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DUCTILE FRACTURE OF U-NOTCHED BEND SPECIMENS IN SPHEROIDIZED AISI 1075 STEEL, AISI 4340 STEEL AND NiCRO.

THE OHIO STATE UNIVERSITY, PH.D., 1978
DUCTILE FRACTURE OF U-NOTCHED BEND SPECIMENS IN SPHEROIDIZED AISI 1095 STEEL, AISI 4340 STEEL AND DS NICKEL

Dissertation

Presented in Partial Fulfillment of the Requirement for the Degree Doctor of Philosophy in the Graduate School of the Ohio State University

By

Tsvi Goldenberg, B.Sc, M.Sc, Met. E.

The Ohio State University
1978

Reading Committee

J.P. Hirth
P.G. Shewmon
J.A. Begley
F.H. Beck

Approved

Department of Metallurgical Engineering
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July 24, 1949

- Terguniamtz, Rumenia.

1974

- B.Sc., Technion, Israel Institute of Technology, Haifa, Israel.

1977

- M.Sc., Ohio State University, Columbus, Ohio.

1975-1978

- Research Associate, Metallurgical Engineering, Ohio State University, Columbus, Ohio.

Publications


Presentation

Abstract of the first publication was presented at the 1977 ASM annual meeting in Chicago.

Fields of Study

Studies in Physical Metallurgy
Professors John P. Hirth, Glyn Meyrick, Gordon W. Powell and P.G. Shewmon.
Studies in Mechanical Metallurgy
Professors J.W. Spretnak and J.A. Begley

Studies in Corrosion and Oxidation
Professors Mars G. Fontana and Robert A. Rapp

Studies in Thermodynamics
Professor George R. St. Pierre

Studies in Metals Physics
Professors Rudolph Speiser and David A. Rigney
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I. INTRODUCTION

It is now well established\textsuperscript{1-5} that the ductile fracture mode which terminates deformation prior to complete necking down to a point involves internal hole growth. In materials of interest here, the holes nucleate at inclusions or second phase particles, although a few instances have been reported of nucleation in the absence of particles in high purity materials.\textsuperscript{6} The holes have been observed to nucleate by decohesion of the particle-matrix interface\textsuperscript{3-5,7,8} or by cracking of the particles\textsuperscript{2,7,9,10}. The voids then grow and coalesce to form a propagating crack.

For both types of void nucleation, cracks occur preferentially at particles which are large,\textsuperscript{7-12} have large aspect ratios\textsuperscript{7,9,10} and have low cohesive strengths of the particle-matrix interface.\textsuperscript{9,10,13} The particle size effect for brittle particles which crack to nucleate a void is associated with the statistically greater likelihood of finding flaws in larger particles.\textsuperscript{7,9,10,14} The particle size effect for particles which decohere has been found to be associated with: (a) internal localized slip with stress concentration at the interface caused by pileups of dislocations, for deformable inclusions such as manganese sulfide,\textsuperscript{7}
and (b) the interaction between the plastic fields of neighboring inclusions to provide a stress concentration at the interface of nondeforming particles. 

In this study a detailed investigation was carried out to gain a better understanding on the behavior of the microstructure of three metals during deformation prior to ductile fracture. Special attention was focused on the role of void initiation and growth in the plastic zone of notched bend specimens. The materials chosen for the study was spheriodized AISI 1095 and AISI 4340 which was quenched and tempered. The third metal employed was a dispersion strengthened nickel-base alloy, DS Ni. The specimens used were U-notched bar type and testing was done in three point bendings. The localized stress-raiser effect of the U-notch added a parameter to the straining procedures. The other parameter was the change in the load during the three point bending test. The incorporation of various notch sizes with different loads allows one to relate changes in the microstructure to parameters such as local stress, nominal stress, strain and stress gradient.

Objectives of the present work include:
1) Examination of the sites most susceptible for void formation in AISI 1095, AISI 4340 and DS Ni.
2) Determination of the relations among the void initiation and growth stages and strain and stress.
3) Study of the role of plastic instability produced by a notch in the microstructural response to deformation.

4) Determination of possible mechanisms of crack formation in the three different alloys.

5) Provision of base-line data for a similar study on specimens charged with hydrogen.
II. LITERATURE REVIEW

The Role of the Second Phase Particles in Ductile Particles

Carbon Steel

It is well known that the second phase particles play a crucial role in ductile fracture. The inclusion and second phase particles are the main source of hole nucleation and growth which leads to the dimple type of fracture. Most of the investigators in presenting their observations have shown relations between plastic strain and fracture, volume fraction of the particles and void nucleation. Hole growth was treated from the point of view of plasticity. Palmer and Smith's \(^{(12)}\) work on hole formation at SiO\(_2\) particles in a copper alloy showed that although the small particles were the predominant type of inclusion, the voids at the second phase formed near the larger particles.

Argon et al\(^{(9,10)}\) treated the void formation problem quantitatively. They investigated the behavior of equiaxed second phase particles in a plastic deformation zone formed by localized plastic flow and with a hydrostatic tension field (negative pressure). The experiments were carried out on spheroidized AISI 1045 steel, a Cu-0.6 Cr alloy and maraging steel, containing Fe\(_2\)C, CuCr and TiC, respectively.
They calculated the interfacial strength by evaluating the stress needed to decohere the matrix from the second phase particles. Their final results showed that void formation at second phase particles or inclusions took place when localized plastic flow of the matrix around the particle produced local stresses which combined the elastic with the hydrostatic tension stress component to reach the interfacial tensile strength. They asserted that the critical stress for decohesion of Fe$_3$C, Cu-Cr and TiC in their respective matrices were 1694 MPa, 1010 MPa and 1850 MPa. In the AISI 1045 steel, the average diameter of the separated particles was considerably larger than the average diameter of the whole population. The rationale for the enhancement in interfacial failure is that the high volume fraction of Fe$_3$C causes a large interaction stress between those pairs of particles separated by distances of 1-2 particle diameters or less.

The local interfacial stress for separation of inclusion from the matrix was found from plotting the density of separated inclusions lying along the axes of the fractured bar vs the distance, $2p$, away from the fracture surface, Fig. 5, ref. 1.

The Fe$_3$C in AISI 1045 gave a concentrated 2nd phase of 0.005 volume fraction. The mean diameter of the

*Calculated from the carbon content. The value measured by quantitative metallography was 0.125 but this is thought to be unrealistically large because of distortion in the SEM caused by magnetic surface effects.
Figure 1 - Cumulative density distribution of diam of Fe₃C particles: upper curve, prior to straining; lower curve, separated particles underneath fracture surface. (10)
particles were found to be 0.443μm (Fig. 1, ref. 10). Argon et al. showed that the spacing of the fracture surface dimples matched the spacing of the particles, implying that, the main source of the ductile fracture voids was the Fe₃C particles. Detailed calculations were carried out for the local plastic, strain and triaxial stress. They were compared to the average equivalent strain to fracture and leading to the conclusion that a substantial portion of the plastic flow occurred during the generation and growth of holes.

Argon et al. (10) computed the distribution of plastic strain and negative pressure using a finite element technique for inhomogeneously deformed bars, both for initially smooth specimens and for those having an initial notch. Among the conclusions that they drew in their paper was that "critical local elastic energy conditions are found to be necessary but not sufficient for cavity formation." Ashby's (29) model was formulated for local plastic flow being produced during deformation near incompatible inclusions and second phase particles. The local flow field is determined from a necessary geometric dislocation model for relaxing local deformation around the particles in the matrix: It is not a general flow criterion. Argon et al. worked with this model to find the critical stress for cavity formation. The size of the
particles in their study was very small (≤ 100 Å) and the bulk lacked dislocations. It is possible that the elastic stress is responsible for opening voids around the particles. This would occur if the theoretical shear stress were reached at the interface particle/matrix, without plastic relaxation. However, they found there was insufficient elastic energy stored in the surroundings of the small inclusion and cavities so the increase in incompatibility stresses produced by local plastic flow near larger particles is an important requisite for hole formation. At low volume fraction of particles, where particles are far away from each other, the stress at the interface has no contribution from interparticle interactions but is composed of stresses arising from localized plastic flow and local triaxial stress (hydrostatic tensile stress). When the volume fraction of second-phase particles is high and the particles are uniformly distributed and close to each other, interparticle interaction stresses of both elastic and plastic origin arise. Thus, the total stress exerted on the interface becomes dependent also on the volume fraction and particle size, as they affect the variation in average size of the second phase particles.

Their calculation based on continuum plasticity for a pure shear mode of deformation indicates a mild stress concentration of the order of √3, which in turn indicates
that the interfacial stress is of the order of the local plastic flow stress in tension. They present the following equation

\[ \sigma_{rr} = \sigma_T + \gamma(\epsilon^b) \]

where \( \sigma_T \) - negative pressure
\( \gamma(\epsilon^b) \) - flow stress in tension

The hydrostatic tension field must be added to the local plastic flow if the distant deformation field is one of impure shear. The stress on the interface particle/matrix will increase with strain hardening and with triaxiality. Both are known to reduce ductility and probably to promote cavity formation.

**Alloy Steel**

Cox and Low\(^{(7)}\) studied the mechanism of ductile fracture (dimpled rupture) in high purity and commercial 18Ni-200 grade marging steel, quenched and tempered AISI 4340. The emphasis in their work was plastic fracture occurring by void nucleation and growth. The fracture of AISI 4340 steel at a yield strength of 1400 MPa was found to occur by nucleation and growth of voids at the interface between MnS inclusions and the matrix. The growth of these cavities was interrupted and coalescence did not occur.
Instead a void sheet lying along a characteristic slip line trace formed and connected the large cavities produced at the sulfides. The small voids in these sheets nucleated at cementite precipitates formed by quenching and tempering the steel.

