

The Structure Sensitivity of the Effects of Pressure upon the Ductility of Fe – C Materials

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ABSTRACT. The effects of a superposed hydrostatic pressure of up to 24 kbars upon the ductility of a series of annealed and spheroidized Fe – C alloys ranging in carbon content from 0.004 to 1.1% are investigated. The effects of pressure upon ductility are found to be highly structure sensitive in terms of the presence, amount and distribution of the cementite. In the absence of cementite, or when the cementite is in spherical form in a ferrite matrix, the relationship between pressure and true strain to fracture is linear, with the slope decreasing with carbon content. When the cementite is in platelet form (pearlite), or in a continuous network along prior austenitic grain boundaries, the relationship between pressure and true strain to fracture is nonlinear, and varies considerably with carbon content and structure.

The relationship between pressure effects upon ductility and atmospheric pressure mechanical properties is examined. It is found that no single linear relationship or proportionality between pressure coefficient of ductility and strain hardening coefficient exists for the range of Fe – C materials investigated.

The effects of pressure upon macroscopic fracture appearance is described and discussed. The absence of a linear or continuous relationship between fibrous to total fracture area and pressure is demonstrated.

THE EFFECTS of high superposed hydrostatic pressure upon the mechanical properties, particularly ductility, of metals has been the subject of extensive investigation. Although the effects of pressure upon the ductility of numerous metals have been well documented, little information has been generated with regard to the mechanism or mechanisms responsible for the observed effects, or to why the pressure-ductility behavior varies so widely for different materials. Summarized herein are the results of the first phase of an investigation aimed at gaining a fundamental understanding of how and why pressure affects the ductility of metals. Considered particularly is the question of the structure sensitivity of the phenomenon in terms of the presence, amount, shape and distribution of a brittle second phase in a ductile matrix.

The flow characteristics and ductility of a variety of metals exposed to superposed pressures to in excess of 20 kbars have been measured by Bridgman (1-3). Although he examined a variety of materials,

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a significant portion of his work dealt with steels. His observations and conclusions concerning the effects of hydrostatic pressure on the mechanical properties, particularly of steels, were quite extensive. Those considered most pertinent to the subject of this current investigation will now be summarized.

Bridgman observed that the mechanical property most sensitive to pressure is ductility. To a first approximation, he found that the form of the true stress-true strain curve is unaffected by pressure; the principal effect of pressure being to extend that portion beyond maximum load (onset of necking). For the case of steels he found that the strain-hardening rate was effectively independent of pressure and remained nearly linear.

Although the rate of increase of ductility in terms of the true strain to fracture ($\ln A_0/A_f$) is a function of the material, Bridgman concluded that it is, in all cases, a linear function of the form

$$P = \alpha + \beta \epsilon_f \quad \text{Eq 1}$$

where ϵ_f is the strain to fracture at some pressure P ; β the pressure coefficient of ductility which is a function of the material; and α a material constant. He observed that, even in the case of steels, β varied widely between materials.

For steels, Bridgman proposed that the true stress-

true strain curve as measured under pressure was linear, thus of the form

$$\sigma = \sigma_0 + N\epsilon \quad \text{Eq 2}$$

where N is the strain hardening coefficient, σ_0 = yield stress, and σ the flow stress at some strain ϵ . This is of the same form as Eq 1 for the relationship between pressure and ductility. Substituting in Eq 2 the fracture strain ϵ_f and fracture stress or flow stress at fracture (σ_f) and equating Eq 1 and 2, he obtained as a relationship between pressure coefficient of ductility, β , and strain hardening coefficient, N

$$N = 0.7\beta \quad \text{Eq 3}$$

Examination of Bridgman's own data shows that, although β appears to increase with N , there is considerable scatter. In the region below 10 kbar, the correlation is very poor. This region encompasses effectively all of his data for plain carbon and low carbon alloy steels of a variety of compositions and heat treatments. The poor correlation at the lower pressures, along with the very limited amount of data at the high pressure, sheds considerable doubt on the existence of the stated relationship or, if it does exist, what it actually is.

For steels, Bridgman further observed that the ratio of what he called the tensile or flat portion to the total fracture area of cup-cone type fractures decreased linearly with increasing pressure. He observed that in most steels the tensile region, which could more appropriately be called the fibrous region, disappeared at pressures in the range 10 to 20 kbars. Beyond the pressure at which the fibrous region was completely suppressed, the fracture became a planar shear.

Subsequent to the work of Bridgman, numerous investigations into the effects of pressure upon the ductility of many metals have been reported (4-9), one of the most extensive being that by Pugh (10, 11). Pugh observed that although two alloy steels, magnesium and cast iron, did exhibit a constant pressure coefficient of ductility, β , there were many exceptions. For example, copper and aluminum exhibited a β that decreased with increasing pressure, whereas zinc and bismuth exhibited an abrupt discontinuity in ductility in which the strain to fracture increased abruptly over a very narrow pressure region.

The results of other investigators shed doubt on the assumptions by Bridgman that ductility is a linear function of pressure even for all steels. Beresnev et al (5) observed that in a 0.46% C steel (R. 20), the ductility did not linearly increase with pressure, but above 13 kbar leveled off. In contrast, Pelczynski (6) reported that the ductility increased faster with pressure than a linear rate for a 1.1% C steel. In the case of a 1045 steel in the quenched and untempered condition, Davidson et al (9) observed effectively no increase in ductility up

to a pressure of 17 kbars, after which an abrupt increase over a narrow pressure region occurred.

In the works of prior investigators, structure was not thought important, thus was not controlled. In this current investigation the form of the ductility-pressure relationship as a function of the presence, amount, shape and distribution of cementite in a series of annealed and spheroidized Fe-C alloys was examined. The relationship between the pressure coefficient of ductility, β , and the strain hardening coefficient along with the effects of pressure upon macroscopic fracture appearance for these materials were also examined and compared to the observations of Bridgman.

MATERIALS, EQUIPMENT AND PROCEDURES

Materials

COMPOSITION AND HEAT TREATMENT

The materials utilized in this investigation consisted of a series of iron-carbon alloys of the following composition: Fe + 0.004% C; Fe + 0.40% C; Fe + 0.83% C; Fe + 1.1% C.

The base material, in all instances, was electrolytic iron of 99.94% purity. The materials were vacuum melted and subsequently reduced from a 5-in. ingot to $\frac{1}{2}$ in. diam by hot forging and swaging. In the case of the 0.004% C material, the final forming step was cold swaging from 1 in. diam in order to facilitate grain size control.

The materials were heat treated in argon, using the following procedures:

0.004% C

1. Annealing—1200 F—1 hr

0.40% C

1. Annealing—1650 F—1 hr—furnace cool
2. Spheroidization—1650 F—1 hr—oil quench + 1320 F—54 hr

0.83% C

1. Annealing—1700 F—1 hr—furnace cool
2. Spheroidization—1700 F—1 hr—oil quench + 1320 F—54 hr

1.1% C

1. Annealing—1850 F—1 hr—furnace cool to 1320 F—air cool
2. Spheroidization—1850 F—1 hr—oil quench + 1320 F—54 hr

The grain diameter obtained using the above heat treatment ranged from 2.4×10^{-3} to 2.7×10^{-3} in. for all materials.

