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# On the Feasibility of Quantitative Ultrasonic Determination of Fracture Toughness - A Literature Review

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## I. INTRODUCTION

It is known that the fracture toughness of materials can be improved
by specific microstructure modifications. For example, the covalent electron
bond, low crystal symmetry, and long range order in a solid lead to increasing
tendency for brittle fracture. The study of stress corrosion gives evidence
that migration of hydrogen atoms to the crack tip embrittles the material.
This implies that a change of microstructure tends to change the fracture
behavior of the material.

There have recently been a number of studies to relate fracture toughness
and specific microstructural factors. A survey of the work is presented in
Section II with emphasis on particular models devised to describe the process
of fracture.

Ultrasonic non-destructive methods are commonly used to detect the presence
and size of flaws. But the criticality of crack-like flaws depends upon the
microstructural environment in which the flaw resides. For this reason
considerable attention is currently being given to the ultrasonic characteri-
zation of microstructural factors that govern such properties as fracture
toughness. A review of some recent work in this area, with emphasis on the
ultrasonic determination of fracture toughness, is given in Section III.

Finally, Section IV presents a review of mathematical methods available
for solving boundary value problems related to scattering of ultrasonic waves
by microstructural factors that govern fracture toughness and ultrasonic wave
propagation. The purpose of this paper is to delineate a basis for the
empirically observed correlation between fracture toughness and ultrasonic
factors.

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## II. FRACTURE TOUGHNESS AND MICROSTRUCTURE

### 2.1 Background

The fracture toughness is an index that is a measure of material resistance to catastrophic crack propagation. This index can be given in terms of the critical stress intensity factor $K_{1c}$, the critical energy release rate $G_{1c}$, the critical J-integral $J_{1c}$ or the critical crack opening displacement $\delta_c$. The numeral "1" is referred to the normal separation mode of crack face displacement. The fracture toughness is recognized as a material constant and is normally evaluated by measuring the onset of crack extension in small laboratory specimens for which $K_1$ is defined as a function of specimen geometry, crack length and applied load or load-point displacement. Standard-ized methods for conducting the tests, instrumentation, and interpretation are described in ASTM specification E-399-74. Due to different methods in computa-tion and measurement, one of the indice of fracture toughness may be easier to obtain and once one of them is found the others can be obtained from the following relations valid within the scope of linear elastic fracture mechanics:

$$G_{1c} = J_{1c} = [(1 - \nu^2)/E]K_{1c}^2 = 2(1 - \nu^2)\delta_c Y \tag{2.1}$$

where the plane strain condition prevails, and $\nu$, E and Y are Poisson's ration, Young's modulus and 0.2% yield stress in uniaxial tension, respectively. Among the indice of fracture toughness, the critical crack opening displace-ment is the one that is physically meaningful and is directly related to the microstructure of the solid.

Before taking up the mechanics and metallurgical considerations of the

correlations between the fracture toughness and microstructures, it is of interest to point to the different regions of stress-strain behavior near the crack tip and the order of magnitude of their sizes, see Fig. 2.1. It is noted that there exists a plastic region near the crack tip even for brittle materials unless the yield stress is not defined. Nevertheless, a general trend is that the more brittle the material, the smaller the plastic region at the crack tip.

## 2.2 Mechanics and Metallurgical Considerations

The criterion for crack extension in a solid was originally proposed by Griffith [2.1]* for glass and later modified by Orowan [2.2] and Irwin [2.3] for metals. The principle states that the elastic energy is released in the vicinity of the crack tip when the crack grows, that the energy consumed per unit area of crack extension is a material property, and that the total energy is conserved during the process of crack extension. The elastic strain energy release rate G for crack extension can be defined for a static crack system as:

$$G = \frac{dW}{dA} - \frac{dU}{dA} \tag{2.2}$$

Where W, and U are work done to the solid and internal energy in solid, respectively. When the applied load reaches a critical value, the crack extends, and the G-value corresponds to the critical situation is equal to the measured value of the fracture toughness, $G_{1c}$. Assuming the crack extends in such a

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* A bracket indicates number of references at end of the section.

manner that the work done is equivalent to the plastic work plus the energy to create new surfaces, i.e., neglecting kinetic energy carried by the out-going elastic waves, the fracture toughness can be related to the tensile properties of the materials. Depending upon the failure criterion used, a few models were proposed.

(i) The Continuum Plasticity Model (Fig. 2.2)

Let the crack extend by a distance $\delta a$, the energy release per unit thickness is

$$\delta W_2 = \frac{\pi a \sigma^2 (1 - \nu^2)}{2E} \delta a = \frac{\pi K_1^2 (1 - \nu^2)}{2E} \delta a \qquad (2.3)$$

If the plastic work is done within the region of size R, see Fig. 2.2., we have

$$\delta W_2 = 2RY\varepsilon_f \delta a + 2\gamma \delta a \qquad (2.4)$$

in which $\gamma$ is specific surface energy and $\varepsilon_f$ is true strain at fracture.

Employing Eqs. (2.3) and (2.4), Krafft and Irwin [2.4] obtained:

$$K_{1c} = \left[ \frac{4E}{\pi (1 - \nu^2)} \right]^{1/2} (RY\varepsilon_f + \gamma)^{1/2} \qquad (2.5)$$

(ii) The Ligament Model

Setting the elastic stress $\sigma_{ij} \propto K_1 / \sqrt{r}$ equal to $E\varepsilon_f$ for failure of a ligament ahead of a crack and letting r equal to the ligament spacing $\ell$, see Fig. 2.3, Krafft [2.5] derived the following relation for $K_{1c}$

$$K_{1c} = C_1 E \varepsilon_f \sqrt{\ell} \qquad (2.6)$$

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where $C_1$ is a constant and he later replaced $\epsilon_f$ by n, the hardening exponent in the Ludwik type hardening law.

Hahn and co-works [2.6, 2.7] showed, at temperature just below the transition temperature, certain grains will fail to cleave and will abosrb a large amount of plastic work and act as ligaments. They obtained the following for $K_{1c}$

$$K_{1c} = \frac{d}{\ell} \sqrt{EY\epsilon_f} \tag{2.7}$$

in which d is the diameter of the cracked particles.

(iii)  Linking-up Model (Fig. 2.4) (Fig. 2.5)

McClintock, Kaplan, and Berg [2.8] proposed that voids near crack tip interact thereby favoring localized intense shear between neighboring voids. In addition to promoting void growth, the intense shear can damage small precipitate particles causing void nucleation, growth and joining-up on an even finer scale. As already shown in Fig. 2.1, this involves more complicated stress-strain environment of the crack-tip as the finer scale is approached. Using a detailed continuum plasticity analysis, Rice and Johnson [2.9], Fig. 2.5, found that crack extension proceeds when the extent of the heavily deformed region is comparable to the width of the unbroken ligament separating cracked particles:

$$\ell \cong \delta_c \tag{2.8}$$

The quantity, $\delta_c$ can be related to $K_{1c}$ by way of Eq. (2.1)

$$K_{1c} = \sqrt{2YE\ell} \tag{2.9}$$

These relations can then be related in terms of $f_c$ and d, the volume fraction and diameter of cracked particles, respectively, to fracture toughness, and Hahn and Rosenfield [2.10] showed:

$$K_{1c} = 2YE \left(\frac{\pi}{\sigma}\right)^{1/2} d^{1/2} f_c^{-1/6} \qquad (2.10)$$

(iv)  Crack Path Tortuosity Model (Fig. 2.6)

McClintock [2.11] proposed that if interface decohesion is important, then a large aspect ratio of second phase particles would promote crack path tortuosity and hence increase $K_{1c}$. The relation is expressed by

$$K_{1c} = A\epsilon_f \qquad (2.11)$$

where A is the particle aspect ratio [2.12]

2.3  Results of Related Metallographical Studies

In what follows, metallographical studies of certain alloys of interest are sampled and viewed in the light of the above mentioned models.

