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ON THE INFLUENCE OF THE ENVIRONMENT ON MODELING THE FATIGUE CRACK GROWTH PROCESS

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Summary

The effect of the ambient environment at room and elevated temperature is considered with respect to the influence exerted on the basic mechanical aspects of the fatigue crack growth process. An experimental assessment of this influence is obtained by conducting companion fatigue crack growth tests in air and vacuum and the results of such experiments are described. Topics considered include crack closure, short crack growth in notched and unnotched specimens, Mode II crack growth, and the effects of oxidation at elevated temperatures. It is shown that the basic mechanisms of fatigue crack growth can be greatly altered by the presence of oxide films at the fatigue crack tip. Modeling the mechanical aspects of the crack growth process is of itself a challenging task. The addition of environmental considerations adds to the complexity of the modeling process.

Introduction

The modeling of the fatigue crack growth process is generally an a posteriori process which depends strongly upon experimental observations for guidance. For example, Forsyth and Ryder's observation of fatigue striations and Elber's observation of crack closure have had significant effects on the modeling of the fatigue crack growth process. Most of the experimental work, however, has been done in environments which themselves may have had an influence upon a result without that fact being fully realized. For example, theories may attempt to predict the experimentally determined exponent of the Paris law without taking into account the possible influences of the ambient environment on the experimental value. It seems evident therefore that if the mechanical aspect of the process is to be modeled, that the guiding experimental observations be obtained under conditions where the environment does not play a role. By carrying out companion tests in an environment of interest an assessment of the influence of the environment can then be made. In this paper examples will be given of the influence of environment on crack closure and crack growth characteristics by comparing results obtained in vacuum with those obtained in air both at room and elevated temperature.

Keywords: Fatigue crack growth, environmental effects, crack closure, oxidation, elevated temperatures.

Crack Closure

Fig. 1 is a schematic depiction of the fatigue crack growth plot. The lower bound, designated as ΔKth, corresponds to a growth rate of 10^{-8} mm per cycle. The upper bound is set by the fracture toughness of the material under cyclic loading conditions. The right band ordinate gives the time required for a 1-mm increment of crack growth at a fixed frequency, and is included to emphasize the long times involved at low crack growth rates. Since corrosion is a time dependent process we can expect that the effects of corrosion may be more pronounced in the near-threshold region than at higher growth rates. The intermediate region, or Paris regime, is the region which in the past had been most closely studied and modeled and a review of this work is given in ref. 1. More recently, attention has been focused on the near-threshold region. In the near-threshold region, crack closure under plane strain conditions can be important at low R values in reducing the level of ΔK_{eff}, where ΔK_{eff} is defined as $K_{max}-K_{op}$ and K_{max} is the maximum stress intensity factor in a loading cycle and K_{op} is the stress intensity factor at which the crack tip exceeds the closure effect and begins to open in the rising part of the loading cycle. The overall crack closure process can be influenced by surface roughness which includes the effects of crack branching and crack deflection (2), as well as by oxidation and fretting (3-5). An example of the dependency of K_{op} on the range of the stress intensity factor ΔK, for R=0.05 loading conditions is shown in Fig. 2 for several ferritic steels tested at room temperature in air (50% relative humidity) or in vacuum (3×10^{-5} Torr) (6). It is seen that in the near-threshold region in vacuum the level of closure is insensitive to ΔK. The materials of these tests had relatively fine microstructures as the result of heat treatment and the K_{op} levels obtained

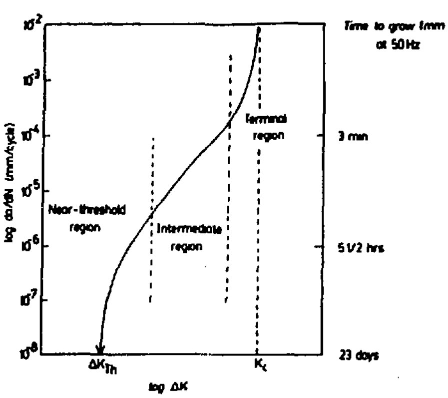

Fig. 1. Schematic of rate of fatigue crack growth as a function of ΔK. Ordinate scale at the right indicates the time required to increase crack length by 1 mm at a given crack growth rate.

