MECHANISMS OF
STRENGTHENING AND FRACTURE
IN COMPOSITE MATERIALS

Second Progress Report
to NASA

December 1, 1964 - May 31, 1965

Contract No. NSG-622

A. S. Tetelman
Associate Professor
Department of Materials Science
Stanford University
Stanford, California

MECHANISMS OF

STRENGTHENING AND FRACTURE

IN COMPOSITE MATERIALS

Second Progress Report
to NASA

December 1, 1964 - May 31, 1965

Contract No. NSG-622

A. S. Tetelman
Associate Professor
Department of Materials Science
Stanford University
Stanford, California

INTRODUCTION

This report describes research activities that were performed during the second phase of a program designed to investigate the mechanism of strengthening and fracture of composite materials. As mentioned in Progress Report #1 (subsequently referred to as PR 1) three different types of composites are currently being investigated. These are discussed separately in the following paragraphs.

I) The Mechanisms of Strengthening and Fracture in Composites Containing Hard Particles Dispersed in a Softer Matrix

a) Tensile Deformation of Hypo-Eutectoid Steels (with Darel Hodgson, Graduate Student)

Hypo-eutectoid steels are composite materials containing varying volume fractions of hard particles of cementite (Fe_3C) dispersed in a ferrite matrix. These systems are interesting from a practical point of view because they are used extensively in structural application. They are also interesting for use as model systems in studies of composites because 1) the shape and distribution of the hard particles can be varied by heat treatment and 2) because the ferrite matrix itself can behave in a ductile or brittle manner, depending on test temperature.

The purpose of this phase of the program was to determine the effect of Fe_3C morphology and dispersion on the tensile properties of these steels, over a range of temperature in which the ferrite matrix is both ductile and brittle (susceptible to cleavage fracture). Flat tensile specimens, 0.040" thick and .250" wide, were machined out of stock containing 1.0 per cent manganese and 0.4 per cent carbon (5.9 + 8.6 vol per cent Fe_3C). Each steel was given two heat treatments. The first (Q. T) consisted of austenitizing at 1150°C for one hour, quenching into oil and then tempering at 695°C for 24 hours. This treatment gave a uniform dispersion of spheroidal Fe_3C in the ferrite matrix. The average particle size was about 3-5 microns diameter. The second heat treatment (S.C) consisted of austenitizing at 1150°C for one hour, slowly cooling (approximately 1/4° C/min.) through the eutectoid transformation at 705°C, and holding at 695°C for 24 hours. This treatment produced islands (53 & 80%) of coarse, partially degenerate pearlite nodules (alternate plates of ferrite & Fe_3C) and islands of free ferrite (47 & 20%).

Tensile tests were carried out between -196°C & +85°C in a specially constructed cryostat. Figure 1 shows that the yield stress of all composites increased with decreasing temperature. Furthermore, for either heat treatment, the yield stress of the 0.6% carbon steel is slightly higher than that of the 0.4%C steel, at any test temperature. Finally, at any test temperature, the (Q.T) steels are stronger than the (S.C) steels. This may result from the finer ferrite grain size in the Q.T steels, carbide morphology, or both effects. The point is currently under investigation.

The effect of test temperature on the ductility of the two types of composite is shown in Figure 2. The (Q.T) steels were ductile at -196°C, fracture initiating by fibrous cracking. The (S.C) steels showed a sharp ductile to brittle transition at -160°C. Again, ferrite grain size and carbide morphology are probably responsible for these differences. The ductile-brittle transition also is evident from the fracture stress curves shown in Figure 3. Of immediate interest is the fact that particle shape plays a significant role in determining ductility, even when the composites fracture by fibrous propagation only. At 85°C the (Q.T) steels containing spheroidal particles show greater tensile ductility than steels of the same composition containing plate-like particles (S.C).

During the next six months extensive metallographic investigations will be carried out to determine the reasons for this behavior, as well as additional tensile testing on manganese free steels to determine the effect of manganese. Microstrain studies will also be carried out to determine the effect of particle shape on the tensile yield stress.

b) Tensile Micro-Strain Studies of TiC-Ni-Mo Cermets
(with F. Darwish, Graduate Student)

The purpose of this investigation is to determine the mechanism of fracture in a composite containing large volume fractions of hard phase dispersed in a soft, FCC matrix. The initial investigations were described in PR 1. During the past six months the following progress has been made.

The Tuckerman micro-strain gauge and collimator have been set up and micro-strain measurements were made on an Fe-3% Si alloy to check out the apparatus. Our results were found to be in excellent agreement with those obtained elsewhere, indicating that our rig is in good working order.

Composites containing 60% by weight of the hard phase (TiC), and 40% Ni-Mo matrix are currently being prepared from powders of TiC, Ni, and Mo_2C. The powders are being milled for varying period of time (12 hours up to 3 days) to obtain a range of TiC particle sizes, dried in a vacuum furnace, and stored.

