Characterization of Fracture Process in Ceramic-Metal Functionally Graded Material under Three-Point-Bending

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This paper deals with fracture process of a ceramic-metal functionally graded material (FGM) under three-point-bending. The used material was fabricated by powder metallurgy using partially stabilized zirconia (PSZ) and stainless steel (SUS 304), and has a functionally graded surface layer (FGM layer) on a SUS 304 substrate. In order to investigate the fracture process of the FGM, three-point-bending tests of rectangular specimens and numerical analysis are carried out. During the three-point-bending tests, crack initiation and unstable crack growth occur in the FGM layer, and the crack is arrested at the interface between the FGM layer and the substrate. Then, the crack branches and both crack tips grow stably along the interface with increasing deformation. After some amount of crack growth, both crack tips are arrested, and a new crack is initiated and grows into the SUS 304 substrate ahead of the initial cracking of the FGM layer. The finite element analysis taking account of gradation of material composition and plasticity of SUS 304 phase is carried out for each stage of fracture process. Based on the numerical results of the stress intensity factor, plastic zone and stress distribution, the fracture behavior of the FGM is discussed in detail.

Key Words: Functionally Graded Materials, Fracture Process, Three-Point-Bending, Fracture Mechanics, Finite Element Analysis

1. Introduction

Functionally graded materials (FGMs) are well known as materials in which a material composition varies gradually in some direction to derive unique mechanical, thermal and electrical performances different from those on homogeneous or joined dissimilar materials. FGMs consisting of ceramics and metals, for example, can be designed to reduce the thermal stress and to maintain the heat and corrosion resistances of ceramics and the mechanical strength of metals. Therefore, ceramic-metal FGMs are promising in thermal and structural applications such as thermal barrier coatings, wear and corrosion resistant coatings and ceramic/metal joining. In order to apply these kinds of materials to engineering structures, their fracture behavior and strength under mechanical and thermal loading should be made clear, and a method to evaluate their strength should be established.

The fracture behavior in FGMs is very complicated because the mechanical properties describing deformation (Young’s modulus, Poisson’s ratio, yield stress, etc.) and strength (tensile strength, ductility, fracture toughness, etc.) vary with the gradation of material composition. In a ceramic-metal FGM plate, surface cracks emanating at the ceramic side behave in several ways depending on the compositional gradation and loading condition. Multiple cracks, crack arrest, crack bowing, crack branching and spallation are well observed in ceramic-metal FGMs under thermal shocks or thermal fatigue. In order to elucidate the fracture process and to evaluate the strength of FGMs, the following three issues should be considered.

The first issue is the description of deformation properties varying with a material composition. Since FGMs are regarded as a kind of composites from their microstructures, models and equations proposed for composites can be applied to predict the deformation properties in FGMs by changing the material composition. The microstructure in FGMs also varies with the material composition. In ceramic-metal FGMs, for example, ceramic particles are dispersed in a metal matrix in the metal-rich region, while metal particles are dispersed in a ceramic matrix in the ceramic-rich region. Leßle et al. and Tohgo et al. developed models, respectively, which can describe the deformation behavior in a full range of ma-
terial composition taking account of the variation of microstructure.

The second issue is the applicability of fracture mechanics to FGMs. Eishen(15) and Jin and Node(16) demonstrated that a crack tip singular field in FGMs can be described by the J-integral and the mechanical properties of the material composition taking account of the variation of microstructure.

The third issue is the evaluation of strength properties varying with the material composition. This is very important in the elucidation of fracture process and strength of FGMs. The rule of mixture between the properties of constituent materials was applied as a simple way to evaluate the distributions of strength properties(25). For the distribution of fracture toughness or R-curve behavior in FGMs, prediction method based on the crack bridging model or cohesive model were proposed by Jin and Batra(25),(26), Jin et al.(27) and Cai and Bao(28). The distribution of strength properties, however, should basically be determined by experiments for a given FGM. The experimental investigations on the fracture behavior and strength of FGMs were relatively limited compared with the theoretical predictions on the fracture behavior and strength of homogeneous materials.

