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LOW-STRESS BRITTLE FRACTURE IN MILD STEEL

SSC-158

By

R. Dechaene, W. Soete, and A. Vinckier

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30 August 1963

Dear Sir:

Professor ir. W. Soete, Director of the Laboratorium voor Weerstand van Materialen, University of Ghent, Belgium, accepted the invitation to participate in the Annual Meeting (held on March 5 and 6, 1963 in Washington, D. C.) of the Committee on Ship Steel of the National Academy of Sciences-National Research Council, one of the principal advisory committees to the Ship Structure Committee. The enclosed report entitled Low-Stress Brittle Fracture in Mild Steel was prepared by Professor Soete and his associates, R. Dechaene and A. Vinckier, to summarize his remarks for the Committee on Ship Steel.

Please send any comments on this report addressed to the Secretary, Ship Structure Committee.

Yours sincerely,

T. J. Fabik
Rear Admiral, U. S. Coast Guard
Chairman, Ship Structure Committee

SSC-158

Special Report

on

LOW-STRESS BRITTLE FRACTURE IN MILD STEEL

by

R. Dechaene, W. Soete, and A. Vinckier
University of Ghent
Belgium

Washington, D. C.
U. S. Department of Commerce, Office of Technical Services
August 30, 1963

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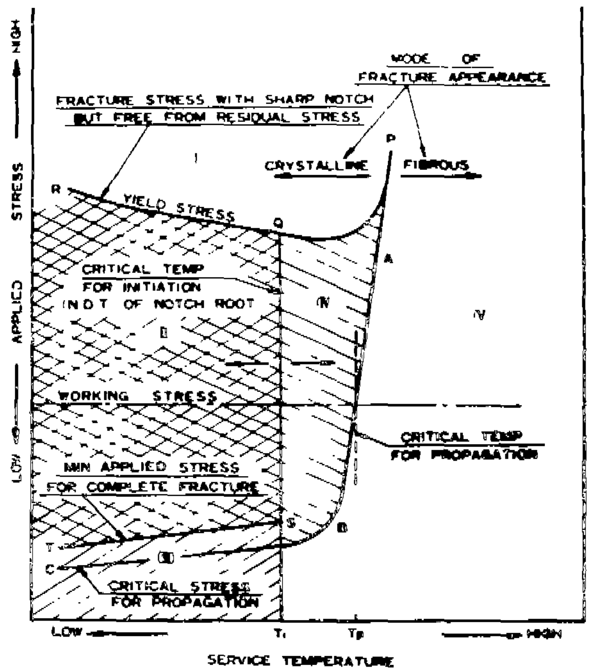
May I start this lecture by expressing my sincere thanks to you Mr. Chairman, and the members of the Ship Structure Committee for their kind invitation. You know that we Europeans are looking up with admiration to the scientific accomplishments of the United States. Each time the news of a new American realization reaches us, we feel happy and proud. We know that the equipment of your laboratories and the staff of your investigators are so large that any contribution from outside seems almost needless. Despite these tremendous research possibilities, you still send specialists to Europe to attend scientific meetings and even invite Europeans to the United States for discussion. This is typical for your scientific objectivity and at the same time a demonstration of democracy--a treasure so carefully protected in this country.

I was asking myself why you chose me to speak here on the problem of brittle failure and I could see only one reason, and that is that I am from a country where all the trouble started. Belgium was the first country which suffered seriously from the catastrophic failure of welded structures. It was way back in March 1938 that the Hasselt bridge collapsed. Since then research work on brittle failures has started all over the world. Despite tremendous efforts no satisfactory explanation why, and no practical solution against such failures have been found. In June 1962, a bridge in Melbourne suffered serious damage, and I am practically sure that the severe winter now in Europe caused a lot of damage on a good many of our steel structures. The phenomenon of brittle fracture does not limit itself to welded steel structures: even prestressed concrete structures suffer from the same trouble. A few weeks ago I was indeed confronted with rather mysterious failures of prestressed bars, which suddenly broke and show all the well-known characteristics of brittle failure.

When reviewing the extensive literature on brittle fracture it is possible to class the papers in two major groups: one dealing with the scientific aspect of the problem, the other one with the technological aspect. In the first group concepts such as dislocations, state of stress, energy-balance, etc. are considered, while the second group relates empirical test results, expressed in the most odd quantities such as foot-pound, energy absorption as in a Charpy test, percentage crystallinity, percentage of contraction, and angle of bending. Unfortunately it must be said, that neither group gives results which are readily usable for the steel user. This statement may not be interpreted as a criticism on research work, but one must recognize that for the moment no relevant link exists between the scientists and steel users while on the other side technologists interpret their test results in quantities which cannot be introduced in formulas for use when designing a steel structure.

A general feeling is growing that after breaking impact specimens for more than half a century time has come to interpret these test results in absolute units rather than in relative ones. In other words the steel user is no longer satisfied with Charpy values and the statement of the steel manufacturer declaring that steel A is better than steel B. He wants to know exactly in what service conditions he can use steel A or B. To reach this aim, test results



Description :

Region (I): **Initiation and propagation (With sharp notch free from residual stress)**

Region (II): **Single stage complete fracture under low applied stress (With residual stress and sharp notch)**

Region (III): **Partial brittle fracture under low applied stress (do)**

Region (IV): **Non-initiation but propagation (do)**

Region (V): **Non-initiation and non-propagation (do)**

FIG. 1. SCHEMATIC DIAGRAM ON FRACTURE STRENGTH OF WELDED STEEL PLATE.

should be plotted in units familiar to the steel user: such units are for instance ultimate load and elongation. In the last two years a serious step forward has been made in this direction. It is indeed hopeful to see that more and more test results are being plotted in a diagram strength versus temperature. Of course, these curves relate only facts, and they do not explain the phenomenon; but for the steel user they have a direct engineering significance. A typical example is given by Fig. 1 taken from a publication of Kihara.[1] The coordinates, respectively strength and temperature, are quantities suitable for design purposes. Characteristic temperatures which can be deduced from this figure are respectively:

t_i : temperature of initiation of a brittle crack

t_a : temperature of arrest of a brittle crack

Of course there is still some doubt about the method of how to determine these temperatures. Some simple but realistic conventions should once and for all make this point clear. For the sake of safety it is logical to determine t_i on a specimen in which a crack, for instance a fatigue crack, has been introduced. Using a crack instead of a notch excludes any discussion about the shape of the notch. Tests have shown that the determination of t_i can be done quite accurately and that this temperature is independent of crack length or specimen width, but is dependent on the thickness of the specimen. On the other side, the arrest temperature t_a can be determined on a Robertson-type test plate. One temperature-gradient test and four isothermal tests are sufficient to determine t_a within 5 C.

As far as the brittle-fracture strength is concerned the determination is a bit more troublesome. Of course in the ductile zone, that is above the temperature t_a, the strength of a precracked wide-plate specimen is above or equal to the yield strength. However when the specimen is cooled below the temperature t_i, the phenomenon is more complicated, and no clear answer can as yet be given. Experiments have shown that for practical purposes it may be assumed that the so-called brittle strength equals the yield strength at this temperature. However there are some most discouraging exceptions and we can almost be sure that all catastrophic failures of welded structures were low-stress brittle fractures, that is the fractures occurred in the structure for average stresses well below the yield strength of the virgin metal. In his lecture given at Cambridge in 1959 for the Admiralty Advisory Committee on Structural Steels,[2] Dr. Weck claimed very accurately: "The really disquieting feature in the brittle fractures of welded structures was not that they were brittle but that there was evidence of an apparent loss of strength." It is along these lines of thought that we focused our research work on brittle fracture.

DELINEATION OF THE PROBLEM

Low-stress brittle fractures illustrate very clearly that the cause of such premature failures lies in the impossibility for the material to sustain overall strains, which can normally be absorbed by the ductility of the steel.

Reasons for this lack of ductility are:

1. State of stress and strain
2. Low temperature
3. Plastic straining prior or during loading
4. Strain rate.

There is of course an academic interest in knowing how much each of these factors enhance brittleness, but from the point of view of the steel user, this interest is rather limited. His problem is indeed simple. All discussions about the state of stress or triaxiality are unnecessary by using a specimen with a crack; nature is indeed so kind to introduce the triaxiality in which we are interested at the bottom of the crack when the specimen is loaded.

