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JOINING OF DISSIMILAR METALS USING TRANSITION PIECES WITH COMPOSITIONAL GRADIENTS

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ABSTRACT

The need for higher efficiency in power plant will result in higher operating temperatures and more sophisticated materials being used in their construction. The dissimilar metal welds between differing materials are already recognised as critical components. The problem arises from the differing coefficients of thermal expansion on each side of the weld which produce thermally induced stresses within the weld each time it is thermally cycled. With the introduction of martensitic 9 and 12%Cr creep resistant steels, the levels of thermally induced stresses are likely to increase as these steels have a lower thermal expansion than conventional ferritic steels thus increasing the mismatch. One method of reducing these thermal stresses is to replace conventional dissimilar metal welds with transition pieces having continuous compositional gradienta along their lengths. Several such graded composition joints were produced in the 2¼%Cr/316, mod 9%Cr/316 and 2?4 %Cr/mod 9 %Cr material combinations.

These graded composition joints were extensively modified to produce optimised joints with creep strength above those currently found in diasii weldments. These jointa also had good toughness and oxidation resistance comparable to existing dissimilar metal weldments. This work has validated the concept of using transition pieces with compositional gradients produced by Hot Isostatic pressing as a means of obtaining improved dissimilar joint creep life and thermal cycling resistance.

INTRODUCTION

In the past the dominant role in power generation has belonged to the fossil fuel sector [1]. Whilst in Europe the role of fossil fuel varies from country to country, it remains the single largest source of electricity generation. Currently, some 80% of world-wide electricity production is consumed within Europe, Japan and the USA and the former Soviet Union although they represent only 23% of the world population. As the economies of other countries develop their demand for electricity must increase greatly.

Such a global increase in demand for electricity will be far beyond that possible on the basis of nuclear, hydro or renewable power sources and an increase in fossil fuel burning is inevitable.

The efficiency of boiler systems is controlled by thermodynamic limitations. Effectively the higher the pressure and temperature achievable, the higher the thermal efficiency of the system and the lower the consumption of fuel, and the production of waste gases, for a given output of electricity. Much larger increases in efficiency can be achieved if, again by increasing the operating temperatures and pressure, the boiler can be made to operate in the supercritical regime (i.e. where liquid and vapour phases no longer exist as separate entities). This increase in efficiency demands significant changes be made from natural convection recirculating boilers (present conventional design) to forced convection once-through boilers for supercritical operation. This requires the use of high chromium creep resistant steels together with austenitic materials for these higher temperature higher pressure applications.

Considering the case of pulverised fuel fired plant, the realisation of high efficiency supercritical plant is estimated to produce overall increases in thermal efficiency of around 6% when conditions change from 160 bar 538°C to 356 bar 649°C. In terms of fuel consumed, and waste gas produced, the savings are substantially greater being in fact in excess of 14%. Such fuel savings alone have a large impact on the economics of such plant. However, further economic advantages accrue from the reduction in emissions, which substantially reduces the cost overhead for coal fired plants.

The high chromium steels can suffer from type IV cracking which results in failure in the parent plate immediately behind the visibly transformed HAZ i.e. the intercritically annealed zone of the HAZ. In this region creep properties may decrease by 30% which is dealt with by careful design ensuring that weldments are made in such an orientation that they are not subject to cross weld hoop stresses but only to cross weld axial stresses. As the axial stresses are equal to only 50% of the hoop stress the weldments therefore have a creep life above the design life of the boiler.

Testing on dissimilar metal weldments (DMW's) has shown the decrease in creep strength due to Type IV failures can be particularly pronounced and has been attributed to the creation of a complex stress pattern due to the interaction between the thermally induced stress arising from differential thermal expansion and the (possibly tri-axial) residual stresses remaining after welding and post weld heat treatment. Hence an improved method of joining 9 and 12%Cr steels to austenitics which addresses and reduces the above effects would be very beneficial, not least in relaxing the constraints placed on the designers.

Originally ferritic/austenitic joints were made by fusion welding using austenitic filler materials. On occasion these joints failed prematurely (within the design life of the plant containing them) resulting in expensive repairs and costly loss of production. This arose through a combination of the generation of high thermal stresses, caused by the abrupt change in thermal expansion coefficient at the ferritic/austenitic weld interface and the development of a decarburised zone due to carbon migration from the ferritic material into the austenitic weld metal as a result of the high affinity of carbon to the chromium rich austenitic material[2]. Such joints are now made in MBEL using a high nickel filler metal (Inconel 82 or 182) which acts as a diffusion barrier to prevent carbon migration and has a coefficient of thermal expansion intermediate between those of ferritic and austenitic steels. Whilst these joints are operating successfully in current power plant, they are widely recognised to be a critical component within the plant.

A survey of dissimilar weldments[3] determined that of 320 generating units containing austenitic material some 60 had experienced dissimilar weld failures. Whilst the majority of these failures (38) were in weldments made with austenitic filler metal some seven units containing dissimilar metal welds made with nickel filler metal experienced failure. However, it must be recognised that while the nickel weldments exhibit a longer mean time to failure (100,0OOhrs) than the austenitic filler weldments (74,000hrs) they had also seen a lower total time in service. It is therefore, surprising that 12% of all failures reported occurred in welds made with nickel fillers. Whilst more recent designs of transition weld appear to have an acceptable creep life they remain a critical component and, in view of the problems previously found, a new approach to the joining of dissimilar metals would be beneficial.

In this programme graded joints have been developed between 2¼%Cr, austenitic stainless steel and mod 9% Cr. The 2¼%Cr/austenitic transition is commonly found in existing plant whilst the Mod 9% C/austenitic joint will be used in advanced supercritical plant. Mod 9%Cr has a lower expansion than 2¼% Cr and thus higher thermal stresses will be produced *in* Mod 9%Cr/316 joints.

The 2¼%Cr/Mod 9 %Cr joints (between ferritic 2¼%Cr and mod 9% Cr) have the lowest difference in thermal expansion coefficient and the lowest thermal induced stresses. These joints are not yet common but are likely to be used in some plant.

The principal objective of the programrne was to show that the problems of differing thermal expansion coefficient, creep strength and of carbon migration in dissimilar metal transition joints could be minimised by the adoption of graded joints manufactured in such a way that the composition varies along their length.

The joints developed in this way were produced using powder metallurgy. Two different fabrication routes were developed using Hot Isostatic Pressing (HIPping) and Vacuum Plasma Spraying (VPS). Despite extensive development of the VPS fabrication route it was not possible to achieve the required mechanical properties using this route. In particular joint toughnesses were unacceptable. It is therefore not intended to produce graded dissimilar joints using this fabrication process.

The HIPped graded joint comprised a graded zone some 50-100mm long with ends produced of pure HIPped parent metal. These joints were typically 300mm long and were designed to be welded as a transition piece between the two materials to be joined. All the joints tested in this work were produced in 50mm diameter bars but, in commercial production, graded tubular transition pieces would be fabricated by HIPing.

PRODUCTION

Both HIPped parent materials and the HIPped graded joints were produced at Krupp Pulvermetall. The porosity in this HIPped material was too low to measure and was well below the 1% level. Due to the slow cooling rate the HIPped material exhibited microstructure different to those found in wrought materials. 'It was therefore necessary to heat treat the bars prior to testing. The heat treatments used are given in Table 1 and those for HIPped parent material correspond to those for normal wrought parent material. It was recognised that actual HIPped graded joints were relatively small and both parent materials would in practice be subject to the same heat treatment regimes which would include a post weld heat treatment of the terminal welds. The heat treatment giving the optimum microstructure for the lower creep strength material was applied to each joint. These heat treatments are shown in Table 1 and two possible heat treatments existed for the HIP/wrought 316 weld. The regime furthest from the normal heat treatment (and likely to give the lowest mechanical properties) was used in testing this weldment.

All the HIP joints were produced from powders with a maximum diameter of less than 500pm. Stepped graded bars were produced from a series of powder blends corresponding to the required band composition. Sufficient of each blend is filled into the capsule to give the required thickness of band after compaction and to locate the graded zone approximately in the centre of the bar. Care was taken to ensure that the interfaces between individual layers remained planar during fabrication.

During the fabrication of a stepless gradient two powders have to be filled into the capsule independently with controlled feeding rates. Additionally, the powders must be mixed during feeding, before they enter the capsule, to give a homogeneous mixture with minimal separation of the two powder streams. Significant separation would lead to clusters of parent material in the HIPped state. For this purpose a special feeding device normally employed on other equipment was adapted.

MICROSTRUCTURES

A characteristic feature of the HIPped graded joints is the lack of alloying between the different materials used to produce the graded zone (figure I). Thus, in a mod 9%Cr/316 joint the 25% austenitic composition corresponds to a matrix of mod 9% Cr containing 25% (by volume) of 316 particles and not to a homogeneous alloy containing 11.25% Chromium.

