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OF METALS AND THEIR OXIDE SCALES

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MECHANISMS OF EROSION/CORROSION OF METALS AND THEIR OXIDE SCALES

A. Levy, R. Bellman, Jr., and J. Maasberg
Lawrence Berkeley Laboratory
University of California
Berkeley, California 94720

I. INTRODUCTION

The presence of small particles of coal and char in moving reactive gases at elevated temperatures in coal gasifiers creates a hostile environment that has caused surface degradation of metallic components in pilot plant operations. This has generated research projects to determine how materials resist the combined erosion-corrosion surface degradation that can occur. Investigations of comparative materials behavior in char-reactive gas erosive streams (1) and as valve seats and stems (2,3) have determined which current materials have the most promise for service in coal gasifiers. The objective of the work reported herein is to provide some insight into the basic mechanisms of degradation that occur when a small particle impacts on a bare metal surface or on one that contains a thin oxide scale. It is hoped that understanding how material is removed from the surface will provide the basis for selecting operating conditions and developing more resistant materials that will result in less materials degradation.

II. DUCTILE METAL EROSION MECHANISM

Background

Recent work in studying the erosion behavior of materials (4-6) has relied upon older, established mechanisms of behavior that were based on research performed in the 1950's and 1960's in support of oil catalytic cracking and helicopter engine problems. The mechanisms were based on a micro-machining concept where the impacting particle penetrates the target a small amount, translates along the surface removing material ahead of it in a machining mode and finally leaves the surface (7).

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A refined model was developed for this mechanism by Finnie that utilized the equation of motion of the particle tip to define an amount of target material that was removed (8). The scanning electron microscope (SEM) with its capability to achieve great depth of field at high magnifications which makes it possible to observe what is physically happening on an eroding surface was not available to Finnie and others at the time of their pioneering work.

The micro-machining model has had several problems in predicting the amount of material lost from a surface under a variety of test conditions. Further, it never was able to properly account for erosion at high impingement angles (9,10). Investigators all tended to use the equation of motion of the particle translating along the surface of the specimen as a starting point for developing an analytical model.

Some investigators have attempted to look at the mechanism of erosion from a physical standpoint. To be able to observe what was happening, they used large single particles, millimeters in diameter, and fired them individually at target surfaces (11,12) or even held the particle fixed and fired the target material at it (13). In these types of tests, the concentrated kinetic energy from the large mass of the impacting particle was not representative of the much smaller particles (1-500 μm) that occur in real erosion environments. Hence, such phenomena as surface melting of the target was observed, an occurrence never seen on actual eroded surfaces.

In early erosion mechanism work performed in this laboratory (13,14) on eutectoid carbon steel, 1075, the threshold phenomena and the effect of cold working on erosion were measured. In both instances the measured erosion was opposite to that which the micro-machining mechanism would account for. Figure 1 shows the amount of erosion that occurred upon successive 60 gm quantities of 250 μm SiC impacting on three different eutectoid steel microstructures. It can be seen that there is a period at the beginning of the tests where no erosion occurred followed by a period where increasing amounts of erosion occurred. A steady state amount of erosion was finally reached. The same pattern of behavior was observed in aluminum alloys.

If micro-machining was the primary mechanism of erosion, the initial impacts should have eroded as much or even more material than subsequent impacts. Further, continued micro-machining should have work hardened the surface, decreasing the amount of erosion as the amount of eroding particles increased. The opposite occurred. To investigate the effect

XBL775-5486

Fig. 1. Erosion of 1075 steel

of work hardening on the erosion behavior of the steel, the spheroidized steel was cold rolled up to 80% reduction prior to the initiation of erosion. Table 1 shows the results of the cold rolling experiment. As the amount of cold rolling and accompanying hardness increased, the amount of erosion increased toward the steady state level. This is opposite to the behavior that should have been achieved by cold rolling as increased hardness should have made micro-machining more difficult.

In order to determine what actually occurs when particles strike a ductile metal surface, carefully controlled experiments were conducted wherein fresh and steady state surfaces

were eroded on an interrupted basis with SEM observations of the same small surface area made between each erosion sequence. This required indexing of the surface so that an area as small as $1000 \mu\text{m}^2$ could be precisely relocated after repeated removals and replacements of the specimen on the SEM stage. The result was what is believed to be the first documented history of an eroding ductile metal surface. This painstaking experimental work showed that the basic removal mechanism was not micro-machining but rather a combined forging-extrusion action that produced thin platelets of highly distressed metal that were finally knocked off the surface by succeeding particles.

