HIGH TEMPERATURE LOW-CYCLE FATIGUE OF AUSTENITIC AND FERRITIC WELDMENTS

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The low-cycle fatigue behaviour of 2 1/4Cr1Mo (10 CrMo 910) and of 304 L Stainless Steel (X2CrNi 189) weldments was investigated in strain controlled tension-compression at elevated temperatures. It is shown that the low-cycle fatigue life decreases with decreasing strain rate. The experimental results show that the influence of the strain rate at elevated temperature (590 °C respectively 600 °C) can be well described by the frequency-modified Coffin model and the Ostergren hysteresis energy model.

INTRODUCTION

Failure due to low-cycle fatigue has been recognized as an important design criterion for structural components in a wide range of practical performance. In case of pressure vessels, boilers and many other power station units, fluctuating thermal loading and gradients associated with pressure vibrations cause local plastic flow in the region of stress raisers such as nozzles and welded connections. Therefore the effect of welding on life of structure subjected to cyclic plastic loading has to be considered for economic and safe design. In spite of the practical importance, studies on the weldments in the past have been confined to the life prediction at room temperature.

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The aim of the investigation is to find out whether models developed by others [1], [2] for fatigue life estimation of base metal at elevated temperatures are applicable for weldments.

The previous work [3], [4] has shown that the LCF life of weldments at elevated temperatures can be described by the analogeous equation to Coffin-Manson rule [4]. The effect of loading frequency on LCF of weldments has not been studied intensively in the past. In this paper, therefore, the results of the investigations on the effect of frequency at elevated temperatures on LCF of weldments are reported and the applicability of the models proposed by others is examined.

EXPERIMENTAL

The austenitic steel Type 304 L Stainless Steel (X2CrNi 18 9) and the low alloyed steel 2 1/4 Cr1Mo (10 CrMo 9 10) were chosen for study. The austenitic specimens were machined from 22 mm thick plate keeping their axis parallel to rolling direction of the plate and solution treated. The specimens of the low alloyed steel were machined from a pipe and after welding post-heat treated. All welding was in the flat position using metal arc processes with basic coated electrodes for 2 1/4 Cr1Mo and with rutile coated electrodes for 304 L as double V butt weld.

Both longitudinal and transverse butt welds were used for tests. All butt welds were checked radiographically according to German Film Classification D4. Only specimens showing no radiographically detectable defects were used for the tests. The specimen size and type of welds are illustrated in Fig. 1. Chemical analysis and mechanical properties of base metal and weldments are given in Fig. 2. For tests at high temperature a multi-zone radiant heater was used. The test temperature was 600 °C for the austenitic steel and 590 °C for the low alloyed steel. The specimen surface was chemically and electrolytically polished before testing. The strain controlled fatigue tests were carried out on a servo-hydraulic push-pull machine ($R_\varepsilon = -1$), the strain rate was between 8.34×10^{-3} and 4.17×10^{-5} 1/s.

RESULTS

304 L Stainless Steel

Fig. 3 illustrates the results of the strain controlled tests under axial loading on specimens with transverse and longitudinal butt welds. The effects of frequency are shown for comparison. Decreasing the strain rate from 8.34×10^{-3} 1/s to 4.17×10^{-5} 1/s at total strain range of 1.6 % results in a reduction of fatigue life up to 60 % and 50 % for longitudinal and transverse welds respectively. It can be seen that not only the reduction of fatigue life but also the decrease of the tensile stress σ_t occurs due to the additional time-dependent damage at high temperature.

In Fig. 4a-c the cyclic stress-strain curves are shown for base metal and welded specimens. It should be mentioned that the value $\Delta\sigma/2$ at $N_c/2$ has been chosen for plotting the cyclic stress-strain curve. The fatigue life curves, i.e. plots of log $\Delta\varepsilon_a$ vs. log N_c are given in Fig. 5. N_c is defined as the number of cycles leading to the formation of a macro crack which noticeably weakens the specimen [6] (Fig. 3). Fig. 5a illustrates the results for base metal at high temperature, room temperature results are given for comparison. It can be seen that the fatigue life reduction due to temperature increase is more pronounced at high strains than at low strains; the slopes of the high temperature curves are not so steep as that of room-temperature curve. The decrease of the strain rate leads to a remarkable decrease of the fatigue life especially at low strain ranges.