One consequence of this is that the properties found within the joint vary approximately as a linear fraction of their composition. The situation is, however, complicated by the diffusion of carbon from the lower chromium material within the joint to the higher chromium material. This ternary diffusion alters the composition of both materials and can result in large changes in the coefficient of expansion and the creep strength and toughness of the materials. This carbon migration can be seen in figure 2 where the 2¼%Cr had become entirely decarburised with the carbon forming a reaction zone around the exterior of the 316 particle.

MODELLING

Based on the results of dilatometry to determine the linear coefficient of thermal expansion (in the temperature range 20°C-1000°C), tensile testing of the 0.2% yield strength at 500°C, 700 ‘C and 900 “C and limited short term creep testing, a Finite Element Model of the internal stress distribution developed during the fabrication of a graded HIP joint was produced.

This confirmed that high stress developed at areas of rapid change in the coefficient of linear expansion (i.e. abrupt compositional steps). This model was used to optimise the design of the compositional gradient to minimise thermal expansion induced stresses.

It was also necessary to produce a model capable of predicting the extent of decarburisation which could develop in the joints. This consisted of a ternary diffusion model, based on thermodynamic and kinetic principles, which predicts the precipitation and diffusion phenomena occurring in the graded joint due to the different chromium content of the parent material powders. The original model applied to a simple interface and required adaptation to the more complex case of powder particles. Considering the 2?4%Cr/316 HIP joint, areas which correspond to the former austenitic powder particles can be identified within the 2¼%Cr matrix (figure 2). With increasing austenitic volume fraction the distance between austenitic particles reduces and the diffusion path for carbon becomes smaller. Thus, the size and volume fraction of the particles must be quantified and the average distance between particles and the particle size must be taken into account. Due to the small particle size (<500 µm) the diffusion of chromium required to be addressed. It was also necessary to model the different volubility behaviour and differing diffusion coefficients of carbon in ferrite and austenite.

The final model comprised the following steps

- Determination of average distance between particles as a function of volume fraction
- Calculation of chromium diffusion during HIPping
- Determination of initial and boundary conditions
- Solution of some twelve equations describing the diffusion of carbon with the inclusion of an interaction coefficient to take the thermodynamic influence of nickel within the 316 into account.

This work predicted that, in the 2¼%Cr/316 HIP joint, all carbon would diffuse from the 2¼%Cr to the 316 particles. Due to the lower diffusion rate in austenite the carbon did not diffuse into the centre of the austenitic particles but reacted with chromium to form bands of carbides at the interface. This resulted in the decarburisation of the 2¼%Cr matrix, as verified by metallography. This work showed that an equilibrium condition was obtained, after HIPping, and very little change was predicted after the heat treatment, as was again verified by metallography.

The main difference predicted by the model between the 2¼%Cr/316 and 2¼% Cr/mod 9%Cr system arose from the ferritic nature of the mod 9 %Cr. This material combination had a completely different solubility for carbon and a diffusion coefficient the same on both sides of the 2¼%Cr/mod 9 %Cr interface. The model again predicted that the 2¼%Cr matrix would be depleted of carbon and precipitates as was found. This resulted from the steep gradient in carbon activity between 21A% Cr and mod 9% Cr. However, due to its ferritic nature, carbon can easily diffuse within the mod 9%Cr and carbon enrichment and intensive precipitation occurs across the entire mod 9% Cr particle. As in the case of 2^{1}A%Cr/316 the model predicts little change during heat-treatment due to the negligible chromium diffusion which occurs.

Within the mod 9%Cr/316 material combination the gradient in carbon activity across the interface is much smaller than found in the combinations containing 2¼%Cr. As a result the mod 9%Cr matrix is not decarburised and, although some diffusion of carbon to the austenitic particles does occur, the carburisation at the interface is much less than that found for the 2¼%Cr/316 combination. Once again the carbon has achieved equilibrium after HIPping and the only subsequent changes predicted are those resulting from the negligible chromium diffusion which can occur during the post-fabrication heat treatments.

On the basis of this modelling work, the final optimised joint designs were selected. These optimised joints had complex compositional gradients and modified production parameters and metallurgy to avoid, or minimise, the extent of decarburisation during the fabrication of the joints. The carbon diffusion model predicted the establishment of an equilibrium distribution of carbon during the high temperature (>1000°C) portion of the HIPping cycle. As chromium migration is negligible during post HIPping heat treatment and during service at temperatures around 600°C, further decarburisation was not predicted to occur. This was verified by extensive metallography.

TESTING

Extensive testing was carried out during all stages of this programme. As each type of joint was intended to be located in a temperature regime suitable for the lower creep strength material in the joint, the stress rupture testing of the2Y4%Cr/316 and 2¼%Cr/mod 9 %Cr joints was conducted at 550°C whilst the mod9%Cr/316 joint was tested at 600°C.

Toughness and oxidation testing were carried out on samples made from joints with 10 compositional step i.e. the graded zone consisted of 8 distinct compositions lying between the two HIPped parent composition. For example in the2Y4%Cr/316 system 11%, 22%, 33%, 44%, 56%, 67%, 78% and 89% austenitic material were used. Creep and tensile tests were carried out on joints with a range of compositional gradients and the results for the optimised designs, giving the , best tensile strength and creep properties, are included within this report.

TOUGHNESS TESTING

The toughness of 10 step joints in each material combination (2'%% Cr/3 16, 2¼%Cr/mod 9 %Cr and mod 9% Cr/3 16) was evaluated by charpy testing at room temperature. For each compositional band three specimens were tested with the notch located in the centre of the band and the average value quoted.

The toughness results for the 9%Cr/316 system are illustrated in figure 3 which shows the variation in impact energy with composition across the joint. The main feature of this result is that impact energies for the transition zone are not significantly lower than the values obtained for the mod 9% Cr parent material and, close to the austenitic end of the bar, a linear relationship between impact energy and austenitic volume fraction exists.

The toughness results for the 2[1]A%Cr/316 joint is given in figure 4. In this case the impact energy within the transition zone is less than that of either parent material. However, with a minimum value of 84. 3J the toughness is impressive.

The toughness results for the 10-step 2¼%Cr/mod9%Cr joint is shown in figure 5. A surprising feature of this combination is the existence of a local maximum of 107J in the impact energy at 22% mod 9 %Cr. The minimum impact energy of 56J was found at 89% mod 9 %Cr. This compares favorably with typical DMW values of around 40J.

OXIDATION TESTING

Smooth billets, containing both parent and graded transition zones, were produced from HIPped and VPS joints. These were tested at 550°C and 600°C. In addition a further specimen was given 100 cycles comprising a 1 hour ramp from 150 °C to 600°C, a 30 minute hold at 600°C, and a 8½hr ramp back from 600°C to 150°C. These conditions were selected in such a manner that the total time for the 100 cycles equal led the isothermal exposure time. The oxide thickness in each compositional step was obtained by averaging three single measurements. A 10 step design of graded joint was tested in each material combination.

The 2¼%Cr/316 material combination is shown in figure 6. The oxide thickness is shown as a function of composition/distance across the transition joint. The oxide thickness is of the order of 0.02mm (316 parent) to O. 16mm within the graded zone.

Microstructural investigation revealed preferred oxidation of the decarburised 2¼%Cr matrix under all three testing conditions. This produces a relief effect causing the 316 particles to protrude from the joint surface.

The oxide thickness for the mod 9% Cr/316 HIPped joint is shown in figure 7. The oxide thickness is of the order of 0.01mm (316 parent) to 0.06mm (within the graded zone). This is significantly less than found in the21A%Cr/316 combination.

The oxidation behaviour of the 2¼%Cr/mod 9 %Cr material combination is shown in figure 8. The oxide layer is of the order of 0.05mm (mod 9 %Cr) to O. 16mm (2¼%Cr). No significant selective oxidation of mod 9Cr was found during microscopic examination. However, it is very likely that selective oxidation will occur in the longer term as, due to its higher chromium content, the mod9Cr powder particles are much more oxidation resistant than the 2¼%Cr.

It is obvious from the results that thermal cycling did not produce any significant problems and the joints showed oxidation rates typical of those found in the corresponding wrought materials

A number of other more rapidly cycled tests were carried out. In all cases, failure was accompanied by strains in excess of 10% and occurred within the graded zone. One such bottle was subject to hi-axial testing **coupled** with rapid thermal cycling. This bottle was identical to those used in the slower **cycled** test but was pressured to **82MPa** at **650°C.** No **axial** load was applied and the axial stress was therefore 42MPa. The specimen was cycled between 650C and 200 *C by raising the specimen into an **infra-red furnace** and then lowering it into a stream of air. The overall cycle time was 9.5 minutes with a hold time **at** 650 "C of 3 minutes.

