

Load-Compression Behavior of Brittle Foams

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Synopsis

Quantitative relationships between the load-compression behavior and the physical characteristics of the foam matrix, previously reported for flexible systems, have now been extended to brittle foams. The shape of the compression curve is expressed in terms of $\psi(\epsilon)$, a dimensionless function of the compressive strain, while the stiffness, or load-bearing capacity, is defined by E_f , the apparent Young's modulus. Because the brittle matrix breaks—rather than flexes—when compressed, a brittle foam exhibits a flatter and wider plateau in the load-compression curve than a rigid (but ductile) foam of equivalent density, cell geometry, and E_f . These differences are expressed quantitatively by $\psi(\epsilon)$. It is important to distinguish between brittle foams and rigid, but ductile, foams. Since both types may exhibit the same stiffness, this distinction, particularly significant in energy absorbing applications, often is not considered in designing foam structures. Using the relationships established in this report, it is now possible to delineate precisely the characteristics a brittle foam must possess to meet a given load-compression specification.

INTRODUCTION

The load-compression behavior of a foamed polymer, important for design and performance characteristics of a foam, is dependent upon the geometric structure and physical properties of the matrix polymer. To specify the optimum foam for a given application, it is important to understand, in a quantitative manner, the relationship between compressive stress and the physical characteristics of the matrix. It was demonstrated previously¹ that the compressive stress, σ , can be factored into the product of (1) a dimensionless function of the compressive strain, $\psi(\epsilon)$, and (2) a factor ϵE_f , where ϵ is the compressive strain and E_f the apparent Young's modulus of the foam. This is expressed as follows:

$$\sigma = \epsilon E_f \psi(\epsilon). \quad (1)$$

The foam modulus, E_f , is a function primarily of Young's modulus of the matrix polymer, E_0 , and volume fraction of polymer, φ :²

$$E_f/E_0 = [\varphi(2 + 7\varphi + 3\varphi^2)]/12 \quad (2)$$

Experimental compression data for several flexible polyurethane foams demonstrated¹ that $\psi(\epsilon)$ is highly sensitive to the specific details of the matrix geometry, only moderately dependent on density or cell size, and

independent of E_0 (and hence independent of temperature or strain rate). In this report, the analysis of compression behavior is extended to other foam systems, with particular emphasis placed on the influence of matrix brittleness on $\psi(\epsilon)$.

The function $\psi(\epsilon)$ reflects the load-bearing capacity of the collapsed matrix and hence must depend on the strength and brittleness of the matrix. A brittle matrix is broken, rather than flexed, during compression, leading to a load-compression curve with a flatter and wider plateau than that characteristic of a flexible matrix with a similar geometric structure. The magnitude of the influence of matrix brittleness on compressive behavior, however, has never been evaluated quantitatively.

In this report, $\psi(\epsilon)$ is calculated from experimental load-compression curves for several rigid (brittle) polyurethane foams, and compared to that calculated from data for flexible polyurethane foams of similar matrix geometry; $\psi(\epsilon)$ also is calculated from compression data for ABS, phenolic, polyethylene, and polystyrene foams. The critical features of $\psi(\epsilon)$ will be shown to vary in a regular manner, reflecting the geometric and physical properties of the matrix polymer.

EXPERIMENTAL

Rigid polyurethane foams generally possess a different matrix geometry than flexible polyurethane foams because of differences in the foaming reaction. To evaluate the influence of matrix brittleness without altering matrix geometry, several polyurethane foams (flexible at 25°C) (Mobay Chemical Company and Scott Paper Company) were compressed at -196° (immersed in liquid nitrogen), a temperature at which the matrix is brittle, and the data were compared to those obtained at 25°C. Load-compression

TABLE I
Physical Characteristics of Polyurethane Foams

Sam- ple	Type	E_f , psi			ϕ	Cell size, in.	
		+25°C	-60°C	-196°C		av.	95% range
B	I ^a	77.	—	—	0.24	0.007	0.001-0.02
C	I	25.	350.	1500.	0.11	0.012	0.003-0.03
D	I	11.5	150.	770.	0.065	0.020	0.007-0.05
E	I	4.5	105.	400.	0.043	0.022	0.008-0.06
F	I	3.5	89.	350.	0.037	0.025	0.01-0.06
G	II ^b	15.	380.	750.	0.033	0.020	0.01-0.04
H	II	9.0	220.	530.	0.028	0.013	0.005-0.025
L	III ^c	18.	400.	450.	0.028	0.10	0.04-0.2
M	III	14.	280.	400.	0.028	0.050	0.02-0.1
N	III	9.4	210.	280.	0.028	0.035	0.01-0.08
Q	III	11.	320.	660.	0.032	0.012	0.005-0.02

^a Normal polyurethane foam with an irregular cell structure.

^b Normal polyurethane foam with a highly regular cell structure.

