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CRACK VELOCITY MEASUREMENTS IN SOLITHANE 113.(U)

JAN 82 S R SWANSON

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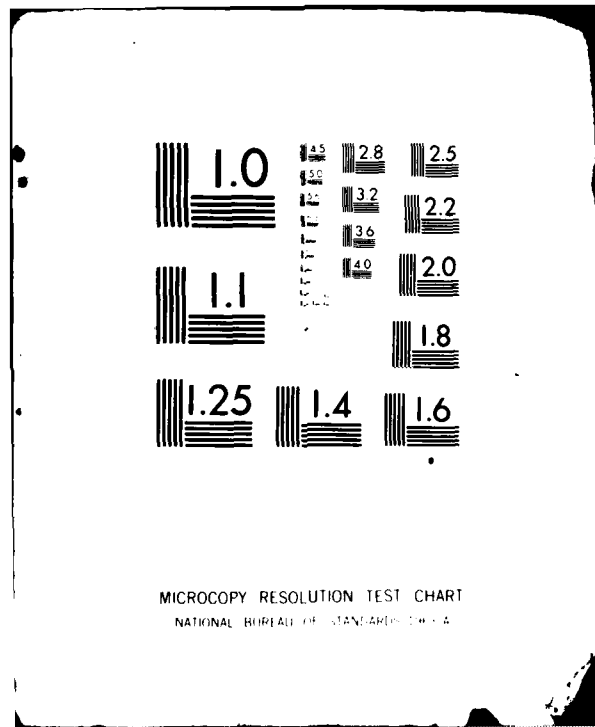
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The results showed that the crack speed varied with stress intensity factor raised to a power of 6, in good agreement with previous studies for this material. Only small deviations from this relationship were observed, even at the highest crack speed measured of 1600 in/sec (40 m/s), which is approximately 12% of the calculated glassy modulus Rayleigh wave velocity. In contrast, a previous study on an inert solid propellant had shown an apparent limiting crack velocity of approximately 6% of the calculated glassy Rayleigh wave velocity. Among other differences, this latter material had a much broader viscoelastic modulus transition range.

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Crack Velocity Measurements in Solithane 113

S. R. Swanson

Department of Mechanical and Industrial Engineering
University of Utah

January 1982

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Annual Report on
Fragmentation in Viscoelastic Materials

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ABSTRACT

An experimental study of crack propagation speeds in a highly viscoelastic polymer, Solithane 113, is presented in this report. Loadings were applied to precracked sheet specimens over a wide range of rates. Resulting crack speeds were found to vary from .003 to 1600 in/sec (8×10^{-5} to 40 m/s).

The primary purpose of this study was to investigate the transition from quasistatic to dynamic crack propagation in viscoelastic materials. It is believed that this transition depends strongly on the viscoelastic material properties. These properties vary widely with time for strongly viscoelastic materials such as Solithane 113 in the modulus transition range above the glass transition temperature.

The results showed that the crack speed varied with stress intensity factor raised to a power of 6, in good agreement with previous studies for this material. Only small deviations from this relationship were observed, even at the highest crack speed measured of 1600 in/sec (40 m/s), which is approximately 12% of the calculated glassy modulus Rayleigh wave velocity. In contrast, a previous study on an inert solid propellant had shown an apparent limiting crack velocity of approximately 6% of the calculated glassy Rayleigh wave velocity. Among other differences, this latter material had a much broader viscoelastic modulus transition range.

INTRODUCTION

This is the first annual report on an investigation of dynamic crack propagation and fragmentation in viscoelastic materials. The primary objective of this year's work has been to measure the dynamic crack propagation behavior of an unfilled polymer, Solithane 113. Of particular interest is the behavior during the transition from quasi-static to dynamic crack growth. The techniques used and results of this work are given in this report.

The overall motivation of the present study is an attempt to understand the conditions under which polymeric materials can fracture into a large number of small pieces or fragments, a process loosely described as fragmentation. Analytical modeling of the process by the author [1] and others [2] has appeared to indicate that the details of high velocity crack propagation may be of key significance.

Studies of high velocity crack propagation in polymeric materials have been primarily restricted to the hard plastics [3-8]. The materials have been often selected to serve as an analogy for metal fracture, with the polymeric material selected on the basis of some convenient experimental property such as birefringence or lower wave speeds as compared with metals.

Theoretical analyses of elastic stress distributions around the tip of a rapidly moving crack [9, 10] have shown that the Rayleigh wave velocity should set an upper bound to crack propagation velocity. The Rayleigh wave velocity is slightly lower than the shear wave velocity, approximately $.95 \sqrt{G/\rho}$ for typical values of Poisson's ratio, where G is the elastic shear modulus and ρ the mass density. The experiments

cited above show that a more typical maximum is on the order of 50% of the Rayleigh wave velocity, with the discrepancy ascribed to various factors such as nonlinear material behavior at the crack tip, crack branching, and the effect of stress waves reflected from the specimen boundaries.

Significant developments have taken place in the understanding of quasistatic fracture in highly viscoelastic polymers. Using the idea of a process zone of finite size at the crack tip, Knauss [11, 12] and Schapery [13] have developed a model for quasistatic crack propagation that is described by the equation

$$\dot{a} = A K_I^Q \quad (1)$$

where \dot{a} is crack velocity, K_I is the opening mode stress intensity factor, and A and Q are experimental constants. The value of Q can be related to the slope of the stress-relaxation modulus, and has been found to have a value of approximately 6 for Solithane 113. Applications of this crack propagation law have been made to solid propellants [14, 15]. A typical value of the exponent for solid propellant is approximately 7.

The high value of the exponent Q of Equation (1) typically found for highly viscoelastic materials in the transition range means that the crack velocity is strongly dependent on the stress intensity factor. This relationship poses a dilemma for fragmentation models using crack growth as part of the fragmentation process. The difficulty can be described as follows. If Equation (1) is used to describe the crack propagation process, it means that cracks will grow with a velocity proportional to their characteristic size raised to the $Q/2$ power. Thus

the largest flaw (usually described in a statistical sense) will quickly grow under load to dominate the fracture process. As a result the usual modeling assumptions of crack interaction will predict very low fragmentation under applied load. To avoid this difficulty Equation (1) has been ignored and an alternative such as constant crack speed used. Another approach has been suggested by the author [1], who used a transition from Equation (1) behavior to constant crack velocity, based on an approximate dynamic model for energy release at the crack tip. While the model may be considered highly approximate, it is interesting to note that a fragmentation model based on this crack velocity relationship does appear to show a transition from low fragmentation failure at low applied stress, to high fragmentation at high stresses in accord with some experiments [1]. Perhaps more importantly, it serves at present to focus attention on the requirement for a definition of crack propagation behavior in the region of the transition from quasistatic to dynamic conditions.

