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Corrosion Response of Downhole Steam-Generator Assembly and Instrumentation and Supply-Line Tubings at Long Beach Field Test

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And Supply-Line Tubings At Long Beach Field Test*

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ABSTRACT

Three families of metals were used to make the non-combustor components for Sandia's downhole steam generator assembly and supply and instrumentation lines. These three families were: first, plain carbon steel (API Grade J 55); second, austenitic stainless steels (316 and 310); and third, a nickel-base superalloy (Inconel 600). The metals in all three of these families were found to be deficient in their corrosion response. J 55 exhibited minimal to severe general corrosion (rusting). The austenitic stainless steels pitted and cracked. Inconel 600 showed both severe pitting and some intergranular attack. For the most part, these materials were found to be unsuitable for extended life in a downhole steam generator.

It is recommended that Inconel 625 be used in future systems where a moderate strength material is specified and Inconel 718 be considered where a high-strength material is necessary.

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INTRODUCTION

The goal of the U.S. Department of Energy's Project DEEP STEAM is to develop the technology necessary for economical recovery of heavy oil ($<18^\circ$ API) from deep reservoirs (>2500 feet).¹ Two approaches are being evaluated: optimizing injection string components for surface generated steam (this approach will not be further discussed within this report) and developing downhole steam generators to deliver the steam at the producing formation. Sandia National Laboratories entered into an agreement with the City of Long Beach to test two generators on wells operated in the Wilmington Field by Long Beach Oil Development Company (LBOD). One of the units burned diesel fuel in oxygen and was surface-operated. The details of the design, operation and system performance of this unit are given in a report by J. B. Moreno.² The other unit burned diesel fuel in air and was operated downhole. The details of the design, operation and system performance of this unit are given in a report by B. W. Marshall.³ Analyses of the materials performance of the combustors of the two systems is given in two reports.^{4,5} Analyses of the materials performance of selected non-combustor materials are reported herein.

The history of the oxygen/diesel steam generator unit is summarized as follows.² The steam generator was operated at ground level, with the steam and combustion products delivered to the reservoir through 2022 feet of calcium-silicate insulated tubing. Typically, the combustion efficiency was 95% and the steam quality was 56%. Excluding interruptions not attributable to the oxygen/diesel system, steam was injected 78% of the time over a four month period. System lifetime was limited by the combustor, which required some parts replacement every 2-3 weeks. The test was extensively instrumented to provide the required data. Problems with downhole instrumentation, particularly thermocouples, were experienced.

The history of the air/diesel downhole steam generator (DSG) units is summarized as follows.³ Equipment was assembled on site and checked out in the spring of 1981. The first DSG unit was placed in LBOD well SZ1-44 and combustion initiated in June 1981. After a period of operation mixed with periods of inoperation due to surface air compressor problems, loss of cooling water, debris buildup in downhole air and water filters, and gradual degradation in the downhole steam temperature and pressure measuring devices, the unit was removed from the well in August for inspection. Evidence of combustor overheating and melting (due to the loss of cooling water), filter rupture, and some corrosion resulted in design modifications for a second unit to be placed downhole.

The second unit, with the combustor and steam generator section identical to the first unit, was returned to the site and surface tested in late October. It was placed downhole with no difficulties and combustion initiated on November 15, 1981. Operation continued at an average of about 80% of the time until the end of February 1982 with the major portion of the down time again due to surface air compressors. Typical operating conditions were: energy output - 2 to 4 MMBtu/hr; air pressure - 1200 to 1400 psi; steam quality - 40 to 50%; and steam temperature - 440°F.

The unit was removed from the well in March 1982 without incident; all tubing remained banded together and the only significant external corrosion occurred on a carbon steel tailpipe below the generator. However, some internal corrosion and cracking of the stainless steel surfaces that form the combustion chamber was observed.

