

INTERNATIONAL INSTITUTE OF WELDING

GUIDE
TO THE WELDING AND WELDABILITY OF
C-Mn STEELS
and
C-Mn MICROALLOYED STEELS



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High strength steels for welded structures
and approved by Commission IX
Behaviour of metals subjected to welding

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Commission IX

"Behaviour of metals subjected to welding"

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FOREWORD

The document which Commission IX (Behaviour of metals subjected to welding) presents under the title "Guide to the welding and weldability of C-Mn and C-Mn microalloyed steels" is the result of several years' work by Sub-Commission G (High strength steels for welded structures), under the chairmanship of Dr G Kullberg (Sweden).

After collecting a great deal of information on C-Mn and C-Mn microalloyed steels, the Sub-Commission attempted to classify them in relation to their characteristics of weldability, which led to the appraisal of the fundamental problems (transformation, cracking, evolution of the mechanical properties, etc.) relevant to this notion, of the conventional criteria (chemical composition, carbon equivalent, underbead hardness) and of the present day ideas regarding the optimum welding conditions.

In the course of this work, Sub-Commission G requested the assistance of other groups of Commission IX, in particular as regards the continuous cooling transformation diagrams and the cracking tests. The Sub-Commission noted the opinions expressed by the members of the Commission during the Annual Assemblies, so that this document can really be considered as the general expression of the views of Commission IX.

The quality of a welded construction is influenced by various characteristics depending upon the base steel, the constructional details and the welding procedure used; this quality is evaluated by means of various tests and inspection operations carried out before or after welding.

In this document, the readers interested in one or more of these four aspects of the general problem of weldability will find information which will be useful to them.

Furthermore, it must be underlined that the Sub-Commission has not wished to delay for too long a time the availability to users of the results of its work. Therefore, the Sub-Commission did not wait to prepare formal recommendations, the drafting of which is, meanwhile, under investigation; it here presents the essential ideas which will govern the content of such recommendations.

H. GRANJON
Chairman of Commission IX

CHAPTER 1

INTRODUCTION

This guide to the welding and weldability of carbon-manganese (C-Mn) and C-Mn microalloyed high strength steels is based on information gathered from steel producers, steel users and from research bodies inside and outside the IIW; it also gives a theoretical background for the standardization of such steels for use in national and international specifications.

Higher strength steels may be obtained in differing ways: by increasing the C and Mn content, by microalloying techniques, precipitation hardening, heat treatment, etc.

It is hoped that, on the basis of the understanding of different weldability criteria as given here, the complexity of, and necessity for guidance in present day welding of modern steels in modern designs will be better understood. This guide also gives the background for the classification of such steels to assist future work in Sub-Commission IX-F (Study of recommendations relating to steels for welded constructions) of the IIW.

The ISO recommendation R 185/1967 defines weldability in the following way:

"A metallic substance is considered to be weldable to a stated degree by a given process and for a given purpose, when metallic continuity can be obtained by welding using a suitable procedure, so that the joints comply with the requirements specified in regard to both their local properties and their influence on the construction of which they form part."

According to the general views expressed by steelmakers, fabricators and users, weldability is a processing property for providing welded joints with the desired qualities. This statement is considered from the points of view of the steelmakers, fabricators and users in different ways; the steelmakers and fabricators expect from the knowledge of the processing properties a guide to the classification of the steels and freedom from

any trouble in welding. All parties look for the optimum of reliability and economy. These points of view frequently conflict with each other.

Welding procedures used in the fabrication of high strength steel structures are moving towards higher heat input processes in line with an increase in welding efficiency, available equipment and the requirements of new design concepts, viz pre-fabrication.

Design, for instance of pressure vessels, is also moving toward the introduction of heavier thicknesses, prefabricated components and more elaborate assemblies where the use of higher strength materials is intended to contribute to a better use of the steel in relation to the service requirements.

Thus the welding of high strength steels involves consideration of several extrinsic factors peculiar to each combination of material, design and welding procedure, comprising the composition of the steel, its initial mechanical properties and thickness, type of joint, welding conditions (e.g. on site), availability of pre- or post-heating facilities, welding process, choice of electrodes and fluxes, etc.

On the other hand, the judgment of the weldability of a steel under given conditions of design and welding must take account of both the overall and the local aspects. The former is a combination of certain local properties intended to prevent the initiation and propagation of unstable cracks. This is normally assessed on large scale specimens, but may ultimately be determined on the construction itself. The latter is more related to the behaviour of the separate components of a welded joint: parent metal, heat-affected zone, and weld metal, including the study of the mechanical properties of the HAZ, hot crack and cold crack initiation, influence of hydrogen, inclusions, etc. together with the effect of stresses and the thermal cycle. This aspect is studied by so-called weldability tests, involving consideration of metallurgical, physio-chemical and physical factors.

Unfortunately, the tests performed and the results available do not cover all the requirements of each class of construction. To prescribe the best choice of material, design and welding conditions on an overall scale is not possible with available data. Frequently, there is a lack of scale effect between the laboratory test and the actual construction. Restraint and shrinkage

stresses are examples of factors whose influence may be underestimated in the evaluation of high strength steel for welding. It is thus unreasonable to look for a "miracle" test of weldability, but it is reasonable to define the limits of each of the existing tests and to promote tests in which the measuring of an intrinsic factor like stress could be developed.

Moreover, reverting to the user's decision for selecting a steel for a given application, this will finally be based on existing specifications or classifications of available materials. It is thus imperative to build up suitable criteria for classifying high strength steels, but it is advisable to qualify these criteria with compatibility tests to take account of the complexity involved in the term weldability.

In the present document, numerical data are expressed according to the metric system, with units such as kgf/mm^2 , kgfm , etc. The conversion of these units into the SI system would have created a risk of confusion, since the old system is used in the commercial designations and symbols of the products mentioned.

THE METALLURGICAL BACKGROUND OF MICROALLOYED STEELS

GENERAL

Microalloyed steels are defined as steels treated with small additions of elements which are carbide-, nitride- or carbonitride- forming such as Nb, Ti, V, Al¹⁾, Ta and Zr, the total content of these elements seldom exceeding 0.15 %. These microalloyed steels are usually delivered in the normalized or controlled rolled condition as fine grained steels with good notch toughness properties. This document considers only C-Mn and C-Mn microalloyed steels delivered in the normalized or controlled rolled condition.

Microalloyed steels have been developed to meet the demands for strong and tough steels with good weldability. To produce a steel with certain desired properties steelmakers can, by using metallurgical techniques, influence the microstructure, which in microalloyed steel consists generally of ferrite and pearlite. In some cases there is also bainite present, but as this bainite does not differ very much from pearlite in its mechanical properties, its presence can usually be neglected.

In general the mechanical properties of a steel depend on the following features of the microstructure:

1) From this definition it will be clear that the aluminium killed carbon-manganese steels belong to the category of micro-alloyed steels, although the Al-addition is not as effective as the strong carbide forming elements in increasing the tensile properties.

- The relative proportion of pearlite which increases with increasing amounts of carbon and manganese.
- The structure of the pearlite which improves the mechanical properties as the spacing between lamellar of ferrite and cementite decreases. A finer pearlite can be obtained by adding manganese and other elements which retard the austenite transformation during cooling.
- The nature of the ferrite which can be influenced by elements in solid solution in the ferrite (Si for instance) or by elements which precipitate within the ferrite as carbides or nitrides (Nb and V for instance).
- The grain size which for a given analysis is strongly dependent on the mechanical and thermal treatment during and after rolling.

The tensile strength and yield stress are higher when:

- a) the relative amount of pearlite is increased (this increases the tensile strength, but not so markedly the yield stress)
- b) the pearlite is refined
- c) the grain size decreases
- d) a greater quantity of elements is in solid solution
- e) precipitated particles are present in quantity.

The toughness properties are improved when:

- a) the relative amount of pearlite is decreased
- b) the pearlite is refined
- c) the grain size decreases
- d) the quantity of elements in solid solution is decreased (nickel being an exception)
- e) there are fewer precipitated particles.

The requirement of good weldability limits the quantity of pearlite (as this is principally dependent on the carbon content

of the steel). Particularly interesting is grain refining, because this gives an improvement in yield strength and in transition temperature. Raising the yield stress by other means than grain refining or refining of the pearlite is always accompanied by a rise in the transition temperature. An exception is nickel; nickel in solid solution in the ferrite raises the yield stress and lowers the transition temperature.

GRAIN SIZE

The most frequently applied methods of obtaining a fine grained steel are controlled rolling and normalizing. With controlled rolling the percentage reduction and the temperature of the final passes are prescribed. The finishing temperature is often about 800–850°C. This is appreciably lower than the finishing temperature with normal rolling. Normalizing is most effective when the normalizing temperature is as low as possible above the A_{c3} -temperature and the time at temperature is short. If the steel contains small precipitates which inhibit austenite grain growth, then the finest possible microstructures after normalizing are obtained. A well known example is aluminium killed steel in which aluminium nitrides are responsible for the production of a fine grain after normalizing.

PRECIPITATES

Precipitated particles can work in two ways:

- The direct way in which they influence the mechanical properties by local deformation of the ferrite matrix.
- The indirect way, that is, by their influence on the grain size.

Precipitates can inhibit the grain growth of the austenite, or influence the recrystallisation behaviour during rolling, or act as nuclei for ferrite crystals during the austenite to ferrite transformation.

The behaviour of precipitates during the production of steel plates may be described in the following general picture:

Heating the slab up to ca 1300°C → all elements enter into solid solution

Cooling during and after rolling → little or no grain refinement, but some slight precipitation of carbides

Heating to and holding at normalizing temperature → precipitation of particles and grain refinement. Coarsening of the precipitates if the temperature is too high. Resolution of precipitates if the temperature is excessively high

Cooling from normalizing temperature → some re-precipitation of particles which were taken into solution if normalizing temperature has been excessively high.

As examples of the above procedure steels containing aluminium, niobium and vanadium are discussed below.

a/ Steels containing aluminium

Steels with normal Al-contents (0.02 – 0.08%) and normal nitrogen contents (< ca 0.013%), which have been heated up to 1300°C before rolling, show in the as-rolled condition only a small proportion of Al and N in the form of nitrides. The direct effect of the particles on the mechanical properties is negligible. During the heating part of the normalizing cycle the maximum possible quantity of aluminium nitride particles precipitate. After cooling, relatively coarse AlN particles are present which cause grain refinement but produce no appreciable increase in yield stress.

b/ Steels containing niobium

Niobium is added to structural steels in quantities of 0.005 – 0.08% and gives carbonitride precipitates in the steel. In the as-rolled condition there are a great number of very small particles which have a very strong direct effect. The yield stress is raised appreciably, but the notch-toughness is lowered appreciably except in plates thinner than about 10 mm. Refining

of the grain by controlled rolling counteracts the unfavourable influence of the precipitates on notch-toughness, so that in this condition acceptable steels can be produced in thicknesses up to about 25 mm. During normalizing, coarsening of the precipitates takes place, but there is no significant resolution, so that in the normalized condition there are only coarse precipitates and the strengthening effect of the precipitates has disappeared almost completely. Grain refining is however very great and in the normalized condition niobium steel is a steel with a very fine grain.

c/ Steels containing vanadium

Vanadium, added in quantities of 0.05 - 0.20%, gives in the as-rolled condition the same effects as niobium, by forming carbo-nitrides. In the normalized condition there is a slight difference from niobium bearing steel as, during heating, some of the VCN-particles go into solution and are reprecipitated as fine particles during cooling. So there is in vanadium steels an increase in strength, combined with a pronounced grain refining effect.

Niobium and vanadium can be added to semi-killed steels as their relatively small affinity for oxygen ensures that their function as carbide or nitride formers is fully utilized. For example, semi-killed and silicon-killed niobium-steels are produced as well as aluminium-killed niobium-steels, and in the latter case the effects of aluminium nitrides and niobium carbo-nitrides are combined to give a steel with very good notch toughness. The combination of V and Nb is also used with similar effect.

There are a few steels delivered in the normalized and tempered condition; tempering after normalizing in the case of microalloyed steels is sometimes used because during cooling after normalizing only a part of the precipitates have been formed. After a tempering treatment at 500°C - 600°C, the precipitation is completed and optimum conditions of particle size and number of particles are reached.

DATA FOR CARBON MANGANESE AND CARBON MANGANESE MICROALLOYED STEELS

GENERAL OUTLINE OF THE DATA ON STEELS COLLECTED BY SUB-COMMISSION IX G

In 1962 Sub-Commission IX-G of the IIW started to collect data on high strength steels for welded structures. Steels of the St 52 type were the basis of the steel data collection and no upper strength limit was fixed. The collection included C-Mn and C-Mn microalloyed and low-alloyed steels with a C-content less than 0.30 %.

No limit was set for the total content of alloying elements, but it was understood that maximum 5% would be appropriate. The following data were collected on a special form:

- Trade mark
- Specification No. (ASTM, BS etc.)
- Manufacturer's name
- Product: plate, tube and/or section
- Chemical composition, including fine grain additions
- Heat treatment and mechanical properties, including transition temperature
- Transition curve Charpy V
- Recommendations for welding.

Each member of the Sub-Commission who represented a steel producing country contributed and up to July 1970 the collec-

tion included about 350 steels. The following countries have participated:

Austria	Japan
Belgium	Netherlands
Czechoslovakia	Sweden
France	United Kingdom
Germany	USA
Italy	

For many steels a very thorough documentation is given, for others only a few data. For the steels manufactured in the USA the information available is collected in tables published in ref. (5). For most steels maximum and minimum values for the chemical composition are found, but definite values for some residuals such as Cu, Cr and Ni are often missing. In the case of a few steels a maximum carbon equivalent is noted. Information about steel making processes is very sparse. Tensile properties are fairly well documented for different thickness ranges. Typical transition curves are sometimes drawn, but for several steels information on the brittle fracture properties is missing. Recommendations for welding are recorded for most steels, but they are often rather brief and not specially informative.

The Sub-Commission decided for its own purpose to classify the collection using an arbitrary system based on the minimum upper yield point followed by a current steel number in each yield strength class. In steel 3613, for instance, the figure 36 refers to an upper yield strength of 36 kgf/mm² minimum and the current number in the 36-class is 13. The yield point figure used for the classification refers to a thickness of 12.5 to 15 mm (about 1/2").

The yield point classes used are found in Table 3:1.

All steels with an upper yield point less than 36 kgf/mm² or 50,000 psi are collected in class 00.

The delivery treatment types for specified mechanical properties are marked as shown in Table 3:2.

In the classified collection there is information on guaranteed minimum temperatures by Charpy V-testing

- at the level of 2.8 kgfm or 20 ft. lbs (t_{KV} 2.8/20°C)
- at an other level, e.g. 4.8 kgfm or 35 ft. lbs.

Table 3:1

Class	UYP kgf/mm ²	UYP psi
36	min. (between 36 and 40)	min. 50,000
40	" (" 40 " 45)	" 55,000
45	" (" 45 " 50)	" 65,000
50	" (" 50 " 55)	" 70,000
55	" (" 55 " 60)	" 80,000
60	" (" 60 " 65)	" 85,000
65	" (" 65 " 70)	" 90,000
75	" (" 75 " 90)	" 105,000
90	" (90 and above)	" 130,000

Table 3:2

A	No heat treatment: (hot rolled)
B	1. Controlled rolled (hot rolled under controlled conditions to produce a fine grain) or 2. Normalized with normal or forced air cooling
C	Normalized and tempered
D	Quenched and tempered

Most of the steel data were collected and classified during the years 1962 to 1966, but new steels have continuously been added and old data sheets have been revised. In this way the Sub-Commission has a classified file with uniform data sheets in which each sheet gives reasonably good information about a particular steel. Data for steels mentioned in this document do not take into account changes made after July 1970.

It should be mentioned that some good surveys of high strength steels have been published during recent years and these have also been included in the files of Sub-Commission IX-G and used in its work (1) to (5). Standards for high strength steels have been published and references (6) to (13) are some of the more important. Only a few standardized steels have been included in

the collection, since the specifications have only been issued recently.

When the steels were grouped in the different yield strength classes, it was found that about 40 % were in the 36-class. About 75 % of all steels are found in the three lower classes: 36, 40 and 45.

In the lower YP classes, the 36, 40 and 45 classes, hot rolled or normalized conditions are used to about the same extent. The normalized and tempered condition is used in some cases. It can also be noted that quenched and tempered steels are found even in the lowest class, YP class 36. All steels with a minimum YP 55 kgf/mm² and higher (with a few exceptions) are quenched and tempered.

As this report aims at giving recommendations for C-Mn and C-Mn microalloyed high strength steels, only the 36, 40 and 45 YP classes are of interest.

The steels in these three classes can be divided in several groups:

1. C-Mn steels. Treatment: A, B or C.
2. C-Mn microalloyed steels. Treatment: A, B or C.
3. Low-alloyed steels, mainly alloyed with small amounts of Cr, Ni, Mo and Cu. Many of these steels are also fine grained. Treatment: A, B or C.
4. Quenched and tempered steels. Treatment: D.

As recommendations for welding low-alloyed and quenched and tempered steels will be published later; only the first two groups are considered here.

DATA FOR THE STEELS

Data for C-Mn and C-Mn microalloyed steels (with the exception of the quenched and tempered variants) will be found in Appendices 1 to 3. All data refer to plate material.

Appendix 1: 36 kgf class

Appendix 2: 40 kgf class

Appendix 3: 45 kgf class

These tables provide the following data:

- Identification number in the IX-G collection
- Manufacturing country
- Trade mark (or in a few cases specification-designations)
- Chemical composition: C, Si, Mn, P, S and additional elements
- C-equivalent according to the IIW formula or other formulas (if given by the manufacturer)
- Treatment type:
 - A = no treatment
 - B = controlled rolled or normalized
 - C = normalized and tempered
- Mechanical properties (1/2" thickness):
 - Yield point kgf/mm²
 - Tensile strength kgf/mm²
 - Elongation 5xd %
- Maximum thickness manufactured
- Yield point guaranteed at this thickness
- Guaranteed minimum temperature for different energy levels by Charpy V-testing (longitudinal test pieces).

