Modeling Failures of Thermal Barrier Coatings

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Modeling failures of Thermal Barrier Coatings
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Abstract. Thermal barrier coatings are commonly used in high temperature parts of gas turbines, to protect the underlying metal substrate from deterioration during high temperature exposure. Unfortunately, the coatings fail prematurely, preventing the design engineers to fully utilize their implementation. Due to the complexity of the coatings, there are many challenges involved with developing failure hypotheses for the failures. This paper reviews some aspects of the current state-of-the-art on modeling failures of thermal barrier coatings, focusing on mechanics based models (such as finite element simulations) where the material physics is incorporated (such as oxidation and diffusion).

Introduction

Thermal Barrier Coatings (TBCs) are commonly used to protect high temperature components in gas turbines from the harsh environment associated with elevated temperatures. The coating reduces the degradation rate of the superalloy and allows for higher operating temperatures, thus making the gas turbine both more economical and fuel efficient. Unfortunately, the full potential of TBCs is not utilized due to premature failures (i.e., in spallation of the coating), leaving the superalloy exposed to the high temperatures. The evolution of the failures is not completely understood, limiting the abilities to redesign and improve the coatings. The lack of understanding of failures in TBCs is primarily due to its complex structure. In order to understand the failures, it is important to correlate experiments and theoretical evolutions by conducting careful experiments and develop mechanics based model. The mechanics based models must incorporate the relevant physics governing the evolution of the failures. [1-6].

In this paper, we will discuss some strategies on how to model failures of TBCs. To this end, we will first discuss the most common TBCs and then discuss some selected failure modes and how these are most successfully modeled according to the current understanding. The discussion is limited to mechanics based simulations, assuming the basic theories of continuum mechanics holds.

Thermal Barrier Coatings

It is pertinent to understand the behavior of the components that constitute the TBC, and how the components interact and evolve during use, to achieve a successful failure model of a TBC. In this chapter, we will summarize some of the key constituents of the coating-system.

A TBC is a multilayered system where a bond coat is deposited on the superalloy, along with a ceramic top coat, see Fig. 1. The ceramic top coat provides thermal protection by having a low thermal conductivity. A thermal gradient over the top coat, of up to 150°C, is achieved from active cooling of the super alloy [1-6]. Due to the high temperatures the top coat is subjected to, the top coat typically sinters, resulting in increasing elastic modulus and potentially a transformation strain associated with the shrinkage due to sintering [7]. The metallic bond coat provides oxidation protection to the superalloy by providing aluminum to form an alpha-alumina scale (α-Al2O3) between the bond coat and the top coat (Fig. 1). Thus, the bond coat sacrifices itself, resulting in decreasing aluminum content and thus changing properties of the bond coat. The chemical content of the materials are important to control, since even small amounts of critical trace elements can enhance or reduce the interfacial fracture toughness of the structure.
The Ceramic Top Coat. The predominant material currently used for the top coat is 7-8wt.% yttria (Y$_2$O$_3$) stabilized zirconia (ZrO$_2$). There are two major processing methods for the top coat: Depositing the top coat through air-plasma spray (APS) or through electron-beam, physical-vapor-deposition (EB-PVD) [3]. Both methods are developed so that the top coat is porous, thus introducing a “strain tolerance.” The strain tolerance allows the top coat to be more compliant, reducing the risk of instantaneous spallation due to thermal mismatch and thermal chock. The processes result in distinct structures and interface morphologies between the top coat and bond coat, Fig 1. In APS top coats, “splats” parallel to the interfaces are formed, and microcracks associated with these splats that governs the strain tolerance, Fig 1A. In the EB-PVD system, the strain tolerance derives from a columnar structure and a micro-porosity (Fig. 1B). Due to the different morphologies, the failure mechanisms tend to be different between the two, as will be addressed below.

Bond Coat Materials: Two classes of intermetallic materials, MCrAlY (M = Ni and/or Co) or Pt-modified Aluminide (Pt-Al), are currently predominantly used as bond coats. The MCrAlYs are normally deposited by plasma spray or EB-PVD where the former case tends to give a rougher surface. Pt-Al made thought a diffusion process where a platinum layer is first placed on top of the substrate and is then aluminized by a thermal process, allowing for diffusion of the constituents, forming the bond coat [4]. The Pt-Al bond coat tends to results in more uniform TGO than the MCrAlYs. Infact, the MCrAlYs can form “pegs” which are localized very thick imperfection of TGO that penetrate the bond coat and may serve as nucleation sites for cracks [6], even though some models suggest that they might serve as crack stoppers [8]. To develop failure models of TBCs it is important to incorporate the evolution of the bond coat during the use: the microstructure can change significantly as the it is depleted from aluminum [9-13].