The die, in the shape of a dog-bone tensile specimen, has now arrived and compacts will be pressed as soon as the sintering furnace is completed. Pre-sintering will be carried out in a hydrogen train furnace described in PR 1. The induction generator is expected to arrive shortly, and as soon as the heat shields are completed, specimen preparation will begin.

During the next six months the tensile specimens, containing varying volume fractions of TiC and varying TiC particle size, will be prepared by sintering, ground, and tested on the microstrain apparatus. Replication studies will be carried out at various microstrains to determine the density of microcracks in the TiC particles as a function of stress and microstrain.

II) The Initiation of Fracture in a Composite Material Containing a Pre-Induced Crack (with D. Barnett, Graduate Student).

During the last six months Mr. Barnett was completing a theoretical problem on the effect of boundaries on crack propagation, supported by an ARO-D contract, and has done no additional work on this problem. It is expected that he shall return to this problem within one month, and take up the program described in PR 1.

III) The Mechanism of Fracture in Composites of Mechanically Drilled Holes and a B.C.C. Matrix (with Charles A. Rau, Jr., Graduate Student)

Small, second phase particles have been shown to increase the resistance to cleavage crack propagation in some brittle materials. This project is designed to investigate whether mechanically drilled holes can similarly improve properties in the presence of a notch without substantially lowering resistance to crack nucleation. During the past 6 months we have

(1) defined the strain concentration around a single hole in 0.040" thick sheet

(2) related the measured, nominal, fracture strain to the local strain around the hole at fracture

(3) determined the effect of holes on nucleation and propagation of cleavage from a sharp notch.

Above the ductile-brittle transition temperature (T_D), a critical fracture strain (ϵ_f) is required for cleavage initiation. This strain was measured in sheet tensile samples of Fe-3% Si for different standard conditions of grain size and test temperature. When single holes of various sizes are introduced into the sheet and the tensile tests are repeated, the <u>nominal</u> fracture strain is reduced (Figure 4). This indicates that the holes act as strain concentrators, i.e., the critical fracture strain ϵ_f is produced locally, and an unstable cleavage crack is nucleated, at a lower nominal strain than if the hole were not present.

A strain concentration factor $K_\epsilon(H_D)$ may be defined as the fracture strain without the hole divided by the nominal strain in the presence of a hole. At -148°C, and with this specific specimen size (w = 0.435", t = 0.040", gauge length 1.25") an explicit relationship was obtained from the data in Figure 4 for both grain sizes studied:

$$K_\epsilon(H_D) = \frac{\epsilon_f}{\epsilon_f(H_D)} = \frac{0.07}{0.07 - H_D}$$