The present report gives the results of experimental measurements of crack propagation velocity for the unfilled polymer Solithane 113. The preparation and characterization of the Solithane 113 specimens are also presented.

EXPERIMENTS

The experiments performed consisted of the preparation and characterization of the Solithane 113 material, and the crack propagation tests on Solithane sheet specimens. The experimental details are given below.

Solithane Preparation and Characterization

Procedures for the preparation of Solithane 113 have been documented by Knauss and Mueller [17] and were followed closely in our work. Sheets of material 1/4 by 12 by 12 inch were cast, using the 50/50 composition.

Characterization of the Solithane 113 was accomplished using a commercial Rheovibron, in which an oscillatory tensile strain is imposed on a specimen at various frequencies and temperatures. The ratio of dynamic stress to strain is E^* , the dynamic complex modulus. A plot of E^* determined at three frequencies and a range of temperatures between -40° and 22°C is given in Figure 1. Using the usual time-temperature superposition [18], these data are shifted to form a master curve of E^* as a function of frequency. A plot of the experimental shift factors is given in Figure 2, along with an equation of the WLF form given by Knauss and Mueller [17]. As can be seen, the experimental shift factors agree quite closely with Knauss and Mueller's results. The WLF equation was therefore used to construct the master plot of E^* versus frequency. This plot was then converted to $E_{re}(t)$ versus time, using standard relationships between the viscoelastic properties, a Prony series representation of the stress relaxation modulus, and a computer based least squares procedure. The resulting master curves of stress

relaxation modulus are shown in Figure 3 for three batches of material. The variation of these three curves gives an indication of the possible effects of batch to batch reproducibility of material coupled with the accuracy of material properties determination. The relaxation modulus curves are reasonably consistent among the three batches of material and also agree well with data presented by Knauss and Mueller [17]. It is therefore concluded that the Solithane 113 prepared in our laboratory is reproducible and representative of that studied by Knauss and Mueller.

Crack Propagation Experiments

The main thrust of the research accomplished to date has been the determination of crack propagation velocities in Solithane 113, over a range of loading rates including high speed tests. The apparatus used for the high speed tests is shown in Figures 4 and 5. It consists of an air driven cylinder, frame and quick release valve or rupture disc. Load is measured by a Kistler crystal load cell, and signals are recorded on a Nicolet digital memory oscilloscope.

A unique feature is the use of event gages to signal passage of the crack tip. While becoming more common in crack propagation experiments with more rigid material, the low modulus and high elongation of materials like Solithane and solid propellants places special requirements on the event gage. These problems were solved by the use of a conductive (silver filled) RTV. A specimen is shown in Figure 6.

A typical experimental result is shown in Figure 7. In this figure a recording of force in the specimen is given, along with the step-like response of the event gages that indicate the passage of

the crack tip. Tests of this type were run on strip specimens, with a central starter crack cut in with a razor blade. Data reduction involves the calculation of stress intensity factor K_1 at each of the known crack positions. The crack position also is plotted versus time and differentiated graphically to get crack velocity. Finally crack velocity is plotted as a function of the applied stress intensity factor.

The applied stress intensity factor is calculated from the load and crack position by use of the formula

$$K_1 = \sigma_0 \sqrt{\pi a} F(a/b, c/b)$$

where σ_0 is the average applied stress at the edge of the strip sample, and F is a geometry correction factor given in Figure 8. This correction factor is based on an interpolation of the work of Isida [19] and finite element results. The analysis is based on quasistatic linear elasticity.

A summary plot of the results obtained is shown in Figure 9. This plot is the major result of the present research. A discussion of the significance of these results will be given in the next section.

DISCUSSION

The data plotted in Figure 9 represent crack velocities varying over six orders of magnitude. The significance of this plot can be assessed by comparisons with the predictions of quasistatic and dynamic viscoelastic crack propagation theories. The quasistatic theory of Knauss [11, 12] and Schapery [13] give the governing equation for crack propagation as

$$\dot{a} = A K_I^Q \quad (1)$$

where \dot{a} is crack velocity, K_I is the applied stress intensity factor, and A and Q are material constants. A fit of this law to the present data is shown in Figure 10. The value of Q thus obtained is 6.0, which is in good agreement with that obtained in quasistatic tests by others [11, 15]. Also shown in Figure 10 is the prediction of the dynamic theory of Swanson [16], applied to Solithane. A final theoretical value shown in Figure 10 is the Rayleigh wave velocity associated with the glassy modulus, which is given by

$$V_{Rg} = \beta \sqrt{G_g / \rho} \quad (2)$$

where V_{Rg} is the Rayleigh wave velocity, β is a numerical constant equal to .96 for Poisson's ratio equal to .5, G_g is the glassy shear modulus (calculated by means of the approximation $G_g = E_g/3$) and ρ is mass density. The value shown for V_{Rg} is 13,700 in/sec (350 m/s), calculated from a glassy tensile modulus of 60,000 psi (410 MPa) and a density of .0378 g/cc. It is also interesting to note the maximum crack speed of 11,500 in/sec (290 m/s) obtained by Kim and Knauss [20] at a temperature of -60 C, also shown in Figure 10.

The scatter in the results shown in Figures 9 and 10 is larger than

would be desired. The major cause appears to be batch to batch variation of the Solithane 113, and can be attributed to variation in modulus. The earlier tests performed in this study showed larger test to test variation. However refinements in test technique, primarily with respect to the crack event gages, have reduced this source of variation and allowed the batch to batch variation to be seen more clearly. Further work is thus necessary to eliminate this variable from the test results.