Another aspect of the investigation dealt with size of the non-metallic inclusions. They demonstrated that voids initiated at larger inclusions and that growth propagated more rapidly from the larger particles. By comparing notched and smooth cylindrical tensile specimens. They concluded that an increase in the multiaxial stress does not affect the void nucleation process for these alloys. However, they found that the triaxial tension greatly increases the rate of void growth. No void initiation at the sulfide-matrix interface was observed unless the matrix had flowed plastically. Van-Vlack et al. were first to report that MnS shear occurs by planar slip which can be seen as straight slip steps on an oriented free surface. This observation might rationalize the effect of inclusion size on the nucleation of large holes.

The rationale for this observation is that as material is strained, more plastic deformation occurs in the MnS than in the matrix. Planar slip provides rather long narrow slip bands within the inclusions which are blocked at the inclusion-matrix interface.
The average length of slip band is obviously related to the size of the manganese sulfide. Thus, in a dislocation pileup model, the local stress increases with size of particles and the boundaries at larger inclusions fail first. Cox and Low's\(^7\) experiments show that the strain for fracture is much lower for the notched tensile specimens than for the smooth specimens. This is due to higher stress in the notched bar for a given strain, Fig. 2, ref. 7. The second stage of the ductile fracture is void growth by some kind of mechanism involving local plastic flow of the matrix. A close look of the microstructure of a necked region shows that the number and size of voids increased with an increase in strain. This is an indication that smaller diameter inclusions are responsible for forming new voids whereas, at the same time, larger voids exist as a result of growth of voids formed earlier at larger particles upon continuous deformation of the metal. Cox and Low\(^7\) reported that the voids in the center of their tensile specimen show growth in the direction of the applied tensile force. In a calculation of the average size of voids it can be seen that the fastest growth of cross sectional area of voids (for the nucleation, growth and coalescence stage) is in the notched AISI 4340 steel specimens. They demonstrated explicitly that an increase in the triaxial stress produced by the notch increases the rate
Figure 2 - Pct of inclusions with voids as a function of true strain for AISI 4340 steel ("F" indicates value taken from only one-half a fractured specimen). 7
• Smooth specimens, △△ Notched specimens, Solid symbols represent commercial purity, Open symbols represent high purity.
of void growth. This is the conclusion made after making a comparison of voids in a notched and smooth specimen at the same strain level. The metallographic work on the different alloys led to the conclusion that voids nucleate at the MnS inclusions and grow to some critical size until the spacing of voids is suitable for coalescence. The deformation seems to concentrate on bands between the large voids. The bands indicate a localized plastic flow which results in the formation of small voids near the cementite particles formed in tempering. Subsequently, a crack connects the large holes.

Mechanisms in Ductile Fracture

Rosenfield\(^{(42)}\) reviewed the status of ductile fracture theories. The survey concerned several aspects of the problem. To simplify the matter, his review emphasizes that when the temperature is sufficiently low, materials must be homogeneous and tough to prevent brittle behavior caused by grain boundary cracking. The variables affecting the ductile fracture in this case are second-phase particles, hydrostatic stress, temperature and several second order effects.

The primary factor that influences ductile fracture is the presence of second-phase particles like inclusions, precipitates and dispersed particles. The important points to
be mentioned are: 1) the nature of the second phase may range from hard and brittle refractory compounds such as \( \text{Al}_2\text{O}_3, \text{SiO}_2 \) to extremely soft ones, the limit being a pre-existing hole; 2) as volume fraction goes up ductility goes down; 3) the adherence between particles and matrix is crucial.

According to Rosenfield\(^{(42)}\) there is enough evidence in the literature to delineate the particle size-hole formation relation and the role of particle composition. On the other hand particle shape has great influence on the anisotropy of mechanical properties. It is known that toughness and ductility are directionally sensitive. The transverse direction is the weak direction whereas usually the longitudinal direction is the working direction and has superior properties.

The effect of hydrostatic tension on necking in a tension test was investigated\(^{(33)}\) after straining a tensile specimen until necking occurred and the specimen was unloaded. The neck was removed by machining. This smooth specimen showed an increase in reduction of the area. In the previous work of Rosenfield\(^{(42)}\) et. al., the effect of notch acuity on ductile fracture was evident. Table I ref. 42 elucidates that an increase of notch sharpness reduces the ductility drastically. Thus, the hydrostatic tension reduces the ductility in engineering alloys.
Another hypothesis is that local stress concentration at inhomogeneities like notches and crack-tips can be released either by competing process dislocation motion or crack propagation. Thus higher strength metals, where dislocations are relatively immobile are less ductile.

Rosenfield\(^{[42]}\) points out that two important parameters emerged from metallographic studies: 1) nucleation and growth of holes around second phase particles; 2) localized flow by formation of shear bands. Each feature may be independent of the other or may intensify one another. The holes may concentrate on the strain in a band or the shear strain might assist N&G of voids around inclusion. Based on Ashby's\(^{[29]}\) model, Rosenfield\(^{[42]}\) elaborates on a possible mechanism that causes fracture. "At onset of yielding slip lines will begin to extend from particle to particle. The particle will block the path of the slip lines, resulting in large stress concentrations. In turn these stresses can partially be relieved by localized hole formation. However, because the metal retains a high degree of plasticity, the hole does not immediately grow into a crack but is blunted by local plastic flow."

The main ideas is that the slip introduced by plastic strain produces some stress intensification at the end of the slip band that may form a hole at a blocking particle. Upon increasing the strain and the time, the holes grow...
and coalesce.

While discussing the origin of holes he claims that most of the alloys that exhibit non-brittle fracture are defect free, based on examination at magnifications up to X 100,000. Hence, the ideas of forming a fracture by the opening of pre-existing cracks is ruled out. Another possibility is build-up of elastic stress at particle/matrix interface sufficient to form decohesion. A variety of researchers worked on this topic and provided observations and calculations of elastic separation. Although this evidence shows that in some alloys elastic stress is responsible for void formation, plastic flow appears to be requisite for hole formation in most materials. Assuming that plastic flow is essential in ductile fracture mechanisms, Ashby's\(^\text{29}\) dislocation model seems the most applicable for plastic rupture. Ashby\(^\text{29}\) showed that the dislocation loop formed to relieve the stress at the interface
\[ \sigma_T = \alpha \frac{r d}{\varepsilon} \]

where
- \( \varepsilon \) - strain
- \( d \) - particle size
- \( \varepsilon \) - distance between loop and particle
- \( \alpha \) - constant

Rosenfield claims that this model is the only one to provide a good explanation and account for several features.

1) Triaxial compression allows the particle to flow plastically and form holes; here the elastic triaxial stress would tend to hinder hole formation.

2) Hydrostatic pressure can reduce the flow of the matrix near the particles and reduce hole formation as observed.

3) When a tensile stress is applied, "... slip induced tension at particle/matrix interface augments the elastically induced tension." The formulation for stress build-up at the particle/matrix interface is

\[ \sigma_I = \sigma_T + g \cdot \sigma + \sigma_H \]

where
- \( \sigma_I \) - is the stress at the interface
- \( g \cdot \sigma \) - elastic stress concentration
- \( \sigma_H \) - Hydrostatic tension
- \( \sigma_T \) - tensile stress caused by dis loop
When $\sigma_1$ reaches a critical stress, $\sigma_f$, a hole is formed. In recent years many investigations have studied hole growth in ductile fracture and produced much evidence, but so far there is not a reasonable model to elucidate the hole growth process. The most detailed description was given by McClintock (5) based on continuum mechanics. The importance of his model arises when he takes into account the triaxiality of stress. Therefore the role of necking, notch geometry and crack tip acuity can be understood better. He based his analyses on an experimental model of plasticene containing polystyrene spheres. The he applied all his computation to the metals. The explanation for this is based on the analogies in the observations of hole growth in plasticene and the metals. McClintock considers only the void growth case where voids form at zero strain. This is in contrast to experimental results (12, 14), for example Gurland and Palmer's detailed work on the Cu-SiO$_2$ system. On the other hand his analysis takes into consideration the triaxiality at high strains where the hole is formed already.
The only model to apply the dislocation theory to hole growth is that of McLean.\(^{(43)}\) His model is a qualitative one and considers the fact that an edge dislocation can be looked at as a miniature crack. Motion of the dislocation toward an existing hole represents a small crack joining a larger one; see Fig. 12, ref. 43.

Hole coalescence is a very crucial stage and the least understood. This is an extremely fast process. Once holes grow enough to impinge on one another, there is a catastrophic propagation of holes to form a crack. McClintock\(^{(33)}\) suggested that once the hole size reaches the hole spacing, the catastrophic failure will take place. Rosenfield and Hahn\(^{(44)}\) modified McClintock\(^{(33)}\) calculation to

\[
\frac{K}{\sigma^*} = \frac{n}{D^{1/2}}
\]

where

- \(K\) - material property
- \(\sigma^*\) - uniaxial critical stress
- \(D\) - critical hole size
Figure 3 - Schematic description of hole growth based on McLean's dislocation-absorption model (43)
Figure 4 - The maximum principal strain below the notch at various loads.
Figure 5 - Contours of equal stress and strain at $L/L_GY=0.673$.\(^{(23)}\)
Elastic-Plastic Stress Analyses in:

Notched bending bar in plane strain

Griffiths and Owen (23) carried out an elastic-plastic stress analysis on a notched bar using a finite element method. They treated a linear work hardening material in their calculation. It was necessary for them to use the analysis to determine the stress and strain distributions with sufficient accuracy since model with work hardening included was so complex, particularly when a large stress and strain gradient existed.

The results obtained from the computer on the strain distribution show the strain to be highest at the notch root and to fall steeply with distance into the specimen. (Figure 4, ref. 23)

The stress distribution picture indicates that the highest stress exists below the notch, Fig. 5, ref. 23.
The variation of the maximum principal stress, $\sigma_1$, is plotted as a function of distance below the notch in Fig. 6, ref. 25. It can be seen that the maximum stress does not occur at the free surface as in the case of the total strain distribution, but rather at some distance from the free surface. This phenomenon is attributed to a build-up of hydrostatic tension components away from the notch surface.

Their results are somewhat different from those predicted by the (linear) characteristic slip-line field theory. According to their findings at the yield point the stress intensification, $\sigma_1 \max / \sigma_y$ is higher than that by the slip-line field theory. (Figure 7, ref. 25).

