

Dynamic Atomic Reconstruction: How Fe₃O₄ Thin Films Evade Polar Catastrophe for Epitaxy

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Polar catastrophe at the interface of oxide materials with strongly correlated electrons has triggered a flurry of new research activities. The expectations are that the design of such advanced interfaces will become a powerful route to engineer devices with novel functionalities. Here, we investigate the initial stages of growth and the electronic structure of the spintronic Fe₃O₄/MgO (001) interface. Using soft x-ray absorption spectroscopy, we have discovered that the so-called A-sites are completely missing in the first Fe₃O₄ monolayer. This discovery allows us to develop an unexpected but elegant growth principle in which, during deposition, the Fe atoms are constantly on the move to solve the divergent electrostatic potential problem, thereby ensuring epitaxy and stoichiometry at the same time. This growth principle provides a new perspective for the design of interfaces.

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Physical properties of surfaces and interfaces of solids could markedly differ from those in the bulk, especially in cases when the surface or interface involves non-neutral crystal planes. For insulators, these polar planes cause the electrostatic potential to diverge and thus to destabilize the system dramatically. The surface or interface is then forced to reconstruct, for example, by forming facets or defects. Or more spectacularly, it could give up a substantial amount of charge, thereby altering its electronic structure completely [1–3]. Recently, claims have been made that this happens at the SrTiO₃/LaAlO₃ interface since the interface is conducting while the two constituents are good insulators separately [4,5]. In fact, the formation of a two-dimensional electron gas and the occurrence of superconductivity in

this interface have generated frantic efforts worldwide to explore the potential of these interfaces for device applications [6–12], as well as to search for new emergent phenomena such as quantum criticality in two-dimensional electron gas systems [13–15]. However, the growth process of these interfaces and polar interfaces, in general, is a mystery. How do the atoms rearrange themselves during the deposition or growth such that a well-ordered and smooth interface is formed, despite the destabilizing forces due to the catastrophic electrostatic potential? Understanding the growth principles will widen the scope of interfaces that one may want to design: Interfaces that appear impossible to grow at first sight may now be tried.

Here, we investigate the polar interface between Fe₃O₄ and the MgO (001) substrate, one of the most used interfaces in the research field of spintronics [16–24]. This interface is completely not understood in terms of atomic structure, electronic structure, and growth mode. Figures 1(a) and 1(b) illustrate how the Fe₃O₄ inverse spinel crystal structure, consisting of Fe³⁺ ions in tetrahedral coordination (A-sites), and Fe²⁺ and Fe³⁺ ions (or, on average, Fe^{2.5+}) in octahedral coordination (B-sites) with O²⁻ ions in a fcc lattice, builds up the polar catastrophe problem. We have prepared the system using molecular beam epitaxy (MBE). This deposition method allows for a layer-by-layer growth of the Fe₃O₄ under ultrahigh vacuum conditions, which facilitates the use of *in situ* characterization techniques with surface monolayer (ML) sensitivity.

