ENHANCEMENT OF THE QUALITY OF THE REACTOR PRESSURE VESSEL USED IN LIGHT WATER POWER PLANTS BY ADVANCED MATERIAL FABRICATION AND TESTING TECHNOLOGIES

K. KUSSMAUL, J. EWALD, *et al.*

*Staatliche Materialprüfungsanstalt (MPA), Universität Stuttgart,
Pfaffenwaldring 32, D-7000 Stuttgart 80 (Vaihingen), Germany*

SUMMARY

The fracture safe assessment of nuclear pressure vessels (RPV) is based upon: (i) an adequate stress analysis, (ii) reliable material characteristics, (iii) acceptable defect sizes.

There may arise problems which are related to the inhomogeneity, low toughness and crack phenomena sometimes observed in the base material and heat affected zone (HAZ). Due to this it is difficult and in some respects even impossible to measure the decisive values of (fracture-) toughness and defects. Apart from the significance of those facts for existing RPVs, all efforts were directed to provide a steel which should be non-susceptible to embrittlement and/or cracking in the HAZ, and simultaneously yielding in a higher upper shelf toughness of base and HAZ material. These objections were pursued in cooperation with manufacturers, vendors and inspection authorities by the following activities:

- (i) Detailed investigations to obtain information on: occurrence and size of inhomogeneities and defects, especially stress relief cracking (SRC), toughness properties adjacent to defects;
- (ii) improvement of: chemical composition, steel making processes, welding procedures, optimum temperature cycle and level for stress relief heat-treatment.

In order to solve these tasks it was necessary to develop additional tools and to correlate all partial results which were newly elaborated:

- common destructive and non-destructive testing methods,
- metallographic tangential cut techniques,
- small HAZ-tensile test procedures,
- HAZ-simulation techniques representing the thermal-mechanical history.

The results were as follows:

- In the base material multiple cracking was detected in the micro and mm-range masking large size cracks. These results did not correspond with the ultrasonic examination, the indication of which was said to be the response of slag inclusions.
- In the HAZ of the 22 NiMoCr 37 steel (similar to A 508 Cl 2) it was found that stress relief cracking (SRC) in the micro and mm-range cannot be detected by ultrasonic testing. About 50% of 100 investigated component strength weldments were affected by SRC.
- The HAZ of 22 NiMoCr 37 steel with optimized chemical composition as well as the 20 MnMo-Ni 55 steel (similar to A 508 Cl 3, resp. A 533 B Cl 1) proved to be not susceptible to SRC and stress relief embrittlement (strain induced embrittlement, in the relevant range).
- It is possible to limit size and number of SRC by reducing non-transformed coarse grain structures in the HAZ through specific welding techniques.
- Proposals could be given for optimum (higher) stress relief temperatures to generate higher toughness properties in the HAZ.

The authors stress the need for proceeding in that successful direction taking other steel making processes into consideration, such as MHKW (electro slag remelting in a forged ring acting as a mould). A recently introduced technology, according to which components are built up only of weld metal, will also be of interest. Another advantage may possibly be expected if for the base material a three step heat-treatment is applied.

For the new generation of components the fracture safe analysis can now be based completely upon well defined and homogeneous and high upper shelf base materials including the HAZ. There has been proposed for the assessment of the safety margin of formerly produced reactor pressure vessels a research program which shall cover also long time in service influences.

1. Introduction

In the endeavour to keep the failure probability of the reactor pressure vessel (RPV) not only low but to create the necessary prerequisites for a technical absolutely doubtless prevention against catastrophic failure it was attempted since the seventies to give the necessary impulses for a further improvement of the quality of the RPV /1-7/.

It became soon apparent that in opposition to the then opinion in the overheated areas of the heat affected zone (HAZ) also for the RPV material 22 NiMoCr 3 7 the relaxation embrittlement and relaxation crack formation to be postulated under stress relief heat-treatment (SRHT) is in the end the root problem. If this problem is solved the whole complex of questions can satisfactorily be answered such as segregation, segregation cracking, the necessary degree of purity (properties in the through thickness direction, ultrasonic testing capability) inclusive long term operation (irradiation embrittlement, corrosion).

The measures and studies initiated by the BMFT (Federal Ministry for Research and Technology), BMI (Federal Ministry of the Interior), and the RSK (Reactor Safety Commission) were discussed and carried out in collaboration with the steel makers, the nuclear system vendors, and the technical inspection authorities (TÜV) /8 - 10/. Extracts of the results are given in the following.

2. Scope of the Investigation

This report is concerned with steels on the basis of 22 NiMoCr 3 7 (similar to A 508 Cl 2) and 20 MnMoNi 5 5 (similar to A 533 B Cl 1 or A 508 Cl 3), Table 1, and their weld joints. In order to solve the problems first of all a general statistical survey over the development of the status of fabrication had to be worked out. A great number of semi-finished products and welded joints coming from the production were investigated for this purpose by means of advanced techniques.

In order to evaluate correctly the observations made on the welded joints, it was also necessary to measure the occurring superimposed temperature cycles of the HAZ due to the subsequent weld passes and finally to simulate the thermo-mechanical processes during welding and stress relief heat treatment. The result of this demand was, among others, the development of a simulation technique allowing to reproduce correctly the critical micro structures being formed in the relatively small volume of the overheated area of the HAZ so that those zones are capable of being tested mechanically and technologically with an adequate specimen size /11/.