The microstructures of the materials utilized are shown in Fig. 1 and 2 for the annealed and spheroidized conditions, respectively. The microstructure of the annealed 0.004% C material, as shown in Fig. 1A, consisted of ferrite with no visible cementite. In the case of the annealed 0.40% C material (Fig. 1B), the microstructure was comprised of pearlite,

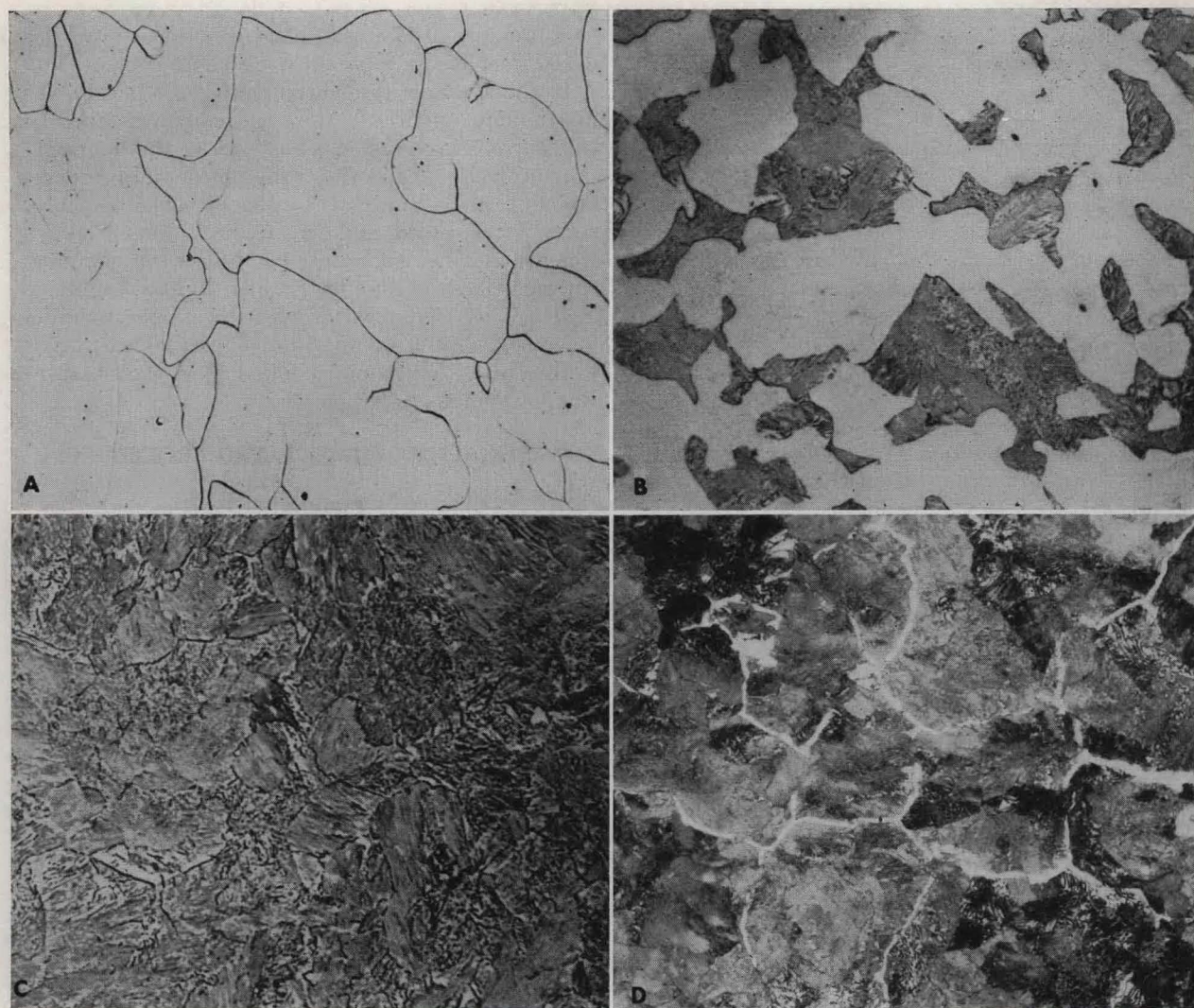


Fig. 1. Microstructure of annealed Fe - C materials. (A) 0.004% C; (B) 0.40% C; (C) 0.83% C; (D) 1.1% C. $\times 200$.

wherein the cementite was in platelet form, in a ferrite matrix. The microstructures of the 0.83% C and 1.1% C materials, as shown in Fig. 1C and D, respectively, consisted totally of pearlite in the former and pearlite with a continuous network of cementite along prior austenitic grain boundaries in the latter.

The microstructures of the spheroidized materials, as shown in Fig. 2A-C, consisted of cementite particles in a ferrite matrix. As shown, the size of the cementite particles increased and the interparticle spacing decreased with increasing carbon content.

SPECIMEN CONFIGURATION

The specimen configuration for the high-pressure testing consisted of a 0.160-in. gage diam with a 0.665 in. nominal gage length. The gage section was ground and polished.

The specimen configuration for atmospheric pres-

sure tensile testing was effectively the same as that used for the pressure tests except that the distance between shoulders was increased to facilitate diameter measurements during loading in order to obtain a true stress-true strain curve.

TABLE 1. Summary of Mechanical Properties

Material, % C	Condition	Yield stress, ksi	Ultimate tensile stress, ksi	Strain hardening coefficient, n
0.004	Annealed	17.2*	31	0.28
0.40	Annealed	25.0†	52.6	0.22
	spheroidized	29.2*	46.2	0.21
0.83	Annealed	33.0†	84.7	0.19
	spheroidized	29.0*	55.8	0.20
1.1	Annealed	53.8†	108.7	0.17
	spheroidized	31.7*	61.3	0.17

* Lower yield stress.

† 0.2% offset yield stress.

ATMOSPHERIC PRESSURE MECHANICAL PROPERTIES

The atmospheric pressure tensile properties were obtained using an Instron Tensile Machine with a 0.050 in./in. strain rate. The measured properties are summarized in Table 1. It should be noted that the actual form of the true stress-true strain relationship for the materials utilized herein was a power function. Thus the strain hardening coefficient (n) in Table 1 is not the same as the N in Eq 2 where a linear relationship has been assumed.

Equipment

The pressure device utilized consisted of a Bridgman piston-cylinder type 30 kbar hydrostatic pressure system (1). The main pressure cylinder, which consisted of an autofrettaged externally tapered inner liner and matching restraining jacket, contained a $\frac{3}{4}$ -in. diam 8-in. long pressure cavity. Both liner and jacket were fabricated of 250 grade maraging steel. Electrical leads were introduced into the pressure cavity through the bottom closure for the purpose of measuring pressure and strain of the sample.

Pressure measurement was accomplished by means of a 120-ohm manganin coil comprising one arm of a bridge which also contained a Foxboro recorder. The estimated inaccuracy in pressure measurement was $\pm 2\%$.

The fixture for the conduct of the tensile test under pressure, which fitted inside of the pressure cavity, consisted of four legs. Two legs were stationary and bore against the bottom closure of the pressure cavity. The remaining two legs were driven downward by the advancing main pressure piston which introduced tensile forces in the sample. The strain in the sample during testing was measured by means of a slide-wire resistance circuit which measured the relative displacement between the stationary and movable legs of the fixture. Displacement as a function of time was continuously recorded using an X-Y recorder. The circuit had an output of 55.0 mv/in. of displacement.

Procedures**HIGH-PRESSURE TENSILE TESTING**

The high-pressure tensile tests were conducted at a constant displacement rate of 0.050 in./min.

Since the permissible displacement of the main pressure piston was fixed by the length of the tensile specimen and associated fixture, a means for varying the test pressure had to be provided. This was accomplished by pre-charging the pressure cavity to various selected levels from 0.5 to 3.0 kbars with nitrogen gas which also acted as the pressure transmitting media.