Table 1. Effect of Cold Work on Erosion of Spherodized 1075 Steel.

% Cold worked	Hardness, VHN 1000 gm. load	Initial erosion rate from 60 gm. of SiC particles
0	162	9.8×10^{-3} mg/g
20	242	10.3×10^{-3} mg/g
40	262	14.9×10^{-3} mg/g
60	288	16.6×10^{-3} mg/g
80	316	17.2×10^{-3} mg/g
		steady state erosion rate
		22×10^{-3} mg/g

Experiment Description

The test materials used were 1100-0 aluminum, a low strength, single phase, FCC metal and 7075-T6, a high strength, multi-phase FCC aluminum alloy. The 1100-0 aluminum is similar in microstructure and strength at the ambient erosion test temperature to the 300 series stainless steels used in coal gasifier internals at elevated temperatures near 1000°C . The 7075-T6 aluminum alloy at its high room temperature strength can be compared to the low strength of 1100-0 aluminum to relate strength and attendant hardness and ductility to the behavior mechanism.

All of the tests were performed in an air blast tester which has a 3/16" dia. nozzle. The particles were 600 μm silicon carbide delivered at a velocity of 100 fps and a solids loading of 0.1 gm/sec. The impingement angles used were 30° and 90°. The tests were conducted in short bursts of particles, generally 0.1 to 1 gm at a time.

A sequential erosion technique is utilized to observe the erosion mechanism. An uneroded specimen is marked at a control point with several microhardness indentations. This control point served as a reference for each relocation of the same areas on the specimen's surface when it is observed in the scanning electron microscope (SEM) after each succeeding burst of particles. The marked specimen is placed in the erosion machine and a small amount of particles, typically 0.1 g or 500-1000 particles, impact the surface. The specimen is then removed from the tester and arbitrary areas within the erosion zone are microphotographed in detail. Usually 1 to 5 areas will be chosen. The specimen is returned to the exact position it occupied previously in the erosion tester and again impacted with the same amount of particles. The previously selected areas are re-examined with the SEM and the damage caused by this second blast of particles is then microphotographed. The cyclic process of erosion--observation is continued until enough data is gathered, usually 3 to 8 cycles.

The method is equally well applied to an already eroded surface. In this case, the specimen is first impacted with a large amount of particles (500 gm) to bring the specimen into the constant erosion rate regime. After achieving this steady state surface, the specimen is then referenced with the microhardness indentations. After these preparations, the sequential erosion technique proceeds.

Results

The surface of an 1100-0 aluminum specimen is shown in Fig. 2. The amount of surface deformation that occurs after 0.1 gm of particles impact at a 90° impingement angle at 100 fps can be seen in the upper left photo. Higher magnification photos are shown in the other 3 pictures in Fig. 2 after 0.1, 0.2, and 0.3 gm of particles have impacted the surface. By studying a single area such as the one outlined a good idea of how the initial platelets form can be obtained.

Figure 3 is a higher magnification picture of the encircled area of Fig. 2. The effect of the sequence of particle impacts on the area of the triangle shaped area in the center of the

Fig. 2. Early erosion sequence on fresh 1100-0 aluminum surface

Fig. 3. Higher magnification photos of early erosion of 1100-0 aluminum surface

picture shows the beginning of platelet formation. Note how sections of the original triangle are separated and flattened by particle impacts. The area beyond the top of the triangle is a previously formed crater. Note that essentially no material has been removed from the surface in this area, which is typical of the whole surface being eroded after 0.4 gm of erodant has impacted it.

Figure 4 shows a similar sequence of erosion behavior, this time on a steady state eroded surface that had previously been impacted with 500 gm of particles at an impingement angle of 30° . The highly plastic nature of the eroding surface can be seen. While material is being removed in the upper left hand corner of the photos, there is no evidence of micro-machining or fracture. Rather the surface is being smeared into craters and the platelets of metal that are formed about the crater are being knocked off by subsequent impacts. Figure 5 shows two typical examples of well formed, severely plastically deformed platelets. At the stage the platelet in the lower picture is in, the next particle that strikes it can knock a good portion of it off the surface. These platelets are formed by a combined forging-extrusion action which occurs at all angles of impingement. The material in Figs. 4-5 is 1100-0 aluminum.