3. Defect Appearance and Toughness Properties

3.1 Metallographical Investigations in the HAZ by means of

Tangential Cut Technique

Relaxation cracks (reheat cracking, stress relief cracking (SRC)) of large dimensions usually appear in multi-layer weldments longitudinal to the seam as long as

there exist sufficiently large stress raisers such as reinforcement or partially welded seams, slag lines, lack of penetration, hot and cold cracking. They pass in many cases the fine grained bridges which are located between the - busk-like - arranged coarse grained areas, whereby the "coarse grained zones" quasi act as a "crack magnet" so that the cracks will run along the HAZ down to considerable depths through the alternating (coarse/fine) micro structure. For a long time such cracks were wrongly related to the (hydrogen affected) cold cracking. Therefore, no consequences were drawn in view of the material and its damage mechanisms: the cracks which easily could be detected by non destructive examination (NDE) were repaired, by which the flaw was formally removed. The possibility of a corresponding crack formation in the strength weldments became again acute after underclad cracking was observed. It had to be assumed that the crack formation in the micro and mm-range could not be detected by means of ultrasonic testing. Since the development of SRC due to the residual stress conditions was supposed to start in form of crack fields, but primarily transverse to the weld direction, and since it had to be assumed that the size of individual cracks was normally confined to mm-size it was obvious to introduce the so-called tangential cut technique /12, 13/ in order to detect the crack formation, and what was even more important, the prestages in connection with the embrittlement. This metallographic technique consists of multistage sections with cuts of large dimensions tangentially to the flank of the seam, whereby a sufficient volumetric examination of the HAZ is possible. The SRC-susceptible micro structure may comprise in multi-layer weldments up to more than 50 % of the wall thickness, which could be proved by transverse cuts.

So far about 80 weldments in 22 NiMoCr 37 and 40 seams in 20 MnMoNi 55 steel with wall thicknesses extending from 50 to 360 mm have been tested. These are prolongations for procedure and production qualification testing. Special examination concerned trepans and other material removed out of components and test weldings on rolled and forged plates and nozzle (disc)weldings in a full scale forged ring for a 1300 MWel RPV.

About 50 % of the 22 NiMoCr 37 steel samples showed SRC, resp. hot cracking. About 30 % of samples had both crack phenomenons. In contrast, no SRC was observed in the material 20 MnMoNi 55. The most that could be found in about 20 % of the samples was a prestage in form of chains of voids. Hot cracking occurred in about 30 % of the samples.

The SRC in the 22 NiMoCr 37 steel and the prestages of it occurred in the form of successive individual cracks (max. length 5 mm) or patterns of cracks (individual pattern max. 5 x 55 mm) with grain sizes from one to six in the ASTM range. The affected micro structure is enveloped by the weld metal and the fine-grained zone. Materials with pronounced hardness differences between weld metal and HAZ, as well as fairly marked hardness differences in HAZ segregation bandings /13/ were likewise inclined to be more susceptible to SRC and hot cracking.

The increase in the susceptibility to SRC above a total sulfide value of 40 in accordance with the Stahl-Eisen Werkstoffblatt 1570 - 71 indicates that the sulphur exercises a distinct influence and in the form of liquation cracking can be a prestage of SRC. This is also true for cold cracking, if the hydrogen influence is not oppressed by adequate measures as required.

A dominant effect was attributed to the chemical composition of the base material, as compared with other parameters such as the steel making process, processing techniques and the welding parameters, SA one wire, SA double wire, manual welding, or heat input. When the elements C, Si, Mn, P, S, Cu, Sn, Al, N, Cr, Mo, Ni, V, Co, As and Sb were considered there was tendency for cracking with increasing contents of the elements Cu, P, S, As, Al, N, Mo and Co. According to the present knowledge on the SRC mechanism the carbide forming elements Mo and Co act in the matrix (secondary grain hardening) and the other elements act mainly in the grain boundaries (grain boundary segregation).

Above all, it turned out that the susceptibility to SRC and hot cracking was determined more by a cumulative effect of the elements listed in Table 2 than by a selective effect of one element. With few exceptions this results in a clear differentiation between crack-affected and crack-free samples in so far as the number of elements exceeding the threshold values is two or more (threshold value criterion) /13/.

The absence of the elements N, Al, Mo and Co in hot cracking can be regarded as a confirmation of the basic correctness of this observation, since the elements N and Al are to a large extent in solution when hot cracking occurs, and that the effect of Mo and Co on SRC behaviour is assessed to be due to carbide precipitation. An effect from V could not be ascertained because the melts under investigation were mostly free of V ($< 0,01\%$). On the basis of the literature /14 - 16/ and of our own HAZ simulation tests a stimulating effect is, however, to be expected from V-contents of more than $0,02\%$.

An application of the threshold value criterion developed for the material 22 NiMoCr 37 in regard to the chemical composition on material 20 MnMoNi 55 is corresponding to the results of the (SRC-free) tangential cuts. Since the Cr-content obviously has no influence on the crack susceptibility and higher Mn-contents are rather positively to be judged, the different behaviour of both steels is mainly due to the Mo-contents. There is another reason for the significance of Mo because hot cracking (not favoured by Mo) was found in both materials 22 NiMoCr 37 and 20 MnMoNi 55 in a similar size and extent. According to the authors' opinion the above defined threshold value criterion is to be employed for low alloyed NiMo- and NiMoCr-steels which do have no V- or Nb-additions.