STEAM GENERATOR DESIGN

A schematic of the air/diesel DSG is shown in Figure 1. This entire section (the steam generator and combustor) was constructed from 310 stainless steel (SS). 310 SS is a high-alloy composition, austenitic stainless steel noted

for its high resistance to corrosion and oxidation at high temperatures while maintaining good strength. It also has good weldability and is commercially available. Table I lists the chemical composition of 310 SS and Table II gives its mechanical properties.

A schematic of the No. 2 air/diesel system is shown in Figure 2. In addition to the steam generator, the outer housings for the instrumentation package and filter cans were made from 310 SS. The sample, caustic and annulus treatment lines and outer sheathings on the RTD's were made from Inconel 600 tubing in the proximity of the steam generator system. Above the steam generator system, 316 SS tubing was used. The thermocouple sheathings for the air/diesel generator were made of Inconel 600 while those for the oxygen/diesel well were made of 316 SS. The chemical compositions and mechanical properties are given, respectively, in Tables I and II for Inconel 600 and 316 SS.

The water and air supply lines were API Grade J 55 pipe, which is essentially plain carbon steel (composition and mechanical properties given in Tables I and II, respectively). In addition, the packer body and exhaust pipe were also made from this material.

A summarization of the environments to which the various components of the steam generator system were exposed is given in Table III. As a function of time, the fluid in the annulus was both untreated and treated with oxygen scavengers and film inhibitors. Also, as a function of time and location, the temperature varied between 100 and ~ 450°F. For the exhaust pipe (the environment below the packer), the pH of the generator effluent varied between 2 and 7 depending upon the effectiveness of the caustic injection.

EXPERIMENTAL PROCEDURES

Standard procedures were used to examine the corroded components. The initial step was to remove pieces of the component to be examined from the bulk, usually by

cutting. A piece was then taken to the metallographic facility to be mounted in epoxy, polished, and subjected to metallographic examination. Another piece was usually examined and photographed using a macrophotographic camera. A third piece, along with the mounted first piece, was examined in a scanning electron microscope (SEM) and subjected to energy dispersive X-ray analysis (EDX) of various microconstituents in the SEM.

Loose corrosion products were chemically and structurally analyzed using EDX and x-ray diffraction equipment, respectively.

RESULTS

Corrosion of J55 Pipes

There were four major components which utilized a plain carbon steel for construction. Three were J55 pipe; the water supply, air supply and exhaust pipes. The fourth application was the packer body. Three of these components (water supply, air supply and packer) were subjected to the annulus environment. None of these components exhibited anything more than superficial, cosmetic corrosion. Apparently, the additives to the annulus fluid, which were oxygen scavengers and film forming inhibitors, performed extremely well, preventing any significant corrosion. The fourth component was the exhaust pipe which was subjected to the generator effluent. The exhaust pipe was found to be very severely and non-uniformly corroded, indeed, perforated in spots. Analysis of the rust-colored corrosion product found it to be primarily a mixture of hematite (Fe_2O_3) and hydrated hematite ($\text{Fe}_2\text{O}_3 \cdot x\text{H}_2\text{O}$). There is reason to believe³ that the caustic (NaOH) injected to neutralize the very acidic (pH of 2) generator effluent was not all reaching the injection point. If the generator effluent was not buffered substantially towards a pH of 7, a plain carbon steel such as the exhaust pipe would corrode significantly at the temperatures and steam and oxygen contents observed.

Corrosion of 316 SS Tubulars

As mentioned previously, 316 austenitic stainless steel was used for supply lines (fuel, caustic, annulus treatment, sample) above the generator system and as sheathing for the thermocouples secured to the insulated tubing in the annulus of the oxygen/diesel well. In addition, some fittings were made from this alloy. Three examples of corrosion induced failures within this material will be given as representative of this set.