As far as temperature is concerned the users' interest is limited to the brittle behavior of the metal. It suffices to run all tests at temperatures below the temperature t_i of the cracked specimen say for instance the lowest service temperature. The real burden of the problem is to know exactly the amount of plastic straining which can occur before or during loading. Unfortunately literature is very scant about the plastic strain distribution in a notched or cracked specimen. If the theory of elasticity has been a useful tool it must be recognized that it has been also a nuisance for a better knowledge of plastic strain distributions. If isoelastics, for instance, is a generally used concept in elasticity there is no similar word

FIG. 2. INTERFERENCE LINES.

for designating the lines of equal strain (iso-strains ?).

An attempt has been made to know something more about the strain distribution during loading in a precracked specimen. It may be of interest to give some results of work done at the University of Ghent. These tests were done at room temperature above the t_1 temperature of the material and aimed to get more information about state and amount of plastic straining.

MEASUREMENTS OF PLASTIC STRAINS IN CRACKED PLATES

Two specimens were prepared, both having approximately the same width (140 mm - 5 1/2 in.) and the same thickness (14 mm = 0.44 in.). Both contained a central crack consisting of a round hole, extended by two saw-cuts and two fatigue-cracks.[3] As far as the strain at the tip of the crack is concerned, such a slit is equivalent to a single fatigue crack, having the same total length. The length of the crack in specimen 1 was 7% of the specimen width, while the crack in specimen 2 extended to 53% of the specimen width.

The strain was measured with the aid of the moiré method. This method has been described by several authors.[4] Suffice it to say that the strains on the surface result in dark and bright interference lines (Fig. 2). The amount of strain is roughly proportional with the inverse of the distance between two neighboring inter-

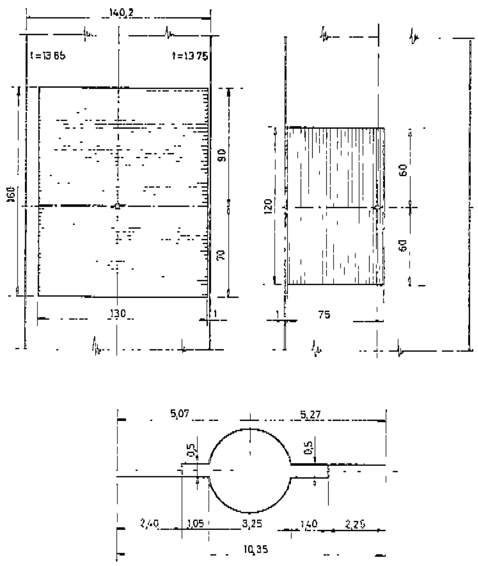


FIG. 3. SPECIMEN WITH SHORT CRACK.

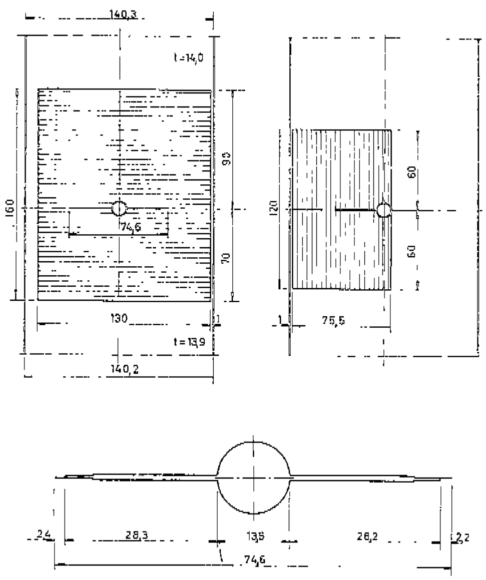


FIG. 4. SPECIMEN WITH LONG CRACK.

ference lines. These lines are in fact loci of points having the same displacement in a given direction.

The longitudinal strains were measured on part of one surface and the transverse strains on a part of the other surface. Figures 3 and 4 show the extent of the grids used and the direction of the grid lines; the distance between the grid lines .006 in. is not drawn to scale. This grid is not sufficiently fine to detect strains of elastic magnitude.

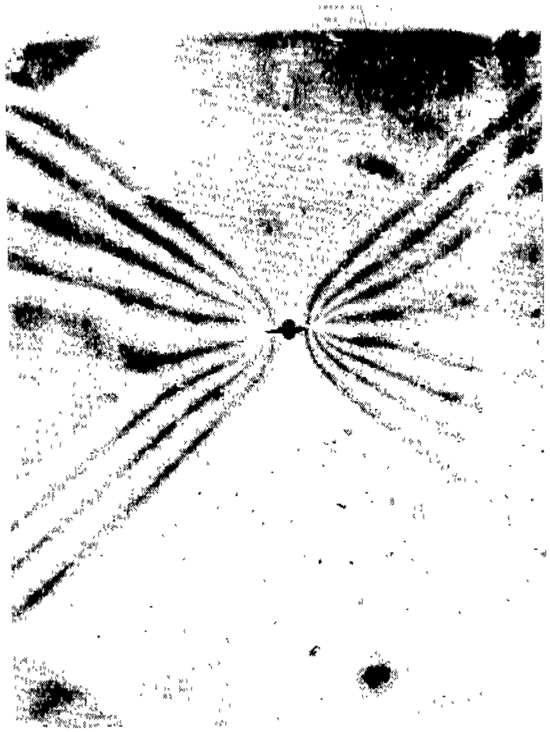
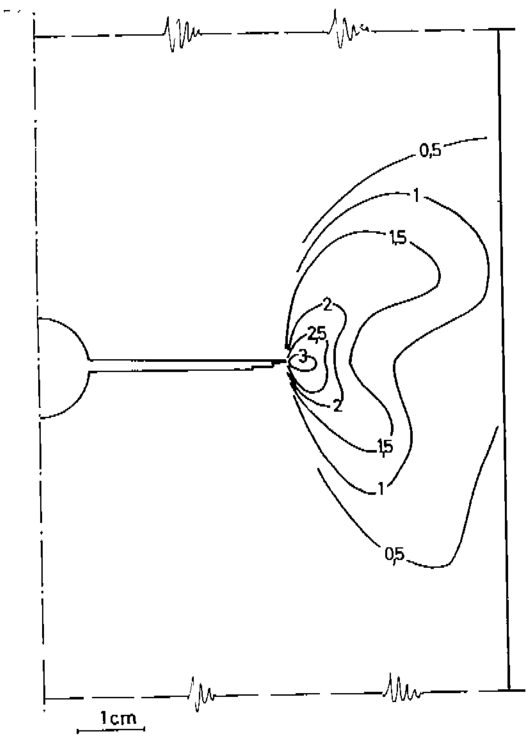


FIG. 5. MOIRÉ PATTERN OF THE LONGITUDINAL STRAINS IN THE SPECIMEN WITH A SHORT CRACK.



1 FIG. 6. CONTOUR LINES OF LONGITUDINAL STRAIN, AFTER UNLOADING FROM 28.9 kg/mm²

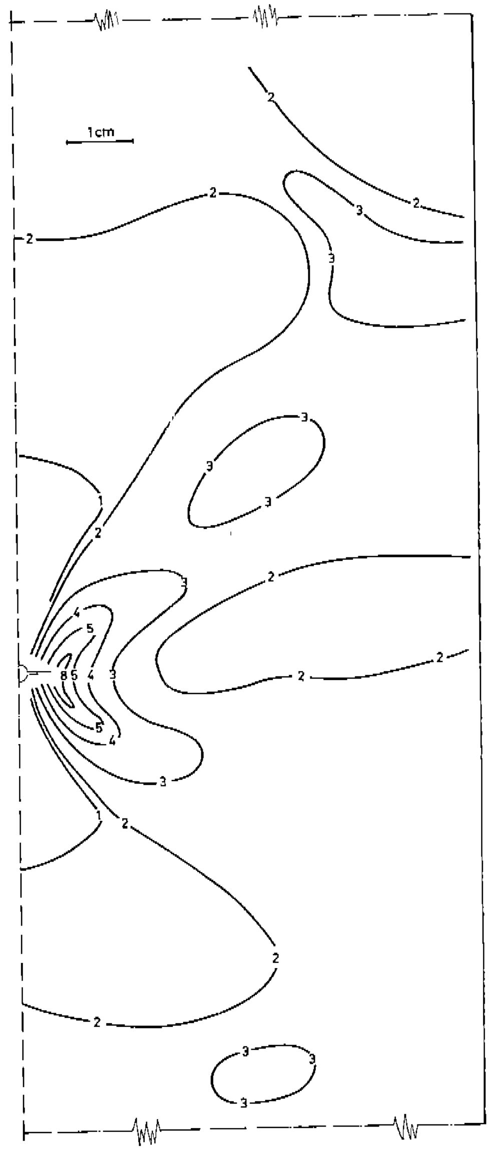


FIG. 7. CONTOUR LINES OF LONGITUDINAL STRAIN, AFTER UNLOADING FROM 29.2 kg/mm².