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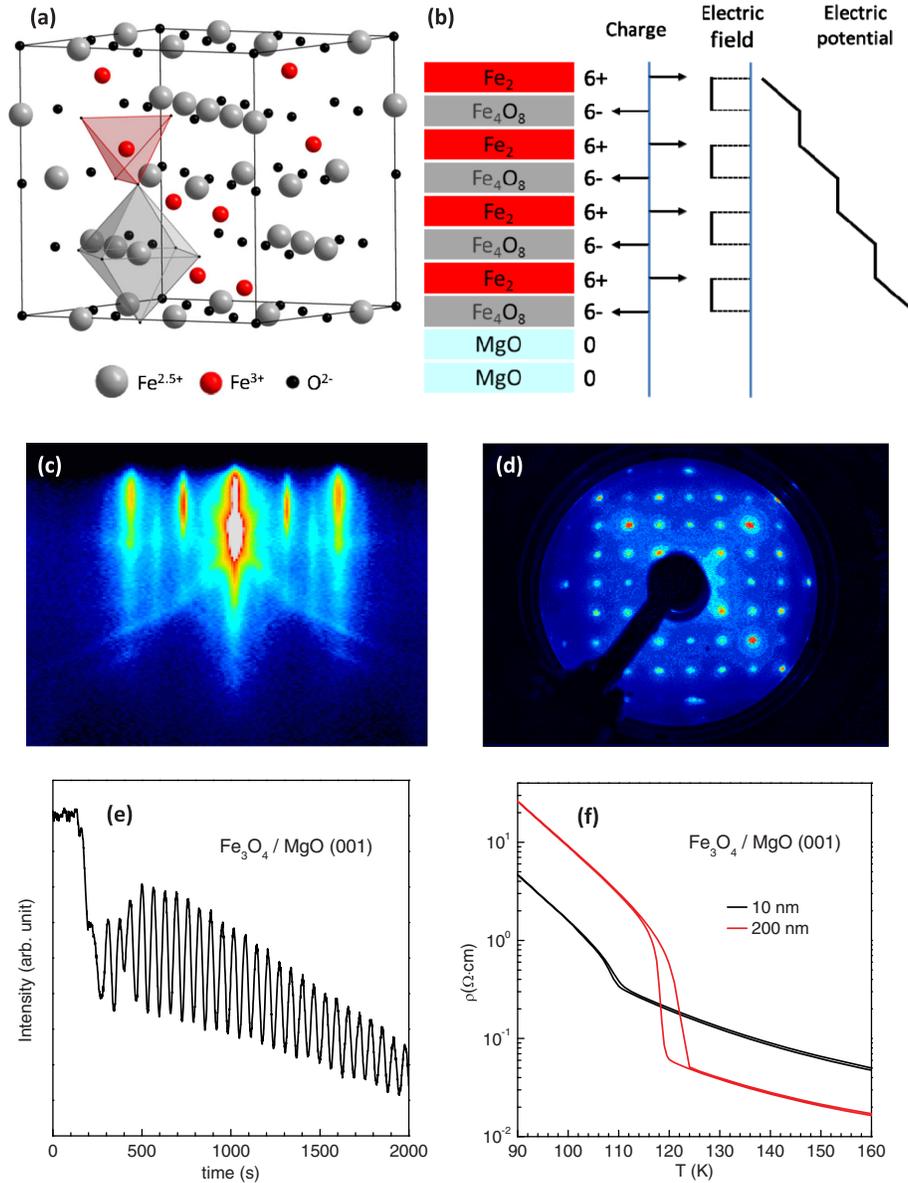


FIG. 1. Fe_3O_4 on MgO (001). (a) Structure of Fe_3O_4 . (b) Buildup of the polar catastrophe of Fe_3O_4 on MgO (001): charged planes, and the corresponding charge, electric field, and electric potential. (c,d) RHEED and LEED patterns, respectively, of a 200-nm epitaxial Fe_3O_4 (001) film showing the characteristic $(\sqrt{2} \times \sqrt{2})\text{R}45^\circ$ superstructure. (e) RHEED intensity oscillations of the specularly reflected beam. The electron beam was incident along the $[100]$ direction, with a primary energy of 20 kV. (f) Resistivity as a function of temperature of a 10-nm and a 200-nm thin film, showing the presence of the Verwey transition.

The MBE growth of Fe_3O_4 on MgO (001) is also known to produce films with excellent physical properties [22]. In order to obtain direct insight into the atomic and electronic structure of the interface, we utilize soft x-ray absorption spectroscopy (XAS) at the Fe $L_{2,3}$ edges. This spectroscopic technique is extremely sensitive to the local coordination and charge state of the Fe ions [25–28].