Thermocouple sheathing - The first example of a corrosion induced failure was the failure of thermocouple sheathings. Examples of cracked sheathing are shown in Figures 3 and 4. It can be seen that the cracking was not confined to a specific area. Figures 5 and 6 are SEM macrographs of cracks showing more detail on the length and depth of the cracks. An end view of a thermocouple separated in an area of a crack is shown in Figure 7. Indeed, cracking has resulted in failure of the thermocouples. Figure 8 shows the chemical analysis spectrum obtained from this surface using the EDX technique. The spectrum identifies the material as a 300 series stainless steel (i.e. Fe, Cr, Ni peaks) and also shows a very prominent peak for chlorine (indicating the presence of the chloride ion). Higher magnification views of the fracture surface shown in Figure 7 are shown in Figures 9 and 10. Figure 9 shows that the area of the transition from environmentally caused fracture to that caused by the mechanical separation of the thermocouple (overload) upon bending open for observation. The overload area is seen as ductile-dimple rupture (top area of photograph). The environmentally induced fracture surface (bottom area of photograph) has the appearance of transgranular (across grains) fracture. Figure 10 shows an area adjacent to the cracked area where the surface has been subsequently corroded. Another area of the fracture surface where the subsequent corrosion has completely obscured the underlying fracture surface is shown in Figure 11. Chemical

analysis of this corrosion product is shown in the spectrum of Figure 12. The film is a chromium-rich oxide, probably a $\text{Cr}_2\text{O}_3 \cdot \text{Fe}_2\text{O}_3$ spinel.

Another technique used to examine the severity and type of cracking is to metallographically cross-section the component of interest. Figure 13 shows the cross-section of a thermocouple sheathing at 100X in the unetched condition. It can be seen that cracks have initiated on both sides of the thermocouple sheathing. This does not mean that the interior of the thermocouple (chromel-alumel wires and alumina insulation) can generate cracks in the sheathing but rather that the sheathing was breached and solution got inside. The etched macrograph of Figure 14 indicates that the cracking is indeed transgranular. The crack path follows the grain boundaries only when directionally convenient but primarily transverses the grain structure.

These results were obtained for a small diameter (1/16") thermocouple used in the O/D well. Similar results were observed for two larger diameter (1/8") thermocouples manufactured in a similar manner with the same materials and exposed to the same environment.

An additional set of data was taken on the small thermocouple sheathings. Microhardness traverses were made to check for any material embrittlement due to hardness increase. No hardness increase over the standard values was measured. Thus, a general embrittlement of the material due to an interstitial species ingress such as hydrogen or oxygen did not occur.

The cause of the cracking was identified as chloride-assisted stress corrosion cracking. This cracking mechanism will be further discussed in the Discussion section of this report.

Supply Line and Thermocouple Protective Tubing - The second example of a corrosion induced failure of a 300 series stainless steel component was the pitting

and cracking of the supply and instrumentation lines. The former were the 3/8" tubes above the generator system, while 1/4" tubes were used to protect the thermocouples in the vicinity of the air/diesel DSG. Figures 15 and 16 show examples of the pitting of one of the thermocouple protective tubes at various magnifications. Figure 15 in particular reveals the linearity of the pits along the axis of the tubular. This linearity is believed to be associated with the mechanical packaging of the tubes in bundles and suggests that the pitting is occurring at a position on a tubing where another tubing is in proximity. Thus, it is the crevice formed between the two tubes which contributes significantly to the corrosion, and this phenomena has been termed "crevice corrosion". Crevice corrosion will be discussed further in the Discussion section.

An SEM micrograph of a pitted area is shown in Figure 17. The feature to be noted in this micrograph is the absence of any appreciable corrosion products within the pits. This event is characteristic of corrosion induced pitting. Figure 18 is a higher magnification SEM micrograph of an area at the edge of a pit. (The edge of the pit is at the right side of the photograph). This micrograph shows the characteristic irregular surface morphology associated with the chemical attack of pitting. This micrograph also shows very distinctively the attack at the grain boundaries on the surface of the tube next to the edge of the pit (left side of photograph). This type of attack has been termed "intergranular attack." Intergranular attack will be discussed further in the Discussion section of this paper. Intergranular attack very often is the initiating process for subsequent stress corrosion cracking. And indeed it could be the cause for the cracking observed in these tubes, Figures 19 and 20. Figure 19 reveals the severity of the pitting and Figure 20 shows that again the stress corrosion cracking path was transgranular.