The specimens were removed from the tensile machine after different amounts of straining and the moiré patterns on both sides were photographed. Moreover, close-up photographs, not showing the moiré pattern, were taken of the vicinity of one end of the crack. They allowed a direct measurement of the distance between the grid lines in the highly deformed region.

Figure 5 shows a typical moiré pattern of the longitudinal strains in the specimen with the shortest crack: the moiré lines and hence the plastic strains are concentrated within a fan on either side of the crack. As the strain is inversely proportional with the vertical distance between these lines, the strain is much larger near the crack tip than on the edges of the specimens.

The moiré patterns in the specimen with the long crack were quite similar. But when the nominal stress of the specimen with the short crack rose to the yield point, plastic deformation also took place outside the fan.

From careful measurements of the distance between the moiré lines and of their inclination, it was possible to map the plastic strains on the specimen and to calculate the principal strains and their principal directions. Figure 6 shows the contour lines of the longitudinal strains existing in the specimen with a long crack when the average stress on the net section was 28.9 kg/mm^2 (40,100 psi). For comparison, the contour lines of the longitudinal strains in the specimen with the short crack are shown in Fig. 7: the average stress on the net section was 29.2 kg/mm^2 (40,500 psi).

These figures clearly show that the highest strains are located in two narrow regions extending from the crack tip under an angle of about 45° with the axis of symmetry. Calculations of the principal directions show that, with the exception of the immediate neighborhood of the crack, they do not deviate much from the directions of symmetry of the specimen, while the ratio between principal strains is closer to -1 than to -2. This confirms the point that the deformation occurs mainly by shear in one direction.

The first observable plastic strains occurred in Lüdersbands forming an angle of about 48° with the longitudinal axis.

Figures 8-11 show close-up pictures of the crack tip. Here again, it is seen that the strains are concentrated along two regions extending under about 45° from the deformed crack tip.

From the measurements it appears at once that fairly large strains, up to 20%, are found on a gauge length of 0.154 mm (0.006 in.) after only a very small overall elongation of the specimen (0.08 mm or 0.003 in.). Any fracture occurring at this moment should still be called "brittle". All the material in the moiré fan has yielded for an average stress of 29 kg/mm^2 in both cases. However the moiré fan is much larger for the short crack, thus a larger overall extension is needed to bring it to full yield than is the case for the long crack. A larger overall extension means more moiré lines, which, as all moiré lines converge at the crack tip, leads to higher local strains at the tip of the shorter crack. Although one crack was seven times shorter than the other one, both cases have exactly the same boundary conditions: the shear zones extend from the tip of the crack to an unloaded edge. The strain distribution seems in this case to be primarily governed by the absolute value of the uncracked part, rather than by the length of the crack or the ratio of the crack to the net section. The boundary conditions of both our specimens being the same, basic differences in behavior were not observed.

It is thought that this might not be the case if the shear zones had to develop in an infinite plate, or if they extended to a loaded edge, or if their development was obstructed by more rigid parts of the specimen. If this is true, the plastic strain concentration near a crack would not only depend on local conditions, as in elasticity, but also on the dimensions of the specimen and on its boundary conditions.

Summarizing, it was found that plastic deformations of both specimens occurred mainly by shear in narrow regions extending from the crack tip to the free edge, under about 45° with the direction of the applied load. Near the crack tip, where two such regions converge, extremely high strains are set up, even for small overall extensions. Moreover for the same average stress, the strains and the extension in the specimen with the shorter crack were higher.

BEHAVIOR OF CRACKED PLATES

From the study of the moiré lines it may be concluded that at full yield the plasticity required at the root of a crack is proportional to the uncracked width. This entails further that for specimens with the same width more plasticity is required at the root of a short crack, and that for specimens with different widths, but with the same length of crack, the wider specimens will exhaust more plasticity at the root of the crack than the narrow one.

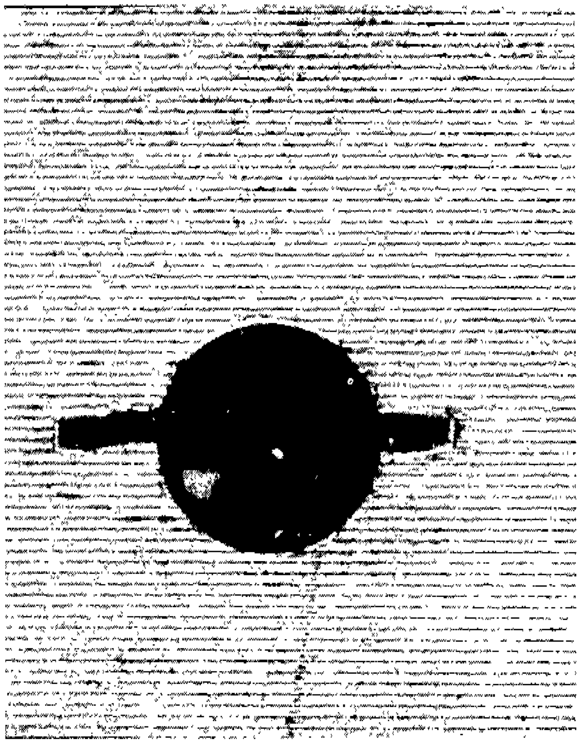


FIG. 8. CLOSE-UP OF THE HOLE, AND SAW CUT. AS THE SPECIMEN IS UNLOADED, THE FATIGUE CRACK IS INVISIBLE.

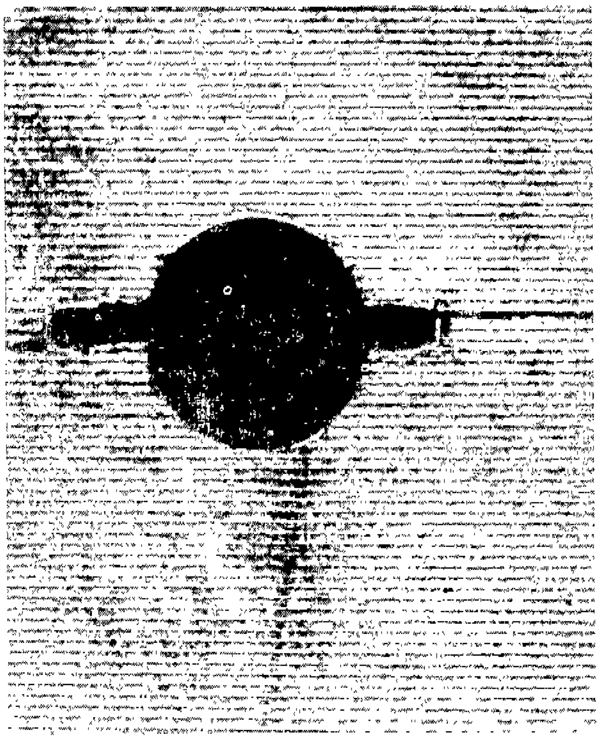


FIG. 9. HOLE, SAW CUT AND FATIGUE CRACK UNDER SMALL EXTERNAL LOAD.

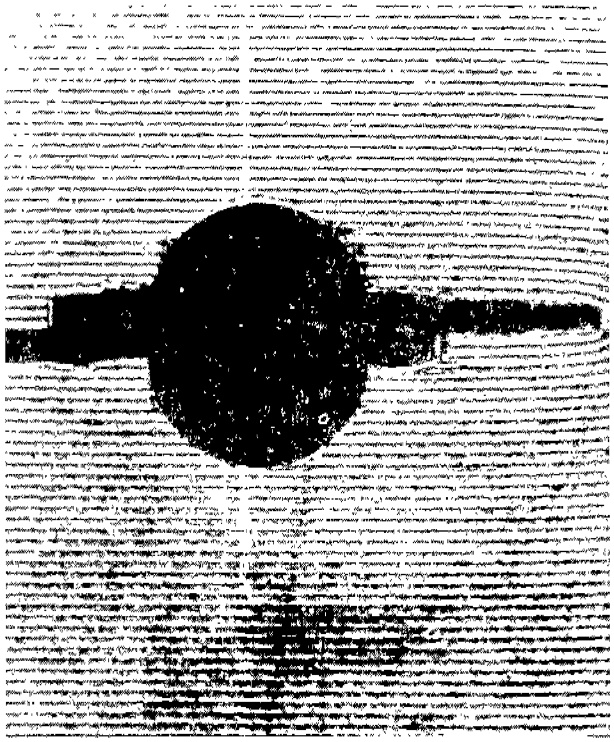


FIG. 10. HOLE, SAW CUT AND FATIGUE CRACK UNDER EXTERNAL LOAD. NOTE THE PLASTIC DEFORMATION AT THE TOP OF THE CRACK.