(i)  Aluminum

In Table I, the mechanical properties of some commercial aluminum and corresponding metallographic data for large inclusion particles are given by Low, van Stone and Merchant [2.13]. Note the numbers for $\delta_c$ and the center-to-center spacing of inclusion are roughly equal. Fracture toughness essentially increases as $\delta_c$ increases. The number $f_c^{-1/6}$ is about ten times less than $\ell$ except for 2124-T851 and 7075-T7351. Particle diameter does not seem to play an important role in $K_{1c}$.

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(ii)  Titanium

Another material of interest to aerospace is titanium.  It is corrosion
resistant as well as having beneficial strength to weight ratio.  Large
amount of work is done for titanium alloys.  During the course of three
separate research programs, Froes and co-workers [2.14, 2.15] and Chesnutt,
etc. [2.16] measured a large number of fracture toughness values, both valid
$K_{1c}$ and $K_Q$, for titanium alloys, and corresponding tensile properties.
They first describe the variations of microstructure and the second describe
the effect of microstructure on fracture toughness and topography.  A summary
of the mechanical properties and corresponding microstructures features for
three different alloys under specific process and heat treatment is given in
Table II [2.17].  It is again seen that for constant aspect ratio, larger
dimple size, implying larger center-to-center spacing of inclusions would give
larger fracture toughness.  On the other hand, for the same dimple size,
larger aspect ratio of the second phase particle would yield larger fracture
toughness.

(iii)  Alumina $Al_2O_3$

For a number of years the influence of microstructure such as grain
size on the fracture toughness of polycrystalline $Al_2O_3$ has been unclear.
Although the material is brittle, it is highly anisotropic.  Some sources
suggested an increase of toughness with grain size [2.18] [2.19] and others
a decrease [2.20].  Furthermore, the results appear to be dependent upon
the specimen geometry used, for example, the D.C.B. test gives results for
coarse-grained alumina almost twice those of the S.E.N.B. test [2.21].
New specimen geometry is currently under development [2.22, 2.23].  It
appears that there is not enough evidence to draw any conclusion as to

how the microstructure of the polycrystalline $Al_2O_3$ influences its toughness. However, the general trend described in (i) and (ii) is expected to be followed. A text given by Lawn and Wilshaw [2.24] is useful for the understanding of the fracture of brittle solids. Microstructural aspects and atomic apsects and atomic aspects of crack propagation are discussed, however, no indications are given in regard to the correlation between fracture toughness and micro-structural factors.

## 2.4 Deformation and Fracture Processes

As early as 1928, metallurgists already were aware of the phenomenon of "tin cry". During heat treatments of steels, audible sounds of "clicks" were noted and were found to be related to the martensitic transformation [2.25]. When materials undergo plastic deformation of fracture elastic energy is spontaneously released in the form of waves and is detectable away from the sources. A study known as "acoustic emission" has been used as a means of detection and determination of flaw size in materials [2.26] [2.27].

Two types of acoustic emissions were detected: a quasi-continuous signal and a burst-type signal. It was found that the first type signal was strain-rate sensitive and hence might be related to dislocation pinning and cross slip, and the burst type signal was probably related to the rapid deformation mechanisms responsible for the formation of stacking faults and mechanical twins [2.28].

It appears that there should be no doubt that the deformation and fracture process during the initiation and subsequent crack propagation is a dynamic one. It appears reasonable to hypothesize that the process involves not only local plastic instability but also local sudden unloading. This sudden

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unloading causes waves to be sent out through the material medium.

2.5  Summary

The above review suggests that fracture toughness is influenced by the microstructure of alloys in such a manner that:

(a)  it increases as the plastic zone size, or equivalently, the crack tip opening displacement, increases,

(b)  the crack tip opening displacement should roughly equal both the spacing of inclusions or second-phase particles which determine hole spacing and the size of the intensely deformed plastic zone at the crack tip, and

(c)  the aspect ratio of second-phase particle that causes crack tortuosity and crack branching should increase the apparent macroscopic $K_{lc}$.

The deformation processes associated with crack initiation involves local plastic instability and sudden unloading which induce waves to be sent through the material.  It is suspected that the interaction of these waves with neighboring particles or inclusions may have caused crack propagation under critical situations.  It is indeed this dynamic nature of the deformation process associated with crack initiation that serves as the key to the quantitative ultrasonic determination of fracture toughness.

III.  QUANTITATIVE ULTRASONICS AND MICROSTRUCTURE

3.1  Background

With ultronic measurements it is possible to investigate those properties of materials which determine the velocity of propagation and the attenuation of ultrasonic waves in them.  The intensity of ultrasonic waves is influenced

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not only by the medium of propagation but also by a geometrical factor--the divergence of the beam of waves. The ultrasonic attenuation caused by the microstructure can be divided into scattering and absorption. Recent progresses in quantitative NDE techniques have generated great interest in the field of quantitative ultrasonic evaluation of mechanical properties of engineering materials [3.1 to 3.3]. An extensive and useful review on very recent advances is given by Vary [3.4]. Further reference can be found in [3.5].

3.2 Tensile Properties

An elastic medium that is isotropic possesses only two independent elastic constants. Thus if the density has already been measured, all the elastic constants for this medium can be determined from two different types of waves propagated in it, $C_L$ the longitudinal wave speed and $C_T$ the shear wave speed, as follows:

$$E = \rho C_T^2 (3C_L^2 - 4C_T^2)/(C_L^2 - C_T^2) \qquad \text{(Young's modulus)} \qquad (3.1)$$

$$\mu = \rho C_T^2 \qquad \text{(Shear modulus)} \qquad (3.2)$$

Because these relations are not linear, a careful study must be made of the effects of errors in the measurements of the velocities on the resulted errors in the calculations. These constants that are obtained from wave velocity measurements are called dynamic elastic constants. For more references, see [3.6-3.12].

3.3 Yield Stress and Fracture Toughness

Due to the extremely complex nature of the deformation mechanism

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involved during yield and fracture, there is no theoretical relation between Y, $K_{1c}$ and ultrasonic measurements, such as Eqs. (3.1) and (3.2), available at this time.