3.2 Mechanical Properties of the HAZ

The larger the cross sections of the specimens removed from the weldments, the less it can be succeeded in meeting the critical areas in the HAZ. In so far, it is not surprising that the HAZ does generally yield no lower fracture toughness values than the base or the weld material, Fig. 1. The C_V -test selects already better, Fig. 2, which, however, yields

considerably higher values than the comparatively recorded results on simulated specimens (sect. 4). Small size tensile test specimens, which were specially removed, reveal that the crack susceptible zones are more or less breaking in a brittle manner at higher or lower loading levels according to the different degree of the damage /6/. (In the scope of the present paper it cannot be dealt with special techniques as how to force the crack initiation in the critical HAZ range.)

4. HAZ-Simulation

The HAZ simulating-procedure was divided into (a) overheating simulation, (b) simulation of stress relief heat-treatment and (c) subsequent mechanical testing (post damage testing (PDT)) /11/. As shown in Fig. 3, two standard cycles, in particular, a coarse grain cycle (cycle 13) and a multipass cycle are performed in the course of overheating simulation, apart from a number of other significant single cycles and multi cycles. To simulate the thermal mechanical history (plain annealing, creep/relaxation) of stress relief heat-treatment, the simulation of stress relief heat-treatment is in its turn broken down into three types of tests:

- (a) Common annealing tests (without external stressing) by means of which the temperature induced embrittlement (TIE) can be made visible during the subsequent material testing /17, 18/. A distinction must here be made between embrittlement that occurs during the heating and the holding phase and that which occurs during the slow cooling-off after the stress relief annealing.
- (b) Short time constant load and relaxation testing up to fracture which the temperature and time dependant deformation capacity characteristics of the cracking susceptibility can be determined.
- (c) Short time constant load tests not up to fracture with decreasing stresses at equal loading times by means of which stepwise decreasing creep straining can be produced, the possible effects of which can then in turn be made visible during the subsequent mechanical testing.

It can be assumed that in a component creep strains ranging from 0.4 to 1.0 % may occur. The tests dealt with in the following were carried out up to 2 %. The degradation of toughness by strain exhaustion has been termed strain induced embrittlement (SIE) /17, 18/.

During the constant load and relaxation tests carried out to determine the cracking susceptibility (b) a distinction is made between:

Isothermal tests in which the specimen is rapidly heated up (600°C/h) and stressed when the testing temperature has been attained.

Heating up tests in which the specimen is stressed at ambient temperature and heated up as the component would be heated in reality.

For all tests (HAZ simulated PDT) specimens were as far as possible removed transverse to the main rolling respectively forging direction (axially in forged rings).

4.1 Crack Susceptibility Tests

Characteristic test results of creep tests up to fracture on two melts of materials 22 NiMoCr 37 (plate of 240 mm thickness), and 20 MnMoNi 55 (plate of 103 mm thickness) for coarse-grain overheating (cycle 13) are shown in Fig. 4, and the effect of other cycles on the elongation at fracture at 610°C are represented in Fig. 5.

In order to characterize the fracture behaviour (creep strain capacity) of different melts so called specific time depending elongations at fracture ϵ_{ZDB-t} are deduced from the fracture for $t = 1\text{ h}, 6\text{ h}$ or 20 h . Temperature specific minima of the elongation at fracture in the form of a creep embrittlement occur at all stress relief heat-treatment temperatures, here denoted as ("V"-curves-minima). The "V"-curve minimum $\epsilon_{ZDB-min}$ shifts to lower values and longer times with decreasing holding temperature (650 - 550°C).

After different overheating cycles (Fig. 5, single and double cycles) characteristic differences are again found in the elongations at fracture. Peak temperatures $> 1200^\circ\text{C}$ are responsible in each case for low elongation at fracture values, while double cycles with a peak temperature of 1300°C and additional cycles with 1050°C or 900°C (cycle 15) result in high elongations at fracture. All specimens with low fracture strains ($< 3\%$) underwent inter-crystalline fracture. There was general agreement that initial micro cracking occurred preferentially in the segregation bands in both materials. Fig. 6 shows the elongation at fracture values at 610°C after coarse grain overheating (cycle 13) for all 22 NiMoCr 37 and 20 MnMoNi 55 melts investigated hitherto. This survey illustrates that the results for the individual materials overlap but the 20 MnMoNi 55 steel displays in line with the tangential cut results, a greater elongation at fracture, i.e. low susceptibility to stress relief cracking.

Another interesting result is shown in Fig. 7 in which the elongation at fracture values are plotted against the Mo-content of both materials and of laboratory melts. Where the Mo-content is less than 0.5 %, minimum elongations at fracture $> 1\%$ are obtained in every instance. In so far as the materials show low levels of trace elements and such steel impurities as P, S, Sn, N, Cu (e.g. within the limits set out in Table 2), even about 0.6 % Mo will produce a minimum elongation at fracture of $> 1\%$. In combination with Mo, contents of $V > 0,02\%$ yield again distinctly lower minimum elongation values.

By means of further tests which were just finished, /19/, it could now also be clarified that SRC in melts with low minimum elongation values ("V"-curve minimum at $610^\circ\text{C} < 1\%$) can already occur during a component-like slow heating up during stress relief heat-treatment (30°C/h) while passing through the temperature range of $500\text{--}600^\circ\text{C}$.

