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SSC-140

A NAVAL ARCHITECT'S REFLECTIONS ON SOME RESEARCH PROBLEMS WITH SHIP STEEL

by

Georg Vedeler

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ADDRESS CORRESPONDENCE TO:

SECRETARY
SHIP STRUCTURE COMMITTEE
U. S. COAST GUARD HEADQUARTERS
WASHINGTON 25, D. C.

August 4, 1961

Dear Sir:

Dr. Georg Vedeler, Managing Director of Det Norske Veritas, Oslo, Norway, recently participated in a meeting 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 A Naval Architect's Reflections on Some Research Problems with Ship Steel, was prepared by Dr. Vedeler to summarize his remarks for the Committee on Ship Steel. This report is being distributed by the Ship Structure Committee because it represents an important current approach to some of these problems.

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

Sincerely yours,

J. A. Alger, Jr.
Rear Admiral, U. S. Coast Guard
Chairman, Ship Structure Committee

Serial No. SSC-140

Special Report

on

A NAVAL ARCHITECT'S REFLECTIONS ON SOME RESEARCH PROBLEMS WITH SHIP STEEL

by

Georg Vedeler
Managing Director
Det Norske Veritas
Oslo, Norway

Washington, D. C.
National Academy of Sciences-National Research Council
August 4, 1961

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INTRODUCTION

I have been asked to speak to a group of the greatest steel experts in America, and I feel greatly honoured. But I am afraid that I shall not be able to tell you anything which you do not know before. I come from a small country with only an insignificant steel-making industry. Our industry mainly works in other directions. Norway may be better known as an owner of merchant ships, which have been built in many different countries.

As a naval architect I have been occupied with designing and supervising ships. I therefore perceive my task here to be an attempt to tell you about some of my experience with ships in relation to the steel they have been built of and to try to suggest some problems of research which may be of mutual interest to your steel industry and our ships.

The importance of fatigue in ships

I begin by showing two slides, Figs. 1 and 2, which are diagrams meant to give an idea of the number, size and position of cracks which have been reported during the 4 years ending in September 1960 in shell and deck of ships more than 400 feet long and built after the second world war. Figure 1 concerns tankers, Fig. 2 dry cargo ships. The latter have usually five big rectangular hatches in the deck and machinery amidships, while the tankers have about 30 small oval or elliptical hatches and machinery aft.

When these diagrams were made we had in our files damage cards for 210 tankers of the size mentioned. 66 of them had had cracks in shell or deck. They have all been shown in the diagram. Each of the numbers in small circles represents a ship. Numbers 1 to 49 are ships built before 1 January 1954, while numbers 50 to 66 represent ships built after that date. The ships from before this date were built to rules which had existed unaltered since before the war. In 1953 we had new rules printed including special requirements to the chemical composition of and the Charpy V-notch impact value for certain plate strakes and also improvements to the design. Now of these 66 ships there is only one, viz. No. 37, which has had a brit-

