

Communications

Observations of Stress Corrosion Crack-Front Bowing in High-Strength Aluminum Alloy Double Cantilever Beam Specimens

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The bolt-loaded double cantilever beam specimen, DCB (Fig. 1) has been used in the past for measuring the growth rates of stress corrosion cracks in high strength aluminum alloys (*e.g.*, Refs. 1 to 4). In a number of previous studies, the crack progress was optically monitored on the sides of the specimens. These crack lengths were used to calculate the dependence of crack growth rate on the applied stress intensity at the crack tip.

Examination of broken open high strength aluminum alloy DCB stress corrosion specimens has revealed their innate propensity toward crack front bow-out, or tunneling. This phenomenon may be due in part to the change in the state of stress with depth in the sample and to the presence of surface residual stresses in the wrought product from which the DCB was machined.^{1,2} Whatever the cause, the development of crack front bowing during the stress corrosion test would lead to errors in the crack growth rate (da/dt) determination, especially if the bow out increased at a greater rate than did the external crack length. The purpose of this communication, then, is to display an example of how the crack front may develop internally during the testing of aluminum alloy DCB specimens.

The specimen used herein was part of a study being conducted in-house at the Air Force Materials Laboratory on the reliability of high strength aluminum alloys.⁵ Its composition is Al, 5.68 Zn, 2.13 Mg, 1.51 Cu, 0.04 Fe, 0.02 Si, 0.12 Zr, 0.005 Ti, and <0.01 wt pct of Cr, Ni, and Mn. It was processed as 32 mm thick plate to the standard T651 temper. The longitudinal and long-transverse tensile yield strengths of plate from the same ingot were both 75 ksi (516 MPa).⁶ The DCB specimen, machined to the specifications of Fig. 1, was tested in constant immersion in 3.5 pct NaCl solution. The plane of the crack lay along the rolling direction normal to the plate surfaces. Four values of crack length were obtained during each measurement procedure. Two values were measured on a calibrated optical microscope on alternate sides of the DCB, and the other two were measured on the sides parallel to the plane of the crack using an ultrasonic pulse-echo nondestructive inspection technique. The pulse-echo technique had an empirically determined uncertainty of about ± 0.03 in. (± 0.76 mm).

The crack front historical profile for the selected DCB specimen was plotted in Fig. 2 assuming that the

longest internal crack length occurred at the center of the specimen. Table I lists the data used in plotting Fig. 2. Optical "Side 1" data was plotted on the left of the figure, "Side 2" data was plotted on the right, and the average of the two ultrasonic values was plotted in the center. A best naked-eye fit was drawn in each case. It should be noted that the crack position

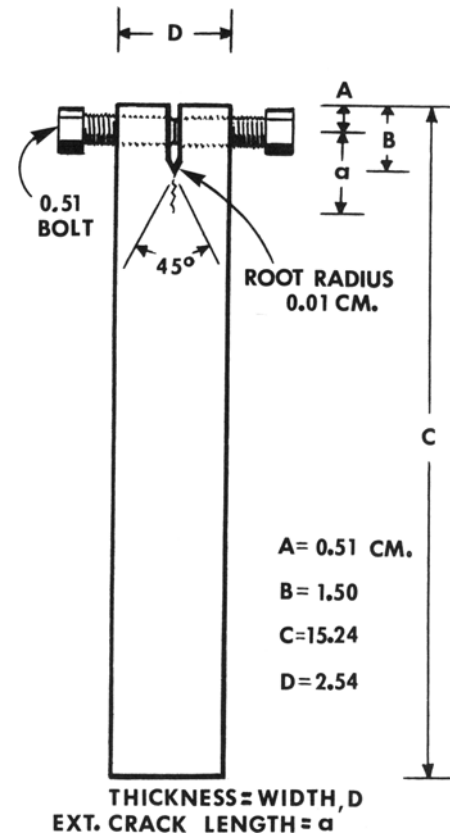


Fig. 1—The double-bolt-loaded double cantilever beam stress corrosion specimen.

Table I. Crack Growth Measurements of a High-Strength Aluminum Alloy DCB Specimen

Time, h	Sides Optical Measurements, mm		Center Ultrasonic Measurements, mm	
	Side 1	Side 2	Top	Bottom
0*	15.0	12.7	†	†
50	20.3	21.1	23.3	23.4
96	22.1	24.1	26.3	27.8
168	24.6	26.6	28.2	27.8
241	25.2	26.7	31.1	28.4
361	26.3	27.3	33.2	32.8
432	28.1	28.7	35.6	35.3
528	30.4	30.8	36.7	36.7
743	31.0	31.5	37.4	37.1
910	31.4	32.4	37.7	37.3
1102	31.9	32.4	38.1	37.8
1506	35.0	32.7	38.3	39.6
Broken Specimen Optical Data				
1506	36.5	32.3	38.8	

*"Pop-in" crack length.

† Too short to measure with available test stand.

