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(M.S. Thesis)

May 1986

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THE RELEVANCE OF INCLUSIONS ON

FORMABILITY IN PUNCH-STRETCHING OF

LOW-CARBON, AK, DQ STEEL

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Master's Thesis

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THE RELEVANCE OF INCLUSIONS
ON FORMABILITY IN
PUNCH-STRETCHING OF
LOW-CARBON, AK, DQ STEEL

ABSTRACT

To investigate the relevance of inclusions on automotive sheet metal formability, void volume fractions and void size distributions from a set of punch-stretched, low carbon steel specimens were measured. Measurements from strain states corresponding to instability in uniaxial tension, plane strain, and positive biaxial stretching were compared. Overall void volume fractions were low (on the order of 1 to 2 tenths of a percent), as were the void sizes (mean size of 8 square microns) for all strain states examined. The growth of voids was measured and was found to be in agreement with the model adopted by Jalinier & Schmitt [8]. Based on experiments in punch stretching and analysis for instability in uniaxial tension, it was concluded that the current level of inclusions found in low carbon AK, DQ, sheet steels does not impose a significant limit on formability.

ACKNOWLEDGEMENTS

I am indebted to many people for their assistance in making this work both possible and better. My warmest gratitude is for my wife, Katie for her unending support, patience, and understanding through all the ups and downs of my research and studies.

I would also like to express my appreciation to my advisor, Prof. J. W. Morris, Jr. for his patience and guidance throughout, and to Dr. Robin Stevenson at General Motors for his interest in, and support of this project. I am indebted to Patricia Sing and Mark Donohue, also of the GM Research Laboratory, for providing the punch stretched specimens.

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Special commendation goes to my research project co-worker Marianela Ledezma, without whose encouragement, perserverance, and useful discussions over the past two years, the task of beginning the forming program at U.C. Berkeley would have fallen much to heavily upon my lone shoulders.

Thank you all.

CONTENTS

1 INTRODUCTION

The elimination of failures or simply improvements in automotive sheet metal formability are generally achieved through one (or both) of two methods: 1) Changes in the forming process (e.g. die shape, corner radii, lubrication, etc.) or 2) selection of an inherently "more formable" material. While the final shape of the part often limits possible tooling changes, it is primarily a question of economics in the selection and use of alternate, more formable alloys. One area of interest in this regard is the influence of inclusions on failures and to what extent they may limit possible improvements in formability. The purpose of this work is to determine the relevance of inclusions on formability in punch stretching for today's low carbon, aluminum killed (AK), drawing quality (DQ) sheet steels.

The industrial definition of failure for sheet forming operations involving stretching is <u>the onset of localized necking</u> which, in the analytical sense, means instability. Therefore it is instability in stretching which is of interest. In the automotive industry, instability in hemispherical punch stretching is of particular interest as this type of test has shown the best correlation with press shop performance [1]. It is instructive however to first examine instability in general.

1.1.1 INSTABILITY IN UNIAXIAL TENSION

It is theoretically possible to predict instability in uniaxial tension using the Considére Criterion[2] ($\partial\sigma/\partial\varepsilon=\sigma$). For example, if one applies this criterion to a material which can be described by the Hollomon relation $\sigma=K\varepsilon^{n}$ (where 'n' is the work hardening coefficient, K is the Strength constant, and no 'strain rate' sensitivity is assumed ('m'=0)) one obtains diffuse necking at $\varepsilon=n$ (fig. 1)*. For a sheet specimen, the expression $\partial\sigma_{1}/\sigma_{1}=\partial\varepsilon_{1}/2$ is used and localized

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* In this paper, "ε" refers to true strain, while "e" refers to engineering strain.

necking is predicted to occur at $\varepsilon=2n$. Thus one can define or establish the **Theoretical Forming Limit** for uniaxial tension. However this theoretical limit is not often achieved.

Some years ago it was found that the removal of inclusions (the so called "clean steels" achieved through better, though more expensive melting practices) led to increases in the forming limit. Today's formable steels are relatively clean compared to those of 10 years ago. It is of interest however to know whether **extremely** clean steels will perform significantly better than those presently used, or are the improvements so minimal that it would simply be a waste of effort and money to produce or use super-clean steels for such things as automotive skin panels where low cost is a significant factor?

1.1.2 BIAXIAL STRETCHING

Most forming operations are not done in uniaxial tension, but rather under conditions of plane stress and a variety of strain states. Failure for such strain states can be plotted on a Forming Limit Curve (FLC), which is a mapping in e_2 - e_1 space of strain states which will lead to success or failure in sheet forming. Such FLC's are empirically determined for a specific material and depend on the particular test [3]. Figure 2 shows a typical curve for mild steel determined by punch stretching. For punch stretching tests, variations in minor strain for all negative ρ (where $\rho = \varepsilon_2/\varepsilon_1$) and for minor strains up to about +12% are achieved by increasing the width of rectangular blanks, thus increasing the amount of lateral constraint supplied by the circumferential locking ring. Minor strains from +12% up to equibiaxial stretching ($\rho =1$) are achieved using fully constrained square blanks and different lubrication techniques. A complete description can be found elsewhere[4].

For use in industry, one would like to be able to predict the FLC and know how inclusions

affect it. For predictions of the Left Hand Side (LHS), the Hill [5] criterion can be used. This essentially states that the local neck must lie along a direction of zero extension, i.e. localized necking must occur in a state of **plane strain**. However for ρ positive, where all strains in the plane of the sheet are positive, the Hill theory does not apply and there is no criterion for localized necking.

Almost 20 years ago, Marciniack and Kuczynski[6] (hereafter referred to as M-K) proposed that some initial (unspecified) inhomogeniety (e.g. variations in grain size, yield strength, n, texture, etc.) existed in the sheet, which leads to the formation of a trough perpendicular to the major strain axis. They reasoned that, as deformation proceeds, the strain **inside** the trough accelerates relative to that **along** the trough and thus locally the strain ratio drifts towards plane strain. These analyses are done incrementally on a computer and the "forming limit" is defined when the ratio reaches some arbitrarily close value, say 10^{-4}, to plane strain.

With this type of analysis the prediction of localized necking for forming operations involving strain ratios of positive ρ finally seemed possible. Unfortunately, while the mathematicians and modellers found this quite satisfying, for the metallurgists it was still unclear as to the exact nature of the "initial inhomogeniety". Careful studies in this area by Azrin and Backofen[7] showed that any initial inhomogeniety found was much smaller than that required by the M-K analysis.

Recently however, a refinement of this type of approach has been used by Jalinier and Schmitt[8]. In this analysis, a random array of voids ("internal damage") is used to model and equivalent thickness defect (i.e. the M-K trough). Incorporating a **void growth** equation, which is updated and implemented at each strain increment in the calculation these researchers seemed to be able to correctly predict the FLC on the RHS. This approach has also been used recently by Barlat and Jalinier [9] for predicting the FLC for dual phase steels.