^c Reticulated polyurethane foam with a regular cell structure.

data also were obtained at -60°C , an intermediate temperature at which the matrix is "rigid" but still ductile. Table I summarizes the physical characteristics of these foams. The compression data at 25°C , together with photomicrographs showing the differences in cell structure, were reported previously, and the foam-sample designations in Table I correspond to those used in the earlier article.¹

Compression data also were obtained for a series of rigid (at room temperature) polyurethane foams (Mobay Chemical Company) varying in density from $\varphi = 0.037$ to 0.79 and several other rigid foams of varying matrix characteristics (Table II). The foams examined were (1) high-density ABS (acrylonitrile-butadiene-styrene) from expandable beads (Marbon Chemical Company), (2) low-density polystyrene from expandable beads (Sinclair-Koppers Company), (3) low-density polyethylene extruded sheet (Dow Chemical Company), and (4) low-density phenolic (Union Carbide Company).

Rigid foams are frequently completely closed cell (cells not interconnected). Air is unable to escape from these closed cells and is compressed as the foam matrix is compressed. The contribution of entrapped air to the compressive stress can be approximated by modifying eq. (1):

$$\sigma = \epsilon E_f \psi(\epsilon) + Pf\epsilon/(1 - \varphi - \epsilon) \quad (3)$$

where P is the atmospheric pressure and f , the volume fraction of closed cells. The latter term in eq. (3) is negligible if (1) the stiffness of the matrix is much greater than the apparent stiffness of the compressed air ($E_f \gg P$), (2) the foam is open cell ($f \approx 0$), or (3) the matrix is very brittle, such that the closed cells are broken during compression (releasing trapped air). The contribution of compressed air is most significant for a material

TABLE II
Physical Characteristics of Rigid Foams

Sample	Type	E_f , psi	φ	Av. cell size, in.
S	Polyurethane	740	0.037	0.020
T	Polyurethane	1,100	0.078	0.015
U	Polyurethane	3,800	0.13	0.010
V	Polyurethane	17,500	0.30	0.010
W	Polyurethane	32,000	0.41	0.008
X	Polyurethane	56,000	0.54	0.006
Y	Polyurethane	140,000	0.79	0.005
Z	Polystyrene	370	0.020	0.003
AA	Polyethylene	110	0.033	0.040
BB	Polyethylene	250	0.054	0.030
CC	Polyethylene	1,200	0.13	0.025
DD	ABS	42,000	0.44	0.006
EE	ABS	56,000	0.51	0.005
FF	ABS	80,000	0.62	0.005
GG	Phenolic	1,050	0.032	0.003

such as low-density polyethylene foam where the matrix is ductile, $E_f \approx P$, and $f \approx 1$.

The function $\psi(\epsilon)$ is slightly dependent on the bulk dimensions of the foam test piece. In the present work, the test pieces measured $2 \times 2 \times 2$ in., $2 \times 4 \times 4$ in. (compressed in the 2-in. direction), and $4 \times 2 \times 2$ in. (compressed in the 4-in. direction). The h/w ratio, where h and w are the dimensions parallel and perpendicular, respectively, to the compression direction, is indicated for all data.

Unless noted otherwise, the compression data presented were obtained at room temperature and a linear strain rate of about 0.5%/sec; the data are characteristic of the first compression only.

For the rigid polyurethane foams (samples S through Y), E_f was measured in tension and compression. The values were similar in all cases, confirming the linearity of the stress-strain curve through $\epsilon = 0$. An identical result was reported previously¹ for flexible foams.

RESULTS AND DISCUSSION

Effect of Matrix Brittleness on $\psi(\epsilon)$

As a foam is compressed to small strains, the struts comprising the matrix bend and compress in an elastic manner ($\psi(\epsilon) = 1.0$). Then, at some characteristic compression, the matrix structure begins to buckle, or collapse, and $\psi(\epsilon)$ starts to decrease. The strain at $\psi(\epsilon) = 0.95$ has been defined¹ the "critical buckling strain," ϵ_b . The point at which the struts buckle and the load-bearing capacity of the collapsed structure are dependent on the flexural strength and ductility of the struts. If a portion of the matrix structure breaks brittlely, the compressed structure will support a lower load than an equivalent structure which does not fracture during compression. This effect is reflected in the shape of the $\psi(\epsilon)$ function.

Earlier work¹ demonstrated that $\psi(\epsilon)$ is independent of E_0 , the modulus of the matrix polymer; and, in the absence of a ductile-brittle transition, $\psi(\epsilon)$ thus is independent of temperature. The influence of changes in matrix brittleness on $\psi(\epsilon)$ is presented in Figures 1-4, which show the $\psi(\epsilon)$ functions calculated from compression data obtained at 25°, -60°, and -196°C for the polyurethane foams listed in Table I. These foams are flexible at 25°C (exhibit 100% recovery after compression), partially flexible at -60° (exhibit about 50% to 80% recovery), and brittle at -196°C (exhibit negligible recovery). The changes in $\psi(\epsilon)$ accompanying the decrease in temperature are qualitatively similar in all cases and are attributed to increased matrix brittleness only.