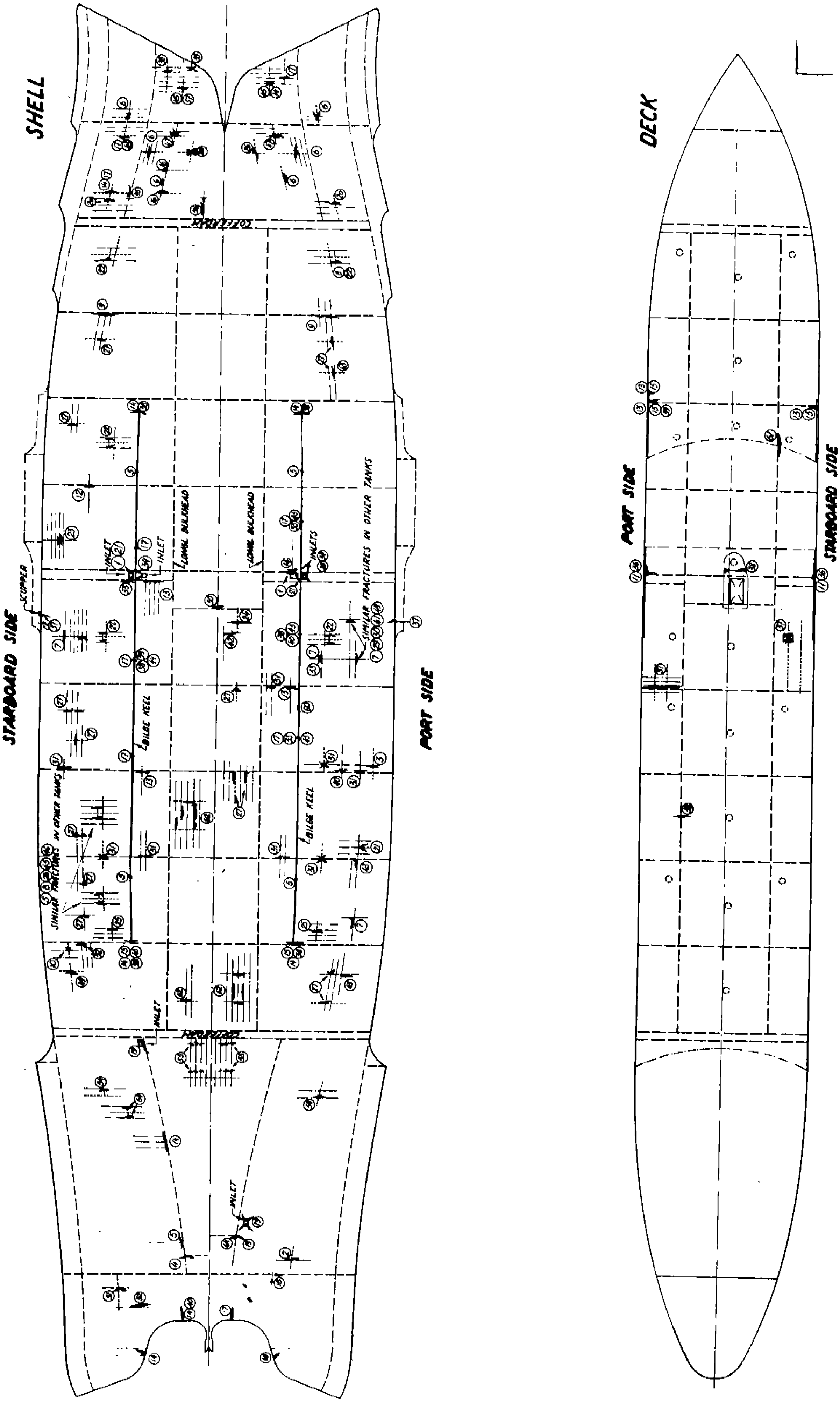
SHELL
STARBOARD SIDE
PORT SIDE
SCUPPER
INLET
INLETS
LONGL BULKHEAD
BILGE KEEL
SIMILAR FRACTURES IN OTHER TANKS
DECK
PORT SIDE
STARBOARD SIDE

Fig.1.

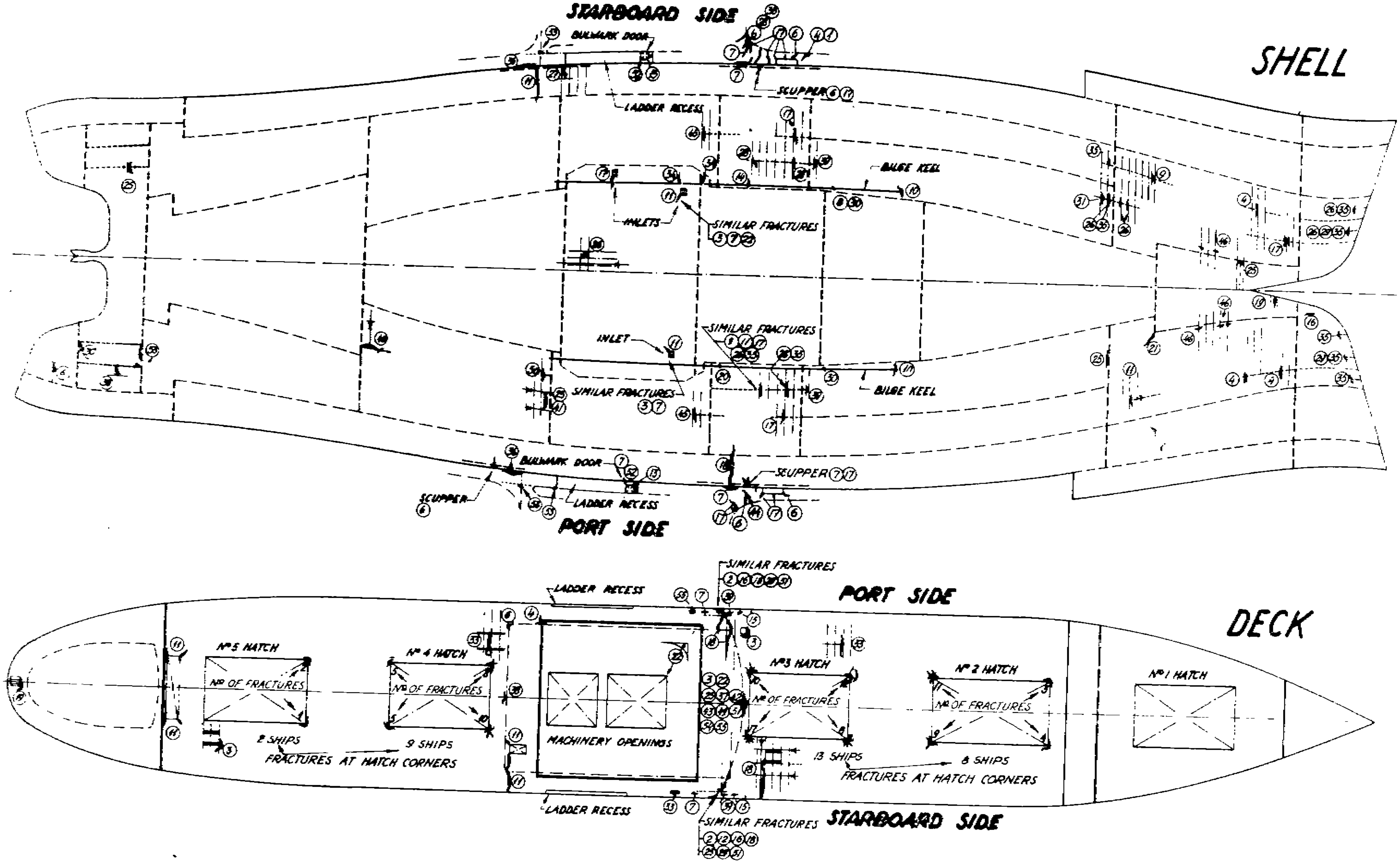
STARBOARD SIDE
SHELL
BULWARK DOOR
SCUPPER
LADDER RECESS
BILGE KEEL
INLETS
SIMILAR FRACTURES
INLET
PORT SIDE
DECK
No 5 HATCH
No 4 HATCH
No 3 HATCH
No 2 HATCH
No 1 HATCH
No OF FRACTURES
MACHINERY OPENINGS
2 SHIPS
9 SHIPS
FRACTURES AT HATCH CORNERS
13 SHIPS
8 SHIPS