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2. Material and Experimental Procedure

The used material was fabricated by powder metallurgy using partially stabilized zirconia (ZrO₂-3molY₂O₃, PSZ) and austenitic stainless steel (SUS 304). The mean particle sizes of the powders were 0.07 µm for the PSZ and 3 µm for the SUS 304, respectively. The PSZ and SUS 304 powders were mixed in volume ratios of 10-0, 8-2, 6-4, 4-6, 2-8 and 0-10, respectively. Preforms of the FGM were made by layering the powder blends to form a graded composition and then sintered by hot isostatic pressing under the condition of 1 473 K and 196 MPa for one hour. The material had a functionally graded surface layer (FGM layer) with a thickness of 0.75 mm on a SUS 304 substrate with a thickness of 4.25 mm. Figure 1 shows the configuration of three-point-bending specimen and details of the FGM layer. The FGM layer consists of multi-layers in which a volume fraction of PSZ is varied stepwise from 100% at the surface to 20% at the interface between the FGM layer and the substrate. Figure 2 shows the microstructure of each layer in the FGM observed by a scanning electron microscope. The white and gray regions exhibit the PSZ and SUS 304 phases, respectively. The SUS 304 particles are dispersed in the PSZ matrix in the layers from 80% PSZ to 60% PSZ, and on the contrary, the PSZ particles are dispersed in the SUS 304 matrix in the layer of 20% PSZ. The layer of 40% PSZ exhibits the microstructure in which both PSZ and SUS 304 phases interpenetrate each other. The groove observed in 100% SUS 304 shows the grain boundary by etching.

Three-point-bending tests were carried out on three specimens as shown in Fig. 1 at 0.05 mm/min crosshead speed at room temperature and in air. During the tests, the relationship between load (P) and load-point-displacement (δ) was recorded, and the image of fracture process observed on the side surface of specimens was recorded in a video recorder through a CCD camera. The test and video recording were started at the same time and the time was measured through the test so that the fracture process could be investigated in detail after the test. On the fractured specimen, details of the fracture surface were examined by a scanning electron microscope.

3. Fracture Process

Figure 3 shows an example of the load-displacement relations obtained by three-point-bending test, and Fig. 4 exhibits the observation of the side surface of the specimen by a CCD camera. The load-displacement relation of the specimen was slightly nonlinear after the elastic linear re-

![Fig. 1 Specimen geometry and details of FGM layer](image-url)
At the maximum load (stage B in Fig. 3), the crack initiation and unstable crack growth occurred in the FGM layer, and the crack was arrested at the interface between the FGM layer and the substrate. During this event, the load dropped from stage B to stage C in Fig. 3. With increasing displacement, the plastic deformation developed from the crack tip, and the crack branching and growth occurred along the interface between the FGM layer and the substrate. After a certain amount of crack growth, both crack tips were arrested (stage E), and a new crack was initiated and grew into the SUS 304 substrate ahead of the initial cracking of the FGM layer (stage F). This crack growth led to the final failure of the FGM. The almost same fracture process was observed in the other two specimens.

The behavior from crack initiation (stage B) to fracture of the FGM layer (stage C) is one of the most noticeable events from a viewpoint of the capability of the FGM layer for crack arrest. In the present investigation, the unstable fracture through FGM layer might be caused by the high load for crack initiation in the smooth specimen and the elastic strain energy stored in the specimen enough to draw the unstable crack growth through the FGM layer.

More detailed observation of crack tip regions in the above fracture process was conducted on another specimen whose surface was etched to expose the microstructure. The specimen was loaded to some stages, unloaded and then observed by a scanning electron microscope. Figure 5 (a) and (b) shows the crack tip regions of an edge crack corresponding to the stage C and a cruciform crack corresponding to the stage F in Fig. 3, respectively. In
Fig. 5 (a), although the crack is macroscopically arrested at the interface between the FGM layer and the SUS 304 substrate, micro damage is observed in the substrate. In Fig. 5 (b), interface branched cracks develop remarkably as compared with a substrate crack.

Micrographs of the fracture surfaces taken at each layer of the FGM are shown in Fig. 6. Morphology of the fracture surface of the FGM depends on the material composition. The fracture surface of 100% PSZ layer shows typical appearance of the brittle fracture of ceramics. In the 80% PSZ layer and 60% PSZ layer, the debonded SUS 304 particles and their traces are observed in the brittle fracture surface of the PSZ matrix. This means that the interfacial strength between the SUS 304 particles and the PSZ matrix was relatively low. In the 40% PSZ layer and 20% PSZ layer, the angular fracture surfaces with sharp ridges are created by the brittle fracture of the PSZ phase and the fracture of the SUS 304 phase. The appearance of these fracture surfaces suggests the brittle-like fracture. On the other hand, the 100% SUS 304 substrate exhibits the ductile fracture surface with dimple-pattern. From the observation of fracture surface, it is found that the FGM layer from 100% PSZ to 20% PSZ was brittle and the SUS 304 substrate was ductile, and consequently the fracture behavior under three-point bending became very complex.