Swagelok Nut - The third example of corrosion induced failure of hardware made from 300 series SS was a cracked Swagelok nut. A macrograph of the fractured nut is shown in Figure 21. This particular fitting was on the caustic line bringing NaOH to the generator effluent for neutralization. The crack encompassed approximately 270° of the nut circumference. Also, it was noted that the crack direction changed from circumferential to longitudinal at both ends of the crack. This crack path directional change correlates well with a predicted change in primary tensile stress orientation as the crack propagated and the nut loosened. Figure 22 is a high magnification micrograph taken in the SEM showing significant secondary cracking in the SS adjacent to and parallel with the primary crack. Metallographic cross-sections of this Swagelok nut in the area of the primary crack, Figures 23 and 24, show the complete penetration of the primary crack across the nut wall, Figure 23, the secondary cracking associated with the primary crack, Figures 23 and 24, and the transgranular paths of the cracks, Figure 24.

A SEM macrograph of the primary crack in the Swagelok nut is again shown in Figure 25. Associated with the crack is a material deposit. The composition of this deposit was determined by using X-ray fluorescent analysis, the results of which are shown in Figure 26. The deposit is sodium hydroxide which this tubular was carrying. Thus, the cracked nut did allow caustic to exit the tubing at this location. What percentage of caustic was lost through this crack and how much remained in the tubing and was injected into the generator effluent is indeterminate.

There are two possible causes of the cracking in this Swagelok nut; chloride-assisted stress corrosion cracking and hot caustic cracking. Type 300 series austenitic stainless steels are subject to stress corrosion cracking in hot concentrated caustic solutions. For this to have happened, of course, the caustic

would have had to leak from the inner tube by the Swagelok fitting into the nut before the cracking took place. If the nut was loose enough to allow this, there would probably not be sufficient tensile stress on the nut to cause cracking. However, it is possible that the fitting was defective or assembled incorrectly such that the caustic leaked into a properly tightened nut, resulting in stress corrosion cracking.

The other possible cause of the cracking was chloride-assisted stress corrosion cracking. This cracking phenomena will be discussed further in the Discussions section of this report.

Corrosion of 310 S.S. Generator Components

As mentioned previously in the Steam Generator Design section, the entire air/diesel steam generator was constructed from 310 SS. Some additional components attached below the generator were also made from 310 SS. Specifically, a 310 SS outlet pipe from the generator was welded to another 310 SS pipe attached to a temperature measurement chamber and centralizer assembly. The only visual evidence of any problem with any component made from 310 SS was observed in the weld between the two 310 SS pipes. Figure 27 is a macrograph of this weld showing the severe cracking. It was conjectured that perhaps the weld was not properly done and lacked full penetration. Indeed, Figure 28 shows that this was the case, i.e., the weld did not fully penetrate and close the joint to the inner surface. However, the lack of full penetration was not the only cause of the cracking, although an important contributor. Figures 29 and 30 show metallographic cross-sections of the weld zone. It can be seen that, although the primary crack is associated with the remaining gap, there are numerous other cracks within and just outside the weld zone. Thus, many cracks have initiated and propagated but most likely the first, the fastest, the longest and the

largest was associated with the mechanical gap. A more detailed look at the metallography and location of the crack paths both within the weld metal and in the adjacent parent metal are shown in Figures 31 and 32 and in Figures 33 and 34, respectively. In both cases, the crack path followed very distinctly the grain boundaries, that is between the grains or intergranular. The cause of the crack is believed to be chloride-assisted stress corrosion cracking analogous to that seen in the 316 SS components but intergranular rather than transgranular. This cracking phenomena will be discussed further in the Discussions section of this report.