Fig. 2.

tle fracture. This ship was finished in 1945, immediately after the war. The cracks in the other 65 ships were probably all due to fatigue and appeared at points of stress concentration.

Figure. 2 for dry cargo ships gives a similar picture. It has emerged from the study of damage cards of 144 such ships all built after World War II. 55 of the ships had cracks in deck and shell, all shown here. 44 of them were built before 1 January 1954. Only two of the ships, viz. Nos. 11 and 18, had brittle fractures. They were both finished in 1949. The cracks in the other 53 ships were, we believe, due to fatigue.

The two figures, which concern ships built to Norske Veritas class, give a similar picture as equivalent figures published by Hodgson and Boyd in the Trans. I.N.A. 1958 for ships classed with Lloyd's Register. The Lloyd's ships may have had a somewhat larger percentage of brittle fractures, but as far as I can judge the overwhelming number of cracks also in those ships seem to have been due to fatigue.

My conclusion is that from a practical point of view the problem of brittle fractures in ships has been solved by the present regulations. We still, of course, have to keep an eye on the problem for ships sailing in extraordinary cold climates or with very cold cargo like e.g. liquified gas. But for ordinary ships in ordinary waters the problem for 95 per cent of the cracks appearing in open water seems to be high-tension low-cycle fatigue. As shipbuilders we would have less trouble if steel makers could offer us a ship steel with better qualities in respect of this type of fatigue and which we could use at points of stress concentration without incurring other risks.

I have to add that we are not very interested in ordinary rotary bending fatigue tests with smooth specimens of finger thickness. We like to see the results of tensional fatigue tests with notched specimens of full plate thickness with two of the surfaces left rough in the as-rolled condition.

I have seen from your yearly report that high-tension low-cycle fatigue is one of the many items on your research program. I should wish it would be given a very high priority. Among the enormous number of fatigue tests which have been published I have up till now found practically none of practical value to a shipbuilder. It seems to be very difficult to differentiate between the different kinds of ship steel with regard to their fatigue properties. With the number of fatigue cracks in our ships it is a pity that we are left in a mist here.

Also in those cases where we have to consider the possibility of brittle fractures, fatigue comes into the picture, because a brittle fracture may well start from a fatigue crack.

Among ship research people in Scandinavia we have started a collaborative program on high-tension low-cycle fatigue in ship steel to try to obtain figures which may be of value to designers. Because of the time needed and the few machines available for the fairly large tests specimens, and also because of the small amount of money which we have been able to raise in each country for this purpose, we have considered it necessary to divide the testing between us, i.e. at the moment between Norway and Sweden, hoping that we shall gradually get also other countries interested. We have started with a fairly long test piece with a number of equal small holes in the centerline. We think we can considerably reduce the number of necessary tests when each specimen has several notches between which the crack can choose.

A great deal of research work is being done--and rightly so--to ascertain the forces and stresses which occur in ships, with the aim to improve design. The question naturally arises: What is the maximum stress which can be allowed? As long as we have points of high stress concentration--and with the rough methods of production used in shipbuilding they seem to be unavoidable--we must at least for these points use a figure for the high-tension low-cycle fatigue obtained with a notched specimen of the material.