The identification of the applied stress intensity factor in the test specimens is subject to several uncertainties. The large deformation of the rubbery Solithane 113 certainly makes the application of small deformation theory questionable. The obvious motivation is the lack of a practical alternative, and the success reached in quasistatic applications. An additional factor is the identification of stress variation from grip to grip within the specimen, due to dynamic effects. In the tests performed so far, the specimen appears to have reached a quasistatic state by the time the crack reaches the first event gage, as judged by the non-oscillating load cell output. Thus the quasistatic assumption of load application away from the running crack appears to be justified. However further study of the stress distribution is underway, using numerical techniques [21].

The prediction of the dynamic theory for Solithane is quite different from that given previously for an inert solid propellant [22] in an important respect. This previous prediction and experimental data are shown in Figure 11. The important point to note is that the dynamic theory predicts that the limiting crack velocity for Solithane

is essentially equal to the glassy Rayleigh wave velocity, while for the inert propellant the limiting crack velocity is predicted to be an order of magnitude lower than the glassy Rayleigh wave velocity. This predicted difference in behavior is due entirely to the difference in the relaxation modulus of the two materials. The inert propellant has a much broader transition in time from glassy to rubbery behavior than does the Solithane. A typical slope of the $\log E_{rel}(t)$ versus $\log t$ curve is .5 for Solithane and .1 for the inert propellant. The dynamic theory is sensitive to this parameter, as it determines the amount of relaxation in the material surrounding the postulated zone of cohesive forces at the crack tip.

The experimental evidence shown in Figure 10 appears to support the theory, although the comparison is not totally definitive. That is, the theory predicts that a terminal crack velocity would not have been reached under the experimental conditions imposed on the Solithane, and this appears to be the result obtained. However some deviation from the extrapolation of the quasistatic curve can be seen. This could be due to scatter or systematic error, wave propagation effects not accounted for in the sample, or could simply reflect disagreement between the dynamic theory and experiment.

In addition to the crack velocity results, it is important to note that the inert propellant showed evidence of being driven at a terminal velocity, in that crack branching was observed at the highest speeds. No evidences of this kind were observed in the Solithane tests.

It would seem then that the apparent absence of a terminal velocity in Solithane that is significantly lower than the Rayleigh wave speed is

a significant factor in dynamic viscoelastic crack propagation. This result occurred under experimental conditions that did produce a limiting velocity in an inert propellant, in accordance with prediction. It would appear that dynamic crack propagation is governed by relaxation at the crack tip (i.e. in the cohesive zone) as is quasistatic crack propagation. Further the interaction of dynamic effects and relaxation processes appears to be dependent on the details of the material relaxation behavior. However it would be strongly desirable to have a more direct confirmation of this. This could be accomplished by selecting a material or materials with the appropriate viscoelastic behavior for further dynamic crack propagation tests. By "appropriate" it is meant that the new material would have a broader viscoelastic transition region than Solithane, perhaps approaching that of the inert propellant tested previously. It is known that particulate fillers tend to give this effect and thus filled Solithane is a prime candidate for future consideration.

SUMMARY AND CONCLUSIONS

Crack propagation speed and applied load have been measured for Solithane 113 over a wide range of loading times at ambient temperature. The data are presented in the form of crack velocity versus applied stress intensity factor. Crack speeds up to 1600 in/sec (40 m/sec) were observed. The primary motivation was to investigate the transition region from quasistatic to dynamic fracture.

A logarithmic plot of crack speed versus stress intensity factor has been presented. Previous work as well as the present study show that crack speed in Solithane 113 varies as stress intensity factor

raised to a power of 6 in the region of quasistatic response. Only small deviations from this relationship were observed even at the highest crack speeds measured of 1600 in/sec (40 m/sec). Thus if strong dynamic effects are to be observed, they will presumably be in the range of crack speeds between that measured and the calculated Rayleigh wave velocity of 13,700 in/sec (350 m/s).

The behavior observed for Solithane 113 is significantly different than that observed previously for an inert solid propellant, where an apparent limiting velocity was reached more than an order of magnitude lower than the glassy modulus Rayleigh wave velocity. Among other differences, the inert propellant had a much broader viscoelastic transition range of mechanical properties. It appears that further investigations of dynamic crack velocities using alternate viscoelastic materials will be informative.

Acknowledgements

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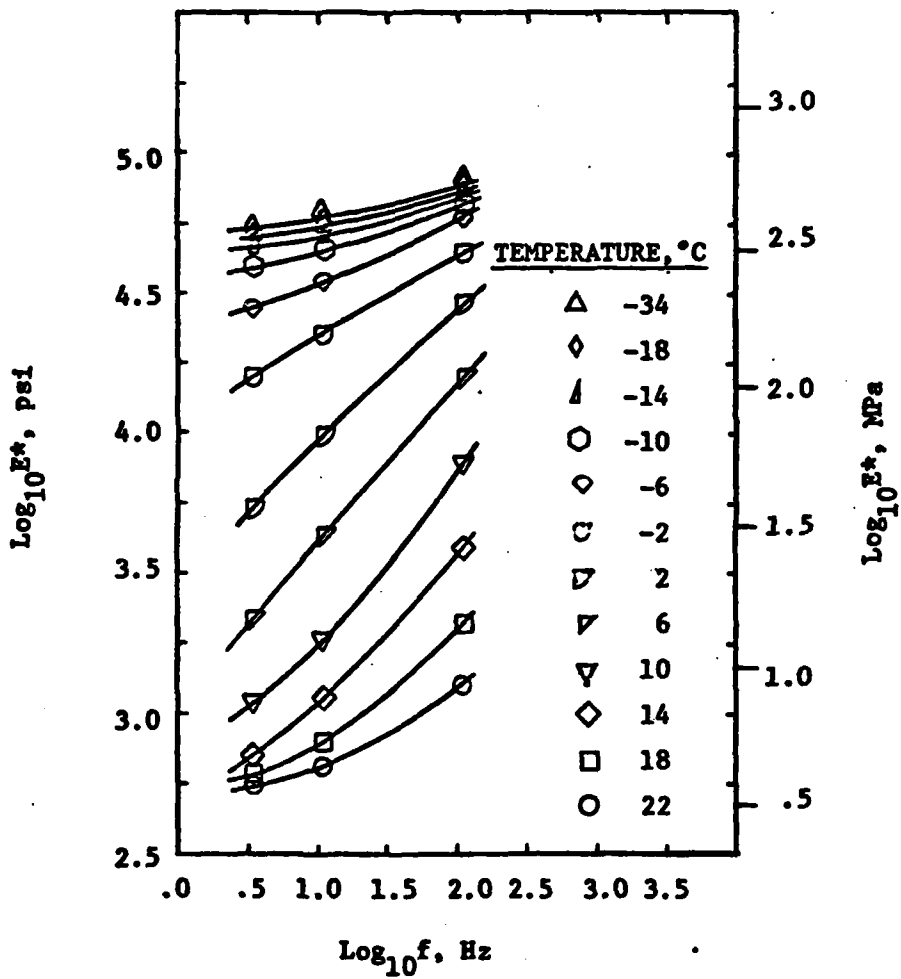