4. Constitutive Relations

Tohgo and Chou(11) developed an incremental damage theory, which describes the plasticity of a matrix and progressive debonding damage of particles from the matrix, based on Eshelby’s equivalent inclusion method(34) and Mori and Tanaka’s mean field concept(35). This theory is applicable not only to particulate-reinforced composites but also to hybrid composites containing particles and voids and porous materials. Here, the incremental theory used in the following finite element analysis is briefly explained for the case of particle-reinforced composites as shown in Fig. 7, in which elastic spherical particles are homogeneously dispersed in an elastic-plastic matrix and any damage never develops during deformation.

In the composites, the microscopic stress/strain in the particles and matrix are generated due to the material heterogeneity, in addition to the overall macroscopic stress/strain. The stress and strain of the particles and matrix are represented with the superscripts “p” and “0”, re-
respectively, and those of the composite are shown by symbols without superscript.

4.1 Properties of particles and matrix

The elastic incremental stress-strain relation of the particles is given in the isotropic form:

$$d\sigma^p = D_0' (E_0', \nu_0') d\varepsilon^p,$$

(1)

where $D_0' (E_0', \nu_0')$ is the elastic stiffness described by the Young’s modulus $E_0'$ and Poisson’s ratio $\nu_0'$ of the particles. The elastic behavior of the matrix is also taken to be isotropic with the elastic stiffness $D_0 (E_0, \nu_0)$ described by $E_0$ and $\nu_0$ of the matrix. Prandtl-Reuss equation (the J2-flow theory) to describe the elastic-plastic deformation of the matrix is approximated by the same form as the elastic relation(11):

$$d\sigma^m = D_0 (E_0, \nu_0) d\varepsilon^m,$$

(2)

where

$$E_0' = \frac{E_0}{1 + \frac{E_0}{H'}}, \quad \nu_0' = \frac{\nu_0}{1 + \frac{E_0}{H'}},$$

(3)

$D_0 (E_0, \nu_0)$ is the equivalent tangent modulus obtained by replacing $E_0$ and $\nu_0$ with $E_0'$ and $\nu_0'$ in $D_0 (E_0, \nu_0)$. $E_0'$ and $\nu_0'$ are the tangent Young’s modulus and tangent Poisson’s ratio under elastic-plastic deformation. $H'$ is the work-hardening ratio of the matrix:

$$H' = \frac{d\sigma^m}{d\varepsilon^m},$$

(4)

where

$$\sigma_e^0 = \left\{\frac{3}{2} (\sigma_{kl})^0 (\sigma_{kl})^0 \right\}^{1/2},$$

$$d\varepsilon_e^0 = \left\{\frac{2}{3} (d\varepsilon_{kl})^0 (d\varepsilon_{kl})^0 \right\}^{1/2}.$$

$\sigma_e^0$ and $d\varepsilon_e^0$ are the von Mises’ equivalent stress and incremental equivalent plastic strain, respectively. Equation (2) is strictly valid in the case of monotonic proportional loading.

4.2 Constitutive relation of particle-reinforced composites

For the particle-reinforced composite containing spherical particles with a volume fraction $f_p$ in a matrix, an incremental stress-strain relation is given as follows(11):

$$d\sigma = D d\varepsilon,$$

(6)

$$D = D_0 \left[ (1 - f_p) (D_p - D_0) S + D_0 \right]^{-1} A,$$

(7)

$$A = \left[ (1 - f_p) (D_p - D_0) S + D_0 \right] + f_p D_p,$$

(8)

where $d\sigma$ and $d\varepsilon$ are the incremental stress and strain of the composite. The stiffness tensor of the composite $D$ is given as a function of the stiffness tensors of the particles and matrix $D_p$ and $D_0$, a particle volume fraction $f_p$ and Eshelby’s tensor for a spherical particle $S^{(34)}$. $S$ is described by the Poisson’s ratio of the matrix $\nu_0$. Furthermore, incremental average microscopic stresses and strains of the particles and matrix are given as follows:

$$d\sigma_e^p = D_0 (S - I) A^{-1} D_p (S - I)^{-1} D_0^{-1} d\sigma,$$

(9)

$$d\varepsilon_e^p = D_0^{-1} d\sigma_e^p,$$

(10)

$$d\sigma_e^m = D_0 (S - I) A^{-1} \left[ (D_p - D_0) S + D_0 \right]$$

$$\times (S - I)^{-1} D_0^{-1} d\sigma,$$

(11)

$$d\varepsilon_e^m = D_0^{-1} d\sigma_e^m.$$

(12)