The strain in the specimen was accomplished by the advance of the main piston. Thus, the pressure

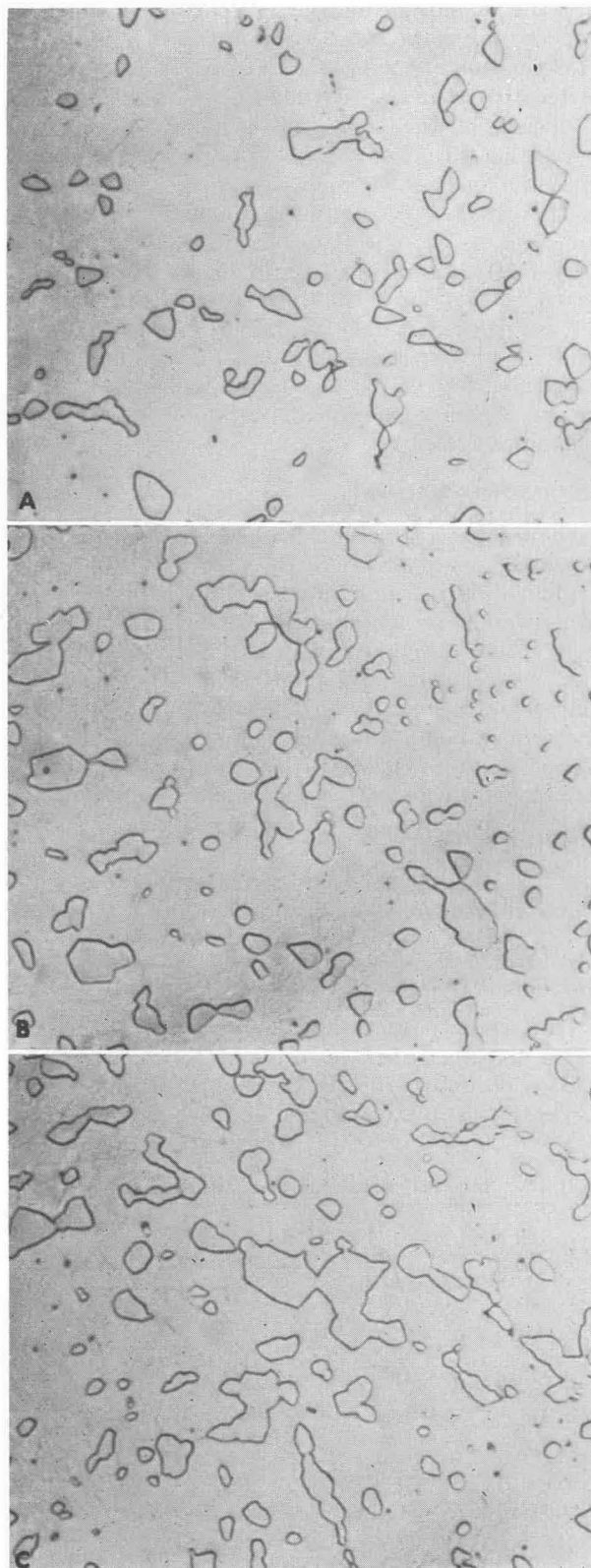


Fig. 2. Microstructure of spheroidized Fe-C materials. (A) 0.40% C; (B) 0.83% C; (C) 1.1% C. $\times 1000$.

continuously varied during the conduct of the test with the magnitude of the pressure change being proportional to the total strain in the sample. The determination of the pressure at which the specimen started to strain and at fracture was accomplished by means of the slide-wire displacement circuit. By plotting the output of the slide-wire circuit versus time, the onset of strain was readily detectable. Except in the very ductile materials at high pressure, such as the 0.004% C alloy, fracture was indicated by an abrupt discontinuity in the displacement-time plot due to the stored energy in the sample at the time of fracture. In the case of the 0.004% C material at high pressure, the pressure at fracture was determined by comparing the total elongation after fracture to the displacement-time plot and the corresponding pressures.

POST-TEST MEASUREMENTS

The total elongation and diameter at fracture were measured using an optical comparator at X20 magnification. A minimum of 3 diameter measurements were made at different orientations and averaged. The inaccuracy in this measurement technique was estimated to be ±0.001 in.

A second technique used for measuring both the fracture diameter and the diameter of different regions on the fracture surface was by means of an optical microscope with a micrometer eyepiece viewing along the normal to the fracture surface.

RESULTS AND DISCUSSION

Effects of Pressure Upon Ductility of Fe - C Materials

RELATIONSHIP BETWEEN PRESSURE AND TRUE STRAIN TO FRACTURE

True strain to fracture (ϵ_f) vs pressure for the materials investigated is shown plotted in Fig. 3. The statistical analysis and relationships for the best fit curves shown are summarized in Table 2.

Of initial concern was the form of the curve that best fit the data. Referring to the statistical results of Table 2, it can be seen that the correlation coefficients (r) are quite high for all data for a linear fit. However, the correlation coefficients are the highest, and the standard deviation (σ) the lowest for the annealed 0.004% C and spheroidized 0.40, 0.83 and 1.1% C materials. This indicates that these materials exhibit a linear relationship between ductility and pressure of the form:

$$\epsilon_f = A + BP \tag{Eq 4}$$

where A and B are the constants listed in Table 2 for a linear fit. It should be noted that since P is the independent variable, for the purpose of this analysis Eq 4 is a more appropriate relationship between strain to fracture and pressure than that given by Bridgman (Eq 1). The relationship between the constants for the two equations is simply:

$$B = \frac{1}{\beta} \text{ and } A = -\frac{\alpha}{\beta}$$

To further examine the form of the best fit relationship between strain to fracture and pressure, a test was made for a polynomial fit of the form:

$$\epsilon_f = A + BP + CP^2 \tag{Eq 5}$$

which is also summarized in Table 2.

It can be seen in Table 2 that the standard deviation (σ) is effectively unchanged from that of a linear fit for the cases of the annealed 0.004% C and spheroidized materials. The deviation of "C" from zero is quite small, and the confidence level is much less than 95% which is the lower limit for acceptable fit. It is apparent then that no improvement results by using a polynomial rather than a linear fit for these materials.

In contrast to the above materials, there is a considerable decrease in σ , and the deviation of "C" from zero is quite large in the cases of the annealed

TABLE 2. Statistical Analysis of Test Data

Material, % C	Condition	Coefficients			Correlation coefficient, r	Standard deviation, σ	Confidence level, %	Deviation of "C" from zero, σ units
		A	B	C				
<i>A—linear fit ($\epsilon_f = A + BP$)</i>								
0.004	Annealed	1.968	0.341	...	0.984	0.158		
0.40	Annealed	0.726	0.218	...	0.990	0.216		
0.40	Spheroidized	1.339	0.263	...	0.999	0.036		
0.83	Annealed	-0.043	0.157	...	0.968	0.327		
0.83	Spheroidized	0.668	0.216	...	0.997	0.107		
1.1	Annealed	-0.014	0.045	...	0.951	0.133		
1.1	Spheroidized	0.470	0.195	...	0.995	0.175		
<i>B—Polynomial fit ($\epsilon_f = A + BP + CP^2$)</i>								
0.004	Annealed	2.005	0.276	0.011	...	0.160	64.04	.916
0.40	Annealed	0.867	0.117	0.006	...	0.015	100.00	39.05
0.40	Spheroidized	1.342	0.258	0.0005	...	0.038	35.20	0.467
0.83	Annealed	0.194	0.043	0.006	...	0.119	100.00	9.58
0.83	Spheroidized	0.673	0.212	0.0003	...	0.115	13.19	0.166
1.1	Annealed	0.083	0.003	0.002	...	0.082	99.99	3.967
1.1	Spheroidized	0.526	0.162	0.001	...	0.163	84.4	1.418