**Notched tensile bar**

A more accurate approach was taken by Wilkins. They employed computer programs employing a finite element method, including second order effects, and equations of continuum mechanics. The program can incorporate nonlinear models of material behavior to calculate strain fields ranging from those characteristic of linear elasticity to those produced in large scale plastic deformation. The plasticity for nucleations follow the Von-Mises hypothesis and the flow law associated with it.
Figure 6 - The variation of the maximum principal stress below the notch root at various loads. (25) -- slip line field solution, --- stresses in plastic zone, --- stresses in elastic zone. Labelled with values of $\frac{\sigma_{nom}}{\sigma_Y}$ and $(L/L_{GY})$. 

Plastic stress analysis for notched bar in bending

PRINCIPAL STRESS/UNIAXIAL YIELD STRESS

DISTANCE BELOW NOTCH (X ROOT RADIUS)

2.292
(1.065)

2.051
(0.953)

0.274
(0.336)

0.965
(0.448)

1.448
(0.673)

(1.065)
Figure 7 - The variation of stress intensification with applied load. Fine Mesh, Coarse Mesh, Slip Line Field Prediction.
The program uses a power law to calculate the flow stress as a function of the equivalent plastic strain.

Tension tests were simulated by the computer for different geometries of 6061-T6 Aluminum specimens. The calculated stress and strain state in two or three dimensions for a tensile V-notch specimen with 1.01mm root radius can serve as a parallel to the sharpest notch used in this research see e.g. Fig. 8, ref. 24.

A study done by Mohamed and Tetelman showed that the central region of a Charpy bar in a three point bend test remains in a plane-strain condition well beyond general yield. Using this concept they determined the stress and strain field near the root of a notch by the visioplastic technique for a fully plastic notched bar deformed in plane-strain bending. They used the slip line theory analysis to predict the local strains. The visioplasticity technique allowed a direct determination of local stress and strain beyond the general yield load by means of the measurement of stress and strain. A convenient parameter used to express the different stages of deformation was the angle of bend. The materials used in their studies were A533B, quenched and tempered AISI 4340 steel with 414 MPa and 1104 MPa respectively. The specimens were standard Charpy bars with various notch radii. Figure 9, ref. 26 shows the variation of the
Figure 8 - Notched tension test, contours of (a) equivalent plastic strain and (b) axial stress. V-notch specimen with a 1.016mm root radius.
Figure 9 - Variation of the longitudinal strain at the notch root with angle of bend for different root radii. \( \rho (\text{mm}) \); \( \phi = 0.254 \), \( \lambda = 0.127 \) and \( \lambda = 0.057 \).
maximum longitudinal strain at the notch root, $\varepsilon_{yy}$, with the plastic angle of bend, $\theta$ for different root radii.

The strain distribution in the mid-section of the Charpy bar indicates that the maximum longitudinal strain $\varepsilon_{yy}$ occurs at the notch root. Thus, Fig. 9, ref. 26 is used to evaluate the maximum longitudinal strain at fracture initiation, $\varepsilon_{i}$, from measurements of $\theta$. Fig. 11, ref. 26 shows a plot of $\varepsilon_{i}$, strain, vs $\rho$, the notch radius, for steels with various yield points. It can be seen that $\varepsilon_{i}$ is independent of $\rho$ but is a function of yield strength. To find the normal longitudinal stress distribution at the fracture initiation stage the flow properties were supplied to the computer program which developed it for the different radii, Fig. 12, ref. 26. A comparison of their results on the strain distribution to McClintock's findings may lead to the conclusion that the critical strain for fracture is reached at the notch surface even though the hydrostatic tension at this location is lower than below the notch. Fig. 13, ref. 26.
Figure 11 - Effect of root radius $\rho$ on the local strain $\varepsilon_1$ required for fracture initiation for different steels. (26)

$\Lambda - \sigma_y = 414$ MPa, $\square - \sigma_y = 718$ MPa, $\Diamond - \sigma_y = 911$ MPa and $\bigcirc - \sigma_y = 1104$ MPa.
Figure 12 - Distribution of the normal longitudinal stress component $\sigma_{yy}$ along the notch center line for different root radii at fracture initiation, for a steel having a yield strength of 104 ksi. 

$\rho$ (mm): $\sigma=0.254$, $\Delta=0.127$ and $\phi=0.051$. 
Figure 13 - Distribution of the longitudinal strain along the notch center line at fracture initiation ($\rho=0.051\text{mm}$, $\sigma_y=718\text{MPa}$). (26)
Rice and Johnson\textsuperscript{27} tried to predict fracture toughness by applying a continuum mechanics analysis to microstructural failure mechanism. The use of elastic-plastic small geometry change (sgc) methods analysis of cracks in plane strain perforce neglects the intense straining directly ahead of the sharp crack tip, as occurs for ductile fracture. R&J applied the slip-line theory for filling a plastic material for the nonhardening case to treat this large localized strain. Their prediction of fracture strength is based on a critical strain at a mean inclusion spacing distance from the tip. They concluded that triaxiality can not be maintained locally as the tip blunts. The maximum stress is limited.

They determined the stress deformation fields near the crack in an elastic-plastic material. These solutions were based on the regular small geometry change. Then the large geometry changes at the crack tip were also considered. When the large geometry change occurs in progressive blunting of the tip a different strain map occurs. A highly localized strain ahead of crack tip exists. The size is comparable to a notch with a mean spacing nearly equal to the grain size.

The blunt crack tip in which introduced strain gradient in front of the crack tip as can be seen in
Fig. 14, ref. 27. Based on the slip-line theory they treated the material as a rigid plastic material. The size scale, \( \delta_t = u_0 R / E \) where \( R \) is the plastic zone size. Following Hill's theory they determined the shape of the deformed notch tip from the velocity vector, see Fig. 6, ref. 27. An important result was the computation of \( \epsilon_x^r \) and \( \epsilon_y^r \) the true strain in \( x \) and \( y \) directions on the line ahead of the crack tip. Looking at figure 15 one can see that a large strain exists ahead of the blunt crack tip as predicted. Assuming that most structural metals will fracture at a true strain on the size scale of 0.2-1.0 the crack opening displacement will be

\[
\delta_t \approx 1.0-2.0 \times X_0
\]

where \( X_0 \) - g size, mean spacing of the larger second phase

for a wider true strain 0.1-2.0

\[
\delta_t \approx 0.8-7.2 \times X_0
\]

They checked these empirical equations by using data for a typical high-strength steel. The approach used in this calculation was based on a fracture toughness value, \( K_{IC} \), and yield stress level that are known. \( \delta_t \) is calculated as following:
Figure 14 - Modification of slip line field in near tip region due to progressive blunting of crack tip with deformation.\(^{(27)}\)
\[ \delta_t = 0.613 \frac{k^2}{E\sigma_0}, \]

\(X_0\) was measured in metallographic examination. The ratio \(X/\delta_t\) was checked against values appearing in Fig. 15, ref. 27 and found to be in close agreement.

**Dispersion Strengthened Alloy**

A different group to be employed in the research of ductile fracture were the dispersion strengthened alloys. One particular case is dispersion strengthened nickel or DS Ni. The strengthened materials due to the presence of a dispersed second phase has been the subject of considerable experimental and theoretical investigation in recent years. The Ni system contained about 2 vol. pct ThO\(_2\) dispersed in a matrix of pure Ni. The content is not more than 2 vol pct because of the great reduction in ductility that occurs with higher content of thorium.

Lutjering\(^{(34)}\) investigated the mode of deformation in TD Ni and the relation of the macroscopic mechanical properties to changes in the microstructural parameters during deformation.

Dispersion-hardened alloys containing a small volume of small (average size 250\(^\AA\)) hard particles causes a homogenous slip distribution during deformation. The microstructural effects for such alloys can only be studied
Figure 15 - True strain on line ahead of crack, as a function of distance X of a material point from tip before deformation. (27)
by a TEM investigation for ductile fracture. During plastic deformation of these alloys voids occur at the interfaces between the hard ThO$_2$ particle and the Ni matrix. These voids grow as deformation increases, then coalesce and eventually cause dimple rupture to take place.

Lutjering used tensile test pieces as a mean of deformation. At a strain of 0.07, voids were observed adjacent to some of the ThO$_2$ particles. As the plastic strain increased, these voids grew in the direction of the principle tensile force. The final fracture of these tensile specimen took place at a true strain of $\epsilon_t = 1.7$

Lutjering$^{34}$ showed that the volume of voids, $V_i$, normalized by the volume of the particle, $V_T$, is independent of the particle size and depends only on the local plastic strain, e.g. Fig. 16, 17; ref. 34. Since the void initiation stage is at a low plastic strain one concludes that the particle matrix interface has a low strength. As expected in the dimple size on the fracture surface corresponds to the ThO$_2$ particle spacing.
Figure 16 - TD $N_i$, relative volume of holes $V_L/V_T$ plotted against particle diameter, $2r$, at constant plastic strain ($\epsilon_p=0.4$) in the tensile test; $V_T$, volume of the particle with which the hole is associated; $V_L$=volume of the hole. (34)
Figure 17 - TD Ni, tensile test; relative hole volume $V_L/V_T$ against the true plastic strain $\epsilon_t$. (34)
III. EXPERIMENTAL METHODS

Materials Studied

The compositions of the alloys investigated in this study are presented in Table 1.

Table 1

<table>
<thead>
<tr>
<th>Alloy</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cu</th>
<th>Si</th>
<th>Mo</th>
<th>Cr</th>
<th>Ni</th>
<th>Fe</th>
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<tr>
<td>AISI 1095 Steel</td>
<td>0.04</td>
<td>0.74</td>
<td>0.010</td>
<td>0.014</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.3</td>
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<tr>
<td>AISI 4340 Steel</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>0.4</td>
</tr>
<tr>
<td>DS Ni</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>2.1</td>
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</table>

Specimen Preparation

Test specimens for AISI 1095 steel were machined from 38.1 x 12.7 mm flat plates which had been hot rolled at Armco Steel Corporation. The specimens were rough machined and spheroidized. The spheroidization anneal was to heat at 750°C for two hours, to cool and hold at 704°C for twenty hours and to air cool. Specimens were then machined to the final form shown in Fig. 18 by removing 0.38 mm
Figure 18 - Loading and specimen geometry for U-notched bar.
from all surfaces and machining U-notches with root radii \( \rho \), equal to 0.59, 1.19 and 1.59 mm, respectively.