Figure 1. Rheovibron characterization of Solithane 113.

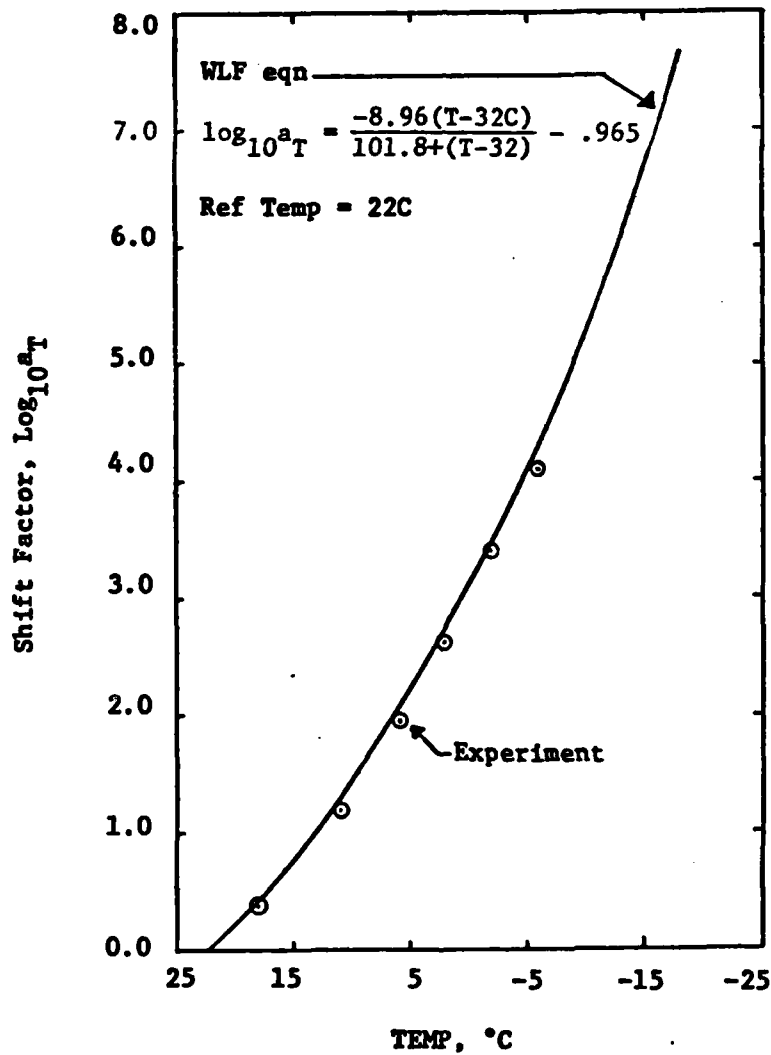


Figure 2. Comparison of Experimental Time-Temperature Shift Factors with WLF Eqn. for Solithane 113.