4.2 Toughness Behaviour after HAZ-Simulation

4.2.1 Temperature induced Embrittlement (TIE)

As far as 22 NiMoCr 37 and 20 MnMoNi 55 are concerned, there is no incidence in the coarse-grain overheated state of temperature-induced embrittlement, in the sense of precipitation hardening with a resulting shift in transition temperature during the notched-bar impact test, as compared with the simple overheated state. On the contrary, notch toughness performance (with respect to transition range and upper shelf) improves with increasing temperature and annealing time, Fig. 8.

However, precipitation hardening (temperature-induced embrittlement) occurs when steels of the type 20 MnMoNi 55 have alloyed Nb in small quantities e.g. in the manner of 13 MnNiMoNb 54, Fig. 9. In connection with the TIE examination, laboratory melts of 22 NiMoCr 37 were also tested for the incidence of temper embrittlement in the event of a high P, Sn, As or Sb content. For the purpose of the tests a normal melt (MPA ref. 16) containing the usual proportions of the four trace elements, Fig. 10, and two melts containing increased quantities of P and, to a certain extent, Sn, As (and also Cu and S) were available. Where, after coarse-grain overheating, the materials were cooled during subsequent annealing of 4 h/610°C by means of the very slow Gould step - cooling process /20/, all the melts displayed a transition temperature shift of at least 100°C to higher temperatures, as compared with the samples which after annealing at 610°C were cooled quickly (300°C/h). It was only with the introduction of slow cooling commensurate with the annealing of heavy forgings (30°C/h) that shifts in the transition temperature of the a_k - T curves were virtually eliminated in the normal melt (MPA ref. 16, P = 0.015%). However, even after cooling at 30°C/h the melt with the highest P, As, (S and Cu) content (MPA ref. 41) still displayed a transition temperature shift of about 150°C compared with the fast-cooled state. When heat-treated as specified in the code, the sensitive melts (MPA refs. 30 and 41) displayed, after step cooling on the Gould model, a maximum transition temperature shift of a mere 20°C.

It can be seen here that, where there is a high trace element content in the coarse-grain overheated state, the material 22 NiMoCr 37 may have a tendency to reversible temper embrittlement. Through the commensurate limitation of the trace elements P, Sn, As and Sb (based approximately on the threshold values found in the 22 NiMoCr 37 tangential cut programme - Table 2) and as a result of a very low acute-carbide-former content ($V < 0.02\%$, $Nb < 0.01\%$ and titanium) both reversible temper embrittlement and precipitation hardening can be avoided. Consequently, it is possible with normal annealing practice, to eliminate the incidence of TIE in 22 NiMoCr 37 and, very probably, in 20 MnMoNi 55 as well (temper embrittlement test currently under way).

4.2.2 Strain induced Embrittlement (SIE)

The following ISO-V-tests (a_k -T-curves) and tensile tests at temperatures of 320 °C were carried out on specimens which were pretreated according to section 4, point c. The results of 22 NiMoCr 37 (reference melt 1) Figs. 11a and 12 are compared with an optimized melt which was received lately (MPA ref. 74), Fig. 11b. Fig. 11c shows a representative melt of material 20 MnMoNi 55.

Only after the formation of stress relief cracks in the mm-range do the tensile strength and yield point begin to drop more sharply. As long as only microcracks are present, the decrease observed in the strength value is negligible, i.e., in spite of fine structural breaks, the structure has a "100 % loading capacity" (high crack potential).

As far as deformation characteristics and impact strength are concerned, the values begin to fall steeply when the creep-strain reaches about 50 % of the available (time-dependent) creep-strain capacity. By the time the first microcracks appear (at 70 to 90 % creep strain capacity) the reduction of area (RA) values have fallen to < 10 % and the C_V -toughness to < 50 J/cm² (at 320 °C). This decrease in C_V -toughness occurs in the same way whatever the test temperature is, with the result that there is a general drop in C_V -impact strength values over the whole test-temperature range (Fig. 12). With regard to melt ref. 74, with optimized chemical analysis of 22 NiMoCr 37, Fig. 11b, the creep strain capacity after 6 h is with 3.2 % considerably higher than for melt 1 with 0.8 %. The steep decrease of RA starts at approximately 0.5 % creep strain and a lower shelf may be reached at 1.5 %. The C_V -toughness after 1. % creep strain may be at just under 100 J/cm². There were no indications for micro cracking in specimens with up to 0.8 % creep strain.

The material 20 MnMoNi 55, MPA ref. 60, has a creep-strain capacity after 6 h of 3.5 %. The RA exhibits a gradual decline. The deepest values are reached at about 2.5 % creep-strain. The C_V -toughness is about 100 J/cm² up to 1 % creep-strain. For this steel first micro cracks were already early observed, namely at 0.7 % creep-strain. In spite of this RA and C_V -toughness were as expected very high.

As a general rule it can be stated the higher the creep-strain capacity, the greater the creep-strain when strain damage eventually occurs.

The degree of strain damage for a given steel (i.e. the decline, for example, of the C_V -impact strength values following the application of a constant amount of creep-strain) is inversely proportional to:

- i) the extent of the creep-strain capacity (ϵ_{ZDB-t})
- ii) the degree of initial toughness in the plain annealing state ($\epsilon_{bl} = 0$)

and vice versa.