From this result, one might draw the conclusion that voids are responsible for failure in biaxial stretching and therefore the removal of inclusions can significantly raise the forming limit.

1.2 PUNCH STRETCHING

Although the "void growth" modified M-K analyses appear to give reasonable and meaningful results, it must be pointed out that there is an important qualification for these type of analyses which is usually not mentioned in the literature. The key limitation on their use is that all **analyses based on an M-K approach apply only to in-plane stretching.** As prevoiusly mentioned however, Ghosh and Hecker[3] showed that punch-stretching forming limits are always higher than in-plane limits. This was demonstrated for steels, aluminum, and brass.

In a subsequent paper[10], it was further shown that punch-stretching is an entirely different case than in-plane strecthing. It was determined that, for punch stretching, the plane strain condition is **not** a necessary condition for failure and that the location of the local neck is controlled by the friction and the geometry of the test and is relatively insensitive to local inhomogenieties. This latter fact was demonstrated by Keeler and Hecker[11] when they were unable to move the location of the neck despite intentionally drilling holes in the blank in a location away from the known necking location. The increased stability of the punch stretching test has been clearly shown by these investigations.

This is not to say that inclusions (and hence voids) have no effect on the level of the strain at necking. Hiam and Lee[12] showed that there was clearly a difference between a "clean" and a "dirty" steel in punch stretching. Though in this study, void fractions were not reported and the forming limit curves for the two inclusion levels differed significantly only for strain states away from plane strain. However the current interest is not between a clean and a dirty steel, but rather between a clean and a very clean steel.

1.3 CURRENT INVESTIGATION

Despite the recent progress in predicting Forming Limit Curves by incorporating the effect of voids, the inapplicability of these analyses to punch stretch-forming processes prevents their

use in gaining insight to the actual effect of inclusions on formability. Instead a deductive method will be utilized to determine the influence of inclusions on the possibility of improving the formability of today's low carbon ,AK, DQ steels, the following approach will be taken:

1) Obtain a range of different strain ratios (without changes in lubrication) in a set of specimens punch stretched to failure.

2) Measure void volume fractions, areas, and distributions at instability.

3) Compare the observed void growth to the model used in the Jalinier and Schmitt calculations as a check on its applicability to this material. Since punch stretching has already been shown to be more stable and insensitive to small local inhomogenieties compared with in-plane stretching, agreement between the two will assure conservative effect predictions.

4) Apply theoretical work for the left hand side of the forming limit curve to the right hand side to obtain an order of magnitude of the effect.

2 EXPERIMENTAL PROCEDURE

2.1 TESTING

Sheet of thickness 0.83 mm with composition and mechanical properties listed in tables 1 and 2 respectively was electroetched on both sides with 2.54 mm diameter circles prior to shearing for the purpose of strain measurement after testing. A total of seven specimen blanks were cut 177.8 mm (7 inches) long in the rolling direction and varied in width from 25.4 mm (1 inch) to 177.8 mm in 25.4 mm increments. Punch stretching was done in an MTS model 866 testing machine with a ram speed of 254 mm per minute. The machine was programmed to automatically terminate testing upon recognition of a drop in load, which was concurrent with necking and/or tearing. A 50.8 mm radius punch was used. All sheets were punched dry but were oiled after punching to inhibit rusting. The sheets were clamped with a fully circumferential locking bead around the die cavity in the usual manner for such LDH-type tests. The seven dome widths gave minor strains ranging from 34% to +15%. (In punch stretching, equibiaxial strain can be achieved only by significantly reducing the friction conditions such as with the use of a polyethelyne spacer. To avoid effects due to change in friction conditions, these spacers were not used.) The domes primarily used in this investigation were selected to have the extreme negative and positive minor strains possible as well as one near plane strain (labelled as domes 1, 7, and 5 respectively in figure 3).

2.2 STRAIN MEASUREMENT

Macroscopic surface strains on the punched specimens were measured using a modification of the circle grid technique[13] to account for the neck and tear (see figure 4). Magnified photographs of the ellipses straddling the neck were measured using a Calcomp series 9000 digitizer which was interfaced to a Northstar Advantage computer. The digitized information was converted into strain measurements by a program written in GBasic by the author. Surface

strains were measured on both sides of the sheet and were found to be consistent.

2.3 SPECIMEN PREPARATION

After surface strain measurements were performed, the necked regions were removed from each specimen. The two halves of the neck were seperated using a fine jewelers saw and mounted "edge on" in standard metallographic mounts. The specimens were carefully ground and polished down to the intersection of the neck and the uniform strain region. This was defined as the instability strain (see fig. 5). No etchants were used proir to observation to avoid void enlargement. The specimen preparation process is illustrated in figure 6a.

2.4 DATA COLLECTION

At least twenty random photographs at magnifications between 350 and 400x were taken from each sample using the backscatter electron signal from an ISI model DS 130 Scanning Electron Microscope (SEM). The negatives from these were then projected onto the digitizer tablet resulting in final magnifications around 1.3 Kx. (Note: Ideally for this type of work, a great many measurements must be made, which is difficult by hand. An aid in this regard would be the use of a video image analyzer system[14] on the SEM which is interfaced to a computer so that a large number of observations and direct data collection is possible without the expense and time involved with photography and data taking as described below).

Void distributions were obtained again using the digitizer-computer set up. Voids areas, numbers and aspect ratios were digitized and input into the computer and stored in data files. File manipulation, data sorting, statistical, and other data reduction programs were then used to generate plots and tables of the results. This process is schematically illustrated in figure 6b.

3 RESULTS and DISCUSSION

3.1 INCLUSIONS and INITIAL VOID COUNT

Initial "inclusion" size distributions are shown in table 3. Triple point grain boundary carbides were also classified as inclusions for this investigation, and no further attempt was made to distinguish between the different types in statistical counting. The unifying feature was the fact that they all lead to void formation, by decohesion or cracking, during deformation. As has been reported elsewhere[15], some voids were present in the as received sheet due to decohesion of the ferrite matrix around the hard alumina and silica particles during cold rolling. The initial "void" volume fraction (4.7 x 10 $^{-4}$) was taken as the sum of these voids and the inclusion volume fraction, thus assuming a zero nucleation strain. This assumption is reasonable as several investigators have predicted and reported void nucleation in steels and other alloys at strains as low as one to two percent[16-19].