Fe_3O_4 thin films with thicknesses varying between 0.67 and 8 ML were grown on MgO (001). Each film was grown on a new and freshly annealed substrate. The substrate temperature was kept at 250 °C during the growth in order to avoid the Mg interdiffusion at the $\text{Fe}_3\text{O}_4/\text{MgO}$

interface [29,30]. Details about the film growth are given in Ref. [31]. One ML consists of one (001)-oriented layer of oxygen anions together with the appropriate number of Fe cations to maintain charge neutrality and stoichiometry, and it has a thickness of 2.1 Å. In Figs. 1(c) and 1(d), we present representative reflection high-energy electron diffraction (RHEED) and low-energy electron diffraction (LEED) patterns, respectively, of a 200-nm-thick Fe_3O_4 film to demonstrate that the surface is still smooth for very long deposition times. The typical $(\sqrt{2} \times \sqrt{2})\text{R}45^\circ$ surface reconstruction is also clearly visible. Figure 1(e) shows the regular oscillations with time in the intensity of

the specularly reflected RHEED beam during growth, indicating a two-dimensional layer-by-layer growth mode. Figure 1(f) displays the resistivity as a function of temperature from a 10-nm and a 200-nm film, showing the presence of the characteristic Verwey transition [22].

Figure 2 depicts the room-temperature Fe L_3 XAS spectra of Fe_3O_4 films with thicknesses varying from 0.67 ML to 8 ML, and of a Fe_3O_4 bulk single crystal. A Fe_2O_3 single crystal was measured simultaneously in a separate chamber to serve as energy reference for a Fe L_3 edge. Further XAS experimental details and display of the entire Fe $L_{2,3}$ spectral range are given in Fig. S1 of Ref. [31]. We also include in Fig. 2 the spectra of bulk $\text{YBaCo}_3\text{FeO}_7$ [28], bulk FeO (reproduced from Ref. [32]) and bulk Fe_2O_3 as references for Fe^{3+} ions in tetrahedral coordination, Fe^{2+} ions in octahedral coordination, and Fe^{3+} ions in octahedral coordination, respectively. The line shapes of the spectra strongly depend on the multiplet