The 36-class includes data for 66 different steels, the 40-class 25 and the 45-class 12.

In the 36-class the most common microalloying element is Al. Several of the steels are Nb-treated and a few have V-, Zr- or Ti-additions.

In the 40- and 45-classes nearly all steels are fine grained by treating with Al, Nb, V and Ti or combinations of two or three of these elements.

Consideration should be given to the influence of microalloying elements on the properties of the weld metal. Generally the figures given in the tables should be satisfactory but in certain cases lower limits may be required.

Maximum carbon contents vary between 0.16% and 0.28%. Most steels have a maximum carbon content of about 0.22%. The silicon content varies between 0.10% and 0.60% and common maximum limits are either 0.30/35% or 0.55%. The manganese contents are fairly uniform: between 0.90 and 1.6% for all steels. The analyses can vary towards the top of the range for thicker materials.

A very few C-equivalent limits are given by the manufacturers; no definite conclusions can be drawn from the tables.

Maximum P- and S-contents are in most cases 0.04% but in a few instances lower limits, e.g. 0.025%, are specified.

The mechanical properties in the tables refer to the thickness 12.5 to 15 mm which was used for classification purposes. The temperature t_{KV} 2.8 varies more widely with guaranteed values of between 0 and -60°C .

As a conclusion it can be said that the majority of the steels in these three groups are microalloyed and normalized steels with the following base analysis:

C max. 0.22% - Si about 0.30% - Mn 1.0 to 1.5% -
P max. 0.04% - S max. 0.04%

Several microalloying elements such as Al, Nb, Ti, V and Zr alone or in combination with each other and with N are used; Al and Nb are the most common.

Specified mechanical properties at 12.5 to 15 mm (1/2") plate thickness seem to be those collected in Table 3:3.

The guaranteed temperature t_{KV} 2.8 lies between -20 and -40°C but there are wide variations.

Table 3:3

Min. yield point (YP)	kgf/mm ²	36	40	45
Min. tensile strength (TS)	kgf/mm ²	55-65	55-70	58-75
Min. elongation	5 x d %	22	20	20

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- (12) L'Association Française de Normalisation (AFNOR): Norme Française A 35-501: Aciers de construction d'usage général - Nuances et qualités. (French Standard A 35-501: Structural steels for common use - types and qualities).
- (13) Rules for the construction and classification of steel ships issued by the Ships Classification Societies.

THE USE OF CONTINUOUS COOLING TRANSFORMATION DIAGRAMS

INTRODUCTION

From the metallurgical point of view, welding is a complex operation, as regards the transformations which occur in solid steel during heating and cooling. In fact, the welding thermal cycle is very unusual when compared with that of ordinary heat treatments, for heating is quick and brief, it brings the metal to a high temperature and the welded zone is the site of a pronounced thermal gradient.

In order to study the influence of these special features, the usual methods for plotting continuous cooling transformation diagrams (CCT diagrams) have had to be revised, since the curves obtained did not fully reflect modifications observed in the vicinity of actual welds (1). Methods specially adapted for welding were developed in various laboratories, using either treated specimens, or "implants" subjected to an actual welding operation. By using such methods it is now possible to plot CCT diagrams for steels subjected to the thermal conditions of welding. For several years, Sub-Commission IX-B (Weldability of steels in relation to their transformation characteristics) of the IIW, has been working on this subject in connection with problems of cold cracking (2) and has organized an international enquiry in which several laboratories from a certain number of countries have participated. At present, it is therefore possible to use the information given by such diagrams to assess weldability and to determine the optimum welding conditions.

METHODS FOR THE PLOTTING OF THE DIAGRAMS

The plotting of the CCT diagrams during welding requires, on the one hand, suitable thermal cycles and, on the other, the

assessment, through their temperature and time co-ordinates, of the transformations undergone by the steel during these cycles.

To obtain the thermal cycles, one operates either on "weld-simulated" specimens, or "in situ" in the vicinity of actual weld beads.

In the "simulated-weld" method, a specimen of the steel under investigation is heated and cooled (in a furnace, by radiation, induction, Joule effect etc.) under conditions simulating as closely as possible those of the welding thermal cycles; certain apparatuses are provided with a programmed heating device, in order to ensure reproduction of the pre-fixed cycles (3).

In the "in situ" method, the work is carried out on actual weld beads. The use of implants makes this method much easier to employ and enables a correlation with cold cracking to be made (4).

The assessment of the transformations in terms of time and temperature is based on changes in a physical property (expansion, magnetism, conductivity) or on the heat of transformation (thermal analysis).

The comparison of results obtained by the various methods proposed was undertaken by Sub-Commission IX-B. Within the framework of a co-operative study, the report of which will be the subject of another publication, the CCT diagram for a C-Mn steel ($C = 0.16\%$, $Mn = 1.08\%$) was determined by seven laboratories in six countries (Belgium, France, Germany, Netherlands, Sweden and the United Kingdom), which compared their results for the co-ordinates of points I to VI of the schematic diagram given in Fig. 4:1. Six other steels were studied (C eq. from 0.39 to 0.52), some of them by two of the laboratories taking part. The conclusion of this comparison was that, subject to certain recommendations for procedure being prepared by Sub-Commission IX-B, it is possible to characterize steels intended for welding by a CCT diagram; in particular, it is possible to determine critical cooling curves and compare them with the conventional criteria based upon maximum underbead hardness or carbon equivalent.

PRESENTATION OF THE DIAGRAMS

For each steel, the transformation diagram can be represented in a conventional manner, as shown in Fig. 4:2, in temperature/

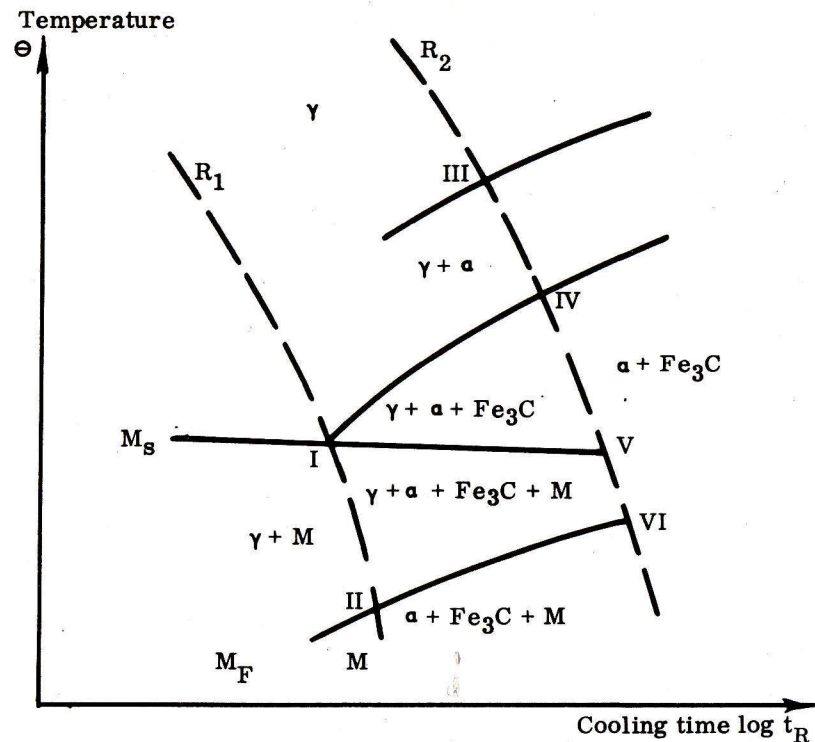


Fig. 4:1. Schematic CCT-diagram showing points I - VI. Controlled in a cooperative study.

logarithm of time co-ordinates, the time being calculated from an arbitrarily fixed point, for instance 1 000°C. Each cooling curve plotted on the diagram corresponds to a welding condition. Thus, in Fig. 4:2 (C-Mn steel, C = 0.19 %, Mn = 1.52 %, Si = 0.50 %), the critical cooling curve from which the martensitic transformation begins corresponds to the depositing on a 20 mm thick plate of a weld bead made with an energy ranging from 30 to 37.5 kJ/cm.

In order to have a better correlation between the indications given in the transformation diagram and the welding conditions, several laboratories present the result (Fig. 4:3) in a diagram giving the transformation temperatures as a function of the

Marking	1	2	3	4	5	6	7	8
Heat input J/cm	6 500	10 000	20 000	30 000	37 500	50 000	40 000	45 000
t_R (800–500°C) s	2.3	5	10.2	16.5	23.5	32	44.5	57
Working temperature °C	Amb	Amb	Amb	Amb	Amb	Amb	200	200

Temperature
°C

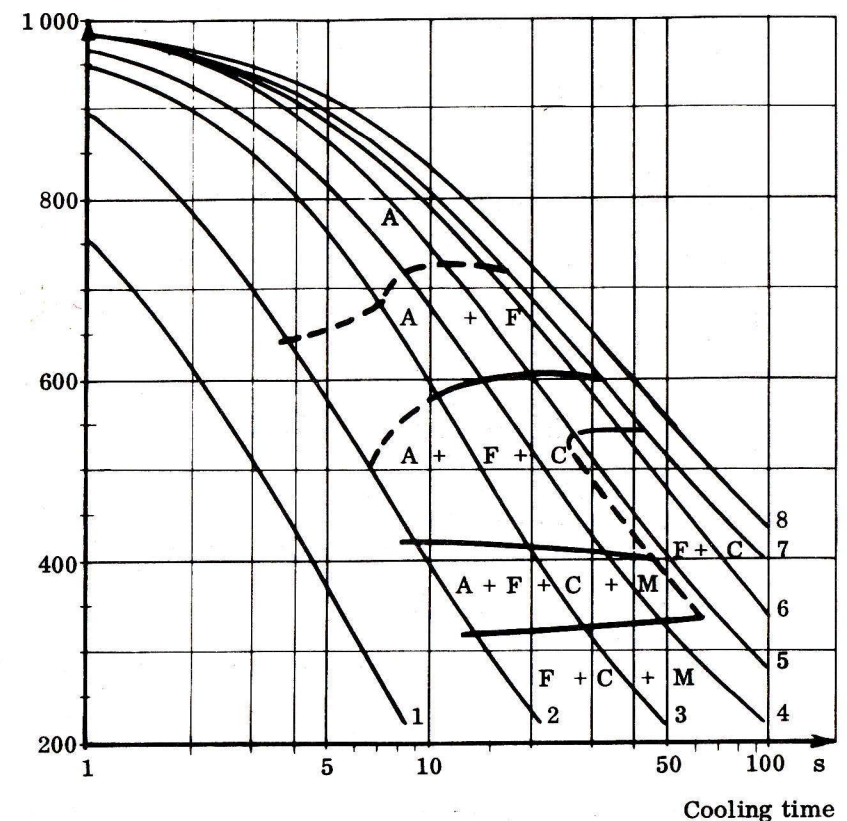


Fig. 4:2. Conventional CCT-diagram for a C-Mn steel (C=0.19 %, Mn=1.52 %, Si=0.50 %).

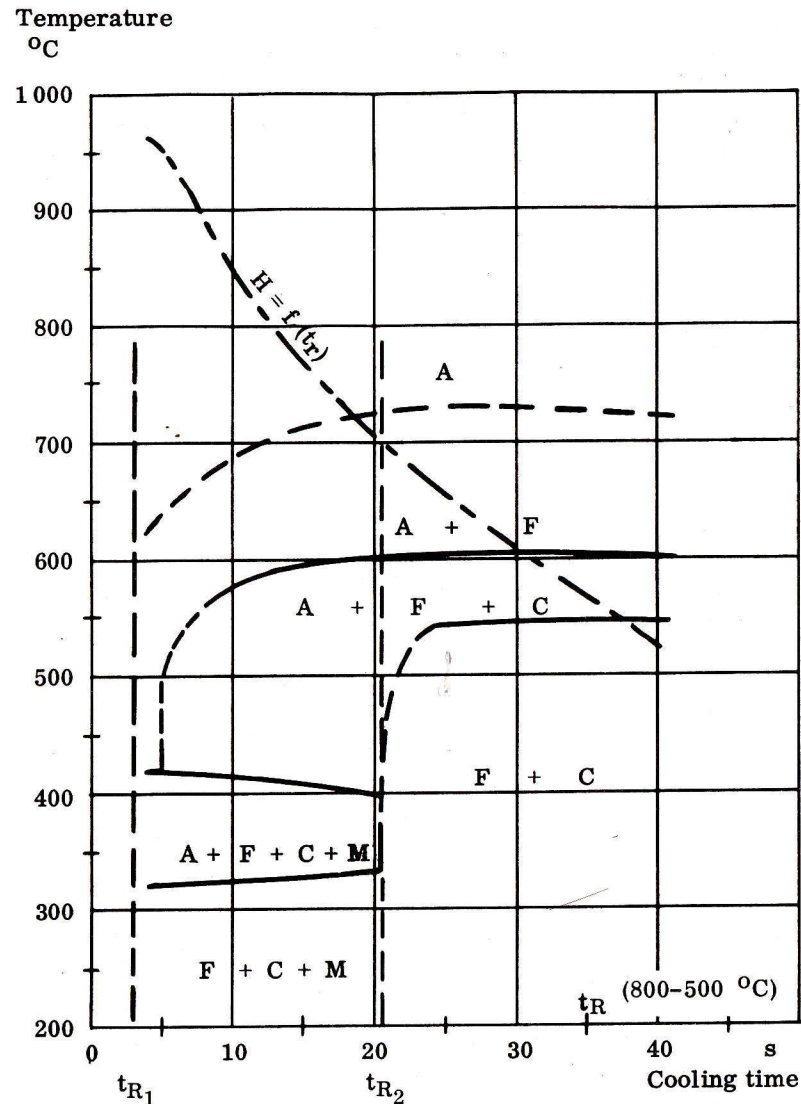


Fig. 4:3. Transformation temperatures as a function of the cooling time between 800°C and 500°C for a C-Mn steel (C=0.19 %, Mn=1.52 %, Si=0.50 %).

cooling time, for example, between 800°C and 500°C. This range is the one which is most likely to offer a constant cooling time over the whole of the heat-affected zone, whatever the temperature attained may be. Furthermore, as the cooling time is a direct function of the energy input (with thick plates, the time is proportional to the energy input), the diagram can be interpreted in terms of the welding energy. Lastly, this presentation allows easy interpretation of the hardness/cooling time curve, which can be superimposed on the diagram. This representation clearly shows the critical cooling time t_{R2} below which the martensitic transformation appears and from which precautions will have to be taken against cold cracking. The critical time t_{R1} corresponds to the completely martensitic transformation.

TRANSFORMATION AND COLD CRACKING

As the presence of martensite is a necessary condition for the occurrence of cold cracking, knowledge of the transformation diagram and, in particular, of the critical cooling time t_{R2} is of great importance: in cases where, with a C-Mn microalloy steel, the welding conditions adopted correspond to a cooling time greater than t_{R2} there is no risk of cracking, whatever the hydrogen content may be. When the cooling time is lower than t_{R2} , it is necessary, for the chosen welding process, to test for the absence of cold cracking. If the test is quantitative, as in the case of the constant-load cracking test on implants (5), a curve may be plotted (Fig. 4:4) giving the cracking stress as a function of the cooling time; by fixing the cracking stress, e.g. at the level of the yield strength of the steel considered, it is possible to determine the corresponding minimum cooling time (i.e. the admissible martensite ratio), as well as the permissible maximum underbead hardness. For each welding thickness it is thus possible to ascertain the welding energy required and, if necessary, the preheat temperature, account possibly being taken of a coefficient representing the effect of the geometrical form of the joint.

It should be emphasized that the cooling time associated with a certain value of the cracking stress and the corresponding underbead hardness depends on the hydrogen content of the weld metal. Thus, as regards the example given in Fig. 4:4, the

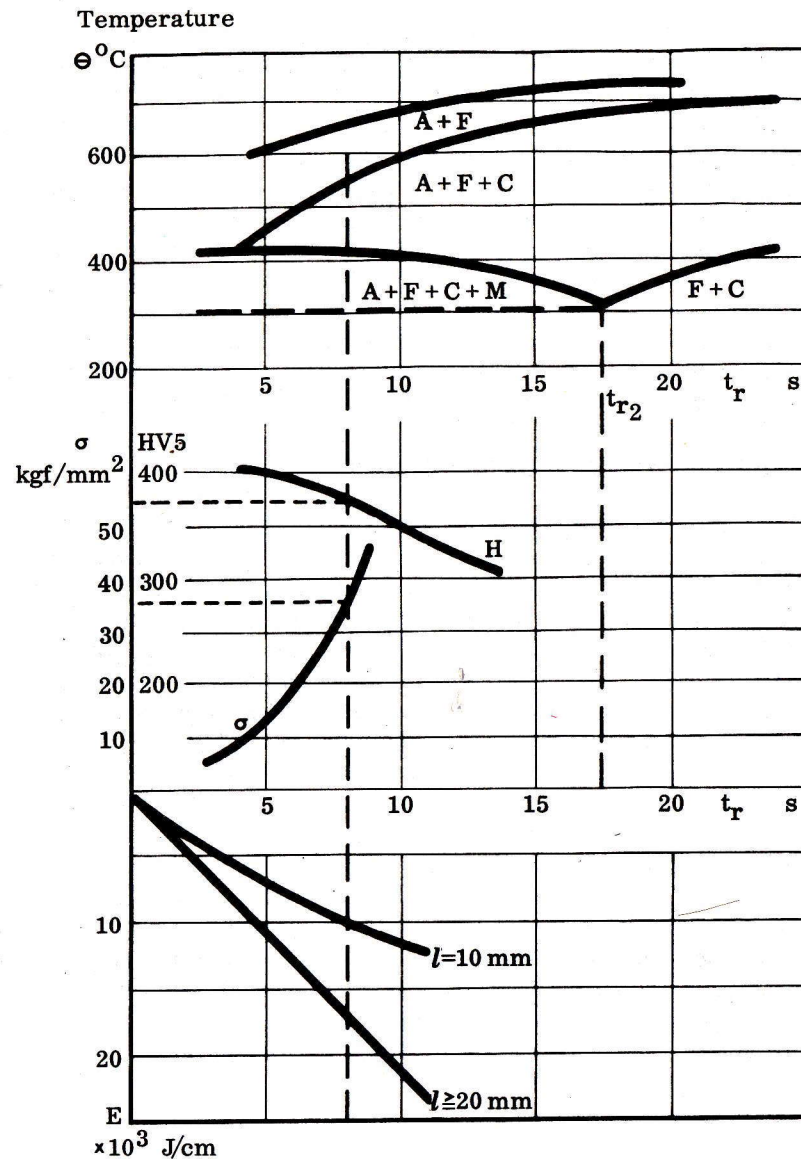


Fig. 4.4. Relationship between transformation characteristics, cracking stress at constant load cracking tests on implants and cooling time.