3.2 VOID FRACTIONS

3.2.1 AREA SIGNIFICANCE

Voids are proposed to initiate instability through an enhancement of geometric softening (i.e. a reduction in load bearing cross section). The magnitude of this loss in load bearing area is determined through cross sectional void area fractions. For large negative ρ, where the major strain is significantly greater than the minor strain, it is clear that loss of cross section along the major strain direction (i.e. that measured in a longitudinal view of area fraction) does not reduce the load bearing ability of the sheet and thus cannot be considered detrimental. It is only the thickness and width of the voids that is important, and thus it is the transverse cross section which is important. This is also true of all negative strain ratios up to plane strain. For positive ρ however, both transverse and longitudinal area reductions are important.

Measurements made on both sections of specimen number 6 ($\rho = +.12$) however, showed differences of only 11% in void area fraction. Hence the order of magnitude is essentially unchanged and the analysis to follow is unaffected by use of transverse section measurements for all strain ratios..

3.2.2 VOLUME FRACTIONS •

For the purposes of simplicity, the first order approximation of volume fraction being equal to area fraction was used[20]. The void volume fractions measured at the instability strain are shown in figure 7. The volume fraction for all strain ratios was between 1 and 2×10^{-3}. In order to be certain that the void volume fraction measured at the "instability strain" (as defined in figure 5) was not greatly different from that at the "point of incipient necking", several measurements from

successive longitudinal sections were taken of domes #2 and #6 (negative and positive ρ) at the neck centerline. These values are also shown in figure 7 and it is clear that no significant increase in void fraction is seen even confined within a region of 100 microns of the centerline of the neck.

3.3 VOID AREAS

Although the total volume fraction of voids is small, the way in which it is distributed can have significant implications. For a given volume fraction, a few closely spaced, large voids will have a much more detrimental effect than the same volume fraction distributed over many smaller voids. Figure 8 is a histogram for nearly one thousand voids taken from the three domes of

negative ρ, positive ρ, and plane strain. The majority of the void sizes (in transverse section) fell between 0.5 and 1.5 square microns with the medians falling around 3 square microns and the means around 8. The tail of the histogram is not shown but continues out quite far, with only the odd one or two voids appearing at areas of up to 190 square microns. This distribution is rather innocuous and is typical of that found in similar steels.

3.4 GROWTH OF VOIDS

As previously mentioned, the predictions of Jalinier and Schmitt (J-S) makes use of the concept of growth of "internal damage" (voids). The growth law which they use is a modification of the Rice and Tracey[21] model for spherical voids growing in a non-hardening matrix under triaxial stress.

$$C_V/C_{VO} = 1.92[exp(1+\rho)\varepsilon_1]$$

This approximation is used to model the growth of a single void formed by decohesion of the inclusion from the matrix. The form of the equation shows that the growth of voids depends exponentially on the strain ratio and the major strain.

Figure 9 shows the (J-S) void growth equation plotted for the three strain ratios used in the current investigation as well as that for equibiaxial stretching for comparison. It is apparent that void growth is rather accelerated for equibiaxial stretching, while it is nearly linear for ρ corresponding to uniaxial tension. This is in agreement with measurements made by Thompson and Nayak[22]. The data point shown on each curve corresponds to the growth of voids measured at the instability strain for this work. The numbers f_{vo} and f_v are defined by the total initial and instability void volume fractions, normalized by the number of voids from which each was measured.

There is good agreement with the Jalinier and Schmitt model, however this agreement may only be fortuitous. The model chosen by J-S was for the non-hardening matrix they preferred the spherical void geometry over the infinite cylindrical holes used by McClintock [23], although the latter case includes the effect of work hardening. Qualitatively, if the effect of work hardening were included in the J-S model, as is applicable for steel, a decrease in the growth rate of voids with strain would be observed.

An additional factor is that the J-S equation was derived for voids formed by decohesion. The present investigation is not restricted entirely to voids formed by decohesion. A small

fraction, estimated to be between 5 and 10 percent, originated from cracked grain boundary carbides. The J-S equation for voids formed by fracture of the particle is of the form :

$$C_v/C_{vo} = \exp(\alpha\varepsilon_1)$$

where α is a strain concentration factor ≈ 3.6, which would then give an _increase_ in the rate of void growth from the equation for decohesion by about a factor of 2. However, this solution is also for a non-hardening matrix so the actual effect will not be as great for a work hardening matrix.

Despite the imperfect applicability of the model to this investigation, the several factors just discussed tend to compensate one another. An important feature to note is that the void growth up to the point of instability does not diverge significantly from linearity for any of the three strain states examined.

3.5 ORDER of MAGNITUDE of EFFECT

As mentioned previously, there is no criterion to predict instability in punch stretching on the RHS of the FLD. The M-K and related analysis simply allow for a local shifting toward the achievement of plane strain, to the limit of where the Hill criterion may be applied. In the only analytical work done for punch stretching, Chu and Needleman [24,25] define instability simply by monitoring the development of a strain gradient (as observed by Ghosh and Hecker), which will occur with or without consideration of voids.

In order to establish the order of magnitude of the effect of voids (inclusions) on forming, analysis done for the LHS of the FLD will be applied to plane strain and the RHS, with consideration of the necessary assumptions.

Analysis by Stevenson and Ghosh [26] of the effect of voids on formability showed that the strain at maximum load (i.e. diffuse necking point) was decreased from "n" (the work hardening coefficient) by an amount equal to the void volume fraction. ($\varepsilon^* = n - f_v$, where ε^* is the strain at maximum load.) Clearly for a small void volume fraction, this effect is quite small. This analysis was done for no initial voids, but assuming a zero nucleation strain. Subsequent analysis by

Stevenson [27] on the effect of initial voids lead to the result that the strain at uniform elongation

was decreased from "n" by a factor of $f_v(1 + \varepsilon^*)$. These analyses lead to a decrease in formability

of, at most, 1.5 to 2 times the void volume fraction.

Two assumptions are necessary in order to apply this result to the present work. First; that

the void growth with strain is in the linear region. This was shown to be nearly true in figure 9.

Second; that the behavoir of void growth is similar beyond diffuse necking all the way up to

localized necking. This has been demonstrated in the work of Thompson and Nayak[21] where it

was shown that void growth for uniaxial tension is rather slow and nearly linear up to localized

necking, beyond which point it rapidly increases. For equibiaxial stretching, although faster than

uniaxial stretching, void growth behavior shows no rapid increase even beyond localized

necking. Plane strain must lie somewhere between the two. Thus it can be concluded that the

decrease in forming limit caused by voids is of the same order of magnitude as the void volume

fraction.

3.6 OTHER EFFECTS

Even if the complete absence, or removal of voids from the sheet were economically

feasible, it is clear that one would still be limited to forming limits on the order of 2n. This is

assuming that the sheet was perfect in every other way; surface smoothness, no "weak" grain

orientations, etc. Since the various theories proposed for the RHS require some defect, they

cannot apply to a perfect material. In such a case they would predict infinite formability. Clearly

some small perturbation will exist and instability will occur, perhaps at higher strains in some

forming operations (such as punch stretching) than others, due to geometric stabilities.