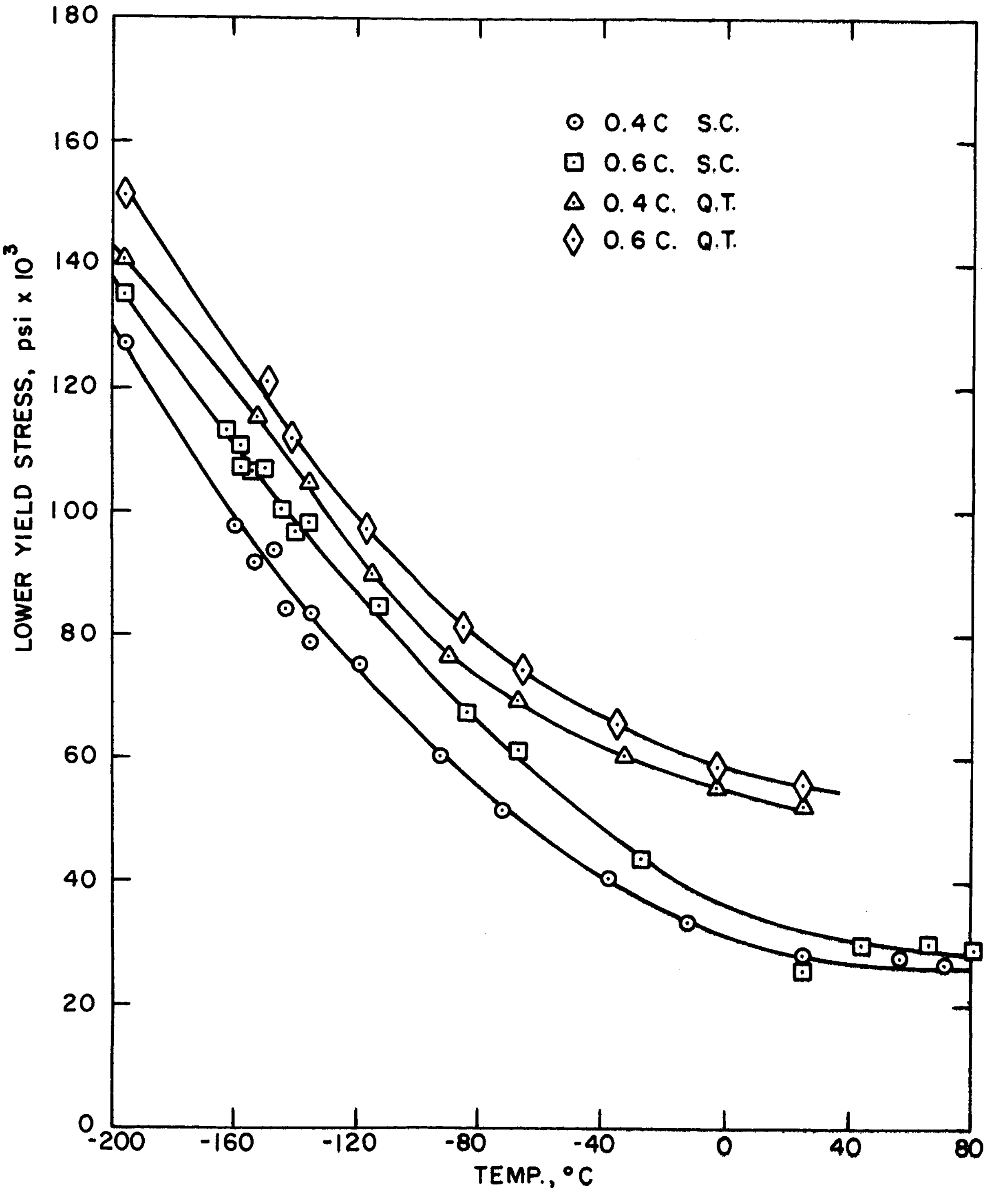
where H_D is the hole diameter in inches. The measured $K_\epsilon(H_D)$ is a result of two effects. First, there is an elastic and plastic stress concentration factor associated with the hole which will concentrate strain in its vicinity. Second, the decreased section area in the sample of finite width tends to concentrate strain in the reduced section until local work hardening raises the applied load to the yield load of the undrilled gauge length. For holes 0.070" and larger in diameter, the critical ϵ_f is reached locally before the yield stress of the undrilled area is reached. Consequently, all strain is restricted to the reduced cross section.

Although drilled holes reduce the ductility of unnotched tensile specimens, they can markedly <u>improve</u> the toughness of specimens containing sharp notches. This results from their ability to concentrate strain and cause the notch to blunt out, by tearing, before a cleavage fracture is initiated below the notch root.

Figure 5 shows that a reduction of about 35°C in charpy impact transition temperature (T_D) can be achieved when small holes are drilled below the notch root. The effect of varying the angular coordinant (θ) for a constant R = 0.0448" is shown in Figures 5 and 6. Figure 6 shows that $\theta = 75^\circ$ produces the greatest lowering of T_D with good improvements at all angles between 45 and 75°. At larger or smaller angles the holes fall partly or totally outside the yield zones of the notch (predicted by theoretical slip line field) and improvement in toughness is not achieved.

The radial coordinant (R) was also increased to R = 0.074" at $\theta = 75^\circ$ and $\theta = 30^\circ$, and no change from standard undrilled samples is produced. In all cases, it is necessary that the two holes lie within the yield zone of the notch and that fracture initially spreads from the root to <u>both</u> holes. These experiments indicate that cleavage initiation is prevented until lower temperatures because the holes in the yield zones of the notch are able to concentrate strain. They produce fracture to both holes before sufficient strain is produced for cleavage initiation at the notch tip. Once fracture has occurred to both holes the "effective radius" of the notch is very large (Figure 7) and cleavage cannot be initiated.

Future work will employ slow-bend tests on Fe-3% Si and subsequent etch-pitting studies to more specifically define this fracture mechanism by determining the effect of holes on the fracture load and on the plastic strain distribution at the notch root.



0.4 C S.C.
0.6 C. S.C.
0.4 C. Q.T.
0.6 C. Q.T.
LOWER YIELD STRESS, psi x 10^3
TEMP., °C
180
160
140
120
100
80
60
40
20
0
-200
-160
-120
-80
-40
0
40
80