4.3 Equivalent stress of matrix

In order to carry out the elastic-plastic analysis of the composite, in-situ von Mises’ equivalent stress of the matrix in the composite should be estimated. The equivalent stress of the matrix deformed heterogeneously in the composite is evaluated based on Tohgo-Weng’s energy approach(12). An initial equivalent stress of the matrix before plastic deformation is given as

$$\left( \sigma_e^0 \right)^2 = \frac{3E_0}{2(1 - f_p)(1 + \nu_0)} \left( \sigma_e - f_p \sigma_e^p \sigma_e^p \right)$$

$$- \frac{9(1 - 2\nu_0)}{2(1 + \nu_0)} \left( \sigma_e^m \right)^2,$$

(13)

where $\sigma$ and $\varepsilon$ are the macroscopic composite stress and strain, $\sigma^0$ and $\varepsilon^0$ are the microscopic particle stress and strain, and $\sigma_m^0 (= 1/3\sigma_{kk}^0)$ is the average hydrostatic stress of the matrix. After incremental deformation the equivalent stress of the matrix is described by $\sigma_e^0 + d\sigma_e^e$, where $\sigma_e^0$ and $d\sigma_e^e$ denote the current equivalent stress before the incremental deformation and its increment, respectively. In the numerical analysis $\sigma_e^0$ is known and $d\sigma_e^e$ is given by

$$d\sigma_e^e = \frac{3E_0}{2\sigma_e^0 (1 - f_p)(1 + \nu_0)} \left( \sigma_e d\varepsilon - f_p \sigma_e^p \sigma_e^p d\varepsilon^p \right)$$

$$- \frac{9(1 - 2\nu_0)}{2\sigma_e^0 (1 + \nu_0)} \sigma_m^0 d\sigma_e^m.$$

(14)

If the matrix is in the elastic state, $E_0'$ and $\nu_0'$ in all the equations in sections 4.2 and 4.3 are changed to their counterparts $E_0$ and $\nu_0$. 
In the deformation and damage behavior of particulate-reinforced composites, particle size is an important factor in addition to the particle volume fraction. However, the present model based on the Eshelby’s equivalent inclusion method and Mori-Tanaka’s mean field concept can not describe the influence of particle size on the elastic-plastic deformation. Implementation of the size effect of particles into the model would be carried out as our future work.

5. Numerical Procedure

In order to make clear the fracture process of the FGM, four stages of the fracture process under three-point bending were separately analyzed. The four stages corresponding to B, C, E and F in Fig. 3 are simulated on the smooth specimen, with an edge crack, with a branched crack and with a cruciform crack, respectively, as shown in Fig. 8. The crack sizes in Fig. 8 are determined from the cracks observed at the stages C, E, and F in Fig. 3. Since the actual fracture process was continuous events, continuous simulation of the four stages should be carried out in the numerical analysis. However, it is very difficult to simulate continuously the complicated fracture process including unstable crack initiation and growth, crack branching, crack arrest and formation of a cruciform crack. Therefore, the present numerical analysis was separately conducted for each stage under elastic-plastic plane strain condition, and then the fracture process was discussed based on the numerical results.

The numerical analysis of the fracture process was carried out by using a finite element method (FEM) developed based on the incremental constitutive relation for particle-reinforced composites. The finite element method was formulated on the basis of quadrilateral 8-noded isoparametric elements with integration at four Gauss points. This FEM is easily applied to the analysis of FGMs by setting the gradation of a particle volume fraction\(^{(21)}\). The microstructure of the FGM varied from PSZ matrix structure to SUS 304 matrix structure with an increase in the SUS 304 volume fraction as shown in Fig. 2. However, the present analysis was carried out under the assumption that the elastic PSZ particles are dispersed in the elastic-plastic SUS 304 matrix through the FGM layer, and that the debonding damage between the PSZ particles and SUS 304 matrix never develops during deformation.