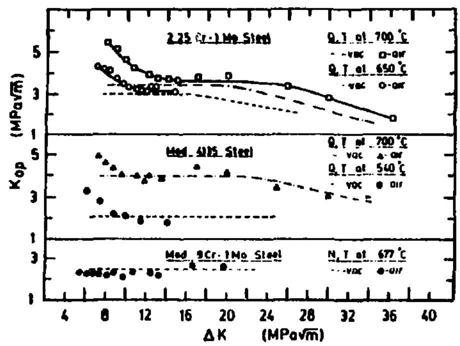

Fig. 2. K_{op} as a function of ΔK for three ferritic steels in vacuum and in air at room temperature.

reflect the roughness of the fracture surfaces which in turn relates to microstructure. In this range Mode II deformation is important; in normalized plain carbon steels in which the ferrite regions are larger, large facets can be developed but that is not the case for the steels of Fig. 2. Incidentally it is sometimes thought that rewelding might occur in vacuum testing, but no evidence of rewelding of the fracture surfaces was found in these tests.

The closure measurements made in the ambient environment reveal some interesting differences. For example both the 2-1/4Cr-1Mo and the 4135 steels showed a rapid rise in the K_{op} level close to the threshold value where fretting was obvious. In the case of the 4135 material this increment in closure could be termed "oxidation-induced" as discussed by Stewart (3) and Suresh et al (4). In this region microstructural detail on the fracture surfaces was eliminated by the effects of oxidation and fretting. For the 2-1/4Cr-1Mo alloy the fracture appearance differed. Microstructural features could be identified but tbey were coarser in air than in vacuum. The work of Majumdar and Chung (7) has shown that oxygen can localize the slip process in ferrite, and it is thought that the "oxygen-influenced" enhanced closure is due to this localization effect of oxygen which is promoted by fretting and the long times spent in establishing the threshold level under decreasing ΔK conditions. A point to be determined is whether these higher closure levels in air are also present in a test conducted under rising ΔK conditions in this region.

The modified 9Cr-1Mo alloy showed little sensitivity to the environment, a fact attributed to the higher chromium content of this steel as compared to the others.

The effects of closure and environment on the overall fatigue crack growth plot are indicated in Fig. 3 based upon studies of ferritic steels. In vacuum four regions can be identified if ΔK is used as the controlling variable. However if the ΔK_{eff} is used then regions II and III merge. In

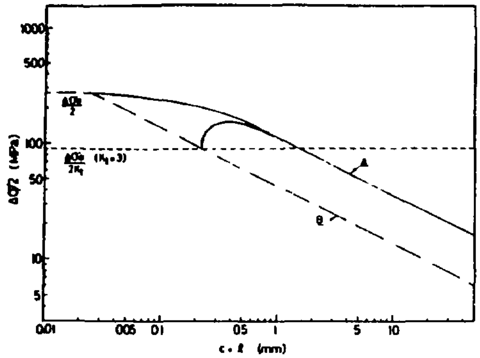

Fig. 4. Threshold condition as a function of notch size, c, and crack length, l. Line A corresponds to the macroscopic threshold, curve B is the line along which ΔK_{eff} is constant. The stress required to propagate a crack from a hole is indicated by the line joining lines B and A.

basis for the development of non-propagating fatigue cracks in the stress amplitude range between the dotted line and the maximum of the rising curve. This circumstance is also related to the fatigue notch size effect and to notch sensitivity as the extent of the rise varies inversely with strength level.

In order to model this process more quantitatively information on the rate of development of crack closure with crack advance is needed. Fig. 5 shows the result of one such experiment (10). In modeling the development of closure process the following expression was proposed (11):

$$K_{op} = K_{op\,max}\,(1-e^{-kl})$$

where $K_{op\,max}$ represents the closure level of a macroscopic crack, k is an experimentally determined material constant of dimensions mm^{-1} which reflects the rate of increase of closure with crack advances, and l is the crack length. To carry out this experiment it was necessary first to grow a crack, then to remove the closure by heat treatment and then to reinitiate the crack and determine the closure level as a function of crack length. Fig. 6 shows how the development of closure can influence the shape of the da/dN vs. ΔK crack growth plot, with behavior such as that shown at A being observed in the experiments related to Fig. 5, and curves B and C relate to other possible types of da/dN vs ΔK behavior for newly formed fatigue cracks. These considerations lead to the overall representation shown in Fig. 7 wherein the region of continuum mechanics is separated from that of micromechanics, a region of importance in considering the growth of short cracks in unnotched specimens.