HV 375 hardness level associated with a cracking stress equal to the yield strength is valid when arc welding with dry basic electrodes; it would be lower if welding with rutile electrodes, and higher with MIG welding. A recommendation to this effect has already been made by Commission IX (6).

TRANSFORMATION DIAGRAMS AND RECOMMENDATIONS REGARDING WELDABILITY

Thus, knowledge of continuous cooling transformation diagrams plotted for welding thermal cycles, and particularly, of the critical martensitic time, makes it possible to lay down recommendations for the welding of steels, including C-Mn or C-Mn microalloy steels. The previous criteria are not superseded, but clarified by these diagrams. This is true also for the underbead hardness criteria, which is dealt with in the hardness/cooling time curve, for the carbon equivalent criteria, already considered by Commission IX which fixed limits for its use (7), or even for more sophisticated criteria showing the influence of the chemical composition on hardenability (8) or on the cold cracking susceptibility which tends more and more to be expressed in a quantitative manner (9).

It is highly desirable that the various grades of C-Mn or C-Mn microalloy steels for welded construction should be defined according to their CCT diagram established during welding cycles. From these diagrams it is then possible to determine and interpret the optimum welding conditions.

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COLD CRACKING IN THE HEAT AFFECTED ZONE /HAZ/ OF WELDS

A. GENERAL CONSIDERATION OF THE PROBLEM OF COLD CRACKING

Definition

Heat-affected zone (HAZ) cold cracks are defined as cracks which occur in the hard HAZ of welds close to the weld metal-steel interface where the microstructure can be coarse.

Survey of the problem

Weld cracking in the HAZ, referred to in the literature as cold cracking or hydrogen induced cracking, is most likely to be present when hard martensitic structures are produced in the HAZ and where the welding process results in high hydrogen enrichment. Restraint of the weldment also has a very marked effect on the formation of HAZ cold cracking; root and toe cracks at marked changes in section are also accentuated. As the carbon equivalent, hardenability and thickness of the weldments increase, cold HAZ cracking is more likely to occur unless suitable precautions are taken.

Complexity of the factors involved

The number of variables involved makes it difficult to quantify all the factors. A survey dividing the main factors as extrinsic and intrinsic is given by Herbiet and Salkin (1). The inter-relationship of each factor is set out in diagrammatic form (Fig. 5:1).

The most important factors involved are the:

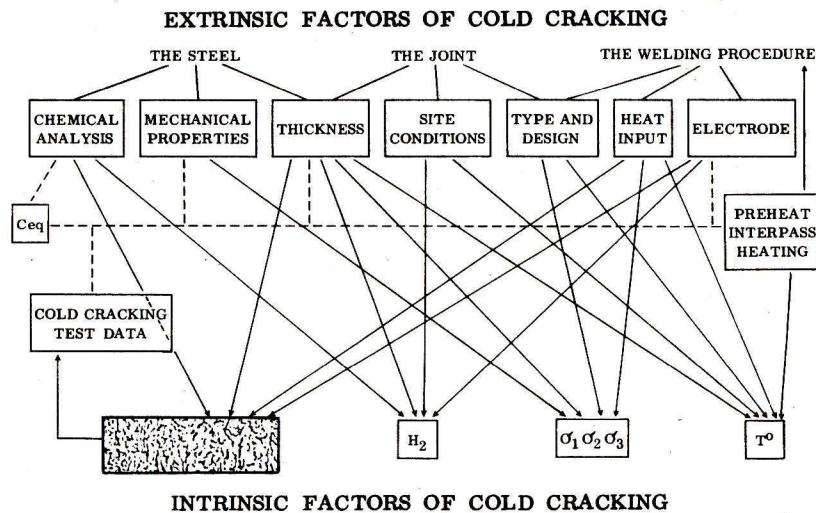


Fig. 5:1. The inter-relationship of factors influencing cold cracking.

- Carbon equivalent
- Rate of cooling after welding, which when too high can give martensitic microstructures susceptible to hydrogen embrittlement
- Hardness of the HAZ
- Hydrogen content of the weld and HAZ
- Severity of restraint of the weldment
- Notch effects

Assessment of cold HAZ cracking

a/ Carbon equivalent

Carbon equivalent is an intrinsic factor; the mechanical properties of the steel required for a construction will in some measure set the level of carbon equivalent required.

The carbon equivalent formula adopted by Sub-Commission IX-G of the IIW:

$$C_{eq} = C + \frac{Mn}{6} + \frac{Cr + Mo + V}{5} + \frac{Cu + Ni}{15}$$

possibly requires an extension as suggested by Ito and Bessyo (2).

b/ Cooling rate

The cooling rate is a controllable factor depending on the section of the component, the size of the weld, the energy input and the initial temperature (ambient or preheat temperature).

The control of cooling rate can be used in many cases to prevent the formation of hard martensitic structures in the HAZ.

The rate of cooling for a given carbon equivalent will control the hardness and amount of martensite in the HAZ, but variations in hardenability will be induced by the deoxidation of the steel.

Continuous cooling transformation (CCT) tests on small specimens heated and cooled at rates typical of welding conditions are helpful in the assessment of the hardenability parameter (3). CCT hardenability data can be used to relate composition, expressed as a carbon equivalent, to critical cooling rates appropriate to different HAZ hardness levels. These in turn can be related to critical welding procedures (4). Bead-on-plate tests are also useful in studying HAZ hardening (5).

It is suggested that the maximum HAZ hardness should not exceed 350 HV5 for C-Mn steels when no special precautions are taken with the hydrogen content in the weld metal. A higher hardness may be acceptable for microalloyed steels with the same amount of martensite, the higher hardness figure being due to precipitation hardening.

c/ Hydrogen content

In the presence of hard martensitic areas in the HAZ, the hydrogen content in the weld zone has a very marked effect in promoting HAZ cold cracking.

According to findings of Cabelka and Million (6), and as quoted by Baker (7):

- a) The fundamental cause of delayed cracking is the mechanism of the martensite transformation itself, and the effect of hydrogen is in lowering the critical stress for delayed fracture.
- b) The most critical zone for the formation of delayed cracks is the region of the HAZ where the prior austenite grain size is largest.

It has been shown that hard martensitic microstructures exhibit severe hydrogen embrittlement at normal ambient temperatures and that the embrittlement is progressively decreased at higher temperatures (8). When the HAZ cools, therefore, a progressive increase in the effect of hydrogen will occur. If the HAZ is made to cool more slowly, then it is possible to diffuse sufficient hydrogen from the HAZ to prevent cracking when the weld cools to ambient temperatures (9, 10, 11).

Welding conditions

The electrodes used and the welding process employed are controllable extrinsic factors. In a difficult or restrained situation where low hydrogen levels are required, controlled or low hydrogen welding rods or inert gas welding can be used.

Welding tests

There are several welding tests which are employed to help in assessing the suitability of steels and welding conditions to prevent HAZ cracking. Their number and their features illustrate the complexity of the problem. In addition to other welding conditions the restraint of the welded joint has a very marked effect on the formation of the HAZ cracking. In their contribution (1) the authors state that "the best weldability tests would be the final weldment itself, with all the effects of welding parameters involved".

Difficulty in correlating cold cracking tests

It is clear from recent studies and from (2) that some of the welding tests at present in use are very severe, particularly in

the restraint factor. A selection of welding parameters according to the data of a specific cracking test may give rise to unduly drastic precautions for some welded assemblies or, on the contrary, to insufficient provision against cold cracking for others. The main reasons of this discrepancy are found in the scale effect between the cold cracking tests and the real joint, and in the influence of restraint.

Hence, if one intends to cover by the appropriate cracking test the overall range of severities of welded joints, it is advisable to bear in mind some flexibility of the test. Sub-Commission IX-G have agreed that such flexibility can be derived from a classification of the welding test in order of severity according to their restraining factors. In this way, the following groups may be suggested:

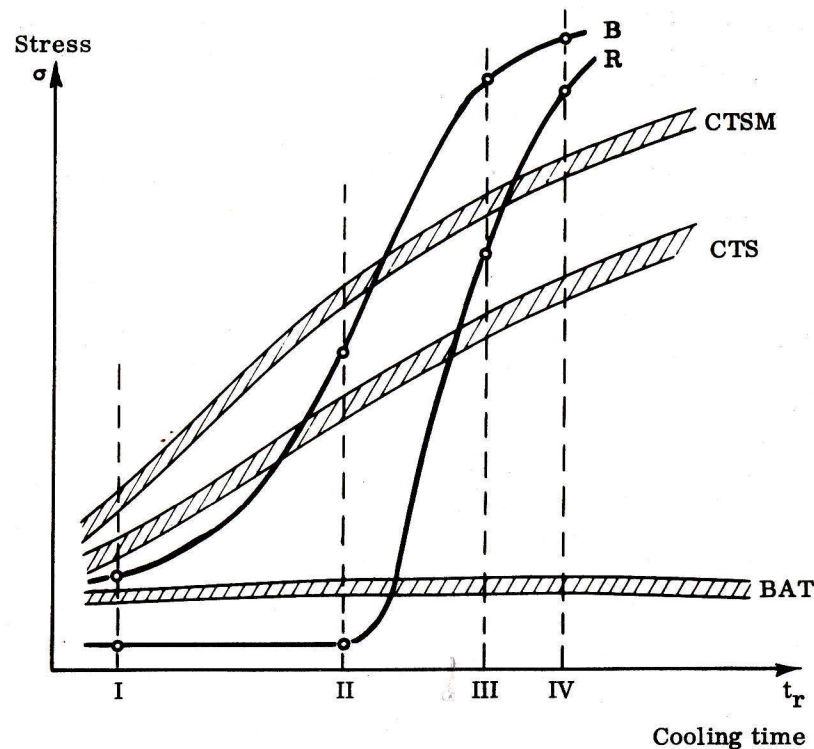
Tekken test: very high restraining factor
 Modified CTS test: high restraining factor
 CTS test: medium restraining factor
 Battelle test: low restraining factor

The selectivity of the cracking test can, in fact, vary quite considerably depending on the level of stresses coming into play.

In order to try and allow for this situation, the stress cracking-temperature curves plotted by Granjon et al (15) using the "Implant Method" can be consulted. For a C-Mn steel these curves assume the shape reproduced in Fig. 5:2 (curve B for a basic electrode and curve R for a rutile electrode). The stress level attained in a less severe test such as the Battelle test (BAT), in a moderately severe test such as the CTS test, or in a severe test such as the modified CTS test (CTSM) can be compared on the same graph.

It strikes one immediately that these three tests give different results depending on the cooling times under the relevant welding conditions. These results have been listed in the table for four welding conditions, I, II, III and IV according to the type of covering (R or B). In this table, F denotes "cracked" and NF "not cracked". In condition II the modified CTS test can be seen to cause cracking for both a basic and a rutile electrode.

It should, however, be understood that this reasoning is, at least for the time being, purely of a theoretical nature. Nevertheless, attention is drawn to the need for a better quantitative



Test	I		II		III		IV	
	R	B	R	B	R	B	R	B
CTSM	F	F	F	F	F	NF	NF	NF
CTS	F	F	F	NF	NF	NF	NF	NF
BAT	F	NF	NF	NF	NF	NF	NF	NF

Fig. 5:2. Relationship between stress and cooling time for implant tests (B=Basic covered electrode, R=Rutile covered electrode), Battelle test (BAT), CTS-test (CTS) and modified CTS-test (CTSM). F=Crack, NF=No crack.

assessment of cracking tests in terms of the restraining factor or deformation.

Another possibility would include the use of variable stress tests, such as suggested by Granjon, Prokhorov (16) and others. By simulating quantitatively the restraining factor, these tests can provide another approach to selecting the most appropriate welding parameters according to the severity of the construction.

Applications of cold cracking tests

Using appropriate cold cracking tests, recommendations on the selection of welding parameters have been agreed in various countries. A typical example can be found in Fig. 5:3 (8), which refers to root runs in butt welds. All these recommendations generally suggest minimum preheat temperatures according to the geometry of the weld, the heat input, the type of electrodes or wires and the composition of the base material, i. e. according to most of the extrinsic factors illustrated in Fig. 5:1.

B. PREHEATING, POSTHEATING AND INTERPASS HEATING

Definition

The preheating and post-heating treatments are aimed at reducing the rate of cooling after welding, in order to modify the microstructure, to promote the diffusion of hydrogen, and, at the same time change the level and distribution of residual shrinkage stresses.

Within this general scope, pre- and post-heating includes interpass heating in multipass welding, which is applied when the heat due to welding is insufficient to maintain the initial temperature of preheating between two consecutive passes.

The preheating temperature is the minimum temperature to be reached through the thickness of the plate just before welding, and which extends over a sufficiently wide zone on each side of the welding line.

The preheating temperature may extend from room temperature when high enough, up to as high as 400°C to 450°C. The temperature may need to be balanced with the welding heat input,

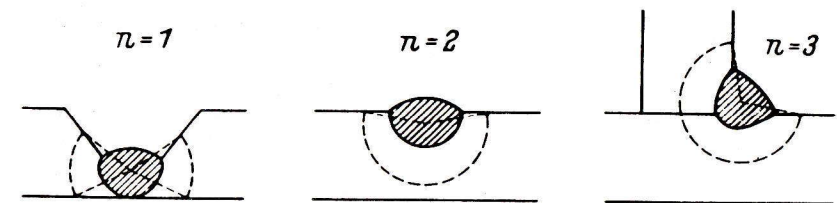
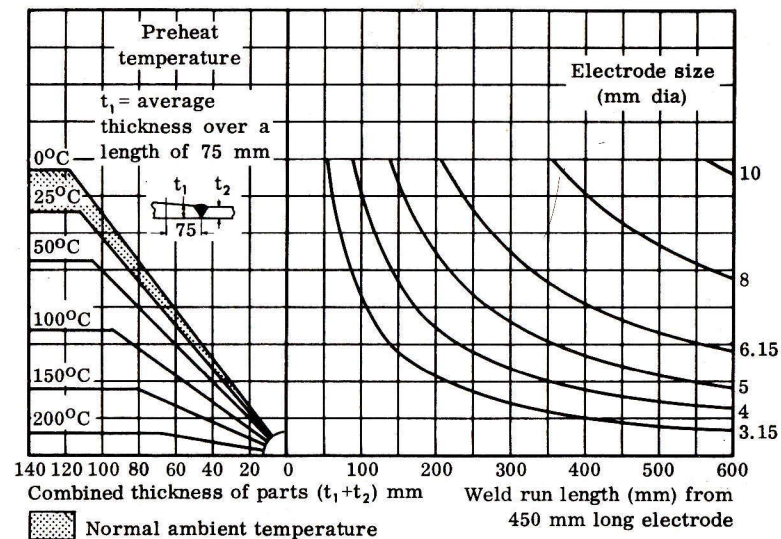
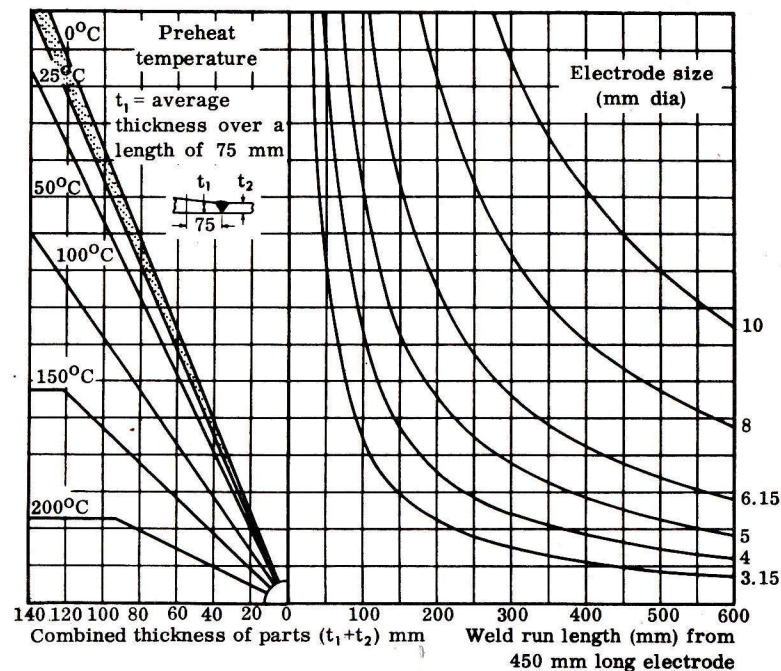


Fig. 5:4. Factor "n" defining the geometry of heat transfer.

according to the steel concerned and to the service requirements if, for example, specific mechanical properties need to be maintained.

Modification of microstructure

If one considers a variable $L = \frac{60 EI}{v/nd}$, where $\frac{60 EI}{v}$ is the input energy in joules per cm, d the plate thickness and n a geometry factor (see Fig. 5:4), it is possible to assess the maximum cooling rate as a function of welding conditions for various thicknesses (12, 13). A large amount of experimental points arising from German, British, Japanese, Russian and American investigations systematically fall within a scatter band, whose left hand limit represents a reliable relationship between the welding conditions and the cooling time between 800 and 500°C in the HAZ (Fig. 5:5).