Therefore if one wishes to improve formability in punch stretching, material inhomogenieties

which increase geometric softening are not the direction to look. The method, if it exists, must lie

in somehow manipulating the inherent strain hardening properties of the material, i.e. manipulate

"n" (and "m").

4   SUMMARY and CONCLUSION

The volume fraction of voids measured at the instability strain in punch stretching was found to be between 1 and 2 x 10 $^{-3}$ for all strain states examined.

The distribution of individual void area sizes showed that they were small, with the most frequently observed size around 0.5 to 1.5 square microns. The median and mean values were around 3 and 8 square microns respectively.

The growth of voids was found to correspond to the growth equation used by Jalinier and Schmitt for the three strain ratios examined at failure.

The decrease in forming limit caused by these voids can be assumed to be of the same order of magnitude as their volume fraction.

Therefore, the following conclusion may be made: In punch stretching, at this inclusion level, inclusions cannot be considered as a major limiting factor to the forming limit, nor does further removal promise to give significant improvements. Thus if further improvements are achievable, the direction of research must lie in the manipulation of the inherent work hardening properties of the material.

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TABLE 1 : Composition of Steel Used in wt. %

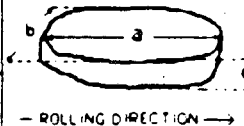| Fe	C	Mn	P	S	Al	Si	Cu	Ni	Cr	Mo	Sn
Bal	.03	.25	.01	.01	.04	.01	.03	.02	.016	.011	.007

TABLE 2 : MECHANICAL AND METALLURIGICAL PARAMETERS

YIELD STRENGTH (MN/m^2)	ULTIMATE STRENGTH (MN/m^2)	STRAIN HARDENING COEFFICIENT	NORMAL ANISOTROPY COEFFICIENT	PLANAR ANISOTROPY COEFFICIENT	GRAIN SIZE (μm)		
S_y	S_{uts}	n	$\bar{r}$	$\overset{\triangle}{r}$	$\bar{a}$	$\bar{b}$	$\bar{c}$
166	300	.22	1.5	.35	41.7	37.2	11.0