Vary [3.13] was able to measure the ultrasonic attenuation dependence on frequencies and discovered some empirical relations between the fracture toughness, yield strength and critical attenuation factors related to grain sizes in two maraging steels and titanium alloys. Subsequently, Vary [3.14] gave some considerations of the stress wave energy required to create a microcrack of "diameter" d to the critical crack tip opening displacement $\delta_c$. Equating the energy loss by Rayleigh scattering, associated with a proposed two-grain microstructure model to $K_{1c}^2/Y$ or critical crack tip opening displacement, he found a quantitative relation between the fracture toughness $K_{1c}$, the yield stress and ultrasonic measurements, as

$$K_{1c}^2/Y = \psi(v_\ell \beta_\delta/m)^{1/2} \tag{3.3}$$

where the RHS is directly related to quantitative ultrasonics in which $v_\ell$ is longitudinal velocity, $\beta_\delta$ is the attenuation slope measured at wave length $\lambda = \delta_c$ and m and $\psi$ are constants. The attenuation factor is assumed to relate to the frequency by power law. The stress wave energy required to create a microcrack of size d is equated to the energy loss by Rayleigh scattering of stress wave as the wave initiates from the source grain and reaches the scattering grain. The model is thus useful in pursuing the understanding of ultrasonic determination of fracture toughness, see Fig. 3.1.

From the experimental results obtained in [3.13], Vary also provided a second equation of the form

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$$Y + AK_{1c} + B\beta_1 = C \qquad (3.4)$$

where A, B, C are dimensional constants and $\beta_1$ is the attenuation slope measured at the frequency for which attenuation factor $\alpha$ is unity, for a complete ultrasonic evaluation of the yield stress and fracture toughness. Currently, quantitative ultrasonic techniques are also being developed for determining strength of composites [3.15-3.18].

3.4 Ultrasonic Attenuation, Velocity and Microstructure

The propagation of high-frequency stress waves in solids is determined by the attenuation and velocity of the stress wave. Various types of defects such as dislocations, and a change in the type or density of defects will usually change the propagation behavior of the stress wave. In general, dislocation damping is of importance in the case of low-frequency waves and the direct scattering by defects and second phase particles is of importance in the case of high-frequency waves [3.2].

Scattering of stress waves in a solid is brought about by differences in mass density and elastic properties from point to point. These differences are related to grain boundaries, precipitates, inhomogeneities in composition, and even to smaller groups of defects whose extent may be measured in terms of lattice spacings.

With availability of higher and higher frequencies for examining material properties, it is now possible to detect directly fewer defects of smaller and smaller sizes [3.15, 3.19, 3.20].

Various methods are now available for measuring velocity and attenuation in solids. A detailed discussion of the methods is given in [3.1], pp. 68-91.

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Vary gave an account for computer signal processing in this respect [3.21].

The general problem of expressing analytically the attenuation by scattering for an ultrasonic beam propagating through a medium containing scatterers with any size, shape, distribution of sizes, density of scatterers, wave length relative to scatterer size, etc. is a very complex one. A brief review is given next in Section IV and some available mathematical methods are discussed.

## 3.5 Summary

Experimental measurements using ultrasonic techniques indicate that there is a correlation between fracture toughness and ultrasonic factors. It appears that ultrasonic measurements at certain frequencies, closely related to material microstructure, can provide means for determining fracture toughness. If a power law relation can be established between the ultrasonic attenuation factor $\alpha$ and the ultrasonic frequency, certain theoretical relations can be derived by using a simple two-grain model. It does seem to be feasible to determine fracture toughness by the quantitative ultrasonic measurements.

## IV. ELASTIC WAVE SCATTERING AND RELATED MATHEMATICAL METHODS

### 4.1 Background

The model discussed in the previous section suggests that elastic wave scattering plays an important role in determining fracture toughness by ultrasonics. This leads to a review of this area.

The recent thrust in studying the scattering of waves in an elastic solid has been highly motivated by its applications in various fields such as seismic explorations, nondestructive testing, material properties, and dynamic stress concentration. An excellent account of history and fundamentals is

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given in [4.1] and a comprehensive discussion of applications from a theoretical viewpoint is found in [4.2 - 4.4].

The deviation of the wave from its original path is known as the diffraction, and the radiation of secondary waves from an embedded obstacle is called the scattering. In elastic medium the obstacle may be in the form of a crack, a rigid body or a second phase particle with different moduli from that of the medium. The problem of an elastic scatterer is more difficult than that of a vacancy or a rigid inclusion in that both the displacements and tractions at the interface are unknowns.

Historically, many theories of scattering, until a few years ago, dealt with scaler waves and simple obstacles. Within that context two regimes were apparently distinct from each other, the long wave length, and the short wave length or imaging regime. The treatment of vector fields in elastic solids is much more complex than classical wave fields. Several useful techniques are extended from the classical fields of scalar waves and electromagnetic waves. Three such methods useful at high frequency range are briefly sketched.

It is interesting to note that the bulk of the boundary value problems studied are not related to the problem of fracture toughness determination but to the application of flaw detection [4.5, 4.6].

4.2 Geometric Diffraction Method

At high frequencies the diffraction of elastic waves by obstacles can be analyzed on the basis of elastodynamic ray theory [4.7]. For time harmonic wave motion, ray theory provides a method to trace the amplitude of a disturbance as it propagates along a ray. The technique of geometrical diffraction theory was introduced by Keller [4.8]. The application of ray theory to diffraction by smooth obstacles was investigated by Resende [4.9] and to

diffraction by cracks was studied by Achenbach, Gautesen and McMaken [4.16].

Geometrical diffraction theory is based on the use of certain canonical exact solutions, for example, the Kirchoff solution for an edge, the diffraction of a plane wave by a semi-infinite crack. These canonical solutions are appropriately adjusted to account for curvatures of incident wave fronts, edges and finite dimensions. The pertinent canonical solution must first be obtained.

Achenbach, et. al. [4.10] showed that the method provides good results for normal incident waves on slits and penny-shaped cracks at relatively small values of the frequency ($K_L a > 1.5$) and relatively close to the crack. For more complicated geometries, say elliptical cracks, the corrections at shadow boundaries and caustics become very cumbersome.

4.3 Transition Matrix Method

The general theory of the scattering of acoustic waves is contained in the mathematical theory of Huygen's principle [4.11]. The scattered waves outside the scatterer are related to the Helmholtz integral over the surface of the scatterer. Waterman was first to introduce the transition matrix (T-matrix) method for acoustic scattering [4.12] and later [4.13] for electro-magnetic waves. He started from the Helmholtz integral formula and expanded both the incident and scattered waves in series of the basis functions. Using the orthogonality of these basis functions, he showed that the unknown coefficients of the scattered waves are related by the transition matrix to the coefficients of the incident waves. This approach was applied, in 1976, simultaneously by Waterman [4.14] and Varatharajulu and Pao [4.15] to the scattering of elastic waves. An excellent summary of how the method can be used for acoustic waves and elastic waves is given by Pao [4.16]. The transfer

matrix is derived for waves scattered by an inclusion of arbitrary shape.