With preheating to T_0 for the constructional part to be welded, the temperature drop becomes smaller and, thereby, the cooling rate is decreased at the same values of L . This may be expressed in terms of analytical developments of T (800 - 500) as a function of L and T_0 , and approaches the curves as shown in Fig. 5:6, each of which corresponds to the left hand borderline of the scatter band. With the same input energy conditions, the cooling time between 800°C and 500°C may be increased more than three times when preheating from room temperature to 250°C.

Fig. 5:3. Relationship between thickness of parts, preheat and interpass temperature and electrode size for root runs in butt welds welded with rutile electrodes (above) and hydrogen controlled electrodes (below).
Transcript from British Standards Institution B. S. 2642:1965.

It should be pointed out, however, that wherever possible it may be easier to employ a higher heat input rather than a preheating temperature of 200 to 250°C for extending the cooling time. For thicknesses less than 20 mm, a high heat input promotes delayed cooling between 800°C and 500°C (Fig. 5:7) which in turn prevents the formation of microstructures sensitive to cold cracking and embrittlement. This is more applicable to higher carbon steels, but it can also apply to plain carbon-manganese steels, where the requirements of underbead hardness

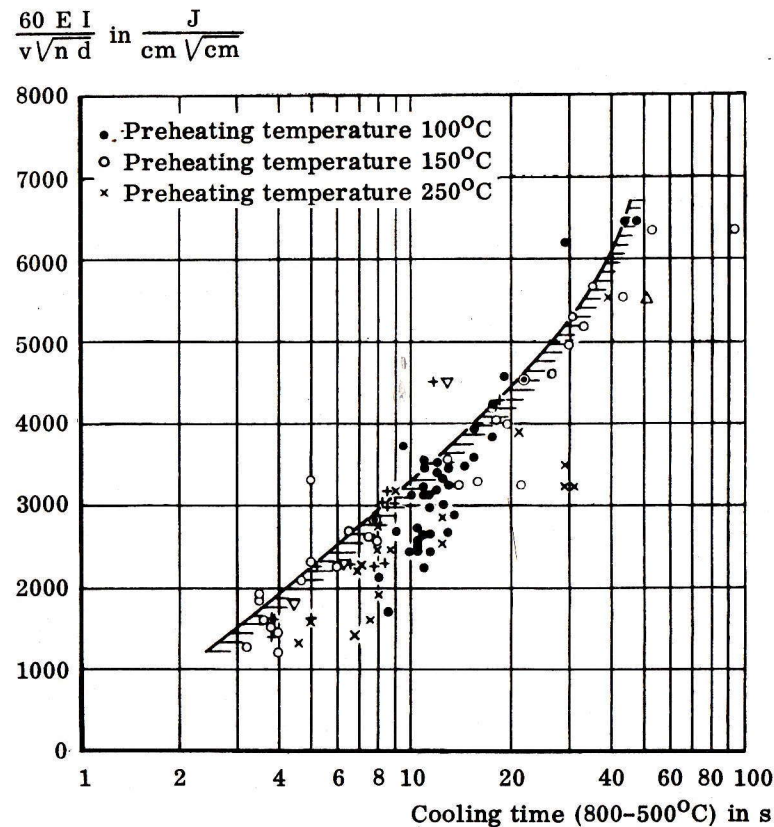


Fig. 5.5. Relationship between cooling time between 800°C and 500°C and the welding conditions.

raise less stringent requirements on the amount of hydrogen in the steel and filler metal.

For heavy thicknesses, it is nearly always necessary to preheat the weldment in order to prevent microstructures sensitive to embrittlement, and also to lower the heat input the higher the preheating temperature.

Diffusion of hydrogen

It is generally agreed that a most important effect of a preheating temperature of less than 200 - 250°C is to increase the time of cooling below 300°C and this enhances the diffusion

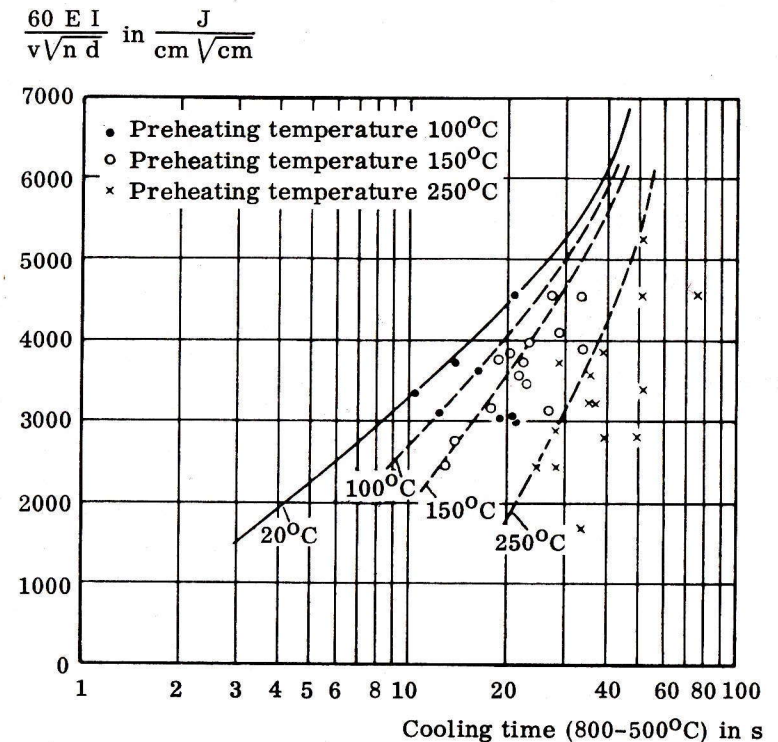


Fig. 5.6. The influence of preheating temperature on the relationship between cooling time between 800°C and 500°C and the welding conditions.

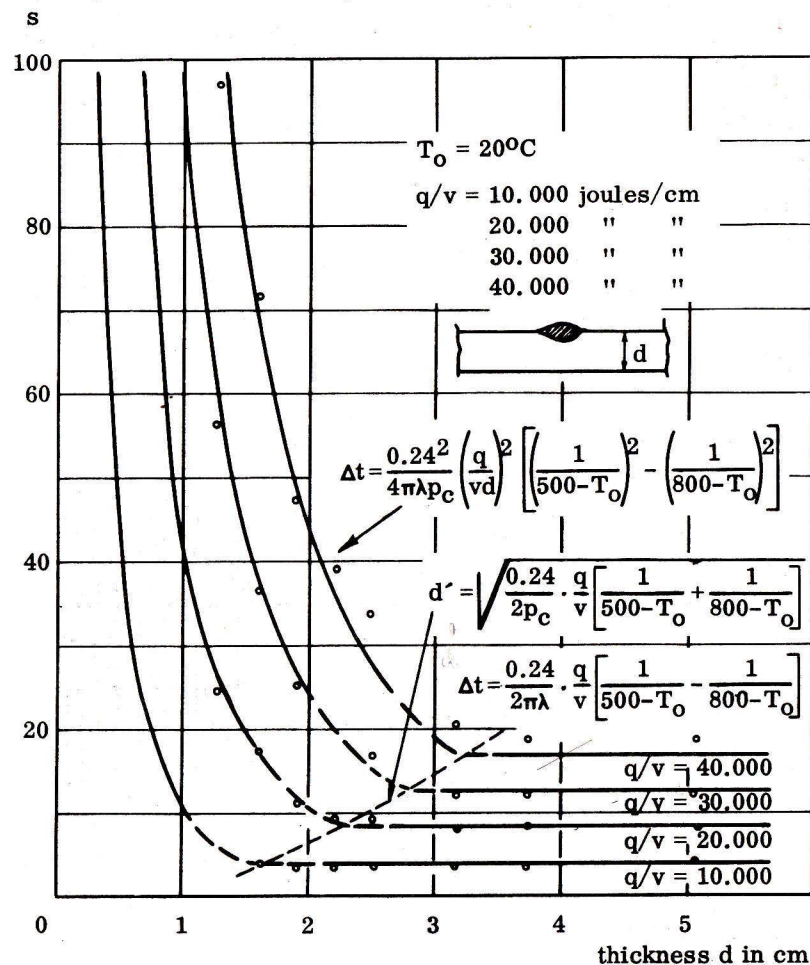
Cooling time Δt (800-500°C)

Fig. 5:7. Relationship between cooling time between 800°C and 500°C and plate thickness when varying the heat input.

of hydrogen from a martensitic HAZ reducing the danger of embrittling the martensite and initiating cold cracks in the HAZ.

Coe and Moreton have established methods for determining the diffusivity of hydrogen at temperatures above normal

ambient (10), and using these data they have worked out detailed methods (11) for calculating the time - temperature - thickness - joint geometry relationships appropriate to the removal by diffusion of a proportion of the hydrogen initially introduced into the deposited weld metal.

When preheating temperatures reach values higher than 250°C, the main effect is to prevent the formation of untempered martensite in high carbon steels by promoting an isothermal transformation of the austenite; the resulting microstructure (bainite) is less sensitive to hydrogen embrittlement and, furthermore, contains less hydrogen owing to its losses by diffusion.

The minimum temperature for preheating and interpassheating under 250°C depends on the following factors:

- susceptibility of the microstructure
- available amount of hydrogen
- heat input
- thickness
- restraint

For a given Ceq, type of electrode and welding parameter, the preheating temperature is dependent upon the thickness (or equivalent thickness) according to a parabolic law (Fig. 5:7). In most cases, the choice of the minimum temperature of preheating following such a diagram was claimed to be satisfactory for mild steels and C-Mn steels (14).

The selection of a preheating temperature based on the HV hardness for preventing cold cracking seems too conservative. In fact, if one considers 350 HV as a critical value, then this is acceptable provided no special precautions are taken on the amount of diffusible hydrogen in the steel and filler metal.

If, according to this level of hardness, preheating temperatures as high as 300°C have to be used, the diffusion loss of hydrogen is so marked during the cooling period that harder and more susceptible underbead microstructures can be allowed. The value of 350 HV corresponds to microstructures which are not susceptible to embrittlement so that in these cases higher

amounts of hydrogen can be allowed. Thus, the two main factors of cold cracking are simultaneously allowed for.

Modifications of level and distribution of residual stresses

A secondary effect of preheating is to reduce the level of residual stresses by reducing the thermal gradients associated with welding. It must, however, be pointed out that:

- Only a preheat of the whole component or an extensive zone near the welded joint is able to decrease both the shrinkage and the local restraint residual stresses, thus reducing the severity of the strain on the embrittled zone.
- A local preheat, or more preferably, an immediate post-heat before the weldment cools, delays the onset of residual tensile stresses until the hydrogen has had time to diffuse, thus reducing embrittlement of the susceptible zone and reducing the danger of cold cracking.

C. CONCLUSIONS

- 1) The sensitivity of C-Mn and C-Mn microalloyed steel to cold cracking depends on the formation of martensite in the HAZ, on the presence of hydrogen and on the restraint conditions. The amount of martensite susceptible to hydrogen embrittlement is related to the hardenability of the steel which is determined by the composition (C eq) of the steel, and to the cooling rate of the HAZ, all of which may be expressed by a hardness value. The hydrogen content of the HAZ may be controlled by using low hydrogen electrodes, preheating, higher heat input and adequate drying of electrodes and base materials. The restraint conditions are related to the yield strength of the steel, to the type, thickness and geometry of welded joint, and to the welding procedure.
- 2) The cold cracking sensitivity of C-Mn and microalloyed steel is at present assessed by the use of welding tests; the severity of these tests is dependent on their restraining and cooling rate features. Hence, the closer these conditions are

to the real joints of the design, the more reliable is the application of their conclusion to the selection of welding parameters. Owing to the wide range of welded joints and design for the same steel, a certain flexibility is necessary in the application of cold cracking tests, either by choosing the most appropriate test or by using a variable stress test for the determination of the characteristics of the steel concerned.

- 3) From the point of view of the transformation of the microstructures, a high heat input gives rise to the same results as high temperature preheating, by reducing the cooling rate between 800°C and 500°C. This is especially applicable to C-Mn steels where the mechanical properties of the HAZ may not be affected by slow cooling in this temperature range.
- 4) Preheating, interpassheating and immediate post-heating treatments have a favourable influence on hydrogen diffusion, permitting the welding of higher Ceq steels with hydrogen controlled electrodes or promoting a greater reliability against cold cracking for microalloyed steels according to thickness, heat input and conditions of site welding.
- 5) The selection of a minimum preheat temperature does not necessarily follow pragmatic rules as is the case with normalizing. Preheat must take account of the other parameters involved in cold cracking: hydrogen content, microstructures, hardness, stress concentration, each varying with the steel, the electrode, the thickness and the welding process. The determination of the cooling time between 800°C and 500°C and hence the application of the CCT data are of prime importance.

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HOT CRACKING

1. HOT CRACKING OF THE WELD METAL

Hot cracks in weld metal occur as a result of the combined effect of thermal contraction, a characteristic of any welding process involving local heating and cooling of the workpiece, and the inability of the metal to withstand this deformation at a temperature at which the last solidifying constituents at the primary grain boundaries are still liquid or too weak to withstand the stresses resulting from this thermal contraction (2, 6).

As a consequence of this combined effect of mechanical and chemical influences, the occurrence of hot cracking is affected by the welding technology employed and by the composition of the weld metal.

The implication of this situation is that an unfavourable composition of the weld metal can be compensated by a careful choice of the welding conditions. Vice versa, lack of care in the execution of the welding operation may result in cracking even in weld metal that satisfies normal standards of composition.

In this connection it should be remarked that most cases of hot cracking which occur in practice do so because of distortion accompanying welding or because some operation in the vicinity of the weld causes a strain across the solidifying weld. In such cases the intrinsic susceptibility of the metal may be extremely low. This demonstrates the necessity for a careful control of the extrinsic conditions during the welding operation.

For the steels considered, the welding processes normally employed do not, as a rule, require excessive measures in the execution of the welding operation with a view to the prevention of hot cracking, as compared with the welding of mild steels. There are, however, some general considerations to be observed as will be explained below.

In the first place it should be remarked that in structures in high tensile steels, higher stresses are applied than in the case

of mild steel, therefore the presence of cracks is, in general, even more harmful as compared with cracks in mild steel.

For welds in high strength steels, therefore, more emphasis may have to be given to the prevention of small cracks than is necessary for certain mild steel applications.

This may necessitate a restriction in the range of welding variables and of filler material in comparison with what can be tolerated for less critical applications. As a matter of course welding conditions that are known to increase the risk of hot cracking should be avoided. To be mentioned are: excessive travel speeds and very thin or very short welds (in relation to the work piece dimensions) and welding sequences that induce the accumulation of shrinkage deformations.

When welding is carried out with manual electrodes, the danger of hot cracking is reduced by using "basic covered" electrodes, which deposit a weld metal of better metallurgical quality as compared with other types of electrodes.

For this and other reasons, such as the prevention of hydrogen induced heat-affected zone cracking, basic covered electrodes are to be recommended, and are indeed widely used, for the welding of high tensile steel.

A second general observation has to be made in relation to welding processes characterized by a weld metal which is wholly derived from the parent metal. Although some of these processes (such as gas- or TIG-welding without filler wire) are only appropriate in a rather restricted range of application for other reasons, it may be in place to emphasize that with such processes hot cracking can only be avoided when the weld metal composition (and consequently, the steel composition) is within rather narrow limits, which in most cases will not be met by the steels considered.

For "deep penetration" welding methods using a filler material, the same argument limits the applicability and a careful choice of the welding conditions may be necessary to avoid cracking, the more so when the contents of harmful elements are relatively high, as will be discussed later.

In discussing the effect of steel composition on weld metal cracking in some detail, it should be kept in mind that eventually only the weld metal composition has a direct effect. Thus, the effect of steel composition depends upon the percentage to which

the filler metal is diluted by the parent steel. It is not possible, however, to mention "critical" values for the weld metal composition because of the very strong interaction of welding parameters and chemical influences, as observed before. As a general rule it may be stated that the deposition of the weld metal in shallow beads tends to favour hot cracking, for a given weld metal composition, especially in the first run in heavy plates or otherwise heavily restrained weldments. As a result, welding procedures for such applications, for a given filler metal and steel composition, normally tolerate less dilution of the weld metal by plate material than welding procedures for the deposition of heavy layers.

Consequently, certain automatic welding processes will allow percentages of dilution which are usually not acceptable for manual electrode welding. As a rough indication, percentages of about 50 % and 25 % may be mentioned as maximum values to be acceptable under average conditions for heavy layer automatic welding and manual electrode welding respectively. One of the consequences of the relatively large "pick up", characteristic of certain automatic welding processes, is that these processes are more sensitive to variations in base metal composition than with manual electrode welding.

For the majority of steels in the group considered, the raised yield strength is at least partly derived from an increased manganese content (as compared with mild steel).

In the percentages that normally enter into the weld metal, manganese by itself is not manifestly deleterious and may even have a favourable influence when a relatively high sulphur content is present. For practical purposes a Mn-S ratio of at least 20 in the weld metal may be taken as a safeguard against hot shortness by sulphur. For the steels considered and for the welding methods usually applied, this limit will as a rule be surpassed if the amount of sulphur in the steel is limited to a maximum of 0.04 % in the ladle.

It must, however, be remarked that at a manganese content higher than about 3.5 times the silicon content (in the weld), failures by hot cracking in submerged arc welds have been reported as a result of the formation of iron manganese silicate, coagulated in liquid form at dendrite interfaces during the solidification of the weld metal (1).

For the joining of manganese alloyed steel, therefore, due attention should be paid to this danger in selecting the filler metal to be applied.

In the normal ranges of composition, carbon, next to sulphur (4), has the greatest effect upon hot cracking.

In the weld metal derived from manual electrodes, carbon contents over about 0.13 % should be avoided. Although the undiluted weld metal usually has a carbon content not higher than about 0.10 % and accordingly will tolerate some admixture of plate material, it will be obvious that a carbon content in the steel of higher than about 0.20 % may impose limitations upon the application of welding processes at percentages of admixture of about 25 %.