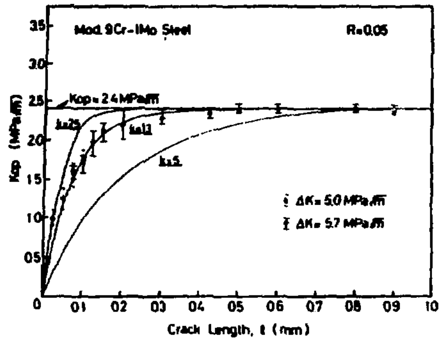

Fig. 5. Comparison of computed and experimental values of K_{op} as a function of crack length. For Mod. 9Cr-1Mo steel the appropriate value of k is 13 mm^{-1}.

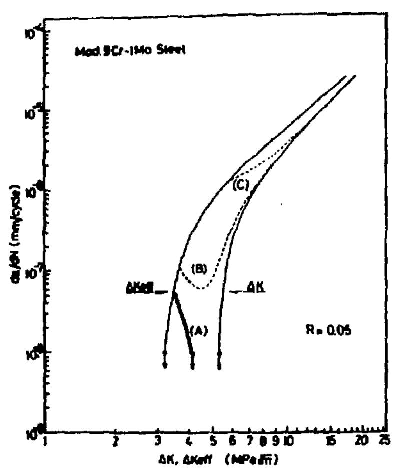

Fig. 6. Observed and calculated types of anomalous crack growth behavior.

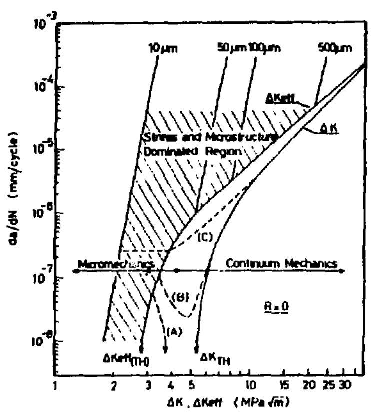

Fig. 7. The rate of crack growth as a function of ΔK for 9Cr-1Mo steel. Continuum-mechanics and micromechanics ranges indicated.

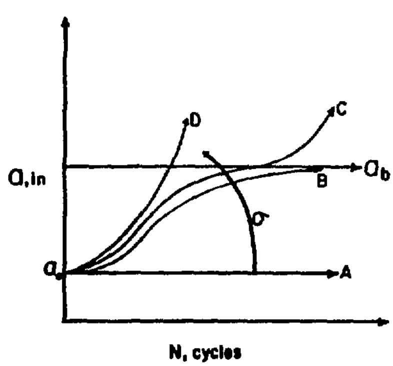

Fig. 8. Schematic of influence of a grain bonding and stress amplitude on fatigue crack growth in the vicinity of the boundary. Lines A-D represent various types of crack growth behavior as influenced by a grain boundary at a length of a_b. The stress amplitude increases from A to D .

Short Crack Growth - Unnotched Specimen

Fig. 8, based upon the observations of Forrest and Tate (12) shows the growth behavior of small surface cracks as influenced by a grain boundary. Lines A, B, C and D represent possible a vs. N plots, with the stress amplitude increasing from A to D. The line a_b represents the distance to the first grain boundary. Under appropriate stress conditions a grain boundary may retard or even arrest the propagating crack. This influence of the grain boundary is most pronounced at very low crack growth rates where the driving force for propagation is weak. The effect can be seen on a more macroscopic level as shown in Fig. 9 for a γ-Mo directionally solidified eutectic alloy (13). As the fatigue crack approaches a grain boundary the crack growth rate drops and in fact a reinitiation process in the next grain may be needed. Recent studies have drawn further attention to this *early stage of crack growth* with results such as those shown in Fig. 10 (14). The crack slows down as it approaches the boundary at a crack length of 10-16 microns and then accelerates once it has penetrated into the next grain. The following expression simulates the growth pattern in the first grain:

$$da/dN \;\alpha\; (\sigma-\sigma_{01})^2 a - (\sigma_{02}-\sigma_{01})^2 a_b \left(\frac{a}{a_b}\right)^m \left(\frac{a_o}{a_b}\right)^n \qquad \sigma_{02}>\sigma_{01}$$