The T-matrix has infinite number of elements. These elements are integrals of the basis functions over the bounding surface of the scatterer and depend upon the incident wave frequency, geometry of the scatterer and the material properties. The integrals are evaluated numerically. Currently, numerical results are available for the scattering of compressive waves and shear waves by a single infinite cylinder or a spheroid [4.17, 4.18].

4.4 The Integral Equation Method

The theory developed by Fredholm [4.19, 4.20] for the solution of certain types of linear integral equations has been well-known and widely used in mathematical physics and mechanics. The main reason being at least two fold: (1) if the kernel is separable, the problem of solving an integral equation of the second kind reduces to that of solving an algebraic system of equations. Any reasonably well-behaved kernel can be written as an infinite series of degenerate kernals [4.21]. (2) The Fredholm theorems provide an assurance of the existence of the solution hence approximate methods can be applied with confidence when the integral equation cannot be solved in close form [4.20].

This method was applied to problems in quantum scattering first by Jost and Pais [4.22] in a discussion of the convergence of the Born series. Reinhardt and Szabo [4.23] suggested a numerical procedure for elastic scatter-ing by constructing the Fredholm determinant which contains all the scattering information. Using a similar approach, Holt and Santoso applied the method first to scalar wave scattering [4.24] and then to vector wave scattering [4.25]. They were successful in both cases. They further studied the simple model of a collinear atom-molecule inelastic collision [4.26]. This results in an integral equation in two variables in which an infinite set of coupled channels

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*occurs. The method was again successful.*

The integral representation technique is closely related to the Fredholm integral equation method. Eyges [4.27] used the integral representation technique to solve of Schrodinger and related equations for irregular and composite regions. He also calculated modes of two coupled electromagnetic dielectric wave guides [4.28], see Fig. 4.1.

Gubernatis, Domany and Krumhansl [4.29] presented a theoretical study of the scattering of ultrasonic waves from a single flaw embedded in an isotropic medium through the use of the integral equation formulation. The integral equation method permits a systemmatic generation of approximations with which the scattering of ultrasonic waves from nonspherical shapes can be treated. They subsequently [4.30] obtained the Born approximation, analogous to that in quantum mechanics. The Born approximation is of broad utility, however, it breaks down for strongly scattering flaws. As a remedy, Gubernatis [4.31] developed a quasi-static approximation. An interesting conclusion he had is that the elastic scattering at long wavelengths is not isotropic. Different shapes produce different angular distributions. This feature is distinctive to elastic scattering while the scattering for acoustical and quantum mechanical problems is isotropic at long wavelengths.

## 4.5 Additional Remarks

Although developments over the past few years have led to promising studies of theoretical methods for treating scattering of ultrasound by defects and second phase particles in elastic solids, the field is still in its infancy. A variety of useful techniques such as partial wave expansions, T-matrix, integral equation techniques, variational methods and geometrical diffraction theory are being studied.

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Krumhansl [4.32] recently commented on the question "which method should you use in which regime?" He argued that, in engineering, a useful theoretical framework whould be one which physical intuition or engineering data can be entered as conveniently as possible, while, on the other hand, for any approximate method, the limits of that approximation must exist. He and his group of researchers, therefore, chose to concentrate, at least in the initial phase, on the integral equation method. He recognized, however, the partial wave expansion, including work by Ying and Truell [4.33] and Varatharajulu and Pao [4.15], will furnish important reference for the theoretician and, eventually, for the experimentalist.

A comparison of the Fredholm integral equation method with the T-matrix approach in the scattering of electromagnetic waves is given by Holt [4.34]. He pointed out that the two methods are entirely different and in many ways complementary. They are independent of each other and can serve as independent checks of the results obtained. Since the Fredholm integral equation method is a convergent method and is numerically stable, it would seem probable that at least for spheroids (of maximum dimension a) the Fredholm integral equation can be used at values ka, where k is the wave vector, greater than those the T-matrix method can treat. For the purpose of completeness in giving a review, his summary of the comparison is given in Table III.

There are currently many efforts relating the theory and some sophisticated interpretation of experimental results on single defects. Not much, however, is done for multiple defects.

## V.  SUMMARY AND SUGGESTED FUTURE RESEARCH

From the reviews given in the previous sections, it is seen that the feasibility of determining fracture toughness by quantitative ultrasonics can be

based upon the findings that fracture toughness is governed by certain material microstructure factors, i.e. aspect ratio and spacing of second-phase particles or flaws. Since interactions of ultrasonic wave with microstructure are measurable in terms of ultrasonic attenuation and velocity, it is possible that these measurements obtained at particular frequencies can be correlated to determine fracture toughness. Such evidences now exist. Confirmation and guidance to experimental set-up is, however, still lacking.

Most of the theoretical studies on elastic scattering emphasize flaw determination, i.e., the reconstruction of flaw geometry, rather than the determination of fracture toughness. As a result, most studies are concerned with a single flaw of size much larger than that of the second-phase particles and their spacings. The reconstruction, or the inverse problem, requires measurements at all frequencies. Although not much attention is given to the problem of ultrasonic determination of fracture toughness, the mathematical methods that have been under development in regard to the flaw determination problem can be used for the fracture toughness determination problem. It appears that the integral equation method is most promising.

In searching for a theoretical foundation for ultrasonic determination of fracture toughness, it is necessary to consider interactions of ultrasonic waves with inclusions at given spacings. Elastic scattering of multiple inclusions will provide essential information for the determination, i.e. the functional dependence of the ultrasonic attenuation $\alpha$ upon frenquency $f$ must be theoretically established. The range of wavelengths of interest must correspond to the sizes of grain, subgrain, inclusions and second-phase particles that comprise the scattering field.

To accomplish these, it is recommended that the two-grain model suggested by Vary and a model where a periodic distribution of grains is present be used

to study the elastic scattering. These models should provide the functional relation between $\alpha$ and $f$. A comparison of the outcome of the two models should provide insight as to how dominant the first neighboring grain next to the source grain is to the scattering field.

The shape of the grains should probably be taken as cylindrical or spherical in the beginning for simplicity purpose. To study the effect of aspect ratio in the value of fracture toughness, more complicated geometries such as ellipses or spheroids, prolate and oblate, will have to be investigated. Since the elastic scattering is anisotropic in many cases, the theoretical study will eventually provide insight and guidance to the actual experimental set-up for an ultrasonic determination of fracture toughness.