The decrease in $\psi(\epsilon)_{\min}$ (minimum value) and the increase in ϵ at $\psi(\epsilon)_{\min}$ are clearly related to increased matrix fracture during compression. The point at which no further collapse of the matrix is possible, because of lateral matrix constraints (perpendicular to applied stress), is represented

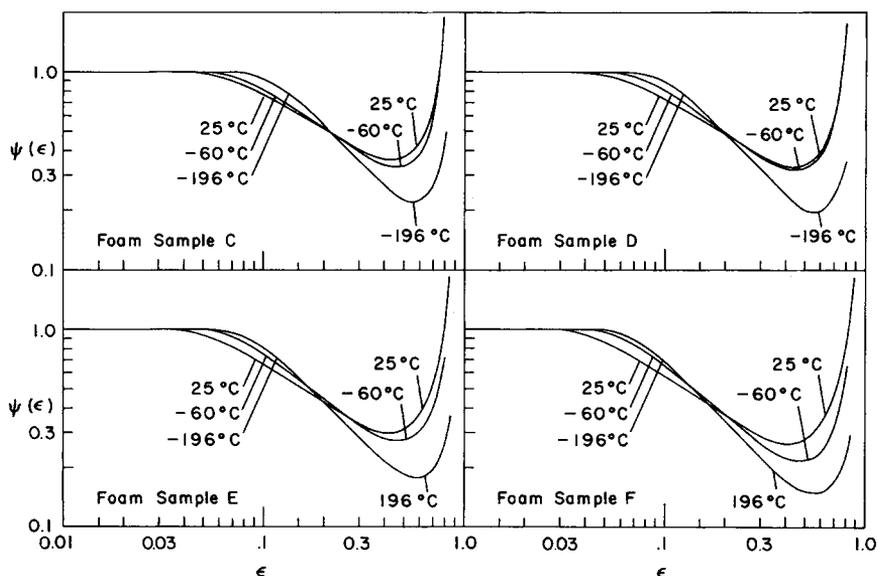


Fig. 1. Dependence of $\psi(\epsilon)$ on matrix brittleness for several Type-I polyurethane foams ($h/w = 1/2$).

by $\psi(\epsilon)_{\min}$. If the degree of matrix fracture is increased, the lateral constraints at large compressions become relatively less significant and the load-bearing capacity is decreased. Thus, increasing matrix brittleness leads to a wider and flatter plateau in the load-compression curve.

Since the qualitative effects of matrix brittleness are the same for all of these polyurethane foams, it can be inferred that the relationships between load-compression behavior and physical characteristics of the matrix are qualitatively the same for brittle foams as those established previously for flexible foams.¹ That is, $\psi(\epsilon)$ for a brittle foam should be highly sensitive to the details of the matrix geometry, only moderately dependent on density or cell size, and independent of E_0 (within the region of brittle behavior).

Density Dependence of $\psi(\epsilon)$

Figures 1 and 5 compare $\psi(\epsilon)$ for four type-I polyurethane foams varying in density from $\varphi = 0.037$ to 0.11. For brittle foams, as in the case of flexible foams, the dependence of $\psi(\epsilon)$ on φ is relatively small. This is confirmed by the data in Figure 6, where $\psi(\epsilon)$ is shown for a series of rigid polyurethane foams (Table II) of similar cell geometry. Increasing φ from 0.037 to 0.79 only doubles $\psi(\epsilon)_{\min}$, while E_f (and hence the compressive stress) increases by a factor of 190. For this particular set of data (Fig. 6), the following proportionality can be written:

$$\psi(\epsilon)_{\min} \propto \varphi^{0.20}. \quad (4)$$

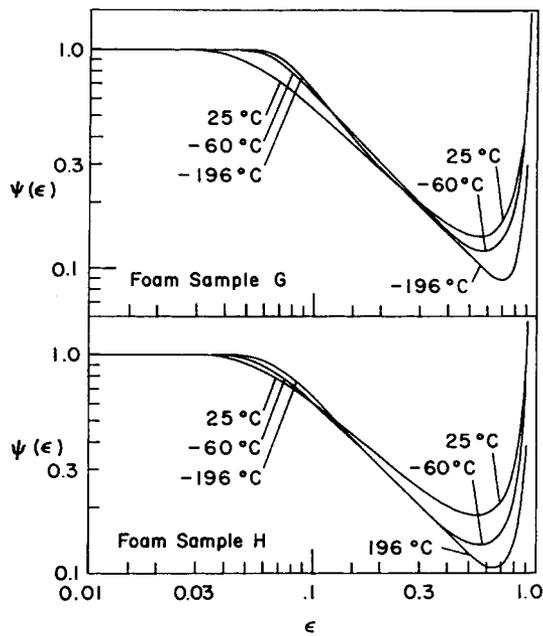


Fig. 2. Dependence of $\psi(\epsilon)$ on matrix brittleness for two low-density type-II polyurethane foams ($h/w = 1/2$).

