SSC-126

INFLUENCE OF HOT-ROLLING CONDITIONS ON BRITTLE FRACTURE IN STEEL PLATE

by

F. de Kazinczy

and

W. A. Backofen

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November 10, 1960

Dear Sir:

Herewith is a copy of SSC-126 entitled Influence of Hot-Rolling Conditions on Brittle Fracture in Steel Plate by F. de Kazinczy and W. A. Backofen. This is the first progress report of a project sponsored by the Ship Structure Committee at the Massachusetts Institute of Technology to determine the relationship of mill-rolling practice to metallurgical structure and properties of ship plate.

The work has been conducted under the advisory guidance of the Committee on Ship Steel of the National Academy of Sciences-National Research Council.

This report is being distributed to individuals and groups associated with and interested in the work of the Ship Structure Committee. Please submit any comment that you may have to the Secretary, Ship Structure Committee.

Sincerely yours,

E. H. THIELE
Rear Admiral, U. S. Coast Guard
Chairman, Ship Structure Committee

Serial No. SSC-126

First Progress Report
of
Project SR-147

to the

SHIP STRUCTURE COMMITTEE

on

INFLUENCE OF HOT-ROLLING CONDITIONS ON BRITTLE FRACTURE IN STEEL PLATE

by

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ABSTRACT

Steel plates processed according to a conventional and controlled (low finishing temperature) rolling practice were studied to establish reasons for a superior notch toughness in the controlled-rolled product. The lower transition temperature (Charpy V-notch 15 ft-lb) in plates investigated is derived largely from a smaller ferrite grain size. Experimental evidence was also obtained to indicate that a part of the improvement results from a microfissuring in the plane of the plate at the notch root, with the effect that stress triaxiality is relieved and transition temperature depressed. The origin of the flaws responsible for the fissures was not determined with certainty, but there was good indication that they were simply inclusions in a fiber structure too fine in scale for observation with normal metallographic techniques. The criteria presented for microfissuring were: (1) a flaw structure dispersed on a scale no greater than the size of the plastic volume from which the brittle crack originated; (2) a ratio of the critical fracture stress in thickness (Z) to rolling (R) direction no greater than about 1/2. The necessary fracturing anisotropy was favored by more intense fibering (lower σ_Z) and finer grain size (higher σ_R). The criteria were most nearly satisfied by controlled-rolled plate. Correlations between σ_Z and visible inclusion content showed the decrease in σ_Z to be paralleled by more elongated inclusions, as if visible changes are indicative of changes in the fine-scale fiber structure.

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INTRODUCTION

In general, processing materials by plastic deformation produces structural changes and, therefore, changes in mechanical properties. An example of interest is the hot rolling of steel at lower-than-normal temperatures, in the austenite and ferrite region. Such processing finds some use in the production of ship plate, and, as practiced with regulated temperature-reduction programs, has been termed controlled rolling. Improved notch toughness is attributed to controlled-rolled plate,[1] yet little in the way of a basic explanation for the improvement has been given in the literature.

Various structural differences could contribute to the superior properties. Finer grain size from lower finishing temperature would be expected to be one of the most important. Several investigations have been made of the dependence of transition temperature on ferrite grain size. Barring complication from subgrain or Widmanstätten structure,[2-5] the decrease is 8-15 C per ASTM grain-size number.[4-7]

A more strongly developed fibering or laminated condition is another difference that might contribute to improved notch toughness. Mechanical fibering commonly associated with ferrite banding and alignment in the direction of working of inclusions or other phases is the type to be considered. Crystallographic fibering or preferred orientation as found in hot-rolled plate should be much less significant. Because of this "lamination," properties relating to fracture are generally lower when measured across the fiber; as a common example, resistance to fracture over a surface in the plane of a rolled plate is practically always the least, with the greatest resistance being measured along the rolling direction and an intermediate level in the transverse direction. The more obvious fiber elements are reasonably expected to contribute to a fracturing anisotropy.[8-12] However, in work by Backofen et al,[13,14] fracturing anisotropy was clearly demonstrated in a variety of relatively pure metals and single-phase alloys. It was concluded that wrought metals in general are characterized by a structure of highly aligned submicroscopic and crack-like flaws. The detailed nature of the structure has not

been established, but one reasonable view is that the flaws are simply very small inclusions which are not detected by usual metallographic examination.

The presence of a fiber structure might be expected to depress transition temperature only when fracture occurs in the presence of triaxial stress with cracking across the fiber. Then, as suggested by Soete[15] and Matton-Sjöberg,[16] tensile stress over the weakest plane may become sufficiently high to produce cracking or fissuring at right angles to the main crack path. The result would be some relief of the triaxiality, which acts to elevate transition temperature.

A number of experiments have been made in the past relating to this view. Matton-Sjöberg[16] compared two fine-grain steels and found a lower transition temperature consistent with narrower ferrite banding, although he pointed out at the same time the difficulty of separating such an effect from one caused by slag inclusions. Matton-Sjöberg and others[17,18] as well have also established a dependence of transition temperature on the orientation of specimen and notch in a given plate. Different results were obtained, however, in a study by Owen, Cohen, and Averbach[19] with a coarse-grain steel showing broad ferrite banding. In that work, the Charpy-V 15 ft-lb transition temperature was insensitive to banding, and there was no evidence of anisotropy in this property. Inconclusive experiments by Mangio and Boulger[20] give another example of attention focused only on coarse-scale structure; these followed a suggestion from Orowan[21] that transition temperature might be lowered by artificially generating weak planes parallel to the rolled surface. Accordingly, plates were prepared by pack-rolling oxidized sheets and by intermittent casting to produce oxidized layers in the ingot.

A preoccupation with coarser details of structure has been common to all such work. As a result, the current experimental basis for appraising any effect of fissure formation on transition temperature still seems quite inadequate. For elements of fiber to be at all effective, they must be sufficiently small and finely distributed to compare in size with the small plastic volume beneath a notch or crack in which the brittle failure originates. The fine-scale flaw struc-

ture referenced previously, and generally characteristic of wrought material, would appear able to satisfy this requirement; the larger and more obvious inclusions and regions of banding would not. It was the principal aim of this research to separate, if possible, these effects of grain-size and fibering on ductile-to-brittle transition. In this way it was hoped that an explanation could be found for all of the improved notch toughness to be expected of "controlled-rolled" plate.

MATERIALS

Source: Plates were produced by the Royal Netherlands Blast Furnace and Steel Works, Ltd. Two different analyses were included; these are given in Table I.

TABLE I. CHEMISTRY AND DESIGNATION OF TEST MATERIALS

Designation	C	Mn	Si	P	S
H, h	0.15	1.18	0.03	0.017	0.026
L, l	0.19	0.74	0.04	0.017	0.028

H and L were 1-1/2 in. thick; h and l were 3/4-in. thick. Taking adjacent ingots from the same charge, one was rolled in a conventional manner (designated S), while the other was controlled-rolled (designated C) with the last 30-35% of the reduction in the two-phase region and with a finishing temperature of 720 C. The detailed rolling schedule is shown in the Appendix. Most tests were carried out on HS- and HC- plates.