During some years large scale fatigue testing has been carried out in Scandinavia to see what could be obtained with different kinds of steel for welded A-brackets or columns for large diesel engines. I have not seen the final results, but I understand from the people who have been doing the job that it is very difficult to draw definite conclusions from the results.

Another experience may be of interest in this connection. A steel manufacturer in Norway has been making cast nickel steel propellers of considerable size for many years. But it has happened several times that such a propeller lost a blade in the open sea. No doubt--and this was confirmed by tests--the fatigue strength of the material decreased considerably in service due to the presence of sea water. Recently some of these propellers have been given protection against corrosion by means of platinized titanium anodes fitted outside the hull, but isolated from it and fed with electric current from a battery. The return current from the propeller passes through brushes on the shaft. As is well known from similar protection of ship steel the propellers after a while become greyish in colour or are covered by a thin grey substance.

Propeller steel protected against corrosion in this way has shown a somewhat higher fatigue strength than the original steel before it was exposed to sea water. It would be interesting to try to find out if the fatigue strength of ship steel can be improved in a similar way. The wetted shell of the ships in question was protected against corrosion by the same arrangement as the propellers and the bare hull steel also got the same grey coating. It is not difficult to protect the under water shell in this way. It only remains to find out if it improves the fatigue strength in comparison with the reduction in fatigue qualities which may be expected where steel is exposed to sea water or moisty sea air.

High yield point and cold-working

It seems to be in vogue to recommend the use of the yield point as the guiding property of the material when choosing the maximum allowable stress. There are of course certain reasons for that. But when the conclusion is drawn that we therefore should aim at using a steel with a higher yield point than the ordinary ship steel so as to be able to reduce scantlings and save weight, I am inclined to say that one should look thoroughly into the matter before one yields to this conclusion. First of all in a welded and not stress annealed structure like a ship there are residual stresses as high as or actually higher than what we call the yield point obtained by an ordinary tension test. The higher the yield point the higher are also the residual stresses. What this means in connection with fatigue I do not know. That may have to be studied. But if one should be close to the transition temperature of the structure with that material, I have no doubt that such a high residual stress may be dangerous. I am inclined to think that for steel with a high yield point one should have a larger margin to the transition temperature and the definition of the transition temperature by means of a Charpy test should be at a higher energy than for ordinary ship steel.

The Belgian Professor Soete in his Houdremont lecture at the International Institute of Welding meeting in Liege in June last year drew the conclusion from some of his experiments that to avoid brittle fracture it would be an advantage to utilize steel with a high elastic limit. But he rightly added that this conclusion is valid only when there is no residual stress and when all the stresses are known. I am afraid we have to admit that this is not the case with ships.

Professor Soete's conclusion was drawn from experiments where he raised the elastic limit of a steel by cold working and thereafter found that the stress necessary to obtain brittle fracture at -20°C was equal to the new and higher yield point, in his case 50 kg/mm^2 (71,000 lbs/sq. in.).

This seems to be contrary to practical experience. In practice there are many examples of cold working which have resulted in brittle fractures at assumedly low stresses. Here it may suffice to remind of one of the early American tests with a 2 feet wide specimen with a slot in the middle and welded at top and bottom to somewhat heavier plates of a better steel. When the specimen broke also one of these heavier plates of better steel broke simultaneously in a brittle way at a stress of only 20 kg/mm^2 (28,000 lbs/sq. in.). This was explained by the statement that the plate had been cold worked by having been cut by an insufficiently sharp shear.*

Comparing this with Soete's experiment I have a feeling that here is something important which has not been fully explained yet. If my guess is right an explanation may, however, be found in some of the results given in N. P. Allen's lecture "The Mechanism of the Brittle Fracture of Metals," printed in the book FRACTURE, Proceedings of the conference in Swampscott in April 1959. A cold-

*N. M. Newmark: "Review of Brittle Fracture Research at the University of Illinois," Ship Structure Committee Serial No. 69, May 17, 1954 p. 197.

worked iron (75% compression) tested transversely to the cold-working was ductile at all temperatures and the yield point increased regularly right down to -200°C. But when tested longitudinally the specimens became brittle around room temperature and the fracture stress decreased considerably with falling temperature. A similar behaviour was found with 80% cold-rolling, but here ductility was retained along the direction of rolling and brittleness occurred during transverse testing.

A third alternative may be cold-working by shear. This may be done in two different directions, either at right angles to the plate surface or by planing parallel to the surface. The question arises whether brittleness will then always occur with testing in tension. There seems to be room for tests of considerable practical importance as a supplement to the tests made by Professor Soete.

With the importance of fatigue and with the amount of shearing done in shipbuilding, e.g. manhole punching or the like, it would also be very valuable to get some information about the influence of cold working on the low-cycle fatigue. The question raised by Soete in connection with brittle fracture may be just as important in connection with fatigue.