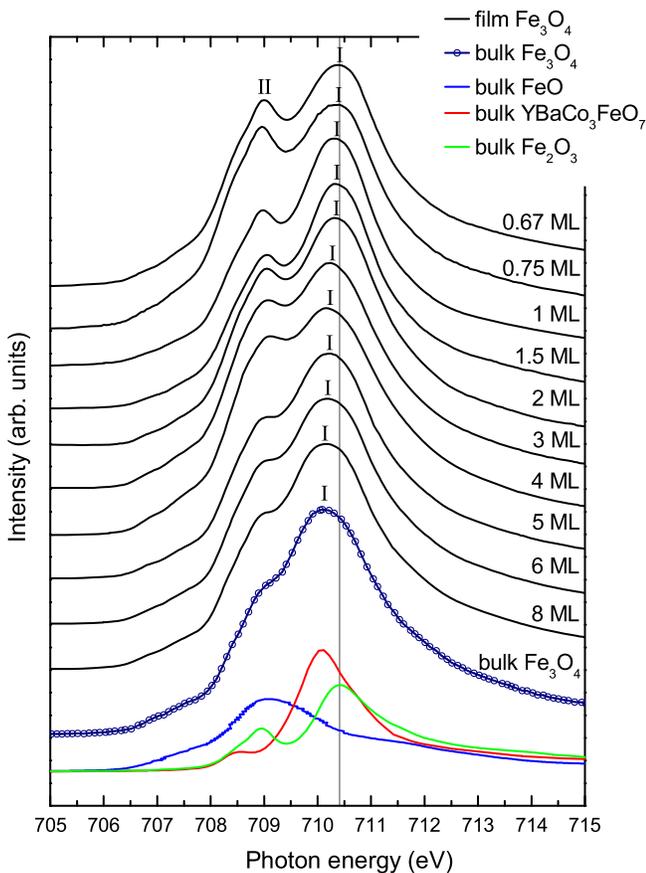


FIG. 2. Fe L_3 XAS spectra of Fe_3O_4 films. The film thickness varies from 0.67 to 8 ML. The reference spectra of bulk Fe_3O_4 , bulk $\text{YBaCo}_3\text{FeO}_7$ (Fe^{3+} in tetrahedral coordination) [28], bulk FeO (Fe^{2+} in octahedral coordination) [32], and bulk Fe_2O_3 (Fe^{3+} in octahedral coordination) are also included. All spectra were measured at 300 K. The gray line indicates the energy position of the main peak in the spectrum of bulk Fe_2O_3 . The full Fe $L_{2,3}$ spectral range is presented in Ref. [31].

structure given by the atomiclike Fe $3d - 3d$ and $2p - 3d$ Coulomb and exchange interactions, as well as by local crystal fields and the hybridization with the O $2p$ ligands [25–28]. Here, we note the striking similarities of the spectral features of the 8-ML Fe_3O_4 thin film and the bulk magnetite, which confirms that our Fe_3O_4 films have the correct stoichiometry.

We now focus on the thickness dependence of the spectra. Clear and systematic changes can be observed, in particular, in the peak position of the spectral feature labeled as (I) and in the intensity of the spectral feature labeled as (II) relative to that of peak (I) (see Fig. 2). The position of peak (I) of the thinnest Fe_3O_4 films, i.e., of the 0.67, 0.75, and 1-ML films, is the same as that of bulk Fe_2O_3 , while for the thicker films, i.e., 2 ML and beyond, it is more similar to that of bulk $\text{YBaCo}_3\text{FeO}_7$. This gives a first indication that the thinnest films contain only tiny amounts of Fe^{3+} ions in tetrahedral coordination and implies that such A-site Fe ions could essentially only be present for films of 2-ML thickness and beyond. This would then also explain why for the thinnest films one can see two separate peaks (I) and (II) like in bulk Fe_2O_3 (green curve), while for thicker films, the appearance of an in-between peak associated with the Fe^{3+} ions in tetrahedral coordination (red curve) will fill up the valley between peak (I) and (II), causing peak (II) to become a shoulder and the position of the larger peak (I) to shift to lower energies. It is important to note that the foot at the onset of the Fe L_3 edge, i.e., the feature between 706 and 707.7 eV, which is part of the spectral feature characteristic for Fe^{2+} ions in octahedral B sites (see blue curve), is thickness independent. All these ideas strongly suggest that the spectral weight of the A-site and the B-site Fe^{3+} ions varies strongly with thickness.

To interpret and better understand the XAS spectra and their thickness dependence, we have performed calculations using the well-established configuration interaction cluster model, which includes the full atomic multiplet theory and the local effects of the solid [25–28]. We have simulated each of the XAS spectra shown in Fig. 2 and obtained the spectral weight of the different Fe sites; computational details and fits are given in Refs. [31] and [33], and in Figs. S2–S4 of Ref. [31]. The results are plotted as closed squares in the top panel of Fig. 3, where the error bars reflect the deviations of the fits to the experimental data. From the relative concentrations of the constituents, we have calculated the average valence, or, equivalently, by taking the oxygen lattice to be complete, we have determined the Fe content y in our Fe_yO films. These y values are plotted as black closed squares in the bottom panel of Fig. 3. We can observe that all points are very close to the $\text{Fe}_{3/4}\text{O}$ (gray) line, which confirms the correct stoichiometry of our films through the entire thickness range and is very consistent with the RHEED intensity oscillations that have a constant time period, i.e., independent of the film thickness.