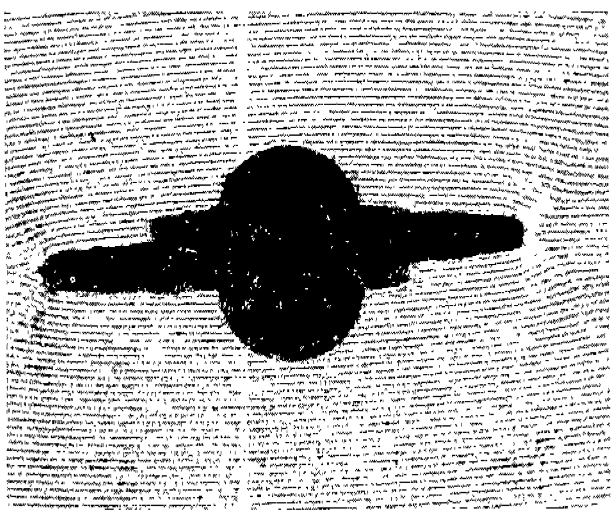


FIG. 11. HOLE, SAW CUT AND FATIGUE CRACK UNDER HIGHER EXTERNAL LOAD. NOTE THE PLASTIC DEFORMATION AT THE TOP OF THE CRACK.

These conclusions influence directly the total elongation and the load at fracture, and therefore are of significance for the engineer. The steel user is indeed unaware of the presence of cracks or notches in his structures, so

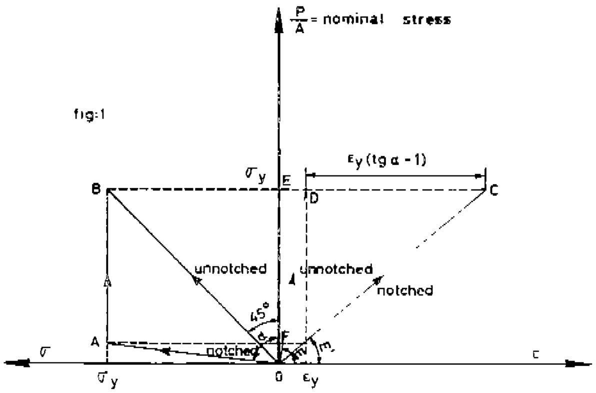


FIG. 12a. On the left side diagram σ versus nominal stress, for unnotched and notched specimen.

On the right side diagram ϵ versus nominal stress, for unnotched and notched specimen.

The strain which must be available for full yielding is given by

$$\epsilon_y \; (\text{tg } \alpha - 1)$$

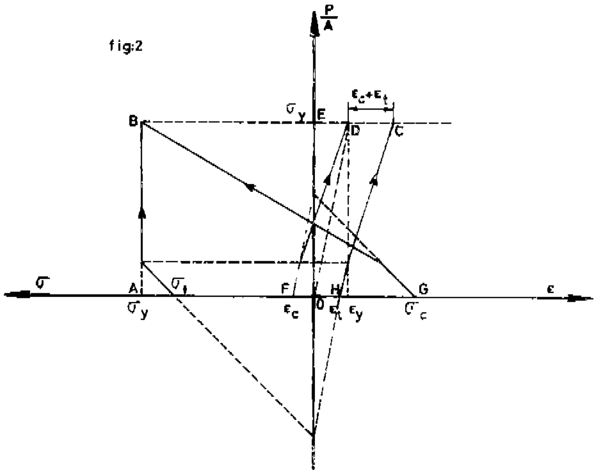


FIG. 12b. On the left side diagram σ versus nominal stress, for unnotched specimen with residual stresses (σ_t and σ_c).

On the right side diagram ϵ versus nominal stress, for unnotched specimen with residual stresses or strains (ϵ_c ϵ_t)

The strain which must be available for full yield is given by

$$\epsilon_c + \epsilon_t$$

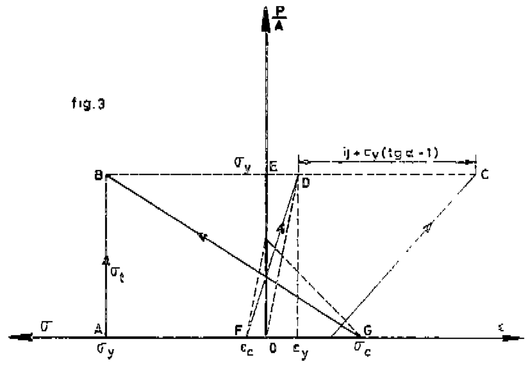


FIG. 12c. On the left side diagram σ versus nominal stress, for notched specimen with residual stresses of yield point value.

On the right side diagram ϵ versus nominal stress, for notched specimen with residual strains y.

The strain which must be available for full yield is given by

$$y + \epsilon_y \; (\text{tg } \alpha - 1)$$

it is necessary to interpret these test results in terms of nominal stresses. He also expects as a minimum requirement that his structure will sustain a nominal stress equal to the full yield strength of the steel used. But this supposes that at the root of all flaws enough plasticity should be available. As we have seen, the required plasticity depends on the width of the specimen and the length of the crack. If this plasticity is not available a crack will initiate and will propagate if Robertson's conditions are met. The nominal strength at fracture therefore depends on the available plasticity, thus also on the width of the specimen and the crack length. A schematic diagram of the elastic-plastic behavior of a crack flat specimen is given in Fig. 12. In this diagram we plotted nominal stress versus local stress (left diagram) and versus strain (right diagram). As long as the whole specimen behaves elastically the stress at the tip of the flaw moves up along this line OA. When the yield stress is reached the local stress does not change until full yield occurs. The amount of plasticity required during first yield and full yield can be estimated on the diagram for nominal stress versus local strain. The elastic deformation at the tip of the flaw is given by the straight line OC. If we now assume that during further loading up to full yield the amount of plastic deformation at the tip of the flaw remains proportional to the applied load,

Influence of crack length (yield stress at +20°C=26kg/mm²)

Temperature	Thickness	Width	Crack length	Nominal fracture stress
0°C	mm	mm	mm	kg/mm²
-20	14	140	114	42,3
-20	14	140	10	30,2

FIG. 13. INFLUENCE OF CRACK LENGTH.

Influence of specimen width (yield stress at +20°C=26kg/mm²)

Temperature	Thickness	Width	Crack length	Nominal fracture stress
0°C	mm	mm	mm	kg/mm²
-20	14	140	10	30,2
-20	14	50	10	36,6

FIG. 14. INFLUENCE OF SPECIMEN WIDTH.

then the plastic strain when full yield occurs will be given by the expression ϵ y(tg α - 1) in which tg α is the so-called shape factor of the flaw. *Figure 12b and 12c give the same diagram* respectively for an unnotched specimen with residual stresses and for a notched specimen with residual stresses. However as we can deduce from the moiré line pictures, the reality is more complicated than that; this schematic diagram gives indeed no answer on the influence of length of the crack or width of the specimen. The assumption of proportionality between load and plastic strain at the tip of the flaw seems to be too simple. It must perhaps be corrected so that the local plastic strains in a wide plate with a short crack should be larger than in a narrow plate with a long crack.

Kihara[5] has carried out tensile tests on wide plates with different crack lengths: all his specimens failed before full yield occurred. He found that fracture stress decreases slightly with increasing crack length; on the contrary we found on rather narrow specimens which all failed after full yield occurred that the fracture strength increases with increasing crack length or decreasing specimen width (Fig. 13-14).

If there can be some doubt about the influence of crack length on the fracture strength, all authors agree on the deleterious effect of the width of test specimens. It is indeed interesting to note that independently of crack length all wide steel specimens fail with very low overall elongation. It is even possible to reduce the nominal fracture stress below the yield stress, by lowering the test temperature.