Discussion of details of impact tests

As I have tried to explain we have apparently obtained very good results with regard to avoiding brittle fractures in ships which have been built during the last eight or ten years. The reason for this is the present requirements for ship steel of thicknesses above 1/2 inch. In the many countries of Europe and Asia steel production practice differs a great deal. In view of this it has been of great value that the steel mills of these countries have been willing to adopt an acceptance test of the Charpy V-notch type for the better qualities of steel. This is now a daily routine. Just because the impact test is so extensively used it may be of interest to consider some of its details.

The ordinary S-shape energy-temperature curve usually has a lower and an upper part which are nearly horizontal interconnected through a steeper part. Some people refer the upper part to crack initiation, the lower part to crack propagation. But I have also seen the upper part referred to crack stopping, the lower part to crack initiation. There are of course reasons for both these sets of references, but I doubt that they are adequate in all cases. The level of the upper part of the curve may not be very important in connection with brittle fracture. A high energy here may tell something about properties of interest in connection with say cold flanging, which is not necessarily of interest for ordinary ship plates.

The different parts of the energy-temperature curve may also be described by their fracture appearance. In the upper part the fracture is completely ductile, in the lower part completely crystalline, while the intermediate part has a mixture of both. The upper end of the completely crystalline appearance will roughly speaking for ordinary steel coincide with an energy level of some 15 or 20 ft-lb which is often used as a criterion for the transition temperature for rimmed or semi-killed steel. It will probably be higher for fully killed fine grain steel. The mixture of the two types of appearances in the intermediate part of the curve may be of different natures. The appearance may be ductile below the notch and brittle at the opposite side. This is often the case over a large part of the steep curve when the notch radius is large. The opposite, viz. a crystalline part below the notch and a ductile part along the other side, may often characterize the steep part of the curve when the notch is very acute. For intermediate notch acuities there may be several cases with a crystalline appearance in the middle of the fractured surface and a ductile appearance along all four sides.

Now an important observation in connection with acceptance tests is that the lower and upper parts of the curve, characterized by completely crystalline and completely ductile fracture appearances respectively, are well defined with a small scatter in results, while the steep part of the curve characterized by a mixed appearance usually has a large scatter in results. If one defines the required transition temperature by an energy value within the intermediate zone with mixed fracture appearances one may therefore, due to the scatter, run the risk to reject a suitable cast or vice versa accept an unsuitable one, because it will seldon be possible to get a sufficient number of tests to obtain a statistically reliable value in this zone. This is one of the reasons why, for acceptance tests of ordinary ship steel, a fairly low energy value such as e.g. 20 ft-lb should be used in combination with a suitably low temperature.

The Charpy impact test like all other small scale laboratory tests does of course not simulate the full scale conditions in a ship. Transition temperatures mentioned in specifications are meant to distinguish between different steels and may not coincide with the lowest temperature expected to be met in practice. 1 3/4 in. plates of E quality may have to be used for certain strakes in a ship expected to meet a temperature minimum of say zero Centigrade, while 3/4 in. plates of the same quality may be required for internal structures in contact with refrigerated cargo of say -20°C. The so-called transition temperature defined by a certain energy value of say the Charpy V-notch impact test must be considered in relation not only to the plate thickness, but also the type of steel used. And as mentioned before the margin which one desires to have between this transition temperature and the working temperature of the structure may also depend upon the type of steel.

It is well known that impact energy-temperature curves will be shifted to higher temperatures when the bottom of the notch is made sharper. From test results with different steels it looks as if the Charpy V curve, except for its upper part, will nearly coincide with the Schnadt $K_{0.5}$ curve, which has the same notch root radius. This may be surprising because in the Schnadt test the compression side has been substituted by a hardened steel pin and the remaining area has therefore been considerably reduced. Judging from the results the pin does not seem to be of much importance. The compression zone seems to play a subordinate role except when the steel is fully ductile.

For broken notched specimens the appearance is often given a number called percentage crystallinity. As far as I know it was introduced with notched tensile tests, probably because it was difficult to measure the energy with such tests. And it has lately been carried over also to Charpy V-notch impact tests.

With a notched tensile test, say with a notch on each side as with the Tipper specimen, we know that to start with the highest stress will be at the root of the notch. But as soon as the yield point has been reached at the bottom of the notch flow lines will quickly develop right across, from notch to notch, and the highest stress will be in the middle of the specimen. By cutting through test pieces before they break Mrs. Tipper has shown that they actually start to open at the center. This is said to be the case also when the fracture is completely brittle.

In the impact bending tests the flow lines develop from the notch in both directions around a circular central solid pin to meet again near the opposite surface. There is no flow line through the hinge pin, but the specimen usually breaks right through this pin.

When comparing the appearance of broken surfaces of the Tipper test with those of the Charpy test it should be borne in mind that in the first test the material has yielded right through the section where it breaks, whereas in the second test it has only yielded for a short distance right below the notch, but otherwise breaks through unyielded material. Especially for the intermediate part of the curve for the two tests the percentage crystallinity must mean different things and one should be careful when comparing them. I believe it would pay to study this in more detail.