The result of this test is also shown in figure 12 where it can be seen that the high rate of cycling has had little if any effect on the life of the bottle. Due to the higher peak temperature used in the rapid cycling test direct comparison is only possible via a parametric approach such as the Larson-Miller relationship

$$P_{Lm} = T.(2O + \log_{10} t)$$

PL_m = Larson - Miller Parameter

T = Temp in °K

t = time in hours

If this formula is used an equivalent life of 1,735 **hrs** at **625°C** can be calculated for the rapidly cycled bottle. The failure in-this bottle with a life which, when corrected to 625*C, **is** virtually the same as that of the more slowly cycled bottles indicates that these mod 9%Cr/316 joints are **highly** resistant to thermal cycling. This was a principal aim of the **programme** and has been **successfully** achieved.

CONCLUSIONS

The main aim of this work was **to** demonstrate that a transition piece, with a **graded** composition along its length, could give improved mechanical properties when compared 'with traditional dissimilar **metal welds** made **with high** nickel fillers. 'Not **only** have the properties of these joints exceeded those of the dissimilar welds they were designed to replace, but-they have also equ-died, or exceeded, the properties of normal like/like welds in the weaker of the materials within the joint. As a result **failure** of such joints will not occur in normal service during the life of the plant. "

The oxidation, toughness and tensile properties of the joints have proved to be equal to those expected of normal wrought products. The hi-axial pressurised and thermally cycled testing has **confirmed** the high resistance of these joints to thermally induced darnage under the most severe of simulated operating conditions.

These joints therefore offer improved joint efficiency particularly for the new 9% and 12 **%Cr** creep resistant steels currently being developed for use at high temperatures. They offer a **ferritic/austenitic** joint capable of sustaining continued thermal cycling during service in the creep range. It is believed these joints can make a contribution in the design of new power plants with improved efficiency.

ACKNOWLEDGEMENTS

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OPTIMISED JOINTS MECHANICAL TESTING

As in service these joints would be welded onto normal wrought materials, not only the graded joint but also the wrought/HIP weldments were fully tested. In addition, testing was carried out on the parent HIP bar in the appropriate heat treatment condition specified in Table 1, for comparison with normal wrought materials.

TENSILE RESULTS

In all three materials the parent HIP properties were generally comparable, though slightly below those typical of the corresponding wrought material. Both the parent HIP material and the wrought/HIP weldments exceeded the minimum tensile values specified for that material in BS 3059 part 2.

Surprisingly all wrought/HIP cross-weld tensiles failed in the wrought parent material even when tensile testing had shown this to be stronger than the HIPped material to which it was welded. It appears that the HIPped material suffers a less pronounced decrease in properties in the HAZ than is found with normal wrought materials.

Testing of the graded joints demonstrated that, in all three material combinations, the strength of the joint exceeded the minimum properties (specified in BS 3059 part 2) for the weaker material in the joint.

CREEP RUPTURE TESTING

The results of the creep rupture testing in 2'A% Cr/3: 6 and 2¼%Cr/mod 9%Cr are shown in figures 9 and 10 respectively. For the2Y&%Cr/316 joint the lines representing the creep strength of the joints and the wrought/HIP terminal welds are parallel and the separation is less than the scatter in the data such that the points could be represented by a single line. In the 2¼Cr/mod 9 %Cr joints the creep strength at the joint is significantly higher than that of the wrought/I-HP weldment at times comparable with service life (> 100,000hrs). In both cases the controlling factor was not the creep strength at the joint itself but rather that of the least creep strong (i.e. 2¼%Cr) terminal weld. During rupture testing of this 2¼%Cr wrought/HIP weld failure always occurred on the wrought side of the weldments. Thus this weld is at least as strong in creep as normal 2¼ % Cr/2¼ % Cr weldments and in service failure of these graded joints would not occur.

Due to the high interest in the mod9Cr/316 material combination additional testing was undertaken on this material combination. In addition to the creep rupture testing at 600°C, two series of tests were carried out at 8 lMPa and 59MPa over a range of temperatures. Such iso-stress tests give straight line plots of test temperature against creep life which can be extrapolated from test temperature to service temperature. This gave estimated creep lives at 600°C of 32,700hrs at 82MPa and 194,614hrs at 59MPa.

These values are plotted together with the results of testing at 600°C in figure 11. Whilst the iso-stess estimates lie below the predicted line based on the 600°C testing they are within the scatter of the data and confirm the good performance of the joint out to 200,000hrs.

Once more the limiting factor on the creep strength of the mod 9%Cr/316 joints is not the creep life of the joint itself but that of the lower creep strength (mod 9 %Cr) terminal weld. During testing of this wrought/HIP mod 9% Cr weld failures had occurred in the H.A.Z. of both the wrought and the HIP materials. Analysis of the development of the strain distribution across these weldments was accomplished by depositing a chain of gold markers along the gauge length of the specimens. Measurements were taken before the start of the test, during tests interrupted at 80-90% of their life and after failure. These confirmed that failure occurred in the type IV position in both wrought and HIP material but no differences in the rate of strain accumulation at the type IV position were found. It was concluded that the creep rupture lives of the wrought/HIP weldments were identical to those of normal mod 9% Cr/mod 9 %Cr weldments.

THERMAL CYCLING TESTS UNDER BI-AXIAL LOADING CONDITIONS

The aim of these tests was to check the performance of mod9%Cr/316 joints under the combined effects of hi-axial stress and thermal cycling. In the tests short tubular specimens containing the compositional gradient were pressurised to produce a calculated hoop stress in the specimens. The specimens acted as models of graded joints in typical tube sizes used in power station water walls. They were subjected to additional factors such as end loads (to simulate system stresses in the water walls) and thermally cycled to simulate the thermal transients which could be expected in power station operation.

Values of applied hoop stress were calculated using the thin cylinder formula

$$\sigma_{hoop} = \frac{PD}{2t}$$

P = Internal Pressure

D = Internal Diameter

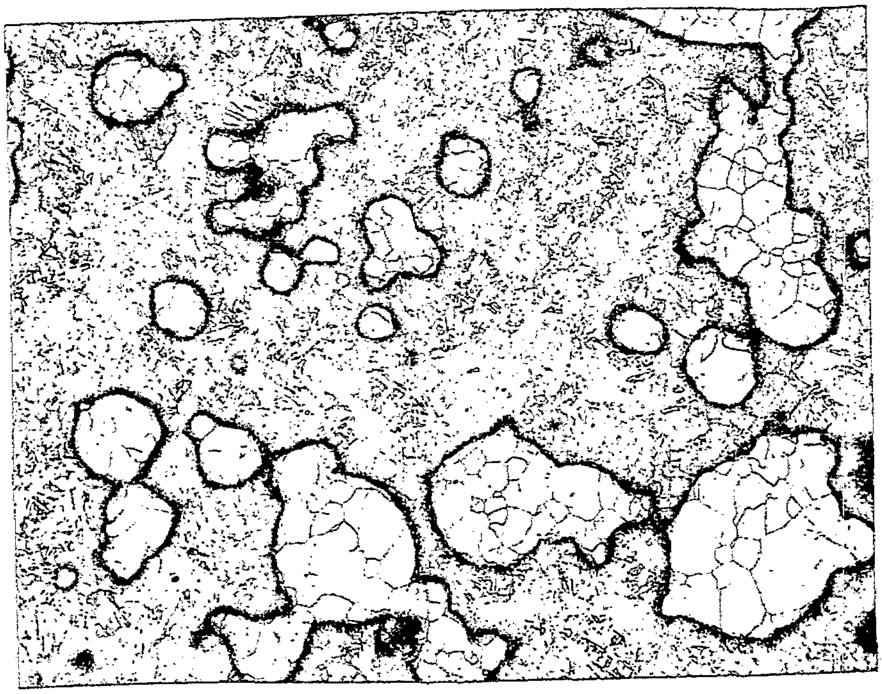
t = Tube Thickness

Initially a 12 hour thermal cycle was adopted. This consisted of a four hour heating period from 230°C to 625°C, a four hour hold at 625°C and a four hour cooling period to 230°C. Due to the delayed start of this testing it appeared that insufficient cycles would be amassed during the programme and after 58 off 12 hour cycles the system was altered to give an eight hour cycle consisting of a 2 hour heating period from 230 to 625 ‘C, a 2 hour hold at 625°C and a 4 hour cooling period to 230°C.

The hoop stress used in these tests was 82MPa. A small program of uni-axial iso-stress testing had also been undertaken on this joint design at the same stress and the results are compared in figure 12. As can be seen the hi-axially tested and thermally cycled bottles failed before the predicted time based on the uni-axial testing. This was expected for three reasons. Firstly, the more complex hi-axial stress field would induce creep damage faster that a simple uni-axial stress field. Secondly, the endurance quoted in the bottles corresponds only to the time at peak temperatures whilst creep damage will continue to accumulate at elevated temperatures during the heating and cooling cycles. Thirdly, thermally induced stresses, produced by thermal cycling, will lead to the accumulation of additional damage.