General Approaches for Modeling

We will here discuss some general modeling guidelines for the various layers in a thermal barrier system. As mentioned previously, the discussion is focused on “mechanics based” simulations, where a continuum mechanics approach is used. Thus, we will focus the discussion towards finite element analyses (FEA), which is currently the most popular approach. Fig. 2 shows a typical finite element model of a TBC, where the imperfection in the interface of the bond coat and top coat is studied [14].

![Figure 1](image_url).

Figure 1. Schematics of thermal barrier coatings: (A) Air plasma sprayed top coat and (B) Electron-beam, physical deposited top coat.
Figure 2. Example of finite element model of a TBC, using an axisymmetric model with continuous boundary conditions. [14]

**Initial State of Stress.** When starting the simulations, it is important to establish the true state of stress for the conditions considered. The most direct way is to establish the initial stress-free state and then perform the simulations. With few (if any) exceptions, the initial stress-free state is at the deposition temperature. For EB-PVD it can be assumed to be around 950°C and for APS around 350°C. However, process techniques vary, and it is important to make sure the correct temperature is used. A typical simulation will start with a stress-free state at this temperature, followed by a computation step that brings the system to ambient. This will result in residual stress. Thereafter the thermal exposure can be simulated, whether it is isothermal or cyclic condition. The cool down step can be ignored if only elastic properties are assumed, but the correct stress free conditions must be used. (A purely elastic simulation is unlikely to result in the correct material response.)

**Thermal cycling.** Thermal cycling (after the initial cooling step from processing conditions, described above) is typically simulated by three general steps: (i) heating (ii) thermal exposure and (iii) cooling, repeated to make up the number of cycles to be simulated. Isothermal conditions are simulated with heating, thermal exposure, and cooling. If the material properties are not changing with time (even though they may change with temperature), cyclic conditions will result in an initial evolution of stresses and strains, but these values will reach shakedown after a few cycles, that is, the stress and strains will cycle between two boundary values corresponding to the conditions at high and at low temperature [15, 16]. If all properties are elastic, shake down is reach immediately. Significant long term changes are numerically induced if there is no change imposed in the structure. By imposing a change in properties with time, such as TGO growth or top coat sintering, significant evolution can be obtained.

**Top coat.** The ceramic top coat is by design porous, Fig. 1, making it strain tolerant so to absorb some of the thermal mismatch strains introduced during cooling. Depending on the failure mode investigated, one can either (i) model the top as a homogeneous solid with low elastic modulus or (ii) introduce the overall shape of the splats (APS) or columns (EB-PVD). It is pertinent to keep the model as small as possible, in order to keep the computational effort reasonable and to retain a tractable numerical scheme. It may be sufficient to model the top coat homogeneous, see for example
[14, 17, 18]. The evolution of the top coat properties due to sintering can readily be introduced as a time dependent elastic modulus and an eigenstrain [7]. The details of the microstructures only need to be modeled if the internal morphology is crucial to the response, for example for foreign object damage or erosion. In most cases, the top coat can be assumed elastic and linear fracture mechanics theories can be utilized.

**Thermally Grown Oxide.** The TGO is the component that drives most of the failures: (i) the TGO grows due to oxidation at thermal exposure and (ii) the TGO has significantly higher elastic modulus than the companion materials in the coating system. The compressive stresses are caused by a combination of the thermal expansion mismatch between the TGO and the rest of the system during the cooling/heating sequence, and by the growth stress, $\sigma_G$, that is associated with of new alumina. The compressive stresses can reach several GPa at ambient conditions [15, 19, 20]. Localized tensile stresses can also appear in the TGO, due to the morphology, see for example [11, 15, 21]. An estimate of the stress in the TGO, $\sigma_{TGO}$, due to thermal mismatch and growth strain can be obtained by [16, 22]

$$\sigma_{TGO} = -\frac{E_{TGO}}{1 - \nu_{TGO}} \Delta \alpha \Delta T + \sigma_G$$

where $E_{TGO}$ and $\nu_{TGO}$ are the elastic modulus and Poisson’s ratio for the TGO; $\Delta \alpha$ is the difference in thermal expansion coefficient between the substrate and the TGO, and $\Delta T$ is the change in temperature. Typical properties as given by table 1 and growth stress ranges from 300 MPa to 1 GPa. Fig. 3 show some typical results from FEA of the stresses in the TGO [15, 21] for two possible imperfection shapes of the bond coat – TBC interface. The results are based on a model without the top coat (the bond coat and substrate are not shown for simplicity).