Metallography: Ferrite grain size was determined by linear analysis in both the rolling (R) and thickness (Z) directions. Since the grains were not equiaxed, the average grain diameter has been represented as $\bar{d} = \sqrt{d_R d_Z}$. The banding ratio was defined as the ratio of free ferrite path in the rolling direction to that in the thickness direction. Care was taken to compute the ratio only

after sufficient counting had taken place to insure that the percentage of ferrite (or pearlite) was the same (within 0.5%) in both directions. With this procedure, the ferrite and pearlite banding ratios are identical.

Results of the grain-size determinations are given in Table II.

TABLE II. FERRITE GRAIN SIZE FROM LINEAL ANALYSIS

	HC	HS	hC	hS	LC	LS	lC	lS
ASTM No.	8.5	7.5	9.5	8.6	8.2	6.7	9.2	7.8
d_R/d_Z	1.26	1.14	1.35	1.25	1.28	1.19	1.33	1.19

These measurements show that an important effect of controlled rolling in the plates under study was a reduction in ferrite grain size by about one ASTM Number for the high manganese (H, h) and 1.5 for the low (L, l). From Fig. 1, the banding ratio is seen to vary in an essentially inverse way with grain size, banding being more pronounced in the steel of finer grain size, irrespective of rolling process. Although the ratio is about 1.4 for the coarsest grain size, the structure would be rated non-banded from visual examination. Another trend in the structures is the nearly direct proportionality between ferrite grain size and ferrite patch size, shown plotted in Fig. 2. In the coarse-grain structures (LS and lS), however, patch size is smaller than predicted by the trend; this deviation is interpreted as marking the start of a change towards a structure of Widmanstätten type, even though there was no direct visual indication of such a change.

Transition Temperatures: A lower transition temperature in all C-plates is to be expected from Table II. If a factor in addition to grain size is contributing, however, a different grain-size dependence of transition temperature in C- and S-plates could be expected. Accordingly, the various heat treatments in Table III were carried out to alter ferrite grain size of HS and HC; all were terminated by furnace cooling. The resulting structures were banded with no indication of Widmanstätten structure. Charpy-V 15 ft-lb transition temperatures were determined by testing 25-30 specimens for each grain size with results as shown in Fig. 3* For HS,

*Specimens in rolling direction; notch in thickness direction.

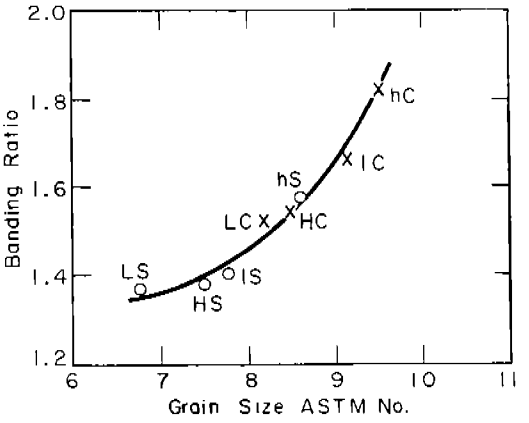

Fig. 1. Dependence of banding ratio on grain size. C: controlled, S: conventionally rolled. Trend not influenced by rolling practice.

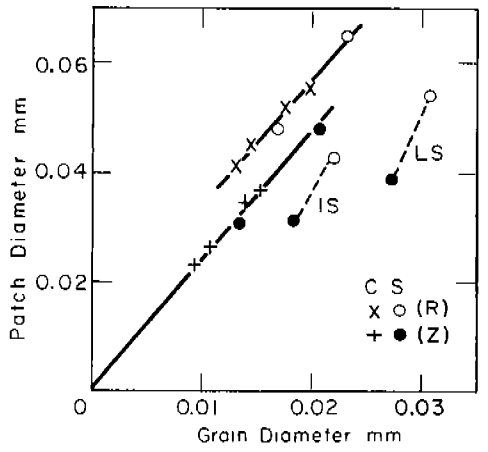

Fig. 2. Relationship between ferrite patch size and mean grain diameter from measurements in thickness (Z) and rolling (R) directions. Only coarse-grain lS and LS, low Mn and conventionally rolled, do not fit pattern.

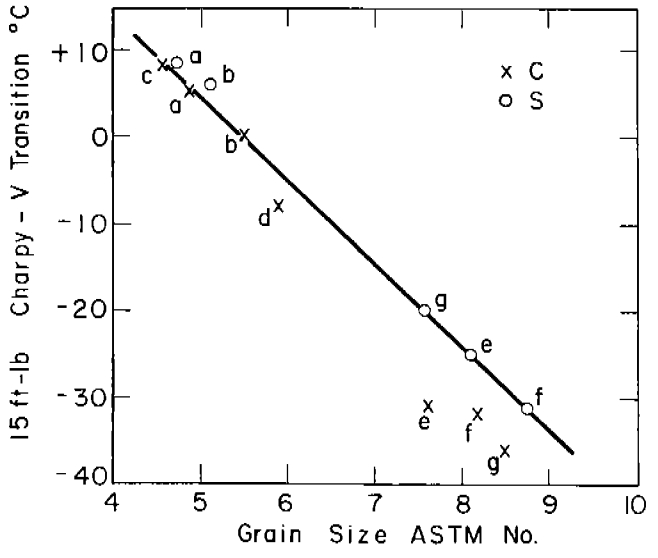

Fig. 3. Grain-size dependence of 15 ft-lb Charpy-V transition temperature determined after various heat treatments on plate HC and HS. Letters indicate treatments listed in Table III.

TABLE III. TREATMENTS OF HS AND HC FOR CHARPY TESTS

Treatment	T (°C)	Time	Thickness
a	1200	3 hr	full-plate
b	1100	3 hr	"
c	1050	24 hr	"
d	900	24 hr	"
e	900	1 hr	"
f	870	5 min	11 mm
g	-------- as received----------		

the slope is 10 C/ASTM Nq; for HC, it is > 10 C/ASTM No. Two features are of special interest: With coarse grain size from prolonged high-temperature annealing, transition temperature is about the same; for finer grain size, however, which is more characteristic of the as-rolled product, HC has the lower temperature. A conclusion is that some amount (approximately 6 C) of the difference between as-rolled HC and HS is the result of an extra-grain-size effect.

These measurements on as-rolled HC and HS and other independent measurements are compared with good agreement in Table IV. Transition temperatures

TABLE IV. CHARPY V-15 TRANSITION TEMPERATURES OF HC AND HS

Source	HC	HS
This work	-36 C	-20 C
Producer	-35 C	-17 C
New York Naval Shipyard	-36 C	-20 C

of the other plates as determined by the producer are plotted against grain size in Fig. 4. Here it is seen that the value for LS does not compare well with that for LC. However, other tests at the New York Naval Shipyard give values of 1°C (LS)

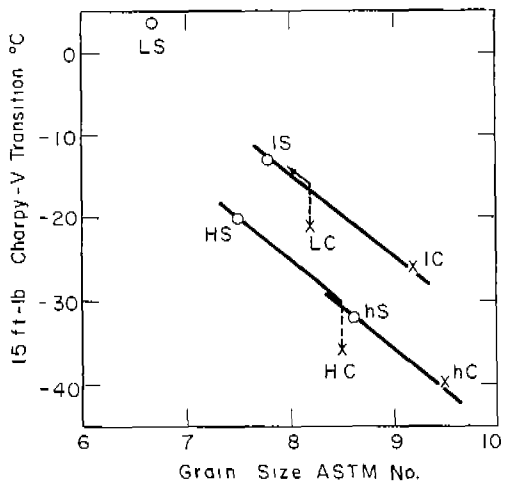

Fig. 4. Transition temperature versus grain size for all as-rolled material.

and -15 C (LC) for a difference of about 10 C/ASTM No. Similarly, for both sets of 3/4-in. thick plates, the differences in transition temperature correspond to 10 C/ASTM No. Such differences can be attributed entirely to grain size and are considered later in the discussion.