AISI 4340 steel specimens were made from cold worked 19.1 mm round bar. Specimens were rough machined prior to heat treatment within 0.38 mm of final dimension. The heat treatment of the AISI 4340 steel consists of austerization in a protective atmosphere at 850°C for 1/2 hour followed by oil-quenching. Tempering was done in a salt bath at 455°C for 2 hours followed by air cooling.

DS Nickel powder was produced by Sherritt Gordon by a process proprietary to them. The powder was pressed isostatically into 45.4 kg "billets" about 145.9 mm inches in diameter and 560 mm length, at about 228 MPa. The billets were canned in mild steel and evacuated. After preheating to about 1065°C, the billets were extruded into rods at 25 times reduction. Following can removal, the rods (about 28 mm in diameter) were cold swaged to about 22.8 mm inch diameter. The swaged rods were then heat treated in hydrogen at 1093°C for 1 hour. Reswaging brought the rods to 19 mm diameter. Finally, they were reheat treated under the preceding conditions. So called TD nickel is processed similarly but given further working and heat treating operations which markedly change the mechanical properties. Hence the behavior of DS nickel should not be associated with that of TD nickel. The DS, Ni specimens were machined
to their final shape as shown in Fig. 1 from 19 mm diameter bar.

**Tensile Testing.**

All tensile testing for the three alloys was conducted at room temperature using an Instron machine with cross-head speed of 8.3 mm/s. The applied load and total deflection were recorded on an x-y recorder. The tensile specimen was an ASTM standard 6.35 mm diameter bar, Fig. 19. Tensile properties are presented in Table 2 for AISI 1095 steel, AISI 4340 steel and DS Ni. These results are calculated from Figures 20, 21 and 22 respectively.
Figure 19 - Specimen Used in this study: ASTM 6.35mm Tensile Specimen.
<table>
<thead>
<tr>
<th></th>
<th>Yield Strength, MPa</th>
<th>Ultimate tensile Strength, MPa</th>
<th>Reduction in area, %</th>
<th>Uniform elongation, %</th>
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</thead>
<tbody>
<tr>
<td>AISI 1095</td>
<td>385</td>
<td>618</td>
<td>66</td>
<td>25</td>
</tr>
<tr>
<td>AISI 4540</td>
<td>1500</td>
<td>1375</td>
<td>55</td>
<td>9</td>
</tr>
<tr>
<td>DS Ni</td>
<td>560</td>
<td>820</td>
<td>75</td>
<td>8</td>
</tr>
<tr>
<td>TD Ni(53) (0.26)</td>
<td>380</td>
<td>450</td>
<td>85</td>
<td>20</td>
</tr>
</tbody>
</table>
Figure 20 - Load-deflection curve for ASTM 0.35mm tensile specimen AISI 1095 steel.
Figure 21 - Load-deflection curve for ASTM 635mm tensile specimens AISI 4340 steel.
Figure 22 - Load-deflection curve for ASTM 6.35mm tensile specimen DS N1.
Fig. 23  Details of Slow Bend Testing: (a) Bend Test Apparatus, (b) Alignment of Specimen.
Slow Bend Testing

All notched bar testing was done at ambient temperature in 3-point bending on an Instron Testing machine using a 5000 kilogram compression load cell. A steel test anvil with hardened steel pins was fabricated to support the specimens during testing. The load was administered by means of a hardened plunger attached to the moving cross-head of the testing machine. The distance between specimen supports (pins) and the geometry of the plunger were similar to those employed in the conventional Charpy impact test. A centering block was used to insure proper alignment of the specimen with respect to the plunger and test anvil. The test configuration and centering device are shown in Figure 23.

The applied load and cross-head motion were monitored on an x-y recorder. The recorded cross-head motion comprises both elastic and plastic deflection of the bend specimens. Load-deflection curves were obtained for AISI 1095 steel, AISI 4340 steel and DS Ni for each of the three U-notch sizes. Specimens were loaded to various stages of load-deflection, unloaded, and the load and plastic angle of bend θ recorded.
was determined by clamping a ground steel bar to the tension surface of the specimen and measuring the displacement, \( \varepsilon \), of this surface with respect to the ground bar, as shown in Figure 24. This measurement was made on a traveling-stage microscope at 100X. Simple trigonometry then enables the bend angle to be determined.

Surface strain \( \varepsilon_{yy} \) at the mid-plane of the notch was determined by measuring the displacement of grooves cut parallel to the notch axis (z direction) by a Tukon diamond indenter. Strains \( \varepsilon_{xx} \) and \( \varepsilon_{yy} \) were measured, in the mid-plane, both on a specimen split, scribed, and glued together with Eastman 910 having a cohesive strength of 34.5 MN/m^2, and on the free surface of a scribed specimen.

**Electron Microscopy**

After deformation specimens were sectioned longitudinally at the notch mid-plane, polished, and etched in a mixture of nital and picral. They were then coated with a thin film of gold and examined in a Cambridge S4-10 Scanning Electron Microscope (SEM). The various phases in the microstructure were identified by means of an energy-dispersion spectrometer attached to the SEM.

In the DS Ni case the preparation after deformation was different. The strained specimen was put in a spark machine
Fig. 24 Determination of Plastic Angle of Bend.

\[ \theta = \sin^{-1} \left( \frac{8}{1083} \right) \]

Grain Steel Parallel Bar

Clamp

Specimen

(27.5 mm)

(\text{Measured})
and a cylinder was cut (extracted) below the notch in a transverse cut normal to the z axis by a tool made from brass, Figure 26. The cylinder, 3 mm in diameter, was sliced by an Isomet, low speed diamond saw into foil of 0.13 mm thickness. The foils were polished from both sides to a final thickness of 0.08 mm, then electropolished in a twin-jet electropolisher manufactured by E.A. Fischione (37), Figure 25. The electrolytic solution was 80% ethyl alcohol with 20% Perchloric acid, 60%. The optimum conditions for electropolishing the 0.08 mm disk were determined from the voltage current curve in Figure 27. The working voltage for final thinning of the cylinder was 10V which corresponded to a current of about 20 MA with a solution temperature of -15°C. Once the disk produced a hole in the center of disk, the hole edges were assumed to be thin enough for viewing in the TEM. The Transmission Electron Microscope used was a Philips EM-300 machine.
Figure 25 - Arrangement of the Jet-Electropolishing System.
Figure 26 - Location of the cylinder in the DS Ni specimen, cut for TEM examination.
Figure 27 - Current Voltage curve for calibration of the twin jet-electropolishing unit.
IV. THE NOTCH EFFECT

Elastic-Plastic Solution

When a bar (as in Fig. 28) is loaded, the external load produces a stress gradient that is directly related to the bar dimensions, Fig. 29. The load applied, \( 2F \), at the center of the bar is assumed to be uniformly distributed over a length of 1 mm (\( a/10 \)) which is equivalent to \( \sigma_p = 20F/a \). The bending shear stress distribution was determined by elementary elastic beam theory. A stress concentration is produced at the vicinity of the notch root.

The resulting stress distribution\(^{20}\) is given in Fig. 2, which shows that the longitudinal stress, \( \sigma_{yy} \), is a maximum at the notch tip

\[
\sigma_{yy} = \sigma_{\text{max}} = K_E \sigma_N(x=0)
\]

where \( K_E \) is the elastic stress concentration factor and \( \sigma_N \) is the nominal stress. As the notched specimen continues to deform, local yielding occurs and a small plastic zone appears at the notch root. Upon increasing the bending load, the plastic zone spreads through the matrix, Fig. 30. For a rigid-perfect-plastic material under plane strain conditions, Hill\(^{21}\) determined the stress distribution in the elastic-plastic region on the basis of slip line field
Figure 28 - Formulation of boundary condition.
Figure 29 - Elastic stress distribution in notched bar at the midplane for a notch similar to that in Figure 1 but with ρ=1mm.
theory, finding for the maximum principal stress

$$\sigma_{yy} = Y[1 + \ln(1 + x/p)], \quad 0 < x < R \quad (2)$$

where $Y$ is the tensile yield strength and the factor in brackets is a measure of the degree of local triaxiality of stress.

$$\sigma_{yy} = \sigma_{\text{max}} = Y[1 + \ln(1 + R/p)] = K_p Y \quad (3)$$

where $K_p$ is the plastic stress concentration factor. Hill's model was specifically elucidated for V-notch bend specimens of steels of various root radii by Wilshaw et al.\textsuperscript{22}. Griffith and Owen\textsuperscript{23} carried out an elastic-plastic, plane-strain analysis for a bent V-notch bar of 0.25mm notch radius using a finite element method, with parameters in the analysis corresponding to carbon steel and considering a linear hardening material. They obtained results differing in detail, but in rough general agreement with the results of Hill\textsuperscript{21} and of Wilshaw et al.\textsuperscript{23}, Griffith and Spretnak\textsuperscript{17,20} performed finite-difference, non-work-hardening, elastic-plastic calculations for U-notched specimens of AISI 4340 steel with a notch radius $p=1$mm and again found good agreement with the results of Hill.\textsuperscript{21} For purposes of the present analysis, the non-work-hardening results predict the stress distribution in Fig. 30, with $K_p$ increasing with $R$ to a value $K_p_{\text{max}} = 2.571$ at a critical value of $R$ given
Figure 50 - Distribution of stress $\sigma_{yy}$ in the elastic-plastic bending case. (22)
by \((R/\rho) = 5.81\). The finite element calculations of Griffith and Owen\(^\text{23}\) indicate a somewhat larger value of \(K_p \max \geq 2.62\). Some added support for the above results is provided by the power-law hardening, finite-element calculations of Wilkins\(^\text{24}\) for a plane-strain, tensile specimen of 6061T6 aluminum alloy with a V-notch of root radius 1.02 mm. In the region near the notch surface, the stress distribution was in agreement with the bend calculations and the value of \(K_p \max \) was 2.5.