where σ_{01} is the endurance strength of the first grain and σ_{02} is that of the second, and a is the crack length at the grain boundary. Here m was taken to be 4 and a_0 was taken to be 16 μm. It is noted that if the macroscopic da/dN vs ΔK and ΔK_{eff} (15) are plotted, the experimental results fall within these bounds. This can be interpreted to indicate that a boundary can inhibit crack opening because of limited plastic deformation in the next grain which in effect reduces the ΔK_{eff} range in the same sense that closure does. This is an interesting area for further study, particularly with respect to the influence of the environment.

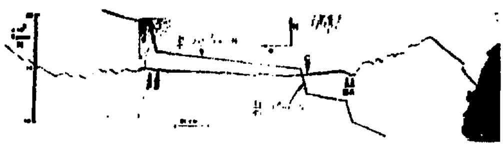

Fig. 9. Fatigue crack growth in a γ'-Mo directionally solidified eutectic alloy. Crack growth from right to left.

Mode II Growth

A dramatic example of the importance of the environment is furnished by results of Mode II tests of single crystals of MAR M-200, a nickel base superalloy (16). Due to the influence of the γ' phase, the glide process is extremely planar in this material. When this crystal was oriented such that the Burgers vector was in the direction of shear and the slip plane was in the plane of shear, Mode II fatigue crack growth occurred in air. However in vacuum, although some cracking occurred along the side faces, overall crack growth was arrested. Fig. 11 shows the appearance of the crack surfaces for these two cases. For the vacuum test the fracture

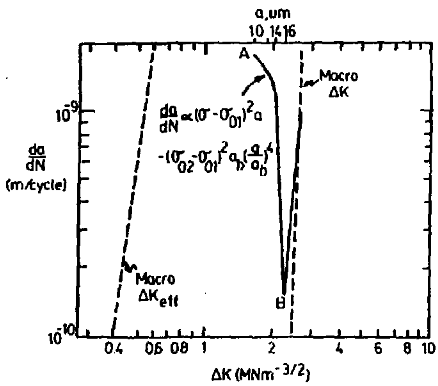

Fig. 10. Influence of first grain boundary encountered by a fatigue crack on the role of fatigue crack growth in the aluminum alloy 7075-T6.

except along the edges of the specimen was developed by pulling the specimen apart. In air the crack was able to propagate by the repeated rupture of the oxide that kept forming at the crack tip. In vacuum, crack growth occurs by an extrusion process which was unable to operate for this particular orientation. Clearly in this case the environment exerted a tremendous effect on the fatigue crack growth mechanism.

Oxidation at Elevated Temperatures

Oxidation can exert a strong influence on fatigue crack growth behavior at elevated temperatures. For example compare the fatigue crack growth curves for the titanium alloy Ti-6Al-2Sn-4Zr-2Mo-0.1Si obtained in air and in vacuum at 811K (538°C, 1000°F) shown in Fig. 12 (17). In vacuum the curve is similar in shape to that obtained at room temperature in air or in vacuum. However at 811K the curve for air differs markedly in shape from that obtained in vacuum. In other studies of this type of alloy at elevated temperatures in air we have noted that the crack closure level is about 3/8 of K_{max} at threshold and is constant in the near-threshold region (15). Therefore we do not attribute the change in shape of the curves in going from vacuum to air to crack closure alone. Rather we think that the steep transition in air from no growth just below threshold to growth on the order of 10^{-5} mm per cycle just above threshold is due to the rupture of the oxide film formed at the tip of the crack. Below threshold the

11a

11b

Fig. 11. Fracture surface appearance of single crystals of MAR M-200 nickel-base superalloy tested in Mode II cyclic loading.
a. Fracture surface of specimen tested in air.
b. Fracture surface of specimen tested in vacuum. Note that fatigue cracking is confined to thin layers along the sides of the specimen. The remainder of the fracture surface is due to tensile fracture.

