A combined dislocation–cohesive zone model for fracture in nanocrystalline materials

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A combined dislocation–cohesive zone model was proposed to describe the fracture toughness of nanocrystalline (nc) materials. In the framework of the model, cohesive stress near crack tip initiates edge dislocations, which move to the opposite grain boundaries. The emitted dislocations provide a shielding effect of the crack. The dependence of both the maximum number of dislocations, emitted by a crack, and the critical stress intensity factor on grain size d (ranging from 20 to100 nm) for Cu was calculated. The calculated results show that (i) nc materials have low fracture toughness, (ii) the critical stress intensity factor decreases with decreased grain size, and (iii) the grain size effect is not high; for instance, increasing the grain size from 20 to 100 nm increases the value of critical stress intensity factor only by 0.035 MPa/m$^{1/2}$.

I. INTRODUCTION

Nanocrystalline (nc) materials have been a source of great interest currently because of their unusual mechanical and physical properties.[1–8] In general, nc materials show superior strength but low tensile ductility and low fracture toughness, which limit their applications.[1,3,4,6,9,10]

Although many experiments have been performed to study the fracture toughness of nc materials, the theoretical and quantitative descriptions about this issue are very limited. In nc materials, the conventional toughening associated with the lattice slip is suppressed. It has been proposed that the alternative deformation modes, such as grain boundary (GB) sliding, GB diffusional creep, stress-driven GB migration, twin and rotation deformation, can play an important role in the toughening of nc materials. This is in agreement with experimental data showing the enhancement of fracture toughness of several nc ceramics,[11–13] nc copper,[14,15] and nc nickel[16] compared to those of their microcrystalline counterparts. These experimental results serve as the basis of theoretical mechanisms hampering crack growth in nc materials. In particular, recently, nanoscale twin deformation carried by partial dislocations emitted from GBs[17] has been theoretically revealed to be the specific toughening mechanism in nc ceramics and metals. In addition, Bobylev et al.[18] built a theoretical model to describe the effects of GB sliding (GBS) on the crack growth in nc metals and ceramics. Within their model, the GBS in nc materials results in crack blunting which, in its turn, leads to an increase (by a factor ranging from 1.1 to around 3, depending on temperature) of fracture toughness. Ovid'ko and Sheinerman[19] have studied the combined effects of GBs and blunting of cracks on the fracture toughness of nc and ultrafine grained materials. They found that nc metals exhibit a ductile-to-brittle transition with decreased grain size. Liu et al.[20] developed a model to study the combined effects of GBS and emission of edge dislocations from crack tip on the toughness of nc materials. Bobylev et al.[21] suggested a new physical mechanism of plastic deformation in nc metals and ceramics to study the cooperative GBS and stress-driven GB migration process. It was theoretically revealed that the new deformation mode was more energetically favorable than "pure" GBS and enhances the ductility of nc solids in wide ranges of their structural parameters. Morozov et al.[22] developed a theoretical model to describe the effect of special rotational deformation on crack growth in deformed nc ceramics and metals. The special rotational deformation releases, in part, local stresses near the crack tip, thus serving as a toughening mechanism in nc materials. Creep deformation conducted by GB processes was also suggested as a toughening mechanism providing crack blunting (at grain size scales of 10 nm or more) in nc metals.[23,24]

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During the last few decades, cohesive model of fracture has gradually evolved as an attractive tool to model crack initiation and propagation. Needleman[25] developed a cohesive model to describe void formation from initial particle debonding through complete separation to subsequent void growth. By reproducing a number of experimentally observed trends for elastic–plastic fracture in homogeneous and bimaterials systems, Tvergaard and Hutchinson[26–28] further increased the attractiveness of cohesive models. More recently, cohesive models of fracture have been increasingly used to study crack propagation in the condition that the plasticity is accounted for by modeling individual discrete dislocations. For instance, Cleveringa et al.[29] reported that the simulated mode-I fracture toughness of a homogeneous elastic–plastic solid increases with increasing the number of dislocations. O'Day and Curtin[30] have found that the bimaterial interfacial toughness increases with increasing mode mixity. Chng et al.[31] studied crack growth in a metal layer confined between two brittle media and reproduced experimentally observed trends in fracture toughness with respect to the thickness and residual stresses of the layer. Broedling et al.[32,33] also studied the fracture toughness of a confined metal layer. Bhandakkar et al.[34] studied the dislocation shielding of a cohesive crack; they have shown a surprising result: while the classical singular crack model predicts that, depending on the sign of the Burgers vector, a dislocation shields or antishields a crack, at low cohesive strengths a dislocation always shields a crack irrespective of the Burgers vector. The discrete dislocation–cohesive zone models suggested by Deshpande et al.[35,36] are well consistent with experimental observations.