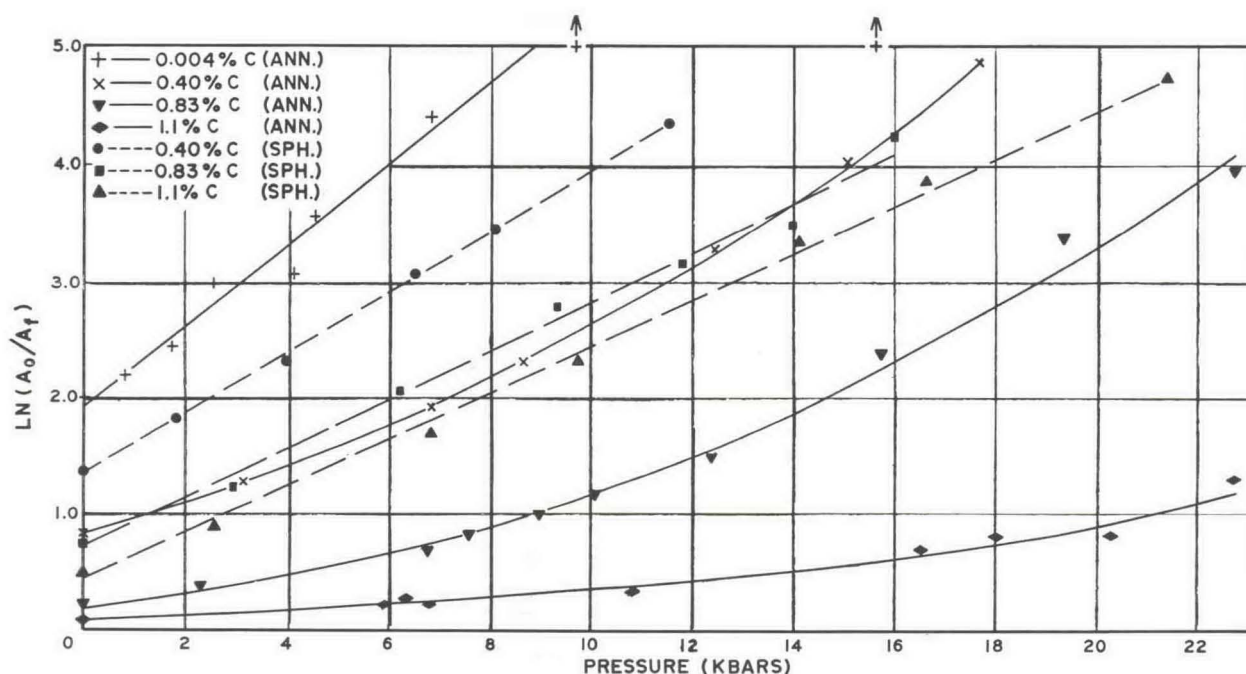


Fig. 3. Ductility vs pressure for Fe-C materials.

0.40, 0.83 and 1.1% C materials when a polynomial fit is attempted. Furthermore, from Table 2 it can be seen that the confidence levels of the polynomial fit exceed 95% and, in fact, are effectively 100% for all three cases. It is apparent, therefore, that the strain to fracture-pressure data for the annealed 0.40, 0.83 and 1.1% C materials are best described by a polynomial rather than a linear relationship.

Several other points are important to note in connection with Fig. 3. In the case of the materials exhibiting a linear relationship, the slope (B) progressively decreased with increasing carbon content. Similarly, if one assumes a linear relationship between strain to fracture and pressure for the remaining materials rather than a polynomial fit, they also exhibited a decrease in slope (B) with increasing carbon content. The significance of this result with respect to the relationship between the pressure coefficient of ductility and strain hardening coefficient will be discussed subsequently.

A second important point concerns the form of the curve for the annealed 0.40, 0.83 and 1.1% C materials. In these materials, the pressure at the beginning of substantial deviation from linearity increased with increasing carbon content.

Finally, in the case of the materials exhibiting a nonlinear relationship between pressure and strain to fracture, the slope or pressure coefficient of ductility at the higher pressures tended to approach that for the spheroidized materials of the same carbon level. This is readily seen in the case of the annealed 0.40 and 0.83% C materials. It is likely that the slope of the annealed 1.1% C material would also approach that of the spheroidized materials at higher superposed pressures.

In summary then, the effects of pressure upon the true strain to fracture was found to be highly structure sensitive, both in terms of the slope B and the form of the relationship between strain to fracture and pressure. Annealed 0.004% C and spheroidized materials exhibited a linear relationship between pressure and strain to fracture, whereas the annealed materials containing substantial carbon exhibited a definite nonlinear polynomial relationship with B increasing with increasing pressure. The slopes of the curves B all decrease with increasing carbon content. In the case of the annealed carbon containing materials, the slope at high pressure approached that for the spheroidized materials of equivalent carbon content.

As previously discussed, Bridgman primarily used materials that were in the "as-received" or quenched and tempered condition. Of all of his data, two plain carbon steels, that were supposedly in the annealed and/or spheroidized condition (the actual structure is unknown), can be used for comparison with the results of this current investigation. These data of Bridgman are shown in Fig. 4, along with the pertinent curves from our investigation. Bridgman's actual data points are shown and the best fit curves drawn through these points. For simplicity, the data points from the current investigation have been omitted with the points shown being used only in order to identify the curves. The curves from the current work are dashed with those from Bridgman being solid.

It can be seen in Fig. 4 that the best fit curves for Bridgman's data are not linear as he has stated, but deviate considerably from linearity. In the case of his annealed 0.90% C material, the agreement is

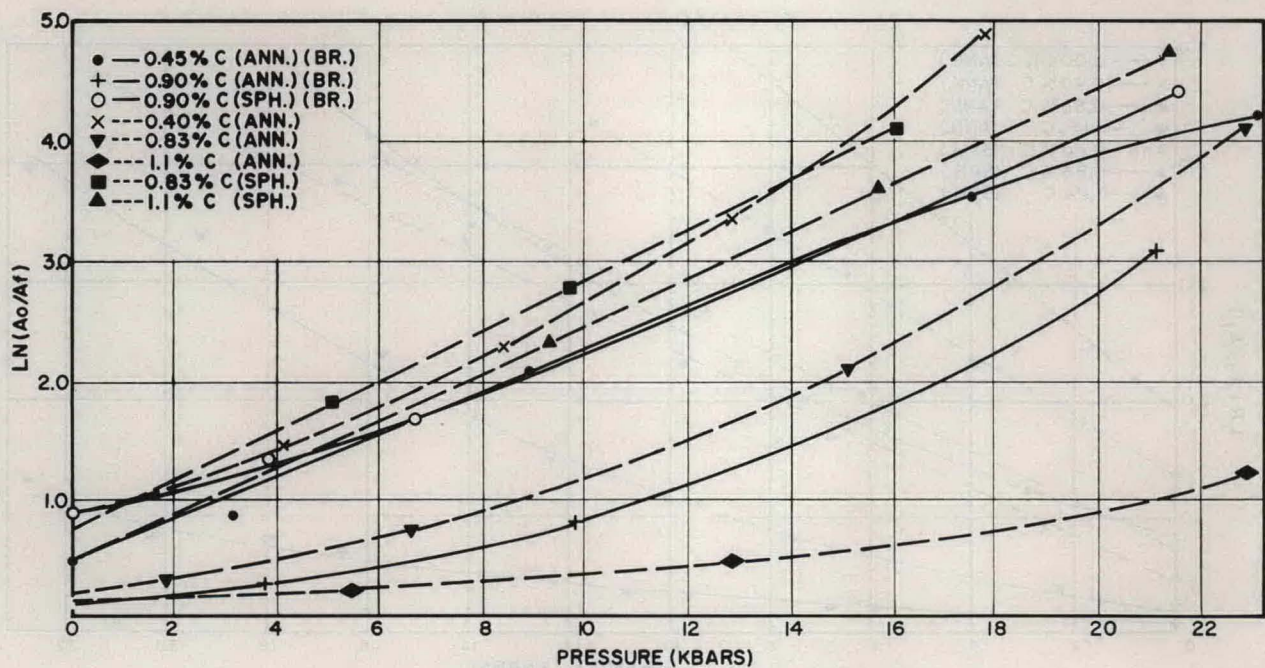


Fig. 4. Ductility vs pressure for Fe-C materials including Bridgman's data.