Note on i)

Improved creep-strain capacity (less SIE, less susceptibility to SRC) may be achieved by:

- the application of higher stress relief annealing temperatures (Fig. 13a and 13b) (longer annealing time, cf. below);
- structural improvements in double cycles (multilayer effects) (Figs. 13c and 5).

Note on ii)

Increases in the basic toughness values (annealing state) can be achieved by:

- a higher stress relief annealing temperature (Fig. 13b);
- longer times at a constant stress relief annealing temperature (Fig. 13a).

A further series of 22 NiMoCr 37 tests was conducted to monitor changes in toughness characteristics when, in the constant load test (6 h/610°C), samples with different degrees of strain damage (ϵ_{bl}) (Fig. 11a) are subjected to subsequent annealing (without load) for over 20 h at 610°C. As with the performance recorded in Fig. 11a, there is an improvement in toughness values for samples without ($\epsilon_{bl} = 0$) and/or very slight (reversible) strain damage. However, in so far as the subsequent tensile test at 320°C had shown a drop in the RA values as a result of greater (irreversible) strain damage, it follows that not even additional annealing for over 20 h/610°C is capable of improving the values again. There are plans to examine the effects of further annealing at 650°C, but it is probable that the same tendencies will be observed.

There is still a gap with regard to this way of observation in as far as SIE can also develop during the heating up period (compare 4.1.). The damages occurring in this connection mean that the degradation of the toughness during the holding phase, Fig. 12, already follows after appropriately low creep strains. Investigations are still under way as to whether and how far the creep strains generated during passing through of the critical temperature range, are to be equated with those in the isothermal test at 610°C.

It should be pointed out in this connection that the fall in deformation and toughness values is not the result of microcracking. Although 20 MnMoNi 55 displays initial microcracks at relatively low elongations, the two materials differ fundamentally in regard to toughness resp. strain damage: Fig. 11. Whereas microcracks do not occur in 22 NiMoCr 37 until the deformation and toughness values have decreased (rule of thumb: by the time microcracking begins, the RA has fallen to ca. 10 %), the deformation and toughness values in 20 MnMoNi 55 remain virtually unchanged despite incipient microcracking.

This behaviour on the part of both steels becomes understandable if one takes into account the chemical composition (especially Mo) and also the fact that microcracks have a greater tendency to occur in segregation bandings.

As far as 20 MnMoNi 55 is concerned, electron beam micro analyser tests show that the "negatively" segregated zones have a molybdenum content of 0.3 and 0.35%. The segregation coefficient ¹⁾ is about 2, with the result that about twice the Mo content was present in the segregation bandings ²⁾. It may be assumed that in those zones Mo content (likewise with a low trace element content) must have fallen short of a threshold value for SIE and SRC susceptibility. As far as 22 NiMoCr 37 is concerned, the Mo content is generally so high that in the event of cracking in the positively segregated zones significant strain damage (strain exhaustion as a preliminary to cracking) also occurred in the surrounding (negatively segregated) structure. Thus, even before the initial microcracks occur, the material is more or less subjected to strain damage (clear decrease in toughness values prior to initial microcracking).

All previous crack-sensitivity investigations have shown that materials with a very high creep strain capacity ϵ_{ZDB} tend relatively early to sustain microcracks. Owing to a high creep-strain capacity there is a high degree of crack tolerance. Materials with a lower creep-strain capacity, on the other hand, have high notch sensitivity (low crack tolerance), with the result that sample fracture occurs shortly after the first microcracks appear.

5. Correlation of Results and Proposals for Manufacturing

A comparison of the results obtained by the tangential-cut technique with that for the creep-strain capacity values determinant of crack susceptibility showed, Fig. 14, that tangential cuts from the material 22 NiMoCr 37 displayed no cracking if coarse-grained overheated specimens (cycle 13) showed a "V"-curve minimum at 610 °C ($\epsilon_{ZDB-min}$) greater than 0.9 - 1%. From this it can be concluded that materials with a minimum creep-strain capacity $< 1\%$ are subject to relaxation cracking during intermediate annealing at 550 °C or final annealing at 610 °C as well as, under certain circumstances, during the heating phase. It must be said by way of clarification that initial inter-crystalline microcracks are to be expected with elongations at fracture ($\epsilon_{ZDB-min}$) in the order of 1% from 0.7 - 0.8%.

The critical minimum creep-strain capacity determined for the material 22 NiMoCr 37, was confirmed during the investigations on 20 MnMoNi 55 to the extent that this steel displayed minimum values equal to or greater than 1.2% at 610 °C and, in line with these values, no microcracking was found in the tangential cuts, Fig. 14.

The two criteria for crack susceptibility (expressed by the critical chemical composition and the minimum creep-strain capacity at 610 °C of 1%) in Fig. 15 separate two areas in which results are included that showed in the tangential cuts either SRC or No SRC.

$$\frac{1) \quad C_{max}}{C_{min}}$$

- 2) In addition, the crack-inducing trace elements Sn and As and the residual elements in the steel (P, S and Cu) are presumably concentrated.

Since melts with more than 1.0% minimum creep-strain (Fig. 15) do not tend to SRC it is suggested to require at least 1% for the qualification test of RPV steels. Melts of material 20 MnMoNi 55 and optimized melts 22 NiMoCr 37 (Table 2) meet this crack susceptibility criterion.