As can be seen from the above literatures, most of the cohesive zone models are used to study the fracture of coarse grained materials rather than nc materials. The present article is aimed at studying the fracture toughness of nc materials by using a combined dislocation–cohesive zone model.

II. DISLOCATIONS-COHESIVE MODEL

In this article, we will focus on the dislocation shielding on a Dugdale[37] cohesive crack under mode-I dominant conditions (Fig. 1). The crack lies at the interface of two adjacent grains with the crack tip intersecting the GB of the third grain. The rectangular cohesive law of Dugdale, as shown in Fig. 1(b), is assumed to govern the traction within the cohesive zone OO′ of length L. The origin of the (x, y) coordinate system (point O) is located at the tip of the cohesive zone. With increased external load, N edge dislocations will emit from the crack tip, and it is assumed that the emitted dislocation is retarded at the opposite GB, as shown in Fig. 1(a).

The stress intensity factor K_{IC} at O is calculated as

$$K_{IC} = \sqrt{(K_{\mathrm{IC}}^{c})^2 - (k_{II}^{d})^2} - k_I^d \quad , \tag{1}$$

where K_{IC}^{c} is the intensity factor for the cohesive stress σ_{c}, and k_I^d and k_{II}^d are, respectively, the intensity factors for the normal (to the crack line) and shear stresses created by the generated dislocations. The mode-I stress intensity factor for a semi-infinite crack subjected to a distribution of normal traction $p(x)$ along the crack surfaces[34] is

$$K_I = \sqrt{\frac{2}{\pi}} \int_{-\infty}^{0} \frac{p(t)}{\sqrt{-t}} dt \quad . \tag{2}$$

In the Dugdale model, the cohesive traction is $p_{\mathrm{c}}(t) = -\sigma_{\mathrm{c}}$ for $-L \le t \le 0$ and vanishes over the rest of the crack faces, as shown in Fig. 1(b). The corresponding stress intensity factor at point O is[36]

$$K_I^c = -\sigma_c \sqrt{\frac{8L}{\pi}} \quad . \tag{3}$$

Note that in the Dugdale model, the cohesive stress σ_{c} is equal to the yield strength σ_{y}. However, all other cohesive zone models suggest that in the traction–separation process, the cohesive stress varies with the opening displacement, and there exists a maximum cohesive stress $\sigma_{\max}$, with value being a few times of the yield strength σ_{y}[38] In the present article, we will use Eq. (4), which provides a relation between the cohesive stress $\sigma_{\max}$ and the yield strength σ_{y}[39]

$$\sigma_{\max} = \frac{\sigma_y}{\sqrt{\alpha^2 - \alpha + 1}} \left[\frac{2E}{\sigma_y} \frac{\alpha + 1}{2 - \alpha} C e^{-1.5 H_0} + 1 \right]^n \quad , \tag{4}$$

where $\alpha = \sigma_{\mathrm{II}}/\sigma_{\mathrm{I}}$ is the biaxiality ratio that defines the stress triaxiality, which is shown in Fig. 2. H_0 is set to a constant value of 0.5, an average value chosen to represent the equivalent plastic strain failure locus for lower triaxilities encountered in plane stress, and C is a material dependent nondimensional parameter. σ_{y} can be obtained by the well-known Hall–Petch relationship:

$$\sigma_{\mathrm{y}} = \sigma_0 + \frac{k}{\sqrt{d}} \quad , \tag{5}$$

where σ_0 is the lattice friction stress, k is the H–P constant, and d is the grain size. It should be noted that (i) many nc materials do not obey the Hall–Petch relation when the grain size decreases to a critical value (3–20 nm)[40–43]; therefore, in this article, we will not consider the nc materials with a grain size below 20 nm, and (ii) we assume that the crack tip curvature radius $\rho = 0$, and it can be seen from Fig. 1(a) that in the cohesive zone, the maximum opening displacement $\delta_{\mathrm{c}} = L\beta$, where β is the crack tip opening angle. In general, 1 nm $\le \delta_{\mathrm{c}} \le$ 10 nm in the cohesive models[38]; thus, in the present article, we fix the