Fracturing Anisotropy: Information about fracturing anisotropy was obtained at room temperature with tension tests on specimens of 0.125-in. dia. gage section taken along rolling (R) and thickness (Z) directions in all plates. Samples of HC and HS were also heat treated to obtain differences in grain size and degrees of banding. A pronounced banding followed furnace cooling with virtually none after air cooling. Fracture stress in the Z-direction was determined only with some difficulty. The break was not abrupt as in the rolling direction. Rather, there was a gradual tearing with cracks in the rolling plane being formed soon after yielding. Initial strain hardening was essentially the same in both R- and Z-directions, however. Therefore, a Z-direction fracture stress, σ_Z, was computed by measuring fracture strain at the minimum section on a broken Z-specimen and taking the associated stress from the true stress-strain curve in the R-direction. Such a procedure was the least subjective and still gave values, although on the high side, that were reasonably close to a true fracture stress. All results are summarized in Fig. 5, where each point represents the average of 3 tests.

The slope of the σ_R vs $d^{-1/2}$ line, 8400 psi/mm$^{-1/2}$, is in good agree-

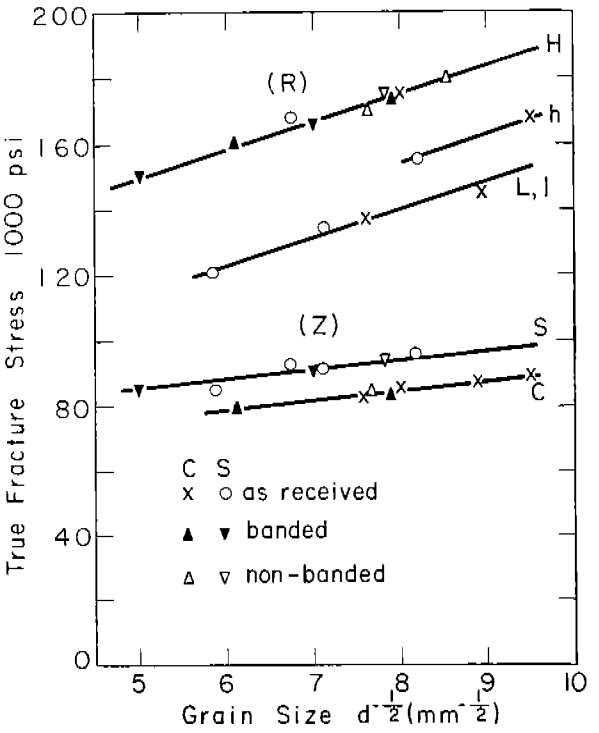

Fig. 5. True tensile fracture stress at 20 C in rolling (3 upper plots) and thickness direction (2 lower plots) as a function of ferrite grain size and with different degrees of banding.

ment with a value found by Heslop and Petch.[7] In the thickness direction, the slope is about 2400 psi/mm$^{-1/2}$, which is closer to the grain-size dependence of yield stress. At equal grain size, σ_Z is smaller for C-plates, but there is no systematic difference between banded and unbanded structures. Thus fracturing anisotropy is not influenced by the intensity of ferrite banding.

NONMETALLIC INCLUSIONS

Samples of all S-materials were analyzed at the Max-Planck-Institut für Eisenforschung.* Two phases were found in each case, pure MnS and an oxide. Composition of the latter is given in Table V. The difference between HS and hS was unexpected, as both were obtained from the same charge although from different ingots. The oxide phase was the more ductile, being flatter and more elongated with sharper edges. When an inclusion contained both phases (duplex), the oxide appeared at the edges.

In investigating effects of rolling practice and heat treatment, an attempt was made to analyze length distributions. Inclusions were counted in an area of 20-30 mm^2 on a surface defined by the rolling and thickness directions. Final polishing was done with a fine (< 1/4 μ) diamond paste. Four groups of different lengths were established: I: 0.01-0.02 mm; II: 0.02-0.04 mm; III: 0.04-0.08 mm; IV: 0.08-0.16 mm. Inclusions smaller than

*Details of procedure are given in Ref. 22, 23, and 24.

TABLE V. INCLUSION ANALYSIS

Plate	Composition** (wt. %)				Total wt. % in.
	SiO_2	MnO	Al_2O_3	FeO	Sample
HS	36	1	56	9	0.004
hS	75	-	18	7	0.009
LS	62	7	18	12	0.007
lS	67	4	18	11	0.012

**Cr_2O_3, TiO_2, MgO, CaO < 1% in all

TABLE VI. EFFECTS OF PROCESSING ON INCLUSIONS

Treatment	Ferrite Grain Size ASTM No.	Inclusion length ratio	
		Oxide	Duplex
HC as rec'd	8.5	1.00	1.00
HC 900 C - 24 hr - F. C.	6.3	0.85	0.85
HC 1050 C - 24 hr - F. C.	4.6	0.65	0.70
HC 1250 C - 24 hr - F. C.*	3.8+	0.35	0.55
HS as rec'd	7.5	0.85	0.95
HS 1250 C - 24 hr - F. C.	4.2+	0.30	0.55
hC as rec'd	9.5	1.25	0.85
hS as rec'd	8.6	0.95	0.85
LC as rec'd	8.2	1.10	0.80
LS as rec'd	6.7	0.80	0.80
lC as rec'd	9.2	1.40	0.80
lS as rec'd	7.8	1.15	0.75

*Furnace-cooled with retarded cooling rate: 900 C to 700 C in 12 hr
+Widmanstätten structure