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TABLE I. MECHANICAL PROPERTIES OF SOME COMMERCIAL ALUMINUM ALLOYS AND CORRESPONDING METALLOGRAPHIC DATA FOR LARGE INCLUSION PARTICLES AFTER LOW, VAN STONE AND MERCHANT [2.13] AND OTHERS

| Alloy | 2014-T6 | 2024-T851 | 2124-T851 | 7075-T7351 | 7079-T651 |
|---|---|---|---|---|---|
| Yield Stress, $MNm^{-2}$ (ksi) | 436 (63.5) | 443 (64.4) | 435 (63.3) | 391 (56.8) | 502 (73.0) |
| $K_{Ic}$, $MNm^{-3/2}$ | 19.5 | 20.6 | 26.2 | 30.9 | 27.4 |
| $\delta_c$, $\mu m$ | 6.3 | 7.0 | 11.4 | 17.7 | 10.8 |
| Average Spacing of Large Dimples, $\mu m$ | 11 | 9.6 | 8.6 | - | 9.5 |
| Identity of Large inclusions | Q-Phase $(Al_4CuMgSi_4)$ $Al_{12}(Fe,Mn)_3Si$ | $Al_2Cu(Mn,Fe)_3$ $CuAl_2$ $Mg_2Si$ | $Al_{20}Cu_2(Mn,Fe)_3$ | $(Fe,Mn,Cu)Al_6$ $Mg,Si$ $CuAl_2$ | $(Fe,Mn,Cu)Al_6$ $Mg_2Si$ |
| Volume Fraction of inclusions, % = f | 2.9 | 4.4 | 1.0 | 2.6 | 0.9 |
| $f_c^{-1/6}$ | 8.4 | 7.8 | 1.0 | 0.85 | 1.02 |
| Center-to-Center Spacing of inclusions, $\mu m$ | 8.4 | 7.4 | 9.2 | 11.6 | 9.4 |
| Average particle diameter, $\mu m$ | 5.8 | 5.6 | 4.5 | 5.7 | 4.2 |

$\delta_c$ is the critical crack-tip opening displacement ($\delta_c = \dfrac{0.5\ K_{Ic}^2}{EY}$) = $COD_c$

TABLE II. MECHANICAL PROPERTIES AND CORRESPONDING MICROSTRUCTURAL FEATURES

| Alloy Condition | Alloy | Direction | Fracture Toughness ($K_{1c}$), MPa m(ksi in.) | Ultimate Tensile Strength, MPa (ksi) | Yield Strength MPa (ksi) | Elongation % | Reduction of Area, % | Primary α Aspect Ratio | Primary α Spacing, pm | Dimple Size µm |
|---|---|---|---|---|---|---|---|---|---|---|
| A | 334 | L | (LR) 92 (84) | 1330 (193) | 1260 (183) | 7 | 15 | 6 | 2 to 5 | 10 to 15 |
|  |  | T | (CR) 63 (57) | 1340 (194) | 1275 (185) | 2 | 4 |  |  |  |
|  | 227 | L | (LR) 65 (59) | 1240 (180) | 1200 (174) | 8 | 24 | 6 | 5 to 10 | 10 to 15 |
|  |  | T | (CR) 56 (51) | 1230 (178) | 1180 (171) | 4 | 12 |  |  |  |
| B | 334 | L | (LR) 64 (58) | 1185 (172) | 1150 (167) | 4 | 14 | 6 | 2 to 5 | 5 to 10 |
|  |  | T | (CL) 50 (45) | 1260 (183) | 1225 (178) | 2 | 6 |  |  |  |
|  | 227 | L | (LR) 63 (57) | 1305 (189) | 1240 (180) | 1 | 5 | 6 | 10 | 5 to 10 |
|  |  | T | (CL) 45 (41) | 1360 (197) | 1295 (188) | 1 | 6 |  |  |  |
| C | 334 | L | (LT) 71 (65) | 1105 (160) | 1095 (159) | 13 | 36 | 3 | 5 to 10 | 10 to 15 |
|  |  | T | (TL) 66 (60) | 1235 (179) | 1180 (171) | 5 | 9 |  |  |  |
|  | 227 | L | (LT) 53 (48) | 1180 (171) | 1130 (164) | 13 | 47 | 2 | 5 to 10 | 10 to 15 |
|  |  | T | (TL) 47 (43) | 1275 (185) | 1225 (178) | 11 | 32 |  |  |  |
| D | 334 | L |  | 940 (136) | 890 (129) | 23 | 52 | 3 | 2 | 10 to 15 |
|  |  | T |  | 930 (135) | 905 (131) | 15 | 44 |  |  |  |
|  | 227 | L |  | 860 (125) | 840 (122) | 20 | 57 | 5 | 2 to 5 | 5 to 10 |
|  |  | T |  | 885 (128) | 855 (124) | 15 | 48 |  |  |  |

TABLE III.  COMPARISON OF BASIC FIM AND T-MATRIX METHODS [4.34]