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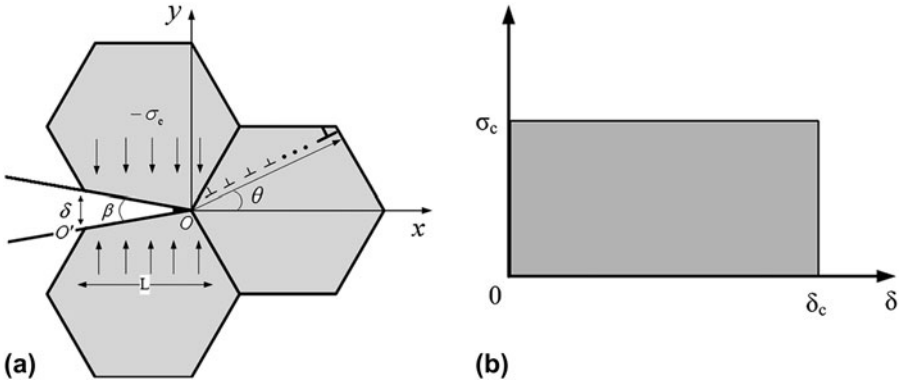


FIG. 1. (a) A Dugdale cohesive crack, subjected to a remote mode-I loading, lies at the interface of two adjacent grains with the crack tip intersecting the grain boundary (GB) of the third grain. *N* edge dislocations with Burgers vectors *b* emit from the crack tip and slip to the opposite GB, and the crack tip cohesive zone size is *L*. The cohesive zone starts at O′ where the crack-opening displacement is calculated. (b) Dugdale's (rectangular) cohesive law with cohesive strength σ_c and critical displacement δ_c.

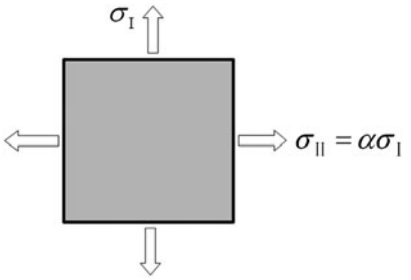


FIG. 2. Schematic of biaxiality ratio, α, under conditions of plane stress.

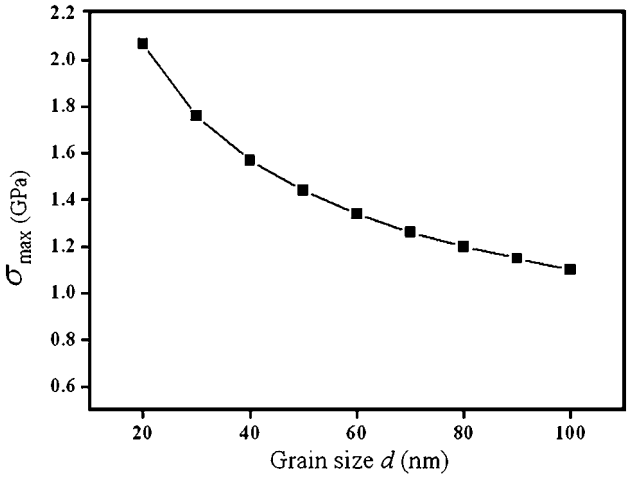


FIG. 3. Dependence of critical cohesive stress σ_{max} on grain size *d*.

cohesive zone $L \leq 100$ nm, and the suggested model is questionable for the solids with grain sizes that exceed a hundred nanometers. By putting $\sigma_0 = 25$ MPa and $k = 0.11 \times 10^6$ for nc copper into Eq. (4), we obtained the dependence of σ_{max} on grain size *d* for nc copper, as shown in Fig. 3.

Next, we will calculate the stress intensity factors k_I^d and k_{II}^d. In doing so, as shown in Fig. 1(a), θ is the angle made by the boundary plane (the dislocation Burgers vector) and the crack propagating direction. The stress intensity factors k_I^d and k_{II}^d for the dislocation-induced stresses can be calculated by[44]

$$k_I^d = -\frac{3\pi D b_T \sin\theta\cos(\theta/2)}{\sqrt{2\pi r_0}},$$
$$k_{II}^d = -\frac{\pi D b_T[\cos(\theta/2) + 3\cos(3\theta/2)]}{2\sqrt{2\pi r_0}} \quad , \qquad (6)$$

where $D = G/[2(1-\nu)]$, *G* is the shear modulus; $b_T = Nb$ is the sum of all the emitted dislocations; $b = 0.256$ nm denotes the Burgers vector magnitude of the edge dislocation; ν is the Poisson's ratio; and r_0 is the distance between the edge dislocation and the crack tip.

Let us calculate *N*—the number of dislocations emitted from a crack tip and retarded at the opposite GBs. In the quasi-equilibrium state, the Burgers vector magnitude *N*b corresponds to a minimum of the energy change Δ*W* within the dislocation generation process (Fig. 1). Δ*W* has three terms:

$$\Delta W = W_s + W_c - A \quad , \qquad (7)$$

where W_s is the proper elastic energy of the dislocation, W_c is the dislocation core energy, and *A* denotes the work spent to the generation of dislocations and the movement of dislocation from the crack tip to its final position.