Thus, the TGO is associated with locally high stresses and strain energy, which has potential to drive failures in the vicinity of the TGO. It follows that it is important to simulate the TGO in a realistic manner. What is “realistic” must be determined from case to case, but in general, the thickening of the TGO and the growth strain has to be incorporated. Furthermore, the TGO is prone to creep at high temperatures and is elastic at ambient. Some aspects of this will be elaborated upon in the section that follows.

**Figure 3.** Typical stress distribution in the TGO for two different shapes of the interface between the bond coat and the TGO (convex vs. concave imperfection) after thermal cycling, for (A) ambient and (B) high (operating temperature). Note the tensile stresses in the TGO. [15, 21] (Stresses tangential to the bond coat interface.)
Sample Material Properties

<table>
<thead>
<tr>
<th>Property</th>
<th>Value at ambient</th>
<th>Value at operating temperature</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Top Coat</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Elastic modulus</td>
<td>40-80 GPa</td>
<td>40-80 GPa</td>
</tr>
<tr>
<td>Poisson’s ratio</td>
<td>0.1-0.2</td>
<td>0.2-0.3</td>
</tr>
<tr>
<td>Thermal Expansion</td>
<td>11 μ/°C⁻¹</td>
<td>12 μ/°C⁻¹</td>
</tr>
<tr>
<td><strong>Bond Coat</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Elastic modulus</td>
<td>180-200 GPa</td>
<td>70 GPa</td>
</tr>
<tr>
<td>Poisson’s ratio</td>
<td>0.3</td>
<td>0.3</td>
</tr>
<tr>
<td>Thermal Expansion</td>
<td>10-14 μ/°C⁻¹</td>
<td>15-16 μ/°C⁻¹</td>
</tr>
<tr>
<td>Yield/Creep strength</td>
<td>500-800 MPa</td>
<td>20-100 MPa</td>
</tr>
<tr>
<td><strong>TGO</strong></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Elastic modulus</td>
<td>380 GPa</td>
<td>320 GPa</td>
</tr>
<tr>
<td>Poisson’s ratio</td>
<td>0.2-0.25</td>
<td>0.2-0.25</td>
</tr>
<tr>
<td>Thermal Expansion</td>
<td>8 μ/°C⁻¹</td>
<td>8 μ/°C⁻¹</td>
</tr>
<tr>
<td>Growth stress</td>
<td>300-1000 GPa</td>
<td></td>
</tr>
</tbody>
</table>

Table 1: Typical properties of the components of a thermal barrier coating. Note: these are approximate values, typically used in generic simulations. Care must be taken when selecting properties.

**Bond Coat.** The bond coat is normally responding elastic at ambient, but yield or creeps at the higher temperatures. As for the TGO, the highest stresses are found at ambient. However, the strength is significantly lower at high temperature, resulting in the inelastic behavior even though the stresses are lower. Many simulations predict overall yielding of the bond coat, not only around the stress risers associated with imperfections [11, 23, 24]. The bond coat might have the most interesting evolution of properties. It provides aluminum to form the TGO, thus slowly changing its composition. This frequently results in phase transformations in some or all of the grains making up the bond coat. This can either be modeled as individual grains that undergo phase transformations, or allow the average properties of the bond coat to change, e.g. [10, 11].

**Superalloy.** The superalloy is frequently modeled as an elastic material with temperature dependent properties. If a single crystal superalloy is considered, the anisotropic nature of the elastic modulus might be important to incorporate.

**Interdiffusion zone.** As the system is exposed to elevated temperatures, the distinct boundary between the superalloy and the bond coat is many times smeared by an interdiffusion zone, eliminating the distinct boundary between them. This zone is may be ignored, unless the failure mechanisms are seen to be related to its presence.