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along the left and right sides of the specimen displayed several instances of arrest and acceleration while the center position advanced relatively uniformly. After 50 h of immersion the difference between the minimum external crack length and the center bow-out point was 0.113 in. (2.87 mm), while the 1102 h difference was 0.203 in. (5.16 mm). The internal crack shape after 1506 h is shown in Fig. 3 where the specimen was broken open by turning the loading bolts. Studying the stress intensity factor scales at the left of Fig. 2 will show the magnitude of the potential errors in using external crack lengths in reporting (da/dt) data. The values on the scale were calculated using the crack opening displacement compliance curve for the DCB specimen¹

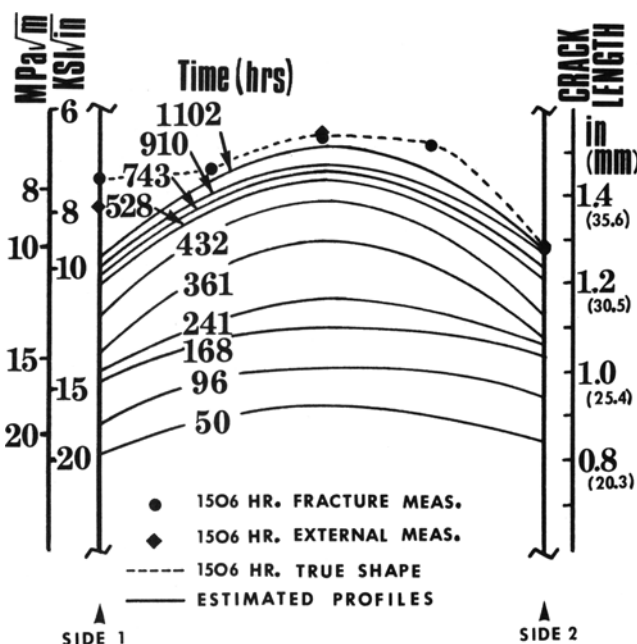


Fig. 2—Crack front profiles as a function of time in a DCB specimen. The last set of side and center readings before breaking the specimen are represented by diamond symbols. Solid circles represent optical measurements on the broken specimen.

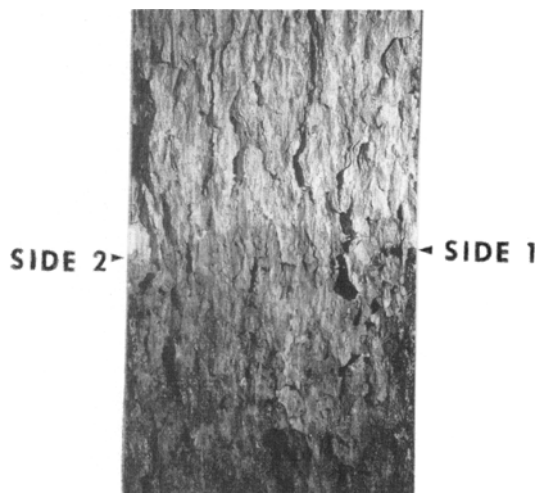


Fig. 3—Post-test appearance of the DCB specimen of Fig. 2 and Table I showing the bowed crack front (darker shaded).

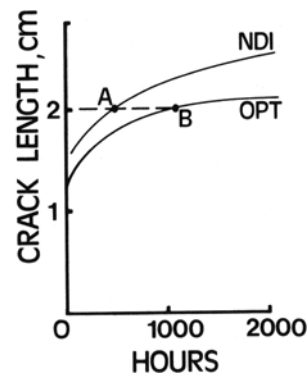


Fig. 4—Typical behavior of a growing crack in a high strength aluminum DCB specimen.

based upon the crack lengths at the right hand scale of Fig. 2.

Another means of displaying the information from Fig. 2 is the more conventional plot in Fig. 4 where cracklength is shown as a function of time in the salt bath. The lower curve depicts optical surface crack measurements and the upper curve depicts the internal crack measurements obtained by nondestructive techniques. Fig. 4 shows how optical data alone may not characterize the stress corrosion crack advance. Not only does the difference between the curves increase with time, but the value of the growth rate (da/dt) at any given value of applied stress intensity (calculated from the instantaneous cracklength) is greater internally than at the surface. For example, consider the slopes of the two curves at points A and B in Fig. 4. It is not known why the bow-out increases as the stress corrosion crack grows (*i.e.*, the stress intensity decreases); however, the possibility of corrodant concentration with time at the center of the crack front should be considered. Further work in this area is indicated.

The proper choice of a crack length to calculate the effective stress intensity factor in the presence of a curved through-crack remains a subject of considerable controversy. It was not the purpose of this short communication to address that subject, but only to provide data supporting the contention that external measurement of DCB crack lengths could lead to certain types of errors in evaluating stress corrosion crack growth rate.

Careful analysis of crack front bow-out development would also provide insight from the standpoint of surface creation during stress corrosion cracking. Of course the amount of surface created would depend not only on the geometrical properties of the crack front, but also on the topography of the new surface. Nevertheless, it is felt that information gained by combined optical and ultrasonic techniques could contribute to the understanding of the stress corrosion mechanism in these and other alloys.