The results obtained with longitudinal welds are shown in Fig. 5b. As for base metal the temperature increase results in a decrease of the fatigue life for longitudinal welds. Again the frequency effect is more pronounced at low strains. It has been found that at low strains the fatigue life of welded specimens is always less than that of base metal at all temperatures and frequencies employed, while at high strains a considerable increase of fatigue life of welded specimens occurred compared with base metal. At 600 °C the fatigue life increase amounts 70 %.

Fig. 5c shows the fatigue life curves for transverse welds. The fatigue life of transverse welds is always less than that of base metal and longitudinal welds at all frequencies regardless of test temperature [7] to [17]. It should be added that in displacement controlled tests on specimens with transverse butt welds, the strains are distributed inhomogeneously over the strain gauge so that the results can be used only for qualitative comparison.

2 1/4 Cr1Mo

In Fig. 6a-c the cyclic stress strain curves of the low alloyed base metal and welded specimens are shown. The values of $\Delta\sigma/2$ were defined in the same way as in Fig. 4. In high temperature region, decreasing strain rate leads to a more pregnant decrease of stress-range than for the austenitic steel.

The fatigue life curves for base metal and longitudinal and transverse butt welds are shown in Fig. 7. In analogy to the austenitic steel increasing temperature leads to decreasing fatigue life and an increasing of the slope of the fatigue life curve. The fatigue life is more reduced at low strains than at high strains. The decrease of strain rate leads especially at base metal and longitudinal welds to a remarkable decrease of the fatigue life, the frequency effect is more pronounced at low strain.

In contrast to the austenitic steel the fatigue life of the low alloyed welded specimens is always less than that of base metal, regardless of test temperature and frequency.

Fig. 7 shows the results of the fatigue tests of the transverse welds. The fatigue life is always less than that of base metal and longitudinal welds at all frequencies.

In opposit to the base metal and longitudinal welds, transverse butt welds show no influence of frequency on fatigue life. The results lay within a relatively wide scatter band.

DISCUSSION

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The cracks of transverse welded specimens start at welds regardless of test temperature. Though all specimens were machined from the part of welds showing no radiographically detectable defects the micro fractographic examination of the fracture surface has revealed that the crack initation occurred often at the site of the micro slag inclusions and micro pores. In case of the longitudinal welded specimens the crack partly started from base metal and partly from welds. Again the metallurgical inhomogeneity was responsible for crack initiation when the crack started from the welds.

According to Munse [15] the effect of weld defects and metallurgical inhomogenities on low-cycle fatigue is more pronounced at low strains than at high strains, and this is probably responsible for the large scattering of the results at low strains (Fig. 5c).

Investigations on AISI 316 SS by Raske et al [18] have shown that the fatigue life of longitudinal welded specimens was 4 to 7 times of the base metal. The finely dispersed delta ferrities within the austenitic cast structures seem to be responsible for the fatigue life increase. They believe that the fine dispersed delta ferrities as well as the phase boundaries retard the crack growth.

In the present work the delta ferrites were measured by means of Förster-Probe. It was found that the welds had 100 times higher delta ferrite content compared with the base metal so that the fatigue life increase of the longitudinal welded specimens seems to be related to the delta ferrite content. Shahinian et al. [19] have investigated the AISI Type 304 and 316 steels and found that the crack growth rate of the submerged-arc welded specimen is lower than that of the base metals, especially at elevated temperature. Their findings can be also interpreted in terms of the delta ferrite content.

The transverse welded specimens showed the lowest fatigue life value. Due to the difference in mechanical properties of weld, HAZ and base metal inhomogeneous straining of the specimen occurs. Because of the metallurgical notches introduced, a quantitative estimate of the fatigue life of transverse welded specimens is impossible. The values of $\Delta\varepsilon_a$ are the mean strain ranges calculated as the displacements devided by 20 mm, the gauge length over which the displacement was controlled. The actual strains will have varied from one part of the weldment to another due to differences in yield strength.