Figure 9 shows a finite element mesh used in the analysis. The PSZ particle volume fraction was assigned in the elements for the FGM layer to be consistent with the actual materials. The first step of incremental analysis was solved for linear problem up to the applied load which corresponds to the plastic deformation at only one Gauss point in the whole region. In the following steps, Newton-Raphson method was adopted to solve nonlinear problem for each load increment.

The mechanical properties of the PSZ particles and SUS 304 matrix used in the analysis were determined from the references\(^{(35),(37)}\) as follows. The Young’s modulus and Poisson’s ratio were

\[
E_p = 133 \text{ GPa}, \quad \nu_p = 0.17
\]

for the PSZ particles and

\[
E_0 = 155 \text{ GPa}, \quad \nu_0 = 0.30
\]

for the SUS 304 matrix. The yield stress \(\sigma_0\) and the equivalent stress \(\sigma_e^0\) vs. equivalent plastic strain \(\varepsilon_p^0\) relation of the SUS 304 matrix were given by
3.27

Fig. 10 Load vs. displacement relations obtained by FEM analyses

\[ \sigma_0 = 245 \text{ MPa}, \]  

\[ (17) \]

\[ \sigma_0' = \sigma_0' \left(1 + \frac{E_0}{E_0} \right)^{0.11}, \quad E_0 = \frac{\sigma_0}{E_0}. \]  

(18)

6. Numerical Results

Figure 10 shows load vs. displacement relations of the four specimens, where the displacement (\( \Delta \)) is taken at the tension side edge of the specimen. In this figure, the following correction was made for the displacement. In the finite element analysis, as the detailed phenomena of solid contact at the loading points was not considered, the load-point displacement contains some amount of error. Therefore, the displacement (\( \Delta \)) at the tension side of the specimen was adopted for the measure of deformation. As the load-point displacement \( \delta \) in the experiment was obtained by the crosshead displacement of the testing machine, \( \delta \) was not consistent with \( \Delta \). In Fig. 10, the experimental relation was modified so that the initial elastic part of the experimental \( P-\Delta \) relation was consistent with the numerical \( P-\Delta \) relation of the smooth specimen, and the numerical \( P-\Delta \) relations for the other cracked specimens were shifted so that they reached the corresponding stages on the experimental relation. Although the actual fracture process was continuous events, the present numerical results for each event were separately obtained for each specimen as shown in Fig. 10.

The elastic and elastic-plastic stress distributions along the center line of the smooth specimen are shown in Fig. 11, where the stress (\( \sigma_y \)) is normalized by the bending stress at the edge of an elastic homogeneous specimen (\( \sigma_b = 3PS/(2W^2B) \)). On the elastic stress distribution, the bending stress at the surface of FGM layer is reduced from \( \sigma_b \) of the elastic homogeneous specimen by the influence of the FGM layer. On the elastic-plastic stress distribution, however, the bending stress is enhanced by the plastic deformation in the metal phase of FGM layer and substrate. This means that the plastic deformation of the metal phase and substrate promotes the fracture of the ceramic surface layer in the FGM under bending.

Based on the elastic analysis of the three cracked specimens, the stress intensity factors were determined for the arrested edge crack (stage C in Fig. 3), the arrested branched crack (stage E) and the growing cruciform crack (stage F). In the cases of arrested edge crack and branched crack, a crack tip or crack line is in the interface between the 20% PSZ layer and the SUS 304 substrate. However, since the difference in Young’s modulus between PSZ and SUS 304, and also between the 20% PSZ layer and the SUS 304 substrate is small, these cracks can be regarded as a crack in a homogeneous body. The stress intensity factors were determined by extrapolation of \( K_I' = \sigma_y \sqrt{2\pi r} \) and \( K_{II}' = \tau_{xy} \sqrt{2\pi r} \) to a crack tip using the distributions of normal stress (\( \sigma_y \)) and shear stress (\( \tau_{xy} \)) along the line ahead of a crack tip. The obtained results are shown in Table 1. From the fact that the crack nucleated in the FGM layer grew unstably and was arrested at the interface between the FGM layer and the substrate (stage C), it is presumed that the fracture toughness is lower than 29 MPa \( \sqrt{\text{m}} \) in the FGM layer, and is higher than 29 MPa \( \sqrt{\text{m}} \) in the SUS 304 substrate. The branched crack tips for the stage E are under mixed mode loading of \( K_I \) and \( K_{II} \), and the mode I and mode II stress intensity factors for the crack arrest are very low; namely the fracture toughness of the interface between the FGM layer and the substrate is very low. The stress intensity factor of the cruciform crack growing in the SUS 304 substrate (stage F) is 21.6 MPa \( \sqrt{\text{m}} \), and lower than that for the stage C.