**Material Properties.** Sample material properties are shown in Table 1. The properties are temperature depended and varies with time. The properties are extremely hard to measure, see for example [12, 13] and careful judgment must be used when assigning properties to the model being developed. When the properties are not well known, it is customary to conduct several analyses, studying the sensitivity for some critical material parameters.

**Examples of Modeling Selected Failure Modes**
We will here discuss some predominant failure modes and current model approaches to capture these failures.
Morphological Instabilities. The evolution of the TGO morphology that occurs due to cyclic thermo-mechanical loading, but not during isothermal conditions, are referred to as morphological instabilities (or sometimes ratcheting), schematically shown in Fig. 4. Surface roughening that occurs during isothermal conditions does not belong to this class of problems. This phenomena has received significant attention over the last few years, where Pt-Aluminide have seen to be most prone to exhibit this behavior [10, 11, 14-16, 21-23, 25], but it has also been seen in NiCoCrAlY-alloys in laboratory environments [24, 26] and on coatings in service or extreme temperatures (~1150°C) [27, 28].

The morphological instabilities evolve due to the compressive stresses that develop in the TGO during thermal exposure combined with thermal cycling, as was discussed previously. The TGO strive to relax the compressive stress by deforming out of its plane, deforming the bond coat. The mechanism is controlled by a combination of three non-linear constitutive behaviors: (1) high temperature inelasticity in the TGO; (2) growth strain in the TGO; (3) cyclic inelasticity in the bond coat. The growth strain is driving the system [15, 16, 22]. If the lengthening component of the aluminum formation is removed, the morphological instabilities cannot occur [15, 23].

The growth strain is induced due to the oxidation process when the new alumina is formed [6, 19, 29, 30]. Most of the TGO growth occurs as thickening, but a small part is distributed in the grains of the TGO leading to a lengthening component. The high temperature inelastic strength of the TGO is often referred to as the “growth stress.” The growth strain is limited by the growth stress, and once the TGO stress reaches the level of the growth stress, the lengthening strain is reallocated into thickening strain. In Fig. 4, the cases of cyclic versus isothermal scenario are compared, as a function of cumulative growth strain, showing the amplitude increase of an existing imperfection, and the tangential stress in the TGO (i.e., stresses parallel to the interface of the bond coat and TGO). The position designated Y refers to the onset of TGO inelasticity, e.g., the growth stress is reached. In the isothermal scenario, the amplitude change essentially stops, whereas under cyclic conditions, displacements and stresses continue to change after the point of TGO inelasticity is reached. This is so, since the stress in the TGO is relaxed during each cycle, allowing for additional accumulation of lengthening strain in the TGO at high temperatures. Morphological instabilities are governed by the state of stress in the bond coat in combination with the inelastic strain of the bond coat at high temperatures, since the inelasticity in the bond coat allows the TGO to relax and deform out of plane – into the bond coat – for each thermal cycle. The growth stress differentiates between isothermal and thermal cyclic exposure: if the growth stress is assumed infinite (i.e., the TGO is fully elastic) isothermal conditions will result in large amplitude changes as well. [16]

Figure 4. (A) Schematic of the development of morphological instabilities. (B) Amplitude change and stress in the TGO tangential to the bond coat surface.
Some factors can suppress morphological instabilities, for example, a full adherence of the ceramic top coat to the TGO/bond coat [21] or high yield strength of the bond coat [16]. Several factors that can enhance the rate of instability growth, including thermal mismatch with the bond coat [11] or thermo-mechanical loading [24], bond coat swelling [31], martensitic phase transformation [11, 23], or due to phase transformations between grains [10]. It is important to note that in all these cases, the rate of amplitude growth changes, but if the lengthening growth strain is not present, morphological instabilities do not occur.

The scenario and understanding of the morphological instabilities, as outlined above, could only be established by combining and coordinating experimental and theoretical investigations. Experimental investigations first indicated the formation of the instabilities and a hypothesis was formulated on how they evolved. The hypothesis was quantified in a numerical simulation (FEA) where some of the critical parts to the hypotheses was included (e.g. growth strain and plasticity in the bond coat). These simulations draw some particular conclusions which were verified in model experiments, e.g. [31].