For a given strain rate the slope of the fatigue life curve for base metal decreases with increasing temperature. This difference between room and high temperature behaviour can be accounted for by the fact that the cyclic stress-strain curves for the base metal at room temperature and at high temperature have various strain-hardening exponents, Fig. 4. At high temperature the plastic strain portion of a given total strain amplitude is larger than at room temperature, thus leading to an increased reduction of the fatigue life. For welded specimens the difference in the cyclic stress-strain behaviour was not so pronounced as for the base metal

(Fig. 4). The increase of the slope α of the fatigue life curve for welded specimens with increasing temperature agrees well with the results of others [4]. Similar results were obtained [1], [20] for other base metals.

An additional increase of the slope α was observed by decreasing the strain rate from 8.34×10^{-3} 1/s to 4.17×10^{-4} 1/s for the welded specimens as well as for the base metal. This increase is thought to be caused by the accelerated oxidation due to environment. The welded specimens are evidently more susceptible to the environmental attack than the base metal. The increased amount of delta ferrite in the welds is resoponsible for the higher susceptibility. Kußmaul et. al. [21] have shown that the environmental effect is more pronounced at lower strain ranges than at high strain ranges. At low strains the environmental attack dominates and reduces the fatigue life remarkably, while at high strains the delta ferrit mostly retards the crack growth and the environment effect is negligable. Decreasing the strain rate to 4.17×10^{-5} 1/s is thought to cause an additional damage by creep.

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In analogy to the austenitic steel decreasing strain rate at elevated temperature results in similar effect on the cyclic behaviour.

The elevated temperature is responsible for the reduction of fatigue life, whereby the reduction at lower strain is more pronounced than at high strain. In contrast to the austenitic steel, the low alloyed base metal and weldments show strong oxidation tendency. For base metal as well as longitudinal and transverse welds accelerated crack initiation and crack propagation occurred by bursting of oxide layers and selective corrosion.

Under all test conditions, transverse welded specimens show the lowest fatigue life value. In contrast to the austenitic weldments, cracks started regardless of test temperature, strainrate and -range in HAZ and partly in base metal.

The cracks of longitudinal welded specimens started in base metal and partly at welds. Microfractographic examiniations of the fracture surface revealed that crack initiation in weld metal often occurred at the site of micro weld defects, which had not been detectable by radiographic examination.

FATIGUE LIFE PREDICTION

The results of the investigations on the base metal and the welded specimens can be described for room temperature as well as for high temperature by the analogeous equation to Coffin-Manson relationship.

$$(N_c)^{\alpha} \times \Delta\varepsilon_a = C \qquad (1)$$

The time-dependent damage processes at high temperature are not included in equation (1). For this reason Coffin [1] proposed a relationship between fatigue life at high temperatures and loading frequencies which includes time dependent processes such as environmental attack and creep. His model has been proved for base metals by others [1], [20], [22]. The fatigue life is expressed in the form

$$(N_c \times f^{k-1})^{\beta} \; \Delta\varepsilon_a = C \qquad (2)$$

where f is frequency, k and C are constant. β is a temperature dependent material constant.

In Fig. 8 the results for the longitudinal and the transverse welded specimens as well as for the base metal are given in the form of the frequency-modified fatigue life as proposed by Coffin. The results for the base metal show a narrow scatter band. The comparatively wide scatter band of the welded specimens is caused by the inhomogeneity of the microstructures within the welds [11], [13], [15]. From the results given above it can be concluded that the fequency-modified fatigue life concept of Coffin may be used for life prediction of the welded specimens of both materials.

The model suggested by Ostergren [2] is based on the assumption that in low-cycle fatigue regime the fatigue crack growth occupies most of the life time and that the crack initiation occurs soon after few cycles. It is assumed that only the plastic portion of the total strain is important for the fatigue damage and for the crack growth. Only the deformation occurring in the portion of the cycle with the crack open is considered to contribute to the state of the damage by propagating the crack. The measure of damage is defined as the net tensile hysteresis energy of the fatigue cycle. As a first approximation it is postulated that crack opening and subsequent propagation occurs for a positive stress value. The net tensile hysteresis energy may be computed from the plastic strain $\Delta\varepsilon_p$ and the tensile stress $\sigma_t = \frac{\Delta\sigma}{2}$ so that a relationship in the form

$$N_c = C\ (\Delta\varepsilon_p \times \sigma_t)^\beta \qquad (3)$$

was obtained.