Figure 1

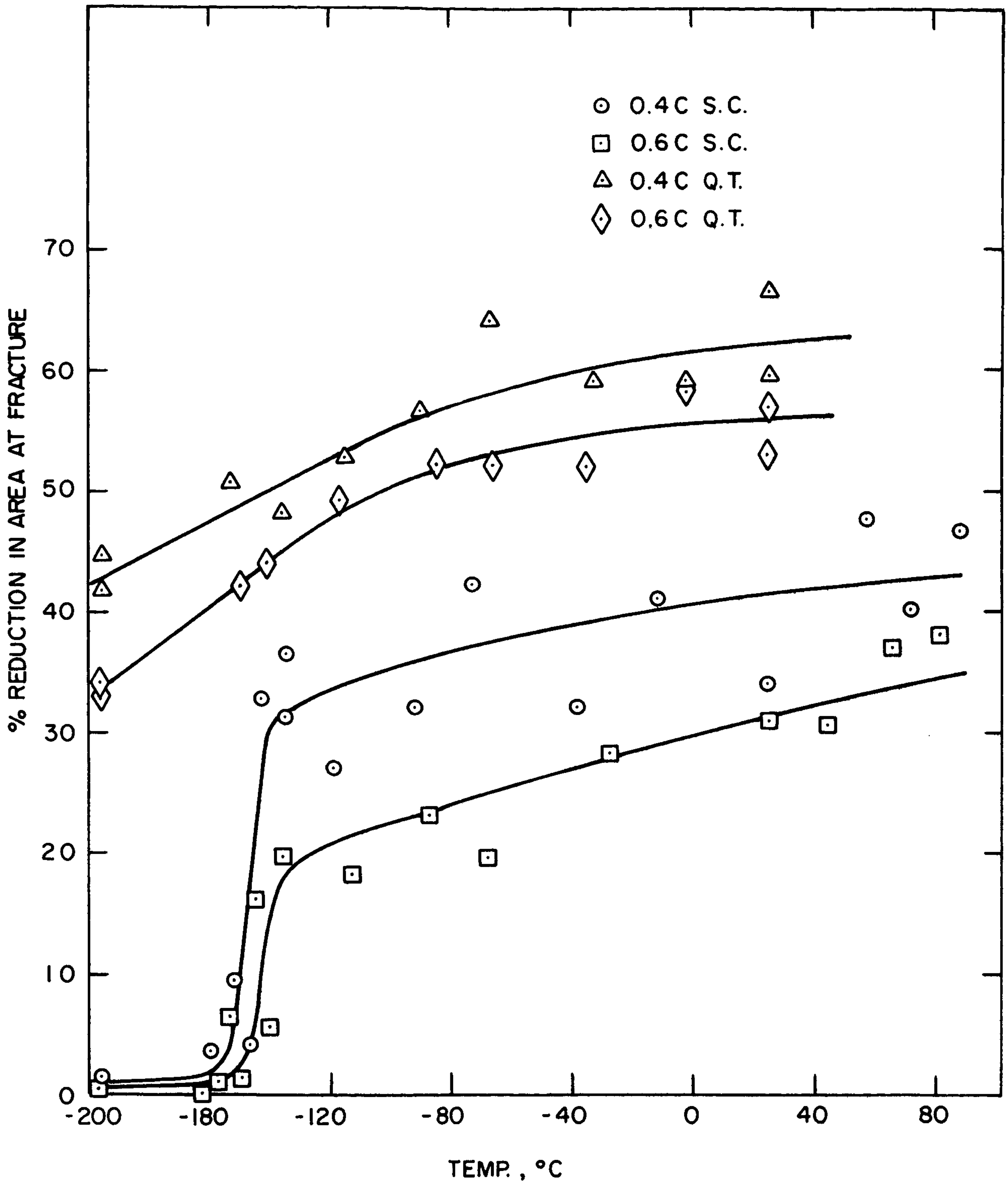