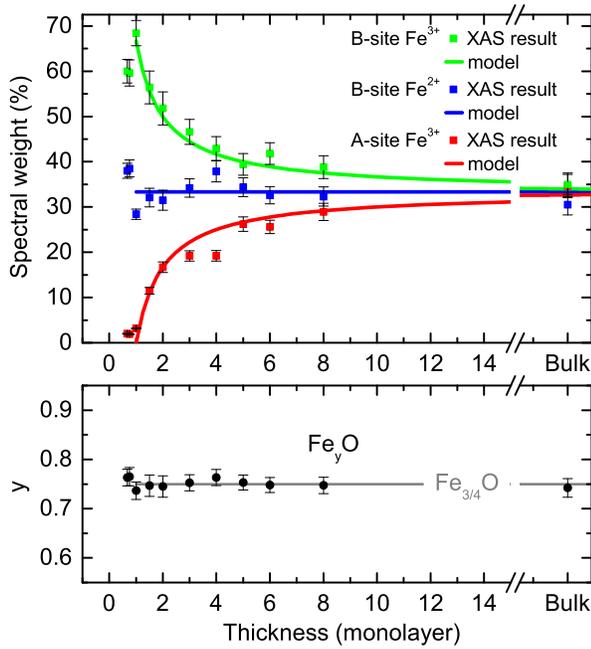


FIG. 3. The extracted spectral weight of each Fe site versus the film thickness. The spectral weight of the B-site Fe³⁺, the B-site Fe²⁺, and the A-site Fe³⁺ are given by the green, blue, and red squares, respectively. Green, blue, and red curves depict the concentrations of the B-site Fe³⁺, the B-site Fe²⁺, and the A-site Fe³⁺, respectively, in our model (see the text). The Fe content (y) of the Fe _{y} O films derived from the average valence is shown in the bottom panel. The error bars reflect the deviations of the fits to the experimental data.

An important aspect that emerges directly from the simulations is the strong thickness dependence of the different Fe constituents (see Fig. 3). We recall that bulk Fe₃O₄ has 1/3 (33%) Fe³⁺ ions in tetrahedral coordination (A-sites), and 1/3 (33%) Fe²⁺ and 1/3 (33%) Fe³⁺ ions in octahedral coordination (B-sites). To our surprise, we found that the amount of A-site Fe³⁺ ions is practically negligible for the thinnest films, i.e., 2%–3% instead of the 33% bulk value. At the same time, the amount of B-site Fe³⁺ in the thinnest films is between 60% and 68%, much larger than the 33% bulk value. We also observe that with increasing film thickness, the A-site Fe³⁺ amount increases and the B-site Fe³⁺ decreases, both approaching the 33% bulk value; see, for example, the 8-ML results in Fig. 3. Interestingly, the amount of B-site Fe²⁺ is rather constant, and independent of the film thickness, it fluctuates around the 33% bulk value.

These spectroscopic findings provide crucial data for the determination of the actual growth process and the interface structure. In particular, the observation that the first monolayer of the Fe₃O₄ film has essentially no A-sites is a surprising piece of information. In fact, as far as the monolayer is concerned, the choice of “nature” not to have A-sites is the simplest way to solve the planar electrostatic

potential problem. As can be seen from Figs. 1(a) and 1(b), it is indeed the presence of the A-sites that causes the polar catastrophe to occur, as there are no negative ions in those A-site planes to neutralize the charges. Thus, since there are no A sites for the first monolayer, there is no electrostatic problem. Then, the first monolayer basically constitutes a charge-neutral nonpolar rocksalt FeO layer with 25% Fe vacancies. All Fe ions occupy the B-site, with 33% of them having a 2+ valence and 67% a 3+ state; the vacancies are not ordered since we did not observe any superstructure. We have also carried out polarization-dependent XAS measurements, and we are indeed able to verify in detail that the dichroic spectrum is also consistent with the 33% Fe²⁺ and 67% Fe³⁺ B-site occupation. See Fig. S5 of Ref. [31]. Please note that these XAS spectra and the dichroism therein are very different from those of Fe atoms on MgO [34]. Moreover, capping this monolayer with a thick layer of MgO induces spectral weight changes that are fully consistent with the presence of 25% Fe vacancies. See Refs. [31] and [33], and Fig. S6 of Ref. [31].