Table 3 : Inclusion Parameters

MnS	Al_2O_3	SiO_2	G.B. Fe_3C
1 - 4 μm x 50 μm	< 1 - 5 μm	< 1 - 5 μm "slag" ~100's μm	10 - 50 μm
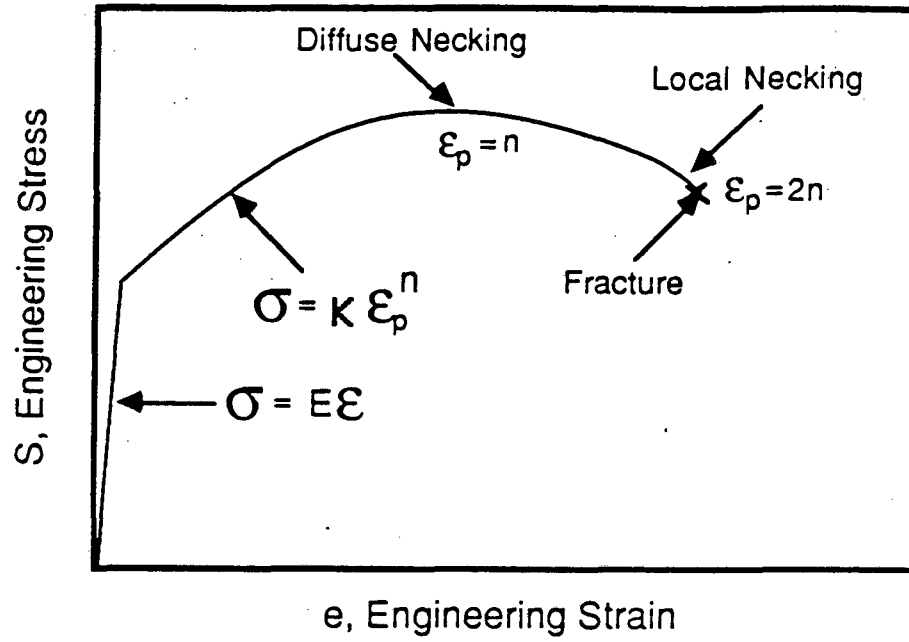
Figure 1. Schematic Engineering Stress-Engineering Strain curve for typical low-carbon steel sheet deformed in uniaxial tension. Theoretical points of diffuse and localized necking are indicated. (Subscript "p" indicates plastic strain.)
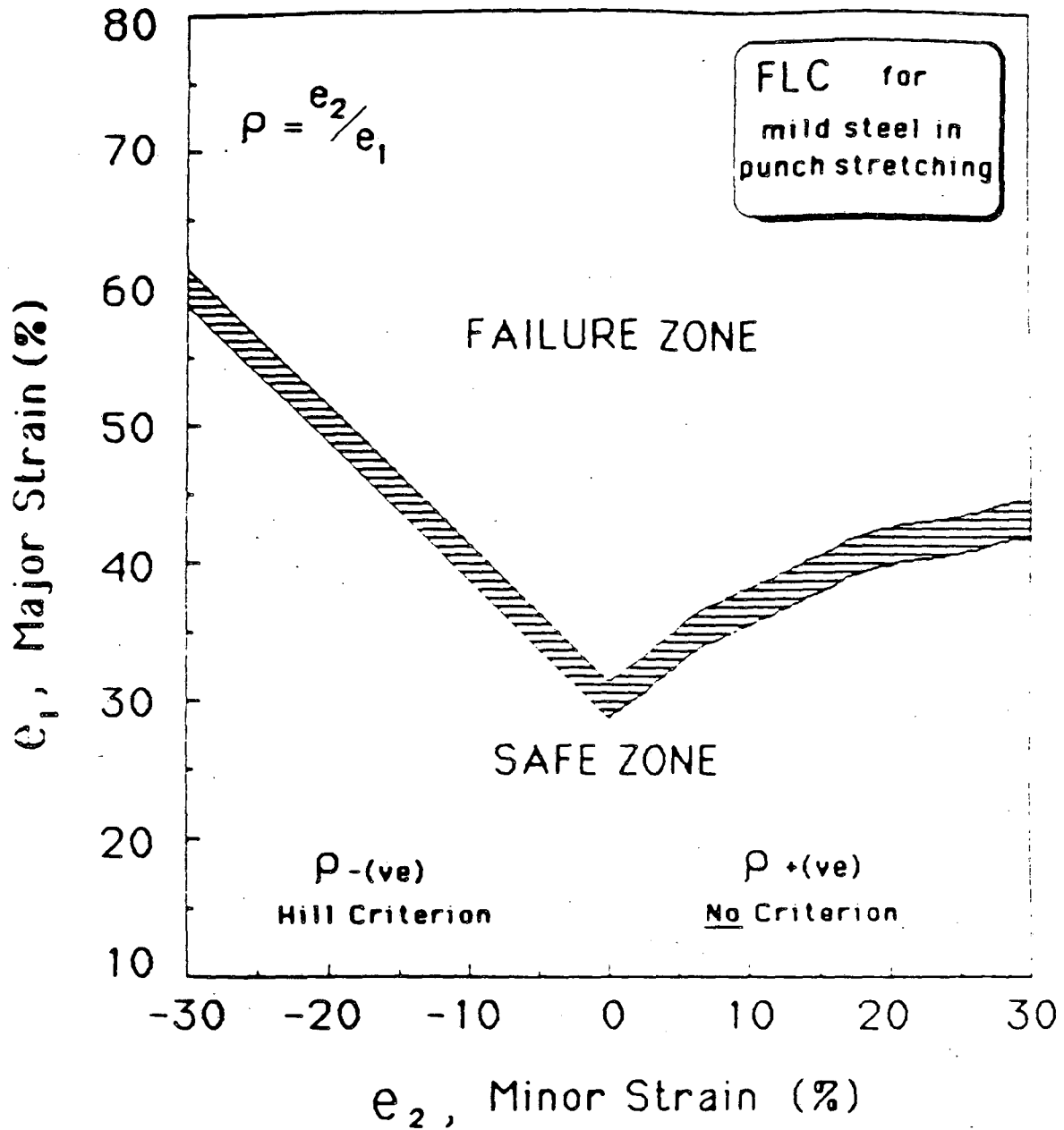
Figure 2. Typical Forming Limit Curve (FLC) for low-carbon steel deformed by punch-stretching.
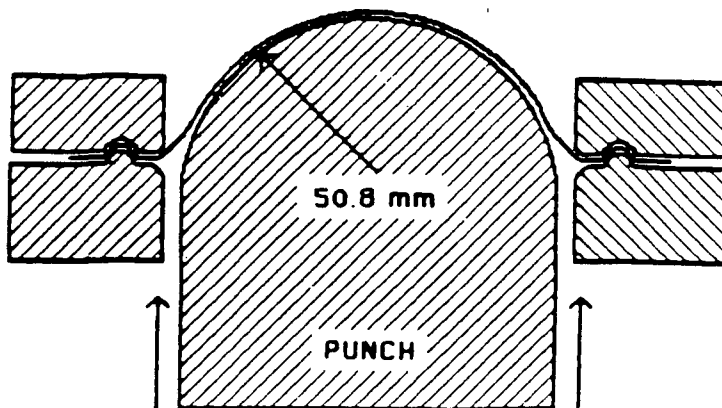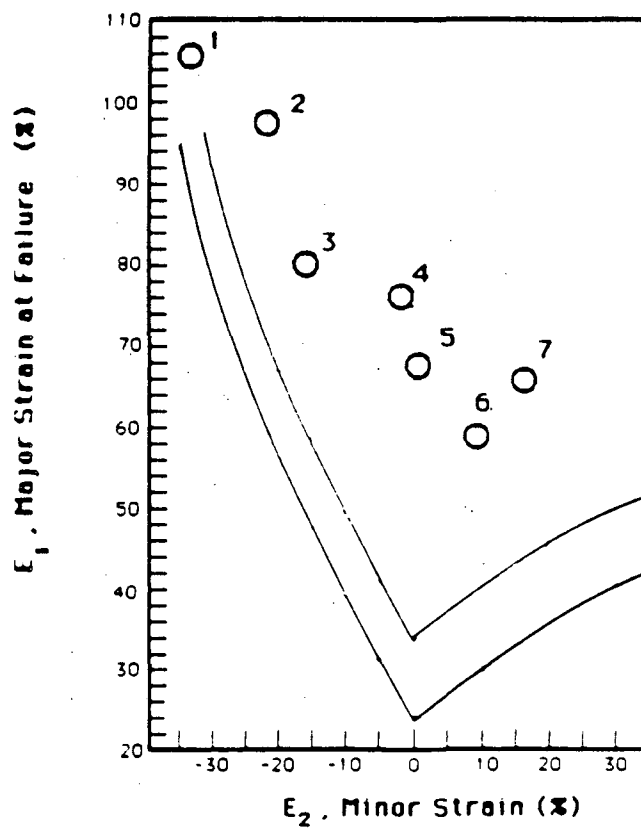