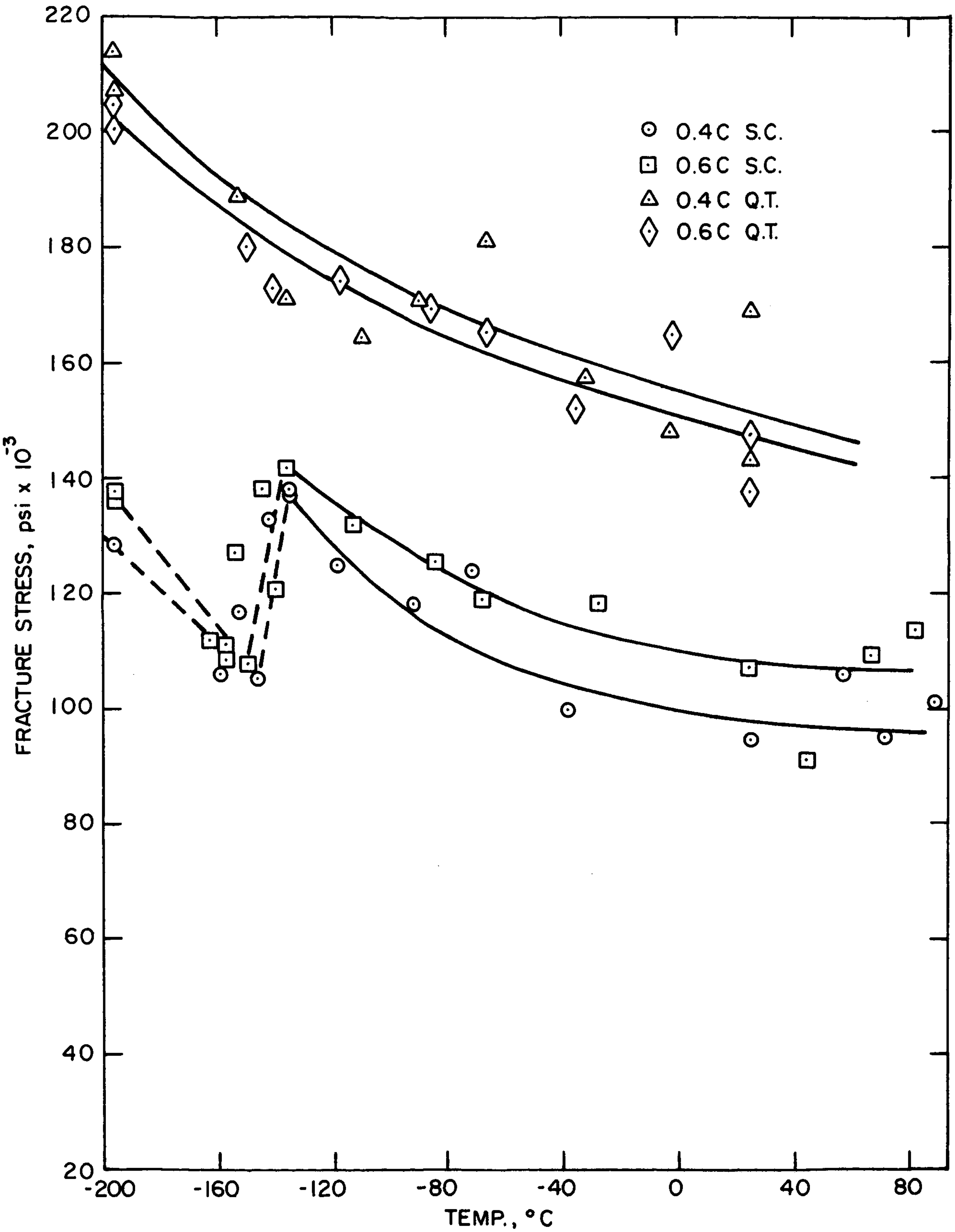
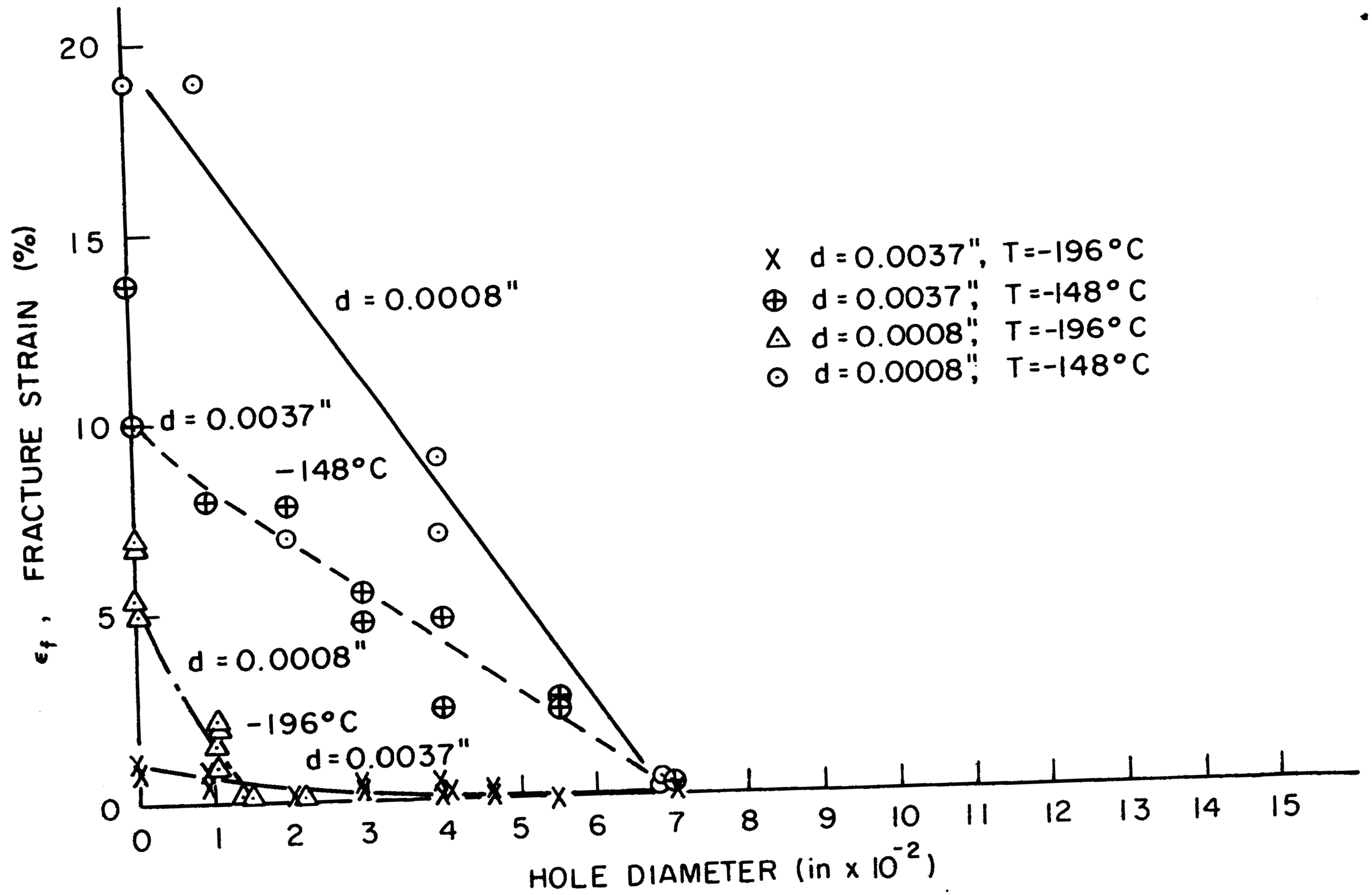