The lack of any appreciable plastic deformation in wide plates with severe stress concentration has also been ascertained by the American researchers W. J. Hall and coworkers.[6] They found very low elongation on their heat-treated welded wide plates and they claim ..."post-heated specimens, did not exhibit much gross deformation before fracture. This lack of sizable deformation is thought to be an important observation", and further "the small amount of deformation is felt to be an important observation in connection with evaluating the benefit afforded by thermal stress relief, the reasons why no more deformation does not result merits additional study." According to our observations heat treatments can increase the plasticity at the root of the crack. However even after normalizing wide plate the ductility at the tip of flaws will still be insufficient to reach full yield before cracking occurs. The reason of the lack of gross deformation is indeed governed by the width of the plates, as shown by the moiré pattern.

Another experimental confirmation of the effect of specimen width is given in the SSC report 135 by J. Ludley and D. Drucker,[7] entitled: "Size Effects in Brittle Fracture of Notched Steel Plates in Tension". The results published by these authors are perhaps more illustrative because they embrittled the steel before fracturing it in tension. The effect of such a treatment was that all plates wider than 6 in. two-thirds broke brittle at nominal stresses far below the yield stress, while in smaller specimens no fracture appeared before full yield.

From both these reports it seems clear that the small overall deformation of a wide plate or, the large overall deformation of a narrow plate with flaws are essentially not influenced by such drastic operations as heat treatments or embrittlement treatments. Size effect seems to give a satisfactory answer.

From these considerations we can conclude that the presence of cracks, notches or flaws

Influence of prestraining by tension
Tension test at -20°C
Natural strains $2\ln\frac{d_0}{d_1}$

A. Unnotched specimens

Steel	Prestraining 0 %	Prestraining 10 %	Reduction %
A	83,4	67,6	19
B	95,6	76,0	21
C	86,4	75,4	13

B. Notched specimens

Steel	Prestraining 0 %	Prestraining 10 %	Reduction %
B	8,38	4,4	47
A	9,80	5,52	43

FIG. 15. INFLUENCE OF PRESTRAINING BY TENSION.

Influence of temperature Prestraining 10%
Natural strains $2\ln\frac{d_0}{d_1}$

Temperature of prestraining 0°C	steel B unnotched %	steel B notched %	steel C unnotched %	steel C notched %
20	76,0	4,4	75,4	11,6
100	78,8	—	59,6	6,6
200	76,0	5,6	54,6	7,6
300	71,0	5,6	66,0	5,4
400	74,2	3,2	75,0	10,0
500	78,2	8,4	74,8	14,4

FIG. 16. INFLUENCE OF TEMPERATURE PRESTRAINING 10%.

in a structure is perhaps the most important reason of the lack of overall elongation at fracture. This reduction of deformation is enhanced by low temperatures and larger width of the specimen. Combination of both factors may even entail fractures for nominal stresses below the yield stress. For this reason good design and non-destructive testing are the best weapons to safeguard structures against low-stress brittle fractures.

BEHAVIOR OF EMBRITTLED STEEL

Up to now we considered only the effect of a geometrical discontinuity such as a crack, a flaw or a notch, but it is most obvious that properties of the steel play an important role in the fracture behavior of a steel structure. Steel can be embrittled before being put into service by several mechanical or thermal treatments.

It is well known that some types of steels are very prone to aging; that aging can play a role is proven by the effect time has between the moment of straining and testing.

Prestraining in tension at room temperature reduces the elongation at fracture for at least a percentage equal to the amount of prestraining. Fig. 15 gives some tensile test results obtained at -20 C on a material before and after straining of 10%; the results obtained indicate that the loss of plasticity is higher than the prestraining of 10%. When the tension tests are carried out on notched specimens the loss in ductility is even higher and reaches average values of 45%.

The effect of straining is enhanced when the tensile tests are carried out perpendicular to the strained grains; this can be done for instance by tension tests on specimens longitudinally strained by compression or transversally strained by tension. In both cases the grains are flattened and all inclusions, precipitates and holes are stretched perpendicular to the tension direction. This orientation has of course a very severe notch effect and hence severely embrittles the steel. This has been shown by Körber and co-authors in 1943.[8] The authors prestrained steel specimens over various amounts respectively by longitudinal or transverse compression. The results of tensile tests on unnotched specimens show clearly the difference in ductile-brittle behavior between the longitudinally and transversely strained specimens.

The material is even more severely embrittled if prestraining is carried out at high temperature. Fig. 16 gives some results of such tests and one can see that the most detrimental temperature range is situated somewhere between 200 and 400 C. Strains at fracture of only 1% on notched specimens tested at -20 C have been obtained by prestraining the material in tension up to 40% at 300 C.

Similar results were found by Körber and co-authors, as can be seen on Fig. 17 and 18. More recently Drucker and co-authors confirmed

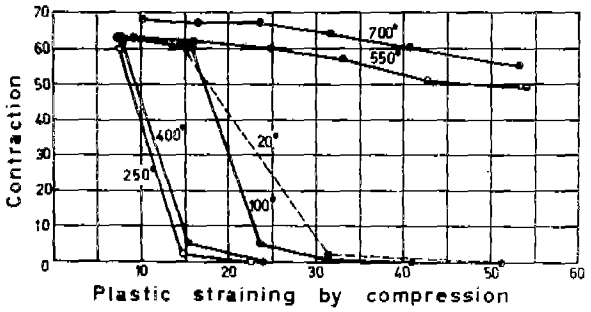


FIG. 17. EMBRITTLEMENT OBTAINED BY STRAINING AT HIGH TEMPERATURES.[8] OPEN-HEARTH STEEL.

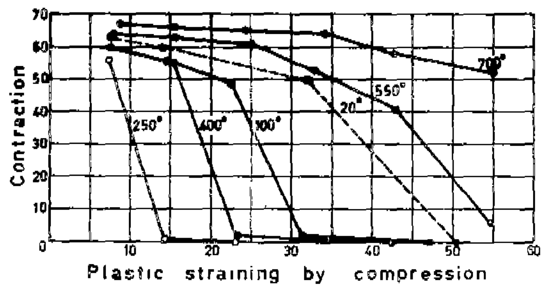


FIG. 18. EMBRITTLEMENT OBTAINED BY STRAINING AT HIGH TEMPERATURES.[8] BASIC BESSEMER STEEL.

these results and proposed a reversed-bend test as a practical reception test.[9-10] From these tests one may conclude that ordinary mild steel prestrained to 40% at 300 C becomes dangerously brittle at lower temperatures.

It is obvious that if such an embrittled material is placed at the tip of a flaw in a wide plate this can lead to very low fracture strengths.

Of course overall prestraining up to 40% is not common practice although it is done in bending steel bars (for instance, for concrete) and forming sheet metal. It is well known that straightening such bent specimens often gives rise to premature fractures. But what is even more important for us is that a prestraining of the order of 40% can easily be obtained when a steel specimen is strained in which flaws are present. As already shown before the plastic strain will concentrate around the discontinuity and it is quite possible that in the immediate vicinity of the flaw strains of more than 40% will occur. Therefore one must avoid straining plastically metal parts in which flaws could be present as for instance in welds.

It must be mentioned that plastic straining can also be produced by local heating, although a maximum of only 2% can be obtained in this way, a severe damage by recycling is possible. Moreover the metal is first strained in compression during heating then in tension during cooling. It is possible by repeating this cycle to obtain spontaneous cracking even during cooling to room temperature especially if strain raisers are present.

Why steels embrittle more in the temperature range of 200 to 400 C is still a mystery; all the more their normal properties are restored when heated above 400 C, as we could see on the Fig. 19. Moreover soaking time does not seem to have any appreciable influence on the ductile behavior. Lagasse[11] has also confirmed the usefulness of such a relatively low-temperature treatment on thin plates (Fig. 20). This metallurgical nature of the embrittlement is actually the subject of research in different laboratories. Investigations with the optical microscope have been unsuccessful to detect any metallurgical phenomenon responsible for this type of embrittlement. However the electron microscope with its resolution approaching the unit cell dimensions of metals might be the right instrument to probe the true nature of the metallurgical damage. An investigation is currently carried out in our laboratory using the microfractographic technique with carbon extraction replicas. Specimens of an ordinary mild steel were slowly pulled in tension to fracture or 50% compressed at temperatures ranging from 150 to 350 C. Carbon extraction replicas were made from metallographic samples taken from the 15 and 30% strained portion and from the necked section of these tensile specimens. We machined also small, notched specimens from the same areas and broke those by impact at room temperature. We then prepared carbon extraction replicas from the brittle-fracture surfaces. Examination of these replicas under the electron microscope revealed several interesting features.