Next to carbon, phosphorus also has an unfavourable effect upon hot cracking (4). For carbon contents of up to about 0.20 %, a limitation of phosphorus to a maximum of 0.040 % in the ladle will be sufficient.

Most of the other elements currently applied in the types of steel dealt with here, can hardly be expected to enter into the weld metal in quantities that would promote the occurrence of hot cracking directly. Some reservation should be made for nickel, an element that promotes hot cracking if sulphur is present. For nickel percentages up to about 1.5 %, however, there seems to be no reason to lower the sulphur limit in the steel of 0.040 % mentioned.

Little is known about the direct effect of the usual micro-alloying elements in the small quantities normally involved. It has, however, been reported that strongly deoxidizing additions in the weld, such as aluminium, may have an unfavourable effect on the distribution of sulphides by the increased sulphur solubility as a result of the lower oxygen content of the metal (1). It is believed that under critical conditions "overkilling" thus may indirectly impair the weld metal ductility. At present it is not possible to give limits for the amount of deoxidizing elements to be tolerated but the argument given may be an extra reason not to overdose the strongly deoxidizing elements.

In the absence of detailed knowledge about the combined effect of elements on the appearance of hot cracking, it is to be recommended that newly developed steels are carefully investigated on their hot cracking sensitivity. As most of the alloying elements,

except manganese, are to be considered as "impurities" from the viewpoint of hot cracking, it seems necessary to include in such investigations steels that are in the upper region of the composition range (except manganese).

For the investigation of the steel for its liability to hot cracking, preferably comparative tests are to be used in which the weld metal is subjected to delayed shrinkage deformation. A number of such tests are described in the document IIS/IIW 93-62 (3) and in (7).

2. HOT CRACKING OF THE PARENT METAL

Hot cracks in the metal adjacent to the weld are less common than weld metal hot cracks. For carbon manganese and micro-alloy steels such cracks, if present, will usually have more serious consequences than similar cracks in mild steel because of the higher reaction and loading stresses which will occur and because of the reduced ductility in the transformed zones.

For the group of steels considered, hot cracking in the parent metal is not to be expected if the levels of sulphur and phosphorus as proposed in the preceding chapter are observed and if the level of the other constituents is low enough to prevent the formation of coherent liquid interfaces below the melting temperature of the steels.

Mention should be made of the possibility of crack formation if large non-metallic inclusions are present in the zones of a weld subjected to shrinkage deformation (1).

Because of the reasons mentioned, the arguments for avoiding these coarse non-metallic inclusions in high tensile steel are stronger than in the case of mild steel.

Although parent metal hot cracking is a relatively uncommon phenomenon in the types of steels considered, some concern is justified because cases have been reported in which the primary causes of the phenomenon could not be detected with certainty.

Under these conditions it seems justified to investigate newly developed steels for their liability to this phenomenon.

This can easily be done by careful microscopic investigation of the zones close to the weld metal in cross sections of pads, welded to the plate surface in one layer. It has been suggested that sectioning under an angle of about 10° with the plate surface

increases the possibility of detecting minor cracks (5). It is also advisable not to limit the investigation to welds on the plate surface, but to include pads welded on surfaces in the two other main directions.

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LAMELLAR TEARING

1. DEFINITION

Lamellar tearing is a form of cracking usually associated with highly restrained welded joints whose geometry is such that the material is stressed normal to its rolling plane. This feature (Fig. 1) distinguishes lamellar tears from hydrogen induced HAZ cracking, as the tears are often in the material below the HAZ.

Lamellar tearing is influenced by the following factors:

- Type of joint
- Thickness of material
- Welding procedure variables
- The transverse strength of the material.

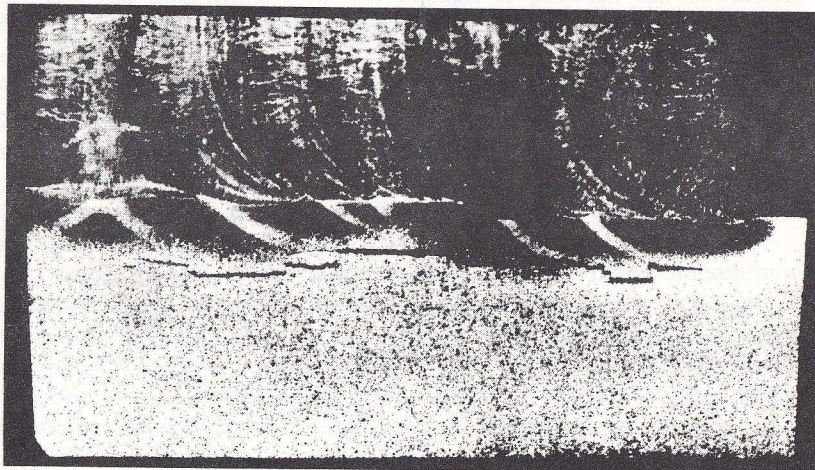


Fig. 7:1. Detail of a lamellar tear showing stepped appearance.
(Original 4x, reproduction at 68 %).

2. IMPORTANT FACTORS

2.1. Joint configuration

Joints with which lamellar tears are commonly associated (see Fig. 7:2):

- a) T joint with fillet welds
- b) T joint with compounded butt and fillet welds
- c) Corner joint with butt weld

Stresses across a weld arise from local shrinkage augmented by any reaction stresses associated with the structure. In a) the stresses act at an angle to the rolling plane of the material. With b) the stress field being more extensive and the stresses acting nearly normally to the base material, both factors may increase strain normal to the rolling plane of the base material.

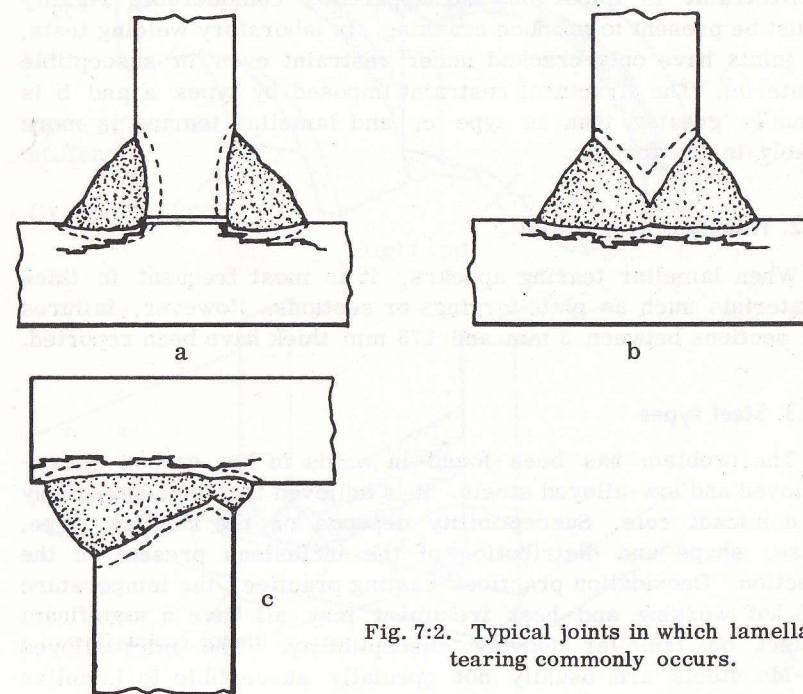


Fig. 7:2. Typical joints in which lamellar tearing commonly occurs.

As expected there is a higher incidence of lamellar tearing with joint b). Corner joints c) can be particularly troublesome.

The position of the joint is an important factor from the point of view of imposed restraint. Restraint may be defined as the restriction of the movement of the various joint components, which would normally occur as a result of expansion and contraction of the weld metal and adjacent regions during welding. Structure types can be considered under three main headings as illustrated in Fig. 7:3:

- Nozzle or penetrator through a rigid curved or flat plate involving joints a) and b).
- Stiffener or rigid end in a cylindrical vessel involving joints a), b) or c).
- T or cruciform joints as part of fabricated box-sections or large I-beams involving joints a) or b).

Restraint is important and apparently considerable rigidity must be present to produce cracking. In laboratory welding tests, T joints have only cracked under restraint even in susceptible material. The structural restraint imposed by types a and b is usually greater than in type c, and lamellar tearing is more likely in the former.

2.2. Thickness of material

When lamellar tearing appears, it is most frequent in thick materials such as plate forgings or sections. However, failures in sections between 3 mm and 175 mm thick have been reported.

2.3. Steel types

The problem has been found in welds in low carbon micro-alloyed and low-alloyed steels. It is believed that inclusions play a dominant role. Susceptibility depends on the number, type, size, shape and distribution of the inclusions present in the section. Deoxidation practice, casting practice, the temperature of hot working and heat treatment may all have a significant effect on lamellar tearing susceptibility. The microalloyed C-Mn steels are usually not specially susceptible to lamellar

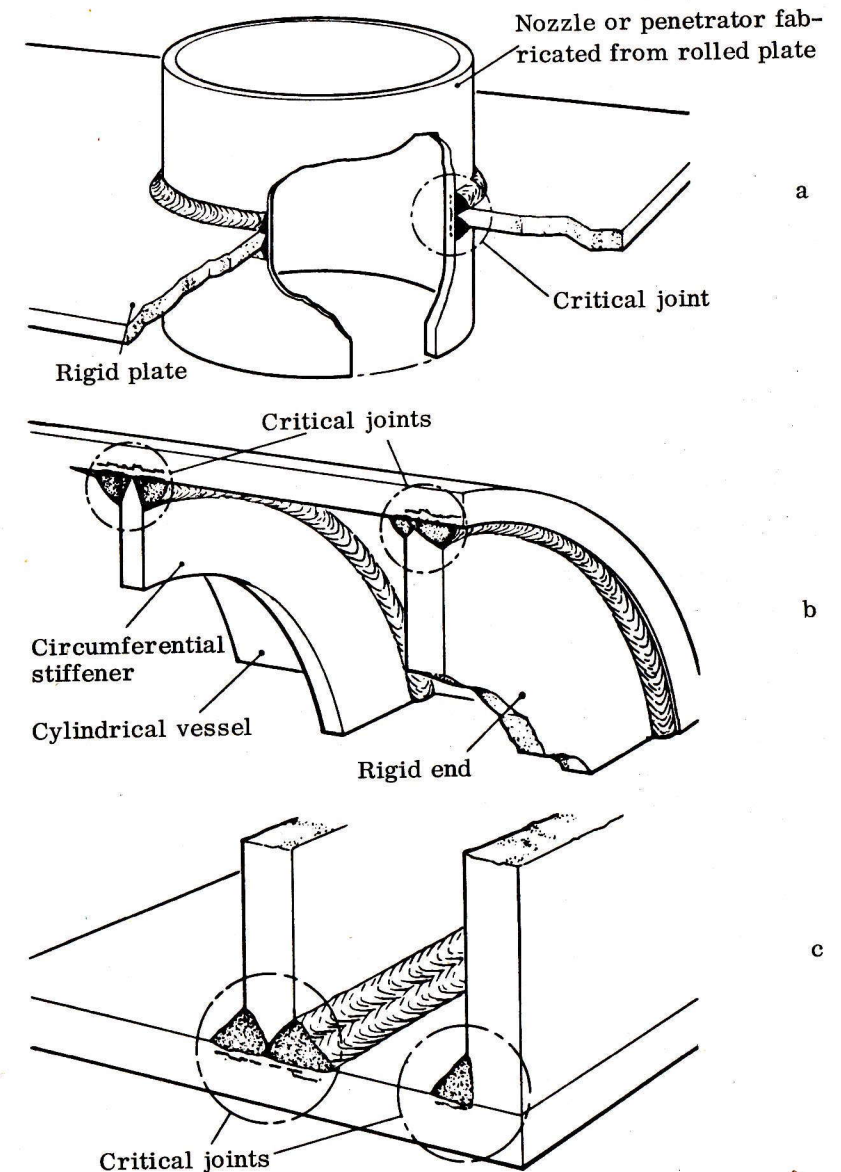


Fig. 7:3. Typical assemblies in which lamellar tearing occurs.

tearing although the residual stresses may often be higher in these steels.

3. PREVENTIVE MEASURES

The welding engineer can take the following measures to reduce the incidence of lamellar tearing:

- 1) Re-design of a particular part or joint of a structure
- 2) Choice of procedural welding variables
- 3) Selection of materials

3.1. Design

Joints where reaction strains are present and which might fail can be divided into two types:

- a) Those which would reduce the efficiency of the structure
- b) Those whose failure would have catastrophic consequences for the structure

Examples of the first type are those associated with nozzles or tubes made from rolled plate in pressure vessels or stiffeners in cylindrical structures, as illustrated in Fig. 7:3.

Examples of the second are attachments for stiffeners or T butt welds for major load carrying joints which are stressed in tension, e.g. bridges, cranes, etc. In both cases the design of the joint should be changed so that the total strain applied through the thickness is minimized (Fig. 7:4).

Cracking in a corner joint is shown in Fig. 7:5. Modification of the preparation is possible to change the orientation of the weld so that reaction stresses are no longer normal to the plate surface.

Summarising the design approach:

- 1) Re-design the critical region of the structure
- 2) Alter the joint preparation.

Approach 1) at the design stage is the more satisfactory where it is practicable.

3.2. Welding procedures

The precautions suggested below to reduce the risk of lamellar tearing are based upon practical experience rather than experimental results.

- a) The use of low strength weld metal.
The strength of the weld metal should be as low as is commensurate with the joint design in order to reduce strain occurring in the subsurface layers of the material.
- b) Modified weld run procedure.
This technique, which is shown in Fig. 7:6a, is probably beneficial because it reduces the short transverse strains and

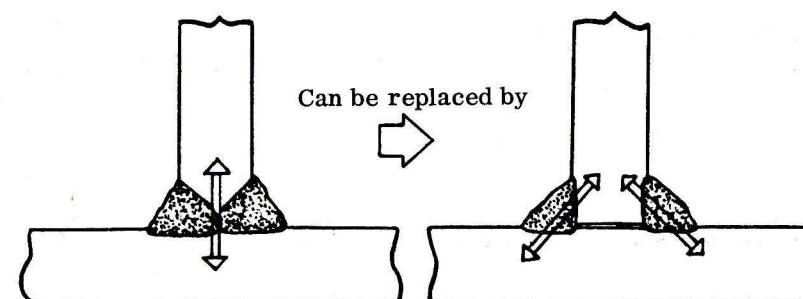


Fig. 7:4. Modification of a T-butt weld to reduce the risk of tearing.

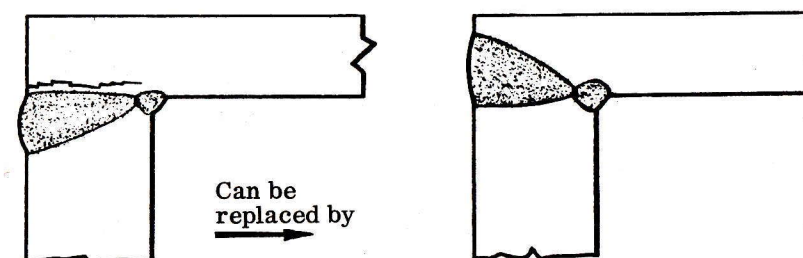


Fig. 7:5. Modification of a corner weld to reduce the risk of tearing.

ensures that the highest welding strains are imposed on pre-deposited layers of weld metal rather than on parent material.

c) Buttering.

This technique (Fig. 7:6b) is often regarded as standard practice when suspect materials have to be used in critical joints, i.e. when re-design of the joints is either impossible or impracticable.

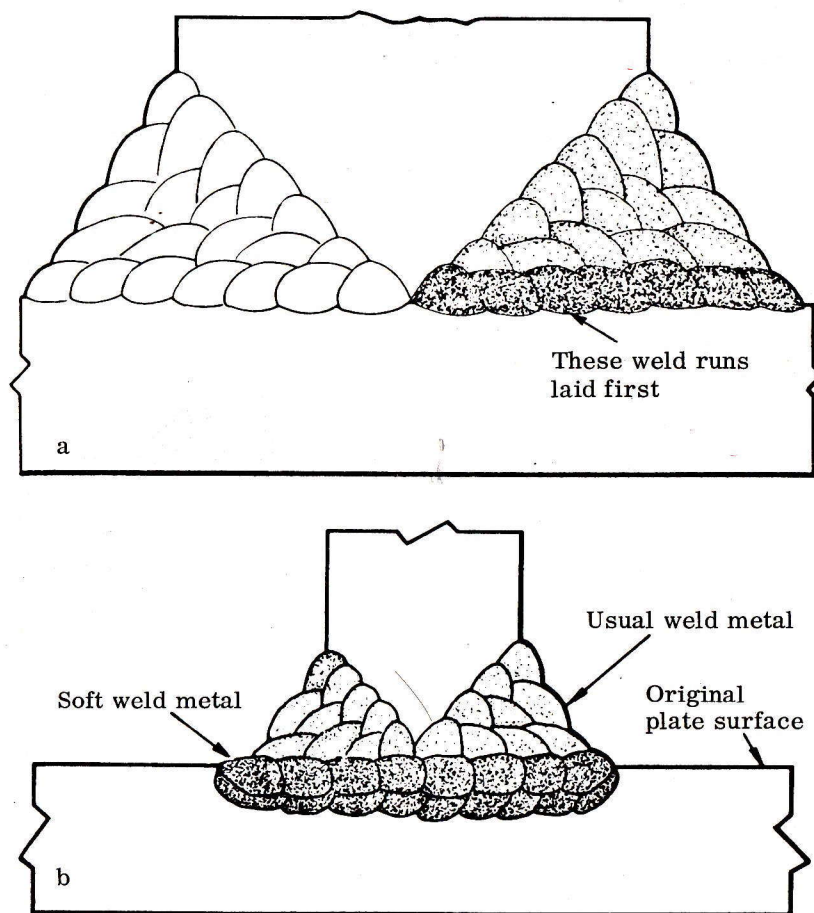


Fig. 7:6. Precautionary and remedial techniques: a) modified run procedure, b) buttering.

d) Balanced welding.