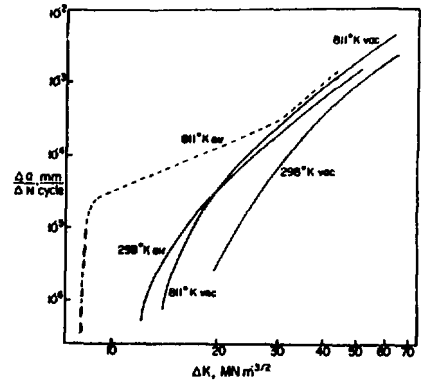

Fig. 12. Fatigue crack growth behavior for a titanium alloy tested in air and vacuum at room and elevated temperatures.

crack tip strain is insufficient to rupture the oxide, whereas above threshold the crack tip strain can rupture the oxide and then allow the crack to propagate. During propagation in air, the rate of propagation is increased by about 10^{-5} mm per cycle, a relatively large increment near threshold but a relatively small increment at high values of the stress intensity factor where the curves for air and vacuum conditions come closer together.

In the case of ferritic steels crack closure in air at elevated temperatures can be more important than in titanium alloys. As shown in Fig. 13 at 538°C the rates of crack growth for the ferritic alloys 2-1/4Cr-1Mo and 9Cr-1Mo are the same in vacuum for a given value of ΔK. Under these test conditions crack closure is virtually absent due to the leveling of contacting asperities by creep during portions of the cycle when the opposing fracture faces are in contact. In air however the behaviors differ. Both exhibit thresholds in air above those in vacuum and both exhibit a sharp rise in the da/dN curves above threshold. The K_{op} level for the 9Cr-1Mo alloy is 1/3 of K_{max}, whereas that for the 2-1/4Cr-1Mo alloy is about 0.9 of K_{max}. Since the oxidation process leading to closure is time dependent there is also a time dependence associated with the threshold level, particularly in the case of the 2-1/4Cr-1Mo alloy. For example after a decreasing ΔK test to threshold it was necessary to increase the ΔK level by 50% above the initial threshold level before the crack began to propagate, for during the hours spent at the initial threshold the oxide had thickened to reduce further the range of K_{eff}. In the case of the 9Cr-1Mo steel the thickening rate of the oxide was less and a pronounced time dependency of the crack behavior was not observed. Because time of exposure was of lesser importance in the case of the 9Cr-1Mo steel as compared to the 2-1/4Cr-1Mo steel, an increasing ΔK test following a decreasing ΔK test resulted in the same da/dN plot. However the frequencies employed were in the range of 30-50 Hz, and in subsequent tests with this alloy we have observed that the crack growth rate increases with decrease in frequency.

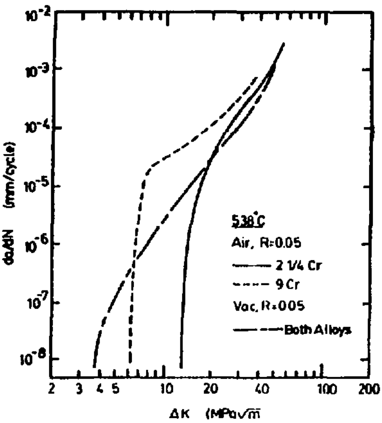

Fig. 13. Fatigue crack growth behavior at 538° of two ferritic steels in air and in vacuum (19).

In the analysis of fatigue crack growth tests at elevated temperature one also has to consider the possible effects of creep on the crack growth process. This will clearly be a factor at very low frequencies where creep crack growth can occur. However in fatigue crack growth tests which we have carried out at frequencies on the order of 30 Hz creep does not appear to be a factor. In tests in vacuum crack growth rates increase with increase in temperature. If the crack propagation process is modeled in terms of a crack opening displacement process it develops that the principal material factor involved is Young's modulus. If we take into account the temperature dependence of Young's modulus then it is possible to express the rate of crack growth as a single-valued function of the strain intensity factor as shown in Fig. 14 (18). In air however the situation is more complex because of the temperature dependence of the oxidation process.

Concluding Remarks

The foregoing examples have been intended to indicate how the environment can influence the basic mechanical mechanisms involved in the fatigue crack growth process. After much research there is still a lack of agreement with respect to the mechanical aspects of the process, an indication of the complexities involved. When we then attempt to include the effects of the environment in a detailed mechanistic sense we are confronted with a formidable task indeed. In each loading cycle new surface is presented to the environment continuously above the K_{op} level so that oxide films may be continually forming, rupturing and reforming in a single cycle at the crack tip. In addition hydrogen embrittlement or anodic dissolution processes may be occurring. Much phenomenological information has been gathered

REFERENCES

1. A. J. McEvily, "On the Quantitative Analysis of Fatigue Crack Propagation," pp. 283-312 in Fatigue Mechanisms, J. Lankford et al. eds, ASTM STP 811, 1983.