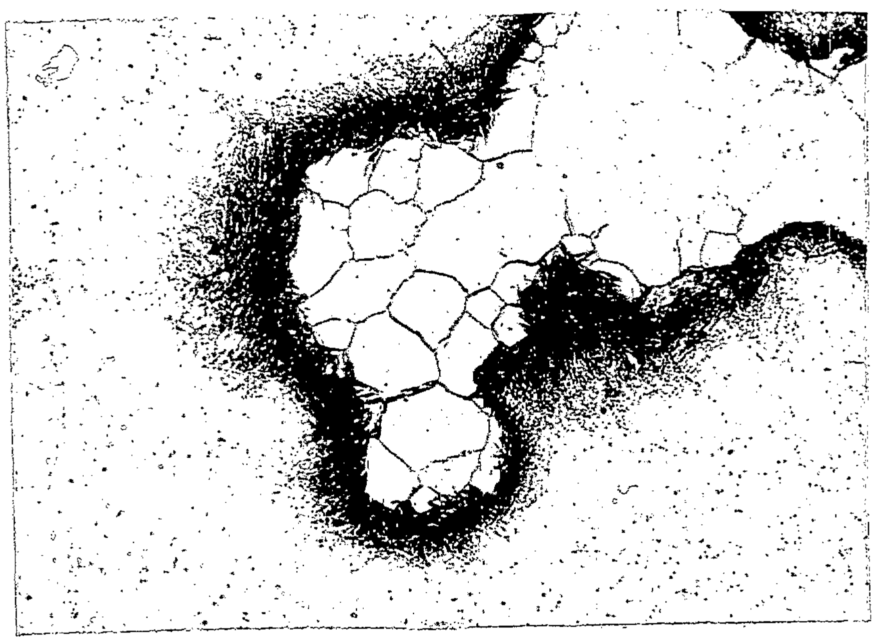
Metallography confirmed that the failure location was not adjacent to the parent HIPped 9 %Cr, as found in the uni-axial testing, but moved to a location well within the graded zone where there were similar proportions of mod 9% Cr and 316. Ductility had been decreased by the imposition of a complex multi-axial stress field and cracking had developed at mod 9%Cr/316 particle interfaces over a large portion of the graded joint. Despite this, strains of 10-20% developed prior to failure on these bottles. Both bottles failed as a result of the development of a through thickness crack within this bulged area.

HEAT TREATMENTS			
MATERIAL	NORMALISE	TEMPER	PWHT
2¼%Cr -9 **%Cr** Graded Joint	**960°C/1** hour/AC	**750°C/2** hours/AC	**720°C/2** hours/AC
9%Cr -316 Graded Joint	**1050°C/1hour/AC**		**740°C/2** hours/AC
2¼%Cr -316 Graded Joint	**960°C/1** hour/AC		7200(2/2 hours/AC
Hipped Parent316	**1100°C/1** hour/AC	None	None
Hipped Parent **2¼%Cr**	**960°C/1** hour/AC	**750°C/2** hours/AC	None
Hipped Parent 9% **Cr**	**1050°C/1** hour/AC		None
Hipped/Wrought Weld **2¼%Cr**	**960°C/1hour/AC**		**720°C/2** hours/AC
Hipped/Wrought Weld **9%Cr**	**1050°C/1** hour/AC		**740°C/2** hours/AC
Hipped/Wrought Weld 316	**960°C/1** hour/AC	**750°C/2** hours/AC	**720°C/2** hours/AC



MICROSTRUCTURE Mod 9%Cr/316 HIP JOINTS
COMPOSITION 25% 316

Figure 1

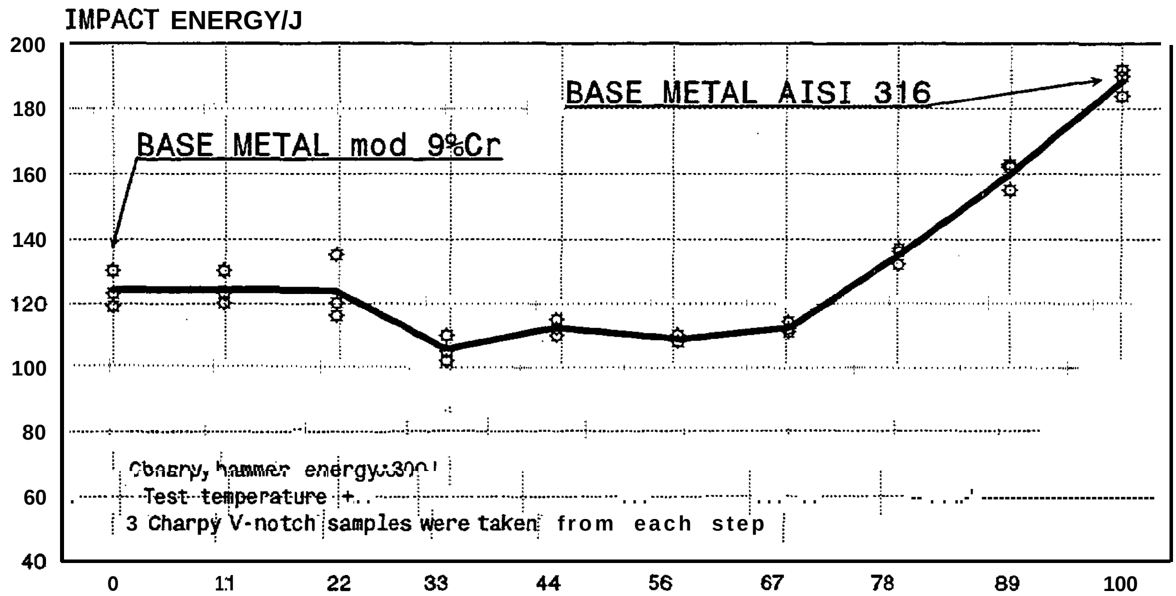


MICROSTRUCTURE OF $2\frac{1}{4}$Cr/316 HIP JOINTS
REACTION ZONE AT $2\frac{1}{4}$%Cr/316 PARTICLE INTERFACE

Figure 2

TOUGHNESS TESTS-IMPACT ENERGY mod 9Cr/316 HIP JOINTS

(HEAT TREATMENT - 1050ºC/1hr/AC,750ºC/2hr/AC,740ºC/2hr/AC)

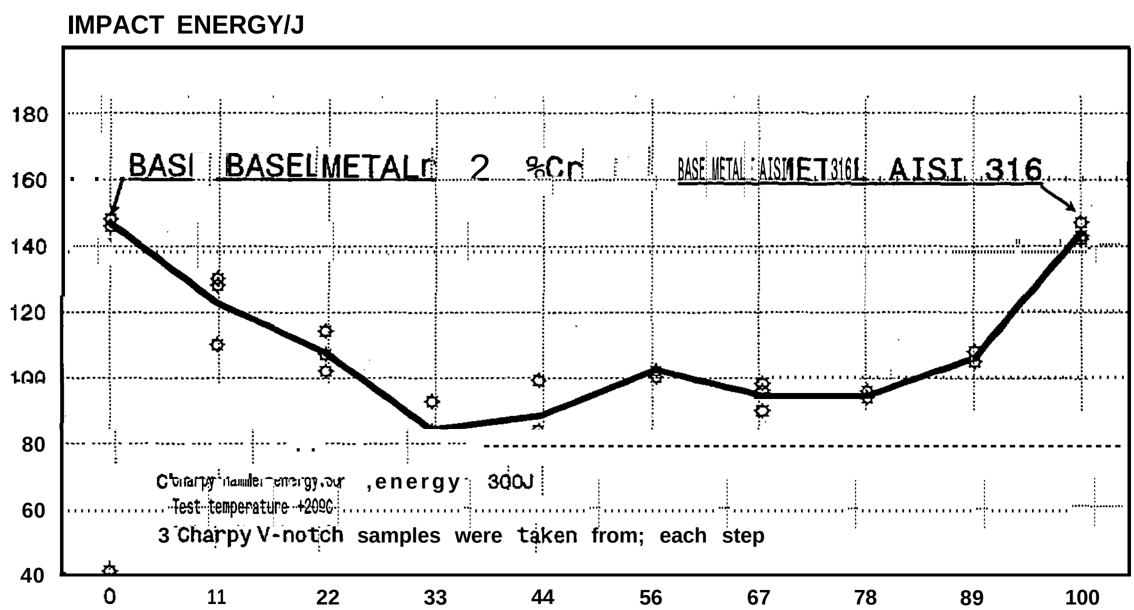


	0	11	22	33	44	56	67	78	89	100
SPECIMEN 1 ☼	130	130	135	105	112	110	114	132	155	192
SPECIMEN 2 ☼	123	123	120	110	115	108	112	137	163	190
SPECIMEN 3 ☼	119	120	116	102	110	108	111	136	162	184
AVERAGE ▬	124	124.3	123.7	105.7	112.3	108.7	112.3	135	160	188.7