When fabricating T butt welds a balanced welding technique should be employed with both fillets of the weld made concurrently. This helps to reduce the high strain at the root of the weld caused by the bending moment.

3.3. Methods of material selection

No selection procedure can be entirely based upon steel class, type and method of manufacture. The susceptibility of individual materials could possibly be measured and checked in ways considered below.

a/ Ultrasonic examination

Normal pulse echo techniques (2.5 to 5 Megahertz = MHz) reveal laminations in plate (>5 mm).

High frequency ultrasonic attenuation techniques (12 MHz) have been reported.

Development of this technique is at an early stage and some results indicate that it needs further work to establish its reliability.

b/ Welding tests

Welding tests have been proposed as a means of assessing susceptibility to lamellar tearing. They have the advantage of simulating practical conditions and enabling the effects of procedures to be assessed. Unfortunately they suffer from serious disadvantages:

- Poor reproducibility
- There is no easily measurable parameter for assessing the susceptibility to tearing of the material
- Reaction strains which occur in practice can only be simulated by reproducing a real joint
- They assess only a small sample of material and give no information about the remainder.

The conclusion is that welding tests may produce information on crack morphology and propagation, which may possibly be used to assess welding procedures. They are not suitable for the selection of materials prior to fabrication.

c/ Destructive testing

A small scale tear test and a notched bend test have been developed which help in a study of both initiation and propagation. The manufacture of the bend test is outlined in Fig. 7:7. Good correlation with this test can be obtained with the short transverse tensile test.

4. CONCLUSIONS

Cracking occurring beneath heavy fillet welds, particularly in thick materials, has become a major problem in the welding of structural and pressure vessel steels. This chapter has described

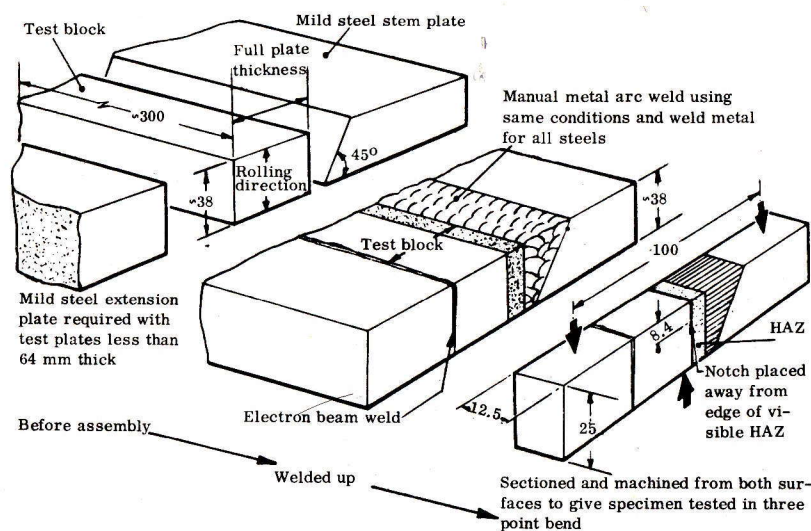


Fig. 7:7. Manufacturing procedure for bend test specimen.

the problems in some detail and has outlined the influence of the joints and the types of structure in which such cracking occurs, the welding procedure and the choice of material.

Some possible approaches to overcoming the problem have been considered, including:

- 1) Design against lamellar tearing
- 2) A number of remedial welding measures
 - the use of low strength weld metal
 - modified run procedure
 - buttering
 - balanced welding.
- 3) The selection of materials which are resistant to lamellar tearing, proposed selection methods based on non-destructive tests, welding tests and small scale destructive tests are mentioned.

If lamellar tearing occurs, a combination of remedies is possible by changing the welding procedure, modifying the joint design and by choosing a less susceptible material. It is emphasized that microalloyed C-Mn steels are not especially susceptible to lamellar tearing.

For more detailed treatment of the subjects the reader is referred to the following list of references.

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MECHANICAL PROPERTIES OF THE HAZ AND THE EFFECT OF POST WELDING THERMAL TREATMENT

The HAZ is currently defined as the zone of the base material in the near vicinity of the fusion bond, in which transformations have occurred during welding resulting from the heat input in the welding process. The present considerations extend the notion of the HAZ towards those regions of material where a simultaneous action of plastic strain and temperatures less than A_{c1} has resulted in modifications of the mechanical properties of the base metal without apparent changes in the microstructures.

A. CASES WITHOUT POSTWELDING HEAT TREATMENT

The modifications of structural and mechanical properties in the vicinity of a weld arise from a rapid thermal cycle occurring with varying configurations away from the fusion bond.

The peculiarities of this thermal cycle (peak temperature, heating time, cooling time and cooling rate) are a function of the heat input, the welding process, the conditions of preheat and the geometry of the weld joint.

A number of tests are or have been used to study the phenomena occurring in the so called heat-affected zone. Continuous Cooling Transformation diagrams of a number of steels are available. However, in a great number of cases when cooling rates typical of welding conditions have been applied, the heating and holding times at temperature were usually not representative of welding. The same is the case for a number of hardenability tests (Jominy), which for welding are useful in rating different steels according to hardenability.

Simulation tests, in which thermal cycles similar to those occurring during welding are applied on test pieces, possibly give better information on the transformation characteristics, hardenability, strength and notch toughness properties.

Finally, the situation of actual joints may be clarified either by sampling conventional laboratory specimens at different locations, or by testing the overall joint in a full scale test, with the possible use of notches positioned at certain distances from the fusion bond.

Within the heat-affected zone, two regions may be separated according to whether the peak temperature reached during welding does or does not exceed the temperature at which the precipitates are taken into solution. Practically, this borderline cannot be clearly determined, owing to the existence of a temperature range in which the carbonitrides are only partly in solution, and to the complexity resulting from the superposition of several welding passes or runs.

Providing the peak temperatures are less than say 1000°C, important changes of the precipitates are unlikely to occur in the corresponding zone of the HAZ. However, in welding processes with high specific heat input, some instability of carbides and nitrides could occur, due to the relatively slow cooling rate below this temperature range.

In the immediate vicinity of the fusion bond, vanadium and niobium carbonitrides are dissolved. The effect of dissolved Nb and V precipitates on the hardenability of steel is claimed to be negligible. The zones of reprecipitation of Nb and V during cooling and their local effect on the mechanical properties will vary according to the cooling time. In other words, the important parameter is the cooling rate, which depends on the welding process and the geometry of the joint, and the contour or zone where a reprecipitation is likely or unlikely to occur.

With ordinary manual welding reprecipitation does not usually occur. There is therefore no significant difference in the strength and toughness between the HAZ of as-normalized microalloyed and ordinary C-Mn steels, because of the small direct effect of precipitates in these steels (see Chapter 2). If as-rolled microalloyed steels are compared with C-Mn steels of the same strength, the difference in base analysis must be taken into account, the microalloyed steels generally having an appreci-

ably lower C content. In this case the HAZ of the microalloyed steel can sometimes have a lower strength and a better toughness. However, actual joints do not show a lower strength than the parent material, probably because of the small size of the softened zone and its position in a residual stress field. With processes with high heat input and associated slow cooling, reprecipitation can occur. This could make the HAZ of microalloyed steel stronger and in certain cases more brittle than the HAZ of C-Mn steel.

In the regions of the HAZ where the peak temperatures are less than A_{c1} , one can anticipate modifications in the distribution and in the density of dislocations, a growing polygonisation of the ferrite, the formation of an oversaturated solid solution of C + N, and, in the case of C-Mn-V steels, a possible reprecipitation of vanadium carbides at temperatures over 650°C. These phenomena occur in conjunction with plastic strain. Their influence on the toughness of high strength steels is claimed to be less pronounced than in the case of mild steels.

Besides the mentioned effects of welding on the changes in the heat-affected zone below A_1 , the changes which can occur when cold worked or bent profiles are welded should be considered.

In the zone in the temperature range 100 to 400°C, signs of strainageing due to the agglomeration of precipitation (C + N) atoms in the form of hexagonal $-Fe_3X$ phase can occur. This strainageing phenomenon has a marked influence on the impact value of the HAZ.

When these steels are welded, the HAZ in the temperature range of 550°C to 750°C sometimes shows signs of the annihilation and polygonization of dislocations. A $\langle 011 \rangle$ polygonization network which forms sections can act as a barrier against dislocation movement. This phenomenon accounts for the plasticity drop which is often observed even during the welding of pure materials. In areas where the deformation reaches the critical value (5 to 15%), namely in the decarburized surface zone with lower pearlite content, recrystallisation and grain growth can also be observed.

Concerning the available data on the mechanical properties of C-Mn and C-Mn microalloyed steels in the HAZ, Aronson (1) observed that the thermal cycle in simulated samples of a 0.03 to 0.04% Nb steel produced a softened zone, but no loss in

strength was observed in actual weldments. Kage (2) investigated the properties of a series of steels; although with a high heat input electroslag weld a softened zone was observed in hardness tests, this did not affect the results obtained when tensile testing butt welds.

Fracture toughness measurements were made in the grain coarsened region 1.6 mm from the fusion face and these showed a significant embrittlement in the higher heat input electroslag welds.

Winn (3) reported for a Q steel (ND1 0.15% C, 1.0% Mn, semi-killed as rolled) and a U steel (ND1 0.14% C, 1.23% Mn, semi-killed as rolled, with a coarse ferrite and pearlite structure) there was a more severe embrittlement in the subcritical zone than in the transformed HAZ in the case of two typical heat inputs (10 kJ/cm and 24 kJ/cm). In contrast, in a T steel (Admiralty B quality, 0.17% C, 1.43% Mn, grain controlled and normalized) fully killed and Al-treated, a significant embrittlement in the transformed HAZ was found, which was considerably more severe than the subcritical HAZ embrittlement. In the three cases, the resistance to fracture initiation was measured using the COD technique of general yielding fracture mechanics.

The properties of the strained and aged zone in carbon and carbon-manganese steels have been extensively studied by a Working Group of Commission IX on Metallurgical Concomitants (4). Lagasse (5) reviewed this work and drew attention to a trough in Charpy impact energy 5 to 10 mm from the fusion face. Recently fracture toughness tests using COD techniques (6) have confirmed the embrittlement in the thermally strained zone below the A_{c1} . In the presence of a notch, thermal cycling below the A_{c1} increased the transition temperature of two semi-killed steels by about 40°C and an aluminium treated steel by 20°C.

Sub-Commission IX-G of the IIW collected a large amount of data in order to synthesise the general trend observed in different cases of weld joints and welding processes (7). The enquiry covered 46 parent metals, which were classified into four categories corresponding to their grades and 84 different butt welded joints.

- 19 joints in C-Mn fully killed and normalized steels (Grades 50/33 and 52/36)
- 15 joints in C-Mn-Nb semi-killed and normalized steels (Grades 52/36 and 56/40)
- 41 joints in C-Mn-Nb fully killed and normalized steels (Grades 52/36 and 56/40)
- 9 joints in C-Mn-V fully killed and normalized steels (Grade 52/42)

Summarized sheets for each grade of steel, containing the essential data on the parent steel, welding conditions and mechanical properties in the HAZ were reported in Doc. IX-G-129-67.

The following comments are noted.

- 1) The parent materials classified according to their grades made up homogeneous groups duly representative of the class to which they belonged, both from the point of view of their chemical composition and from that of their C equivalent.
- 2) A wide range of welding conditions was covered and the heat input values were typical of the welding processes used. In some cases, the same welding process and the same type of joint had been deliberately put to work under varying input conditions.
- 3) The mechanical properties in the HAZ were only given for the Charpy V notch test piece.
- 4) Relatively few transverse weld tensile test results were available. There were still fewer Vickers hardness results and, where there were, hardness numbers had been determined by the various laboratories under loads varying between 5 and 30 kg.

The purpose of the above comments is to call attention to the fact that the conclusions deduced from this study are imperfect as is most often the case when test results collected from an international enquiry are simply compiled without a precise and detailed programme having been drawn up, specifying all the parameters and information required.

It was however possible to draw a few conclusions of general validity:

- a) The tensile properties of the welded joints are usually not adversely affected by the HAZ of C-Mn or C-Mn micro-alloyed steels in the as-welded condition. Failure generally occurs in the parent metal.
- b) The maximum hardness values observed in the HAZ seem to be associated with the C content or in line with the C equivalent value. But due allowance should be made for the variable loads applied to determine the hardness value and for the position of the test (not generally mentioned); this fact makes the interpretation of the test results more difficult.
- c) The notch impact values determined by means of the V notch Charpy impact test for one or more temperatures in the transition zone is generally as high as that of the parent metal when the notch is located in the vicinity of the fusion bond (2 mm).

Further from the fusion bond, decreasing impact values at a distance of 3 to 5 and even 7 mm may be observed. This phenomenon occurs in the thermo-thermal zone of the HAZ.

A few impact values, abnormally low, may be observed in the HAZ in the vicinity of the fusion bond. The joints concerned are generally those with a very high heat input value (e.g. 200 kJ/cm).

B. EFFECT OF POST WELDING THERMAL TREATMENT

General

Residual stresses of up to yield point magnitude can be set up during any welding operation. These can be detrimental to subsequent operations in that they help to promote failure by brittle fracture, fatigue or stress corrosion and may lead to dimensional instability. Although in many cases the presence of residual welding stresses has little effect on subsequent service behaviour, it is often necessary to reduce or remove these stresses. These are generally reduced by means of a heat treatment at a relatively high temperature.

In addition to reducing the residual stress level, this heat treatment is analogous to a tempering treatment and the metallurgical structure and hence the mechanical properties of plate material, weld metal and the heat-affected zone can be changed. Normalizing of welded structures is not considered.

Relief of residual stress

Although specific requirements vary from code to code, most design codes specify a postweld heat treatment of 550–650°C for plate thicknesses in excess of 38 mm.

Isothermal relaxation tests have shown that the relaxation behaviour of C-Mn steels depends on the aluminium addition during steel making (8). Nevertheless, at temperatures in excess of 525°C, appreciable relaxation occurs in all C-Mn steels.

The heating rate of large structures is slow and significant relaxation would be expected to occur during this period. Thus in practice the holding time has little influence on the total relaxation, the isothermal relaxation rate becoming practically zero after a few hours (9, 10).

At temperatures in excess of 550°C Murry and Constant (10, 11) have shown that the relaxation of heat-affected zone and weld metal is comparable to that obtained with the plate.

C. EFFECT OF THERMAL TREATMENT ON PARENT PLATE AND WELD METAL

Tensile properties

Variable results have been observed on the effect of thermal stress relieving on tensile properties. Watkins and co-workers (12) showed that in some instances a prolonged stress relief of 36 hours at 600°C produced an increase in tensile strength of 2.5 kgf/mm², whilst in other cases there was a deterioration of 5.6 kgf/mm². The general effect, however, was a deterioration of 1.5 to 3 kgf/mm² in tensile strength with a corresponding decrease in yield point. The deterioration in properties can be limited by reducing the holding time or reducing the temperature.

A similar deterioration is observed in the tensile properties of weld metals. Again the results are variable; only a slight deterioration is observed in some cases, whilst others show deterioration of 4.6 kgf/mm² (12) and 11.5 kgf/mm² (13).

High temperature properties

Wood and Swift (14) have studied the creep behaviour of 0.15% C, 1.2% Mn, 0.15% Si pressure vessel steels after stress relieving at 575 to 600°C for 5 hours and 36 hours at 650°C. The results show that a prolonged stress relief causes the creep strain at 2000 hours to be increased by a factor of 5 to 10 for silicon killed steels, but there is no significant effect on the aluminium grain refined steel. The creep rupture strength is also reduced.

The authors attribute this phenomenon to the formation of silicon nitride during heat treatment; in the case of the aluminium grain refined steel, the available nitrogen is already "fixed" as the stable aluminium nitride.

Impact properties

A prolonged stress relieving treatment produces an increase in the Charpy V notch transition temperature of both plate and weld metal (12, 15). Shifts of 20 to 40°C have been observed after 36 hours at 650°C but the effect is not as marked at lower temperatures or shorter times.

Temper embrittlement can occur in some steels and a Working Group of Commission X (16) on thermal stress relieving treatments has concluded that, when the manganese content is higher than 1%, molybdenum additions do not make the steel immune from this phenomenon.

Experience gained from large scale tests and service behaviour

The work of the Sub-Commission "Brittle Fracture in Service" of Commission X has shown that the risks of brittle fracture in service are significantly reduced by an effective thermal stress relieving treatment.

This beneficial effect has been confirmed on large scale fracture initiation tests (17, 18). These test results show that whilst large improvements are obtained by stress relieving in the temperature range 600 to 650°C, some benefit is obtained at all temperatures in excess of 450°C.

D. EFFECT OF POST WELD HEAT TREATMENT ON THE STRAINED AND AGED ZONE

The heat treatment for the relief of residual stress produces structural changes in the strain-aged region 5 to 10 mm from the fusion face and in the transformed heat-affected zone. The effects in the strain-aged region have been extensively studied by the Working Group on Metallurgical Concomitants (4); these effects were recently reviewed by Lagasse (5).

In the experiments reported by Dolby and Saunders (6), heat treatment for 1 hour at 650°C removed the embrittling effect, the transition temperature being restored to that of the parent plate. Other investigations (19, 20) have also shown that a heat treatment in the range 600 to 650°C removes the embrittling effect in the thermally strained zone.

E. EFFECT OF THERMAL TREATMENT ON THE TRANSFORMED HEAT AFFECTED ZONE

1. Carbon-manganese steels

Tensile properties

Data are limited on the effect of thermal stress relief on the tensile properties of the heat-affected zone in C-Mn steels. In most cases the tensile properties have been determined by means of cross welded samples and, except in the case of defective welds, these samples generally meet the minimum tensile requirements of the plate.

Some results with small Hounsfield specimens taken parallel to the weld, with the bulk of the sample within the gauge length, show that the yield strength of the HAZ, even after a prolonged

stress relieving treatment, is higher than the corresponding plate material.