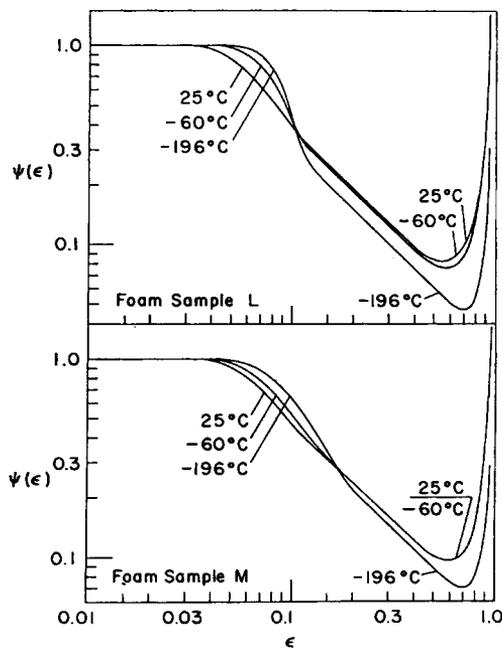


Fig. 3. Dependence of $\psi(\epsilon)$ on matrix brittleness for two low-density type-III polyurethane foams ($h/w = 1/2$).

In the case of flexible polyurethane foams,¹ $\psi(\epsilon)$ approaches unity as ϵ approaches $(1 - \phi)$. This is not observed for brittle foams (Fig. 6), where $\psi(\epsilon)$ is still much less than unity at $\epsilon = (1 - \phi)$. The low values of $\psi(\epsilon)$ at large compressions (brittle foams) reflect brittle fracture of the matrix.

Since changes in $\psi(\epsilon)$ are small compared to those in E_f , the stress required to compress a rigid foam to some given strain, ϵ^* , relative to that

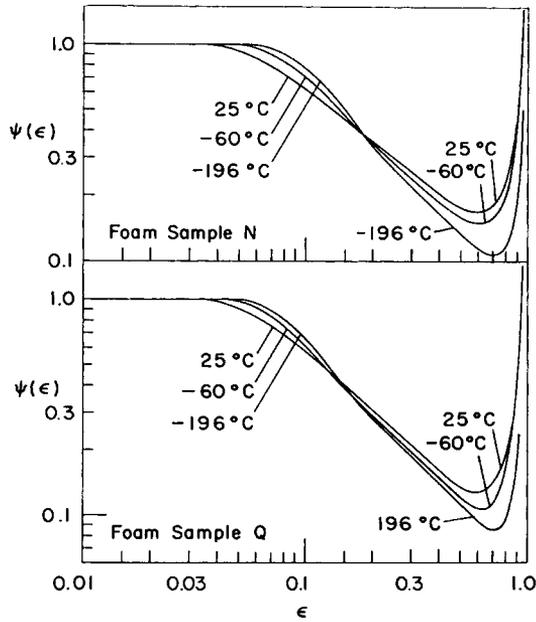


Fig. 4. Dependence of $\psi(\epsilon)$ on matrix brittleness for two low-density type-III polyurethane foams ($h/w = 1/2$).

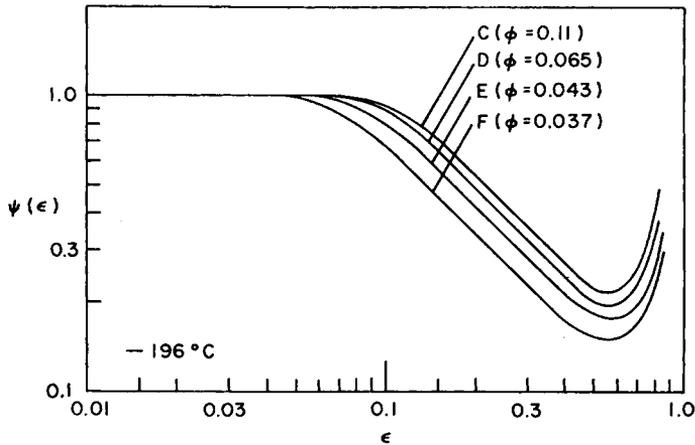


Fig. 5. Dependence of $\psi(\epsilon)$ on volume fraction of polymer for brittle type-I polyurethane foams ($h/w = 1/2$).

