

**The History of Hydrogen Induced Stress Cracking (HISC) Failures Due to
Cathodic Charging of Duplex Stainless Steels (DSS)**

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ABSTRACT

Between 1975 and 1995 roughly half a million kilometers of duplex stainless steel (DSS) pipe had been installed in the North Sea, subsea, with insulation coating and cathodic protection (CP) applied. In contrast to the previous 20 year of good experience, between 1996 and 2004 a cluster of subsea failures of new and relatively newly installed DSS assets occurred. These failures were attributed to HISC as a consequence of CP. The paper reviews the available literature detailing a number of case histories and presents some additional anecdotal information not previously reported. Some similarities between these failures and a cluster of HISC failures of martensitic stainless steel pipelines that occurred shortly after the first DSS failures are detailed. Current methods of mitigation such as those detailed in design codes, the use of hot iso-statically pressed production methods, surface treatments and modified alloys with improved HISC resistance are discussed. The paper considers issues surrounding analysis of the problem and changes in practice that may further reduce the risk of HISC failure in the future.

Key words: HISC, Duplex Stainless Steels, Cathodic Protection, Hydrogen Embrittlement

INTRODUCTION

As long ago as 1973, design codes¹ considered the possibility of hydrogen embrittlement due to CP. Between 1986 and 1995²⁻⁴ the failure of DSS fasteners subjected to CP were reported. These were associated with high ferrite levels in the steel (approximately 70%) combined with precipitation hardening at 475°C to give the high levels of strength desired for fastener applications.

At the same time, the susceptibility of DSS welds to hydrogen embrittlement had been reported⁵. Just like the fastener failures, the hydrogen cracking of welds was associated with high ferrite levels (70%), highly restrained joints and in the case of welds, high levels of diffusible hydrogen.

In 1987 a UK operator investigated the crack susceptibility of 22% Cr (UNS S31803) DSS. This related to 2 x 7km lengths of 6" (150mm) x 12.7 mm wt 2205 duplex stainless steel flow lines for a project in the North Sea⁶. Interestingly in 1989, this work resulted in the deployment of Schottky diodes on the pipeline and the use of very high integrity insulation coating system. The diodes were placed in series with the CP connections on to the two flow lines. The diodes limited the potential of the pipes in seawater to between -450 mV and -800mV (Ag/AgCl), such that the pipe was in the passive range, but more positive than the potential for significant hydrogen evolution. This was the first time that such a system had been used in the North Sea and was some 6 years prior to the first HISC failure of DSS.

The decision to deploy the diodes may have been influenced by the failure in 1985 of 5 joints of 160ksi, cold worked duplex stainless steel downhole tubulars Figure 1.



Figure 1: Cracking in the 22% Cr DSS Cold Worked Tubulars⁷

The tubes had been in service for 3 years and had suffered cracking when pulled from the well and dropped on to the deck. It was noticed that the crack length slowly grew while the joints were lying on the deck (such delayed cracking behavior being indicative of hydrogen embrittlement). It was concluded that the failure mode was indeed hydrogen embrittlement due to cathodic charging with hydrogen as a consequence of corrosion of the carbon steel casing⁷. Improved inhibition of annulus fluids and better tube handling procedures allowed the continued use of the 3,300m of tubing successfully.

The issue of HISC had also been the subject of research^{8,9}. By 1995 the generally held view was that DSS were susceptible to HISC and hydrogen embrittlement (at the time the distinction between these two modes of failure lacked clarity) but, the cracks did not initiate until stresses exceeded the yield strength and that the cracks did not grow until stresses were approaching the UTS of the material. It was believed that the austenite content of these steels was influential in arresting HISC cracking because of its tough FCC crystal structure and its high solubility and low diffusivity of hydrogen atoms. This was based on data from slow strain rate, interrupted slow strain rate, ripple load, double cantilever beam fracture mechanics and corrosion fatigue with CP testing¹⁰. There was also good field experience from DSS pipelines, flow lines, manifolds, pump column pipes and solution annealed fasteners operating subsea with CP dating back to 1975¹¹.