0.4C S.C.
0.6C S.C.
0.4C Q.T.
0.6C Q.T.
% REDUCTION IN AREA AT FRACTURE
70
60
50
40
30
20
10
0
-200
-180
-120
-80
-40
0
40
80
TEMP., °C

Figure 2



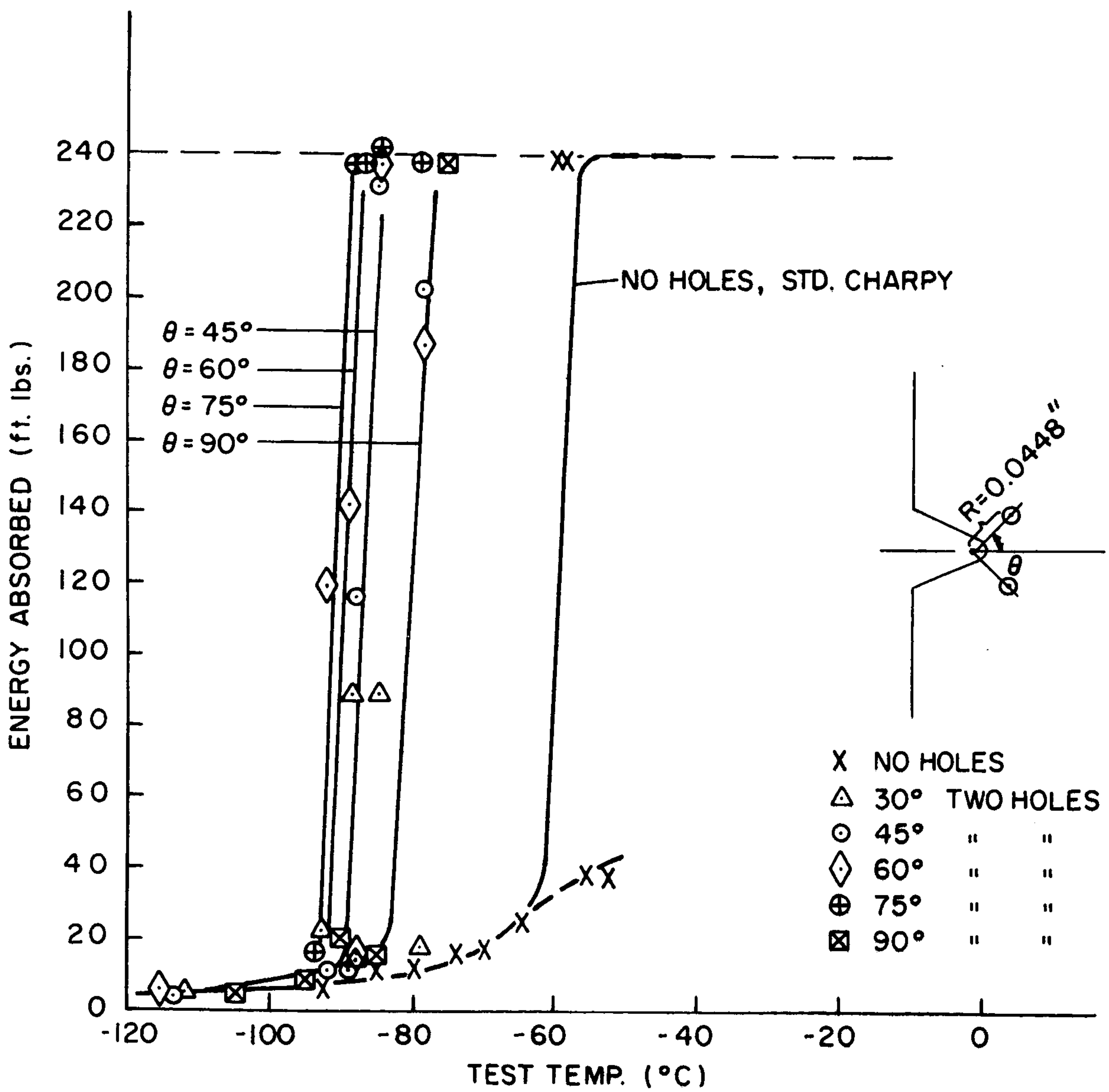
220
200
180
160
140
120
100
80
60
40
20
-200
-160
-120
-80
-40
0
40
80
TEMP., °C
FRACTURE STRESS, psi x 10^-3
0.4C S.C.
0.6C S.C.
0.4C Q.T.
0.6C Q.T.