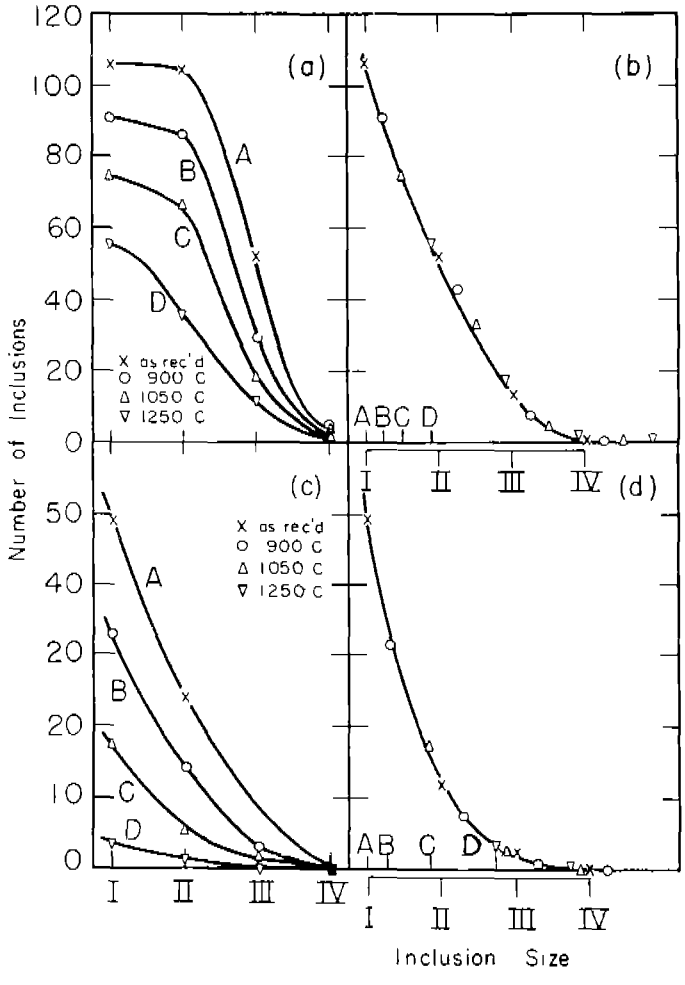

Fig. 6. Distribution of duplex (a, b) and oxide (c, d) inclusions in HC plate after various heat treatments (Table VI). Size groups: I: 0.01--0.02 mm; II: 0.02--0.04 mm; III: 0.04--0.08 mm; IV: 0.08--0.16 mm. Area represented is 10 mm^2 defined by rolling direction and thickness direction. Measurements on the plane surface are shown in (a) and (c); adjustments for volume distribution (see text) are made in (b) and (c).

those of Group I could not be recounted for verification within reasonable limits, probably because of difficulty in retaining them through polishing. Sizes greater than Group IV were also neglected, being too infrequent to be reproduced well. The counting was carried out for all plates in the as-received condition and also for HS and HC after several annealing treatments (see Table VI). An example based on as-received and annealed HC is given in Figs. 6a and 6c, where the distribution is seen to shift towards the left with annealing. Inclusions clearly become spheroidized, the oxides at a somewhat greater rate.

The number counted on a plane section does not represent distribution in volume without some correction for the fact that the probability of cutting an inclusion will diminish with its width. Assuming width and length to be proportional, volume distribution was estimated by multiplying each size group in a

10 mm^2 area by the factor: 1/1 for Group I, 1/2 for Group II, 1/4 for Group III, and 1/8 for Group IV. Counting data for all materials were adjusted accordingly.

Further comparisons were then made assuming a constant number of inclusions and a similar distribution pattern; the only difference considered was that inclusions in one sample may be longer by a certain factor. This length factor could be obtained from the horizontal separation between plots of number vs logarithm of length, which corresponds to the logarithm of the length ratio. Measurements plotted in Figs. 6a and 6c are treated this way in Figs. 6b and 6d where all curves superimpose quite well; length ratios are obtained from the different locations on the abscissa to which the curves have been displaced. Complications are, of course, introduced by differences in chemical analysis; specifically, larger oxide content resulting in a larger number of inclusions would be interpreted as longer inclusions. However, it should be possible to evaluate, at least approximately, effects of annealing and rolling practice if comparisons are made only between the C- and S-plates of any particular pair and if it is assumed that there are no major differences between analyses of controlled and conventionally rolled plate.

Inclusion length ratios, taking as-received HC as 1.00, are listed in Table VI. Conclusions to be drawn are:

(a) Comparing C- and S-plates, oxides were about 30% longer in C. Difference between HC and HS was only 15%, possibly owing to the higher alumina content.

(b) Differences between duplex inclusions in as-rolled plates were negligible. The deformation of sulfides was apparently not affected by rolling practice.

LOW TEMPERATURE TENSILE DEFORMATION

Procedures: Tensile tests were made at temperatures as low as 58 K (-215 C) to investigate fracturing anisotropy as the ductile-brittle transition is approached. Materials were the differently treated HC and HS listed in Table VI. Testing was carried out in an apparatus consisting essentially of a long rod sliding within a tube and positioned vertically in the neck of a liquid nitro-

gen or helium storage vessel. The specimen was gripped between the bottom of the tube and the lower end of the rod while the apparatus was mounted in a hydraulic testing machine. The upper end of the rod was attached through a spherical seat to one crosshead and the upper end of the tube fastened to the other. A round, horizontal cross-bar projected through a universal-type joint in the lower grip for locking the specimen to the tube; the assembly of rod, grips, and specimen was passed down through a slotted plate closing the tube so that by turning 90° and retracting, the bar made contact with the under side of the plate. A linkage between the grips kept a broken specimen from dropping into the vessel. There was no need to move the tube during a series of tests as specimens could be changed by turning the central rod to free the cross-bar and then withdrawing the rod from the neck. The tests were made at a crosshead rate of 0.05 in./min.

Specimens from both thickness and rolling direction were 1-1/2 in. long, with a gage length of 7/8 in. and a diameter of 0.250 in. True fracture stress was calculated from the diameter at fracture as measured with an optical comparator and the load reading at which an audible click indicated fracture. Unlike specimens of 0.125 in. diam., fracture occurred at a fairly well-defined stress even in the thickness direction, indicating that the gradual tearing-apart process is important only with small diameter.

A flow of gas up through the neck for cooling the specimen was obtained by blowing tank gas (N_2 or He) into the liquid. A copper-constantan thermocouple was attached to each specimen near the center of the gage length and, with some experience, temperature could be controlled to $\pm$ 1°C by manually adjusting the flow rate. The lowest temperature conveniently obtained with nitrogen was -182 C; below this, helium was used. Between -182 C and about -205 C, normal evaporation of the helium cooled the specimen to such an extent that the room-temperature gas had to be blown in above the liquid level. Constant temperature was maintained in all cases for 5-15 min before starting a test. The total time required for one test was 40-60 min.

Figures 7-12 represent the temperature dependence of: true tensile fracture stress in both rolling, σ_R, and thickness, σ_Z, directions; lower yield stress,