Figure 3

TOUGHNESS TESTS- IMPACT ENERGY 2¼Cr/316 HIP JOINTS -

(HEAT TREATMENT - 960ºC/1hr/AC,750ºC/2hr/AC,720ºC/2hr/AC)



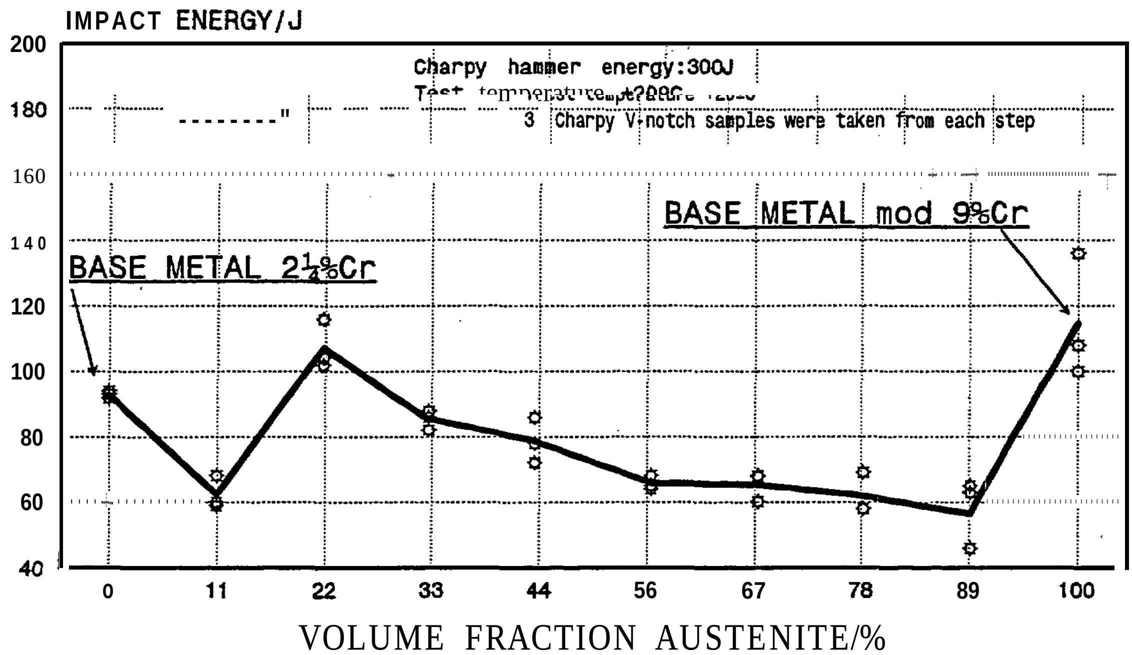
VOLUME FRACTION AUSTENITE /%

	o	11	22	33	44	56	67	78	89	100
SPECIMEN 1 ☼	146	110	i 07	93	84	102	90	94	105	142
SPECIMEN 2 ☼	41	130	102	76	82	100	96	96	105	143
SPECIMEN 3 ☼	148	128	114	84	99	102	98	94	108	147
AVERAGE -	147	122.6	107.6	84.3	88.3	102.3	94.6	94.6	106	144

Figure 4

TOUGHNESS TESTS-IMPACT ENERGY $2\frac{1}{4}$Cr/mod 9%Cr HIP JOINTS

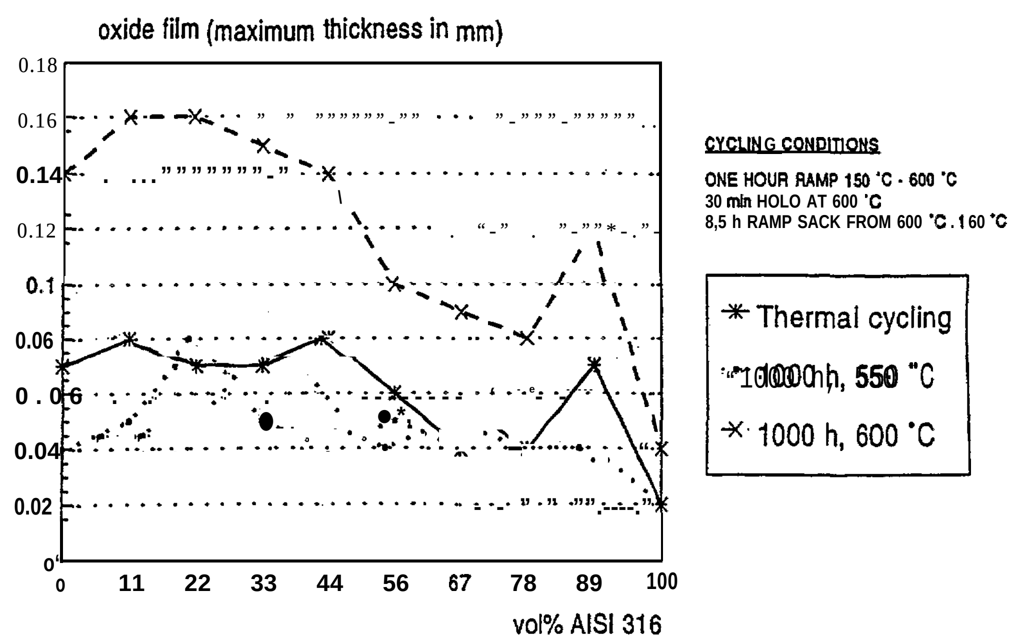
(HEAT TREATMENT - 960ºC/1hr/AC,750ºC/2hr/AC,720ºC/2hr/AC)



	0	11	22	33	44	56	67	78	89	100
SPECIMEN IV	92	59	116	86	78	65	68	58	65	108
SPECIMEN 2 ✡	94	60	i 02	88	86	84	88	58	63	136
SPECIMEN 3 ✡	93	68	104	82	72	68	60	69	46	100
AVERAGE"	93	62.3	107.3	85.3	78.6	65.7	65.3	61.7	56.3	114.7