Figure 3



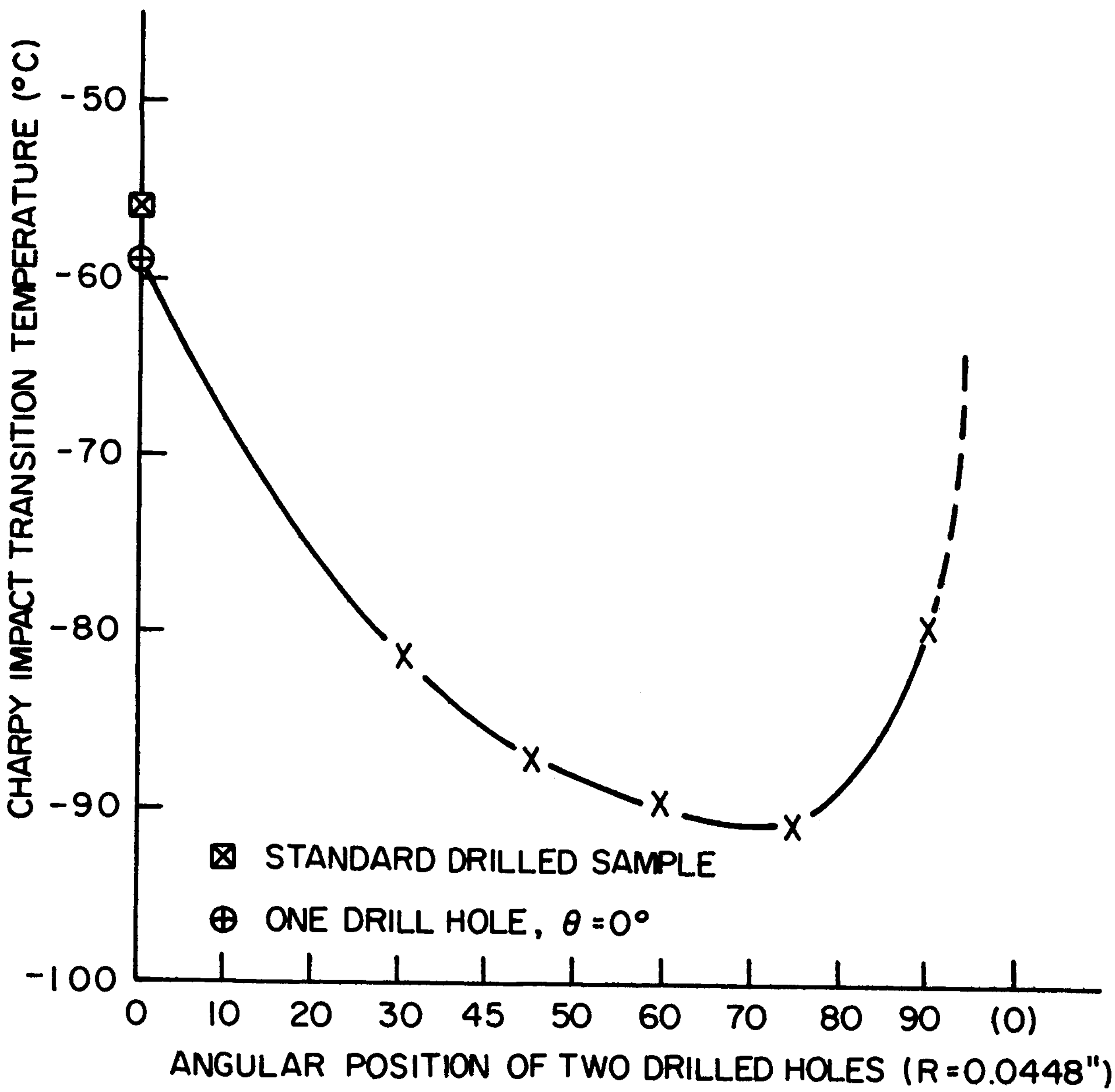
The effect of hole size on the nominal fracture strain in Fe- 3% Si sheet at various conditions of grain size and temperature.