Because of the cracks found in the 310 SS pipe associated with the weld zone, the bottom end of the generator was sectioned at six weld joints for examination. Figures 35 through 37 show typical examples of the microstructure and integrity of these six welds. All six welds exhibited full penetrations, no void formation and an absence of cracking. Further observation of the metallographic mounts containing these weldment specimens uncovered the following surface response. Figure 38 shows an area of the generator surface where significant compositional microsegregation existed in the material. This surface was exposed to the annulus fluid and no deleterious corrosion response was seen. In contrast, Figure 39 shows an area of the generator surface where the microstructure was clean but intergranular attack has initiated. This surface had been exposed to the generator effluent. Intergranular attack can be an initiator of intergranular, stress corrosion cracking but the correlation was not observed in this instance.

The evidence of cracking associated with the weld which did not exhibit full penetration as opposed to no evidence of cracking in welds which had full penetration causes a comparison to be made. Two major differences in these welds are apparent. First, the lack of full penetration in the cracked weld could have

resulted in a much higher tensile stress within the weld region than the full penetration welds. Higher tensile stresses mean increased SCC susceptibility. Second, the cracked weld was subjected to the generator effluent. The generator effluent was also shown to have caused intergranular attack on parent material. The full penetration welds were exposed to the annulus fluid. The annulus fluid had no noticeable effect upon the 310 SS material. These pieces of evidence suggest that full penetration welds may have increased sensitivity to SCC when exposed to the generator effluent environment.

Corrosion of Inconel 600 Tubulars

On the first air/diesel DSG assembly, 300 series SS tubing was used for the entire length of the supply and instrumentation lines. Stress corrosion cracking was encountered in the annulus environment in the vicinity of the steam generator. Thus, on the second air/diesel unit, Inconel 600 tubing was used for the supply and instrumentation lines in the zone containing the generator. This substitution of material did eliminate the stress corrosion cracking problem. Unfortunately, another problem arose which led to similar results as before, that is, leaky tubes. This time the tubes pitted severely, producing holes, as seen in Figures 40 and 41. Figure 42 is an X-ray fluorescent analysis spectrum showing the compositional pattern for Inconel 600. A SEM micrograph of a pitted area in an Inconel 600 tube is shown in Figure 43. The lack of corrosion product within the depression is characteristic of pitting. Figure 44 shows a smaller pit but at higher magnification. Contributing to the pit formation was the intergranular attack which undermined the grains, allowing them to detach from the surface. The susceptibility of this material to intergranular attack is further shown in Figure 45. The phenomenon of pitting will be further discussed in the Discussion section of this report.

DISCUSSION

With the exception of the general attack of the exhaust pipe by the effluent gases, the failures documented in the Results section of this paper were caused by "localized corrosion". Localized corrosion can be defined as selective removal of metal by corrosion at small special areas or zones on a metal surface in contact with a liquid environment.⁶ It usually occurs under conditions where the largest part of the original surface either is not attacked or is attacked to a much smaller degree than at the local sites.

Pitting Corrosion

One common form of localized corrosion is pitting, in which small volumes of metal are removed by corrosion from certain areas on the surface to produce craters or pits. While the shapes of pits vary widely, they usually are roughly saucer shaped, conical or hemispherical.⁶ Pit walls usually are irregular when viewed under a microscope.

Pitting usually occurs on metals that are covered with a very thin, often invisible, adherent protective surface film which may be formed during fabrication or be produced by reaction with the environment. Pits develop at weak spots in the surface film and at sites when the film is damaged mechanically under conditions where self repair will not occur.

The practical importance of pitting depends on the thickness of the metal and on the penetration rate, which usually decreases with time. The rate of penetration generally decreases if the number of pits increases. The reason for this is that adjacent pits have to share the available adjacent cathode area, which controls the corrosion current that can flow. Most cases of pitting are believed to be caused by local cathode sites in an otherwise normal surface.

Since pitting is electrochemical, it can be stopped by cathodic protection.⁷ It can be prevented also by the use of inhibitors which alter the electrode reactions of the local cell or by coating the surface of the metal with a sacrificial layer or a protective coating.