The virgin metal broke almost entirely in a cleavage or transgranular mode. Such a brittle cleavage fracture in virgin mild steel can easily be recognized by the numerous river patterns and tongues, Fig. 21. These river patterns originate from steps on the cleavage plane along which cracks, propagating at different levels, are linking up. The tongues show local deviation of the crack from the

Restoration of embrittled steel
(10 % pretension at 300°C)
Tension tests at -20°C. Natural strains

A. Unnotched specimens

Virgin material	Heated during 1 hour at °C		Virgin material	Heated during 50 hours at °C	
83,2	500	77,2	78,8	500	75,6
88,0	575	72,6	83,6	575	76,0
84,8	650	81,4	92,6	650	81,2

B. Notched specimens

10,4	500	10,8	12,2	500	(5,1)
14,2	575	9,3	15,3	575	9,9
16,0	650	10,5	13,2	650	11,9

FIG. 19. RESTORATION OF EMBRITTLED STEEL.

cleavage plane caused by the presence of twinned lamellae in the matrix. Some areas show also the normal ductile cups with nonmetallic inclusions, such as sulfides, oxides or silicates, in the bottom (Fig. 22, 23).

When examining the fractures in the specimens previously deformed at 150 to 350 C it became quite clear that the mode of failure gradually changes with temperature and amount of deformation from a predominantly transgranular fracture to a predominantly intergranular fracture. Moreover the appearance of the transgranular fracture changes, e.g.: no tongues could be found on the cleavage facets in material previously deformed at temperatures above 200 C, Fig. 24. The intergranular fractures, Fig. 25, 26, are characterized by a rather irregularly undulated surface often showing small precipitates. Moreover, these inter-

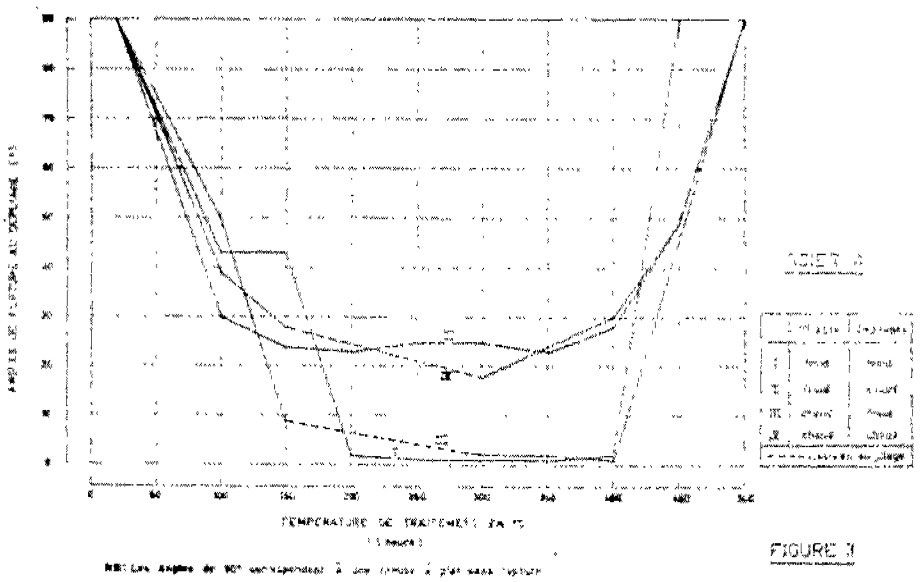


FIG. 20. RESTORATION OF DUCTILITY BY HEAT TREATING AT DIFFERENT TEMPERATURES. ANGLE OF BENDING AFTER RESTORATION VERSUS TEMPERATURE OF HEAT TREATMENT.

I : Cold bended - heat treated - cold straightened
II : Cold bended - heat treated - warm straightened
III : Warm bended - heat treated - cold straightened
IV : Warm bended - heat treated - warm straightened[1]

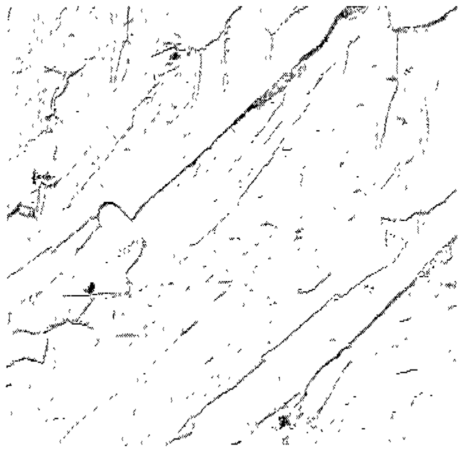


FIG. 21. BRITTLE FRACTURE AT ROOM TEMPERATURE IN UNDEFORMED METAL. NOTE RIVERS AND TONGUES. x 2000

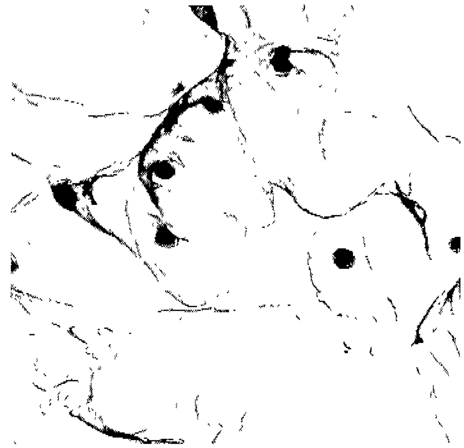


FIG. 23. DUCTILE FRACTURE AT ROOM TEMPERATURE IN MATERIAL COMPRESSED OVER 50% AT 350 C. NOTE CUPS AND INCLUSIONS. x 3200

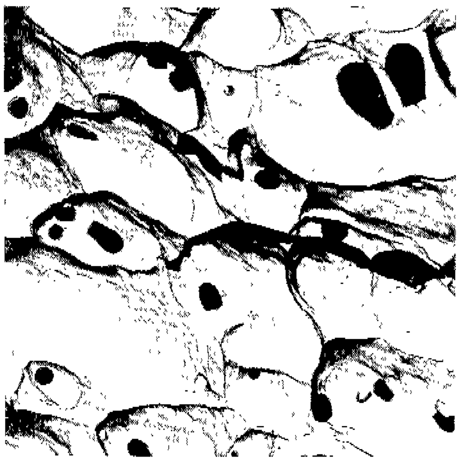


FIG. 22. DUCTILE FRACTURE AT ROOM TEMPERATURE IN MATERIAL PULLED IN TENSION 15% AT 350 C. NOTE CUPS AND INCLUSIONS (SILICATES ?). x 3200

FIG. 24. BRITTLE FRACTURE AT ROOM TEMPERATURE IN MATERIAL COMPRESSED BY 50% AT 350 C. NOTE RIVERS. NO TONGUES. x 1600

granular fractures are often decorated with fragments of a precipitate which is most likely cementite, Fig. 27, 28. Similar precipitate particles can easily be studied on the carbon extraction replicas made from the metallographic samples. They usually consist of large sheets, Fig. 29, which envelope the ferrite grains and have been damaged and shredded apart during the deformation process. The extraction replicas made from the samples treated at temperatures of 350 C show much more of those intergranular carbide precipitates than the virgin metal or samples treated at the lower temperatures. It is most likely that these precipitates grow during the deformation of the metal at higher temperatures. Indeed, one can observe dendritic branches growing out of the intergranular plates and joining up, Fig. 30, 31, 32. These dendritic branches are better developed in severely strained material at the highest temperatures. It is most likely that the presence of these intergranular carbide particles precipitated during the high temperature deformation process are directly