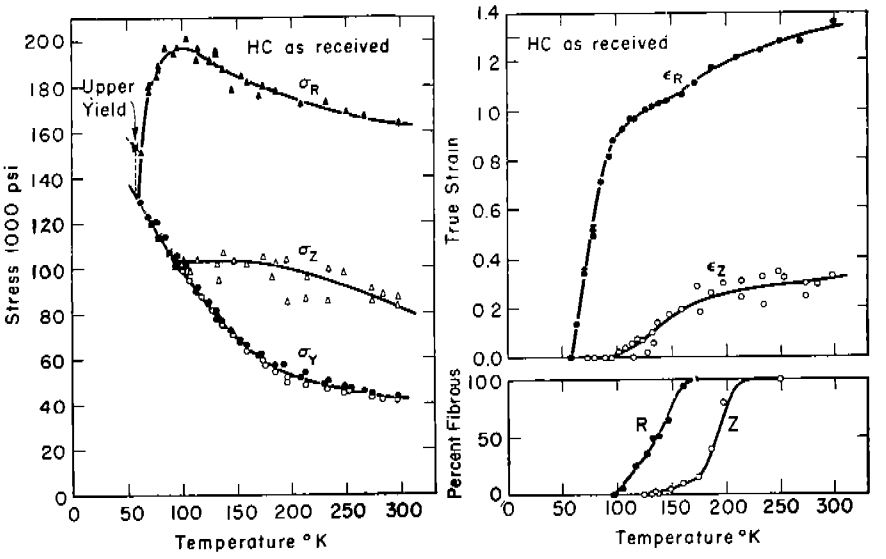

Fig. 7

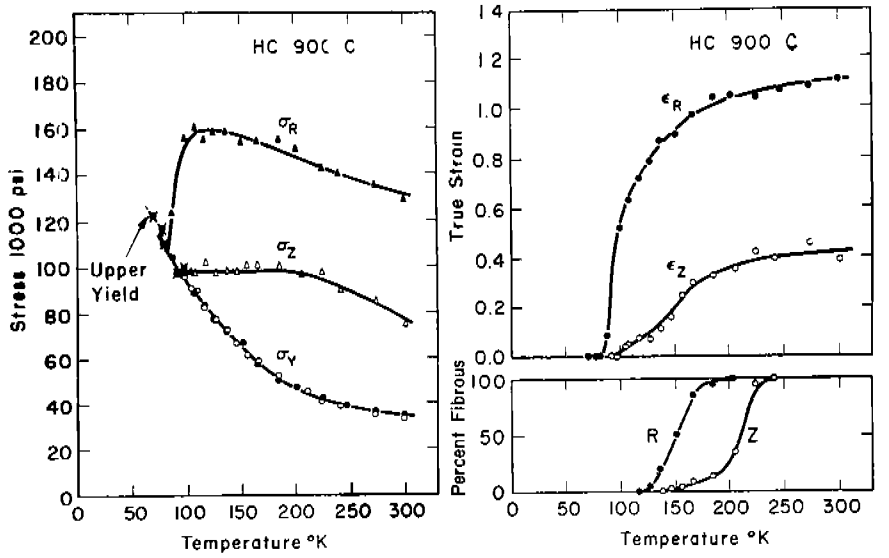

Fig. 8

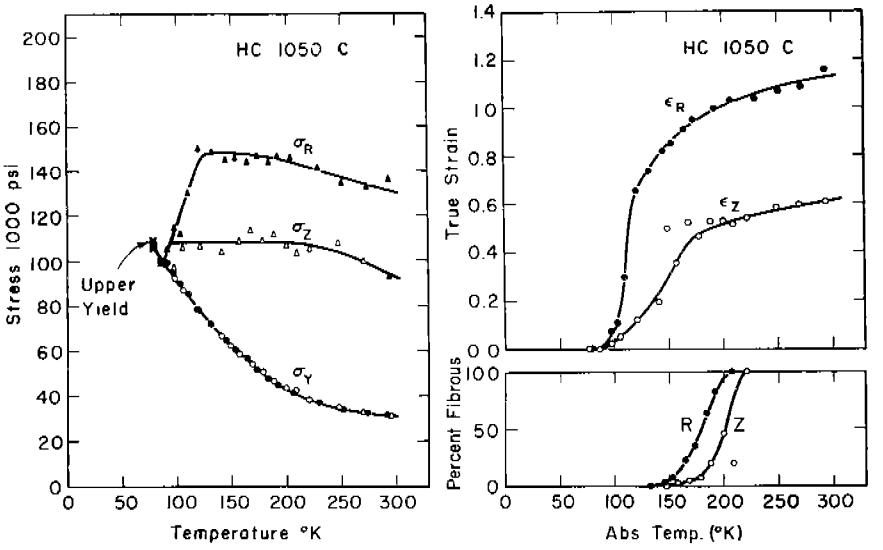

Fig. 9

Temperature dependence of true tensile fracture stress (σ_R,σ_Z), yield stress (σ_Y), fracture strain (ϵ_R, ϵ_Z), and extent of fibrous area on samples from rolling (R) and thickness (Z) directions.