The equation (3) is established only for time-independent damage process, though the measure of fatigue damage is now extended by the stress value. For time-dependent damage mechanisms Ostergren introduced a frequency term similar to the fequency-modified fatigue life of Coffin:

$$N_c = C\ (\Delta\varepsilon_p \times \sigma_t)^\beta\ f^m \qquad (4)$$

where f is frequency and m constant.

Fig. 9 illustrates the results of the investigations plotted using equation (4). The scatter band is comparable with that obtained using frequency-modified fatigue life concept. It can be seen that the two models described are substantially identical. At low strain ranges the effect of strain rate on σ_t is negligible so that a good agreement between the two models is observed. At high strain ranges, however, the strain rate effect on σ_t may be considerable and the equations (2) and (4) can lead to a different life prediction as shown in Fig. 8 and 9.

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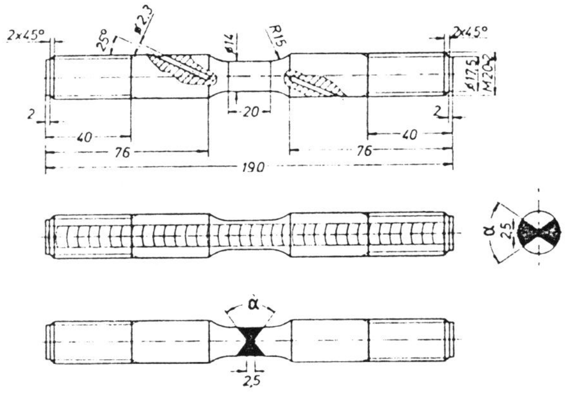


Fig. 1. Specimen size and type of welds
$\alpha = 70^{\circ}$ (304 L)
$\alpha = 60^{\circ}$ (2 1/4 CrlMo)

MATERIAL	MECHANICAL PROPERTIES			
	YIELD N mm^{-2}	UTS N mm^{-2}	ELONG %	IMPACT VALUE J
304 L	280	601	46	DVM 180
FILLER METAL	340	550 - 650	>35	ISO-V >55
2 1/4 CrlMo	355	513	27	
FILLER METAL	>440	575 - 675	>20	ISO-V >120

MATERIAL	CHEMICAL COMPOSITION							
	% C	% Si	% Mn	% P	% S	% Cr	% Ni	% Mo
304 L	0,025	0,37	1,48	0,026	0,003	18,04	10,13	
FILLER METAL	0,04	0,70	0,80			19,0	10,0	
2 1/4 CrlMo	0,10	0,26	0,47	0,013	0,017	2,28		0,97
FILLER METAL	0,06	0,50	0,80			2,4		1,0