A CONNECTOR FORGING FAILURE

In 1996 one of two 200 tonnes production manifolds (Figure 2a) suffered a leak after spending 6 months on the seabed, subjected to a CP potential of $-1050\text{mV (SCE)}^{12,13}$. The leak occurred after a prolonged period of successful subsea hydro-testing, where pressure had been retained. (This delayed cracking, again is indicative of hydrogen embrittlement). The manifold was lifted from the seabed. It was found that two proprietary pull in connectors (hubs), located at one corner of the manifold had suffered cracking. What is not commonly known is that during installation this manifold was dropped from a short height and in a skew position, landing on to the seabed first on one corner. This was the corner where the two hubs that subsequently cracked were located. The incident was not considered serious at the time it happened, but it did result in the shearing off of a carbon steel stinger from the structure.

The hubs had each been made from 25%Cr (UNS S32760) material. A 250kg piece weight billet was upset forged to form “top hat” shape. A total of 186 forgings of this type were deployed in various locations in the 24 well developments. (The procedure used to make the forgings had been prequalified and passed all the customer specification requirements).

However, a late design change added 50mm of material to the heavy end of the top hat. This was to allow for subsequent machining of pipe nipples that would act as butt weld connections to which connecting pipes could be attached by welding Figure 2b. The qualification was not repeated for the larger forging due to time restraints. The failure investigation found cracking in two of the butt weld nipples that had been machined into forgings. Because the forgings were so large, the austenite spacing was high ($100\mu\text{m}$) and because the forgings had been upset, the direction of grain flow in the butt weld nipples was transverse, across the thickness of the nipple rather than longitudinal, along the length of the nipple, as would be the case for an extruded pipe. There was no evidence of plastic deformation around the cracking. Fractography showed that the cracks were brittle in nature. The cracks were not associated with the pipe to hub butt welds.

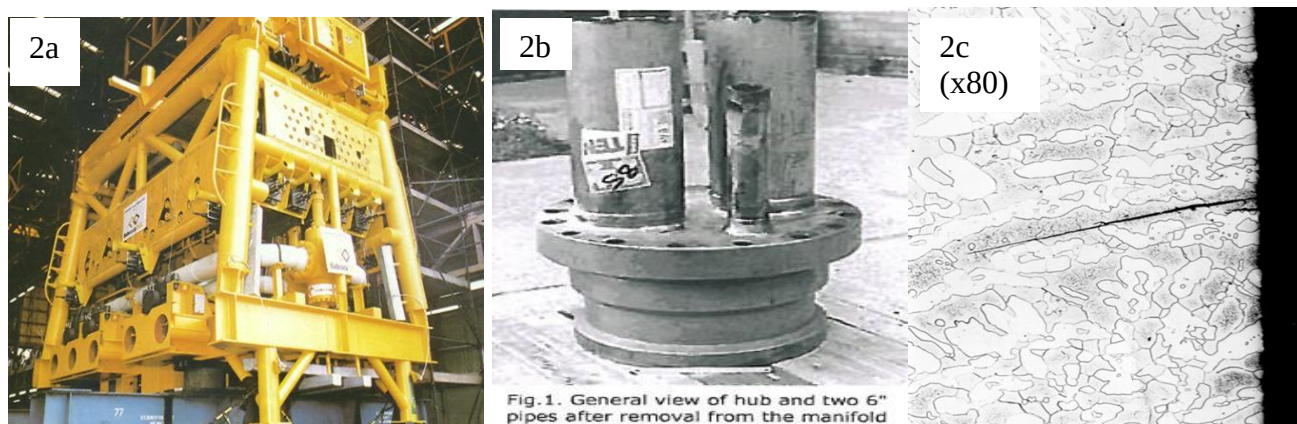


Figure 2a: the manifold, b: the forged connector, c: the microstructure of the nipple showing cracking through the ferrite and chromium nitrides inside the ferrite grains¹²

The cracking ran through the thickness of the nipple, down large ferrite grains and was uninhibited by austenite because the nipples had been machined across the grain flow of the forging. The ferrite grains in the forgings were also found to contain many chromium nitride precipitates (Figure 2c). At the time these precipitates were not considered to have contributed to the cause of the failure.

After a prolonged and technically challenging investigation the metallurgists involved were able to reproduce the failure mechanism and prove that the cracking was due to hydrogen generated by the CP system. Further, the threshold stress/strain for the initiation of HISC was determined using constant load tensile specimens and this was proved to be applicable to hub forgings by several full scale tests.