Fig. 4. Steady state erosion sequence on 1100-0 aluminum

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Fig. 5. Steady state erosion platelets formed on 1100-0 aluminum

Figures 6 and 7 are a sequence observed on a steady state eroding surface of 7075-T6 aluminum. The impingement angle was 30° . The surface was first eroded to a steady state condition by 500 gm of particles and then sequentially photographed after each of four succeeding grams of erodent were impacted against it. The area consists of a number of overlapping platelets of various sizes that have been formed by previous impacting particles. The lower picture in the sequence, Fig. 6,

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Fig. 6. Platelets being formed and removed on 7075-T6 aluminum

shows that the platelets in the upper right hand side of the upper figure have been knocked off. In addition the force of the impacting particles has extruded out new platelet material on the upper left hand side and further deformed the platelets at the bottom of the view. The sequential pictures in Fig. 7

Fig. 7. Platelets being formed and removed on 7075-T6 aluminum

show further formation, extension and removal of the platelet material seen in the upper photograph in Fig. 6. The 4th photo in the sequence shows that the original platelet complex has been essentially completely removed and a new sequence of platelets is taking its place.

This behavior is typical of what was observed over the whole surface of the eroding material; it is not an isolated occurrence. It is observed to the same degree on both the 1100-0 and 7075-T6 aluminum. The difference in hardness and strength between the two alloys (the 7075-T6 has several times the strength of the 1100-0 aluminum) does not affect the platelet forming and removal mechanism. The great force of the impacting particles exceeds the strength of both materials by a large margin. The sizes of the craters and the platelets is essentially the same in both alloys.

Figure 8 is a sequence showing the formation, extension and removal of a single platelet by 4 particle impacts. The sequence is shown in a clockwise positioning of the pictures. The first picture in the upper left of Fig. 8 shows a crater. The 2nd picture shows a new crater formed along the right hand side of the photo and a platelet from it that has been formed and thrown up and to the left. The flattening and extension of the platelet as the result of the impact of two additional particles is shown in the 3rd picture. The 4th picture shows the area after the impact of one additional particle; the platelet has been knocked off the surface, revealing the original crater over which it was extruded. The markings in the crater shown in the 4th picture identify it as the unaffected crater of the 1st picture. This is typical behavior of platelets. They do not bond to the areas over which they are extruded. Their attachment to the surface occurs along the edge of the crater from which they are extruded.

Fig. 8. Formation, extension and removal of a single platelet in 7075-T6 alloy

The plastic deformation that occurs at the surface of the eroding metal is severe enough to have caused some adiabatic shear deformation heating to have occurred. Nowhere on any surface observed was melting seen, but the extent and relative evenness of the severity of the deformation indicates that warm working or even hot working was the reason that such extensive smearing type deformation could occur. Some of the impacting SiC particles were retrieved and examined for evidence of localized heating where they contacted the eroding aluminum. Figure 9 shows the contacting portion of a typical SiC particle. A thin film of melted and resolidified aluminum can be seen on one surface of the particle. The relatively low thermal conductivity of SiC compared to that of aluminum can concentrate the combined frictional and deformation heating that occurs at the aluminum surface upon impact at the tip of the SiC particle. Aluminum is smeared onto its surface and melted. The corresponding heated aluminum surface of the target conducts the heat away enough to prevent melting, but does experience enough heating to result in warm or hot forming of the craters and platelets. Aluminum deposits were found on many of the retrieved SiC particles.

To establish further the nature of the deformation that develops craters and platelets on the surface of the aluminum, a series of experiments was performed where a thin layer of gold, 150 Å thick, was sputtered on to both fresh and steady state eroded surfaces and the coated surfaces subsequently impacted by a single SiC particle. Figure 10 shows the crater formed on a gold plated fresh surface by a particle impacting from the right hand side at $\alpha = 30^\circ$. The SEM x-ray map of gold in the lower picture shows that a significant amount of gold remained on the crater's surface, even in the areas where scribe lines appear to indicate the formation of new surface by a micro-machining mechanism. The presence of the gold on the bottom surface of the crater indicates that the crater was formed by an action that forged the original surface with the gold on it downward to form a crater. Material was extruded from immediately below the initial surface up and out of the crater to form the ridge of material along the left hand edge of the crater. The absence of gold in an extended region to the left of the crater indicates that elastic-plastic deformation from the particle impact extended out beyond the immediate crater site and that the high speed of the deformation deformed the gold in a manner that caused it to be bent up so that the particle leaving the surface could snag it and tear off the piece that is missing.