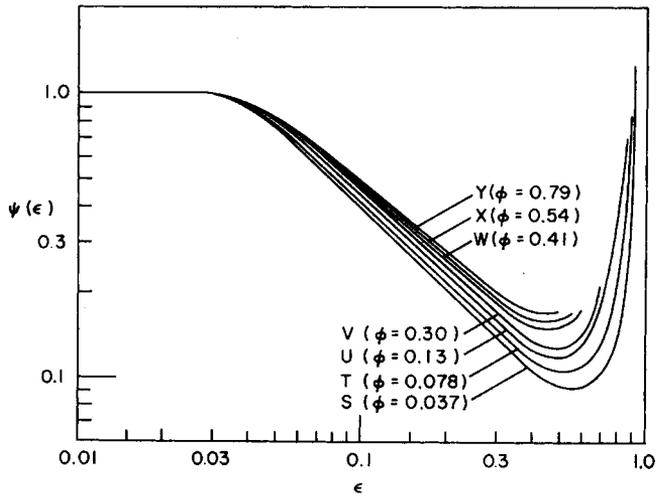


Fig. 6. Dependence of $\psi(\epsilon)$ on volume fraction of polymer for several rigid polyurethane foams ($h/w = 1$).

required to compress another similar foam, is approximately equal to the ratio of moduli:

$$\sigma(\epsilon^*)_1/\sigma(\epsilon^*)_2 \approx (E_f)_1/(E_f)_2. \quad (5)$$

This relationship is valid at any value of ϵ , provided the foams are of similar matrix geometry. The compressive stress, or load-bearing capacity, generally is related to density rather than E_f . But it is important to remember that $\sigma(\epsilon^*)$ can be related to φ only if E_v is assumed constant and cell geometry changes are small.

The density dependence of compressive strength frequently is expressed in the form^{3,4}

$$\sigma(\epsilon^*) \propto \varphi^b \quad (6)$$

where b is a constant. The value of b can be calculated by inserting eq. (2) in eq. (5). In the range $\varphi = 0.08$ to 0.5, eq. (2) is approximated by

$$E_f/E_0 \approx 0.70 \varphi^{1.5}. \quad (7)$$

Therefore we obtain $b = 1.5$; values between 1.4 and 1.6 have been reported for b in this density range.^{3,4}

Dependence of $\psi(\epsilon)$ on Matrix Geometry

The influence of the average cell-size, d , on $\psi(\epsilon)$ for a series of reticulated polyurethane foams ($d = 0.012$ to 0.10) is shown in Figures 3, 4, and 7. The general effect of increasing d is to decrease $\psi(\epsilon)_{\min}$ and to accentuate the maximum in the load-compression curve in the vicinity of 5%–10% compression. The relative position of $\psi(\epsilon)$ for sample N, however, indicates that other geometric features of the matrix, such as cell-size dis-

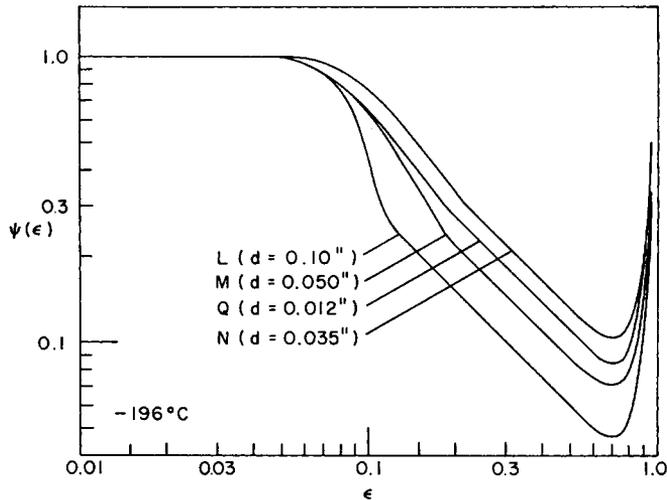


Fig. 7. Dependence of $\psi(\epsilon)$ on average cell size for brittle, low-density, type-III polyurethane foams ($h/w = 1/2$).

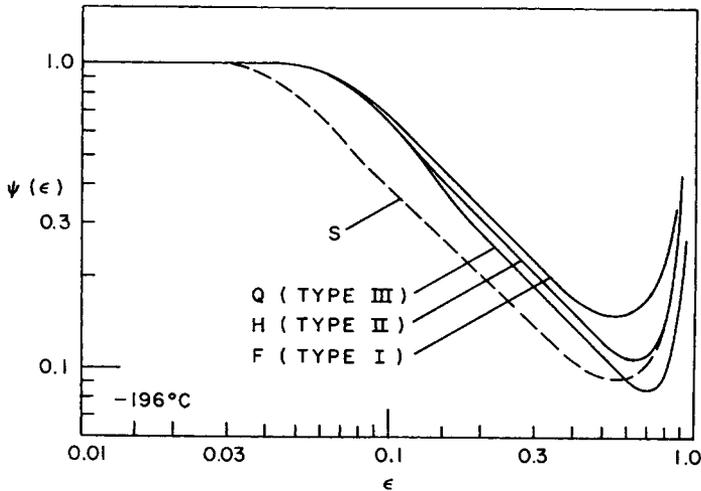


Fig. 8. Dependence of $\psi(\epsilon)$ on matrix geometry for brittle, low-density polyurethane foams ($h/w = 1/2$). For sample S, $\psi(\epsilon)$ was calculated from data at 25°C , at which temperature the foam is rigid; it is assumed that $\psi(\epsilon)$ at 25°C is essentially identical to that at -196°C .