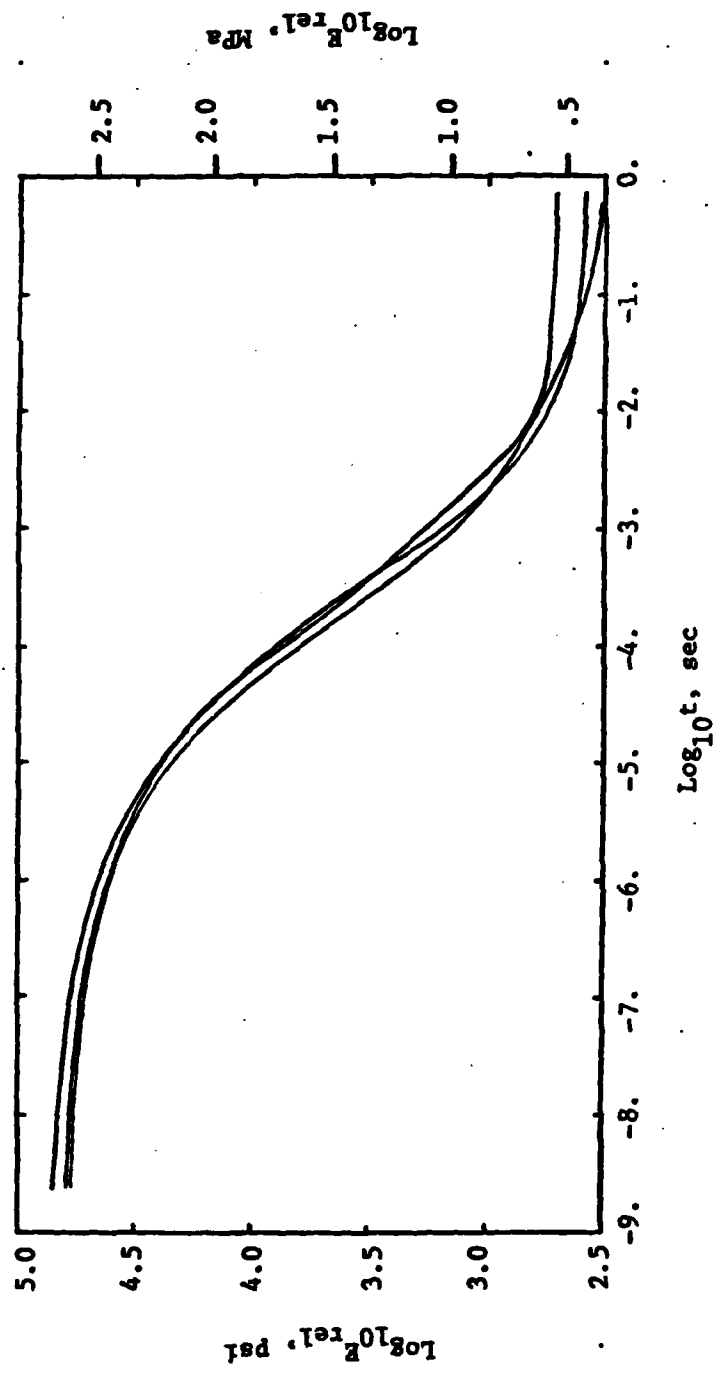


Figure 3. Tensile Relaxation Modulus Derived from Rheovibron Tests for Three Batches of Solithane 113.