The elastic stress concentration factors for the \(\rho=0.59, 1.19 \) and 1.59 mm notches, calculated at the notch root from the results of Peterson\(^\text{25}\), were \(K_p = 2.70, 2.05 \) and 1.90, respectively.

Notch-root principal true strains \(\varepsilon_{yy}\) were determined by the surface groove method as a function of plastic deflection angle \(\Theta\), as shown in Fig. 31. Finite element calculations\(^\text{24}\) and experiment\(^\text{27}\) show that the notch root strain is constant over most of the notch, so there is no problem in associating the strain with the value \(\varepsilon_{yy}(\gamma=0)\). The trend of the results was similar to that measured for V-notch bend specimens of low carbon A533B steel and AISI 4340, tempered to various yield strength levels and with various values of \(\rho\), by Mohamed and Tetelman\(^\text{26}\).
Figure 31 - Notch-root principle strain, $\varepsilon_{yy}$, as a function of deflection angle $\theta$, o for $\phi=1.39$, $\Delta$ for $\phi=1.19$, $\Box$ for $\phi=0.59$mm notch root.
Figure 32 - Strain distribution $\varepsilon_{yy}(y=0)$ for a series of deflection angles for the $\rho=1.59\text{mm}$ specimen.
$\theta - \varnothing = 25^\circ$, $\Delta - \theta = 18^\circ$, $\diamond - \theta = 13^\circ$, $\square - \theta = 9^\circ$, $\bigcirc - \theta = 7^\circ$, $\bigtriangledown - \theta = 5^\circ$
Figure 33 - Principle strain at fracture, $\varepsilon_{yy}$, as a function of $x/\rho$.
$\times o=1.59$mm, $\Box - o=1.18$mm, $\bullet$-ref 24, $o$-ref 26.
**Plastic Strain Determination**

Attempts to use the visioplastic of Mohamed and Tetelman\textsuperscript{26} to determine the strain distribution on split and glued bars, however, were unsuccessful for strains $>0.10$ because of fracture of the adhesive (the same used in ref. 26). Because several studies\textsuperscript{23,24,26} had indicated that $\varepsilon_{yy}$ should vary hyperbolically with $x$, we therefore fitted such curves to the surface strain values and to internal points where $\varepsilon_{yy} < 0.10$. The resultant curves, are shown in Fig. 32 for the case $\rho = 1.59\text{mm}$, together with the unmodified values influenced by relaxation in the $z$-direction. The form of the resulting curves is in good agreement with the finite-element calculation, as shown in Fig. 33. As also shown in Fig. 33, the results of Mohamed and Tetelman\textsuperscript{26} appear to be relatively low at the surface, possibly indicating a problem with the visioplastic method at large strains in their work as well.

Load-deflection curves were determined as shown in Fig. 34. The corresponding load-plastic deflections for AISI 1095 steel data are presented in Table 3. Samples at various stages of deformation as determined by these data were studied metallographically.
Figure 34 - Load-deflection curve: for bend specimens with different $\rho$ values.
Table 3 - The plastic deflection angle, $\Theta$, of the three notches as a function of the bending load.

<table>
<thead>
<tr>
<th>Specimen notch radius (mm)</th>
<th>6.9</th>
<th>7.8</th>
<th>8.8</th>
<th>9.8</th>
<th>10.8</th>
<th>11.8</th>
<th>12.7</th>
<th>13.7</th>
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<th>15.7</th>
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<td>0.67°</td>
<td>1.87°</td>
<td>3.13°</td>
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<td>10.50°</td>
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</tr>
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<td>9.27°</td>
<td>13.23°</td>
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<td>18.12°</td>
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V. RESULTS AND DISCUSSION OF MICROSTRUCTURAL ANALYSIS IN NOTCHED SPECIMENS, AISI 1095 STEEL.

Sites for Voids Formation

The initial microstructure was characterized by a volume fraction of cementite of 0.14, Fig. 35. The cementite particles had a mean intercept length of 0.45μm and a mean section density of particles of 6.6x10^{11} m^{-2}. The predominant inclusions were of two types as shown in Figs. 36 and 37. The first comprised stringer-type manganese sulfide particles with section shapes elongated in the rolling direction. The second group included isolated, angular-shaped silicate inclusions. For each class, carbide particles were frequently found to be in contact with the inclusions.

For each case a series of specimens was tested to various loads in 980 N (100Kg) increments, for example 11 specimens from 5,880 to 14,700 N load for the \( \rho = 1.59 \text{mm} \) case. These specimens were examined for void initiation and crack initiation. Results in terms of critical void formation, profuse void formation and crack initiation as a function of deflection, or strain converted from Fig. 31, are presented in Table 5. Details of the structures are presented first for the \( \rho = 1.59 \text{mm} \) case.
Figure 35 - Microstructural appearance of Fe₃C particles embedded in the ferrite matrix.
Table 4

Angle of plastic (deflection), maximum notch-root strain \( \varepsilon_{yy}(\pm 0.001) \) and load (kN) for initial void formation, profuse void formation and crack initiation for the various specimens.

<table>
<thead>
<tr>
<th>Void initiation</th>
<th>0.59</th>
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<td>( \Theta )</td>
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<td>load</td>
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<table>
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<td>15.3°</td>
<td>17.3°</td>
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<tr>
<td>( \varepsilon_{yy} )</td>
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<tr>
<td>load</td>
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<table>
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<td>7.7</td>
<td></td>
</tr>
</tbody>
</table>

\( \varepsilon_{yy} \) denotes the maximum principal strain at the notch-root.
In the largest notch specimens, the first void initiation occurred near the notch surface at three types of sites, Figs. 36-41. In all figures the principal stress $\sigma_{yy}$ acts left and right as viewed on the page. Fig. 39 shows nucleation at manganese sulfide particles. For these particles, preferred formation sites were the elongated surface normal to the x-direction, and, at such surfaces, positions of three phase contact among inclusion, ferrite and cementite. In agreement with the results of Argon et al.$^8$, but in contrast to the supposition of Ioue and Kinoshita$^{28}$, the predominant mode of void formation was by interface decohesion with only a rare event of cementite particle cracking. Fig. 38 illustrates void formation at a silicate inclusion. Preferred sites for the inclusion were positions of three-phase contact, as shown in Fig. 38, with the ferrite-inclusion interface first decohering. The type of site was between two closely spaced cementite particles, with the direction of closest spacing aligned with the direction of action of the maximum principal stress $\sigma_{yy}$, Figs. 40 and 41. Indeed, the disparate result on void formation by particle cracking$^{28}$, mentioned above, may be a result of misinterpretation of micrographs: many voids in the micrographs shown$^{28}$ which are interpreted to arise from particle cracking appear to represent nucleation
Figure 36 - Manganese sulfide inclusion in the steel.

Figure 37 - Silicate inclusion in the steel.
Figure 38 - Void formation at silicate inclusion. 
ρ=1.59mm case.

Figure 39 - Void formation at MnS inclusion. 
ρ=1.59mm case.
Figure 40 - Void formation between two closely spaced carbide particles. \(d=1.59\,\text{mm}\) case.

Figure 41 - Void formation between two closely spaced carbide particles. \(d=1.59\,\text{mm}\) case.
by decohesion between closely spaced carbides.

**Void Initiation and Growth**

As the strain increased, profuse void formation suddenly occurred, as illustrated in Fig. 42. Moreover, observation in the SEM showed that the increased void density was not random. Rather, sheets of voids were forming with their traces lying along characteristic slip lines, which have a logarithmic spiral shape near the notch root. Void densities in such a zone exceed those in a projection parallel to the root radius vector by a factor of 1.4, Fig. 43. Even this factor does not reveal the full effect, however, since the latter case involves the intersection of characteristic slip lines. While further work with a greater number of deflection levels studied would be required to demonstrate the details, the data of Fig. 42 suggests that the void density for the initial special-site nucleation cases increases about linearly with strain followed by an increasingly abrupt rise which can be associated with the onset of marked plastic instability. The greater scatter when the void density rises is associated with the non-random occurrence of voids along characteristic slip lines.

In contrast to the void density-distance (related to void density-strain) plots for notched specimens by others 7, 10,
Figure 42 - Density of nucleated holes as a function of deflection angle.
\( \Delta \) is for \( \rho = 1.59 \text{mm} \), \( \circ \) for \( \rho = 1.19 \text{mm} \). The density of Fe\(_3\)C particles is 6.6x10\(^{11}/\text{m}^2\). Arrows represent rise of void density to this value as the critical deflection for fracture is approached.
Figure 43 - Cumulative sum of holes per unit length (moved to dimension $x$) with increased distance $x$ from notch root. $p$=1.19mm notch, uncharged specimen, $\theta$=18.7°, load=15.7KN. Hatched distribution normal to notch root, open distribution along characteristic slip line.
the void densities did not drop smoothly with distance along a characteristic slip line but were constant up to the point where the strain decreased to 0.10, whereupon the void density abruptly dropped to zero. Again, this constancy of void density, and difference with prior work, is connected with the nonrandom association of voids with plastic instability zones.

Crack Initiation in Notched AISI 1095 Steel

Only as the strain for crack initiation was reached did extensive void growth occur. As denoted by the arrows in Fig. 42, all particles decohered in the final stage of fracture, as indicated by the correlation of dimple spacing on the fracture surface with carbide spacing, presumably in the intense elastic-plastic stress field at the advancing crack tip. In Fig. 44, both crack initiation at the notch root and growth of several voids to a size larger than the bounding carbide particles are seen. In agreement with the work of Girffis and Spretnak, crack initiation occurred at the notch surface, generally at a site somewhat removed from the notch midpoint, and propagation occurred along one (or a set of) characteristic slip line(s). This stage of fracture involves stable crack growth, with the final ductile fracture dimple spacing
Figure 44 - The Nucleation and Growth of Voids Close to the Notch Root. \( \sigma=1.59 \text{mm} \) case.
corresponding to the cementite spacing, Figs. 45 and 46. As the crack reaches the midsection of the bar, the increased stress intensity causes a transition to unstable crack propagation in a plane normal to $y$, the principal stress axis.