quite good with our data obtained for the 0.83% C material. The slope is slightly lower due to the higher carbon content, but B increases with pressure in the same manner. Bridgman's data for the annealed 0.45% C material is quite linear at lower pressures, but B decreases at higher pressures rather than increases as we have found. His spheroidized 0.90% C material exhibits nonlinearity with B increasing with increasing pressure, which is in contrast to the fact that linearity was obtained for all of the spheroidized material in this work. It should be noted, however, that there is a small number of data points available to describe Bridgman's curves and, in most cases, they are not uniformly distributed with respect to pressure. Based on the data available, however, it is obvious that these materials do not exhibit a linear relationship between pressure and strain to fracture and the forms of the curves vary considerably both from material to material and as a function of microstructure. Except for the annealed 0.90% C material, the agreement with the results of the present investigation is not good. This may be accounted for by the lack of knowledge of the actual microstructure of his materials and, as previously stated, the lack of usable data.

EFFECTS OF PRESSURE UPON ELONGATION

Elongation as a function of pressure is plotted in Fig. 5 for the materials investigated.

In the case of the most ductile of the materials investigated, that is, the annealed 0.004 and 0.40% C, and spheroidized 0.40% C materials, the elongation increased slightly at lower pressures, then rapidly leveled off with no further observed increase in

elongation with increasing pressure. These materials undergo substantial necking even at atmospheric pressure. Pressure does not, in this case, affect the uniform strain, but only enhances the amount of reduction in area obtained in the necked region. Thus, as the reduction in area in the necked region becomes greater as a result of increasing pressure, its contribution to the over-all elongation becomes less significant. This, then, accounts for the effective insensitivity of elongation to pressure at the higher pressures.

In the lower ductility annealed 0.83% C and spheroidized 0.83 and 1.1% C materials, the increase in elongation with pressure was much more extensive and only tended to level off at quite high pressures. This is a manifestation of the smaller amount of necking obtained at atmospheric pressure for these steels. As a result, there is a greater contribution of the reduction in area in the necked region to the over-all elongation as the pressure is increased. The leveling off in elongation observed at high pressure is attributed to the increased degree of necking as previously discussed for the more ductile materials.

For the relatively brittle annealed 1.1% C material, there was a large and continuous increase in elongation with increased pressure with no signs of leveling off. This is likely due to a two-fold effect. First, since there was very little, if any, necking observed for this material at atmospheric pressure, there was a large contribution to the elongation resulting from the substantial necking that occurs under pressure, as will be subsequently shown. The second contribution was due to the fact that this material fractured at very low plastic strains at low or atmospheric pressure. It is, therefore, likely that

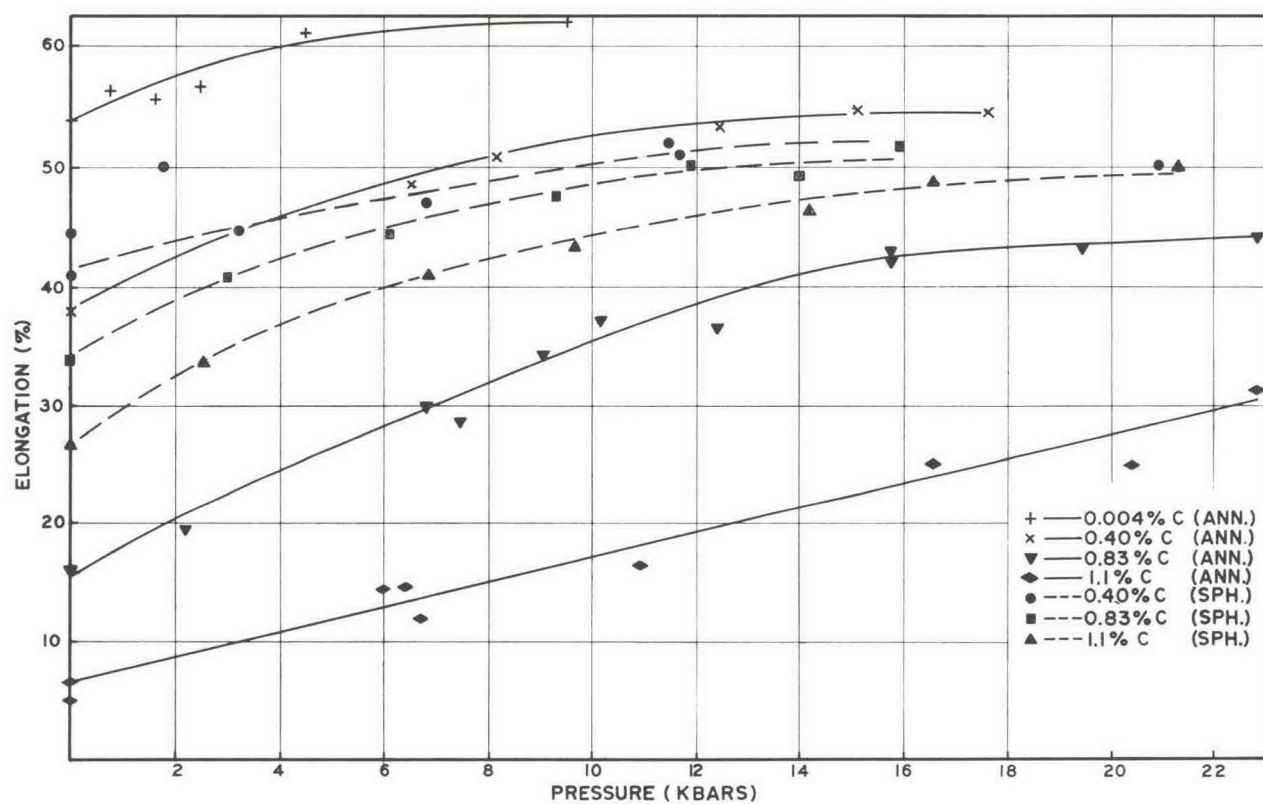


Fig. 5. Elongation vs pressure for Fe - C materials.