Crevice Corrosion

A specific type of pitting corrosion which occurs in crevices is called "crevice corrosion".⁶ Crevice corrosion occurs in cracks or crevices formed between mating surfaces of metal assemblies and usually takes the form of pitting or etched patches. Both surfaces may be of the same metal or of dissimilar metals or one surface may be a non-metal. It can occur also under loose fitting washers and gaskets that do not prevent the entry of liquid between them and the metal surface.

Crevice corrosion is believed to initiate as the result of a differential aeration mechanism. Oxygen in the liquid deep in the crevice is consumed by reaction with the metal. The oxygen content of the liquid at the mouth of the crevice exposed to air, or of bulk liquid in the case of immersion, is greater so a local cell develops in which the anode, or area being attacked, is the surface in contact with the oxygen-depleted liquid. It is also thought that subsequent pH changes at anode and cathodic sites further stimulate local cell action as in the case of pitting corrosion.

Intergranular Attack

Intergranular corrosion is a form of localized surface attack in which corrosion occurs preferentially along the grain boundaries of a metal.⁶ It initiates on the surface and proceeds by local cell action in the immediate vicinity of a grain boundary. The driving force is a difference in corrosion potential that develops between a thin grain boundary zone and the bulk of the immediate adjacent grains. The difference in potential is often caused by a

difference in chemical composition between the two zones. The grain boundaries may even contain precipitate particles which usually have a different corrosion potential than that of the grains, which causes a local cell to form. If the precipitate is anodic to the denuded zone it corrodes preferentially and if it is present in a continuous network throughout the grain boundary, it is removed to produce a thin channel, fissure, or crack between adjacent grains. If the precipitate is cathodic to the denuded zone, it remains intact and local cell action develops a crack in the adjacent denuded zone.

Since most intergranular corrosion is the result of small differences in composition at grain boundaries, the metallurgical history of an alloy can be as important as its average chemical composition. In particular, stainless steels are susceptible to intergranular corrosion when improperly heat treated, that is, "sensitized" (chromium carbide precipitates within the grain boundaries).

Intergranular corrosion is often prevented by ensuring a metal microstructure that is immune to this type of attack. For stainless steels this translates to not heat treating or subjecting the alloy to a use temperature which is within its temperature zone for sensitization, i.e. - carbide formation.

Stress Corrosion Cracking

"Stress corrosion cracking" is defined as the spontaneous cracking of metals as the result of the combined action of a corrosive environment and tensile stresses, either applied or residual.⁸ How much stress and how extensive must the corrosive action be to produce cracking? From extensive laboratory work,⁸ it is believed that at least in the region where the cracks originate, the stresses must be above the yield strength of the material; that is, the stress must produce some plastic deformation of the material in the

region where cracking occurs. The tensile stress necessary to produce the plastic deformation is a combination of residual and applied stresses.

Stress corrosion can occur under very mildly corrosive conditions. The new surfaces resulting from a stress-corrosion crack may show evidence of corrosion but other surfaces of the metal usually do not appear to be corroded. If there is severe general corrosion, stress corrosion cracking usually will not occur. Again, stress corrosion cracking occurs in metals exposed in an environment where, if the stresses were lower, there would be no damage. On the other hand, if the metal, subjected to the higher stresses, were in a different environment, there would be no failure.

Stress corrosion cracks in early stages of their development are microscopic. In many cases they are not evident on the exposed surface by normal visual examination and require special techniques such as scanning electron microscopy. As the cracking penetrates further into the material it eventually reduces the supporting cross section to the point where the structure fails from overload.