| Aspect of Method | Fredholm Integral Equation Method | T-Matrix Method |
|---|---|---|
| Formulation | Volume integral equation for electric field | Surface integral equation and Huygens principle |
| Scattering Parameters determined from | Internal Field via integration | External Field via asymptotic form |
| Part of Field removed from calculation | External Field | Internal Field* |
| Surface enters calculation | Implicitly through volume integrals | Explicitly via surface integrals – implicitly matching occurs on surface |
| Expansion in terms of | Fourier transform variable | Position space variable |
| Numerical Stability | Theoretically stable practically instabilities have not revealed themselves | Instability occurs when ka is increased too far (Ref. 6) |
| Easily Adaptable to various shapes | No | Yes (?) |
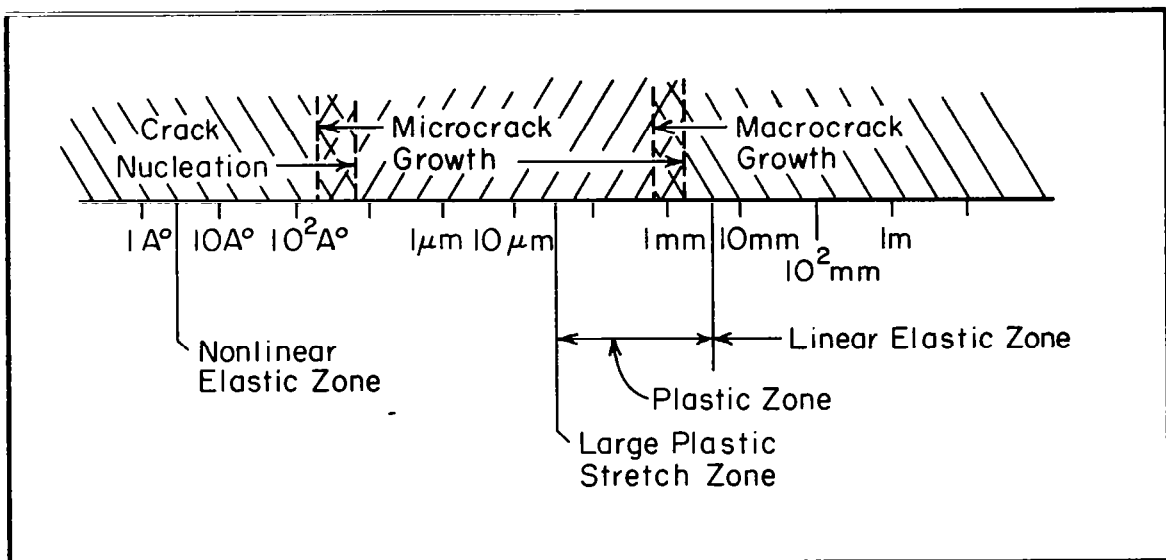
Fig. 2.1   Schijve's scale of crack dimension.
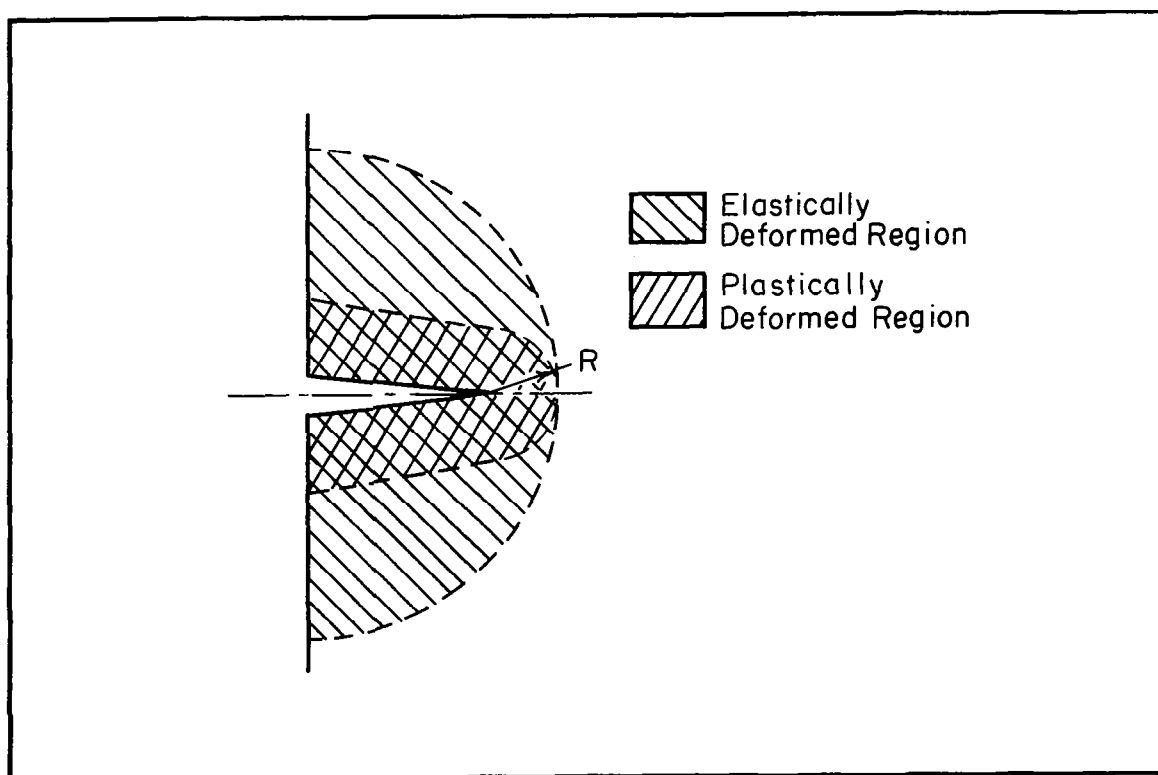
Fig. 2.2    Model for crack extension where the only work done is
the plastic work plus energy to create new surfaces.
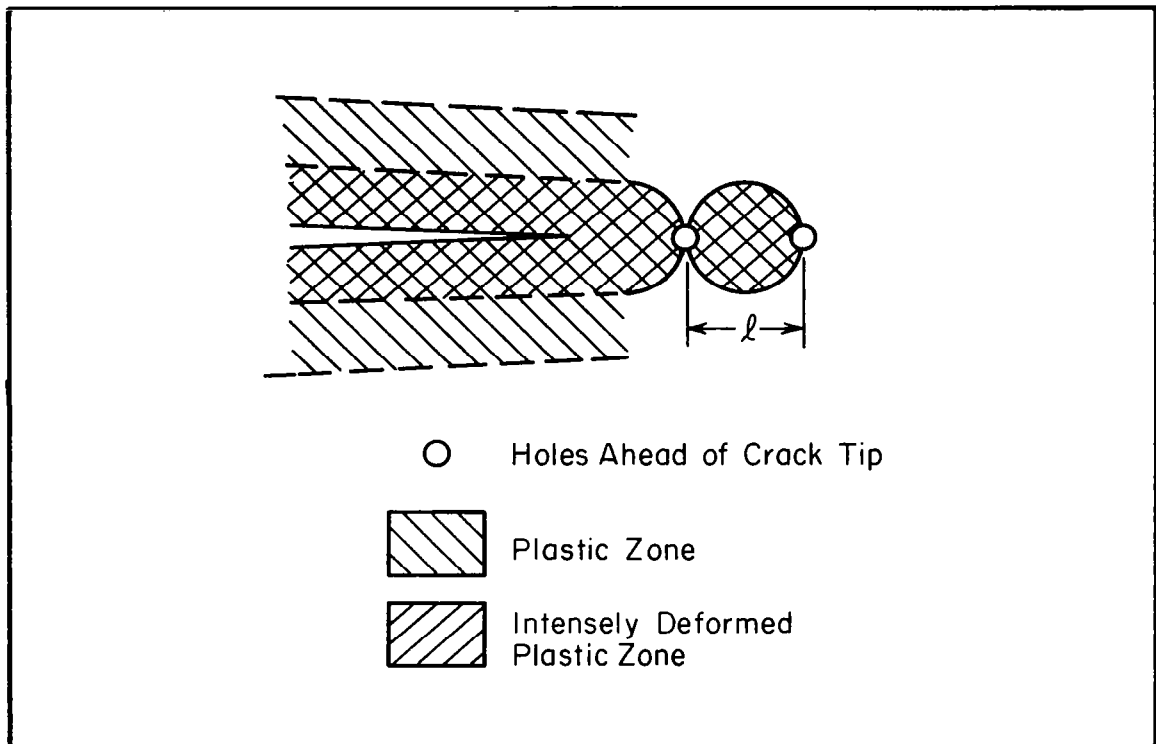
Fig. 2.3    Model demonstrating contained region of intensely deformed material with hole spacing.
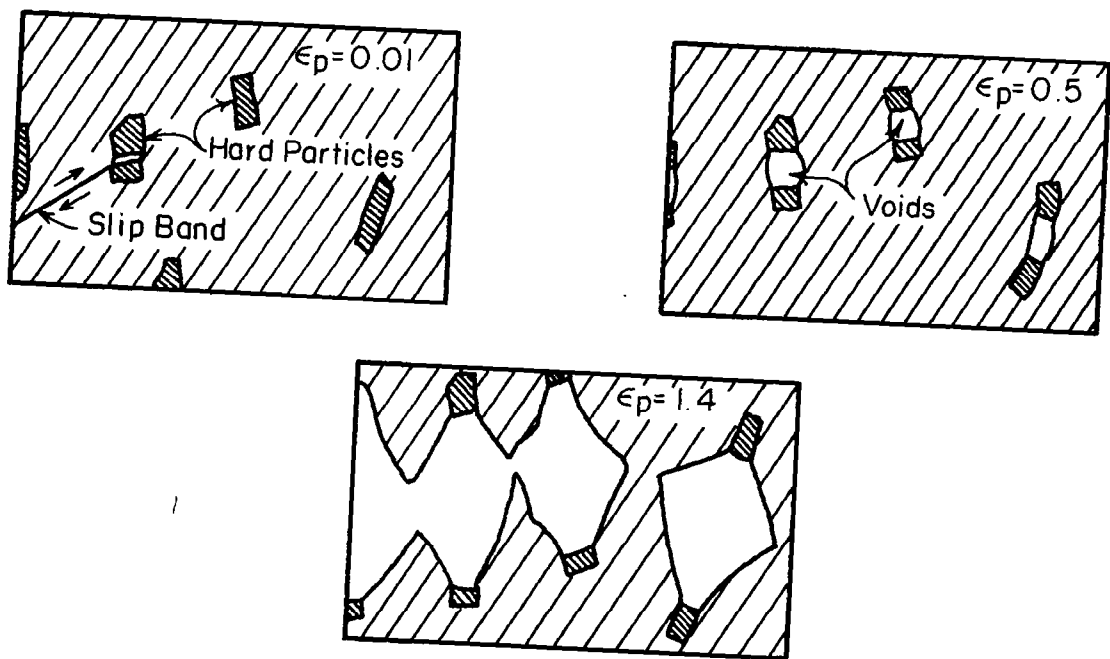
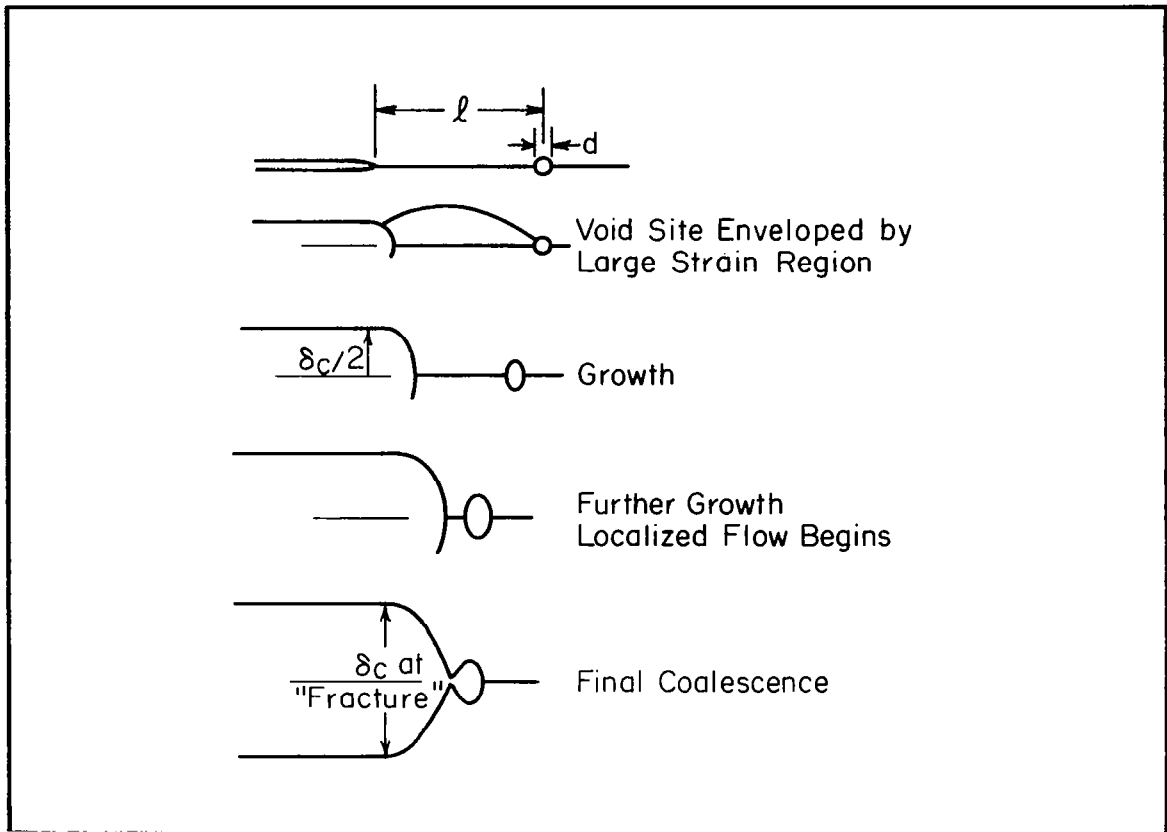Fig. 2.4    Mechanism of fibrous crack extension:   schematic.