It would be desirable, to deviate from the post damage testing after simulation of the stress relief heat-treatment a simple C_V -toughness criterion, e.g. 50 ftlb which also covers large, still realistic creep-strains originated in the heating and holding phase. Microcracks possibly developed in segregation bandings would then be insignificant (compare the results on 20 MnMoNi 55).

The susceptibility for SIE in the holding phase can up to now in each particular case be examined by means of mechanical post damage tests and the melt can accordingly be classified. The improved melts of material 22 NiMoCr 37 and those ones of 20 MnMoNi 55 under realistic maximum strain damages of 0.8% after 6 h at 610°C still exhibited C_V -toughness values of about 50 ftlb during the post simulation tests at 320°C, Fig. 11. However, it is still uncertain, in this connection, whether the mentioned strain damages produced during the heating-up period, which may possibly be irreversible - i.e. no recovery during the holding phase - compare Fig. 11a - can be completely covered by means of the isothermal constant load test. In order to clarify this circumstance appropriate tests are under way.

6. Improvement of the Steel Quality and the processing Parameters

According to the foregoing investigations, which also include unfavourable but realistic material states and crack overall conditions, embrittlement and cracking can be avoided by limiting the Mo-content, the elements P, S, Cu, Sn, N, As, Co as well as Al (Table 2), and further by a sharp reduction of the micro alloying elements V, Nb and Ti as well as the segregation bandings. Special melting processes characterized by favourable ingot solidification factors /21/ and use of the multi- /22/ or multiple-pouring process or the core remelting process MHKW /23/ ensure a relatively uniform distribution of the alloying elements in the ingot and the finished product, as well as high cleanliness and a reduced tendency to segregation /21, 23/. Coupled with good forging, this will consequently ensure that, apart from the favourable effects on susceptibility to stress relief cracking and the accompanying reduction in the tendency to SIE susceptibility to reversible temper embrittlement, in the coarse-grain HAZ is virtually eliminated, hot cracking (liquation cracking due to S and Sn content) largely avoided and the C_V upper-shelf toughness and through-thickness properties /21/ in the base material, substantially improved. Today RA values of 45% are already guaranteed in the through thickness direction /21/. Semi-finished products are not only because of this virtually free of ultrasonic indications, which, in fact is again favourable for flaw interpretation in the welded joints.

The HAZ simulation findings on these steels showed that minimizing the coarse-grain areas in the HAZ on the basis of a favourable layer technique and favourable penetration conditions during multilayer welding in the HAZ has the effect of producing additional, more or less normalized/refined zones with a low susceptibility to cracking and SIE. According to observations to date, a technique which lends itself particularly well to minimizing the coarse-grain proportions is the narrow-gap-seams-with-temper-bead technique familiar from turbine engineering and used recently in Japan and the USA for reactor pressure vessel seams. As far as the present materials are concerned, the emphasis was not to be placed on the recommended use of low heat inputs but rather on the aim of avoiding non retransformed zones in the HAZ, even at the expense of higher heat input. At the same time, a not unduly low heat-input reduces the risk of hydrogen-induced cold cracks.

As regards heat treatment, one of the recommendations to emerge from the HAZ simulation tests is that the highest possible stress-relief-annealing temperatures should be used, since high annealing temperatures produce a lower incidence of crack sensitivity and SIE in coarse-grain zones. As mentioned earlier, however, one should bear in mind that brittle temperatures (550 - 580°C) are traversed during the heating-up stage. Furthermore, attention should be paid at higher stress-relief-annealing temperatures to the effects on resistance to intercrystalline corrosion of the cladding, on the tensile properties of the welding material and on the C_v -toughness characteristics of the basic material /21, 24, 25/.

At all events, where materials of optimum chemical composition with a low segregation tendency are used, it is recommendable to omit the sometimes used intermediate annealing practice at 550°C and to replace it by a shorter stress-relief-annealing time, if at all possible at temperatures above 610°C. For melts with a lower creep-strain capacity in the coarse-grain state, a post weld heat-treatment should be applied, during which the crack-sensitive temperatures between 500° and 600°C are traversed as quickly as possible and without an interval.

Under these circumstances, it is possible to make reliable calculations with the help of fracture mechanics, since the material in question is homogenous and can be expressed in meaningful characteristic values. At the same time there would be no restrictions whatsoever as regards small cracks through the application of the specified mode of operation. Therefore, questionable repair welding, up to now normally applied also in cases of difficulties in flaw interpretation could be avoided.