Figure 3. a) Schematic diagram of punch geometry. b) Set of domes after testing. c) Strain states at failure (numbers indicate width of dome in inches) plotted on FLC for this material.
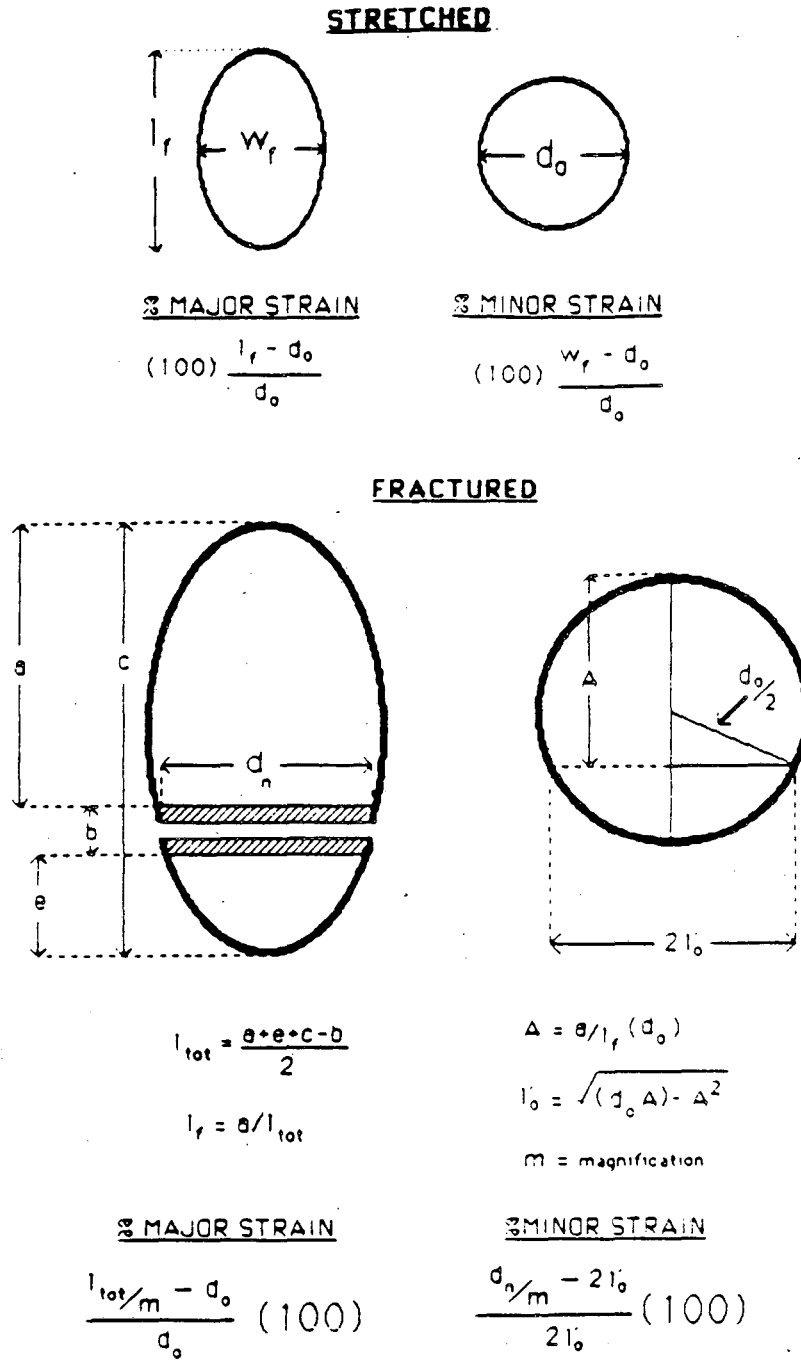
Figure 4. Illustration of surface strain measurement using circle grid technique. Upper method is used for uniform strain, lower method accounts for neck and/or tear.
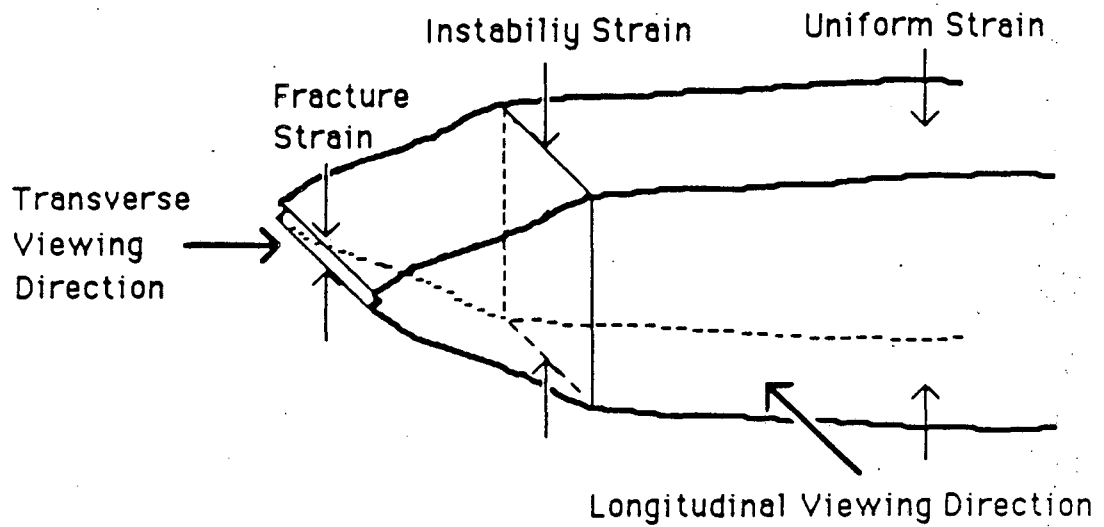
Figure 5. Schematic view of one half of necked section taken from punch stretched sheet illustrating strain location definitions and observation directions.