To model the development of morphological instabilities, the volume change associated with the TGO formation must be incorporated. The lengthening component has been seen to be the most critical parameter driving the system. A convenient way of incorporating this component is to impose a stress free eigenstrain during the high temperature part of the cycle [15]. In this case, the total strain tensor, $\varepsilon_{\text{tot}}$, can be written as

$$\varepsilon_{\text{tot}} = \varepsilon_E + \varepsilon_P + \varepsilon_T + \varepsilon_G \tag{2}$$

where $\varepsilon_E$ is the elastic strain, $\varepsilon_P$ is the inelastic strain, $\varepsilon_T$ is the thermal strain, and $\varepsilon_G$ is the growth strain. The elastic strain and the thermal strain are conservative whereas the inelastic strain and the growth strain (caused by the oxide growth) are non-conservative (i.e., path depended and after unloading still present). Furthermore, the thermal strain and the growth strain are stress-free strains (i.e., if there were no constraints, no stress would be induced).

The implementation of growth strain into a finite element program will depend on the code used. For example, in ABAQUS it is convenient to use the user subroutine UEXPAN that allows the user to prescribe an anisotropic eigenstrain [15, 32]. Even though the thickening of the TGO is – surprisingly – not critical for driving the morphology change, it should be included. The concept of eigenstrains has been used successfully to model the thickening as well, e.g.,[11, 15, 24, 33, 34]. As an alternative, the bond coat elements closest to the TGO can be transformed to TGO as a function of cycles. In ABAQUS, this has successfully been done using “state variables” [13, 33, 35].

As mentioned previously the TGO and the bond coat creeps at high temperatures. This material behavior is important to incorporate. The TGO growth stress (which is the high temperature creep strength) is a function of exposure temperature, creep strength of the $\text{Al}_2\text{O}_3$ and the bond coat. [Here, we differentiate between the creep properties between a material (i.e., $\text{Al}_2\text{O}_3$) and the structure (i.e. TGO): the growth stress in the TGO depends on the properties of the surrounding components, such as elastic modulus and creep strength of the bond coat, since these are constraints for the TGO.] It has been shown that the creep strength of the TGO is what differentiates this system between cyclic and isothermal conditions: if the TGO is assumed elastic, the amplitude change is only depending on time-at-temperature [16, 22].

The most obvious approach is to incorporate the creep behavior through built-in routines in the used finite element package intended to simulate such behavior, which has successfully been done, in related problems, e.g., [36, 37]. However, this may not be very time efficient, since simulations involving visco-plastic effects are very time consuming. A more time efficient way is instead to use the time independent yield strength as the ultimate strength, first introduced in ref. [15]. If using the “end of cycle” stress of the TGO and bond coat, a good estimate of the behavior is found. In the case of morphological instabilities, we are mostly interested in the accumulation of plastic strain, (which will be translated into permanent amplitude change), thus it is not critical how the accumulation...
occurred, just that it did occur. (This scheme works well for cases where the long term cyclic behavior is investigated, but is not appropriate in all cases, particular when only a few thermal cycles are considered).

In this case, the failure behavior was successfully established through qualitative rather than quantitative simulations. For example, simplified material properties were used, and in most cases, limited numbers of cycles were considered (typically about 24 full cycles). By keeping the number of parameters to a minimum, and then carefully investigate which were important for the instability problem, qualitative statement could be made explaining the failure evolution. For example, it was seen that the lengthening component of the oxidation growth is the most important parameter driving the system and that by increasing the bond coat strength at high temperature it would be possible to suppress or at least reduce the rate of growth. To this end, the parameters were varied over a range of values, to study its influence of the structural behavior. Since material properties are many times unknown, a qualitative rather than quantitative approach is a powerful method to investigate failure modes.

We also note here that the modeling techniques used for simulating the oxide growth and the cyclic behavior can be successfully used for a range of cases where the TGO stresses contributes to the failure evolution.

**Plasma Sprayed Coatings.** In APS coatings, the surface is significantly rougher than for EB-PVD, as indicated in Fig. 1, and the morphological instabilities as outlined above are not observed. Nevertheless, it is important to incorporate the growth strain in the TGO, including the inelastic properties. A typical idealization used for investigating the effect the roughness has on the TBC stresses is shown in Fig. 5A, with typical stresses after cycling in Fig. 5B.