Since the dislocation is located near the crack free surface, its stresses are screened. The proper energy of the dislocation can be calculated, with the aid of the formula taking into account the image forces that characterizes the screening effect,[45] as

$$W_s \approx \frac{D b_T^2}{2} ln \frac{r}{r_c} \quad , \qquad (8)$$

where r_c is the dislocation core radius. The dislocation core energy W_c under the standard approximation is given as[19]

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$$W_c \approx Db_T^2/2 \quad . \tag{9}$$

The work A spent on generation of the dislocation and the movement of the dislocation from the crack tip to its final position is calculated in the standard way. The work A is represented in the following form:

$$A = b_T \int_0^r \sigma_{r\theta}^{K_1}(r,\theta)d\rho \quad , \tag{10}$$

here $\sigma_{r\theta}^{K_1}(r,\theta)$, written in the polar coordinate system (r, θ) with the origin at the crack tip, can be given by

$$\sigma_{r\theta}^{K_I}(r,\theta) = \frac{K_I^c \sin\theta\cos(\theta/2)}{2\sqrt{2\pi r}} \quad . \tag{11}$$

According to Eqs. (10) and (11), one will obtain

$$A = b_T r\left(\frac{K_I^c \sin\theta\cos(\theta/2)}{\sqrt{2\pi r}}\right) \quad . \tag{12}$$

Then, with Eqs. (8)–(12), we obtain the following formula for the characteristic energy change ΔW

$$\Delta W = \frac{Db_T^2}{2}\left(ln\frac{r}{r_c}+1\right) - \frac{b_T r K_I^c \sin\theta\cos(\theta/2)}{\sqrt{2\pi r}} \quad . \tag{13}$$

With the condition $\partial(\Delta W)/\partial b_T = 0$ of the minimum of the energy change ΔW, one find the following expression for the number of emitted dislocations:

$$N = \frac{(1-\nu)K_I^c \sin\theta\cos(\theta/2)\sqrt{2\pi r}}{bG[\ln(r/r_c)+1]} \quad . \tag{14}$$

In our work, it is assumed that all the emitted dislocations will reach the opposite GBs; that is, each dislocation will slip the same distance, which equals to the grain size d.

In our previous article,[20] we have calculated the critical crack intensity factor K_{IC} as a function of θ for nc Ni, as shown in Fig. 4. It can be seen from Fig. 4 that the critical intensity factor K_{IC} reaches the maximum value when the boundary plane makes the angle θ of around 70° with the direction of crack growth. By putting θ = 70°, we obtained the relationship between N and d for nc copper, which is shown in Fig. 5. It shows that N increases with increasing the grain size. It should be noted that the value of N is relatively small, dislocation emission along one slip plane cannot lead to a significant crack blunting, and the crack can easily grow. This is a reason why the nc materials have relatively low fracture toughness. Generally speaking, in the case of $d < 20$ nm, one should take into account GBS,[18] diffusional creep,[46] rotational deformation,[47] and nanoscale twin deformation[14,15] effectively operating in nc materials that contribute to the blunting of cracks and increase strain-to-failure. This is another reason why our model is not suitable for nc materials with grain size below 20 nm.

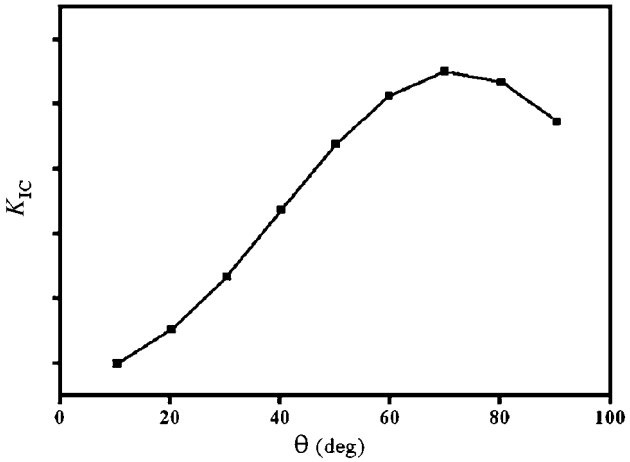


FIG. 4. Dependence of critical crack intensity factor K_{IC} on the angle θ made by the boundary plane and the crack growth direction for nanocrystalline (nc) nickel.