Fig. 9. Melted aluminum remains on a SiC eroding particle

Figure 11 shows a crater that was formed on a gold coated steady state erosion surface impacted by a single, additional particle. The bottom of the crater has significant amounts of gold on its surface. The lower left, scribed side of the crater shows less gold on it because of the shadowing effect of the crater depth interfering with the gold x-rays reaching the counter. The counter is located in the SEM beyond the lower left hand side of the picture. In no instance observed was the bottom of a crater free of gold, as would occur in the creation of new surface by a micro-machining mechanism.

650X

GOLD MAP

XBB790-13305

Fig. 10. Gold coated crater formed on a fresh aluminum surface with gold x-ray map

All of the photographs used in this report are typical of the mechanism occurring on the surface of the aluminum alloys. No other mechanism than that presented was seen. The sequences were selected to clearly show what was occurring. They were randomly selected as any area observed had to be selected prior to the sequence shown. Some sequences selected did not show the mechanism as clearly as others did and they were not used

in this paper. However, they all showed the formation of craters and platelets as depicted in the sequences selected for inclusion in the paper.

1000 X

GOLD MAP

XBB790-13303

Fig. 11. Gold coated crater formed on a steady state aluminum surface with gold x-ray map

Proposed Mechanism

Combining the measured erosion behavior data presented in Fig. 1 and Table 1 with the documentation by SEM photographs of the actions that occurred on the surfaces of the 1100-0 and 7075-T6 aluminum alloys, the following mechanism of erosion of ductile metals is proposed:

Particles initially impact a fresh surface creating craters by a combined forging-extrusion action that displaces material out from directly under the contact area between the particle and the surface. The extruded material forms platelets in the vicinity of the crater from which it was extruded. This activity occurs during the threshold period in Fig. 1. The force of the particle impacts also plastically deforms the metal beneath the craters over a much wider region than that defined by the crater itself.

As additional particles impact, the density of craters and platelets on the metal's surface increases. The severe deformation that is occurring causes adiabatic shear deformation heating of the surface into the warm or even hot working regime of the metal to a depth up to that of the deeper craters, approximately 20 μm in the aluminum alloys tested. The excellent thermal conductivity of the aluminum removes heat from the deformed surface at a rate that prevents actual melting from occurring. The heating of the surface increases the deformation capability of the surface region and thin, highly deformed platelets are generated.

The force of the impacting particles is sufficient to also plastically deform a region of the target material considerably below the immediate surface region. This plastic deformation occurs at lower temperatures where cold working of the metal occurs. A cold worked region is built up beneath the highly deformed surface region as the erosion process continues.

A direct measurement of the soft surface zone and cold worked zone beneath it is shown in Fig. 12. The results of the micro-hardness survey of the cross section of an eroded 1100-0 aluminum specimen is shown. A 5 gm load was used and the diamond pyramid hardness (DPH) determined from a location just below the surface where sufficient material existed to keep the penetrator from breaking down the specimen edge and giving a false reading. Figure 12 shows the results of the hardness survey taken on two specimens, one eroded at an impingement angle of 90° and the other at 30° . The softer area at the surface underlain by a considerably harder zone can be seen.

XBL 8011-2236

Fig. 12. Micro-hardness surveys of eroded material cross sections

As more particles continue to strike the surface, the density of the craters and platelets increases. Some platelets are restructed by succeeding particles and extended to the severely distressed condition where additional particle impacts begin to knock them off the surface and actual erosion begins. While this is occurring the cold worked region beneath the surface is increasing in size and hardness until it reaches a steady state.

When the cold worked sub-surface region reaches steady state, the efficiency of the particles impacting on the surface to extrude platelets is increased. This occurs because the extrusion force in the heated surface region can now be exerted between the hard impacting particles and the hard, sub-surface cold worked region. The increased platelet forming efficiency of the particles brings the erosion surface to a steady state density of craters and platelets. This, in turn, produces a relatively constant number of sufficiently distressed platelets that can be knocked off the surface. This can be seen by the constant amount of erosion loss that occurs with increasing numbers of impacting particles in Fig. 1.