CIRCLE-GRID
PATTERN ON
BLANK

PUNCH 7
DOMES →

POLE
RIM

NECK AND TEAR

PHOTOGRAPGH
FAILURE REGIONS

MEASURE SURFACE
STRAINS AT
FAILURE REGION
USING DIGITIZER

SELECT EXACT REGION
OF KNOWN STRAIN

Pole Side

REMOVE REGION
OF KNOWN STRAIN →

← SEPERATE ALONG
NECK CENTERLINE

Rim Side

STANDARD "EDGE ON"
← TRANSVERSE OR LONGITUDINAL
METALLOGRAPHIC MOUNTING

SEM
METALLOGRAPHY

Figure 6a) Specimen preparation sequence beginning with as rolled sheet, progressing through
SEM metallography to obtain negatives used in data collection.

DIGITIZER

DATA
REDUCTION
PROGRAMS

DATA
COLLECTION
PROGRAMS

COMPUTER
INPUT

COMPUTER

X-Y
PLOTTER
INPUT
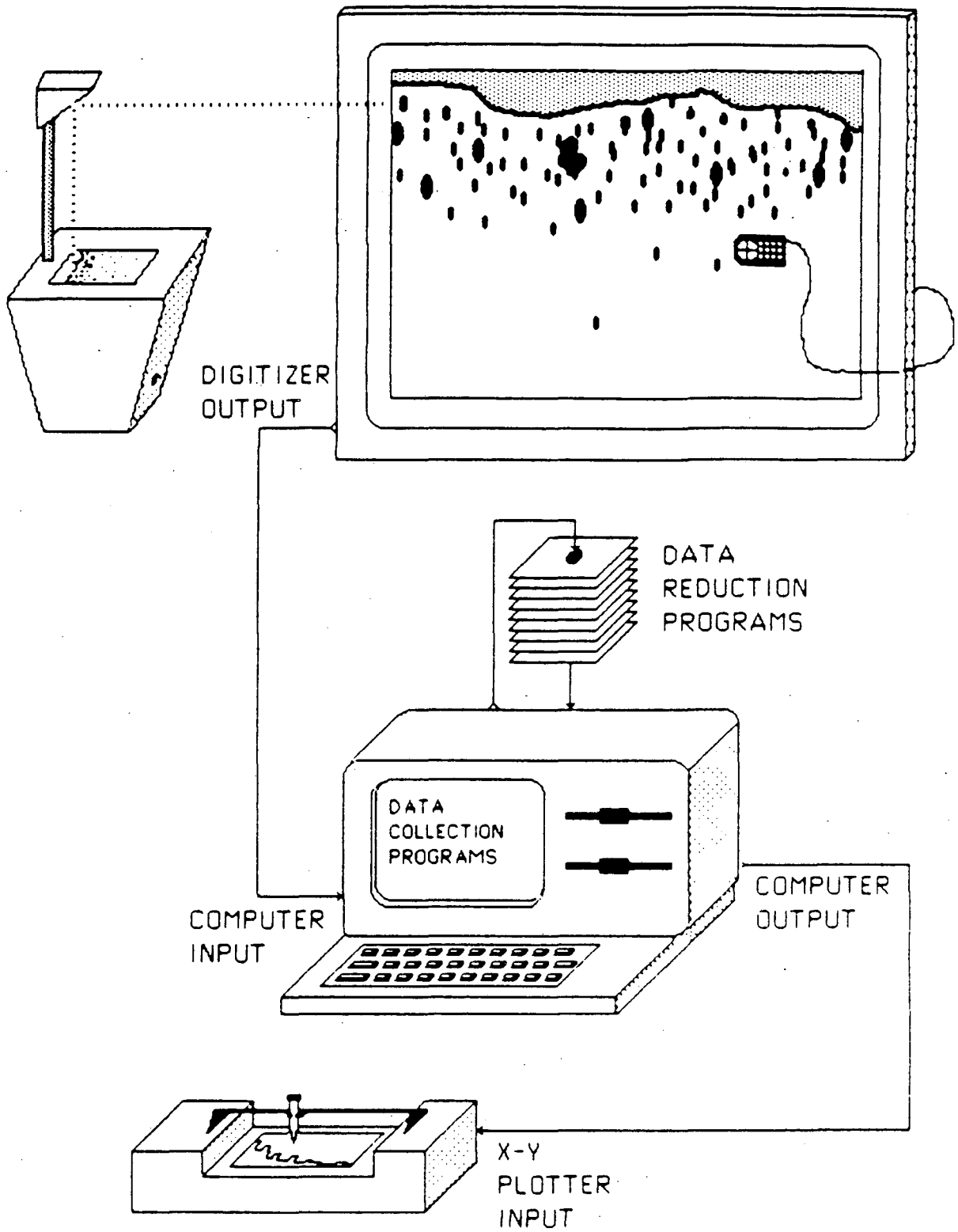
Figure 6b) Schematic illustration of data collection and reduction method beginning with
negatives obtained from SEM metallography.
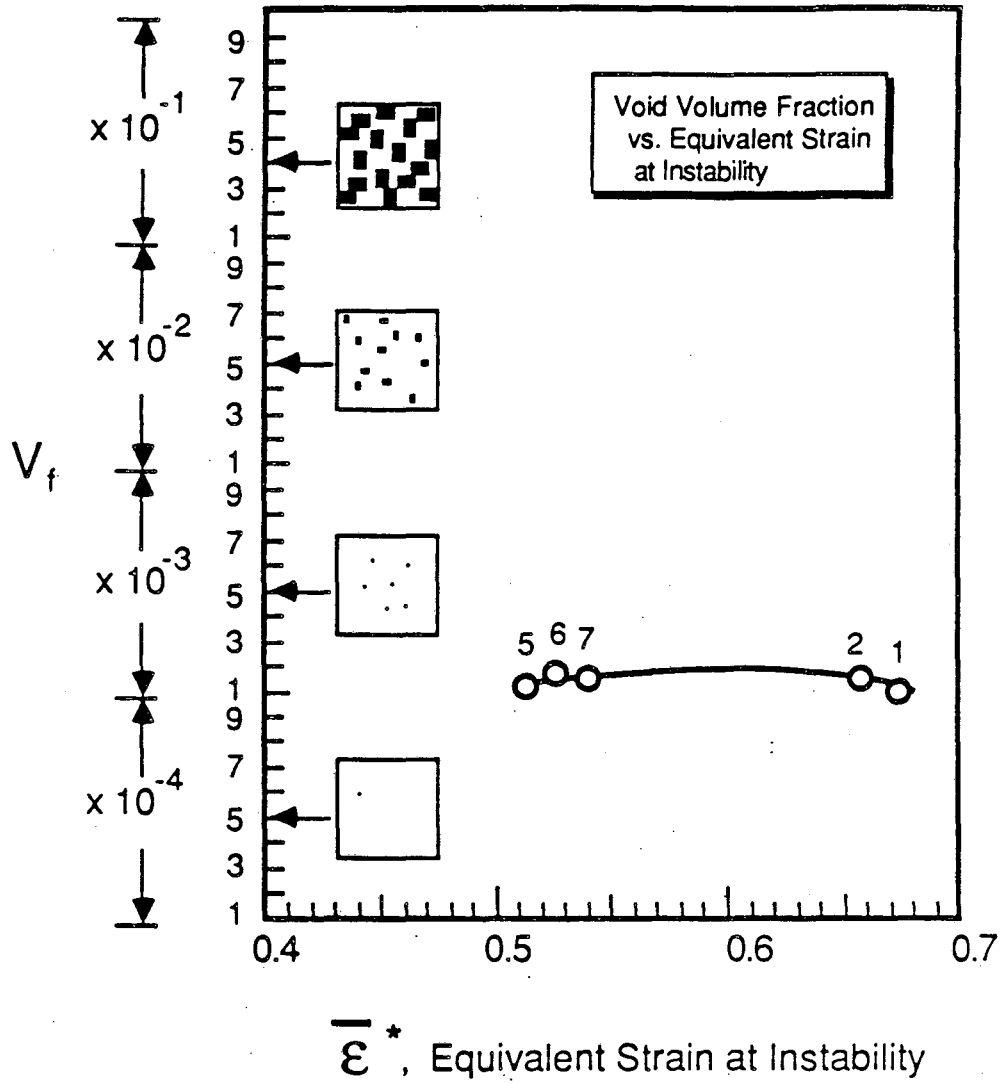
Figure 7. Void volume fraction as a function of equivalent strain. 1, 5, and 7 are taken from transverse sections at instability strain, 2 and 6 are from successive longitudinal sections at neck centerline. Example void fractions are shown at left.
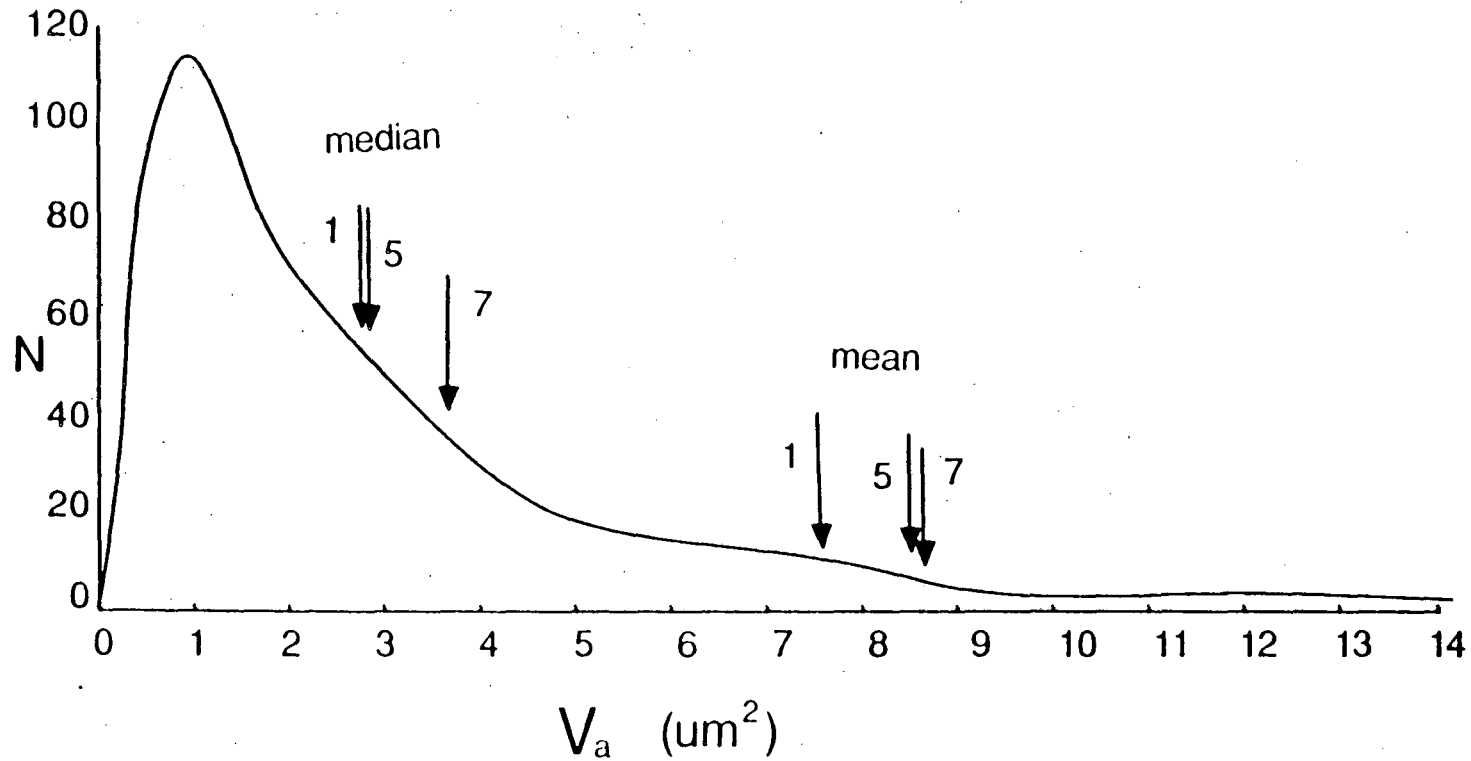
Figure 8. Size distribution of transverse void areas.