References

- /1/ KUSSMAUL, K., Resistance of Welded Constructions in Pressure Vessel and Piping Technology with Special Regard of Defects in Welded Joints, IIW 1970 Public Session, Bd. A, S. 33/60, and Welding Research Abroad Vol. XVII (1971) No. 3, pp. 2/18
- /2/ WELLINGER, K., KRÄGELOH, E., KUSSMAUL, K. and STURM, D., Die Bruchgefahr bei Reaktordruckbehältern und Rohrleitungen, First International Conference on Structural Mechanics in Reactor Technology, Berlin, 20.-24.9.1971, Nuclear Engineering and Design 20 (1972), S. 215 - 235
- /3/ KUSSMAUL, K. and STURM, D., Research Activities at the MPA Stuttgart in the Field of Nuclear Pressure Vessel Technology, Material-Welding-Safety, Paper No. 12, Sixth Annual Information Meeting, HSST-Program, ORNL 1972
- /4/ KUSSMAUL, K., Verfügbarkeits- und Sicherheitsaspekte bei geschweißten Bauteilen mit größeren Wanddicken für Energieerzeugungsanlagen, 8. Allianz-Forum "Technik und Versicherung" 1972, der Maschinenschaden 45 (1972) H. 6, S. 231/242; Betriebs-Ökonom 11/1973, S. 209 und 12/1973, S. 228 - 233, Jap. Journ. for Failures in Mech.Engineering 1975, vol. 3, p. 60 - 70
- /5/ WELLINGER, K., SCHOCH, W., KUSSMAUL, K. and KRÄGELOH, E., Considerations of the Practical Use of Fracture Mechanics Technology with Particular Reference to Pressure Vessels, 3. Internationale Tagung über den Bruch, München, 8. - 13.4.1974 Zusammenstellung der Tagungsberichte Teil IV, S. VIII-525
- /6/ KELLERMANN, O., KRÄGELOH, E., KUSSMAUL, K. and STURM, D., Considerations about the Reliability of Nuclear Vessel Technology, Sec.Int.Conf. on Pr.Ves.Techn., San Antonio, ASME Part I, p. 25 - 38
- /7/ KUSSMAUL, K. and KRÄGELOH, E., Festigkeit und Zähigkeit als Sicherheitskriterien im Apparatebau, Zeitschrift für Werkstofftechnik, 5. Jahrgang, Heft 1 (1974), S. 1 - 8
- /8/ Sofortprogramm (Priority Program) 22 NiMoCr 37 Staatliche Materialprüfungsanstalt (MPA) Universität Stuttgart, Dezember 1973 sponsored by the Kraftwerk Union Erlangen
- /9/ Forschungsprogramm "Reaktordruckbehälter" (Urgency Program) Dringlichkeitsprogramm 22 NiMoCr 37, RS 101, by Order of the Federal Ministry for Research and Technology, Staatliche Materialprüfungsanstalt (MPA) Universität Stuttgart
- /10/ Sofortprogramm (Priority Program) 20 MnMoNi 55 Staatliche Materialprüfungsanstalt (MPA) Universität Stuttgart, Dezember 1976 sponsored by the Kraftwerk Union Erlangen
- /11/ KUSSMAUL, K., EWALD, J. and MAIER, G., Assessing Heat Affected Zone Properties of Pressure Vessel Weldments Using Simulation Techniques, International Journal of Pressure Vessel and Piping, volume 5, No. 2, p. 85 - 101, Verfahren zur Simulation der Wärmeeinflußzonen von Schmelzschweißverbindungen, Schweißen und Schneiden, Heft 7/76, S. 250 - 255

- /12/ KUSSMAUL, K., BLIND, D. and EWALD, J., Investigation for the Detection and Study of Stress-Relief Cracking, International Journal of Pressure Vessel and Piping, to be published in volume 5, No. 2 see: Schweißen und Schneiden, Heft 6/76

- /13/ SCHELLHAMMER, W., MAILE, K. and MAIER, S., Zentrale Auswertung von Herstellungsfehlern und Schäden im Hinblick auf druckführende Anlageteile von Kernkraftwerken (Tangentialschliffprogramm 22 NiMoCr 37 - BMI - TB SR 10), MPA Stuttgart, S. 1 - 114 (Central Evaluation of Manufacturing Defects and Failures in the Pressure Boundary System of Nuclear Power Stations (Tangential Cut Project 22 NiMoCr 37 - BMI - TB SR 10), MPA Stuttgart, p. 1 - 114 Forschungsprogramm SR 10, durchgeführt im Auftrag des Bundesministeriums des Innern

- /14/ VOUGIOUKAS, P., Über die selektive und kumulative Wirkung von Legierungselementen auf das Relaxationsverhalten von Feinkornbaustählen, Diss. TU Clausthal, 1973. (VOUGIOUKAS, P., FORCH, K., PIEHL, K.H.: Stahl und Eisen 94 (1974), S. 805 - 13)

- /15/ NAKAMURA, H., NAIKI, T., and KABAYASHI, H. Fracture in the Process of Stress Relaxation under Constant Strain, Proceedings of the First International Conference of Fracture, V 2, pp. 863 - 878 (1965), Sendai, Japan

- /16/ ITO, Y. and NAKANISHI, M., Study on Stress-Relief Cracking in welded Low-Alloy Steels, International Institute of Welding, Document X-668-72

- /17/ KUSSMAUL, K. and EWALD, J., Assessment of Toughness and Cracking in the Heat-Affected Zone of Light Water Reactor Components, Third Intern. Conference on Pressure Vessel Technology, Tokyo, April 18 - 22, 1977, to be published

- /18/ EWALD, J. and KUSSMAUL, K., Methods for Assessing the Properties of the HAZ, Review Document, CSNI Expert's Group on Mechanical and Material Problems Relating to the Safety Aspects of Steel Components in Nuclear Plants, Specialist's Session March 1976, Report from the Staatliche Materialprüfungsanstalt (MPA), University of Stuttgart

- /19/ KUSSMAUL, K., EWALD, J., und MAIER, G., Einfluß von Aufheizphase und Haltephase beim Spannungsarmglühen auf Ribbildung und Schädigung in der Wärmeeinflußzone, Influence of Heating and Holding Period during Stress Relief Heat-Treatment on Embrittlement and Cracking; to be published

- /20/ Temper Embrittlement in Steel, STP 407, American Society for Testing and Materials, 1967