tribution, may be much more significant than d in determining $\psi(\epsilon)$. The geometric structure of the matrix is difficult to describe quantitatively, but can be defined qualitatively in terms of a comparison to a matrix of known geometry. The marked influence of the details of the matrix geometry on $\psi(\epsilon)$ is further demonstrated (Fig. 8) when $\psi(\epsilon)$ is compared for rigid polyurethane foams of similar density ($\varphi \approx 0.03$) and cell size ($d \approx 0.01$ in.) but different cell structure. Photomicrographs showing the

differences in cell structure between samples F, and H, and Q have been published previously.¹

While $\psi(\epsilon)$ is highly dependent on matrix geometry, the ratio E_f/E_0 is rather insensitive to geometry and depends primarily on ϕ , eq. (2). Therefore, if the shape of the load-compression curve is unacceptable for a

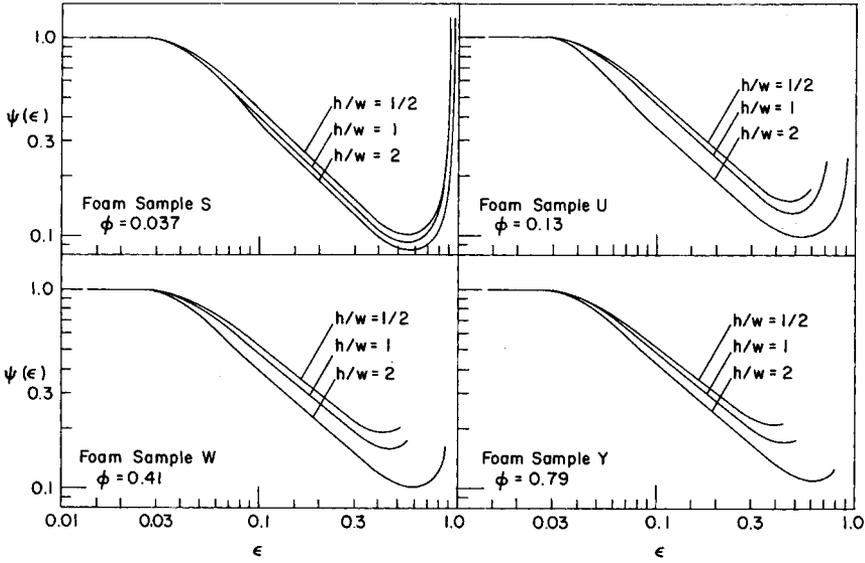


Fig. 9. Dependence of $\psi(\epsilon)$ on h/w for several rigid polyurethane foams.

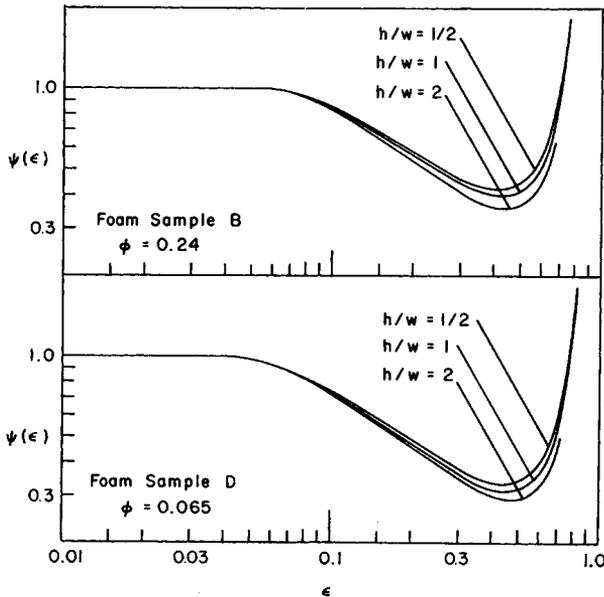


Fig. 10. Dependence of $\psi(\epsilon)$ on h/w for several flexible polyurethane foams.