2. S. Suresh, Crack Deflection: Implications for the Growth of Long and Short Fatigue Cracks, Met. Trans., Vol. 14A, 1983, p.2375.

3. A. T. Stewart, The Influence of Environment and Stress Ratio on Fatigue Crack Growth at Near Threshold Stress Intensities in Low Alloy Steels", pp. 463-478, Eng. Fracture Mech., 13, 1980.

4. S. Suresh, G. F. Zaminski, and R. O. Ritchie, "Oxide-Induced Crack Closure: An Explanation for Near-Threshold Corrosion Fatigue Crack Growth Behavior", pp. 1435-1443 Met. Trans. vol. 12A, 1981.

5. S. Suresh and R. O. Ritchie, "Near-Threshold Fatigue Crack Propagation: A Perspective on the Role of Crack Closure", in Fatigue Crack Growth Threshold Concepts, pp. 227-261, ed. by D. Davidson and S. Suresh, AIME, Warrendale, PA., 1984.

6. W. Zhu, K. Minakawa, and A. J. McEvily, "On the Influence of the Ambient Environment on the Fatigue Crack Growth Process in Steels", submitted to Eng. Fracture Mech.

7. D. Majumdar and Y. Chung, "The Effects of Oxygen on Fatigue Deformation of Iron," pp. 791-794, Scripta Met., Vol. 16, 1982.

8. R. A. Schmidt and P. Paris, Threshold for Fatigue Crack Propagation and the Effects of Load Ratio and Frequency," pp. 79-94, in ASTM STP 536, 1973.

9. A. J. McEvily and K. Minakawa, "Crack Closure and the Conditions for Fatigue Crack Propgation," pp. 517-530 in Fatigue Crack Growth Threshold Concepts, ed. by D. L. Davidson and S. Suresh, AIME, Warrendale, PA., 1984.

10. K. Minakawa, H. Nakamura, and A. J. McEvily, "On the Development of Crack closure with Crack Advance in a Ferritic Steel," pp. 1371-1374, Scripta Met., Vol. 18, 1984.

11. A. J. McEvily, K. Minakawa, and H. Nakamura, "Fracture Mechanics, Microstructure, and the Growth of Short and Long Fatigue Cracks," pp. 215-233, in Fracture ed. by J. M. Wells and J. D. Landes, AIME, Warrendale, PA, 1984.

12. P. G. Forrest and A.E.L. Tate, p. 438, J. Inst. Metals, Vol. 93, 1964-65.

13. C. Hoffman, P. Bhowal, and A. J. McEvily, "Fatigue Crack Growth in Co and Ni Base Directionally Solidified Eutectic Composites," pp. 69-83, in In Situ Composites IV ed. by F. Lemkey et al. Elsevier Science Publishing Co., 1982.

14. J. Lankford, "The Growth of Small Fatigue Cracks in 7075-T6 Aluminum Alloy," pp. 233-248, Fatigue Eng. Mat. and Structures, Vol. 5, 1982.

15. K. Minakawa and A. J. McEvily, Unpublished results.

16. O. Y. Chen and A. J. McEvily, unpublished results.

17. J. A. Ruppen and A. J. McEvily, "The Effect of Elevated Temperature and Environment on the FAtigue Crack Growth Characteristics of Ti-6Al-2Sn-4Zr-2Mo-0.1Si," pp. 63-72, Fat. of Eng. Mats. and Structures, Vol. 2, 1979.

18. A. J. McEvily and R. P. Wei, "Fracture Mechanics and Corrosion Fatigue", pp. 381-393, in Corrosion Fatigue, ed. by O. Devereux et al., NACD, Houston, 1972.

19. R. Nakamura, K. Murali, K. Minakawa, and A. J. McEvily, "Fatigue Crack Growth in Ferritic Steels as Influenced by Temperature and Environment", to be published in the Proc. Int. Symp. Microstructure and Mechanical Behavior of Materials, Xian, China, 1985.