Growth of Voids

$$\frac{C_v}{C_{v_o}} = \exp[\, 1.92\,(1+\rho)\,\varepsilon_1\,]$$
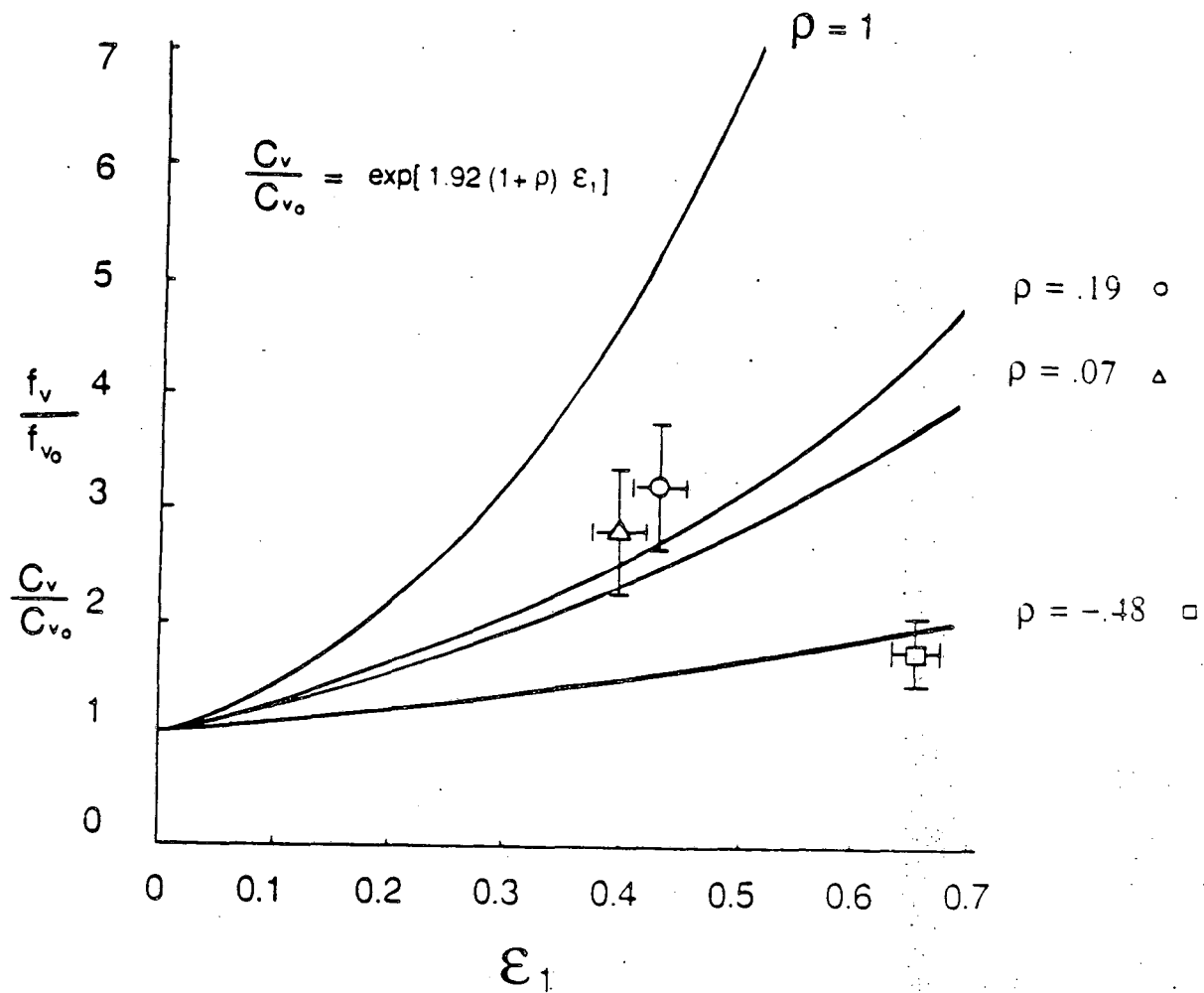
Figure 9. Void growth equation, based on ref. 8, plotted for strain ratios in this work. Upper curve is growth equation plotted for $\rho = 1$ for comparison. Data points indicate measured values of f_v/f_{vo}.