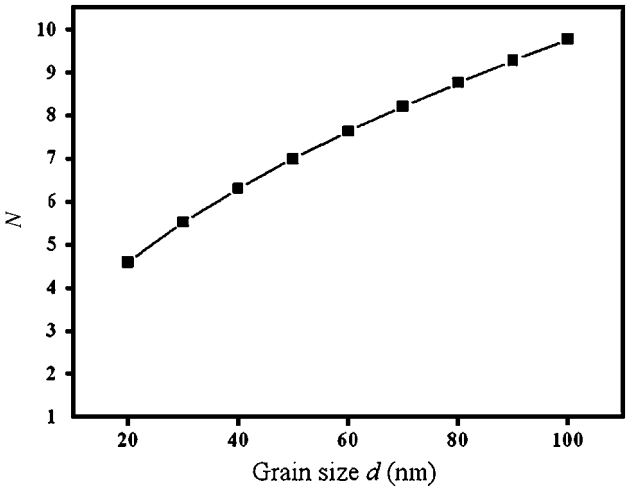


FIG. 5. The maximum number N of edge dislocations that can be emitted from the crack tip along one slip plane as a function of grain size d in nc copper.

III. RESULTS AND DISCUSSION

With Eqs. (6) and (3) substituted to Eq. (1), one will obtain the relationship between the fracture toughness and the grain size. Figure 6 shows our calculating result, which demonstrates that a decrease in grain size dramatically decreases K_{IC} and thus makes the nc materials much more brittle. This is in agreement with the experimental fact that some nc metals with a face-centered-cubic lattice exhibit a ductile-to-brittle transition when their grain size is decreased.[48–50]

We will make a further analysis of K_{IC} expressed in Eq. (1) by considering two cases: in case 1, we neglect the

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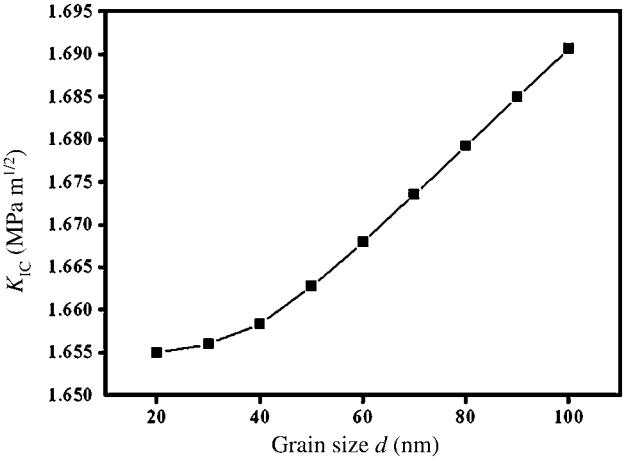


FIG. 6. Critical stress intensity factor K_{IC} versus grain size d in nc copper.

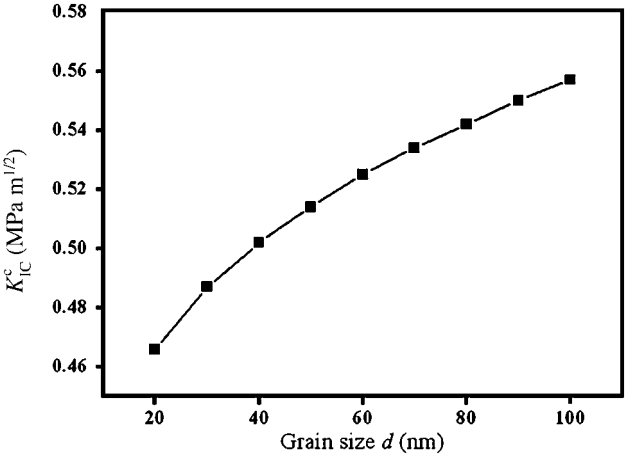


FIG. 7. Dependence of critical crack intensity factor K^c_{IC} on grain size d in nc copper.

shielding effect of the dislocations, and we can obtain the dependence of critical crack intensity factor K^c_{IC} on grain size d by using Eq. (3). The results, as shown in Fig. 7, suggest that K^c_{IC} increases with increased grain size d. This is untreatable since the cohesive stress σ_{max} has been reduced and the cohesive zone size L becomes larger, which results an increase of K^c_{IC}. In case 2, we only consider the shielding effect of dislocation and define

$$K^d_{IC} = k^d_{II} - k^d_I \quad , \tag{15}$$

where K^d_{IC} is the critical intensity factor induced by dislocations. The dependence of K^d_{IC}, for various grain sizes d, is shown in Fig. 8. As can be seen in this figure, K^d_{IC} decreases with increased grain size d. This can be explained by the following reasons: though the number of dislocations that emit from the crack tip increases with increased grain size, the slip distance (being equal to grain size in this article) of each emitted dislocation increases and leads to a decrease of K^d_{IC}. Comparing K^c_{IC} with K^d_{IC}, we know that K^d_{IC} mainly provides the fracture toughness, whereas K^c_{IC} is mainly responsible for the grain size effect.