If the sintering of the top coat is ignored, simulations show that the stresses in the top coat reach a constant value, reaching shake down, e.g. [34]. The change due to the TGO growth is absorbed by bond coat yielding. If the sintering is incorporated, it is expected that the top coat stresses keep evolving. As can be seen from the stress values in Fig. 5B (based or ref. [34]), these stresses are large enough to induce cracks in the top coat. Using linear fracture mechanics, the crack propagation can be simulated and predicted [34, 37].

**Crack Growth.** Crack growth in a TBC is most frequently seen in the vicinity of the TGO. It is not uncommon that the crack grows in the top coat, as indicated in Fig. 4A (for a Pt-aluminide with morphological instabilities) or that the crack moves through several interfaces, such as from the TGO-bond coat interface, growing through the TGO and propagating in the top coat, as indicated in Fig. 5C (for a typical APS TBC). In both cases, the cracks are driven by the overall mode II caused by the thermal mismatch and the TGO growth stress, where will find the path of least resistance. For Pt-

![Figure 5.](image)

(A) A schematic of a thermal barrier coatings and (B) stresses in the top coat after thermal cycling, (C) Schematic of an interfacial crack in an APS TBC with an overall mode II crack.
Aluminide with EB-PVD coating, the top coat tends to have a very dense layer next to the TGO, where the porosity starts maybe 100 nm above the TGO. The cracks are observed to grow at this distance above the TGO. Based on modeling and experimental results, it is seen that the cracks start at the site of imperfections, and grow parallel to the TGO [14, 17, 38]. For the APS TBC, the crack will find its weakest path in combination of the largest crack driving force. Stresses are tensile in the top coat where the top coat is in a valley (Fig 5B), and the TGO can be tensile at its extreme points [34, 37, 39, 40]. Cracks typically starts as micro cracks which coalesce with time, leading to large scale cracks and final spallation [17, 18, 41].

**Externally Caused Damage of the Top Coat.** A range of mechanisms may cause damage to the top coat, which are caused by particles present in the gas flow. The smaller particles tend to cause erosion – slowly wearing down the top coat surface – whereas the larger particle can cause “foreign object damage” (FOD) where significant parts of the top coat is instantaneously impacted and removed [42, 43]. To simulate either of these two related damage modes, the individual columns of the EB-PVD top coat must be modeled (or the “splats” of the APS top coat). A random approach, using Monte Carlo simulations to estimate erosion rates, was conducted by Wellman and Nicholls [44]. The impact of larger particles, causing instantaneous critical damage was investigated in [45] where the individual columns and the friction between them where simulated. A large spherical impression was numerically induced and the deformation of the underlying top coat was investigated.

A second class of external damage to the top coat is caused by infiltration of calcium-magnesium-alumina-silicate (CMAS). This substance is commonly present due to sand dust and similar materials (in particular for gas turbines used for propulsion). The substance melts at temperatures of about 1150°C. During shutdown this may cause cold shock and spallation of the top coat [46]. To simulate this behavior, Chen [47] developed a model where the individual columns of the EB-PVD top coat is modeled, along with the volume change – and therefore the additional strain energy – of the CMAS. The energy release rate of assumed cracks in the interface between top coat and the TGO was determined based on the infiltration rate. It was seen, as maybe expected, that the interfacial cracks were more likely to grow as the infiltration rate increased.

**Concluding Remarks**

In this paper, a brief overview of a range of modeling techniques capturing failure modes in thermal barrier coatings (TBCs) is given. Most failures in TBCs are associated with a combination of mechanical loading and evolution of material properties due to high temperature exposure. The paper outlines some methods to incorporate these evolving properties into mechanics based simulations.

Most simulations discussed in this paper are based on qualitatively approaches rather than quantitatively. In qualitative approaches, simplified material properties and loading conditions are used, and parametric studies are conducted to evaluate the influence particular parameters have on the failure process. This is an excellent and simple approach to establish the overall failure behavior and it can show when one failure mode is dominate over another possible mode. However, if reliable life prediction models are developed, where the true number of cycles before failures is to be predicted, it is important to incorporate the exact material properties into a model, along with how they evolve with time. It is also important to incorporate all the possible failure modes that can occur. To this end, a statistical approach might be needed (which was not discussed in this paper), to randomly distribute the nucleation sites for the failures. A life prediction model will thus require significant experimental work along with a sound numerical scheme. This is still not available.

**References**

[34] M. He, J. Hutchinson, and A. Evans, Materials Science and Engineering A-Structural Materials Properties Microstructure And Processing, 345 (2003), p. 172