For the second monolayer, in the experiment, we observe the appearance of some amount of A-sites, about 16.7%; see top panel of Fig. 3. We now arrive at the following model, Fig. 4, where the left side shows the growth process and the right side the corresponding net charges, electric field, and electric potential of each plane. Since in bulk magnetite a monolayer per unit cell includes two A-site Fe³⁺, two B-site Fe²⁺, two B-site Fe³⁺, and eighth oxygen ions, we use the formula notation Fe₆O₈ instead of Fe_{3/4}O to describe each monolayer. When deposited, the second monolayer will first form a nonpolar monolayer, like the first monolayer. Then, both the first and the second layers give away one Fe³⁺ ion, as shown in Fig. 4(b), to the space in between them to form a layer with two A-site Fe³⁺ ions. Then, the entire film also remains nonpolar: A 6+ charged layer is sandwiched by two 3- charged layers. The electric field oscillates symmetrically around zero, and the electric potential remains nullified. In this model, there are two A-sites per 2-ML formula unit; i.e., the A-site concentration is $2/(2 \times 6) = 16\%$, which is very close to the experiment. The remaining outer Fe₅O₈ (defective rocksalt) layers each contain three Fe³⁺ and two Fe²⁺ B-site ions. The concentration of the Fe²⁺ B-site ions is thus $(2 \times 2)/(2 \times 6) = 33\%$, i.e., the same as for the first monolayer and consistent with the experiment; see top panel of Fig. 3.

For a 3-ML film, the layer added will again form a nonpolar monolayer first. This monolayer and the subsurface monolayer then carry out the same process in which both give away one Fe³⁺ ion to the space in between. See Fig. 4(c). Again, the potential divergence remains nullified after this process, as shown on the rightmost panel of Fig. 4(c). One complete bulk Fe₆O₈ layer is now formed. This growth process is repeated for the subsequent layers, and the model predicts that the concentration of the A-site ions will increase following the geometrical series

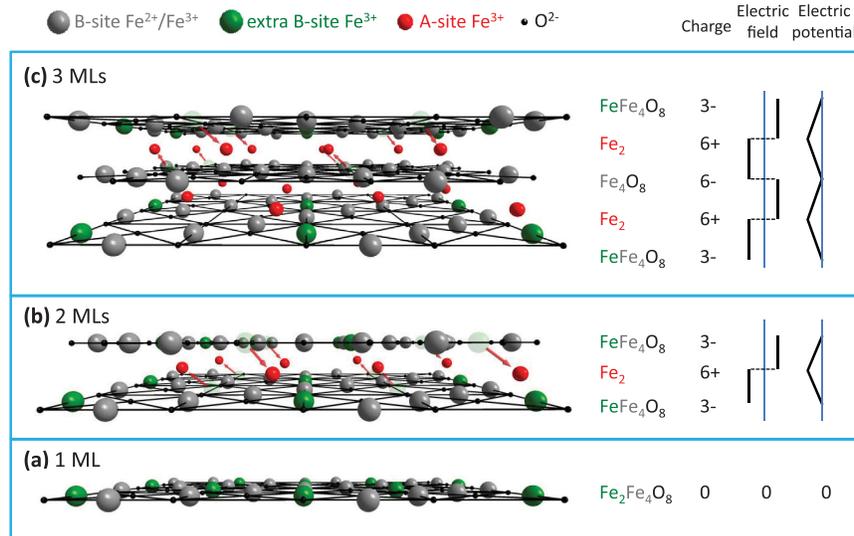


FIG. 4. Model for the growth process of polar Fe_3O_4 (001) thin films. (a) One monolayer, (b) two monolayers, and (c) three monolayers.

