the electrostatic potential in the vacuum gap taken as the reference vacuum level. With this model, they obtained \( \epsilon_{VBM} \) of \(-4.48 \pm 4.96\), \(-5.28\), and \(-5.57\) eV for the PBE+U, HSE, PBE0, and G0W0 (DFT+U) methods. Stevanovic et al. used a similar technique to obtain \( I_a \) = 5.08 eV. However, neither our nor other authors’ results match with the experimental estimates of \(-7.5 \pm 0.8\) eV, taken from photoelectron spectroscopy (PES) measurements by Fujimori et al., though these relate to the room temperature.
PM phase. A position of the VB can also be implied from the PES-BIS (bremsstrahlung isochromate spectroscopy) measurements of van Elp et al., where the CBM was reported as being $-3.6 \pm 1.2$ eV. This again positions the VBM well below our estimates. The offset between experimental and computational results is worth remembering when interpreting our defect calculations: a lower value of $E_{\text{VBM}}$ stabilizes electrons, which will shift the equilibrium between $e^-\text{compensated charged defects such that positively charged defects will be more favored.}$

### 4.2. Intrinsic Defects

The formation energies of neutral and positively charged oxygen surface vacancies are presented in Figure 4. $\Delta H_f (V_O^{0+})$ is almost identical to the bulk at 5.88 eV for the B3LYP (B97-3) XC functional, differing by only 0.02 eV. This difference relative to bulk increases $\sim 0.2$ eV for $V_O^{0+}$ with $\Delta H_f (V_O^{0+}) = 2.81$ (2.37) eV at the surface when considering $\epsilon_f = E_{\text{VBM}}$ for $V_O^{0+}$, $\Delta H_f (V_O^{0+})$ is then almost identical for the surface and bulk, at 0.63 (−0.04) eV for the B3LYP (B97-3) XC functional. Analysis of the defect levels shows that, like the bulk, the favored spin configuration when ionizing to the $V_O^{0}$ states is low-spin (LS): $I_f^L (V_O^{0} \rightarrow V_O^{0+}) = 2.84$ eV compared to $I_f^H (V_O^{0} \rightarrow V_O^{0+}) = 3.13$ eV for high-spin (HS) configurations, using the B3LYP XC functional. Also similar to the bulk, the structural integrity of the void is preserved, with neighboring bond distances remaining close to the bulk values. However, removing a second electron from the oxygen vacancy is more difficult: $I_f (V_O^{0} \rightarrow V_O^{0+})$ and $I_f (V_O^{0} \rightarrow V_O^{0+})$ are $4.06$ (4.22) and $2.60$ (2.73) eV, respectively, with the B3LYP (B97-3) XC functional, which is $\sim 0.6$ eV larger than for the bulk. The structure of the defect remains relatively unchanged.

### 4.3. CO$_2$ Adsorption

Rock salt oxides have recently been proposed as candidate materials for the catalytic transformation of CO$_2$\textsuperscript{10,11,14} and here we investigate the initial step of that reaction: CO$_2$ surface adsorption. Previously, it has been shown that surface defects on MgO, specifically oxygen vacancies, can affect CO$_2$ adsorbate binding\textsuperscript{33} both structurally and energetically, which is further amplified by the presence of extrinsic cation dopants like Mn\textsubscript{Mg}.\textsuperscript{35} A favorable initial step involves charge transfer from the support to the CO$_2$, which results in bending of the linear CO$_2$, and this can be aided by high-energy defect levels associated with oxygen vacancies. There are two possible orientations for a CO$_2$ molecule on the (100) surface: parallel or perpendicular to the surface x-y plane.\textsuperscript{55} Downing et al. showed that, for pure MgO, parallel adsorption is stronger over the pristine surface, whereas perpendicular orientation is more favorable in the presence of $V_O$. In the case of MnO, parallel adsorption is again stronger over the pristine surface than perpendicular, with the C atom interacting with a surface oxygen (Figure 5) with an adsorption energy of $-0.64$ eV (Table 3). This result is very similar to previous work on MgO ($-0.68$ eV) and Mn-doped MgO ($-0.66$ to $-0.79$ eV).\textsuperscript{35} For $V_O$, the energetics of CO$_2$ adsorption in parallel or perpendicular configurations are similar (2.60 eV), with structural analysis showing that, in both cases, the CO$_2$ adopts a bent structure coordinated with both the defect and the surface atoms (Figure 5). In particular, one oxygen from the CO$_2$ fills the void while the second oxygen coordinates with a surface Mn, resulting in $\angle$(OCO) of 115.5°. This angular distortion prevents the adsorption energy from being as strong as for CO$_2$ z-aligned over MgO.\textsuperscript{55}

As we remove an electron from $V_O^{0}$ to form $V_O^{0+}$, the interaction between CO$_2$ and MnO weakens yet the adsorbate is still strongly coordinated with the surface, and the internal angle of $\angle$(OCO) is 119.7°. The maintenance of an adsorbed structure, yet weakening of the adsorption energy, shows how
This case, formation of VO defects on MgO and Mnsurf would be most beneficial.

### ASSOCIATED CONTENT

**Supporting Information**

The Supporting Information is available free of charge on the ACS Publications website at DOI: 10.1021/acs.jpcc.8b07846.

Edited basis set for an Mn atom, as used in our calculations, documentation of periodic calculations for the formation enthalpies of MnO, figures illustrating defect formation enthalpies, and a pictorial representation of the defect levels for neutral and charged defects in MnO when using the B97-3 exchange-correlation functional (PDF)

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