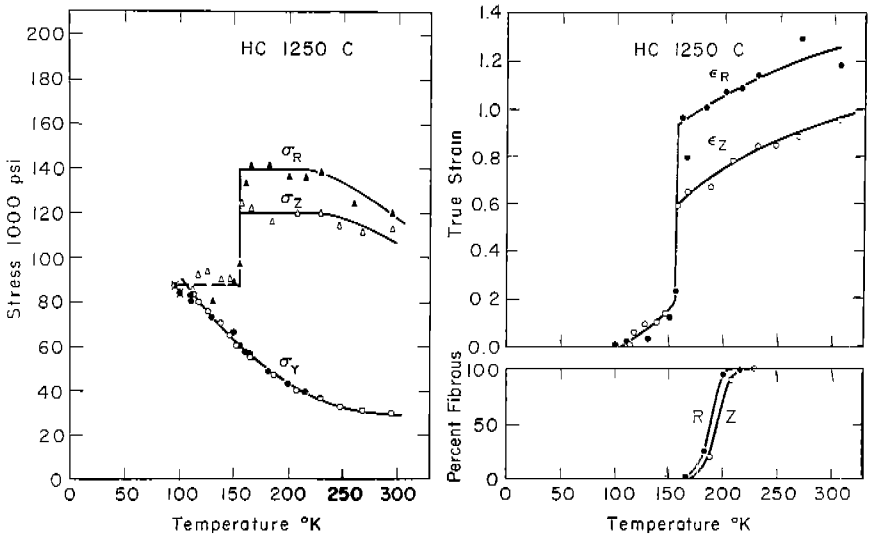

Fig. 10

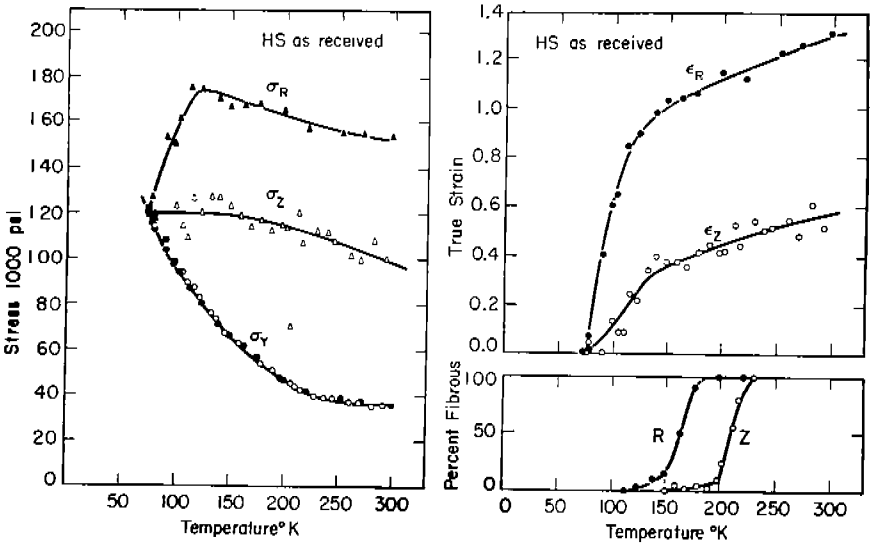

Fig. 11

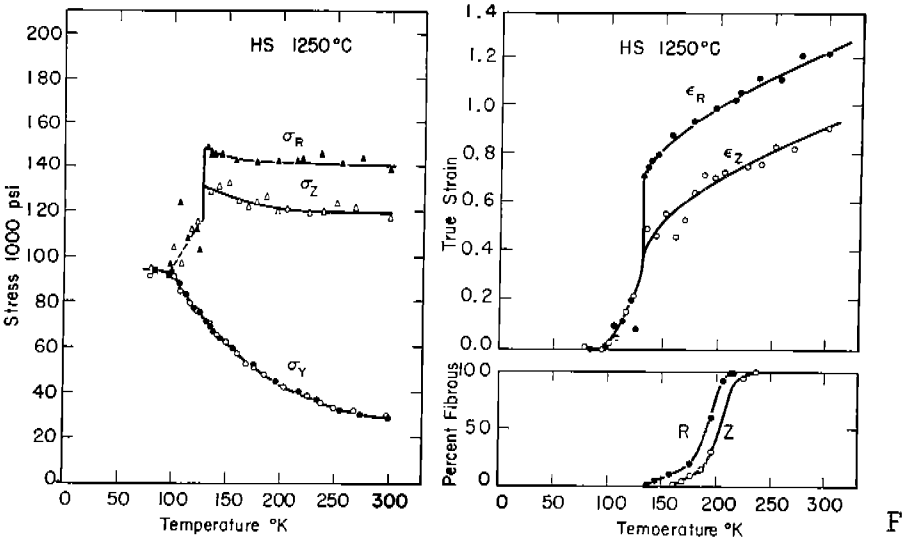

Fig. 12

Temperature dependence of true tensile fracture stress (σ_R, σ_Z), yield stress (σ_Y), fracture strain (ϵ_R, ϵ_Z), and extent of fibrous area on samples from rolling (R) and thickness (Z) directions.

Fig. 13. Cracking around visible inclusions after yielding in a specimen (HS) from the thickness direction (500 X).

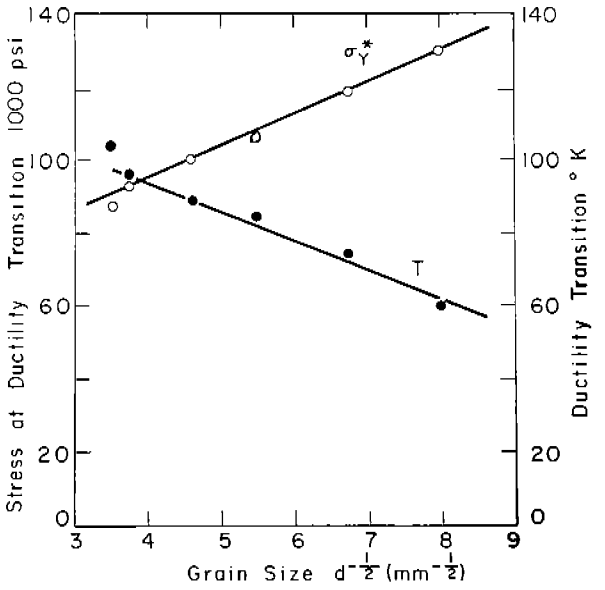

Fig. 14. Ductility transition temperature (intersection of σ_R and σ_Y curves in Figs. 7--12) and corresponding stress σ_Y^* as a function of grain size. Trend is insensitive to rolling practice.

σ_Y; and true tensile fracture strain, ϵ_R and ϵ_Z. These figures also indicate the extent of fibrous area on fracture surfaces of R- and Z- specimens, visually estimated.

Fracture Observations: At low temperatures, as at room temperature, freely visible surface cracks normal to the stress axis in Z- specimens appeared after yielding. Cracks were not found when fracture occurred at the upper yield point; therefore such cracking must be associated with plastic deformation. Microscopic examination showed that the cracks originated at inclusions,

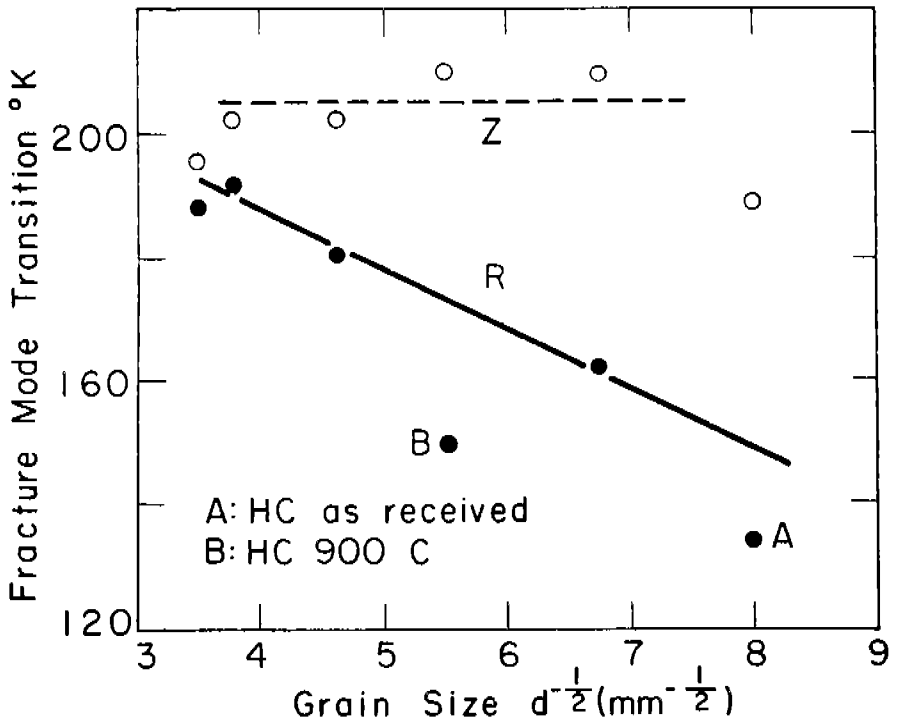

Fig. 15. Grain-size dependence of temperature for 50% fibrous fracture from Figs. 7--12. Differences indicated are consistent with Fig. 3.

as in Fig. 13. Apparently the path of ductile fracture follows along many of these damaged areas, for numerous inclusions may be seen in the fracture surfaces of Z-specimens.

Variation with grain size of the ductility transition temperature, defined as the intersection of σ_R and σ_Y curves, and the corresponding σ_Y* are plotted in Fig. 14. There is no difference between conventional and controlled-rolled specimens in this regard. A fracture-appearance transition temperature was also defined by 50% fibrous fracture area. Its dependence on grain size is shown in Fig. 15, in which the line is drawn with the same slope as in Fig. 14. Here the lower levels of HC-plate as-received and annealed at 900 C suggest a depression in transition temperature for some extra-grain-size reason even more distinctly than in the 15 ft-lb Charpy transition (Fig. 3).

The spread in σ_Z measurements on as-received specimens seems to represent not experimental error but rather a material characteristic, for it is reduced by annealing. A few as-received specimens, especially of HS, broke at abnormally low stresses, even below the yield point. When this happened, an unusually large inclusion could always be found in the fracture. Thus the

very large, sharp-edged (as-rolled) inclusions do occasionally initiate fracture, but there is no indication that they are generally important in this respect. Over a range of low temperatures, σ_Z is essentially constant (Figs. 7-12), even as the fracture mode changes from largely ductile to brittle and the fracture path no longer follows inclusions. This suggests a critical stress criterion for these conditions, with both ductile and brittle modes sharing a common initiation process which determines the level of σ_Z.