The characteristics of the microstructure for the two smaller notches were similar to those of the largest notch. The major difference was that the various stages of fracture occurred at successively smaller deflection angles for the smaller notches, as shown in Table 3. Another difference is illustrated in Figs. 47 and 48. While void nucleation occurred preferentially between two closely spaced carbide particles for the smallest notch, the alignment of the particles with the direction of action of the maximum principal stress $\sigma_{yy}$ was less pronounced.

Simulation of Voids Between Two Fe$_3$C Particles

In order to further understand the localized plastic flow, a simulation of the microstructure and state of stress was built. A 15x10x1.5cm box made from plexiglass was filled with a high viscosity oil which represents the ferrite matrix. Inside the box two steel balls with a diameter of 1.2cm were placed close to each other, representing two cementite particles; figures 49, 50, 51 and 52. On the left and right sides of the box, two
Figure 45 - Fracture surface of the stable crack portion of a crack lying along a characteristic slip line, close to the notch surface in the plane-strain region of the specimen.

Figure 46 - Fracture surface of the stable crack portion transformation into the unstable portion. The dimple rupture changes to a semi-cleavage type.
Figure 47 - Void formation between cementite particles. \( p=0.59 \text{mm} \) case.

Figure 48 - Void formation between cementite particles. \( p=0.59 \text{mm} \) case.
Figures 49, 50 - The simulation of local flow between two closely spaced particles.
Figures 51, 52 - The simulation of local flow between two closely spaced particles.
flexible walls made of hard rubber were installed. These moveable walls were used to apply to tensile stress on the oil and steel ball within. Two small bubbles of air were injected through one of the plyglass walls and were attached to the inner surface of the ball facing each other. Upon pulling the concave rubber walls, the bubbles in the oil attached to the steel ball moved toward the other ball. Increasing the tension on the wall continued to narrow the distance between the air bubbles until they departed from the balls; Figs. 49, 50, 51 and 52. This simple experiment shows that the matrix is capable of transferring a longitudinal stress component into a local stress between two hard particles. These results in the high viscous oil with the steel balls under tensile stress can be applied to results in the alloy itself. As in the simulation model, the principle stress produces a local interaction between the two local stress fields associated with the cementite particles. Upon increasing $\sigma_{yy}$, the plastic flow is introduced to the area. Relaxation of the matrix will take place when overlap of the large localized plastic strain from the two particles reaches a critical value. This shows that once voids are nucleated between two closely spaced particles they should grow to link the particles, producing the configuration of Fig. 11 which was the usual
one observed. Thus, the observations of voids are interpreted to mainly reflect preferential void nucleation.

Discussion

With respect to void nucleation, the present results demonstrate the role of both elastic and plastic incompatibility stresses in augmenting void formation. The three-phase junctions observed to be highly preferential sites for void nucleation are sites of divergent stress singularities in a purely elastic calculation. While the singularity is removed by plastic flow in the actual situation, the combination of the elastic-plastic stress arising from the incompatibility and either a relatively weak interface (iron with MnS in the Fe-MnS-carbide case) or an interface between two difficult-to-deform phases (carbide with silicate in the Fe-carbide-silicate case) results in preferential void nucleation. The specific elastic-plastic solution is not available for this case, so it cannot be analyzed quantitatively.

The observed second type of preferential site for void nucleation, between two closely spaced carbide particles, provides direct confirmation of the calculations and predictions of Argon et al. who demonstrated that plastic incompatibility stresses would be maximized at such sites
where plastic fields of two particles overlapped. They\(^9,10\) indicated that such an effect was the likely cause of preferentially earlier void nucleation for large particles, observed many times as discussed in the literature review.

With a large volume fraction of second phase, both factors would favor plastic zone overlap. The plastic incompatibility stress can be understood from Figs. 53 and 54 in terms of either the continuum model of Eshelby\(^50\) or the geometrically necessary dislocation model of Ashby.\(^29\) For a hypothetical process of removal of the spherical particle, shear of the remaining material and hole, restoration of the particle, and deformation of the matrix to restore compatibility, the stress field in Fig. 54 is the result. Between two particles aligned along the direction of action of the maximum principal tensile stress, the tensile incompatibility stresses of the two particles overlap and void nucleation is favored at the interface. The observation that such nucleation is preferential in the intense plastic shear zone along a plastic instability line (trace of characteristic slip line) confirms that the stresses arising from plastic incompatibility are important, as suggested by Argon et al.\(^9\)\(^-\)\(^10\) Because the shear is localized in the instability line, however, and therefore indeterminate, again we are unable to treat the
Figure 53 - Shear strain around undeformable spherical particle.

Figure 54 - Corresponding distribution of displacements to restore compatibility. T and C represent resultant regions of tension and compression.
nucleation problem quantitatively. Because of this uncertainty in local stress, the present results cannot be used to test the proposal of Argon and Im that void nucleation at the carbide-ferrite interface should occur at a critical normal interfacial stress. The results are qualitatively consistent with their suggestion, however, in that the observed sites for nucleation are the sites where the sum of the concentrated applied stress, the plastic incompatibility stress, and the elastic incompatibility stress (also demonstrated by Fig. 54) give the maximum normal stress on the interface. Also the observation that the orientation of the two particles bounding the resultant void becomes more random for the smallest notch.

The role of the triaxiality in nucleation and growth is not fully understood, but a loss in ductility is observed to exist. In a low notch acuity, the formation of voids at low incompatibility sites was observed. This is a result of the combination of elastic buildup at the particles interface, a low hydrostatic effect, plus a high local plastic strain along a slip line. The elastic stress set-up depends both on the relative rigidity of the particle and the matrix as well as the shape of the particle within the ferrite matrix. At a low notch acuity, the hydrostatic stress is low. To increase the plastic zone size, the general plastic strain (bending load) is
increased. This results in a higher capability to take general plastic deformation.

As noted by Argon et al., the above discussion applies to the class of materials where overlap of particle strain fields is important and not to the class with widely spaced particles such as those studied by Cox and Low. Hence, the results are not inconsistent with the Cox-Low observation that stress triaxiality had little effect on void nucleation at MnS particles, relatively widely spaced, in steel. The present observation of preferential void formation along characteristic slip lines is, in fact, in excellent agreement with the Cox-Low observation of final fracture by cracking along characteristic slip lines in the ligaments remaining between the large voids formed at MnS particles. In their studies, final fracture proceeded in a ductile manner with smaller voids nucleating preferentially at carbide particles along the plastic instability lines (see Fig. 16 in ref. 7).

In final preferential site for void nucleation, on the large face of elongated MnS particles, is consistent with other findings for such large, weakly cohesive inclusions. A possible factor favoring nucleation on the large face is the localized flow by plastic instability. When one of these bands impinges on such a large, large-aspect-ratio,
particle, it is analogous to a blunted dislocation pileup or terminated mode II crack, and produces stress concentration favoring void formation. The likelihood of impingement is statistically reduced for the small face of the elongated particles.

The data of Table 4 show the influence of plastic instability on ductile fracture. Profuse void formation takes place in the planes of shear instability and crack initiation occurs with a small amount of further strain. For the class of materials with widely spaced voids, as discussed previously, Mohamed and Tetelman\textsuperscript{26} found that the strain to fracture initiation depended on yield strength level but was independent of $\rho$, and that crack initiation commenced at the surface, not in the interior region of maximum triaxiality and stress. No mention was made of plastic instability in their work,\textsuperscript{26} but a comparison with the work of Cox and Low\textsuperscript{7} and Griffis and Spretnak\textsuperscript{17,20} suggests that such instability may have been present. The present results are in good agreement with Mohamed and Tetelman\textsuperscript{26} in that crack initiation takes place independent of $\rho$. Moreover, the same trend of independence of $\rho$ follows for the strain for profuse void formation and, thus, the strain for plastic instability which is the precursor of fracture. These results are in agreement
with the postulates of Spretnak, McClintock and others\textsuperscript{15,16,17} that the onset of plastic instability occurs at the free surface of a low-work hardening material and is favored by a large plastic strain gradient, conditions which are met for the present material. However, it has been suggested that an increase in the magnitude of the plastic gradient should favor plastic instability, leading to a smaller strain for its initiation.\textsuperscript{15,31} This postulate is not supported by the results in Table 4 since the plastic strain gradient increases with decreasing $\phi$ while the strain for instability does not.\textsuperscript{*} While the effect of strain gradient has been found in some cases,\textsuperscript{15} a similar independence of strain for instability on plastic strain gradient has been found by Hauser\textsuperscript{15,32} for 304 stainless steel.

The marked differences in deflection to fracture in Fig. 34 simply reflect the large differences in elastic and plastic stress concentration factors for the different sized notches. In terms of the strain data of Table 3, the differences in behavior are absent, within experimental scatter.

\textsuperscript{*} The postulate is not supported in detail for the varying strain gradients in the three sizes of U-notch: it is supported to some extent, however, in that the strains to instability and fracture for the notched specimens were less than the corresponding strains in a smooth tensile specimen.
The void densities for strain less than fracture in the present work are much less than in the work of Argon et al.\textsuperscript{10} and Cox and Low.\textsuperscript{7} (In final crack propagation all agree that all particles in the crack path produce voids as indicated by correlation of dimple size on fractographs with carbide particle spacing). The profuse void formation in Fig. 42 corresponds to a fraction of 0.13 of particles with voids at a strain within 10 percent of the fracture strain. In contrast, Argon et al.\textsuperscript{10} and Cox and Low\textsuperscript{7} find, respectively, fractions of 0.5 and 1.0 of particles with voids at fracture and Argon et al., in the class of material resembling that in the present work, find a slow decay of this fraction with decreasing strain. These differences are associated with the presence of plastic instability.