Figure 4



The effect of two 0.0292" holes drilled at various angles in the yield zones of a notch on the charpy impact transition curve of alloy V- 906.

Figure 5



CHARPY IMPACT TRANSITION TEMPERATURE (°C)
-50
-60
-70
-80
-90
-100
STANDARD DRILLED SAMPLE
ONE DRILL HOLE, θ=0°
0
10
20
30
45
50
60
70
80
90
(0)
ANGULAR POSITION OF TWO DRILLED HOLES (R=0.0448")

Figure 6

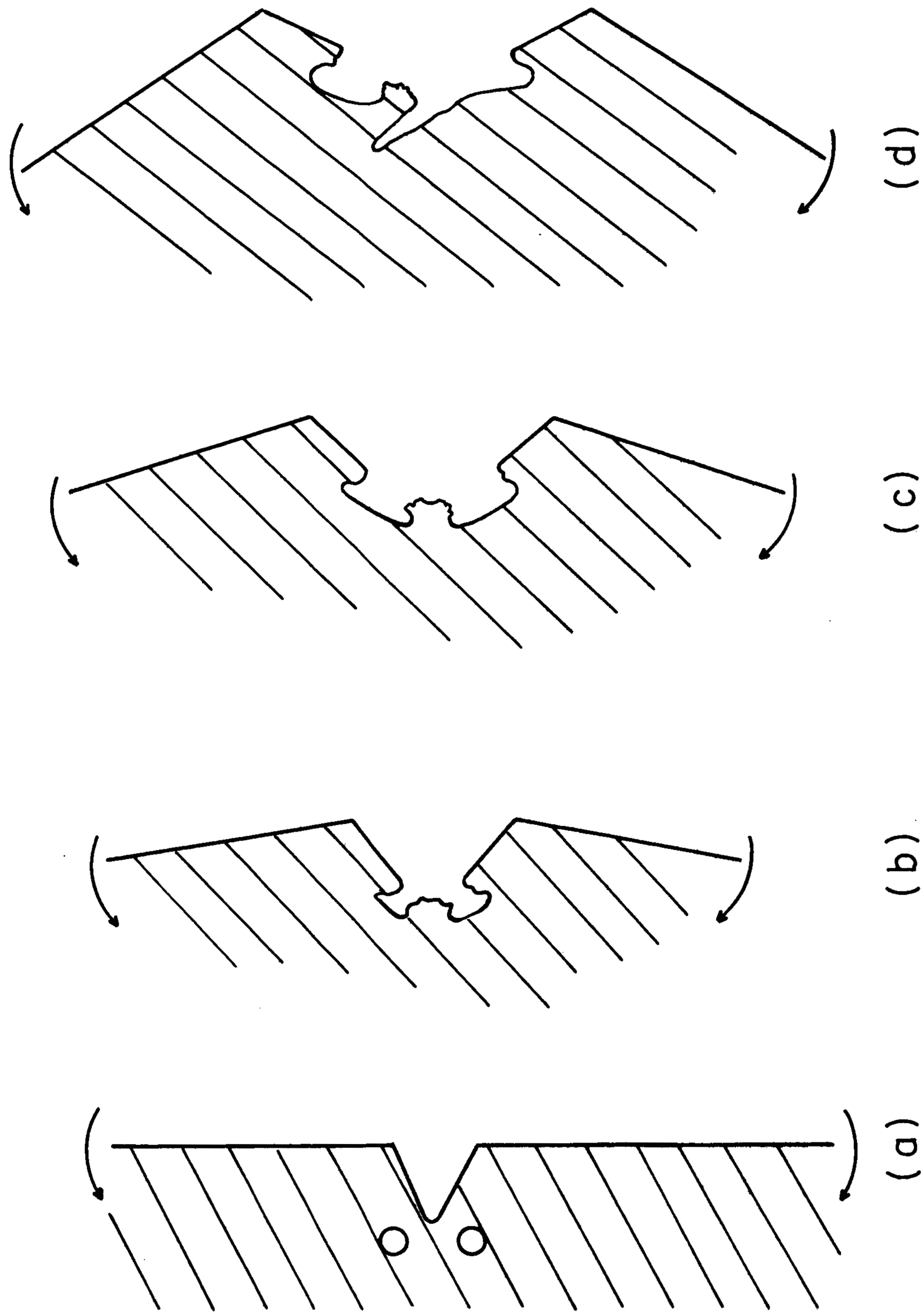


Figure 7