As the ductility transition temperature is approached, both σ_R and ϵ_R drop rapidly, but they do so for most specimens without indication of the discontinuous or bimodal fracturing suggested in results obtained by Wessel[25] and Hahn et al.[26] The clearest example is given by HC as-received for which the ductility drop occurred at liquid nitrogen temperature. A large spread in ϵ_R would be expected for a discontinuous change. Yet at liquid nitrogen, three tests giving an intermediate ductility level were closely reproduced (Fig. 7). Only for HS and HC annealed at 1250 C was the change so sharp that fracturing could, with some confidence, be interpreted as bimodal. The temperature difference between the abrupt drop in ductility and the ductility transition was larger for HC, which was cooled from the annealing temperature at a slower rate and had a larger grain size.

Large differences in fracturing anisotropy, defined as $\sigma_{Z(max)}/\sigma_{R(max)}$ from Figs. 7-12, are listed in Table VII. These derive from at least two sources:

TABLE VII. EFFECT OF PROCESSING ON FRACTURING ANISOTROPY

Plate	Treatment	$\frac{\sigma_{Z\,(max)}}{\sigma_{R\,(max)}}$
HS	1250 C - 24 hr - F.C.	0.88
HC	1250 C - 24 hr - F.C.	0.87
HC	1050 C - 24 hr - F.C.	0.72
HS	as rec'd	0.69
HC	900 C - 24 hr - F.C.	0.61
HC	as rec'd	0.52

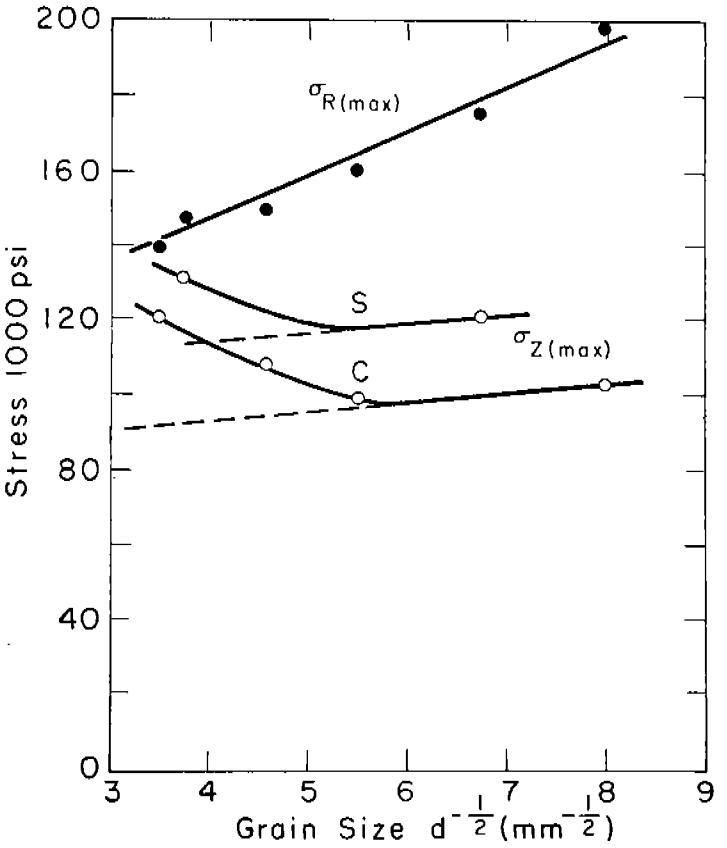

Fig. 16. Dependence on grain size of maximum fracture stresses from Figs. 7--12. Two reasons for reduced fracturing anisotropy are demonstrated: lower $\sigma_{R(max)}$ with increasing grain size and increased $\sigma_{Z(max)}$ from high temperature annealing.

differences in inclusion distribution and in grain size. The grain-size effect is illustrated by room-temperature measurements (Fig. 5) showing σ_Z/σ_R to decrease with reduction in grain size because of σ_R increasing more rapidly than σ_Z. Effects of the annealing treatments on both inclusions and grain size are reflected in Fig. 16. A Widmanstätten structure in the HC- and HS-specimens annealed at 1250 C results in a grain-size uncertainty. However, it is still clear that a consequence of prolonged annealing is an increase in σ_Z superimposed on the room-temperature trend (Fig. 5, slope = 2400 psi/$mm^{-1/2}$) for as-rolled and normalized specimens. This increase follows closely the spheroidizing of inclusions described in Fig. 6 and Table VI. Unfortunately, detailed comparisons between HC and HS are difficult because it is uncertain that they have the same inclusion analysis. Nevertheless, lower $\sigma_{Z(max)}$ for HC does correlate with more elongated inclusions in the controlled-rolled plate, which should contribute to lower fracture stress in the Z-direction. Perhaps the reasonable view is that changes in the more obvious inclusions serve only as qualitative indicators of changes in the fine-scale flaw structure so important to fracturing anisotropy.

DISCUSSION

The experiments reported here have demonstrated an effect of rolling

practice on fracturing characteristics. They have also shown that fracturing anisotropy does not result from ferrite banding, but rather, there is evidence that a structure of elongated flaws is the primary cause. The further indication is that such a condition may contribute, along with reduced grain size, to a lowering of transition temperature by controlled rolling. The argument is that outlined earlier.

With a standard Charpy bar as the example, deformation beneath the notch produces a tension stress σ_3 across the weak plane parallel to the plate surface; σ_3 is only a fraction q of the axial stress σ_1, or $\sigma_3 = q\sigma_1$. A separation or fissuring occurs in this plane before a brittle crack forms if σ_3 reaches the level of Z-direction fracture stress σ_3^* before $\sigma_1 = \sigma_1^*$, the fracture stress in the R-direction. Therefore, the requirement for fissuring is that $\sigma_3^*/\sigma_1^* < q$. The fissure must lower both hydrostatic and axial stress and thus reduce the tendency for brittle fracture. Recent work by Hendrickson, Wood, and Clark[27, 28] can also be interpreted in support of the argument and used in a more specific statement.

It was shown that brittle fracture in both notched tension specimens and Izod bars occurred as σ_1 reached a critical level σ_1^* at the elastic-plastic boundary of a small plastic zone beneath the notch. In general, $\sigma_1 = K\sigma_y$; K depends upon notch geometry, increasing with distance beneath the surface to a maximum at the boundary, and σ_y is fixed by prevailing conditions of structure, temperature, and stress rate. At the transition temperature, the plastic zone is on the verge of propagating across the section, and $\sigma_1^* = K_m \sigma_y^*$, with K_m having the largest possible value for given geometry and method of loading. A corollary is that under fixed conditions the ductile-brittle transition may be characterized by a critical value of yield stress, σ_y^*. In the notched Izod and Charpy specimen, $K_m = 2.4$. For plane-strain deformation, $\sigma_3 = \frac{\nu(2K-1)}{K}\sigma_1$, where ν is Poisson's ratio. Therefore, at the transition temperature, $\sigma_3 \simeq 0.5\sigma_1^*$, and here q is a maximum at $q_m = 0.5$. Below the transition temperature, σ_1^* is still reached at the elastic-plastic boundary. Now, however, $\sigma_1^* = K'\sigma_y'$, where $\sigma_y' > \sigma_y^*$, so that $K' < K_m$, with the two results that the boundary lies nearer to

the notch root and $q' < q_m$. Accordingly, the fissuring criterion, $\sigma_3^*/\sigma_1^* < q$, is most likely to be satisfied around the transition temperature since here $\underline{q}$ is maximum. Fissuring could also be expected at lower temperatures, although $\underline{q}$ becomes less, so that eventually σ_3 may not reach σ_3^* before $\sigma_1 = \sigma_1^*$. In these terms, the effect of fissuring is to lower the constraint, expressed by K, in a small volume around the plastic-elastic boundary. The result with the given testing method is that lower temperature is necessary to obtain higher σ_y, thus compensating for reduced K in reaching σ_1^*.