Fig. 2: Mechanical properties and chemical composition

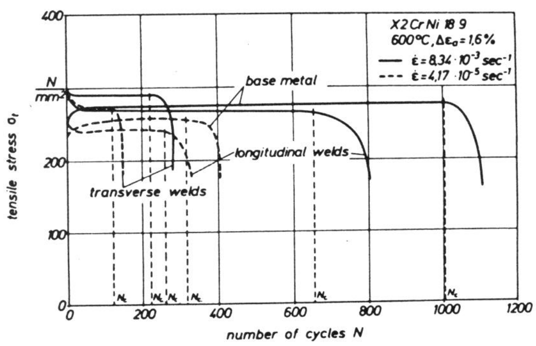


Fig. 3: Effect of strain rate on cyclic flow curve

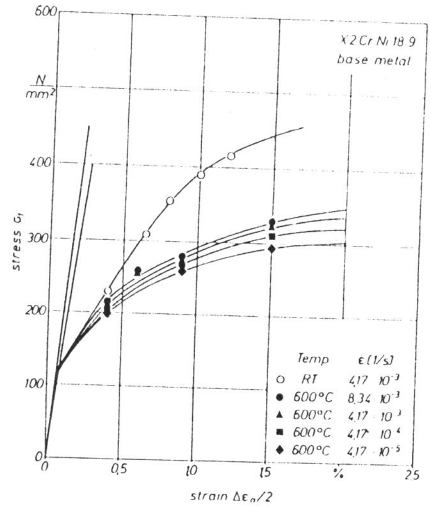


Fig. 4a

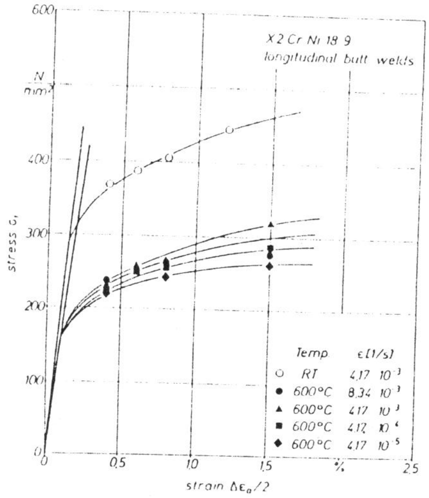


Fig. 4b

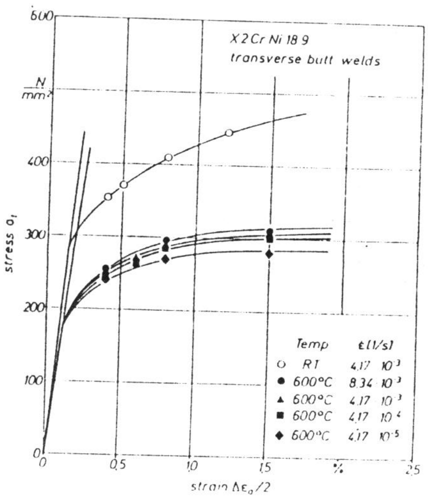


Fig. 4c

Fig. 4a-c:

Cyclic stress-strain curve for base metal, longitudinal and transverse butt welds, 304 L

Fig. 5a

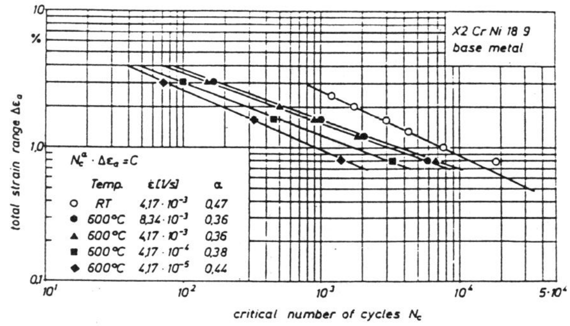


Fig. 5b

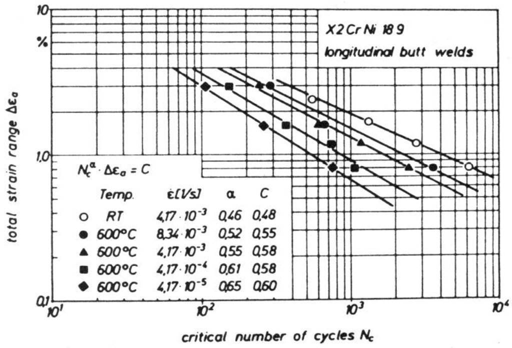
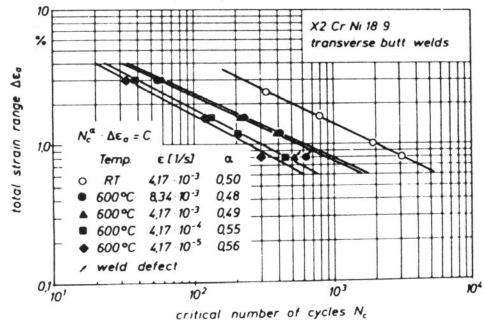


Fig. 5c



<u>Fig. 5a-c:</u> 304 L, base metal, longitudinal and transverse butt welds, total strain range vs. critical number of cycles.