The term hydrogen induced stress cracking (HISC) was used to describe the failure mechanism rather

than hydrogen embrittlement because of the highly localized nature of the degradation and the fact that the bulk toughness of the forgings was not impaired. It was found that under constant load conditions the strain generated in the material could increase with time through a mechanism referred to as cold creep. This phenomenon was found to be important with respect to both initiation and propagation of HISC. It was conservatively established that a threshold total strain (i.e. the sum of strains due to residual, applied, fabrication and installation stresses) of 0.5% was required for the onset of HISC (note, the strain at failure was estimated to be in excess of 3.5%). From this, the mechanical engineers were able to consider the load cases on individual forgings in the field and establish if the threshold strain would be exceeded or not.

Other counter measures included improvements in the type and extent and degree that insulation coating systems were applied to “shield” the material from CP and encapsulation of nipple areas in polyurethane. The initial investigation noted that the coatings were, in some cases, already dis-bonded, even before the manifold had gone into service and were serving no function at all. This analysis and remedial action gave the operators the confidence to start production using the 194 existing hub forgings along with replacements for the 2 that failed. To date there has been no re-occurrence of the problem on this project.

During the investigation, it was found that pre-straining then unloading of the material prior to exposure to CP had the effect of preventing low temperature creep on re-loading under CP. This improved the HISC threshold. Incorporation of this into hydro test procedures could be very beneficial. Also, the investigators had to postulate high levels of residual stress in the forgings in order to realize cracking thresholds. In the analysis the element of loading associate with dropping the manifold was taken only as the force required break off the stinger. But this would be the minimum force experienced by the hubs and may be an under estimate. Any real difference would offset the need to assume very high residual stresses in the component due to manufacture.

A PIPELINE FAILURE

In 1997 a 22% Cr DSS 6” (150mm) NB x14mm thick, seamless pipe leaked¹¹. The pipe was on the seabed and had been in service for 4 years, subject to CP potentials of -900 to -950mV (SCE) and temperatures ranging from 4°C to 100°C (as a consequence of multiple startups and shutdowns). These thermal cycles had caused expansion spools at the manifold end to stretch and at the platform end to concertina, and the pipeline had moved along the seabed (thermally walked) some 2 to 3 meters. One expansion joint was ineffective because it was fouled by workover cuttings that had been dumped on it. The leak was detected during a startup. It was located at the toe of a fillet weld that joined the “standoff” anode double plate to the pipeline. This joint was considered to be metallurgically acceptable but highly restrained. Finite element analysis indicated tensile stresses in bending of 450MPa around the connection. The coating system applied was epoxy phenolic and was blistered over large areas. The pipe material had a fine grain structure containing 48% austenite. The microstructure of the fillet weld was also considered to be satisfactory. Cracking was found to run through ferrite grains but was arrested or its path was diverted by austenite. Strains at failure were estimated to be 2 to 3%. The cracking was found to be brittle in nature. Diffusible hydrogen determinations were made. At the surface of the pipes in areas of coating disbondment hydrogen levels of 29ml/100g of metal were measured. The bulk hydrogen content of the pipe in these areas was 4ml/100g of metal. In areas where the coating was intact it was 2ml/100g of metal. HISC was considered to be the cause of failure.

A MANIFOLD PIPE SPOOL AND METHANOL LINE FAILURE

This was a plinth mounted 5” (127mm) x 15.9mm wt. 22% Cr DSS subsea manifold and associated flow lines^{11,14}. It had experienced temperature cycles between 4°C and 100°C on 60 occasions due to start up’s and shut downs. The pipes were fusion bond epoxy coated and subjected to CP. Surveys indicated a CP potential less than -1V (Ag/AgCl) in the area. Five tie-in spools and flanged saver spools

connected the manifold and flow lines. The saver spools had a 20mm thick, 22% Cr duplex wrapper plate's fillet welded on to them and these were joined by welding to saddle type supports. These spools ran down from the plinth and were tied in to trenched flow lines. The trench was backfilled by rock dumping. The project also utilized a 1" (25.4mm) diameter methanol line, also built in 22%Cr DSS. In 1998, 2 months after start up, the methanol line leaked. 2 months later, after intermittent production, a leak was detected in the primary pipe spools on start up.