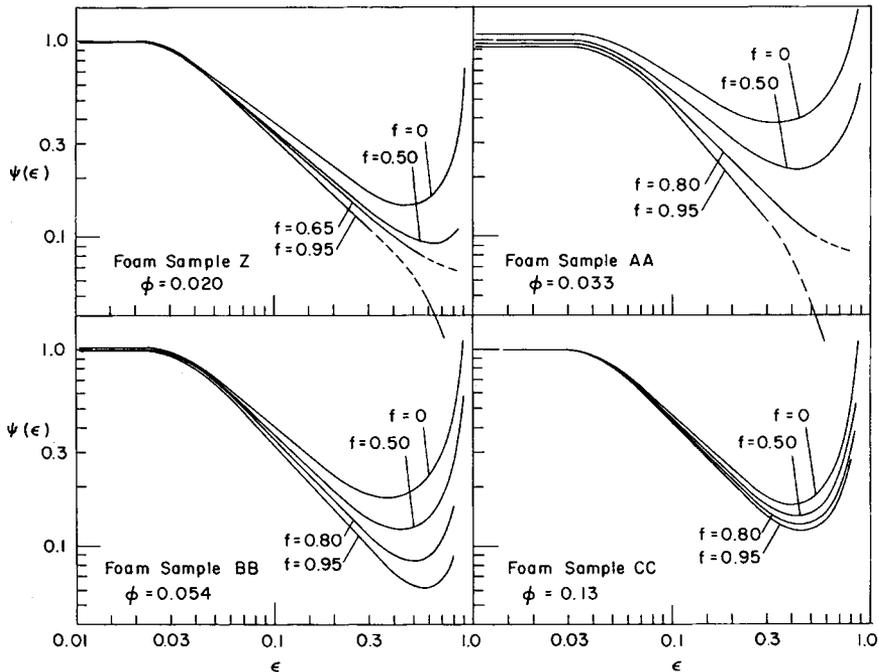


Fig. 11. Approximate $\psi(\epsilon)$ functions for low-density polyethylene and polystyrene foams ($h/w = 1$) calculated from eq. (3). The volume fraction of closed cells, f , is assumed independent of compression.

particular application, the matrix geometry must be altered; but if the shape is acceptable while the stiffness is not, then φ or E_0 must be altered.

Influence of h/w Ratio on $\psi(\epsilon)$

As the foam is compressed, lateral constraints—contribution to compressive stress resulting from deformation (bulging) perpendicular to applied stress—become increasingly more significant. Qualitatively, these lateral constraints would be expected to increase as h/w decreases, analogous to the uniaxial compression of a solid rubber block.^b This is reflected as an increase in $\psi(\epsilon)_{\min}$ and a decrease in ϵ at $\psi(\epsilon)_{\min}$. An experimental evaluation of the influence of h/w on $\psi(\epsilon)$ ($h/w = 1/2, 1,$ and 2) is presented in Figure 9 for four rigid polyurethane foams varying in density from $\varphi = 0.037$ to 0.79 , and in Figure 10 for two flexible polyurethane foams with $\varphi = 0.065$ and 0.24 . The flexible and low-density rigid foams exhibit a smaller dependence on h/w , $\psi(\epsilon)_{\min}$ decreasing only about 15% as h/w is increased from $1/2$ to 2 , than the high-density rigid foams. This difference is a result of the brittleness of the matrix. A high-density ABS foam ($\varphi = 0.51$), which has a ductile matrix, exhibited less than a 10% decrease in $\psi(\epsilon)_{\min}$ as h/w was increased from 1 to 2 .

Other Rigid Foams

Although data presented thus far have been for polyurethane foams only, the conclusions reached should be valid for any foamed material. Figure 11 shows $\psi(\epsilon)$ functions for three polyethylene foams ($\phi = 0.033, 0.054,$ and 0.13) and a polystyrene foam ($\phi = 0.020$). For these, the contribution of entrapped air to compressive stress is significant, and $\psi(\epsilon)$ was calculated from eq. (3), assuming $f = 0.5, 0.8,$ and 0.95 . This separation of the contributions from compressed air and matrix geometry is only approximate, however, since f continually decreases during compression because of rupture of the closed cells. The curves for $f = 0.5$ are considered representative of an open-cell matrix of equivalent geometry. Figure 12 shows $\psi(\epsilon)$ for a low-density phenolic foam; because of the brittle matrix, entrapped air does not contribute to the compressive stress of this foam. The very low ϵ_b and $\psi(\epsilon)_{\min}$ observed for phenolic foams reflect the highly brittle character of the phenolic matrix.

The $\psi(\epsilon)$ functions for three high-density ABS foams ($\phi = 0.44, 0.51,$ and 0.62) are shown in Figure 12. These tend to resemble the $\psi(\epsilon)$ functions obtained for flexible¹ rather than brittle high-density polyurethane foams. Particularly, $\psi(\epsilon)$ approaches unity as ϵ approaches $(1 - \phi)$, indicating that the matrix is not breaking brittlely during compression, although it is permanently deformed. An ABS foam will exhibit a narrower and steeper plateau in the load-compression curve than a brittle foam (such as polyurethane) of equivalent ϕ and E_0 . Comparing ABS and brittle polyurethane foams of identical density and compressive modulus, the ABS foam will possess a higher compressive strength (for example, at 25% compression), but the polyurethane foam will display superior energy-