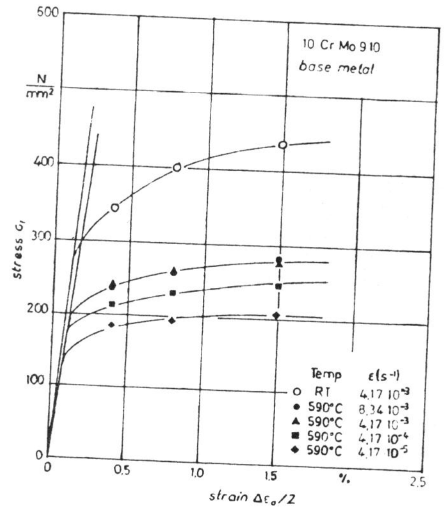


Fig. 6a

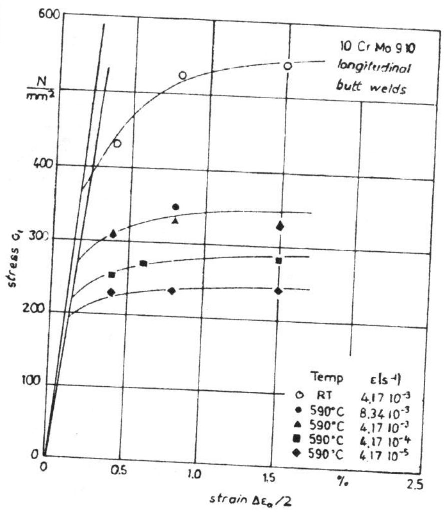


Fig. 6b

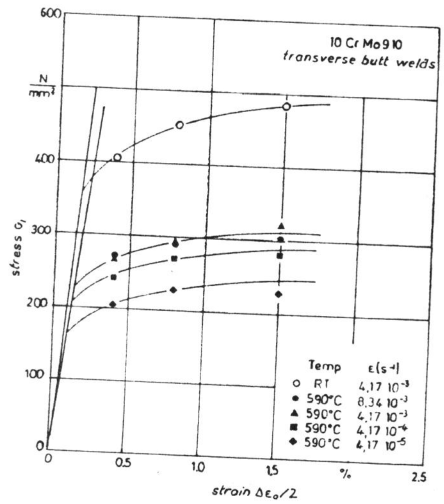


Fig. 6c

Fig. 6a-c: Cyclic stress-strain curve for base metal, longitudinal and transverse butt welds, 2 1/4 Cr1Mo

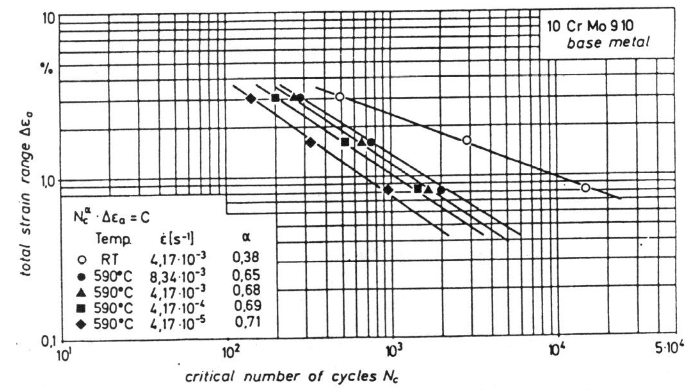


Fig. 7a

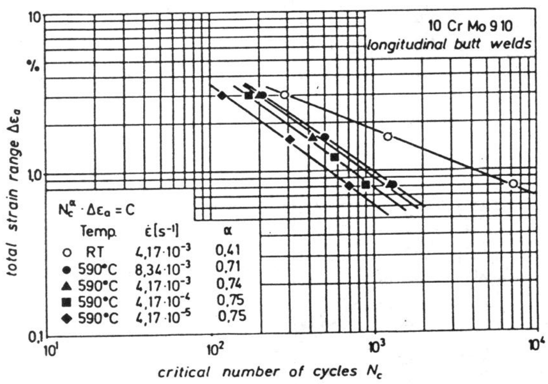


Fig. 7b

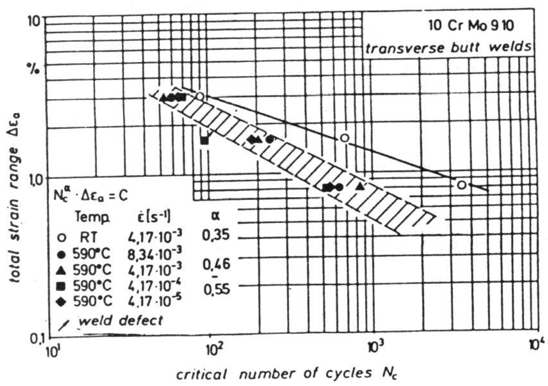


Fig. 7c

Fig. 7a-c: 2 1/4 Cr1Mo, base metal, longitudinal and transverse butt welds, total strain range vs. critical number of cycles.