Figures 6–8 suggest that the fracture toughness changes with grain size; however, the changes are not large. For instance, the fracture toughness only increases by 2.1%, when the grain size increases from 20 to 100 nm. It demonstrates that the fracture toughness of nc materials is not highly sensitive to the change in grain size. In addition, our calculated fracture toughness values are between 1.65 and 1.70 MPa/m$^{1/2}$, which are smaller than the experimental data. This can be explained by two reasons: (i) in the present model, we did not consider the dislocations already existing in real materials but the dislocations emitting from crack tip, and (ii) we did not consider the alternative deformations such as GBS, diffusional creep, rotational deformation, and nanoscale twin, which are capable of causing good toughness of nc materials under certain conditions.

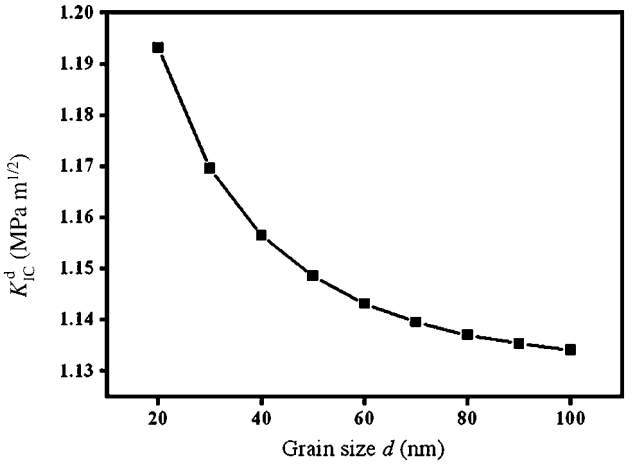


FIG. 8. Dependence of critical crack intensity factor K^d_{IC} on grain size d in nc copper.

IV. CONCLUSIONS

We have developed a theoretical model to study the critical stress intensity factor K_{IC} (which characterizes toughness) of nc materials by considering a typical case where crack lies at the interface of two neighboring grains, and the cohesive zone size is assumed to be equal to the grain size. Blunting and propagating processes of the crack are controlled by a combined effect of dislocation and cohesive zone. Edge dislocations emit from the cohesive crack tip and make a shielding effect on the crack. It was found that the critical stress intensity factor K_{IC} decreases with decreased grain size and leads a ductile-to-brittle transition. It should be noted that the K_{IC} of nc materials is not highly sensitive to the grain size d. For instance, increasing the grain size from 20 to 100 nm increases the value of K_{IC} only by 0.035 MPa/m$^{1/2}$. At the same time, the crack blunting and growth processes are strongly influenced or even controlled by alternate deformation modes (e.g., diffusional creep, rotational, GBS, and twin deformation); these deformation modes may cause high

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toughness of nc materials under certain conditions. However, the influence of alternative deformation modes on the toughness of nc materials is largely unknown and needs further study.