| Stages of fracture process | C | E | F |
|---------------------------|---|---|---|
| \( K_I \) (MPa\( \sqrt{\text{m}} \)) | 29.0 | 7.86 | 21.6 |
| \( K_{II} \) (MPa\( \sqrt{\text{m}} \)) | – | 8.64 | – |
Because the large plastic deformation actually developed in the specimen at stage F, this $K_I$ value might be invalid as a measure of the fracture toughness for the SUS 304 substrate. It is concluded that the above complicated fracture process of the FGM is attributed to the low fracture toughness in the FGM layer and at the interface as compared with the substrate. The distribution of fracture toughness in the FGM layer and interface is necessary to evaluate the strength and fracture behavior of the FGM in more detail\(^{(38)}\).

Figure 12 shows the plastic zones for each stage of the fracture process of the FGM. The plastic zone indicates the region where the microscopic stress of the SUS 304 satisfies the von Mises’ yield condition. The current plastic zone for each stage is indicated by the shading, and the plastic zone of the previous stage is described by thin lines in Fig. 12 (b)–(d). Before the crack initiation (stage B), the plastic zone spreads out from the FGM layer to the SUS 304 substrate, and the crack initiation leads to the stress release on the crack surfaces and enlarges the plastic zone in the direction of 45 degree to the crack line (stage C). At the stage C of the crack arrest, the plastic zone extending from the crack tip is connected with the plastic zone of the backside of the specimen, namely the specimen is under the general yielding. The plastic zone develops furthermore in the stage of the branched crack (stage E), and then it shrinks as the crack grows into the SUS 304 substrate (stage F). These numerical results of the plastic zone suggest that the plastic deformation should be taken into consideration in the evaluation of the fracture behavior of the FGM.

Stress distributions around cracks for each stage of the fracture process are shown in Fig. 13. Figure 13 (a) and (e) describes the distributions of $\sigma_y$ for the mode I cracks in the stages C and F, and Figs. 13 (b)–(d) describe the distributions of $\sigma_x$, $\sigma_y$ and $\tau_{xy}$ for the mixed mode crack in the stage E, respectively. In this figure, the high
stress concentration around a crack tip and the variation of the stress distribution with fracture process are well expressed. After the crack was arrested (stage C), the crack did not extend straight into the SUS 304 substrate but the crack branched and stably grew along the interface. On the other hand, after the branched crack was arrested (stage E), a new crack was initiated and grew into the SUS 304 substrate. Figure 14 shows the comparison of the stress distributions along the centerline between the stage C and stage E. Although in the vicinity of the crack-tip the stress in the stage C is higher than that in the stage E due to the stress singularity, in the wide region ahead of the crack-tip the stress in the stage E is higher than that in the stage C.

From the discussions on the stress intensity factor, plastic zone and stress distribution, the fracture process of the FGM is explained as follows. The branched crack was created due to low fracture toughness of the interface as compared with the SUS 304 substrate after the edge crack was arrested. With increasing deformation of the specimen, the tips of the branched crack grew stably along the interface and the stress intensity factors decreased down to the critical value for crack arrest at the interface. During this period, the SUS 304 substrate ahead of the initial cracking of the FGM layer deformed plastically, and the stress level increased in the wide region. Finally, the new crack was initiated and grew into the SUS 304 substrate.

The complicated fracture process in FGMs as shown in the present investigation would depend on the distribution of fracture toughness in the FGM layer, thickness of the FGM layer and fracture toughness of the interface between the FGM layer and the substrate. The strength evaluation and material design of FGMs would be achieved by elucidating the fracture process based on the distribution of fracture toughness.

7. Conclusions

In order to investigate the fracture behavior of the FGM consisting of PSZ and SUS 304, the three-point-bending tests and finite element analysis were carried out. The obtained conclusions are summarized as follows.

1. During the three-point-bending test, the crack initiation and unstable crack growth through the FGM layer, the crack branching along the interface, and the stable crack growth into the substrate were observed.

2. The complicated fracture process of the FGM was attributed to the low fracture toughness in the FGM layer and at the interface as compared with the substrate and to the plastic deformation of the metal phase.

3. From the above conclusions, the distribution of fracture toughness in the FGM layer, interface and substrate and the plastic deformation of the metal phase should be considered in the strength evaluation and material design of FGMs.

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