Impact properties

An examination of the impact properties of the HAZ of stress relieved 76 mm plate C-Mn steels has shown that the impact properties in the heat-affected zone are better than those of the parent plate. The impact properties of automatic welds are, however, sometimes inferior to those obtained by manual welding.

The effect of a thermal stress relieving treatment on the properties of the transformed heat-affected zone measured in simulated samples (21) and bead-on-plate techniques (6) had been to restore the properties of the transformed region to that of the unwelded plate.

In some instances low impact values are recorded in the HAZ. Thus, there are isolated examples on procedural testing during production welding of 76 mm stress relieved silicon-killed steel where HAZ impact values of the order of 1.4 kgfm at -10°C have been noted compared with 2.8 to 4.2 kgfm at -10°C for the unwelded plate (22).

Crack opening displacement tests

Dolby and Saunders (Welding Institute, Abington) have made a series of bead-on-plate tests on three carbon-manganese steels. They observed (6) that embrittlement occurs in the transformed heat-affected zone and that the production of a mixed structure of upper and lower bainite and auto tempered martensite containing 0.17% C and 1.4% Mn caused an increase in the transition temperature of more than 100°C .

In the three steels investigated, heat treatment for 1 hour at 650°C removes the embrittling effect of welding completely. Stress relieving of the coarse HAZ structure reduces but does not entirely remove the embrittlement.

High temperature properties

Creep rupture tests have been carried out at 375 and 425°C on composite samples taken from externally stress relieved

(5 hours at 600°C) manual and automatic welds in a 76 mm plate of a 0.2% maximum C, 0.7 to 1.0% Mn, 0.15 to 0.25% Si steel.

The specimen which incorporated the HAZ in the centre of the gauge length was taken at plate depths of T/4 and T/2. Failures always occurred in either the parent plate or weld metal; in no instance was failure associated with the heat-affected zone (23).

No stress relief cracking was observed (24) in C-Mn steel, thus confirming the good creep rupture of this class of steel.

2. C-Mn microalloyed steels

Little data are available on the effect of thermal stress relief on the HAZ of microalloyed steels. Evidence of secondary hardening producing an embrittling effect in the HAZ has been observed in higher tensile steels, but no data are available on microalloyed steels. Low impact values have been reported with CO_2 enclosed welding and submerged arc welding of C-Mn-Nb and C-Mn-V microalloyed steels. It has been suggested that the phenomenon is generally associated with welds of high heat input.

CONCLUSIONS

- 1) In usual manual welding conditions, the mechanical properties of the heat-affected zone of C-Mn steels are not usually adversely affected. In cases of very high heat input welding processes, low impact values may be observed using Charpy impact specimens.

In the stress relieved condition in C-Mn steel, the tensile and impact properties of the transformed heat-affected zone are of the same order as those of the parent plate. Isolated instances are quoted where the impact properties are inferior to those of the parent plate and these could be accentuated by high heat input welds.

- 2) In microalloyed steels, the mechanical properties of the heat-affected zone are usually in the same range as the values observed in the parent metal, or, at least, as the mechanical

properties of the heat-affected zone of plain C-Mn steel welded under similar conditions. A softened zone can occur in the transformed heat-affected zone but this only affects the properties of a welded joint at high heat input.

Associated with this softened zone there is sometimes an area of reduced toughness. No data are available on the effect of thermal stress relief on the properties of this softened zone but, from an examination of the metallographic structure no improvement could be expected.

- 3) The transition temperature as determined in COD tests of material strained below A_{c1} , i.e. 5 to 10 mm from the fusion face, can be raised 40°C. Thermal stress relieving for 1 hour at 650°C or longer times at lower temperatures restores the properties to those of the parent plate.
- 4) Thermal stress relieving for 1 hour per 25 mm of thickness in the temperature range 550 to 650°C reduces the residual welding stress in carbon manganese steels. At slow heating rates, the holding time has little influence on the resultant stress level.

Service experience has shown that an effective thermal stress relief significantly reduces the risk of the initiation of brittle fracture in C-Mn steels.

Fracture initiation characteristics, as defined in wide plate testing, are improved by thermal stress relief. Improvements are obtained at temperatures above 450°C but the effect is optimised in the range 600 to 650°C.

- 5) A prolonged stress relieving treatment at temperatures above 600°C can sometimes raise the Charpy V notch transition temperature by as much as 40°C and reduce the ultimate tensile strength by as much as 5 kgf/mm². This deterioration can be reduced by decreasing either the holding time or temperature of stress relief. Prolonged stress relief of microalloyed steels also produces a deterioration in tensile properties.

- 6) Considering the lack of information in some specific cases, where abnormal values of impact properties are recorded, and the necessity of correlating the behaviour of full scale joints with the impact values obtained on small specimens, a test programme was decided in Sub-Commission IX-G and is actually in operation.

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CHAPTER 8

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CHAPTER 9

CORROSION

INTRODUCTION

In connection with welding, various corrosion phenomena appear (1) to (7), (11).

1. The weld metal and the parent metal have different structures and the chemical composition is often different. This can lead to galvanic attack. If the weld metal is anodic to the base metal, rapid corrosion of the weld can result. Different steel qualities such as killed or rimmed steels can have different electrode potentials and thus corrosion cells can be formed.
2. Oxide scale on the surface of the plate can contribute to very heavy corrosive attack. As plate is now often sandblasted, the risks are, however, relatively small for this type of corrosion.
3. The weld bead often has a rough surface and moisture can condense in crevices and beneath loose scale.
4. Slag residues on the surface of the weld and in its vicinity can lead to breakdown of protective coatings on the weld. This is particularly the case when low hydrogen basic electrodes are used. The alkaline nature of the slag particles can result in blowholes in the paint due to a reaction between the slag and components of the paint.
5. In the weld metal itself, different structures can have different potentials and can act as electrolytic cells on a microscale. The same effect can happen in the HAZ.
6. Finally, stress corrosion must be mentioned. Stress corrosion cracking can take place in some solutions if the metal is stressed in tension. As an example, intergranular cracking

can occur in mild steel if it is stressed in and exposed to nitrate solutions. Stress corrosion cracking in contact with NH_3 and H_2S solutions can also occur.

In this chapter, corrosion problems which are met in the welding of C-Mn and C-Mn microalloyed high tensile steels will be described. The weather proof low alloyed steels are outside the scope of this survey and will not be considered here.

CORROSIVE ATTACK ON PAINTED STRUCTURES

As mentioned above, there are sometimes reactions between slag residues on and near the weld and the paint. Since basic low hydrogen electrodes are often used for welding these steels, it is important that the slag residues are thoroughly removed either by sandblasting or by pickling with phosphoric acid before painting. Corrosive attack of this type has mainly been found under the water line on welded ships. In the atmosphere or in fresh water this phenomenon is of minor importance. Careful slag removal in combination with the use of a wash primer is nevertheless recommended as soon as possible after welding to promote optimum corrosion resistance after painting.

CORROSION OF WELD METAL

When a protective coating of paint is absent due to service conditions (for instance in ice-breakers), severe corrosive attack by sea-water, particularly on welds, has been observed. This is due to the difference in potential between the weld metal and the parent metal. The phenomenon has been studied by potential difference measurements. The corrosion can be avoided by choosing a weld metal of the correct composition. Basic low hydrogen electrodes containing small quantities of copper and nickel have been used successfully.

CORROSION IN THE HAZ

A phenomenon which can be of some importance when welding high tensile steels is the increasing sensitivity to corrosion observed in the HAZ of some steels when tested in flowing sea-

water. Previous investigations with normal mild steel ship plates showed that the HAZ was no more sensitive to corrosion than the parent metal. However, recent investigations have shown that if there is a hardened structure in the HAZ, then the behaviour during corrosive attack can be more marked (8 to 10, 12).

When testing C-Mn steels with varying C- and Mn-contents in flowing sea-water, it was found that if untempered high carbon martensite or retained austenite was present in the HAZ, this zone was attacked more than the parent metal (8). The analysis of the steel tested was

C 0.11%, Si 0.35%, Mn 1.62%

The theory is that the untempered martensite acts as local cathodes to the ferrite matrix. If the martensite was tempered the attack was negligible.

In another study (9, 10) the theory is that massive ferrite is formed and corrodes selectively. It is shown that corrosion can be avoided by the use of preheating at somewhat high temperatures. Stress relieving can be beneficial but the result is very much dependant on the choice of a correct temperature.

The results mentioned refer to testing in flowing sea-water and without a painted coating. This means that very severe conditions are seldom met by normal welded structures. The same tests made in flowing fresh-water and stationary sea-water showed no attacks in the HAZ. Since most C-Mn microalloyed steels have a low C-equivalent and relatively low C- and Mn-contents, the risk of attack in the HAZ must be judged as negligible.

Under normal circumstances the corrosion effects when using welded C-Mn microalloyed steels are relatively small and are about the same as with ordinary C-Mn steels. When optimum corrosion resistance is required, care must be taken to remove any welding slag before painting. In special cases, such as unpainted steel in flowing sea-water, a hardened HAZ can sometimes show corrosive attack. In this case the use of steels with low C-equivalent is recommended. In flowing sea-water, when the protective coating can be damaged, as in ice-breakers,

it is also necessary to choose a weld metal which is cathodic in comparison with the base metal. Welded structures which are used where there is a risk of stress corrosion must be considered individually, since the higher yield point of the HT steels can result in higher residual stresses.

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SUMMARY AND CONCLUSIONS

In an earlier section (Chapter 2), we defined C-Mn and C-Mn microalloyed steels according to their chemical composition and the role played by the additional elements with regard to the mechanical properties (tensile and impact characteristics) in the ferritic/pearlitic conditions under which these steels are delivered. This information is collated from the results gathered during an international investigation into steels produced in eleven countries (Chapter 3). The results of this investigation should provide the basis for a classification system. To describe the weldability of C-Mn and C-Mn microalloyed steels it is very helpful to refer to the continuous cooling transformation diagrams since these cover the thermal conditions of actual welding, or simulate these conditions (Chapter 4). The risk of cold cracking during welding can be evaluated by means of quantitative tests, and interpreted in the light of the data provided by the transformation diagrams.

The problem of cold cracking is discussed (Chapter 5) in connection with the formation of martensite (hence the criteria using the carbon equivalent and underbead hardness), the hydrogen content and the restraint level. The corresponding risk is assessed by means of special tests, which should be adapted to the conditions of actual joints. In order to avoid cold cracking, one may act on the cooling time through the preheat, energy input and temperature, as well as on the post-heat temperature.

The risk of hot cracking (Chapter 6) in the molten area or in the heat-affected zone must be evaluated with the aid of specific tests fully to take account of the combined effect of the various elements likely to promote the phenomenon.

Cracking due to lamellar tearing (Chapter 7) is an important risk in welded constructions, but this risk is not peculiar to C-Mn and C-Mn microalloyed steels. The preventive precautions

are the suitable design of joints, the appropriate welding procedure, choice of auxiliary materials and base metals, together with the use of non-destructive and destructive tests.

In the welds not heat treated after deposition, the mechanical properties of the heat-affected zone do not undergo any important change (Chapter 8). However, a tendency for the Charpy V impact results to be lower is observed when high welding energies are employed. It is important to verify whether or not this statement, based on small experiments, is in keeping with a change of behaviour in the joints in full-scale tests. The embrittlement in the area heated to a temperature below A_1 , does not seem pronounced for this category of steel.

In the case of welds which have undergone a stress-relieving heat treatment, there is usually an improvement in the tensile and impact strengths in the heat-affected zone which then returns to the high values of the parent metal. High energy welds sometimes show an exception to this behaviour.

Finally we broach the question of corrosion (Chapter 9). It seems that C-Mn and C-Mn microalloyed steels do not pose any particular problem when welded except for a tendency to corrode in salt water more rapidly in the heat-affected zone. For this reason only the lower strength grades are recommended.

This document, prepared by Sub-Commission G and approved by Commission IX, should, from the theoretical and technical data which it contains regarding the weldability of C-Mn and C-Mn microalloyed steels, serve as a basis for the preparation of more detailed recommendations on national or international levels. These recommendations will concern in particular on the one hand the classification of the steels in question, and on the other hand the most appropriate welding procedure.

Appendix I Steels Class 36 kgf/mm²

Classification No.	Country and Trade mark	Chemical composition %					
		C max.	Si	Mn	P max.	S max.	Other elements
AUSTRIA							
3613	Alfort	0.22	0.45	1.40	0.040	0.040	Al $\leq$ 0.02
3614	Altherm 55	0.23	0.50	1.40	0.040	0.040	Al $\leq$ 0.02
3615	Aldur 50	0.21	0.40	1.3	0.040	0.040	Al $\leq$ 0.02
3616	Aldur 55	0.22	0.45	1.5	0.040	0.040	Al $\leq$ 0.02
BELGIUM							
3602	Feralsim 52	0.20	$\leq$ 0.50	1.20-1.40	0.03	0.04	N $\leq$ 0.010
3603	St 52-C	0.20	0.30-0.50	1.00-1.25	0.05	0.05	
3604	A 52-D	0.22	$\leq$ 0.55	$\leq$ 1.50	0.05	0.05	
CZECHOSLOVAKIA							
3675	CSN 11483	0.20	$\leq$ 0.55	$\leq$ 1.20	0.045	0.045	Ti $\leq$ 0.20
3676	CSN 11523	0.20	$\leq$ 0.55	$\leq$ 1.50	0.050	0.045	Al $\leq$ 0.02
3689	CSN 11524	0.20	$\leq$ 0.55	$\leq$ 1.50	0.050	0.045	
3690	CSN 11484	0.20	$\leq$ 0.55	$\leq$ 1.40	0.040	0.040	
FRANCE							
3619	MS 52	0.22	$\leq$ 0.55	$\leq$ 1.0	0.035	0.035	Nb $\leq$ 0.050
3673	Creuselso 38	0.18	0.4	1.2	0.035	0.030	
3677	Usiten LE 36	0.22	$\leq$ 0.50	$\leq$ 1.5	0.040	0.040	
3678	Usiten LE 36 C	0.22	$\leq$ 0.50	$\leq$ 1.5	0.035	0.035	Nb $\leq$ 0.050, V $\leq$ 0.150
3679	Usiten LE 36 Nb	0.20	$\leq$ 0.50	$\leq$ 1.5	0.035	0.035	
3680	Usiten LE 36 NbC	0.20	$\leq$ 0.50	$\leq$ 1.5	0.035	0.035	
3681	Usiten LE 36 NbF	0.20	$\leq$ 0.50	$\leq$ 1.5	0.030	0.030	Nb $\leq$ 0.050, V $\leq$ 0.150
3682	Usiten LE 36 NbVC	0.18	$\leq$ 0.50	$\leq$ 1.5	0.035	0.035	
3683	Usiten LE 36 NbVF	0.18	$\leq$ 0.50	$\leq$ 1.5	0.030	0.030	
3684	Usiten LE 38 Nb	0.22	$\leq$ 0.40	$\leq$ 1.5	0.035	0.035	Nb $\leq$ 0.050, V $\leq$ 0.150
3685	Usiten LE 38 NbC	0.22	$\leq$ 0.40	$\leq$ 1.5	0.035	0.035	
3686	Usiten LE 38 NbF	0.22	$\leq$ 0.40	$\leq$ 1.5	0.030	0.030	
3687	Usiten LE 38 NbVC	0.20	$\leq$ 0.40	$\leq$ 1.5	0.035	0.035	Nb $\leq$ 0.050, V $\leq$ 0.150
3688	Usiten LE 38 NbVF	0.20	$\leq$ 0.40	$\leq$ 1.5	0.030	0.030	
3699	Usiten LE 36 NbV	0.16	$\leq$ 0.50	$\leq$ 1.5	0.035	0.035	
36100	Usiten LE 38 NbV	0.18	$\leq$ 0.40	$\leq$ 1.5	0.035	0.035	Nb $\leq$ 0.050, V $\leq$ 0.150

Appendix I Steels Class 36 kgf/mm²

Heat treatment	Mechanical properties (12.5 mm)			Maximum thickness manu- factured mm	Yield point at maximum thickness kgf/mm ²	t _{KV} (2.8 kgfm) °C	KV kgfm at °C
	Minimum yield point kgf/mm ²	Ultimate tensile strength kgf/mm ²	Elongation % on 5d				
AUSTRIA							
B	38	52-64	22	50	35	-40	
B	37	55-65	17/TS	100	34	-40	
B	36	50-60	25	100	33	-60	
B	39	55-65	24	100	36	-60	
BELGIUM							
B	36	≅52	≅24	60	35	(-40) -20	4 -20
B	37	≅52	≅22	30	≅36		
B	36	≅52	≅22	30	≅35		
CZECHOSLOVAKIA							
B	38	48-62	22	50	36	-20	
B	36	52-64	20	50	34	± 0	
B	36	52-64	20	50	34	-20	
B	38	48-62	22	60	34	-30	
FRANCE							
C	37	≅52	≅23	80	35	-20	4 -20 4 -20 4 -20 4 -20 4 -40 4 -20 4 -40 4 -40 4 -20 4 -40 4 -20 4 -40 4 -20 4 -40 4 -20 4 -20
	38	55-65		60	≅36		
B	36	52-62	≅22	80	35		
B	36	52-62	≅22	80	35		
B	36	52-62	≅22	50	35		
B	36	52-62	≅22	50	35		
B	36	52-62	≅22	50	35	-50	
B	36	52-62	≅22	50	35		
B	36	52-62	≅22	50	35	-50	
B	36	52-62	≅22	50	35		
B	38	54-64	≅22	50	36		
B	38	54-64	≅22	50	36		
B	38	54-64	≅22	50	36	-50	
B	38	54-64	≅22	50	36		
B	38	54-64	≅22	50	36		
B	36	52-62	≅22	50	35		
B	38	54-64	≅22	50	36		