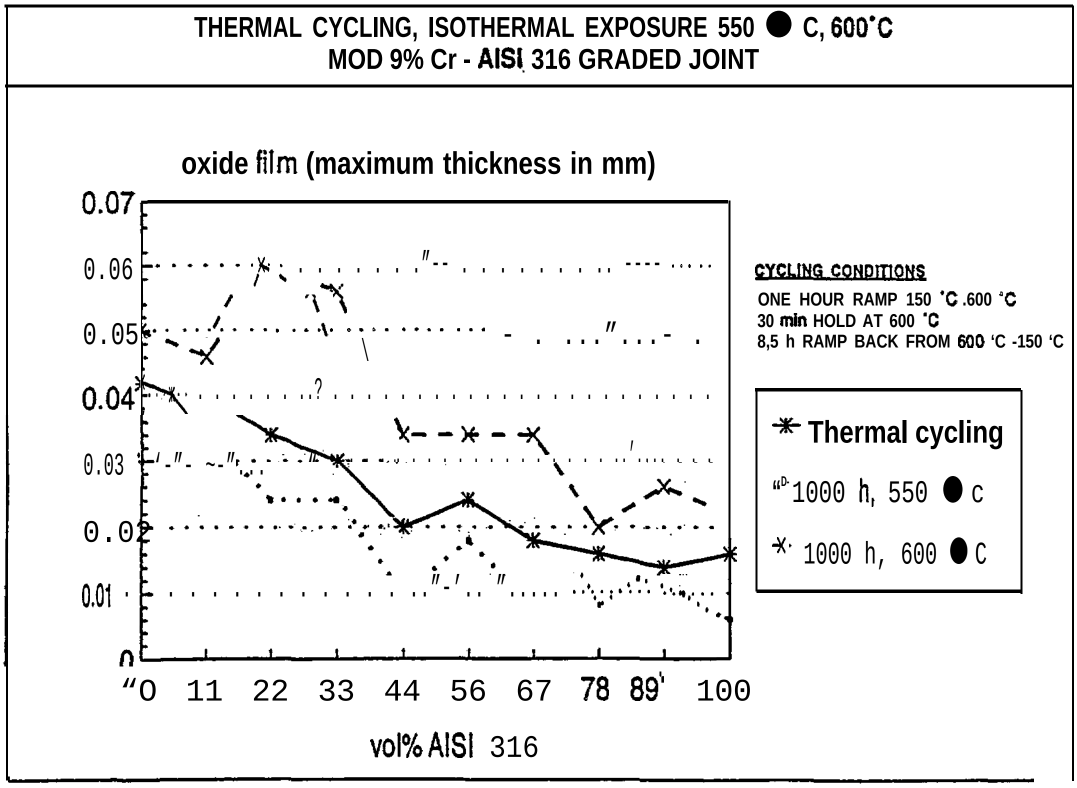
Figure 5

THERMALCYCLING, ISOTHERMAL EXPOSURE 550 °C, 600°C
2,25 % Cr - AISI 316 GRADED JOINT

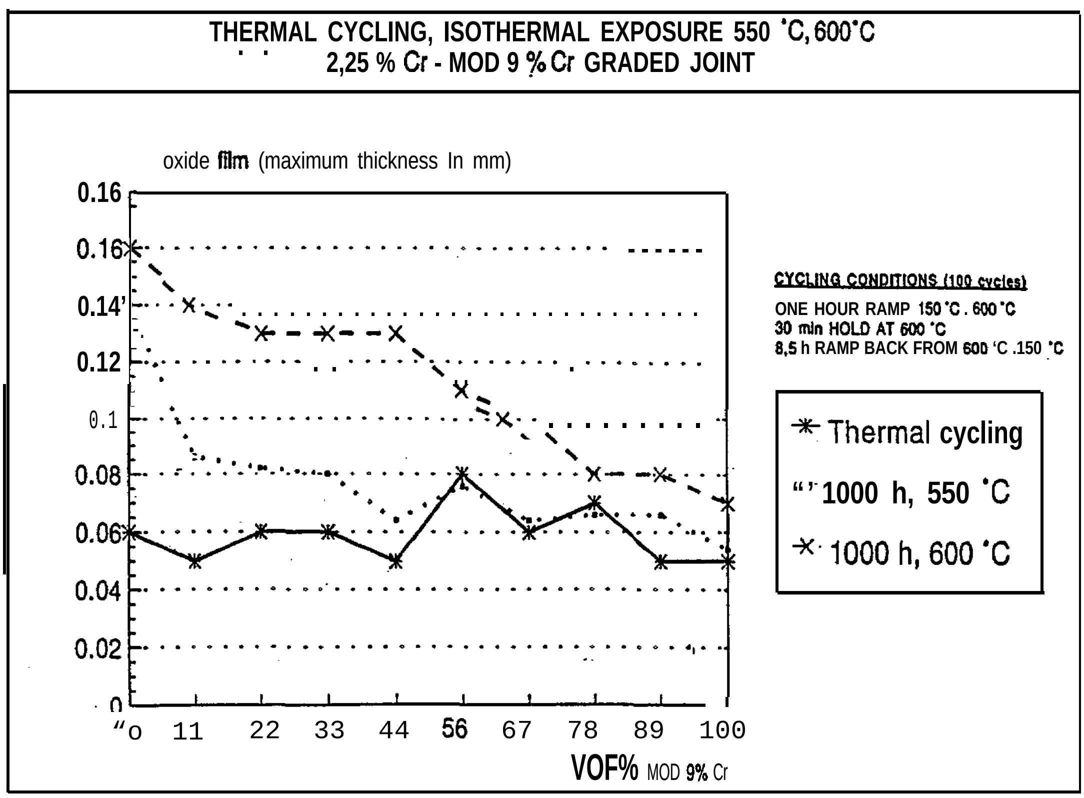


OXIDE THICKNESS 10-STEP $2\frac{1}{4}$Cr/316 HIP JOINTS

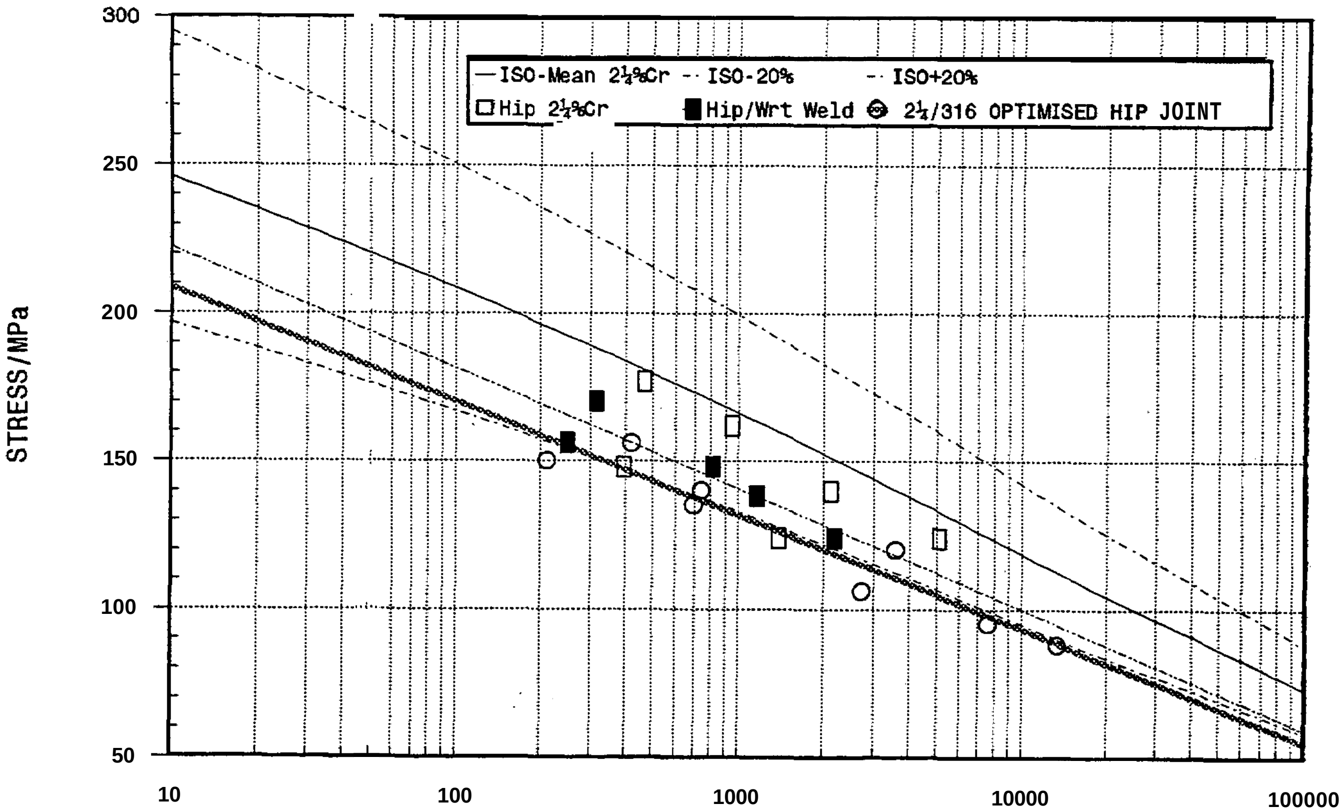
Figure 6



OXIDE THICKNESS IO-STEP **MOD 9%Cr/316** HIP JOINTS **Figure 7**



OXIDE THICKNESS 10-STEP **$2\frac{1}{4}$Cr/MOD 9%Cr HIP JOINTS** **Figure 8**