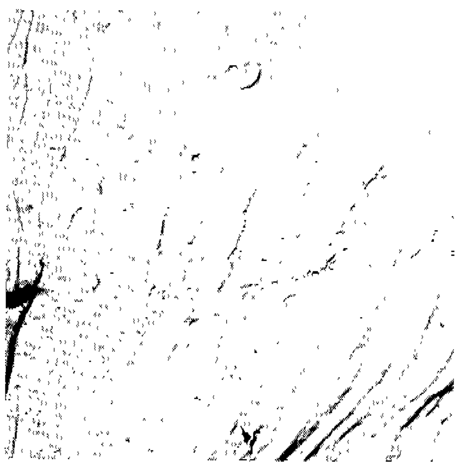


FIG. 25. INTERGRANULAR FRACTURE IN MATERIAL COMPRESSED OVER 50% AT 350 C. UNDULATED SURFACE. x 6400

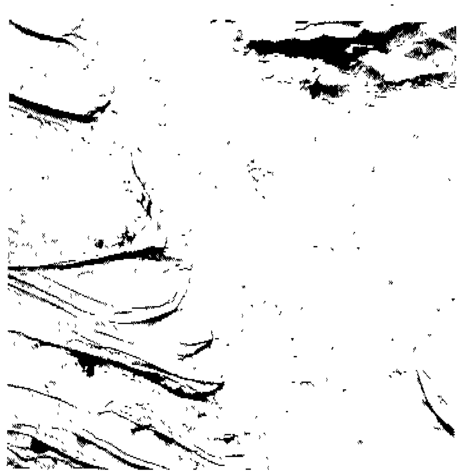


FIG. 27. BRITTLE FRACTURE AND INTERGRANULAR FRACTURE (+ CARBIDES) IN MATERIAL PULLED 15% IN TENSION AT 350 C. x 3200

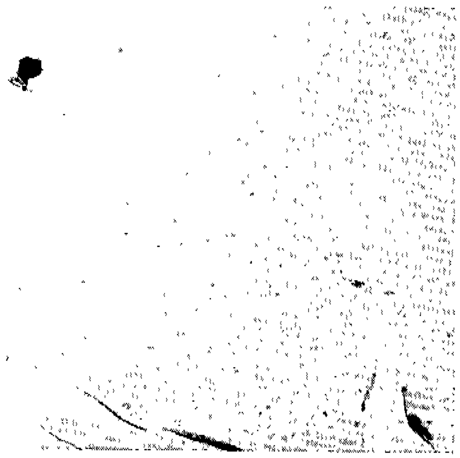


FIG. 26. INTERGRANULAR FRACTURE IN MATERIAL COMPRESSED OVER 50% AT 250 C. NOTE INTERGRANULAR VERY SMALL PRECIPITATES. x 6400

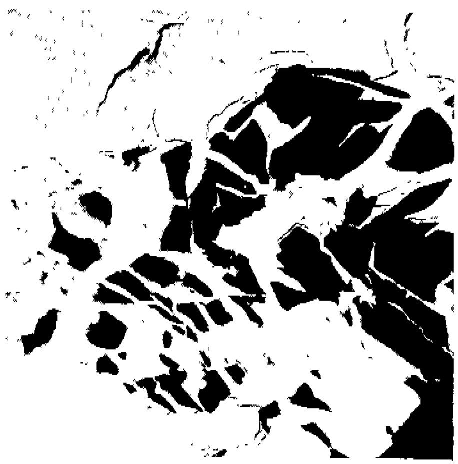


FIG. 28. INTERGRANULAR FRACTURE AND CARBIDE IN MATERIAL COMPRESSED 50% AT 350 C. x 1600

responsible for the intergranular weakening and thus room temperature embrittlement of the material. However one also finds that numerous small intergranular dendrites simultaneously precipitate in the matrix (Fig. 33). They may to a certain extent lower the ductility of the metal.

BEHAVIOR OF THE WELDED WIDE PLATE

Till now we have not yet spoken about the influence of welding on low-stress fractures, although the low brittle strength was ascertained by Wells on welded plates. Since then his tests have been widely repeated in the United States (University of Illinois) and in Japan (Tokyo University). Wells used a specimen in which a saw cut was produced before welding, but at Urbana and Tokyo specimens in which the saw cut was produced after welding were also included. Specimens with the saw cut before welding produced systematically low-stress brittle fracture, while specimens with the saw cut after welding gave rather erratic results.

The behavior of such specimens can now easily be explained. If we consider (Fig. 34) the wide plates to be welded, we can assume to simplify our problem - that zone I is heated, while zone II remains at room temperature. By heating, zone I expands and if it was entirely

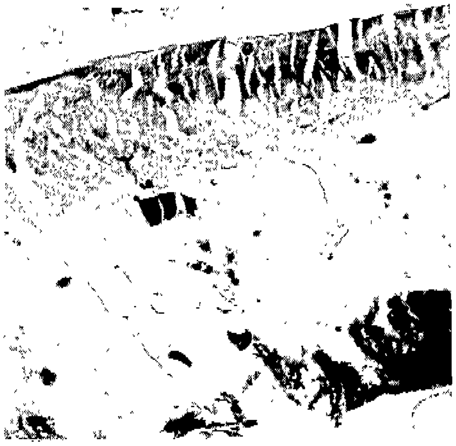


FIG. 29. EXTRACTED INTERGRANULAR CEMENTITE (CARBIDE) PLATE, AND PEARLITE NEST (BELOW). PULLED IN TENSION (NECKED REGION) AT 350 C. x 1600

FIG. 30. EXTRACTED PARTICLE FROM GRAIN BOUNDARY (CEMENTITE) IN MATERIAL PULLED IN TENSION 30% AT 300 C. x 9600

free it would take a length

$\ell_1 = \ell_0 \ (1 + \alpha \ t)$

However this zone cannot expand freely, which implies that I will be compressed and II expanded so that both will have the same length ℓ'_1

$\ell'_1 = \ell_1 \ (1 - \epsilon_I) = \ell_0 \ (1 + \epsilon_{II})$

$$\epsilon_I = \frac{\alpha \ t - \epsilon_{II}}{1 + \alpha_t} = \alpha_t = \epsilon_{II}$$

FIG. 31. EXTRACTED PARTICLE FROM GRAIN BOUNDARY (CEMENTITE) IN MATERIAL PULLED IN TENSION 15% AT 350 C. x 6400

FIG. 32. EXTRACTED PARTICLE FROM GRAIN BOUNDARY (CEMENTITE) IN MATERIAL PULLED IN TENSION 15% AT 300 C. x 4000

II will be small if the plate is wide and stiff. In this case we may even assume that

$\epsilon_I = \alpha_t$

During heating the zone I will be strained in compression over an amount of $\alpha_t = 2\%$. During cooling the same reasoning shows that zone I will be strained in tension over an amount of 2% (Fig. 35).

During welding this mechanical-thermal cycling will be repeated several times and will result in an embrittlement of the steel. If a saw cut has been introduced in zone I

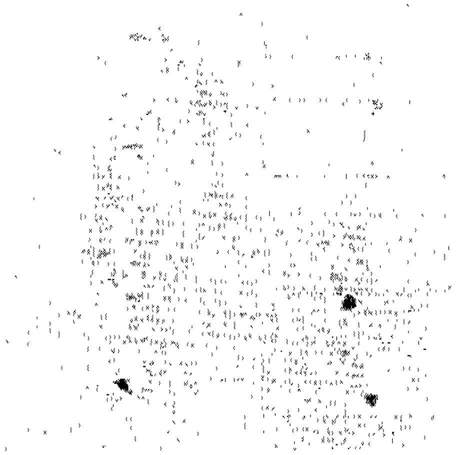


FIG. 33. INTERGRANULAR PARTICLES IN MATERIAL PULLED IN TENSION (NECKED REGION) 350 C. x 16000

before welding much higher straining will occur in the vicinity of the tip of the notch and serious metallurgical damage must occur. Occasionally spontaneous cracking has been observed.

When the saw cut is introduced after welding, the embrittling will be much less severe, but the width of the specimen and overall embrittling may be sufficient to obtain even in this case low-stress fractures. Tests confirm this conclusion: specimens notched after welding behave generally much better than specimens notched before welding.

It may also be concluded from these considerations that the welding sequence has a serious influence on the ductility of the metal in the heat-affected zone. Each welding run entails a mechanical and thermal cycle which embrittles the steel. The worst conditions to embrittle the base metal are obtained by using small electrodes and waiting after each run till the metal is cooled.

The importance of specifications will be illustrated by the following facts: in Europe it is usual to assess the ductility of welds by Charpy tests on the weld metal. In order to obtain the required figures electrode manufacturers are compelled to use multipass welding techniques. As a consequence of this technique the base metal is embrittled and all advantages of automatic welding are practically precluded.

A good welding technique must try to obtain equivalent ductility in weld and base metal and, to satisfy this condition, the heat input should be adequately adjusted.

A wide welded plate with a flaw in the vicinity of the weld will behave brittle for two reasons: one due to the flaw and the width of the plate and the other due to embrittlement caused by welding.

As already mentioned before, the first reason can be kept under control by good design and non-destructive testing so that strain concentrations are avoided but at the same time the severity of embrittlement is reduced during welding. For this reason again the value of good design and workmanship cannot be overemphasized. The only contribution the researcher can do, is to reduce embrittlement by a judicious balance of heat input during welding.