Appendix I Steels Class 36 kgf/mm²

Classification No.	Country and Trade mark	Chemical composition %					
		C max.	Si	Mn	P max.	S max.	Other elements
GERMANY							
3633	FB 50	0.20	0.10-0.50	0.90-1.6	0.040	0.040	Al+
3634	Union 36 G	0.20	0.10-0.50	0.90-1.60	0.040	0.040	Al $\leq$ 0.025
3637	HOAG 50 FK/G	0.20	0.10-0.50	0.90-1.60	0.040	0.040	
3638	HSB 50	0.20	0.10-0.50	0.90-1.60	0.040	0.040	Al+
3639	KHB 36 G	0.20	0.10-0.50	0.90-1.60	0.040	0.040	Al $\leq$ 0.20
3640	BH 39	0.20	$\leq$ 0.40	1.10-1.60	0.035	0.035	V 0.10-0.17
3641	BH 36	0.20	0.10-0.50	0.90-1.60	0.040	0.040	Al+
3691	FB 55	0.20	$\leq$ 0.50	1.10-1.60	0.035	0.035	Al+, N 0.015, V 0.10-0.16
3692	Union 39 G	0.20	0.55	1.40	0.035	0.035	Al $\leq$ 0.025, (Ti alt V 0.10-0.20)
3693	Dillinal 50/36	0.20	0.10-0.50	0.90-1.60	0.040	0.040	Al+
3695	SB 36 F	0.20	0.10-0.50	0.90-1.60	0.040	0.040	Al+
3696	SB 39 F	0.20	0.30-0.55	1.10-1.50	0.035	0.035	Al+, Ti alt V $\leq$ 0.10
3698	KHB 39 G	0.20	0.30-0.55	0.90-1.60	0.035	0.035	Al+, Nb $\leq$ 0.020
ITALY							
36107	SELCO 53	0.20	0.35	1.50	0.030	0.030	Al+
36108	SELCO 53 V	0.20	0.35	1.50	0.030	0.030	V $\leq$ 0.18
36109	ASERA 54	0.22	0.15-0.40	$\leq$ 1.50	0.025	0.035	V $\leq$ 0.20
36110	ASA 52	0.20	$\leq$ 0.40	$\leq$ 1.50	0.030	0.035	(Nb $\leq$ 0.06, N $\leq$ 0.015 C eq $\leq$ 0.45
36111	ASA F 52	0.20	$\leq$ 0.40	$\leq$ 1.50	0.030	0.035	(Nb $\leq$ 0.10, V+(ev), N $\leq$ 0.015 C eq $\leq$ 0.45
36112	Polar 52	0.20	$\leq$ 0.40	$\leq$ 1.50	0.030	0.035	(Nb $\leq$ 0.10, V+(ev), N $\leq$ 0.015 C eq $\leq$ 0.45
JAPAN							
3660	FTW 55	0.18	$\leq$ 0.55	$\leq$ 1.50	0.030	0.030	If necessary
3661	FNB 36 A	0.23		$\leq$ 1.40	0.035	0.035	Nb $\leq$ 0.10
3662	KH 36	0.20	$\leq$ 0.15	$\leq$ 1.40	0.035	0.035	Nb $\leq$ 0.10
3663	HTP 55W	0.18	$\leq$ 0.55	$\leq$ 1.60	0.035	0.035	
3664	HK-Hiten HS-1	0.17	0.43	1.10	(0.023)	(0.019)	
3665	Wel-Ten 50 A/B	0.18	0.25-0.45	1.30-1.50	0.035	0.040	
3668	St-2	0.15-0.20	0.30-0.60	1.20-1.50	0.035	0.035	
3669	HS-A	0.18	0.35-0.60	1.10-1.40	0.045	0.045	
3670	HS-B	0.26	$\leq$ 0.60	1.10-1.40	0.045	0.045	

Appendix I Steels Class 36 kgf/mm²

Heat treatment	Mechanical properties (12.5 mm)			Maximum thickness manu- factured mm	Yield point at maximum thickness kgf/mm ²	t _{KV} (2.8 kgfm) °C	KV kgfm at °C
	Minimum yield point kgf/mm ²	Ultimate tensile strength kgf/mm ²	Elongation % on 5d				
GERMANY							
B	36	≤64	22	150	30		5 -20
B	36	≤64	22	150	30		5 -20
B	36	≤64	22	150	30		5 -20
B	36	≤64	22	150	30		5 -20
B	36	≤64	22	150	30		5 -20
B	39	≤66	20	150	31		5 -20
B	36	≤64	22	150	30		5 -20
B	39	≤66	20	150	31		5 -20
B	39	≤66	20	150	31		5 -20
B	36	≤64	22	150	30		5 -20
B	36	≤64	22	150	30		5 -20
B	39	≤66	20	150	31		5 -20
B	39	≤66	20	150	31		5 -20
ITALY							
B	37	53-62	24	70	34	-50	
B	39	53-62	23	70	37	-45	
B	38	54-64	≥21	100	36	-20	
B	37	52-62	≥22	100	34	-20	
B	37	52-62	≥22	100	34	-45	4.8 -20
B	36	52-62	≥22	50	36	-60	5.6 -20
JAPAN							
A	37	55-63	≥29(2")	50	36		4.8 ± 0
A	36	≤50	≥20(200)				
A	36	≤50	≥19(2")			B ± 0 ± 0	
A, B	36	55-65	≥20(200)				4.8 ± 0
A	38	54		40	36	-40	
A	38	54	≥27(200)	50	37		2.1 -47
A	36	55	≥18(85 √A)			-40	
A	36	55-65	≥18(200)				
A	38	60-70	≥17(200)				

Appendix I Steels Class 36 kgf/mm²

Classification No.	Country and Trade mark	Chemical composition %					
		C max.	Si	Mn	P max.	S max.	Other elements
NETHERLANDS							
3644	Y 36 - 50 N	0.18	≤0.50	≤1.50	0.035	0.035	Nb≤0.020
3645	Y 36 - 50 K	0.18	≤0.50	≤1.50	0.035	0.035	Nb≤0.020
SWEDEN							
3671	{ OX 520 OX 522 OX 525 }	0.18	0.15-0.50	≤1.4	0.04	0.04	{ Nb 0.01-0.04 C eq ≤0.41 }
36106	{ YP 36 D YP 36 E }	0.18	0.25	1.3	0.040	0.040	Nb+, Al+
36313	{ Domex 360 C Domex 360 D Domex 360 E Domex 360 F }	0.20	≤0.60	≤1.60	0.030	0.030	Nb+, Al+
UNITED KINGDOM							
3612	Hyplus 23	0.20	0.50	≤1.50	0.05	0.05	Nb≤0.10
36101	BS 4360 50 B	0.20	≤0.50	≤1.50	0.05	0.05	Nb≤0.10
36102	BS 4360 50 C	0.20	≤0.50	≤1.50	0.05	0.05	Nb≤0.10
36103	BS 4360 50 D	0.18	0.10-0.50	≤1.50	0.04	0.04	Nb≤0.10
36104	Arctic B	0.22	0.10-0.40	≤1.60	0.05	0.05	
36105	{ Hyplus 23 (controlled rolling) }	0.18	0.50	≤1.40	0.05	0.05	Nb≤0.10
UNITED STATES							
Data for steels manufactured in the United States will be							
C eq calculated according to the formula:							
$C eq = C + \frac{Mn}{6} + \frac{Cr + Mo + V}{5} + \frac{Ni + Cu}{15}$							

Appendix I Steels Class 36 kgf/mm²

Heat treatment	Mechanical properties (12.5 mm)			Maximum thickness manu- factured mm	Yield point at maximum thickness kgf/mm ²	t _{KV} (2.8 kgfm) °C	KV kgfm at °C
	Minimum yield point kgf/mm ²	Ultimate tensile strength kgf/mm ²	Elongation % on 5d				
NETHERLANDS							
B	36	50-62	≧23	100	34	-40	4.8 -20
B	36	50-62	≧23	100	34	-50	5.6 -20
SWEDEN							
B	37	52-62	≧22			-40	
B	36	≧52	≧20			{ D -20 E -40 C ± 0	
B	36	≧52	≧24	50	36	{ D -20 E -40 F -60	
UNITED KINGDOM							
B	36.0	50-63	≧20	100	≧33.0	-20	
B	35.5	50-62	≧20.0	100	≧32.5		
B	35.5	50-62	≧20.0	100	≧32.5	-15	
B	35.5	50-62	≧20.0	100		-30	
B	36.2	51-59	≧21(8")	150	≧29.9	-30	
B 1	36.0	50-61	≧20	100	≧33.0	-15	
UNITED STATES							
found in tables published in ref (5) of Chapter 3.							

Appendix 2 Steels Class 40 kgf/mm²

Classification No.	Country and Trade mark	Chemical composition %					
		C max.	Si	Mn	P max.	S max.	Other elements
AUSTRIA							
4006	Aldur 58	0.23	0.50	1.6	0.040	0.040	Al $\geq$ 0.020
BELGIUM							
4003	Feralsim 58	0.25	$\leq$ 0.50	1.30-1.50	0.03	0.04	
CZECHOSLOVAKIA							
4040	CSN 13220	0.15-0.21	0.25-0.50	1.3-1.8	0.040	0.040	V 0.10-0.20, N $\leq$ 0.020
FRANCE							
4007	MS 56	0.24	$\leq$ 0.4	$\leq$ 1.3	0.035	0.035	
4036	Creuselso 42	0.22	0.4	1.4	0.035	0.035	
4039	Usiten LE 40 V	0.22	$\leq$ 0.50	$\leq$ 1.5	0.030	0.030	V $\leq$ 0.15
GERMANY							
4043	SB 43 F	0.20	0.30-0.55	1.20-1.60	0.035	0.035	Al+, V $\leq$ 0.10
4044	KHB 43 G	0.20	0.30-0.55	1.20-1.60	0.035	0.035	Al+, N $\leq$ 0.020
ITALY							
4049	ASA 56	0.21	$\leq$ 0.50	$\leq$ 1.60	0.030	0.035	Nb $\leq$ 0.08, V+(ev), N $\leq$ 0.015
4051	ASA F 56	0.20	$\leq$ 0.40	$\leq$ 1.60	0.030	0.035	V $\leq$ 0.06, Nb $\leq$ 0.06, N $\leq$ 0.015
JAPAN							
4024	FNB 40 A/B	0.26		$\leq$ 1.40	0.035	0.035	Nb $\leq$ 0.10
4025	KH 40	0.23	$\leq$ 0.15	$\leq$ 1.40	0.035	0.040	Nb $\leq$ 0.10
4026	NK-HITEN HS -2	0.13	0.45	1.40	(0.023)	(0.020)	
4027	Wel-Ten 55	0.18	0.35-0.55	1.20-1.50	0.035	0.040	Fine grain steel
4029	YES 40	0.28	$\leq$ 0.15	$\leq$ 1.40	0.035	0.040	Nb+, V $\leq$ 0.15
4031	Her-Ten	0.20	0.30-0.50	0.70-1.20	0.035	0.035	Al 0.015-0.025
4032	St-1	0.13-0.18	0.30-0.60	1.35-1.65	0.035	0.035	
NETHERLANDS							
4019	{ Y 42-56 N } { Y 42-56 K }	0.20	0.50	1.60	0.035	0.035	Nb 0.020

Appendix 2 Steels Class 40 kgf/mm²

Heat treatment	Mechanical properties (12.5 mm)			Maximum thickness manu- factured mm	Yield point at maximum thickness kgf/mm ²	t _{KV} (2.8 kgfm) °C	KV kgfm at °C
	Minimum yield point kgf/mm ²	Ultimate tensile strength kgf/mm ²	Elongation % on 5d				
AUSTRIA							
B	41	58-68	22	100	38	-60	
BELGIUM							
B	40	≧58	≧22				4 -20
CZECHOSLOVAKIA							
B, C	44	60-75	20	80	41	-20	
FRANCE							
C	40	≧56	≧22	80	≧37	-20	6 -10
	42	56-66	≧23				
B	40	58-66	≧22	50	≧38		
GERMANY							
B	43	≧69	19	150	35		5 -20
B	43	≧69	19	150	35		5 -20
ITALY							
B	40	56-66	≧19	100	≧39	-20	
B	40	56-66	≧21	100	≧39		
JAPAN							
A	40	≧55	≧15(200)			± 0	2.1 -30
A	40	≧55	≧17(2")			± 0	
A, B	43	59	36(2")	20	41	-28	
A	40	58	23(2")	30	37		
A	44	57	26(2")	32	40	(-30)	
A	40	51-62	20			-30	
A	41-45	57-60	32(4√A)	25	≧36	(-40)	
NETHERLANDS							
B	42	56-68	22	100	39	(N-40 K-50)	4.8 -20 5.6 -20

Appendix 2 Steels Class 40 kgf/mm²

Classification No.	Country and Trade mark	Chemical composition %					
		C max.	Si	Mn	P max.	S max.	Other elements
SWEDEN							
4018	CD-40	0.18	0.4	1.2	0.03	0.03	Al+, V+
4033	{ Domex 400 C Domex 400 D Domex 400 E }	0.20	≤0.55	≤1.60	0.040	0.040	Al+, Nb 0.02-0.04
4038	{ OX 540 OX 542 } { C D E }	0.20	0.15-0.50	≤1.6	0.04	0.04	{ Al+, Nb 0.03 C eq ≤0.45 }
4046	{ DET. 40 DET 40 K } { C D E F }	0.20	≤0.5	≤1.6	0.04	0.04	Al+, Nb+
4047	{ YP 40 D YP 40 E }	0.20	≤0.25	1.4	0.040	0.040	Al+, Nb+
4054	{ Domex 400 T C Domex 400 T D Domex 400 T E }	0.20	≤0.55	≤1.60	0.030	0.030	Al+, Nb 0.02-0.04
UNITED KINGDOM							
4045	Thirty Oak 11	0.22	0.10-0.60	≤1.5	0.04	0.04	Nb≤0.08
UNITED STATES							
Data for steels manufactured in the United States will be							
C eq calculated according to the formula:							
$C\ eq = C + \frac{Mn}{6} + \frac{Cr + Mo + V}{5} + \frac{Ni + Cu}{15}$							

Appendix 2 Steels Class 40 kgf/mm²

Heat treatment	Mechanical properties (12.5 mm)			Maximum thickness manufactured mm	Yield point at maximum thickness kgf/mm ²	t _{KV} (2.8 kgfm) °C	KV kgfm at °C
	Minimum yield point kgf/mm ²	Ultimate tensile strength kgf/mm ²	Elongation % on 5d				
SWEDEN							
B	40	52	≧25	50	40	-20	{ C ± 0 D -20 E -40
B	40	55	≧23	50	40	-20	
B	40	54	≧22	40	≧40	-20	
B	40	≧55	≧22	50	≧40	-20	{ C ± 0 D -20 E -40 F -60
B	40	55	20			-20	
B	40	55-65	≧23	50	40	-20	
UNITED KINGDOM							
B	42.5	58.0		50		-60	
UNITED STATES							
found in tables published in ref (5) of Chapter 3.							

Appendix 3 Steels Class 45 kgf/mm²

Classification No.	Country and Trade mark	Chemical composition %					
		C max.	Si	Mn	P max.	S max.	Other elements
GERMANY							
4526	KHB 47 B	0.20	0.30-0.55	1.25-1.65	0.035	0.035	Al+, N 0.023, V 0.08
ITALY							
4530	CD 62 V	0.20	≤0.40	≤1.50	0.030	0.030	N ≤0.015, V ≤0.20
JAPAN							
4502	Her-Ten	0.22	0.40-0.70	≤1.3	0.035	0.035	Al 0.015-0.025
4518	HTP 60-W	0.18	0.75	≤1.60	0.030	0.030	Nb≤0.10, V ≤0.10
4519	Hicon 36	0.20	≤0.60	≤1.50	0.035	0.040	{ Nb≤0.10 C eq ≤0.48
4520	Hicon 40	0.20	≤0.60	≤1.50	0.035	0.040	{ Nb≤0.10 C eq ≤0.48
4521	Hibil A Hibil B	0.20	≤0.20	≤1.50	0.035	0.040	Nb≤0.10
4522	YES 36 A YES 36 B	0.23	≤0.15	≤1.40	0.035	0.040	Nb+, V ≤0.15
SWEDEN							
4531	{ Domex 450 C } { Domex 450 D }	0.20	≤0.55	≤1.60	0.040	0.040	Al+, V ≤0.10
UNITED KINGDOM							
4527	BS 4360 55 C	0.22	≤0.60	≤1.60	0.04	0.04	Nb≤0.10
4528	BS 4360 55 E	0.22	≤0.60	≤1.60	0.04	0.04	Nb≤0.10
4529	Hyplus 29	0.22	0.15-0.50	≤1.60	0.05	0.05	V ≤0.20
UNITED STATES							
Data for steels manufactured in the United States will be							
C eq calculated according to the formula:							
$C_{eq} = C + \frac{Mn}{6} + \frac{Cr + Mo + V}{5} + \frac{Ni + Cu}{15}$							

Appendix 3 Steels Class 45 kgf/mm²

Heat treatment	Mechanical properties (12.5 mm)			Maximum thickness manu- factured mm	Yield point at maximum thickness kgf/mm ²	t _{KV} (2.8 kgfm) °C	KV kgfm at °C
	Minimum yield point kgf/mm ²	Ultimate tensile strength kgf/mm ²	Elongation % on 5d				
GERMANY							
B	47	74	17	150	39		4.5 -20
ITALY							
B	≈45	62-75	19	70	≈42	-50	
JAPAN							
A	47	60-70	17			-22	
B	45	60-72	≈26(2")	50	≈45		4.8 ± 0
A	46	59	39(2")	40	≈37		4.8 +15
A	47	59	40(2")	40	≈47		4.8 +10
A	47	58	26(2")	42	≈37	B ± 0	
A	45	55	25(2")	32	≈39	B ± 0	
SWEDEN							
B	≈45	60	21	50	≈36	{ C ± 0 D -20	
UNITED KINGDOM							
A	45.0	55-70	19.0	63		± 0	
B	45.0	55-70	19.0	63	40.0	-50	
B	≈45.0	58-71	≈20	64	≈41.0		3.5 -30
UNITED STATES							
found in tables published in ref (5) of Chapter 3.							