The σ_Z/σ_R values in Table VII indicate that fissuring at and near the transition temperature might be expected for some treatments. Values of σ_1^* for these steels are unknown and apparently cannot be measured under pure tensile loading. In any event, the experimental $\sigma_{R(max)}$ is perhaps not far below σ_1^*. The existence of σ_3^* must be postulated, although the idea seems a reasonable one; σ_Z is relatively insensitive to temperature over the lower end of the range, and the level on approaching intersection with the σ_y curve ($\sigma_{Z(max)}$) may then be viewed as an approximation to σ_3^*. Therefore, the measurements of $\sigma_{Z(max)}/\sigma_{R(max)}$ would seem reasonable (high) approximations to σ_3^*/σ_1^*. On this basis, treatments giving the smallest values of σ_Z/σ_R have the highest probability of fissuring. Accordingly, the materials in order of decreasing tendency to develop fissures may be read from Table VII starting with the most likely, controlled-rolled HC as received ($\sigma_{Z(max)}/\sigma_{R(max)} = 0.52$), and going to the least, HS annealed at 1250 C ($\sigma_{Z(max)}/\sigma_{R(max)} = 0.88$). Of these, the two materials with lowest ratios also have the lowest transition temperatures if comparisons are made after adjustment for differences in grain size (Fig. 3).

The basic requirement in this argument, a finely dispersed flaw structure, has been stressed. Evidence for such a structure exists, and supporting observations were made in the present work. Figure 17 represents the fracture surface directly beneath the notch in a Charpy specimen breaking with energy absorption of 6 ft-lb. Normally this fracture would be classed as 100% brittle, yet a microfissuring in the rolling plane is clear over a region of a size (2-5 μ)

Fig. 17. An observation of the fracture surface of microfissuring beneath the notch in a brittle Charpy specimen (HC) (2000 X).

consistent with the plastic volume that develops before the brittle crack. Microfissures at the notch root could also be observed in somewhat more ductile specimens, but these were rapidly obscured with increasing deformation.

The specimen in Fig. 17 happens to be HC as rolled; similar observations can be expected in other materials with properties favorable for fissure formation. Findings to date show only that microfissuring occurs; further study is necessary before variations in degree can be established and the origin of the fissures made clear. It seems a reasonable possibility that they result from very small inclusions. Although an intense fibering is important, it may not be a sufficient condition for microfissuring. It is clear from Figs. 5 and 16 that fine grain size also contributes to reduced σ_Z/σ_R through an increase in σ_R. In H-steels the probability of fissuring proved greatest, since higher σ_R was realized even for constant grain size. This would seem to be a reason for finding the microfissuring effect only in this material (Fig. 4).

In the heavily deformed regions, beneath surfaces of ductile fracture, cracks were clearly evident around practically every visible inclusion. These areas were sometimes connected to form a large crack lying in the rolling plane which might best be termed a macrofissure. Such fissures are relatively common in the ductile range where they do influence notched-bar energy absorption.

However, their spacing is too great compared with the size of the plastic volume at the notch root to act as efficient stress-relievers in the low energy-absorption range.

As a final observation, there is reasonable experimental support for the view that microfissuring can be involved in determining the level of notched-bar transition temperature. Grain size as well as the fibering necessary for microfissuring are both influenced by processing history. Therefore, critical evaluation of the effect of processing on notch toughness should consider both structural details; advantage may even derive from processing designed for control over both.

SUMMARY AND CONCLUSIONS

1. The lower Charpy-V 15 ft-lb transition temperatures of controlled-rolled plates studied in this program result largely from smaller ferrite grain size. There is experimental evidence, however, that a part of the improvement may result from a microfissuring in the plane of the plate at the notch root; the effect is to relieve stress triaxiality and so depress transition temperature.

2. Microfissuring is not unique to controlled-rolled plate. Two requirements must be satisfied for its occurrence, regardless of processing history: (1) a flaw structure dispersed on a scale no greater than the size of the plastic volume from which the brittle crack originates; and (2) a ratio of the critical fracture stress in thickness (Z) to rolling (R) direction no greater than about 1/2.

3. Microfissuring may be seen beneath the notch root in Charpy specimens broken with little plastic deformation. The origin of flaws responsible for the fissures is not clear but may simply be inclusions too small for observation with normal metallographic techniques. Neither the visible inclusions, responsible for a macroscopic ductile fissuring, nor ferrite banding have an effect on transition temperature.

4. The ratio σ_Z/σ_R in tension over a range of temperatures extending down to the ductility transition is lowest for controlled-rolled plate in the as-rolled condition because of the low values of both σ_Z and grain size (giving high σ_R).

5. Correlations between σ_Z and inclusion content show a decrease in σ_Z to be paralleled by more elongated inclusions, as if the visible changes are indicative

of changes in the fine-scale fiber structure. Annealing above 900 C gradually spheroidizes inclusions and is accompanied by marked increase in σ_Z.

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APPENDIX

Rolling Schedule: Ingots were 80-3/4 in. in height, measuring 31-1/4 x 60-1/4 in. at the bottom and 29-1/2 x 59 in. at the top. Reduction consisted of rolling first to slabs: 98 x 55 x 9 in. for L and H plate, and 61 x 55 x 9 in. for l and h plate. With further rolling in a two-high mill, L- and H-slabs were reduced to 3-1/2 in. and 4-1/4 in., respectively, and l and h were reduced to 3 in. and 2-3/4 in., respectively. The S-plates were finished in a four-high mill according to more conventional practice, the temperature of the finishing pass being between 930 C and 970 C. The controlled-rolled C-plates were finished with the following schedule:

Pass No.	1	2	3	4	5	6	7	8	
Pass Temp.	>910°	890°	860°	830°	800°	770°	740°	720°	After finishing pass, cooling rate to 200 C
Plate	Thickness (in.)								
HC 1-1/2 in.	>2.25	2.08	1.93	1.80	1.67	1.56	1.46	1.46	16 C/min
LC 1-1/2 in.	>2.25	2.13	1.97	1.81	1.69	1.56	1.47	1.47	14 C/min
hC 3/4 in.	>1.20	1.10	1.02	0.93	0.85	0.77	0.72	0.72	16 C/min
lC 3/4 in.	>1.18	1.08	0.99	0.90	0.83	0.75	0.70	0.70	18 C/min

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