The present work, with bending and the larger plastic strain gradient of the smaller notch radii favoring plastic instability, represents void formation primarily in the sheets of intense shear and thus a more localized void formation than in the other work. In this sense, the final localization of fine void formation along characteristic slip lines in the fracture of ligaments in the Cox-Low work is in agreement with the present work.
AISI 4340 STEEL

Defect Occurrence

One of the first observations made in the microstructural examination on the as-received AISI 4340 steel revealed already separated interfaces of the newly equiaxed ellipsoidal MnS particles embedded in the ferrite matrix, Figs. 55, 56. The interfaces of the greatly elongated MnS particles seemed to have a better adherence since they did not exhibit such separation, Figs. 57 and 58. Also, the angular SiO$_2$ particles did not exhibit decohesion prior to the straining process. The occurrence of large separate voids around the almost spherical MnS in as-received material is attributed to the thermo-mechanical process in producing the 19mm round bars. Nevertheless, this early occurrence of holes near the inclusions had little effect on the crack initiation and growth as shown later. Some of MnS particles in AISI 4340 also were cracked prior to straining, again, mainly the ellipsoidal type, Fig. 59. Besides these discontinuities in the matrix there exist other types of zones susceptible to crack initiation. In these zones there is an absence of cementite particles which produces non-uniform distribution of the cementite second-phase in ferrite matrix, Figs. 60 and 61. The lack of cementite particles in these certain areas can be understood on the bases of the diffusion
Figures 55, 56 - Large separation at MnS inclusion within the ferrite matrix in the as-received AISI 4340 steel.
Figures 57, 58 - Non-separated interface at MnS inclusion in the ferrite matrix in the as-received AISI 4340 steel.
Figure 59 - A cracked MnS inclusion in the as heat-treated AISI 4340 steel.
Figure 60 - Non-uniform distribution of cementite particles in the tempered structure of AISI 4340 steel.
Figure 61 - High magnification of the circled area in Fig. 60.
mechanism in the austenizing stage during heat treatment. The as-received structure consists of pearlite within a ferrite matrix. The mean free jump distance of a carbon atom in the austenite phase is calculated to be about 1 nm \( (\bar{x} = \sqrt{2}Dt) \). Hence, ferrite zones which are larger than 1 nm are not homogenized during the heat-treatment. Upon quenching the austenitized area is transformed into martensite which surrounds the ferrite. During the tempering the martensite transforms to cementite whereas the low carbon zone, i.e. ferrite, remains unaffected.

Chemical Analysis of the Microstructure

A detailed chemical analysis of the AISI 4340 steel alloy system was carried out with an energy-dispersive unit attached to the SEM. Each time a second phase particle was detected, a qualitative chemical analysis was performed in two steps. First, a narrow electron beam was focused on the target and the full spectrum of the emitted energy was recorded by the Energy-Dispersive Unit. The spectrum was analyzed by an emission energies chart for the occurrence of the various chemical elements. Second, a method was employed to detail the distribution of a particular element in the scanned area. This method determines the relative concentration of the elements in the matrix. The micrograph
in Fig. 62 shows a second phase particle which was analyzed. Figure 63 depicts a typical diagram of a number of electronic counts vs energy (KeV). The interpretation of the full spectrum relied on reading the energy level of each peak. In the particular case of Fig. 63 the sequence of peaks from the left is $K_a$ for silicon at 1.74 KeV, and $K_b$ at 1.83 KeV. At the center lies iron with $K_a$ of 6.4 and $K_b$ of 7.05 KeV. The last peak belongs to gold with L$_{III}$ series of 137 KeV. Since gold was used to coat the specimen to improve resolution, and iron is the metal base, it is concluded that this particle is a silica type, ($SiO_2$).

The same procedure was followed for MnS inclusions. Figure 64 shows the examined particle and Figure 65 the full spectrum in this case. The peak in the formost left is sulfur with $K_a$ of 2.50 KeV. After it, lie traces of calcium at $K_a$ 3.69 KeV. Then, a high peak at 5.89 KeV which corresponds to $K_a$ of manganese. The two common peaks of iron $K_a$ and $K_b$ at 6.40, 7.05 and KeV, respectively are present immediately following those of manganese. The last peak at the right is that of gold. Thus, this particle is manganese sulfide (MnS).

For the same MnS particle Figs. 66 and 67 elucidate the usage of scanning microscopy for a particular element in the matrix using the second method discussed before. Figure 66
Figure 62 - SiO₂ particle in an illustration for a chemical analyses.

Figure 63 - Full spectrum of the chemical elements of the SiO₂ particle.
Figure 64 - Showing the MnS particle to be analyzed.

Figure 65 - Full spectrum of the chemical elements in the MnS particle.
show a high concentration of sulfur and Fig. 67 a low concentration of iron. In Figure 68 a similar appearing particle, was detected in scanning the microstructure. This suspected area was searched for chromium manganese. The first is a carbide former and the second one an inclusion producing element. Figures 69 and 70 indicate a uniform distribution of these two elements, whereas uniform distribution of iron proves there is an absence of any type of inclusion, Fig 71. Thus the "particle" is a lean carbon area i.e. the ferrite matrix.

Crack Formation

Load-deflection curves were determined as shown in Figure 72. The samples at various stages of deformation were studied metallographically. The microstructure of a heat-treated AISI 4340 was characterized by a volume fraction of 6.5% cementite. The tempered structure consists of a ferrite matrix with elongated manganese sulfide particles as the predominant inclusion, with their major axes oriented in the rolling direction. The other inclusion type consisted of isolated angularly shaped silica particles, Figure 36a. For each notch-size case, a series of specimens was examined at 980N (100 Kg) increments. These specimens were examined to determine conditions for
Figure 66 - Showing a scanning map for sulphur in Fig. 64.

Figure 67 - Showing a scanning map for iron in Fig. 64.
Figure 68 - Examined area for a suspected inclusion.

Figure 69 - Showing the results for scanning for chromium in Fig. 68.
Figure 70 - Showing the results for scanning for manganese in Fig. 68.

Figure 71 - Showing the results for scanning for iron in Fig. 68.
Figure 72 - Load-deflection of AISI 4340 steel in three point bending.
void initiation, crack initiation, and maximum load. The results are presented in Table 5. The strain as a function of deflection angle is taken from Figure 31. The total strain for void initiation-crack initiation is around 0.05 for all three notch sizes. Although, the total strain for crack initiation is almost identical for each notch size, the plastic deflection angle varies markedly from the sharp notch to the large one. This means that there is a high strain-gradient in the smallest notch specimen.

<table>
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<table>
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<th>Max. Load</th>
<th>0.59</th>
<th>1.19</th>
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<td>$L$, Kg</td>
<td>2850Kg</td>
<td>3050Kg</td>
<td>3150Kg</td>
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<tr>
<td>$\Theta$</td>
<td>1.6°</td>
<td>3.5°</td>
<td>6.7°</td>
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<tr>
<td>$\varepsilon_{yy}$</td>
<td>0.10</td>
<td>0.13</td>
<td>0.15</td>
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The pre-deformation separation between the ferrite and the ellipsoidal MnS inclusions remains the same upon increasing the total strain. Hence, a special interest was taken in examining the interface between the cementite particles and ferrite matrix. The predominant mode of void formation was by interface decohesion of the cementite-ferrite interface. In rare cases the boundaries of areas free of cementite served as incompatibility sites for void occurrence.

Unlike the results in the AISI 1095 carbon steel\(^{39}\) there is not a distinct void-initiation stage separate from the crack formation stage in AISI 4340 steel. Microstructural examination did not reveal any voids prior to the crack initiation. When strain for crack initiation was reached, only then were voids observed. The crack started at the notch surface, away from the mid-point in agreement with Girifiss and Spretnak\(^{17}\), Fig. 73. Propagation took place along a set of characteristic slip lines. Figures 74, 75 and 76 show explicitly the presence of one principle crack along \(\alpha\) and \(\beta\) slip lines crack propagation.

This stage of fracture involves stable crack growth, with the final ductile fracture dimple space corresponding to the cementite spacing, Figs. 79 and 80. As the crack reaches the midsection of the bar, the increased stress
Figure 75 - An overview of the shear lip inside the notch in AISI 4340 steel. The two cracks as present on both sides of the mid-plane.
Figure 74 - Electron microscopy of crack formation along an instability line in AISI 4340 steel. Three-point-bending, 0.59mm case.
Figure 75 - Electron microscopy of crack formation along an instability line in AISI 4340 steel. Three-point-bending, 1.19mm case.
Figure 76 - Continuation of the crack initiated at the notch surface shown in Fig. 75.
intensity causes a transition to an unstable crack portion in a plane normal to the \( y \) axis, which is the principle stress axis. The crack is composed of sheets of voids that were produced at the carbide interface, Figs. 77, 78 and 79. An important observation was the absence of a high population of voids throughout the matrix during the cracking of the sample. Few voids are spread near the crack plane, Fig. 77.

The lack of voids near the main crack plane is attributed to the low degree of homogeneous plastic deformation.

The crack along the shear instability line is a consequence of highly localized plastic flow once a sharp crack forms. Evidence for such localized flow is provided by Rice and Johanson\(^\text{(27)}\), whose work (on AISI 4340) shows that large strain exists at the crack tip region for a sharp crack, one order of magnitude higher than the total strain measured here.

The characteristics of the microstructure for three notches were almost similar. The major difference was that the various stages of fracture occurred at successively smaller deflection angles for the smaller notches, although the resulting total strain at the surface was about the same, as shown in Table 5. Another difference is depicted
Figure 77 - Voids lying along the crack-line.
Figure 78 - Crack-tip of the main crack on an instability line.

Figure 79 - Voids at the crack-tip of the main crack on an instability line.
in Figs. 74, 75 and 76. The plastic zone size is smaller for the more acute notch. This results in a shorter stable crack path for the smaller notch.

When the strain exceeds the critical value for crack initiation, propagation takes place along a shear instability band lying along a characteristic slip trace. Sheets of voids formed near the notch root with their loci along these logarithmic spiral characteristic slip trace. The onset of plastic instability causes an intense elastic-plastic stress field at the advancing crack tip which is powerful enough not to be diverted by large inclusions that are surrounded by large voids, Fig. 74. Hence, these inclusions play a minor role in the crack formation and propagation. Only when some of the randomly distributed inclusions were lying on a susceptible shear instability line was crack growth assisted. The influence of such inclusions in the crack growth can be seen in the fracture surface, Fig. 80.