I have read the publications of Mr. Hartbower on his Low-Blow Technique with very great interest. He has no doubt invented an impact test with the sharpest possible notch so that his curve will lie to the right of all ordinary Charpy curves. Very valuable also is the fact that he can cool the specimen between the two blows and thereby get rid of the temperature rise due to the first blow. This makes it possible to refer the test data to a more correct temperature. But I am not entirely convinced that even his low-blow transition temperature rightly deserves the name of a material property. He rids the final act of the test from the size effect of the specimen by a variable first blow. But still other people with say another pendulum weight, another velocity and another machine rigidity may get somewhat different results. And what actually does a material property mean if it cannot be referred to in a calculation of brittle vs. ductile behaviour in a full size structure?

Several investigators have invented other ingenious tests and may for some time have believed that they found a material property. But when they have been able to accumulate a large amount of data containing perhaps thousands of test results with many different steels, they seem sooner or later to have come across inexplainable problems. Steel is such an inhomogeneous material and the variables in its temperature dependence so many that I am afraid one must be very careful before using the term material property in connection with temperature dependent macro tests. We still seem to be some distance away from the complete understanding of the brittle fracture mechanism.

Is the running of a crack a continuous or a discontinuous process?

A threshold in stress or strain must apparently be overcome to start a brittle crack. The initiation of a crack needs a larger energy than the propagation of a running crack. These statements require a kind of continuity in the running of the crack. But there are many observations of discontinuous small cracks, through single grains or through more than one grain, and which exist before the crack starts to run, and which also develop in front of the running crack. Many observers believe that the opening of a running crack consists in uniting the many small and separate cracks which have developed some distance ahead. For these observers the running crack is a discontinuous phenomenon where initiation and propagation follow each other like the links in a chain. There seem, however, also to be other observers who do not believe that the small separate cracks created by the stress field in front of the main crack are able to extend and unite, and that the main crack runs beside them in a continuous process. I am afraid that this is not in accordance with all observations. But I consider it of main importance for the understanding of the brittle fracture mechanism that one should try to reach a decisive conclusion whether the process is discontinuous or continuous.

Importance of defining state of stress for temperature dependent properties of steel

In papers on brittle fracture one sometimes sees two temperature-yield point curves plotted, a lower one for smooth tensile tests, a higher one for tests with notched specimens. One often also hears the statement that the yield point is raised by a triaxiality of stress. To avoid confusion it should always be remembered that yield is due to slip caused by shear stresses the magnitude of which depend upon the differences between the principal stresses. When all principal stresses are the same, there will be no distortion, only geometric similarity before and after application of the stresses, and there can be no yield. A fracture at a sufficiently high stress will always be of the cleavage type.

The theory of distortion energy yielding is usually accepted as sufficiently accurate when the material has not been prestressed in one direction or there are other reasons for anisotropic behaviour. According to this theory the yield point can be expressed by an equivalent stress which is the following function of the principal stresses:

$$\sigma_y = \frac{1}{\sqrt{2}}\sqrt{(\sigma_1-\sigma_2)^2+(\sigma_2-\sigma_3)^2+(\sigma_1-\sigma_3)^2} = \sigma_1\sqrt{1+\xi^2+\eta^2-\xi-\eta-\xi\eta}\ , \text{ where}$$

$\xi = \sigma_2/\sigma_1$ and $\eta = \sigma_3/\sigma_1$. For an ordinary smooth tensile test $\sigma_2 = \sigma_3 = 0$ and $\sigma_y = \sigma_1$. With triaxiality the same σ_y may be obtained with a higher σ_1, but it may be misleading to say that the yield point is higher.

Some years ago the notion triaxiality was used much in discussions on brittle fracture and some people tried to define the word by a formula. In later years one has been more occupied by discussions on energy of the type emerging from an extension of the Griffith theory and on phenomena observed under microscope. These are no doubt very important in the attempt to understand physical behaviour. But the great difficulties in measuring principal stresses should not be sufficient excuse to leave them out of the picture. It may be true that designers do not often know or use principal stresses in their analyses. But nevertheless a good deal of research is being carried out to assist designers in getting a better knowledge of actual forces and stresses. The ultimate aim of research in this field naturally is to get knowledge and give information about which grade of material should be used in each case from a technical as well as from an economical point of view. Designers are no doubt longing for figures which can be used in their calculation.