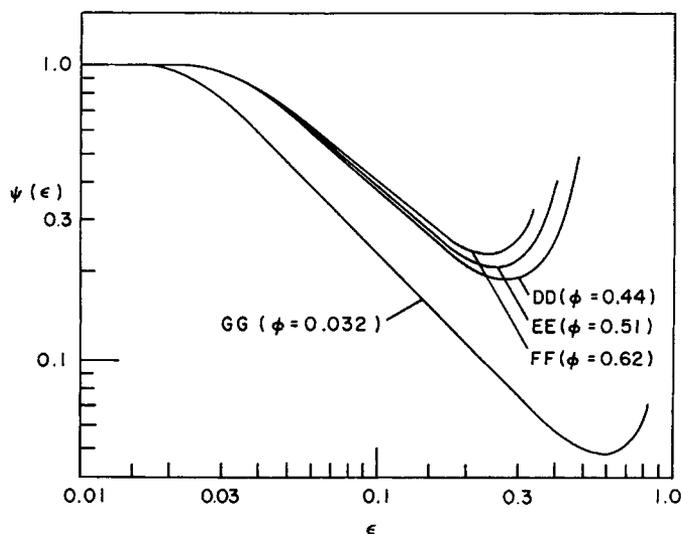


Fig. 12. Dependence of $\psi(\epsilon)$ on matrix geometry for ABS and phenolic foams ($h/w = 1$).

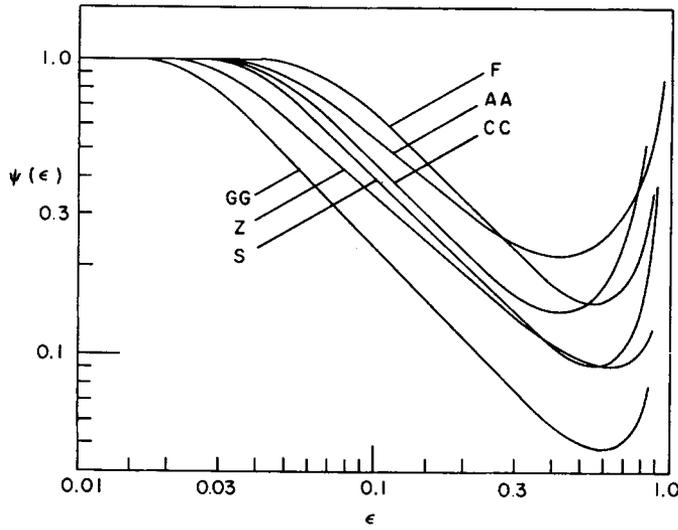


Fig. 13. Dependence of $\psi(\epsilon)$ on matrix geometry for several low-density foams ($h/w = 1$).

absorbing characteristics. Thus, the shape of the load-compression curve, which is expressed quantitatively in terms of $\psi(\epsilon)$, can significantly influence the performance characteristics of a foam structure.

Figure 13 provides an additional comparison of the variation in $\psi(\epsilon)$ which results from differences in matrix geometry and brittleness. Foams commonly are classified as "flexible" and "rigid" (rigid denoting $E_0 > 10^4$ psi); but, since $\psi(\epsilon)$ is dependent on the brittleness of the matrix, and not on the modulus E_0 , it would be better to differentiate the compressive behavior as "flexible" and "brittle." An ABS or polyethylene foam, for example, although rigid, exhibits a $\psi(\epsilon)$ function which resembles a flexible rather than a brittle foam.

CONCLUSIONS

The compressive stress can be factored into the product of (1) a dimensionless function of the compressive strain, $\psi(\epsilon)$, reflecting the buckling of the matrix, which describes the shape of the compression curve and (2) a factor ϵE_f , where E_f is the apparent Young's modulus, which expresses the stiffness of the foam. The $\psi(\epsilon)$ functions were calculated from experimental compression data for several rigid (brittle and ductile) foams and compared to those previously calculated from data for flexible polyurethane foams with the following results:

1. The brittleness of the foam matrix has a significant effect on $\psi(\epsilon)$. Because a brittle matrix is broken rather than flexed during compression, a brittle foam exhibits a load-compression curve with a flatter and wider plateau than that displayed by a ductile foam of equivalent Young's modulus, density, and matrix geometry.

2. It is important to distinguish between brittle foams (exhibiting glassy fracture) and "rigid" foams (exhibiting ductile fracture). Since both types appear equally stiff, this distinction, particularly important in structural or energy-absorbing applications, frequently is not considered when designing foam components.

3. The critical features of $\psi(\epsilon)$ are highly sensitive to the geometric structure and brittleness of the matrix, slightly dependent on foam density, cell size, bulk dimensions, and volume fraction of closed cells, and independent of the modulus of the matrix polymer.

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References

1. K. C. Rusch, *J. Appl. Polym. Sci.*, **13**, 2297 (1969).
2. K. C. Rusch, unpublished work.
3. T. H. Ferrigno, *Rigid Plastics Foams*, 2nd ed., Reinhold, New York, 1967.
4. R. K. Traeger, *J. Cell. Plast.*, **3**, 406 (1967).
5. A. N. Gent and P. B. Lindley, *Proc. Inst. Mech. Eng.*, **173**, 111 (1959).

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