Fig. 2.5    Model of dutile fracture by the growth and coalescence of
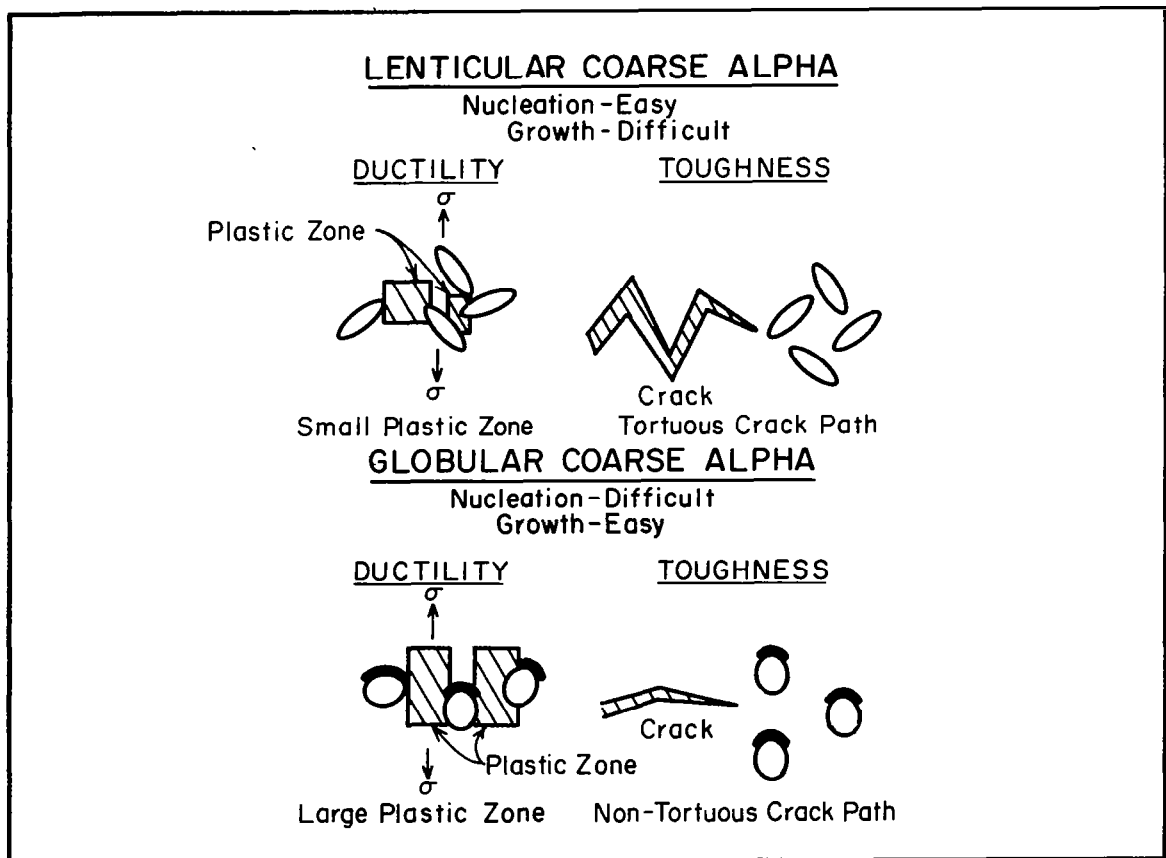an initially spherical void with the crack tip.