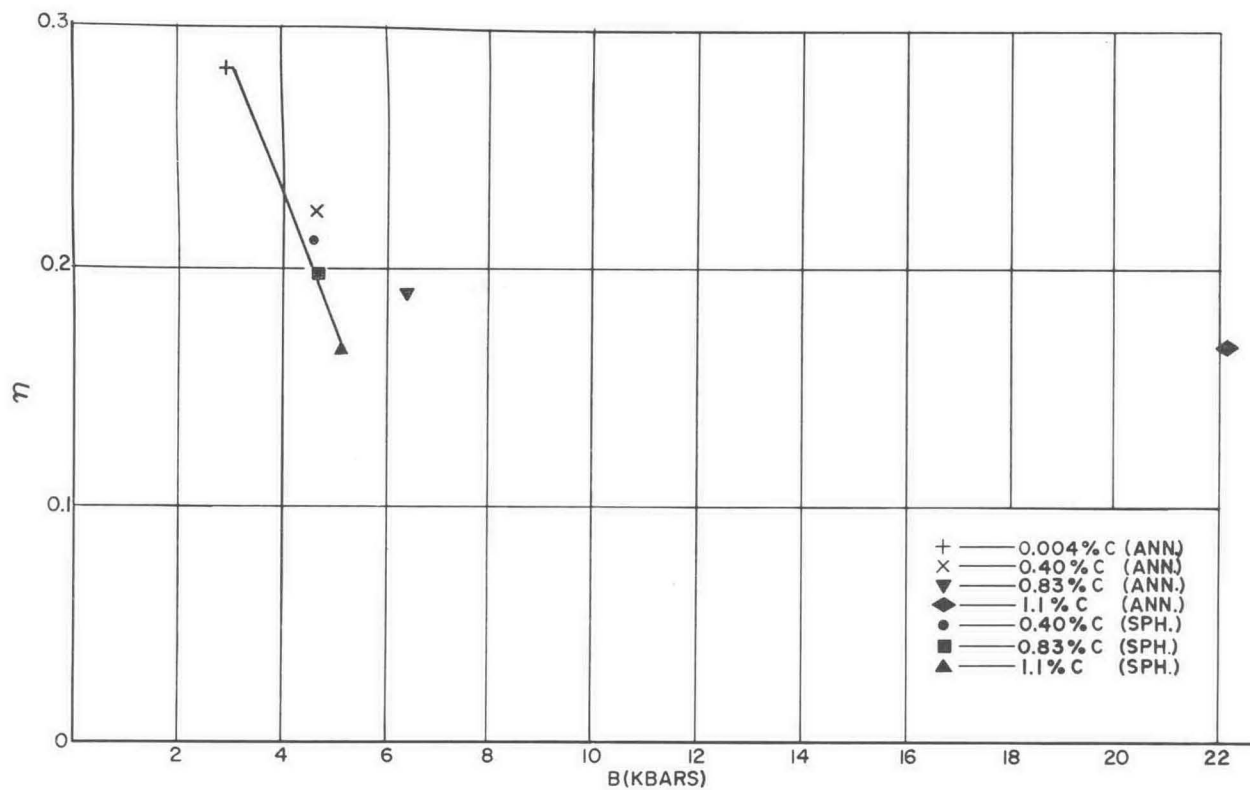


Fig. 6. Strain hardening coefficient vs pressure coefficient of ductility.

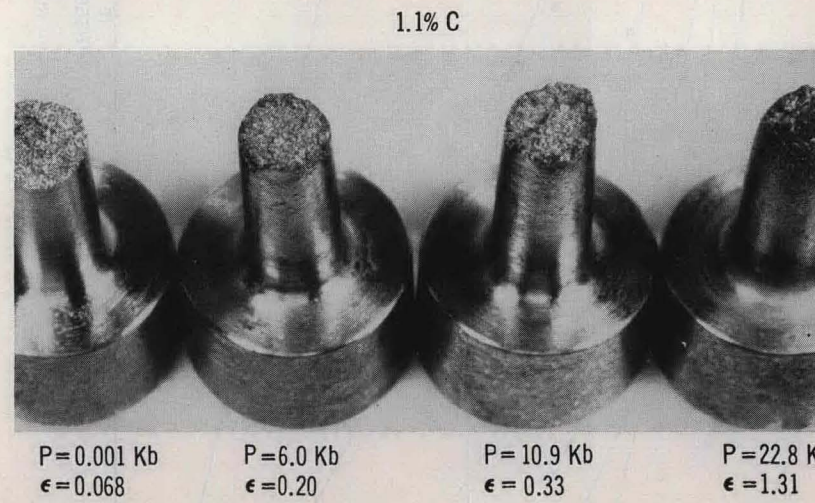
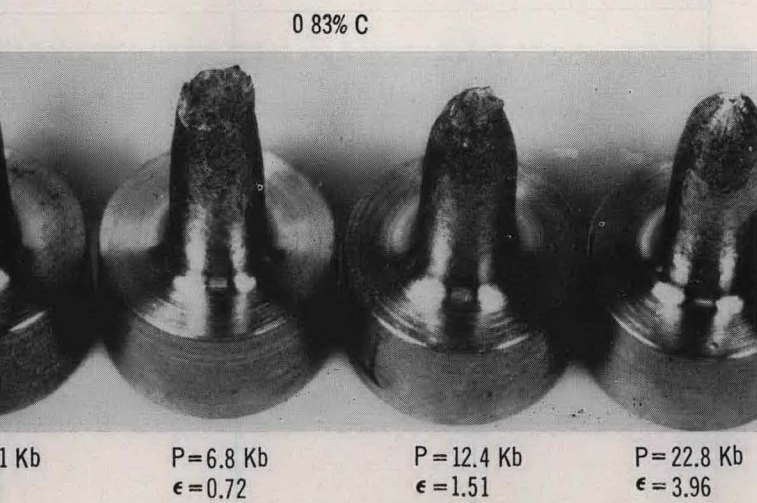
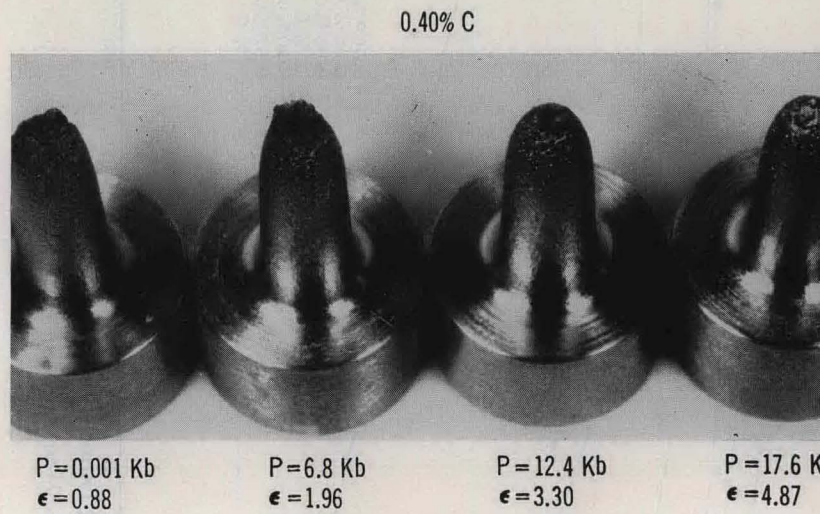
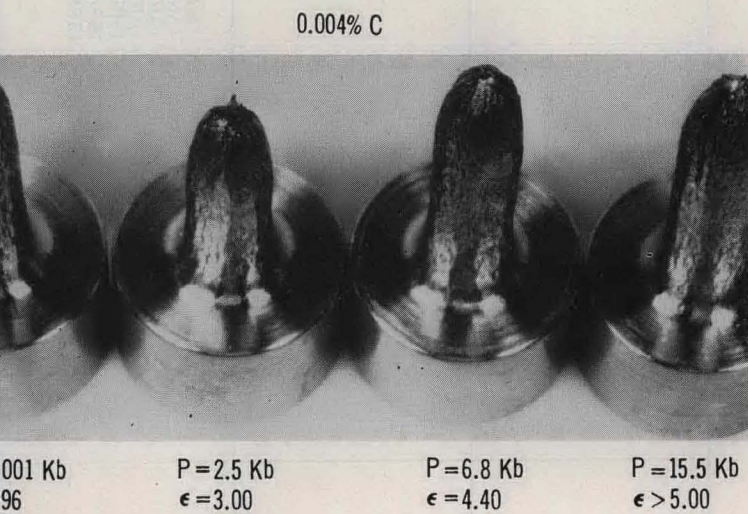


Fig. 7. Macroscopic fracture appearance as function of pressure for annealed Fe-C materials.

pressure retarded the brittle fracture mode, thus initially increasing the amount of uniform strain obtained prior to necking.