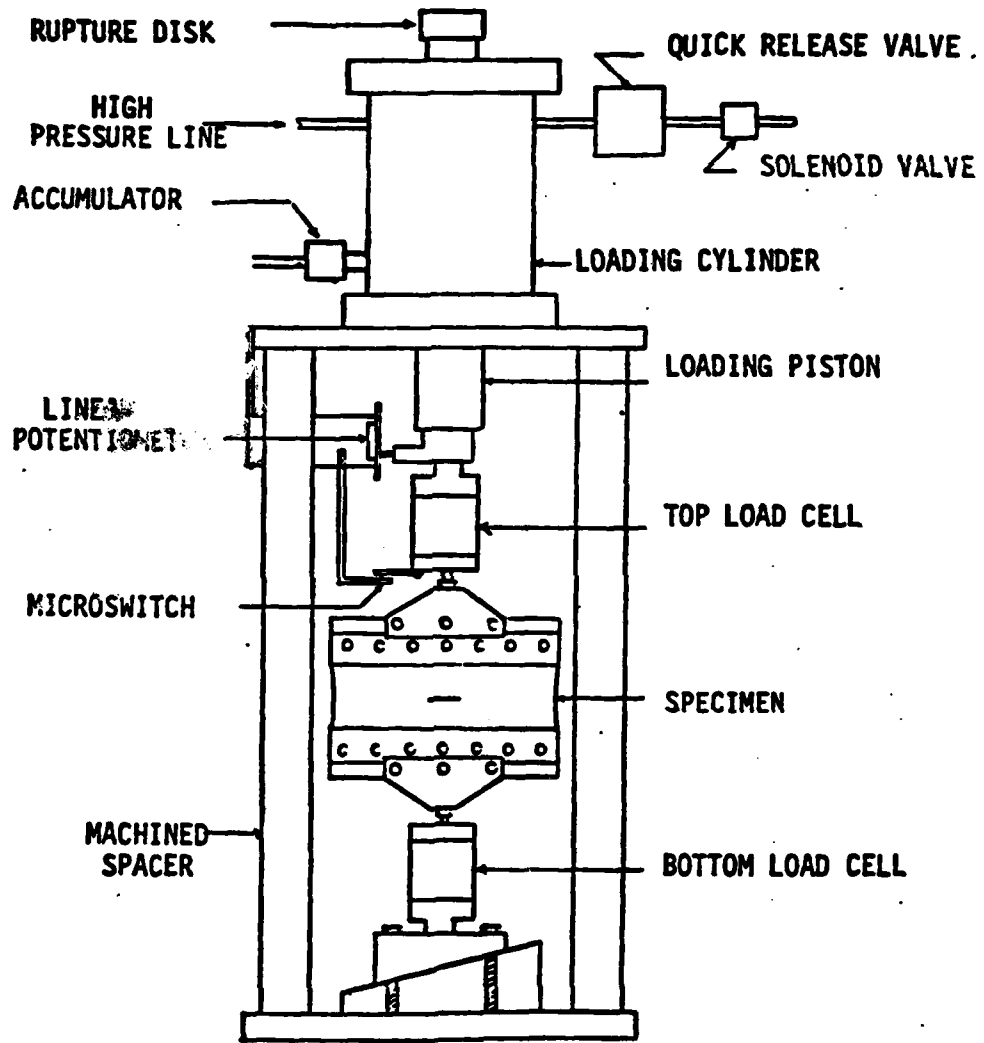


Figure 4. Dynamic Test Apparatus

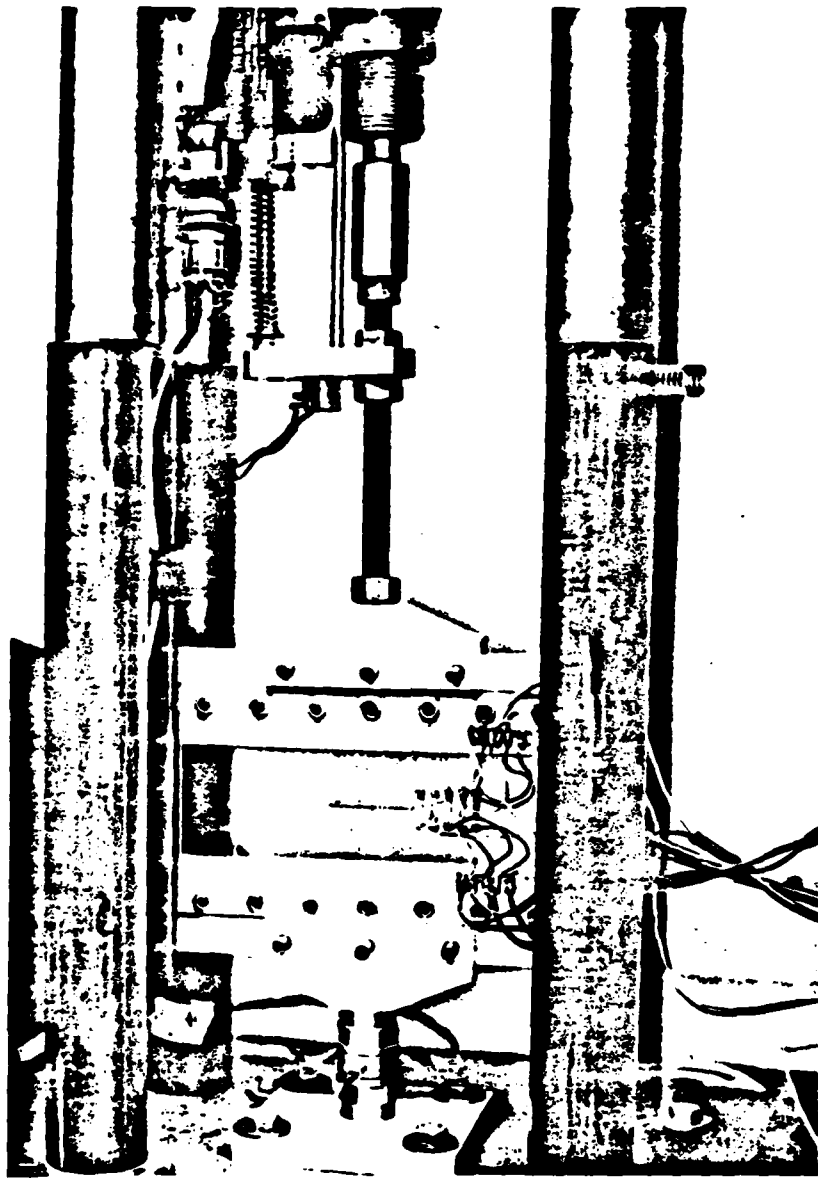


Figure 5. Dynamic Crack Propagation Apparatus

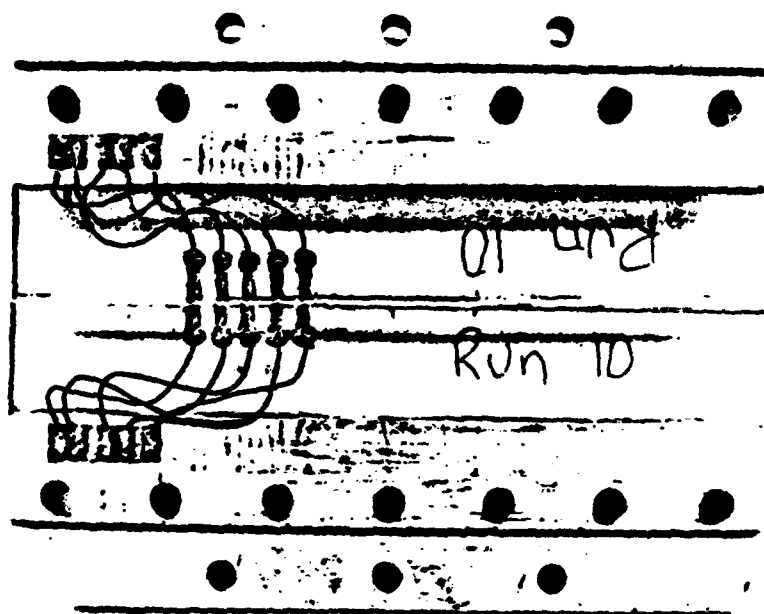
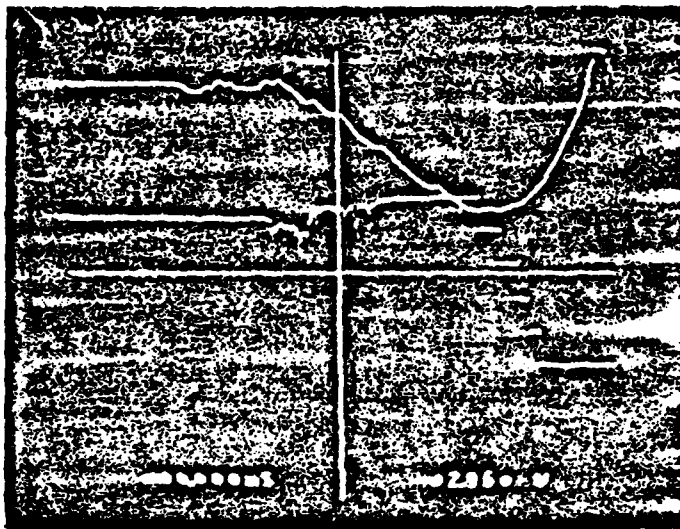


Figure 6. Solithane Crack Propagation Specimen



← 1 ms →

Figure 7. Typical Force and Crack Propagation Gage Record

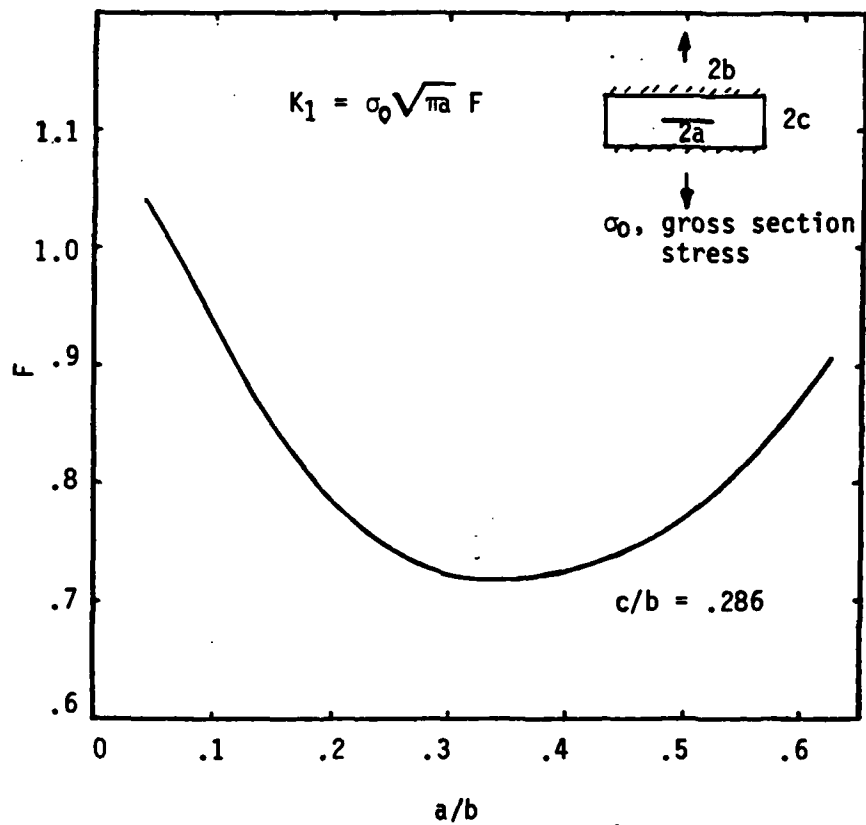


Figure 8. Stress intensity geometry factor for cracked strip specimen with clamped edges.