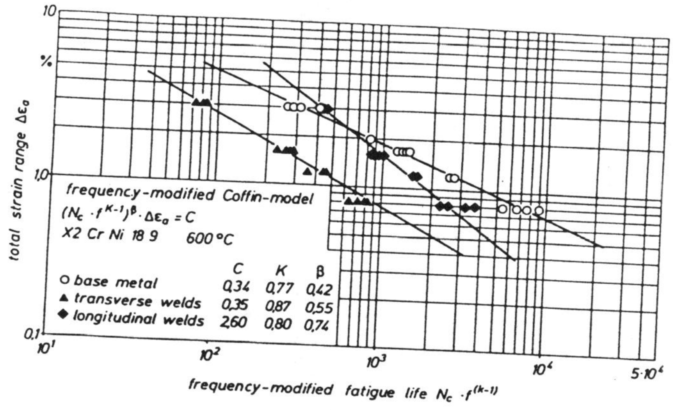


Fig. 8a

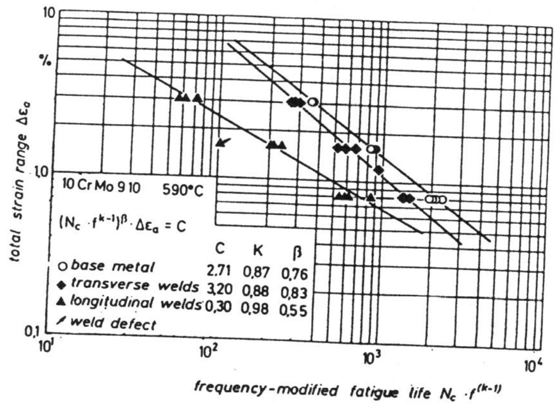


Fig. 8b

Fig. 8a,b: Frequency modified fatigue life vs. total strain range.

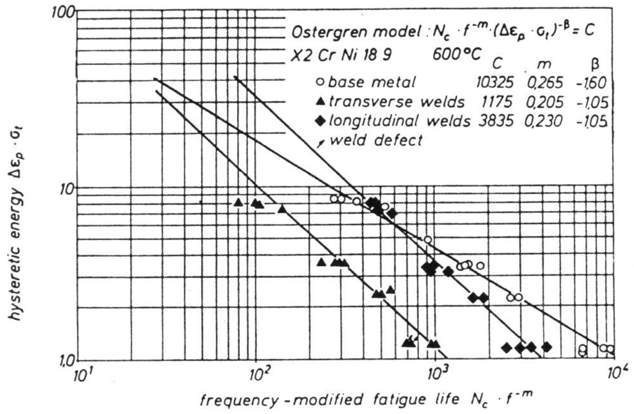


Fig. 9a

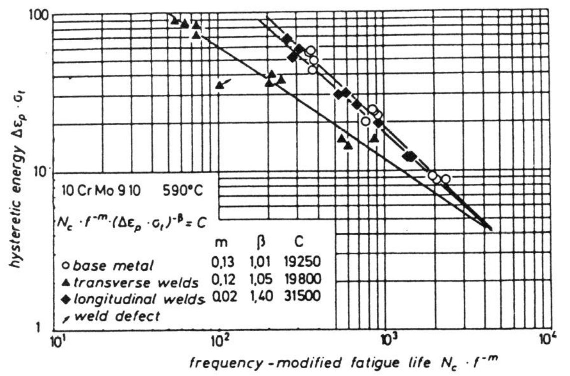


Fig. 9b

Fig. 9a,b: Frequency modified fatigue life vs. hysteretic energy