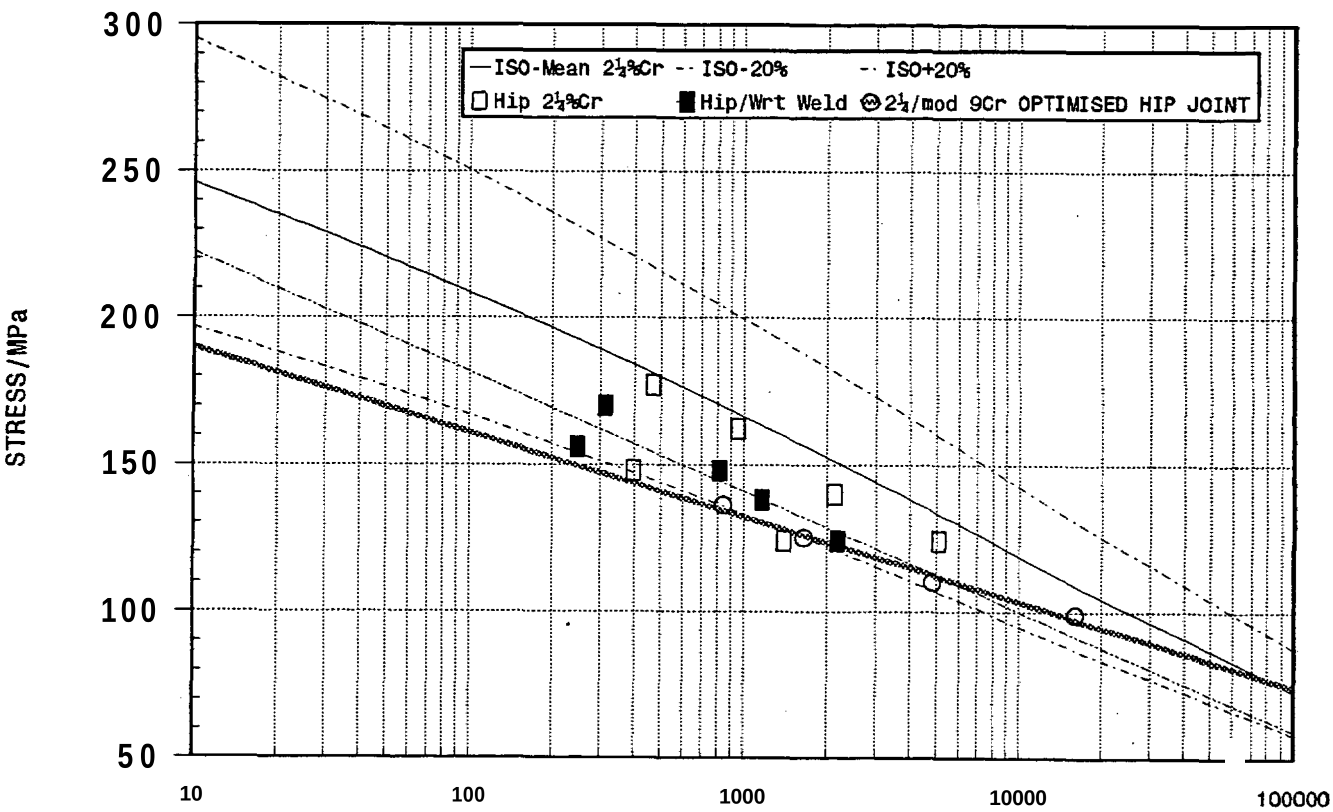
ISO-DATA from PD6525

TIME/hrs

CREEP DATA 2¼Cr/316 TRIAL HIP JOINTS

TEMP 550°C

Figure 9



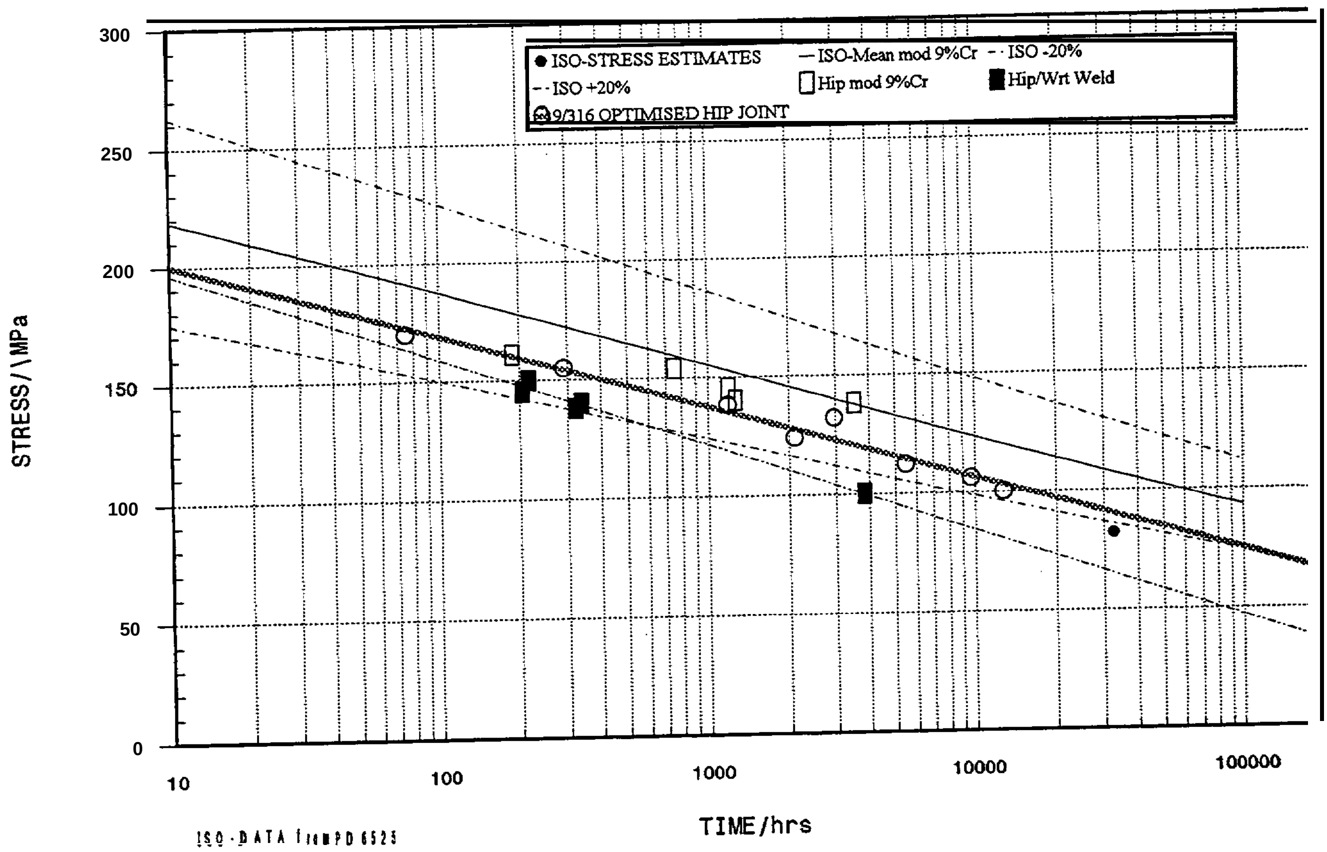
ISO-DATA from PD6525

TIME/hrs

CREEP DATA 2¼%Cr/mod9Cr TRIAL HIP JOINTS

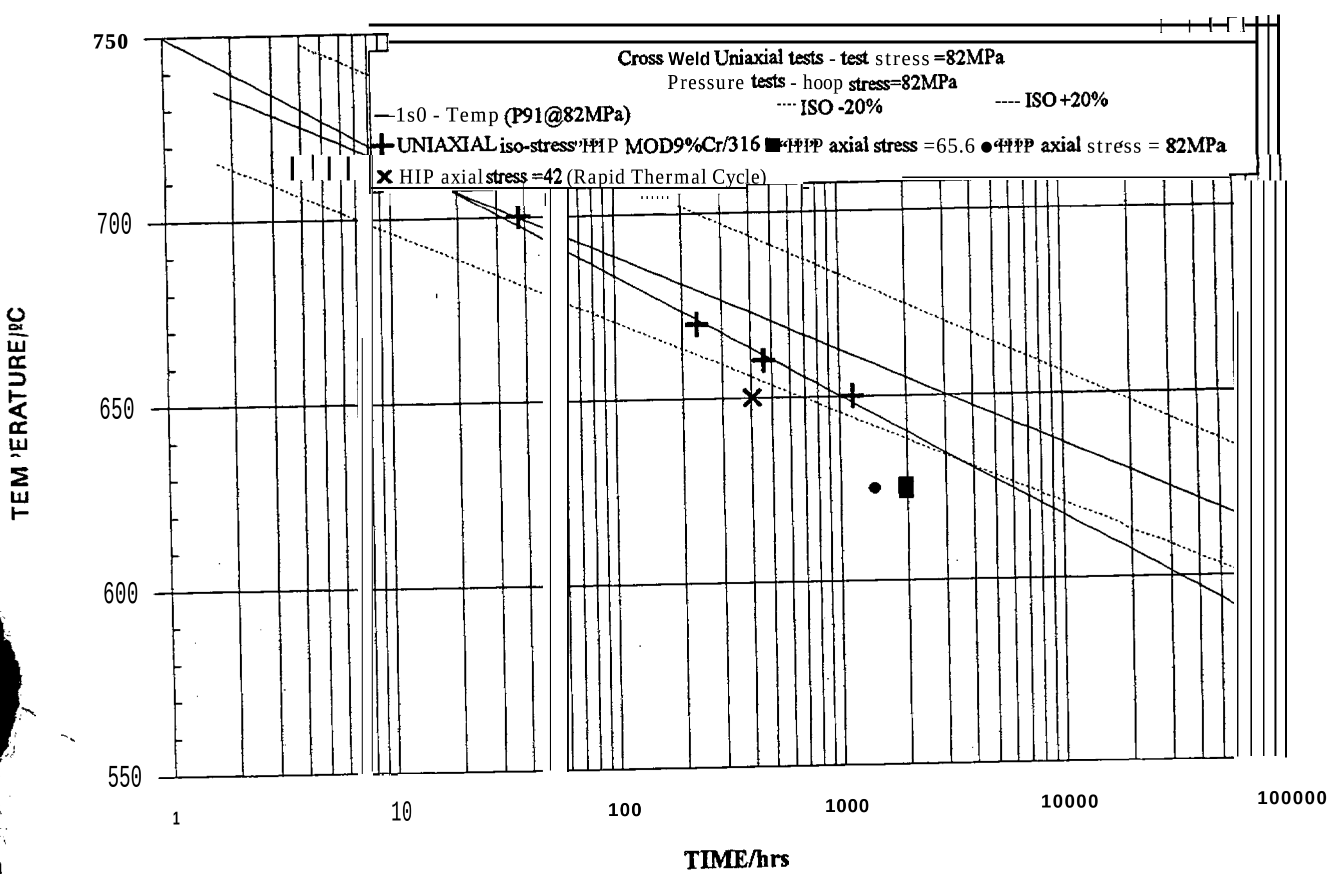
TEMP 550°C

Figure 10

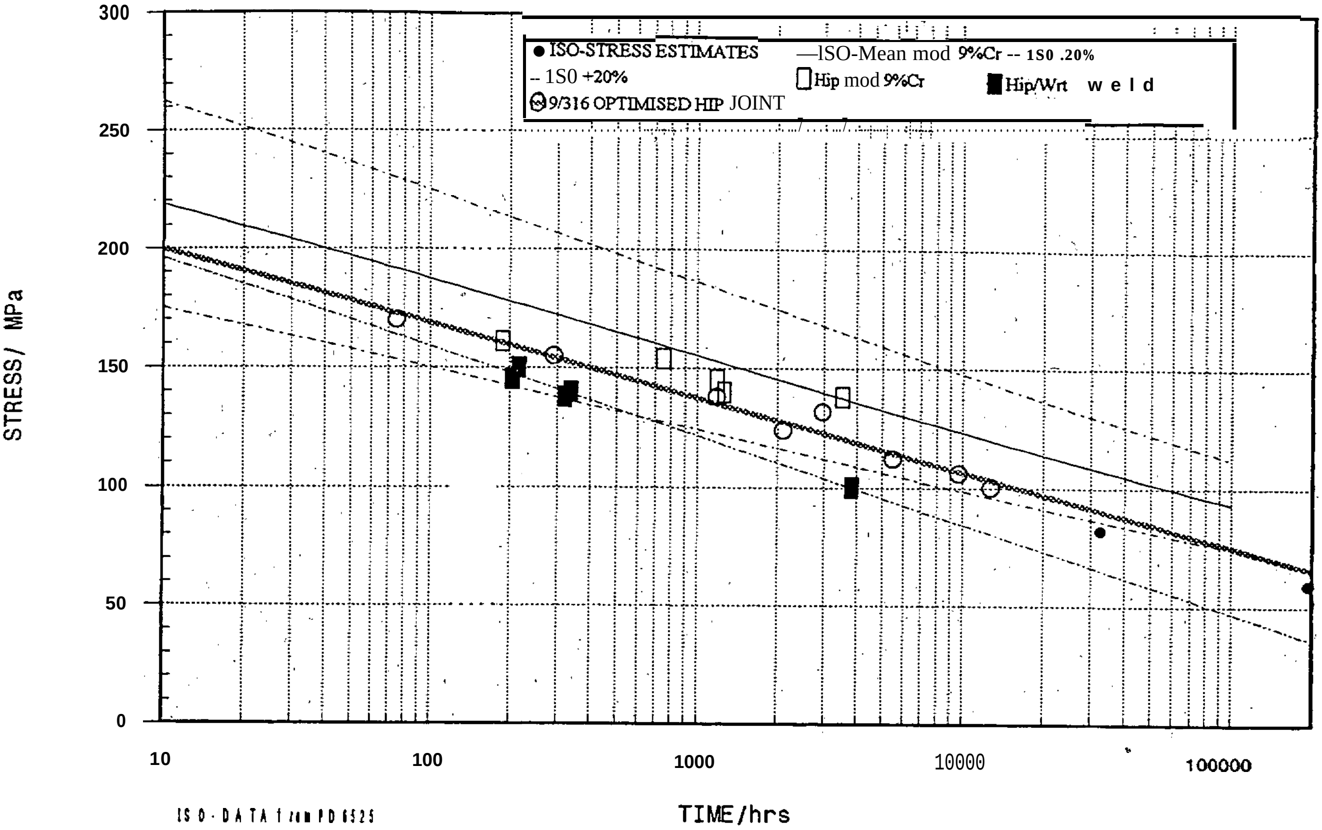