I do not advocate unnecessary details. But I believe that somewhat more understanding could be given to some research results if the question of triaxiality could also be included. The parameter $\Pi = \sigma_y/\sigma_1$ used by Schnadt in his

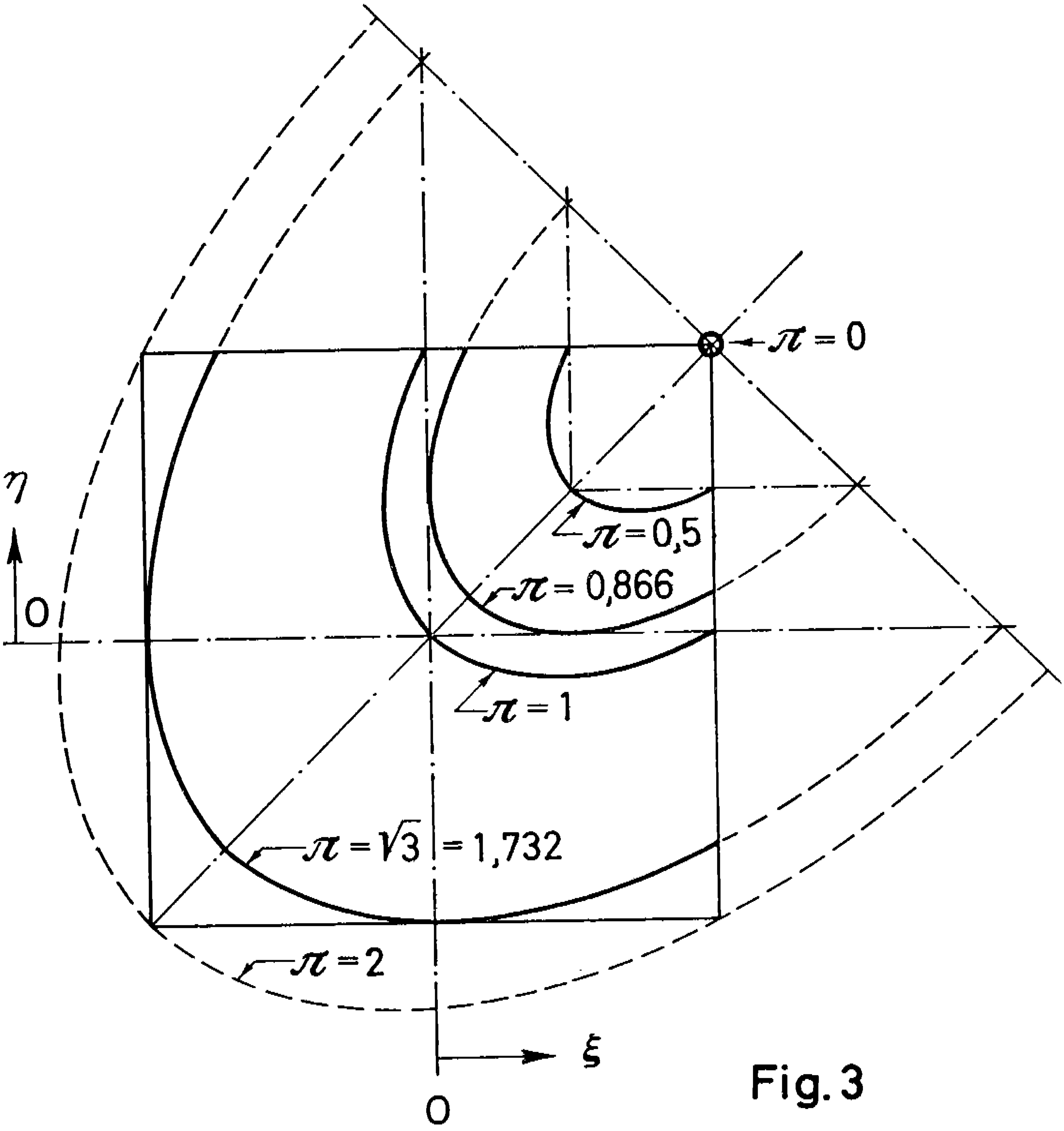

Fig. 3

Π - σ_1 diagram for constant temperature may be a suitable definition of triaxiality. $\sigma_y = \Pi . \sigma_1$ being constant at the yield point the mean Π of a notched specimen may simply be obtained by taking the ratio between σ_1 in yield of a smooth specimen and σ_1 in yield of a notched specimen in an ordinary tensional test. For the smooth specimen in one-dimensional tension $\Pi = 1$, for notched specimens it will be less. Π can vary only between 0 and 2.