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REFERENCES

1. M.A. Meyers, A. Mishra, and D.J. Benson: Mechanical properties of nanocrystalline materials. *Prog. Mater. Sci.* **51**, 427 (2006).
2. T.D. Shen, C.C. Koch, T.Y. Tsui, and G.M. Pharr: On the elastic moduli of nanocrystalline Fe, Cu, Ni, and Cu–Ni alloys prepared by mechanical milling/alloying. *J. Mater. Res.* **10**, 2892 (1995).
3. K.S. Kumar, S. Suresh, and H. Van Swygenhoven: Mechanical behavior of nanocrystalline metals and alloys. *Acta Mater.* **51**, 5743 (2003).
4. C.C. Koch: Structural nanocrystalline materials: An overview. *J. Mater. Sci.* **42**, 1403 (2007).
5. L. Lu, L.B. Wang, B.Z. Ding, and K. Lu: High-tensile ductility in nanocrystalline copper. *J. Mater. Res.* **15**, 270 (2000).
6. M. Dao, L. Lu, R.J. Asaro, J.T.M. De Hosson, and E. Ma: Toward a quantitative understanding of mechanical behavior of nanocrystalline metals. *Acta Mater.* **55**, 4041 (2007).
7. Y.G. Liu, J.Q. Zhou, T.D. Shen, and D. Hui: Effects of ultrafine nanograins on the fracture toughness of nanocrystalline materials. *J. Mater. Res.* **26**, 1734 (2011).
8. Y.G. Liu, J.Q. Zhou, and X. Ling: Impact of grain size distribution on the multiscale mechanical behavior of nanocrystalline materials. *Mater. Sci. Eng., A* **527**, 1719 (2010).
9. B. Jiang and G.J. Weng: A generalized self-consistent polycrystal model for the yield strength of nanocrystalline materials. *J. Mech. Phys. Solids* **52**, 1125 (2004).
10. I.A. Ovid'ko: Review on the fracture processes in nanocrystalline materials. *J. Mater. Sci.* **42**, 1694 (2007).
11. J.D. Kuntz, G.D. Zhan, and A.K. Mukherjee: Nanocrystalline-matrix ceramic composite for improved fracture toughness. *MRS Bull.* **29**, 22 (2004).
12. Y. Zhao, J. Qian, L.L. Daemen, C. Pantea, J. Zhang, G.A. Voronin, and T.W. Zerda: Enhancement of fracture toughness in nanostructured diamond–SiC composites. *Appl. Phys. Lett.* **84**, 1356 (2004).
13. Y.T. Pei, D. Galvan, and J.T.M. De Hosson: Nanostructure and properties of TiC/a-C: H composite coatings. *Acta Mater.* **53**, 4505 (2005).
14. A. Singh, L. Tang, M. Dao, L. Lu, and S. Suresh: Fracture toughness and fatigue crack growth characteristics of nanotwinned copper. *Acta Mater.* **59**, 2437 (2011).
15. A. Singh, L. Tang, M. Dao, L. Lu, and S. Suresh: Enhanced fracture toughness and strength in bulk nanocrystalline Cu with nanoscale twin bundles. *Acta Mater.* **57**, 6215 (2009).
16. R. Mirshams, S.H. Whang, C.H. Xiao, and W.M. Yin: R-curve characterization of the fracture toughness of nanocrystalline nickel thin sheets. *Mater. Sci. Eng., A* **315**, 21 (2001).
17. M.Y. Gutkin and I.A. Ovid'ko: Homogeneous nucleation of dislocation loops in nanocrystalline metals and ceramics. *Acta Mater.* **56**, 1642 (2008).
18. S.V. Bobylev, A.K. Mukherjee, I.A. Ovid'ko, and A.G. Sheinerman: Effects of intergrain sliding on crack growth in nanocrystalline materials. *Int. J. Plast.* **26**, 1629 (2010).
19. I.A. Ovid'ko and A.G. Sheinerman: Ductile vs. brittle behavior of pre-cracked nanocrystalline and ultrafine-grained materials. *Acta Mater.* **58**, 5286 (2010).
20. Y.G. Liu, J.Q. Zhou, L. Wang, S. Zhang, and Y. Wang: Grain size dependent fracture toughness of nanocrystalline materials. *Mater. Sci. Eng., A* **528**, 4615 (2011).
21. S.V. Bobylev, N.F. Morozov, and I.A. Ovid'ko: Cooperative grain boundary sliding and migration process in nanocrystalline solids. *Phys. Rev. Lett.* **105**, 055504 (2010).
22. N.F. Morozov, I.A. Ovid'ko, A.G. Sheinerman, and E.C. Aifantis: Special rotational deformation as a toughening mechanism in nanocrystalline solids. *J. Mech. Phys. Solids* **58**, 1088 (2010).
23. W. Yang and H.T. Wang: Brittle versus ductile transition of nanocrystalline metals. *Int. J. Solids Struct.* **45**, 3897 (2008).
24. F. Yang and W. Yang: Crack growth versus blunting in nanocrystalline metals with extremely small grain size. *J. Mech. Phys. Solids* **57**, 305 (2009).
25. A. Needleman: A continuum model for void nucleation by inclusion debonding. *J. Appl. Mech.* **54**, 525 (1987).
26. V. Tvergaard and J.W. Hutchinson: The relation between crack growth resistance and fracture process parameters in elastic–plastic solids. *J. Mech. Phys. Solids* **40**, 1377 (1992).
27. V. Tvergaard and J.W. Hutchinson: The influence of plasticity on mixed mode interface toughness. *J. Mech. Phys. Solids* **41**, 1119 (1993).
28. V. Tvergaard and J.W. Hutchinson: On the toughness of ductile adhesive joints. *J. Mech. Phys. Solids* **44**, 789 (1996).
29. H.H.M. Cleveringa, E. van der Giessen, and A. Needleman: A discrete dislocation analysis of mode-I crack growth. *J. Mech. Phys. Solids* **48**, 1133 (2000).
30. M.P. O'Day and W.A. Curtin: Bimaterial interface fracture: A discrete dislocation model. *J. Mech. Phys. Solids* **53**, 359 (2005).
31. A.G. Chng, M.P. O'Day, W.A. Curtin, A.A.O. Tay, and K.M. Lim: Fracture in confined thin films: A discrete dislocation study. *Acta Mater.* **54**, 1017 (2006).
32. N.G. Broedling, A. Hartmaier, and H.J. Gao: A combined dislocation–cohesive zone model for fracture in a confined ductile layer. *Int. J. Fract.* **140**, 169 (2006).
33. N.G. Broedling, A. Hartmaier, and H.J. Gao: Fracture toughness of layered structures: Embrittlement due to confinement of plasticity. *Eng. Fract. Mech.* **75**, 3743 (2008).
34. T.K. Bhandakkar, A.C. Chng, W.A. Curtin, and H.J. Gao: Dislocation shielding of a cohesive crack. *J. Mech. Phys. Solids* **58**, 530 (2010).
35. V.S. Deshpande, A. Needleman, and E. van der Giessen: A discrete dislocation analysis of near-threshold fatigue crack growth. *Acta Mater.* **49**, 3189 (2001).
36. V.S. Deshpande, A. Needleman, and E. van der Giessen: Discrete dislocation modeling of fatigue-crack propagation. *Acta Mater.* **50**, 831 (2002).
37. D.S. Dugdale: Yielding of steel sheets containing slits. *J. Mech. Phys. Solids* **8**, 100 (1960).
38. N. Chandra, H. Li, C. Shet, and H. Ghonem: Some issues in the application of cohesive zone models for metal–ceramic interfaces. *Int. J. Solids Struct.* **39**, 2827 (2002).
39. I. Scheider, M. Rajendran, and A. Banerjee: Comparison of different stress-state dependent cohesive zone models applied to thin-walled structures. *Eng. Fract. Mech.* **78**, 534 (2011).