Stress corrosion cracks are designated as two types. Intergranular or intergranular cracks follow the grain boundaries in the metal. Transcrystalline or transgranular cracks generally pass through the metallic grains rather than follow their boundaries. Transgranular cracking (TG SCC) in austenitic stainless steel is usually associated with chloride-containing aqueous environments. However, under certain metallurgical and/or environmental conditions, the fracture mode may be predominantly intergranular (IG SCC). The occurrence of intergranular cracking is usually associated with the presence of a sensitized microstructure, but not always.⁹

For austenitic stainless steels, there is general agreement, that if stress cracking will develop in 304 SS exposed in a given environment, a 316 or 316L SS probably will not be more resistant. Moreover, a higher alloy stainless steel such as 310 SS also will fail in this environment, but after a longer exposure period.⁸

Various schemes have been tried in an attempt at preventing stress corrosion cracking of austenitic stainless steels in contact with chlorides, most unsuccessfully. If possible, such as in a closed system, keeping the chloride and oxygen contents of the water to less than 1 ppm may be successful. Another approach is to keep the applied stresses as low as possible, but of course with the uncertainty of the magnitude of the residual stresses (the usual case), this is a risky approach.

CONCLUSIONS

1. Plain carbon steels can be used for steam generator components when the environment is comparable to the treated annulus water.
2. Some austenitic stainless steels (304, 316, 316L) cannot be used for steam generator components because they undergo transgranular stress corrosion cracking, even in treated water, due to the presence of the chloride ion.
3. Type 310 austenitic stainless steel is more resistant to transgranular stress corrosion cracking than other austenitic stainless steels but is susceptible to intergranular stress corrosion cracking when sensitized such as when welded.
4. Inconel 600 can be used to avoid stress corrosion cracking in chloride environments, however, it does exhibit severe pitting and intergranular attack.

RECOMMENDATIONS

It is recommended that Inconel 625 be used in future systems where a moderate strength material is required and Inconel 718 be considered when a high strength material is necessary.

Inconel 625 is used for its high strength, excellent fabricability (including joining), and outstanding corrosion resistance from cryogenic temperatures to 2000°F. Composition is shown in Table IV and its mechanical properties are shown in Table V.¹⁰

Strength of Inconel 625 is derived from the stiffening effect of molybdenum and columbium on its nickel-chromium matrix. Thus, it is a solid-solution matrix-stiffened face-centered-cubic alloy that does not require precipitation-hardening treatments to obtain strength. It is also stable when exposed for extended periods in "sensitization" temperature zones, 1000 to 1800°F. Inconel 625 is readily fabricated by common industrial practices. Two companion welding products¹⁰ ensure achievement of high-quality joints with strength and corrosion resistance comparable to those of the base material.

The high alloy content of Inconel 625 enables it to withstand a wide variety of severe corrosive environments. In mild environments such as the atmosphere, fresh and sea water, neutral salts, and alkaline media there is almost no attack.

In more severe corrosive environments the combination of nickel and chromium provides resistance to oxidizing chemicals, whereas the high nickel and molybdenum contents supply resistance to nonoxidizing environments. The high molybdenum content also makes this alloy very resistant to pitting and crevice corrosion, and columbium acts to stabilize the alloy against sensitization during welding, thereby preventing subsequent intergranular cracking. Also, the high nickel content provides freedom from chloride-ion stress corrosion cracking.

Inconel 718 is a high-strength, corrosion-resistant material used at -423 to 1300°F. Composition is shown in Table IV and its mechanical properties are shown in Table V.¹¹ Inconel 718 can be readily fabricated, even into complex parts. Its welding characteristics, especially its resistance to postweld cracking, are good.

Inconel 718 is an age-hardenable austenitic material. Strength is largely dependent on the precipitation of a gamma prime phase following heat treatments. It has good tensile, fatigue, creep, and rupture strength. A major part of the development effort with this alloy was concerned with establishment of the proper heat treatments for producing optimum properties.

Inconel 718 has excellent corrosion resistance to many media. Nickel contributes to a resistance to corrosion by many inorganic and organic (other than strongly oxidizing) compounds throughout wide ranges of acidity and alkalinity. It also is useful in combating chloride-ion stress corrosion cracking. Chromium imparts an ability to withstand attack by oxidizing media and sulfur compounds. Molybdenum is known to contribute to resistance to pitting in many media.