To avoid tests on wide plates which are very expensive, we use a special small specimen which allows us to respect the conditions necessary to embrittle the steel namely stiffness and temperature distribution.[12] This specimen is shown in Fig. 36; its width is only 140 mm ≈ 6 in.; two longitudinal slots are machined in the specimen. The stiffness can be expressed by the ratio of the areas: A_c/A_W where A_c and A_W are respectively the areas of the columns and the web of the specimen. While heating the web the columns are kept at room temperature by water cooling. The strain concentration is obtained by a hole drilled in the web, which is then extended by a saw cut which in turn is elongated by a fatigue crack. From the results obtained so far, we may conclude that:

a) No low-stress brittle fracture can be obtained if the cracked section has not been prestrained. This prestraining is obtained by heating the web uniformly with electrical resistance elements. Below a certain heating temperature no low-stress fractures occur when the specimen is pulled in tension.

b) Specimens with the same magnitude of residual stresses (which can easily be measured by putting strain gages on the columns) but which have not been prestrained at critical temperatures could not be fractured at low stress. These specimens were obtained by welding the web in the frame; while the critical pre-cracked section was put in cold water.

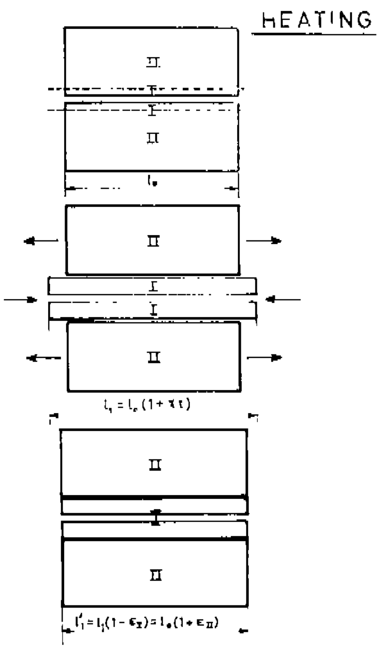


FIG. 34. WELDING OF WIDE OR STIFF SPECIMENS ENTAILS DURING HEATING PLASTIC STRAINING IN COMPRESSION.

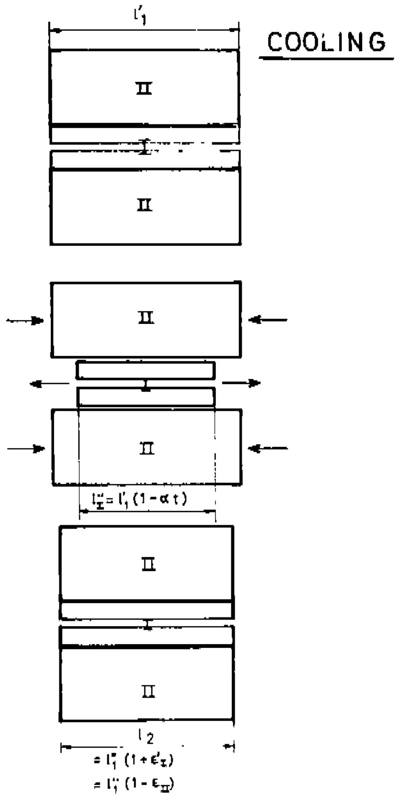


FIG. 35. WELDING OF WIDE OR STIFF SPECIMENS ENTAILS DURING COOLING PLASTIC STRAINING IN TENSION.

c) Some of the welded specimens failed at low stress; this occurred only when different weld runs were laid in the joint or when the critical section was not cooled during welding.

The results of these tests have convinced us that prestraining at a critical temperature obtained by local heating may have a deleterious influence on the ductile behavior of steel, moreover that welding in certain conditions may be directly responsible for the embrittlement.

FURTHER RESEARCH

Guided by these results we hope in a very near future to study the following:

1. The mechanical and thermal cycling which occurs during welding. We want indeed to have more information about the strain and temperature history of the metal in the heat affected zone. Temperature recording is no problem. Strain recording is more difficult. We intend to use the moiré technique in order to have an overall view of the plastic straining. But high temperature necessitates engraving the grid in the steel and special optical equipment to record the moiré pattern during heating and cooling.

2. For a given steel, its brittle ductile behavior when it has been submitted to strains and temperatures recorded during the above investigation. Eventually this study must give us inquiries about strains and temperatures which must be avoided. We hope to realize these strains and temperatures on a hot ductility machine. We hope in this way to be able to give indications about how to balance the heat input during welding in order to save a maximum of ductility of the steel employed.

3. The effect of crack length and plate width on the plastic strain patterns at the tip of a crack and its influence on brittle behavior of the plate.

4. Various steels in more detail regarding the metallurgical nature of the damage caused by straining at different temperature. This investigation will be carried out under the electron microscope with the aid of the extraction

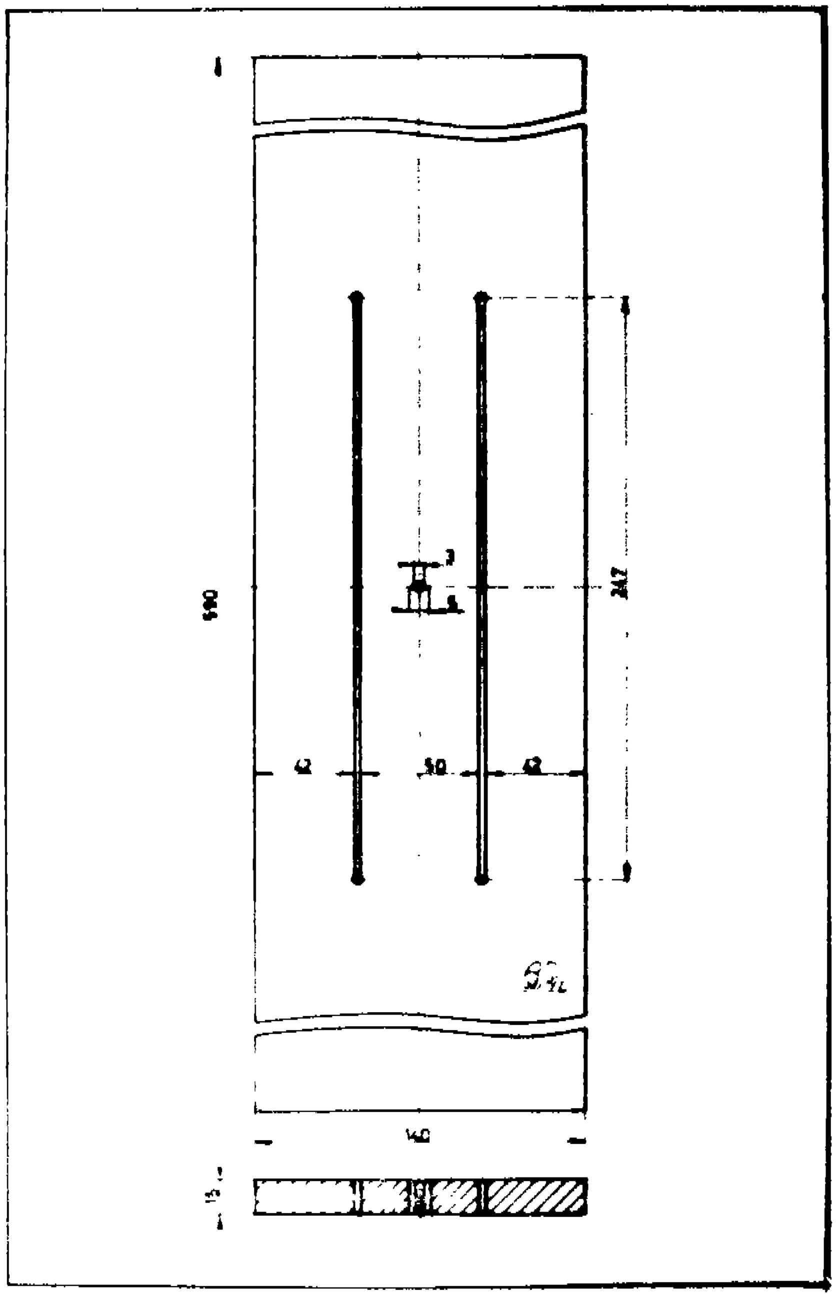


FIG. 36. FRAME-TYPE SPECIMEN USED TO STUDY THE CONDITIONS NECESSARY TO PROVOKE LOW-STRESS BRITTLE FRACTURE.

replica technique on fracture surfaces and the thin-foil technique.

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