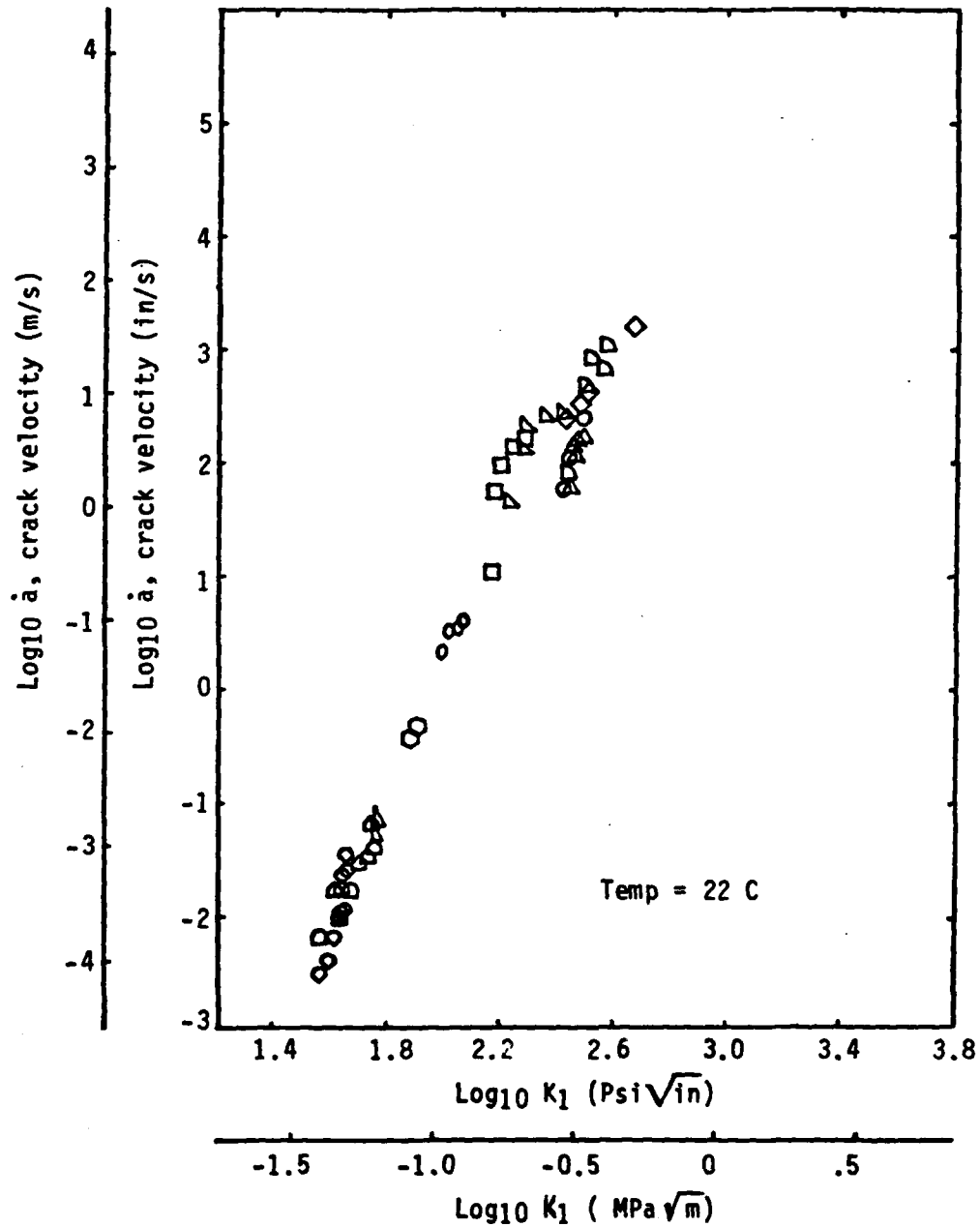


Figure 9. Measured Crack Propagation Velocities for Solithane 113.