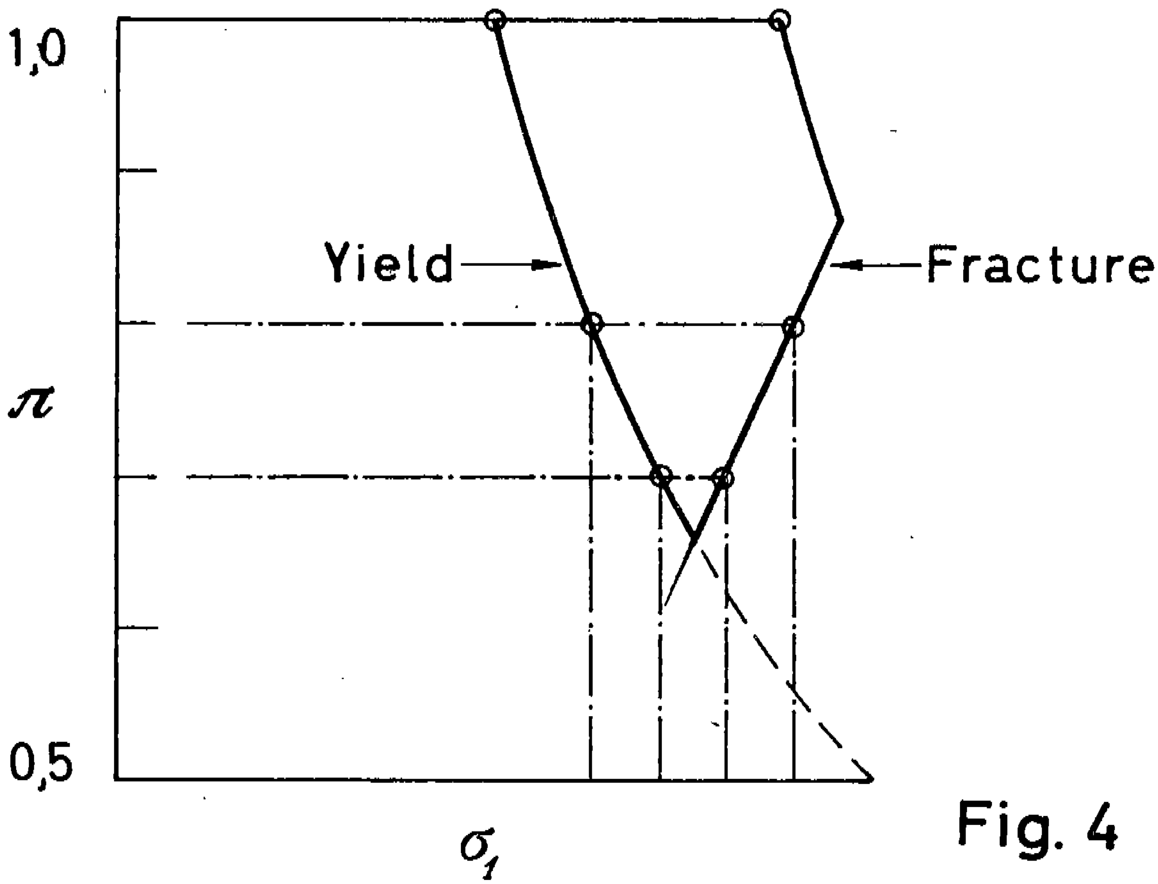

Fig. 4

The curves for Π = constant are ellipses in a diagram with ξ and η as coordinates, as shown in Fig. 3. All the ellipses have $\sqrt{3}$ as the ratio between their large and small diameters and have their axis at an angle of 45° to the coordinate axes. Only the part of the ellipses within the square $\xi = \pm 1$, $\eta = \pm 1$, has a physical meaning. It is remarkable that Π = 2 at three of the corners of the square and Π = 0 at the fourth corner. The upper right quadrant with all principal stresses in tension is the most important in connection with brittle fracture. Notched specimens are expected to have Π somewhere between 0.5 and 1.0.

To get a $\Pi - \sigma_1$ diagram which gives some information about the notch sensitivity of a steel at a certain temperature it will be necessary to measure yield and fracture stress for a smooth and at least two notched specimens with different notch radii. Isosceles hyperbolas are drawn through the two values for smooth specimens, Π = 1.0, Fig. 4. The yield points for the notched specimens will lie on the left hyperbola and thereby determine the Π's for these specimens. A line is finally drawn through the points of fracture stress for the notched specimens. The brittleness of the material is judged by the position of this line. If the area between the curves of yield and fracture is small, the material is brittle. If the area is large it is fibrous.

It would be interesting to have say Charpy V-notch specimens of ship steel broken in slow bending including measurement of the bending moment, with a standard notch as well as with a notch treated according to the Hartbower low-blow technique, so as to get some idea of the Π-values of such specimens. It would give us designers at least a feeling of some understandable stress values.

One reason for my expanding this way of presentation is that it gives me an opportunity to call attention to a thesis by the German scientist Christof Rohrbach.* By modern means he has measured strains over lengths of 1 mm on steel and has been able to map the principal stresses over the least cross section of a notched specimen. His machine for slow bending is interesting and allows the specimen to be surrounded by a liquid of constant temperature.

*"Kennzeichnung der Sprödbruchneigung der Stählen durch Messung der Fliessspannung, Reissspannung und Brucheinschnürung an dreiachsig beanspruchten Proben," Archiv für das Eisenhüttenwesen pp. 213-229, April 1955.

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