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Fig. 2.6    Schematic representation of effect of coarse alpha
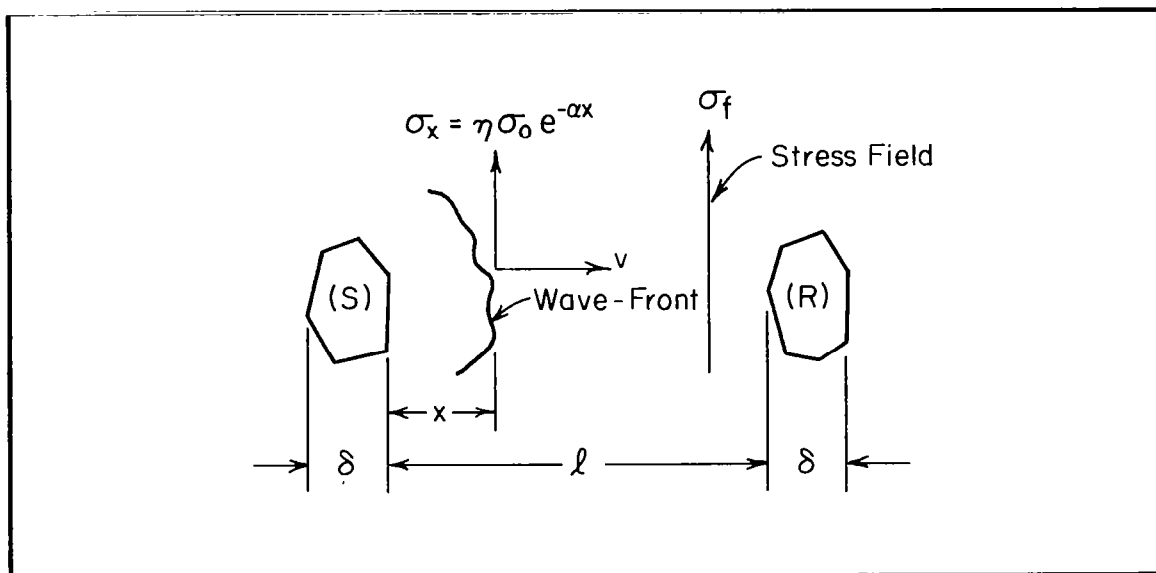shape on ductility and toughness.

Fig. 3.1   Diagram of fracture model.  "Grains" (S) and (R) are im-
bedded in a matrix subjected to a local static stress field
of magnitude $\sigma_f$.  A stress wave of initial amplitude $\sigma_0$
and velocity $v$ is emitted from (S).  The distance between
(S) and (R) is $\ell$ and "grain" size is $\delta$.
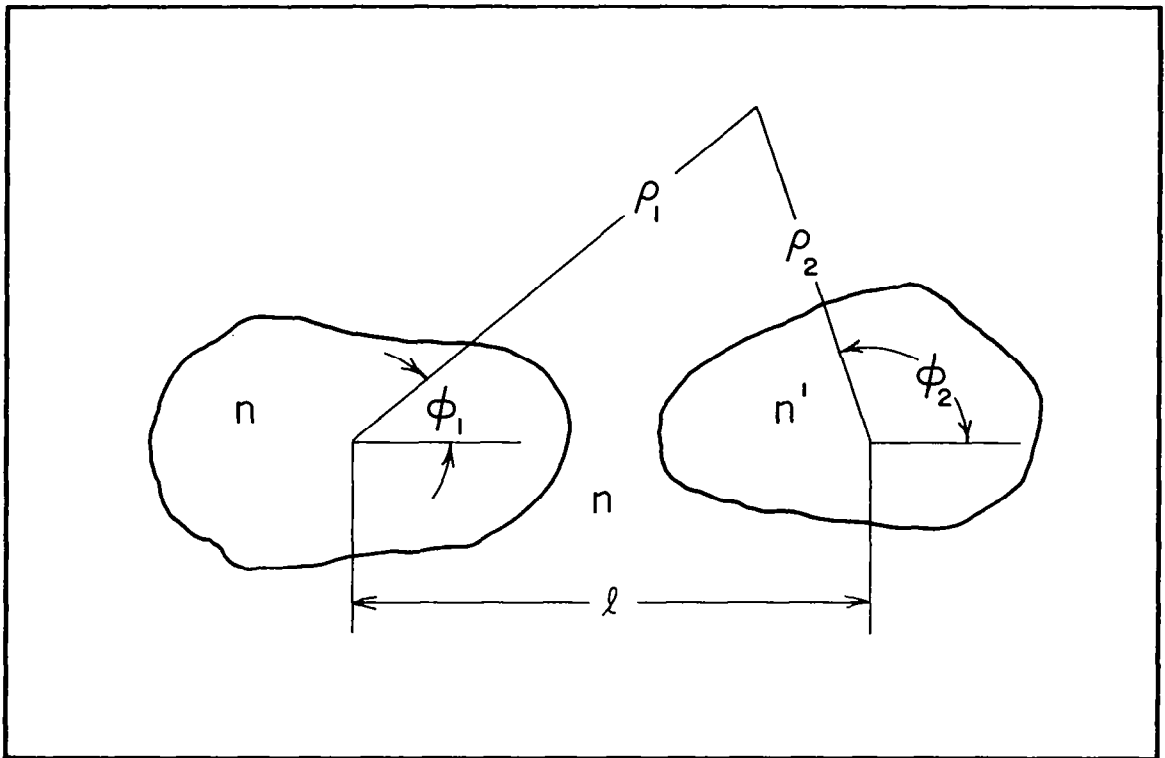
Fig. 4.1   Coordinate system for  two guides, at a distance $\ell$ apart.

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16. Abstract

The report covers three main topics: (a) Fracture Toughness and Microstructure. (b) Quantitative Ultrasonics and Microstructure, and (c) Scattering and Related Mathematical Methods. Literature in these areas is reviewed to give insight to the search of a theoretical foundation for quantitative ultrasonic measurement of fracture toughness. The literature review shows that fracture toughness is inherently related to the microstructure and, in particular, it depends upon the spacing of inclusions or second phase particles and the aspect ratio of second phase particles. There are indications that ultrasonic velocity and attenuation measurements can be used to determine fracture toughness. This leads to a review of the mathematical methods available in solving boundary value problems related to microstructural factors that govern fracture toughness and wave motion. A framework towards the theoretical study for the quantitative determination of fracture toughness is described and suggestions for future research are proposed.

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