Relationship Between Pressure Coefficient of Ductility, β , and Strain Hardening Coefficient, n

The pressure coefficient of ductility, β as a function of the strain hardening coefficient n is shown plotted in Fig. 6 for the materials of this current investigation. It should be noted that in the case of the annealed 0.40, 0.83, and 1.1% C materials, β varied as a function of pressure as was previously discussed. However for comparison purposes, the pressure coefficient of ductility for the best fit linear relationship between strain to fracture and pressure for these three materials are the values plotted in Fig. 6.

As can be seen in Fig. 6, β generally increased with decreasing n . However, the relationship between β and n is not linear, which is in contrast to the proposal by Bridgman (1), and exhibits considerable structure sensitivity. The dependency of this relationship upon microstructure is particularly noticeable in the case of the 0.83 and 1.1% C materials. Whereas the value of β is dependent upon whether these materials are in the annealed or spheroidized condition, n is effectively the same for both conditions. Only if the data for the annealed 0.83 and 1.1% C materials are neglected is a near linear relationship between β and n obtained, as is shown by the solid line in Fig. 6.

It is apparent that the relationship between the pressure coefficient of ductility and strain hardening coefficient is significantly structure sensitive and, based on the materials investigated, is not linear. Apparently then, the considerable scatter in the results of Bridgman, particularly for plain carbon and low alloy steels, is a manifestation of his having a variety of different microstructures.

Effects of Pressure Upon Macroscopic Fracture Appearance

The macroscopic fracture appearance as a function of pressure is shown in Fig. 7 and 8 for the annealed and spheroidized materials, respectively. As can be seen in Fig. 7, the macroscopic fracture appearance for the annealed 0.004% C material changed from a cup-cone at atmospheric pressure to one exhibiting a point or "chisel" point at high pressure. Considerable ribbing also occurred. In the case of the annealed 0.40 and 0.83% C materials, as also seen in Fig. 7, the fracture at atmospheric pressure was cup-cone with the shear portion increasing progressively with increasing pressure at the expense of the fibrous (cup) region. At high pressure, the fracture converted to a planar shear type with the fracture surface being quite flat and oriented approximately along the shear plane. The annealed 1.1% C material exhibited a flat cleavage type of

fracture at atmospheric pressure. With increasing pressure, the fracture progressively converted to a planar shear without any tendency toward a cup-cone type. At the highest pressure (22.8 kbars), the fracture was almost entirely planar shear with a thin peripheral ring of flat or cleavage type fracture. It can also be noted that in the necked region there were numerous fissures on the longitudinal surface.

In the case of the three spheroidized materials, as shown in Fig. 8, the change in macroscopic fracture appearance as a function of pressure was much the same as for the intermediate carbon annealed materials. The fracture exhibited an increase in the shear portion of the cup-cone type fracture with increasing pressure. At high pressure the fracture in all three materials converted to a planar shear type.

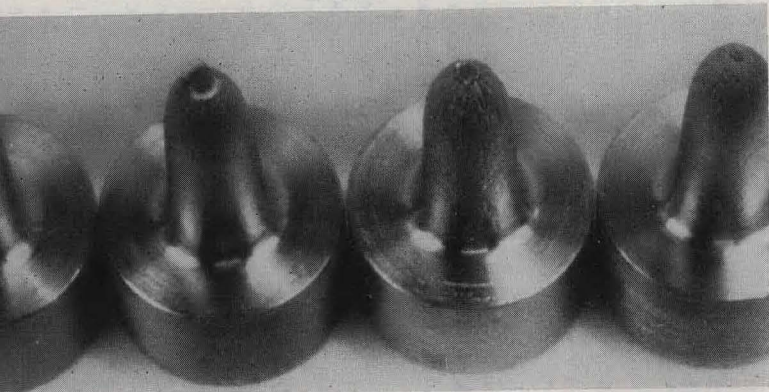
Considering the macroscopic fracture surface appearance, Bridgman reported a linear relationship between pressure and the ratio of tensile (fibrous) to total fracture area. His data, however, exhibited considerable scatter and it is questionable that such a relationship should actually exist.

Plotted in Fig. 9 is the ratio of fibrous to total fracture area as a function of pressure for the materials of this current investigation. The 0.004% C material is not shown, due to the inability of accurately defining the fibrous fracture region because of irregular fracture profile and ribbing. The 1.1% C material is also not shown since, as previously demonstrated, it did not exhibit a cup-cone fracture but went directly to planar shear from a flat cleavage type fracture.

As can be seen in Fig. 9, there is no linear relationship obtained if a best fit curve is used. The scatter would be so great that any attempt toward a linear fit is evidently illogical. As is evident, the curves are of the same general form for all materials. There is an initial rapid decrease in the fracture area ratio with increasing pressure, followed by a leveling off in the range of an area ratio of 0.2 to 0.3. This leveling off, or plateau, which exists over a significant pressure range, is then followed by an abrupt decrease in area ratio over a narrow pressure range in which the area ratio goes to zero. This form of curve is most evident in the annealed and spheroidized 0.83% C material. There is no systematic variation in the form of the curves with respect to material or structure except that the pressure at which the fracture converted completely to planar shear generally increased with increasing strength.

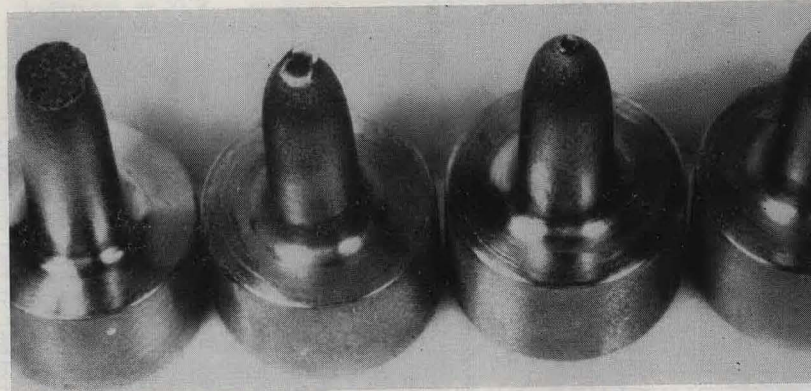
A possible reason for the lack of a continuous relationship between the ratio of fibrous to total fracture area and pressure is that the basic fracture mechanism at atmospheric pressure may be the same for both the fibrous and shear portions of the cup-cone fractures exhibited by these materials. Pressure may, therefore, affect the fracture mode in both of these regions (9). Thus, pressure may not increase the amount of the shear fracture of the type characteristic at atmospheric pressure, but may cause the

0.40% C



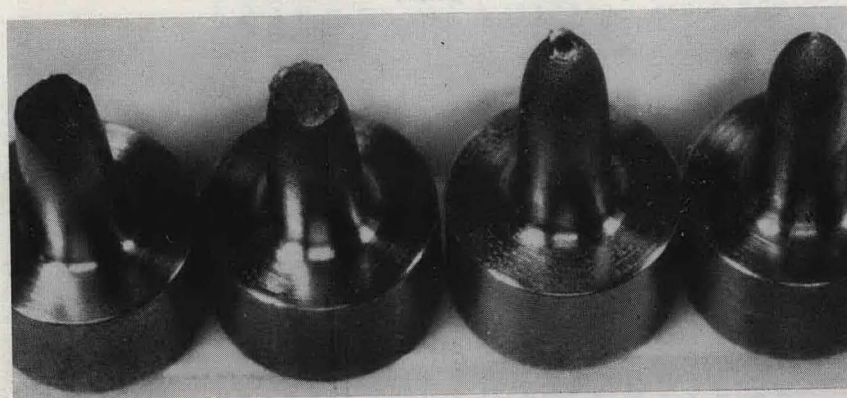
0.001 Kb 33	P=4.0 Kb $\epsilon=2.32$	P=11.5 Kb $\epsilon=4.36$	P=20.9 Kb $\epsilon>5.00$
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0.83% C



P=0.001 Kb $\epsilon=0.68$	P=6.2 Kb $\epsilon=2.06$	P=11.8 Kb $\epsilon=3.25$	P=15.9 Kb $\epsilon=4.25$
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1.1% C



P=0.001 Kb $\epsilon=0.55$	P=2.5 Kb $\epsilon=0.90$	P=9.7 Kb $\epsilon=2.34$	P=21.3 Kb $\epsilon=4.74$
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Fig. 8. Macroscopic fracture appearance as function of pressure in spheroidized Fe - C materials.