The technical assistance of Messrs. Marlin Cook and Charles Smith, University of Cincinnati, is gratefully acknowledged. This project was part of an Air Force Materials Laboratory, Metals and Ceramics Division, Structural Metals Branch, in-house effort investigating the influence of purity level on microstructure-property relationships in high strength aluminum alloys.

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Increasing the Fracture Toughness of a Maraging Steel Type Alloy

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Recent unpublished studies¹ have shown that the fracture toughness of a Fe-16Ni-1.5Ti maraging steel type alloy may be greatly improved by age hardening at higher temperatures (550°C to 600°C) than those conventionally used (~450°C) for maraging steels. The purpose of this note is to describe the mechanisms believed to be responsible for the improved fracture toughness in specimens aged at the higher temperature.

In this alloy, austenite can form at all the aging temperatures investigated (450°C to 600°C); whether or not it does form is strongly dependent on the time at temperature, as shown in Fig. 1, which shows the volume percent of austenite formed as a function of time of aging at 450, 500, 550 and 600°C. The data in Fig. 1 is for the isothermal rate of formation of austenite at the aging temperature. This is not necessarily the same as the rate of formation of retained austenite measured at room temperature subsequent to aging. Thus there is *no measurable retained austenite* in the Fe-16Ni-1.5Ti alloy for any of the aging temperatures and times reported in Table I (450°C 1 h, 550°C 4 min and 600°C 10 min), despite the large amount of austenite formed at temperature during aging for the higher aging temperatures. This result is based on a combination of X-ray and transmission electron microscopy techniques for measuring retained austenite. The formation of austenite in Fe-Ni-Ti alloys has been discussed by Jin² and Peters.³ The data in Fig. 1 was obtained by a magnetic technique reported in detail by Yuen.⁴ It was

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observed that the austenite forms initially at the prior austenite and martensite lath boundaries,¹⁻³ as it has similarly been observed to do in more conventional maraging steels.⁵

The precipitate in this alloy, hexagonal Ni₃Ti (DO₂₄), does not appear to precipitate in austenite. For example, in specimens aged at 650°C (a temperature at which a considerable amount of austenite is formed very rapidly) no precipitation is observed in the austenite; precipitation occurs only in the martensitic

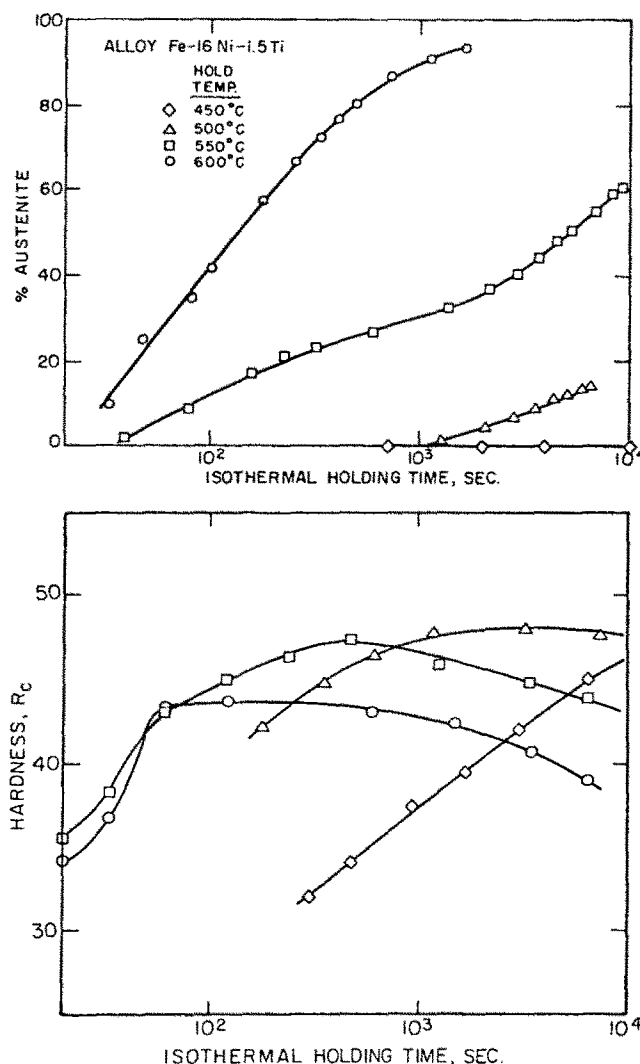


Fig. 1—The vol pct austenite and the Rockwell C hardness of the 16-Ni-1.5Ti alloy as a function of aging time.

Table I. Fracture Toughness for Various Aging Temperatures

Alloy, Wt Pct	Aging Temperature	R _c Hardness (σ _y ~ 1,379 MPa)	K _{IC} MPa·m ^{1/2} (23°C)	CVN N·M (23°C)
Fe-16Ni-1.5Ti (Vacuum induction melted)	600°C	43	~125	37
	550°C	43	~55	19
	450°C	43	—	4

200 ksi = 1,379 MPa.
1 ksi √in. = 1.1 MPa m^{1/2}.
1 ft-lb = 1.36 J.