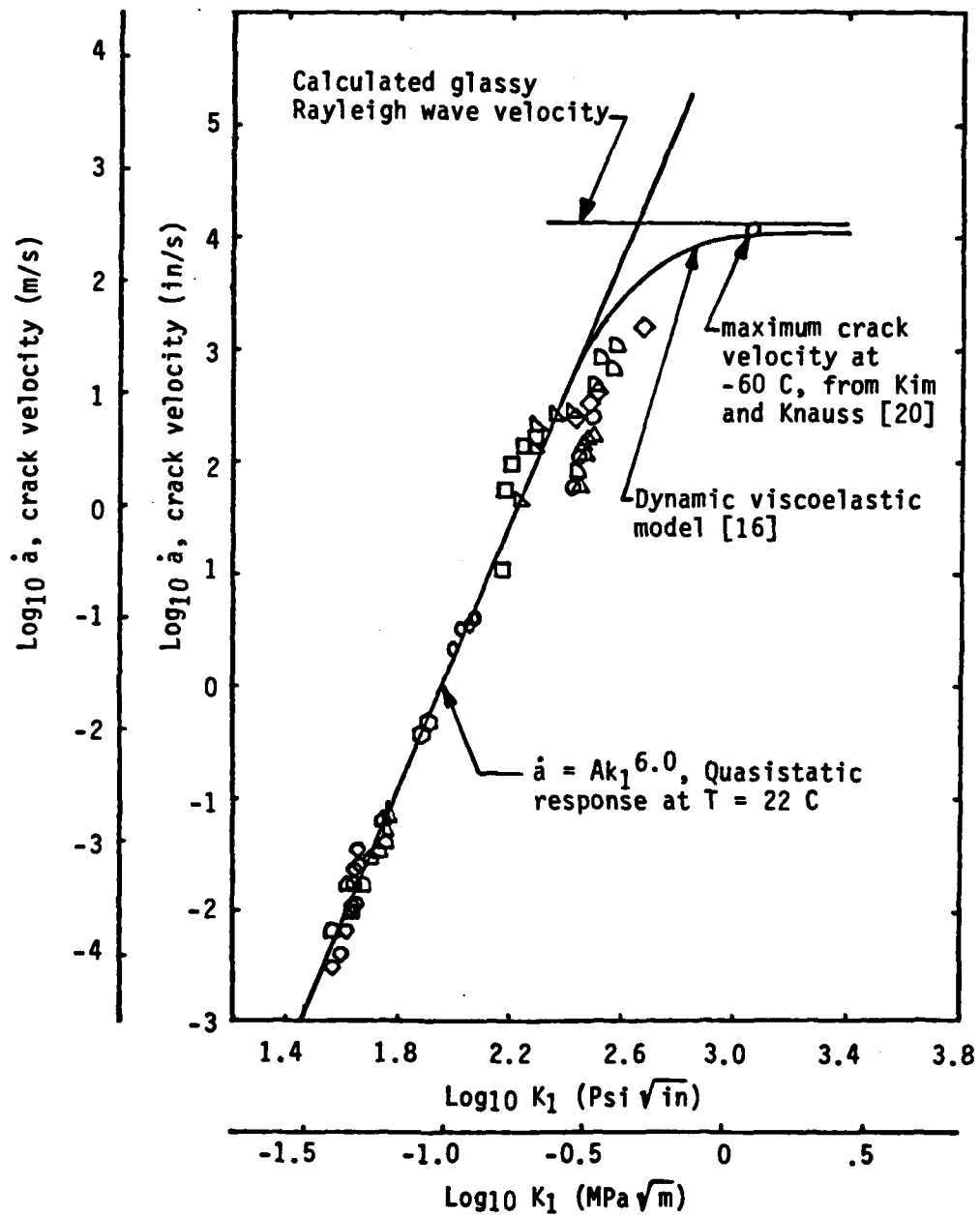


Figure 10. Comparison of Quasistatic and Dynamic Crack Propagation for Solithane 113.

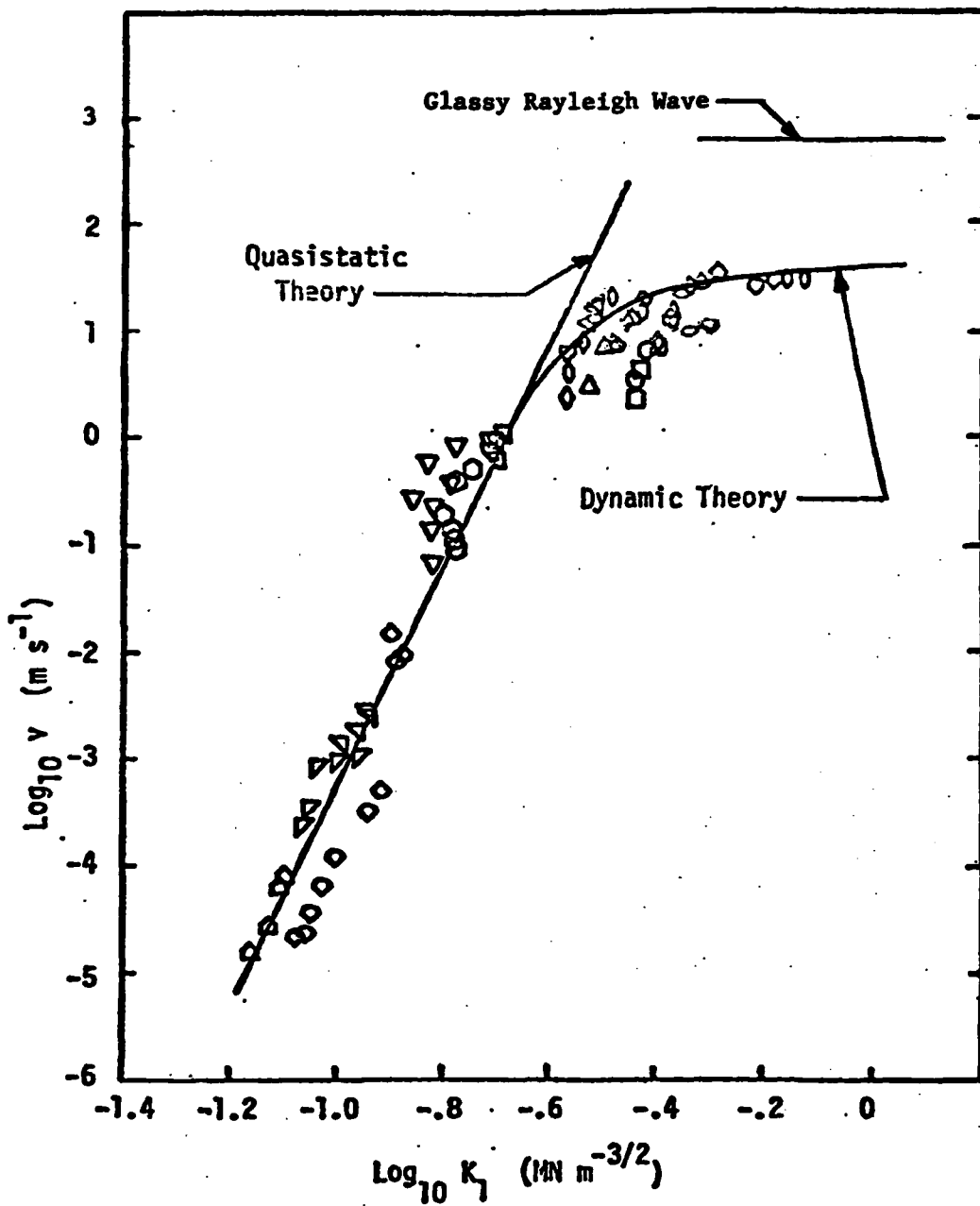


Figure 11. Crack Propagation Velocity for Inert Solid Propellant

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