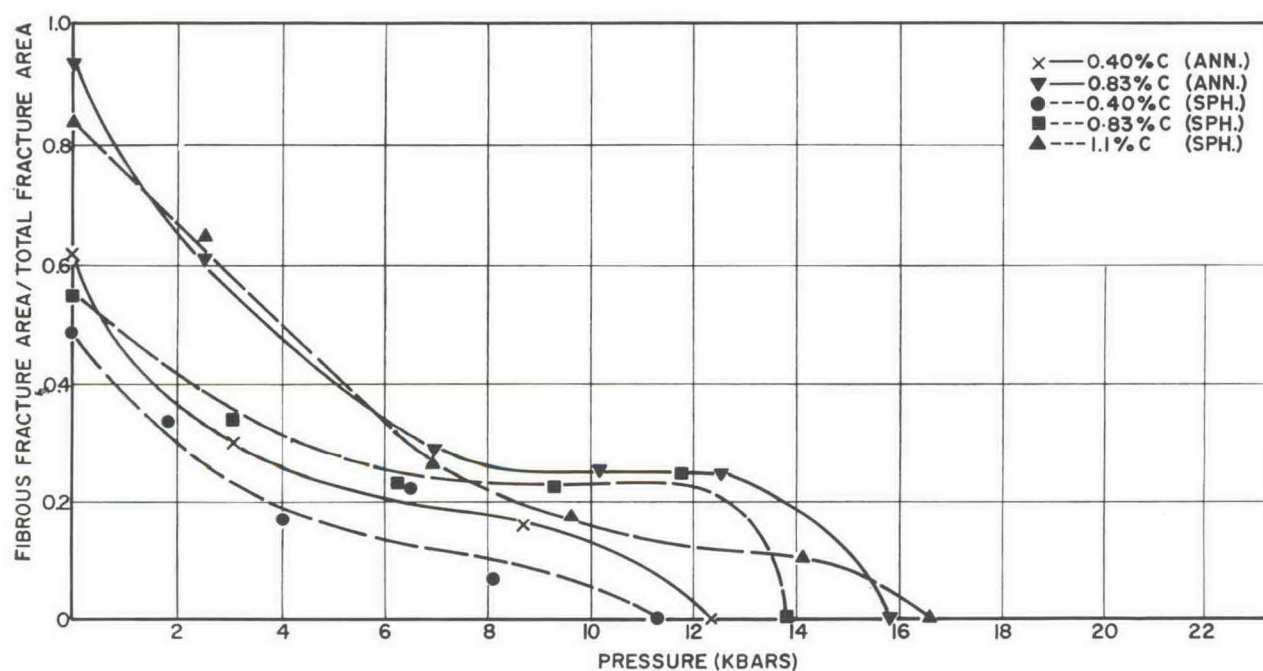


Fig. 9. Fibrous to total fracture area vs pressure for Fe - C materials.

fracture mode to progressively convert to one of planar shear (12). Such a shift in fracture appearance, but is observed by electron fractographic examination.

CONCLUSIONS

1. The effect of pressure upon the ductility of a series of annealed and spheroidized Fe - C materials ranging in carbon content from 0.004 to 1.1% C is highly structure sensitive in terms of the amount, shape and distribution of the cementite phase.

2. Annealed 0.004% C and spheroidized 0.40, 0.83 and 1.1% C materials exhibit a linear relationship between true strain to fracture and pressure with the slope decreasing with increasing carbon content.

3. Annealed 0.40, 0.83 and 1.1% C materials exhibit a polynomial relationship between pressure and true strain to fracture with the slope increasing with increasing pressure. The slope at low pressure and that for the best fit linear relationship for these material decreases with increasing carbon content. The slope at high pressure approaches that for the spheroidized materials.

4. The elongation as a function of pressure increases rapidly at lower pressures, then levels, or tends to level off, at higher pressures in all but the brittle annealed 1.1% C material. The leveling off of the increase in elongation is attributed to the pressure only affecting the area reduction in the necked region and not the uniform strain. In the 1.1% C

material, the elongation progressively increases with no leveling off even at 23.8 kbars. This is a combined effect of an effective increase in uniform strain, since brittle fracture is retarded, plus the contribution of the area reduction in necking.

5. No single or linear relationship exists for the materials investigated between pressure coefficient of ductility and strain hardening coefficient. A linear relationship does exist for the annealed 0.004% C and the spheroidized 0.40, 0.83 and 1.1% C materials, the remaining annealed materials falling far off this linear relationship.

6. No linear or continuous relationship exists between the ratio of fibrous to total fracture area as a function of pressure in those materials exhibiting a cup-cone type fracture at atmospheric pressure. There is a general decrease in the area ratio going to zero for all of the materials at high pressure. The pressure at which the fracture converts entirely to planar shear generally increases with increasing strength.

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REFERENCES

1. P. W. Bridgman, *Studies in Large Plastic Flow and Fracture*, McGraw-Hill, New York (1952).
2. P. W. Bridgman, *The Effects of Pressure on the Tensile Properties of Several Metals and Other Materials*, *J Appl Phys*, 24 (1953) 560.
3. P. W. Bridgman, *The Tensile Properties of Several Special Steels and Certain Other Materials Under Pressure*, *J Appl Phys*, 17 (1946) 201.
4. S. I. Ratner, *Change of Mechanical Properties of Metals Under Hydrostatic Pressure*, *Zh Tekh Fiz*, 19, 3 (1949) 408.
5. B. I. Beresnev, L. F. Vereshchagin, Y. N. Ryabinin, and L. D. Livshits, *Some Problems of Large Plastic Deformation in Metals Under High Pressure*, Pergamon Press (1963).
6. T. Pelczynski, *The Influence of Hydrostatic Pressure on the Plastic Deformation Properties of Metals*, *Arch Hutnictwa*, 7, 1 (1962) 3.
7. L. W. Hu, *Determination of the Plastic Stress-Strain Relations in Tension of Nittany No. 2 Brass Under Hydrostatic Pressure*, *Proceedings of 3rd International Congress of Applied Mechanics* (1958) 557.
8. A. Bobrowsky and E. Stack *Deformation of Metals Under High Pressure*, *Proceedings of AIME Conference on Metallurgy at High Pressure and High Temperature* (1964).
9. T. E. Davidson, J. C. Uy, and A. P. Lee, *The Tensile Fracture Characteristic of Metals Under Hydrostatic Pressures to 23 Kilobars*, *Acta Met*, 14 (1966) 937.
10. H. Ll. D. Pugh, *The Mechanical Properties and Deformation Characteristics of Metals and Alloys Under Pressure*, *Proceedings of International Conference on Materials*, American Society for Testing Materials (1964).
11. H. Ll. D. Pugh, *Recent Developments in Cold Forming*, *Bulleid Memorial Lecture, Vol. III*, University of Nottingham (1965).
12. T. E. Davidson, *The Structure and Stress State Sensitivity of Fracture Mechanisms in Two-phase Materials*, Ph.D. Thesis, Rensselaer Polytechnic Institute, October 1967.

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