$2(n-1)/6n$, while the concentration of the Fe^{2+} B-site ions will remain constant at $2n/6n = 33\%$ and that of the Fe^{3+} B-site ions will decrease following $2(n+1)/6n$, where n denotes the number of monolayers. These predictions of the model are also presented in Fig. 3. We see that the essential behavior observed in the experiment is well reproduced with convergence to the bulk values for thicker films. We also note that an ordering in the outer Fe_5O_8 layer can be made consistent with the often-observed $(\sqrt{2} \times \sqrt{2})R45^\circ$ surface reconstruction in thicker (001) Fe_3O_4 films; see Ref. [31] for details.

We have thus found that the A-sites are absent in the first monolayer or interface, and that Fe ions are on the move while the film is growing to accommodate for the presence of A-sites inside the film having the proper crystal structure and stoichiometry. We clearly have a “dynamic atomic reconstruction” taking place here.

It is interesting to note that the Fe_3O_4 thin films are insulating and that the interface does not induce metallicity, as shown by the resistivity measurements displayed in Fig. 1(f). This result is obviously in contrast with the resistivity measurements on $\text{SrTiO}_3/\text{LaAlO}_3$ [5,35–37] and $\text{SrTiO}_3/\text{RETiO}_3$ [13–15]. In principle, the Fe_5O_8 interface layer could be conducting since this layer can be considered as a defective and doped rocksalt FeO layer. Yet, considering the fact that small polaron effects in bulk Fe_3O_4 are strong and that they hamper the conductivity [38–40], we may expect that this will also be the case for the interface layer. Its resistivity will then be dominated by strong scattering effects due to disorder.

Our findings have direct and important implications for the field of Fe_3O_4 spintronics. There are some reports concerning the possible existence of a magnetically dead layer at the interface [41–43], but others ascribe the

decrease of the magnetization in the thin films to the presence of antiphase boundaries leading to superparamagnetic behavior of the domains [44–46]. Our findings may give credit to the proponents of the dead magnetic layer model. In view of the absence or a low amount of A-sites in the interface region, some of the superexchange paths that determine the ferrimagnetism in Fe_3O_4 are certainly missing. This would also explain why tunneling experiments have spin polarizations that are different than expected from the properties of bulk Fe_3O_4 [47]. We can now propose that the insertion of a monolayer of magnetic metals like Fe, Co, Ni, or even noble metals like Cu, Ag, Au, or Pt between the Fe_3O_4 and the insulating oxide substrate will drastically change the situation: The metal layer inserted will act as a charge reservoir that can accommodate the flow of planar charges required to stabilize a Fe_3O_4 interface layer that has A-sites like in the bulk. The occurrence of a magnetically dead layer can then be prevented, and the spin polarization at the interface can also be increased. A hint that the latter is not unrealistic can be found in an early work by Dedkov *et al.* [48] on oxidized Fe films deposited on metal substrates.

To summarize, using soft x-ray absorption spectroscopy, we find that nature provides us with an unexpected but elegant solution for the polar catastrophe problem at the $\text{Fe}_3\text{O}_4/\text{MgO}$ (001) interface: The A-site Fe^{3+} ions are missing in the first Fe_3O_4 layer, and the growth process involves movements of not only the surface but also the subsurface Fe ions, securing epitaxy and stoichiometry at the same time. Having identified this “dynamic atomic reconstruction” growth principle, we conclude that we really have to think differently and openly about how polar interfaces can grow. Apparently, nature offers us a much wider range of opportunities to prepare unstable polar

interfaces. It would be interesting to put effort into growing a monolayer or a few monolayers of Fe_3O_4 film where the defects are ordered so that diffraction techniques can confirm the growth model.

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- and the entire Fe L_{2,3} spectral range, the XAS spectra computational details, and the $(\sqrt{2} \times \sqrt{2})R45^\circ$ surface reconstruction, which is consistent with the Fe₅O₈ surface as proposed by our model.
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