- /21/ CHEVIET, A., and ROUX, J.-H., Heavy Gauge Plates for Nuclear Application, Second Symposium, Special Steels and Nuclear Energy, Dec. 14 - 15, 1976

- /22/ DEBRAY, W., CERJAK, H., ONODERA, S., TSUKADA, H. and SUZUKI, K., Large Nozzle Belts Forgings for PWR 1200 MWE. European Nuclear Conference, Paris. April 21 - 25, 1975

- /23/ AUSTEL, W., MAIDORN, CH., Metallurgical and Technological Properties of Forgings made by the MHKW-Process. Fifth International Conference on Vacuum Metallurgy and Electroslag Remelting Processes, Munich, 11 - 15 October 1976
- /24/ CERJAK, H., DEBRAY, W., PAPOUSCHECK, F., Eigenschaft des Stahles 20 MnMoNi 55 für Kernreaktorkomponenten, VGB-Konferenz "Werkstoffe und Schweißtechnik im Kraftwerk", Düsseldorf 1976, S. 3 - 44
- /25/ AUSTEL, W., GROTE, G., MAIDORN, CHR., SCHRÖDER, J., Effect of temperature and holding time at stress-relieving on the decrease of internal stresses and the mechanical-technological properties of the steel 22 NiMoCr 37, Archiv für das Eisenhüttenwesen 45 (1974), S. 309 - 315

Tafel 1: Chemical Composition (requirements)

	C	Si	Mn	P	S	Al	Cr	Cu	Mo	Ni	Ta	Co	V
ASTM A 508 C12-74	≤0,27 0,15 0,35	0,15 0,35	0,50 0,90	≤0,025	≤0,025	-	0,25 0,45	-	0,55 0,70	0,50 0,90	-	-	≤0,05
ASTM A 508 C13-74	0,15 0,25	0,15 0,35	1,20 1,50	≤0,025	≤0,025	-	-	-	0,45 0,60	0,40 0,80	-	-	≤0,05
ASTM A 533B C11-74	≤0,25	0,15 0,30	1,15 1,50	≤0,035	≤0,040	-	-	-	0,45 0,60	0,40 0,70	-	-	-
22 NiMoCr 37 Forging Mat. VdTÜV 365 (4.72)	0,17 0,25	≤0,35	0,50 1,00	≤0,025	≤0,025	≤0,05	0,30 0,50	≤0,20	0,50 0,80	0,60 1,00	-	-	≤0,05
22 NiMoCr 37 Plate Mat. VdTÜV 366(1.74)	0,16 0,21	0,15 0,30	0,50 1,00	≤0,020	≤0,010	0,010 0,040	0,20 0,50	≤0,20	0,45 0,70	0,50 0,90	≤0,030	≤0,030	
20 MnMoNi 55 Plate Mat. VdTÜV Tent.401(76)	0,17 0,23	0,15 0,30	1,20 1,50	≤0,015 (≤0,010)	≤0,015	0,010 0,040	≤0,20	≤0,18 (≤0,10)	0,45 0,60	0,50 0,80	≤0,030	≤0,030	≤0,02
20 MnMoNi 55 Forging Mat. VdTÜV Tent.Oct.76	0,17 0,23	0,15 0,30	1,20 1,50	≤0,015 (≤0,012)	≤0,015	0,010 0,040	≤0,020	≤0,18 (≤0,10)	0,45 0,60	0,45 0,80	≤0,030	≤0,030	≤0,02

1) In Core-Materials

Table 1: Chemical Composition

Table 2: Enumeration of elements ("vanadiumfree" melts) with differentiation of specimens with and without cracking and threshold values (%)

Crack advancing elements and threshold values for cracking occurrence

"Threshold value criterion": cracking, if by two or more elements the threshold value is exceeded

	Trace elements and impurities							Carbide forming elements	
Liquation cracking	Cu 0,13	Sn 0,011	P 0,007	S 0,008	As 0,036				
Stress-relief cracking	Cu 0,12	Sn 0,011	P 0,008	S 0,008	As -	Al -	N 0,013	Mo 0,62	Co -

Table 2: Threshold Value Criterion

Fig. 1: Fracture Toughness, CT specimens

Fig. 2: ISO-V-Impact bending Tests on Weldings

Fig. 3

Fig. 3: Overheating Cycles 13 and 15

Fig. 4

Time

Fig. 4: Elongation at Fracture at diff. Temp. (Cycle 13)

Fig. 5: Elongation at Fracture with different Overheating Cycles

Fig. 6: Elongation at Fracture - Comparison of Materials

Fig. 7: Minimum Elongation versus Mo-content

Fig. 8: Annealing of overheated 20 MnMoNi 55

Fig. 9: Annealing of overheated 15 MnNiMoNb 54

Fig. 10: Influence of Cooling Rate after Annealing

Fig. 11a: Post Damage Tests on 22 NiMoCr 37 (1)

Fig. 11b: Post Damage Tests on 22 NiMoCr 37 (74)

Fig. 11c: Post Damage Tests on 20 MnMoNi 55 (60)

Fig. 12: Post Damage Tests on 22 NiMoCr 37 (1)

Fig. 13: Post Damage Tests on 22NiMoCr 37 (1) Variation of overheating and SRHT-Conditions - a_K/T -Curve

Fig. 14: Minimum Elongation at fracture - Comparison of diff. Melts

Fig. 15: Minimum Creep Strain versus Threshold Value Criterion