It was found that the location of the seabed had been misjudged and the foundation support was less than the design case. Also, the tie-in spools had been cut short and as such were unsupported. The intended 600mm depth of rock dumping was 3 meters deep. This rock dump loading caused plastic deformation of the tie in spools, and all of them were bent downwards. Stress analysis indicated that the tie in spools may have been subjected to 10% strain.

A failure investigation revealed that the microstructure and properties of the pipes and plates used were good. Both primary and secondary cracking was found in the fillet weld. The primary crack was in the cap of the fillet weld joining the wrapper plate to the pipe, close to, but not at, the toe of the weld on the pipe side of the joint. The crack mouth had opened up indicating high stress levels. Figure 3c shows evidence of significant plastic straining, opening and blunting the crack, but with continued crack propagation out of the plane of bending and running down the grain flow of the parent pipe. The fillet had an unusually smooth appearance; indicative of a cosmetic weld repair (Figure 3b). Examination of the fillet weld cap microstructure (Figure 4) showed it to have a coarse grained columnar structure, containing 70% ferrite. Secondary cracking across the columnar grains was also found. Its step wise appearance was more like conventional hydrogen cracking in weld metal than HISC. Also, in some areas of the pipe, insulation coating was found to be intact and covering over some of the secondary cracks. In other areas coating residue was found on the sides of the fracture faces of cracks. This indicated that they occurred during fabrication, prior to insulation coating. So their initiation was not associated with hydrogen due to CP.

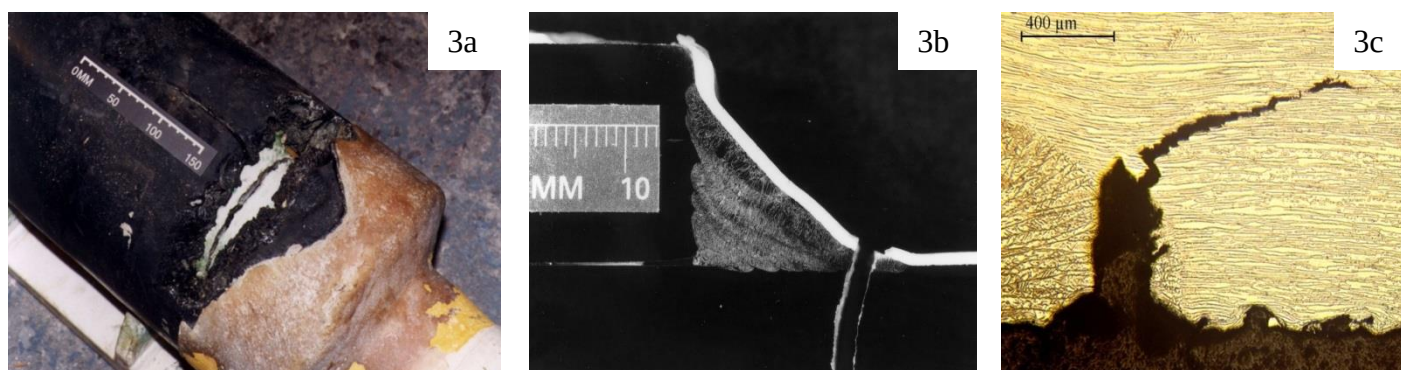


Figure 3a: location of primary cracking, b: macrostructure through the fillet weld and c: microstructure through which the primary crack propagated¹⁴

The methanol line was found to have failed at the toe of a butt weld between the tube and a flange on the flange side of the joint. The microstructure of the materials of construction and the weld was as expected and without fault. The coating applied was 300μm of epoxy phenolic, but on recovery of the spool most of the coating in the area of the leak was missing. It was concluded that plastic deformation of the pipes, due to excessive rock dumping, had caused the insulation coating around the fillet weld to spall off. This exposed the area to full cathodic charging with hydrogen from the CP system. At this point pre-existing hydrogen cracks in the highly ferritic microstructure of the fillet weld could begin to grow by the HISC mechanism until the pipe wall had been penetrated. Further laboratory tests were done and a safe limit of use of the spool material was established at strains below 3.5%. Welding procedures and records were all reviewed, so the cause of the weld cracking during fabrication could

be identified and the risk of similar defects remaining elsewhere in the spools eliminated. The original design case was revisited and as a result the cracked spool was replaced and production resumed under the original design conditions without further problems.