CREEP DATA HIP TRIAL JOINTS mOD 9%Cr/316
TEMP 600°C

Figure 11



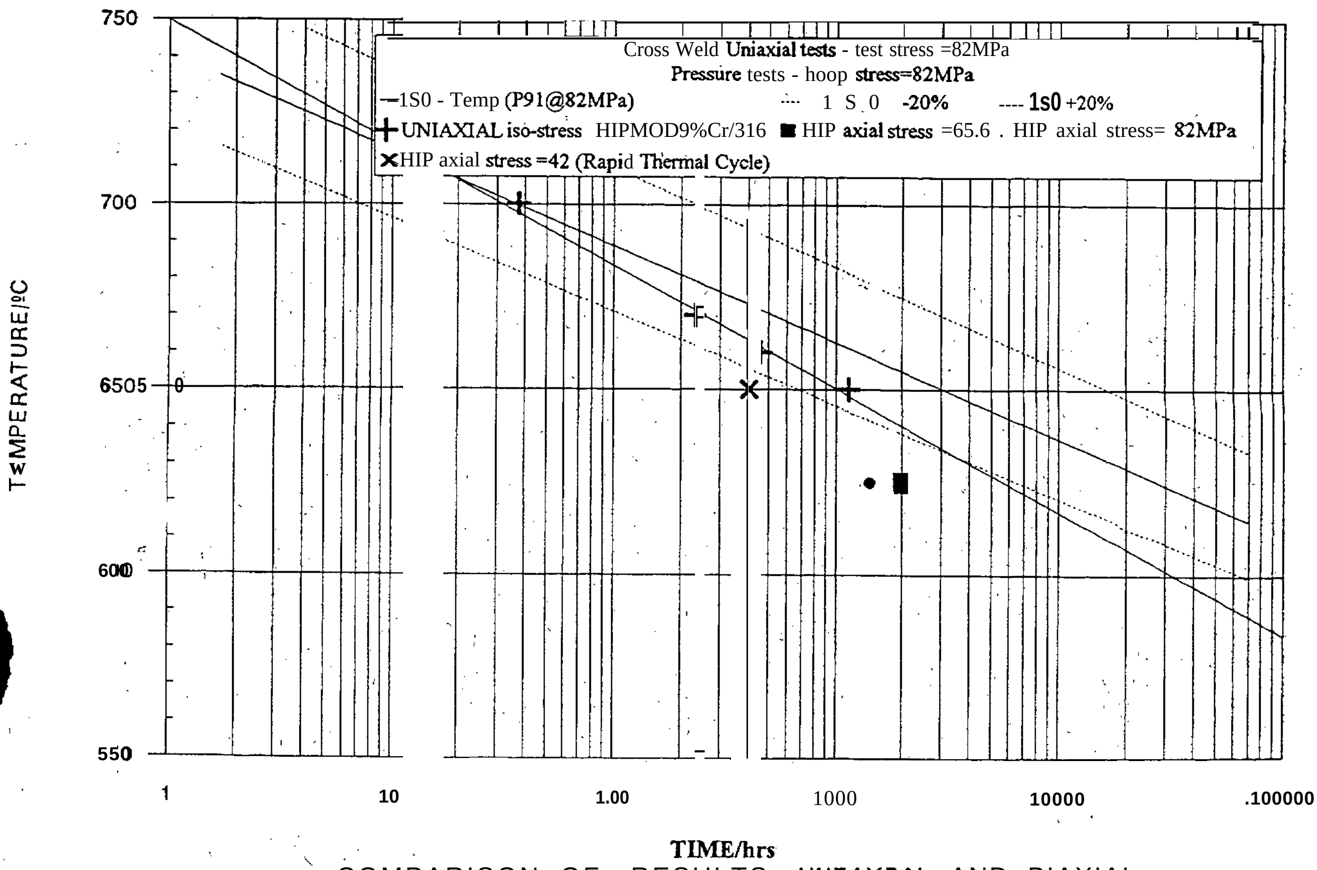
COMPARISON OF RESULTS - UNIAXIAL AND BIAXIAL
THERMALLY CYCLED mod 9%CR/316 STEPLESS TRANSITION JOINT **Figure** 12



CREEP DATA HIP TRIAL JOINTS mOD 9%Cr/316

TEMP 600°C

Figure 11



COMPARISON OF, RESULTS -UNIAXIAL AND BIAXIAL

THERMALLY CYCLED mod 9%CR/316 STEpLESS TRANSITION JOINT

Figure 12