https://doi.org/10.1557/jmr.2011.442 Published online by Cambridge University Press

40. A.H. Chokshi, A. Rosen, J. Karch, and H. Gleiter: On the validity of the Hall–Petch relationship in nanocrystalline materials. *Scr. Metall. Mater.* **23**, 1679 (1989).
41. D.K. Kim and K. Okazaki: Processing of superconducting composite materials by electro-discharge compaction. *Mater. Sci. Forum* **88–90**, 553 (1992).
42. T.D. Shen, R.B. Schwarz, S. Feng, J.G. Swadener, J.Y. Huang, M. Tang, J.Z. Zhang, S.C. Vogel, and Y.S. Zhao: Effect of solute segregation on the strength of nanocrystalline alloys: Inverse Hall–Petch relation. *Acta Mater.* **55**, 5007 (2007).
43. G. Palumbo, U. Erb, and K.T. Aust: Triple line disclination effects on the mechanical behaviour of materials. *Scr. Metall. Mater.* **24**, 2347 (1990).
44. T.Y. Zhang and J.C.M. Li: Image forces and shielding effects of an edge dislocation near a finite length crack. *Acta Metall. Mater.* **39**, 2739 (1991).
45. I.H. Lin and R. Thomson: Cleavage, dislocation emission, and shielding for cracks under general loading. *Acta Metall.* **34**, 187 (1986).
46. I.A. Ovid'ko: Deformation and diffusion modes in nanocrystalline materials. *Int. Mater. Rev.* **50**, 65 (2005).
47. Y.G. Liu, J.Q. Zhou, T.D. Shen, and D. Hui: Grain rotation dependent fracture toughness of nanocrystalline materials. *Mater. Sci. Eng., A* **528**, 7684 (2011).
48. H. Li and F. Ebrahimi: Transition of deformation and fracture behaviors in nanostructured face-centered-cubic metals. *Appl. Phys. Lett.* **84**, 4307 (2004).
49. H. Li and F. Ebrahimi: Ductile-to-brittle transition in nanocrystalline metals. *Adv. Mater.* **17**, 1969 (2005).
50. F. Ebrahimi, A.J. Liscano, D. Kong, Q. Zhai, and H. Li: Fracture of bulk face centered cubic metallic nanostructures. *Rev. Adv. Mater. Sci.* **13**, 33 (2006).

https://doi.org/10.1557/jmr.2011.442 Published online by Cambridge University Press