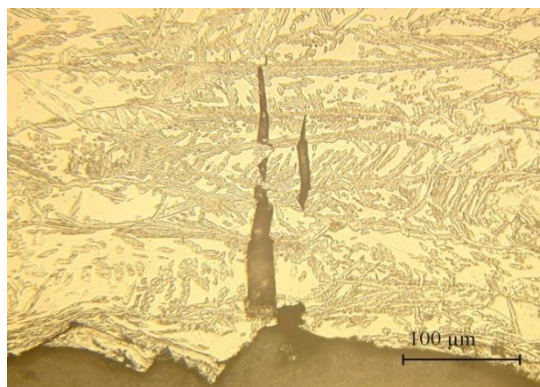


Figure 4 : Showing the ferritic weld metal microstructure and “step wise”, transgranular, cracking that was adjacent to the primary crack¹⁴

A Second Connector Forging Failure

In 2003, there was another HISC failure of a UNS S32760 hub connector forging¹⁵. It was very similar in size, shape and function to the component discussed previously^{12,13} that had failed 7 years earlier, but did not have butt weld nipple connections machined in. This event occurred after 1.5 years of service, while operating between 4°C to 50°C and exposed to a CP potential of -1050 mV (SCE).

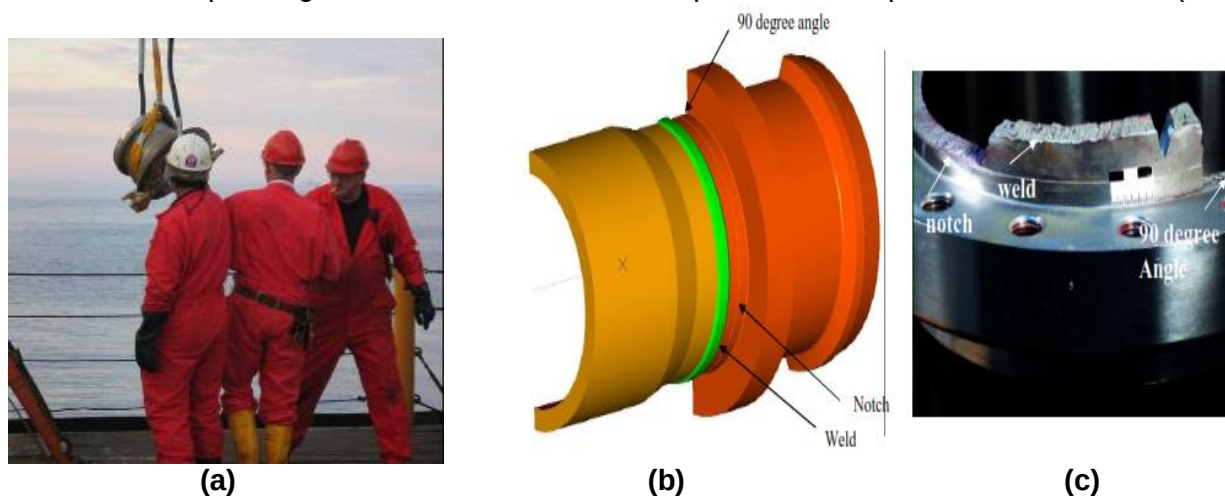


Figure 5: a) recovery of the detached hub, b) design detail of the hub, c) crack initiation sites¹⁶

In this case the hub had fully detached from the structure (Figure 5a) during a planned startup. There was no insulation coating applied to the part in the area of the failure. This is believed to be because this particular hub was mounted in a sliding support structure, fixing the part only at one end. This was unlike other, similar hubs, that that did not crack, which were fixed at both ends.

Sites of crack initiation were found at machined in details that formed, sharp angles and at weld toes (Figure 5b and 5c), all points of stress concentration. The cracking was found to run through the ferrite phase and appeared to be deflected off its path when it encountered austenite grains (Figure 6). Although not explicitly stated in the reports of the failure, looking at the microstructural detail presented, there is evidence of chromium nitride precipitation within the ferrite grains. This is in common with the first hub failure discussed.