Discussion

The incompatibilities throughout the matrix arising near interphase interfaces lead to stress concentrations which produce void formation during crack propagation. However, some sites which are potentially sources of
Figure 80 - Electron fractographic of tempered AISI 4340 steel, showing areas of a shear lip near the notch in the three-point-bending.
incompatibility are less active in the crack propagation: These include the separated interfaces of the various inclusions.

The void initiation stage occurred at a higher fraction of the load for crack initiation than was the case for the AISI 1095 steel. Also the load required to form voids occurred at a higher multiple of the load for yielding for the AISI 4340 than for the AISI 1095. In the case of the more ductile 1095 steel voids occurred at loads closer to the load for mutual yielding, while the crack formed later when a large plastic strain was reached. The AISI 4340 steel, high strength steel, exhibits a different behavior. Because the void initiation and void profusion coincide with the crack initiation stage, observations of crack formation give essentially the same parameters for void initiation as well as crack formation.

As indicated previously, the crack is initiated at the free surface away from the mid-point of the notch root and is propagated along a plastic instability trace. The various figures showed explicitly that the crack path was not diverted by inclusions, even when a large inclusion was lying near the crack line. The fine dimples seen in the fractographs arise from voids found at the cementite particles with only an occasional large dimple formed due to presence of an inclusion.
Thus, the evidence indicates that the mode of cracking is one of development of a shear instability front which leads to a sheet of voids which rapidly coalesce to form a crack. The rapid coalescence of voids on these instability shear lines provides an explanation for the brittle behavior of the high strength alloy. Even though the microscope examination showed the typical features of dimple rupture, the spacing of the voids at cementite particles indicates a highly contained plastic zone of the order of the cementite particle spacing, an attendant minimal plastic contribution to the energy release required for cracking, and hence brittle behavior.
Various workers attributed dispersion strengthening of nickel alloy at room temperature to the Orowan mechanics of hardening by nondeforming particles. However, in addition to this mode of strengthening by second-phase particles it is possible to develop additional strength by controlling the thermomechanical process. Room temperature strength was produced by the thermomechanical process as a result of refining the grain size and substructure spacing (cell size), in accord with usual Hall-Petch analyses. The stress $\sigma$:

$$\sigma = \sigma_0 + k \varepsilon^{-1/2}$$

where $\varepsilon =$ grain or cell size, $\sigma_0$ can be considered as the yield strength of the pure base metal single crystal. Wilcox and Clauer tested the 0.2% offset yield strength of $Ni-2% ThO_2$ (TDNi) for a series of grain sizes at room temperature. The plot 0.2% offset yield strength as a function of $\varepsilon^{-1/2}$ shows a linear relationship, in excellent agreement with the Hall-Petch equation. Thus, a fine, stable, very elongated grain structure is optimum for combined room temperature and high temperature strengthening.

As mentioned in the experimental procedures section, the
Dispersion Strengthen Ni, DS Ni, is processed differently than the nickel base alloy known as Thoria Dispersion Nickel, TDNi. The DS Ni used in this study went through fewer steps in the forming process and the subsequent heat treatment. This results in a marked difference in behavior of the microstructure during deformation when compared to the TD Ni. The more refined grains and substructure spacing increases the room-temperature strength. Higher yield strength and a smaller reduction in area is observed in the DS Ni case.

Figure 81 shows the size and distribution of the ThO₂ particles in the as-annealed sample. The particles are roughly spherical in shape and exhibit dark contrast. The size of the particles is not uniform and the diameter varies between 10 and 200 nm. The width of the elongated grains range from 0.2 nm to 1 nm.

The DS Ni was exposed to the same straining procedures as the AISI 1095 and AISI 4340 steels, Fig. 82. Mechanical properties presented in Table II (Experimental procedure) show a large reduction in area with a small amount of uniform elongation. This is a direct result of a prolonged necking stage prior to complete separation of the fractured surfaces. The ability of the DS Ni to undergo a (highly) localized plastic strain indicates the high
Figure S1 - Electron micrograph of the microstructure of DS $N_1$ as received.
Figure S2 - Load-deflection of DS Ni in three-point-bending.
adherence of the ThO₂ particle/matrix interface.

As described in the experimental procedures section specimens of each notch were strained to various loads in three-point-bending and then prepared for viewing in a TEM. The important stages during deformation are presented in Table 6. Clearly, the main changes in microstructure take place after the yield point. In Fig. 83 the dislocation particle interaction occurring after yielding can be seen. The dislocations are pinned and some are
Figure 83 - Arrays of dislocations pinned by the ThO₂ particles prior to the void initiation stage.
tangled around the particles. This dislocation structure indicate that dislocation lines by-pass the stationary obstacles, i.e. the thorium particles, by cross-slip (and perhaps local climb) producing ragged loops and tangles as proposed by Hirsch and Humphreys. No circular, Orowan loops behind or around the particles were observed. Similar results were reported by Heimerdahl and Thomas. Numerous investigators worked on two phase-alloys with non-deformable particles embedded in a ductile matrix and proposed different mechanisms for the relaxation of plastic strain developed at the interface. Fisher, Hart and Pry reported concentric loops were left around particles as a consequence of cross-slip of the dislocations. Humphreys and Hirsh used copper alloys containing alumina particles and showed that the dislocation structures associated with particles consist generally of prismatic loops and Orowan loops. The formation of the interstitial prismatic loops was explained in terms of a cross-slip mechanism controlled by the misfit strains around the particles. The accommodation of the plastic strain that is primarily confined to the matrix can be explained based on Ashby's model using the concept of geometrically necessary dislocations. Regardless of the mechanism that operates during plastic flow around the particles, the dislocations
giving rise to plastic flow can be interpreted as geometrically necessary.

At larger strains, voids appear at the ThO₂ matrix interface. Figure 84 shows few particles with voids aligned in the rolling direction, which is also the direction of the force. The detection of voids in the TEM was difficult. Two factors were affected the lack of a large number of voids at high strain levels in comparison to the Th nickel case of Wilcox and Clauer. First, the different thermomechanical processing of the DS nickel can result in a change in cohesive strength of the ThO₂/matrix interface. Second, the twin-jet electropolish unit did not remove the hard ThO₂ particles from the surfaces during the thinning process. This gives one the false impression of a large density of second-phase particles in the matrix. Since the few voids were observed at a high strain in the load-deflection curve close to the crack initiation stage no distinction was made between the two different stages.

Crack Initiation and Propagation

Crack propagation then ensued. Figures 85 and 86 show a crack which formed along a shear instability line. This type of crack appearance is consistent with crack behavior in the other two alloys, AISI 1095 and AISI 4340. Further-
Figure 84 - Void initiation at the ThO$_2$ particles and matrix interface.
Figure 85 - Overall view of the U-notch with cracks in DS $N_1$.

Figure 86 - Crack formation along a shear instability line in DS $N_1$. 
more, fractography was done on the stable crack portion, Figures 87 and 88, on a shear lip near the notch surface seen in Figures 85 and 86. The dimples in Figures 87 and 88 show a directionality which clearly, indicate the presence of a shear stress component. Also, Figures 86 and 87 show a typical dimple rupture. The very fine dimples are associated with small size thorium particles distributed within the nickel matrix. The large dimples are caused by the few large ThO₂ particles that reach size of 0.2μm.

In order to obtain a better understanding of the ductile fracture in DS N₁, the fracture surface of the standard tensile specimen was analyzed. A close-up of the fracture in the necked area of the round bar, Figures 90 and 91 depicts the large number of holes away from the fractured surface. A comparison of the dimple rupture surface as shown in Fig. 90 to the one in Fig. 88 reveals a difference in dimple size. The dimples produced during the tensile test are larger in the longitudinal and transverse direction than the dimples produced during the bending test. This implies that the void-growth stage was minimized in the notched specimen in the three-point-bending. Again, as discussed previously, the localized plastic strain is responsible for reducing the degree of homogenous plastic deformation throughout the specimen. It results in lowering the void-growths and initiating a crack at lower strain.
Figures 87, 88 - Electron fractograph of the shear lip in three-point-bending of DS N₁ showing fine and coarse dimples.
Figure 89 - Electron fractograph of fine dimples in three-point-bending in DS Ni.
Figure 90 - Electron microscopy of tensile specimen of DS N\textsubscript{1}, necking area.

Figure 91 - Electron fractograph of tensile specimen of DS N\textsubscript{1} shear lip at edge of fractured surface.
CONCLUSIONS

1. Void nucleation takes place preferentially at sites of elastic and plastic incompatibility stress concentration including sites of three-phase contact, two-closely spaced carbide particles, (in agreement with the postulate of Argon et al\textsuperscript{9}), and large elongated MnS particles in AISI 1095 steel.

2. Void nucleation and growth occurs predominantly along the characteristic slip lines which are traces of sheets of plastic shear instability in AISI 1095 steel.

3. The onset of plastic instability is favored by the presence of a free surface and by a large plastic strain gradient (in agreement with the proposals of Spretnak, McClintock and coworkers\textsuperscript{15-17}) for all three materials studied, AISI 1095 and 4340 steel and DS Ni\textsubscript{j}.

4. Crack initiation begins at the free surface and not in the region of maximum stress or stress triaxiality and at a constant level of plastic strains, in agreement with prior work of Mohamed and Tetelman,\textsuperscript{26} for all three materials.

5. When a notch is present large inclusions play a minor role in crack formation in AISI 1095 and 4340, although separation at the interface occurs at some of the particles. This minor effect is attributed to
intensified local plastic strain at the crack-tip.

6. Decohesion of carbide particles/matrix interfaces produces the dimple rupture appearance of the fracture in surfaces of AISI 1095 and 4340 steels.

7. The fracture of the alloys studied here was a consequence of plastic flow in shear instability bands lying along characteristic slip traces. The localized shear caused void nucleation and growth and, in turn, fracture along these traces.
REFERENCES