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TABLE I

METAL COMPOSITIONS

<u>Designation</u>	<u>Chromium</u>	<u>Nickel</u>	<u>Molybdenum</u>	<u>Carbon</u>	<u>Iron</u>
J-55	-	-	0.15	0.40	balance
316 SS	17	12	2	0.08	balance
310 33	25	20	-	0.25	balance
Inconel 600	15	72	-	0.15	8

TABLE II

MECHANICAL PROPERTIES⁺

<u>Designation</u>	<u>Yield Strength</u> KSI	<u>Ultimate Strength</u> KSI	<u>Elongation</u> %
J-55	55	75	20
316 SS	35	82	55
310 SS	45	95	45
Inconel 600	40	90	40

⁺at room temperature

TABLE III

ENVIRONMENTS

Annulus:

Fluids - Mixture of crude oil, sand and salt water (NaCl ~ 26 g/l)
with oxygen scavenger and filming inhibitor

Pressure - ~ 15 to 1380 PSI

Temperature - ~ 100 to 450°F

Below Packer:

Generator Effluent - N₂, CO₂, CO, O₂, H₂, CH₄, C₂H₆, NO_x, SO_x, H₂S and H₂O

Caustic (NaOH) - 50% Aqueous Solution

Pressure - ~ 1200 to 1380 PSI

Temperature - ~ 440 to 550°F

TABLE IV

COMPOSITIONS OF RECOMMENDED ALLOYS

<u>Designation</u>	<u>Chromium</u>	<u>Nickel</u>	<u>Molybdenum</u>	<u>Iron</u>	<u>Nb+Ta</u>	<u>Al+Ti</u>
Inconel 625	21.5	61.0	9.0	2.5	3.6	0.4
Inconel 718	19.0	52.5	3.0	18.5	5.1	1.4

TABLE V

MECHANICAL PROPERTIES OF RECOMMENDED ALLOYS

<u>Designation</u>	<u>Yield Strength</u> KSI	<u>Ultimate Strength</u> KSI	<u>Elongation</u> %
Inconel 625	55	120	45
Inconel 718	161	190	20

Figure 1. Schematic of Air/Diesel Downhole Steam Generator

Figure 2. Schematic of Air/Diesel Downhole Configuration

Figure 3. Macrograph of Thermocouple Sheathing - 316 SS

Figure 4. Macrograph of Thermocouple Sheathing - 316 SS

Figure 5.

SEM Cracked Thermocouple Sheathing

316 SS - Top View 45X

Figure 6.

SEM Cracked Thermocouple Sheathing

316 SS - Side View 75X

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- Cracked to Overload 1000X

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- Corroded to Cracked 1000X

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Figure 22. SEM of Secondary Cracking at Edge of Primary Crack 1000X

- 316 SS Swagelok Nut

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- Etched 30X

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- Cracked 9X

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- Cracked 9X

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Figure 33. Metallography of Parent Metal in 310 SS Outlet Pipe 25X

Figure 34. Intergranular Crack Path in 310 SS Parent Metal 100X

Figure 35. Weld Joint at Bottom End of 310 SS Air/Diesel Generator 31X

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Figure 37. Weld Joint at Bottom End of 310 SS Air/Diesel Generator

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Figure 40. Macrograph of Inconel 600 Tubing

- Sample Line for Air/Diesel DSG - Hole

Figure 41. Macrograph of Inconel 600 Tubing

- Annulus Treatment Line for Air/Diesel DSG - Deep Pit

Figure 42. X-Ray Fluorescent Analysis of Inconel 600 Tubing

Figure 43. SEM of Pit in Inconel 600 Tubing - 30X

Figure 44. SEM of Pit Showing Intergranular Attack on Inconel 600 Tubing - 100X

Figure 45. SEM of Intergranular Attack on Inconel 600 Tubing - 400X

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