The cold working effect is demonstrated by the data in Table 1. With increasing time of erosion, the cold worked region can be formed by particle impact deformation. However, if the metal is cold rolled prior to erosion, the sub-surface cold worked region is partially formed by the rolling process. Thus, with increasing reductions by rolling, the amount of erosion increases toward that of the steady state erosion level. It can be seen in Table 1 that by the time 80% reduction by cold rolling has been achieved the amount of erosion by the initial 60 gm of particles has doubled over that of the annealed material and is approaching that of the steady state erosion level. It is important to conduct this experiment with a number of particles that are too few to achieve steady state erosion by themselves.

At steady state erosion, four distinct actions are occurring. The impacting particles are forming new craters, extruding new platelets, extending out already existing platelets and knocking severely distressed platelets from the surface. The material behavior characteristics that affect this mechanism include many properties such as forgeability, extrudeability, thermal conductivity, and others that have not previously been considered in the selection and design of erosion resistant ductile metals.

III. EROSION OF THIN OXIDE SCALES

The erosion of brittle materials has been determined to occur in a different manner from the erosion of ductile materials. In brittle materials, the erosion mechanism is a chipping away of material from the surface. It occurs more at high impingement angles than at low impingement angles. The combination of a thin brittle adherent oxide scale (2-5 μm) on a ductile base metal has not been investigated to determine the erosion mechanism that occurs or the rate at which material is removed. In service the brittle surface oxide is growing by a corrosion mechanism that is fairly well defined at the same time that it is being eroded away by impacting particles. In order to gain an understanding of the combined corrosion-erosion mechanism, a first step has been taken wherein pre-corroded surfaces of 310SS and an experimental FeCrAlHf alloy have been eroded at room temperature.

Experiment Description

The erosion test conditions used were similar to those used in the ductile metal investigation described above. 50 μm

size SiC particles were used in an air blast tester operating at velocities of 200 fps. Impingement angles were 30° and 90° . Small quantities of particles, 0.3 g and 1.2 g were directed at the surface of the alloys that had been previously corroded at 900° and 1000°C (1650° and 1800°F) in gases ranging from air to low P_{O_2} oxygen mixtures to H_2 , H_2S , H_2O combined oxidizing sulfidizing gas mixtures for exposure times of 100 hrs. The scale formed on the 310SS was primarily a thin, adherent Cr_2O_3 layer with a few specimens showing the presence of iron sulfide compounds. The scale formed on the FeCrAlHf alloy was a pegged Al_2O_3 layer.

Results

Figure 13 shows the type of Cr_2O_3 scale that was formed on the 310SS in an air exposure at 1800°F and the surface of the steel after the scale had been eroded off. The test was designed to erode away the scale but not to initiate erosion of the base metal; 1.2 gm of erodent was required. The threshold quantity for erosion of the base metal is in excess of 5 gm of erodent.

Figure 14 plots the amount of material eroded at a 90° impingement angle for each 0.1 gm of erodent. It can be seen that the conditions under which the 310SS was oxidized have a major effect on the amount of erosion that occurs at any time during the erosion cycle. The thicker scale that formed in the air oxidation naturally lost more weight than did the thinner scale which forms at the low P_{O_2} . The early erosion at 0.1 to 0.4 gm of impacting particles is expected to be representative of that which occurs in a combined erosion-corrosion environment where the scale will not grow very thick. The low P_{O_2} formed scale was much more resistant to erosion than the scale formed in air. The differences in the morphology, however, were not readily discernible under the microscope.

The erosion of the scale formed on the FeCrAlHf alloy was less than that formed on the 310SS. Differences affecting the adherence of the scale may have contributed to the greater erosion resistance of the Al_2O_3 scale. Al_2O_3 itself may also be more erosion resistant than Cr_2O_3 . Figure 15 shows the cross sections of the two scales. The nature of the scale on the 310 was the same at 1800°F as at the 1650°F shown, only thicker. The black internal oxides on the 310SS sample are SiO_2 . Their presence in the immediate vicinity of the scale-metal interface degraded the adherence of the scale, resulting in increased erosion. The cross section of the FeCrAlHf alloy, which did not contain any silicon, does not have degrading

Fig. 13. Cr_2O_3 scale on 310SS before and after erosion

internal oxides. Rather it exhibits the enhanced adherence microstructure developed by the presence of the hafnium which causes pegs of Al_2O_3 to extend into the alloy surface, binding the scale to the metal more tightly.

XBL 793-803

Fig. 14. Material loss vs amount of erodent used for several scales on 310SS and on experimental FeCrAlHf alloy

Table 2 presents data on several variables used in the erosion tests of low P_{O_2} and simultaneous sulfidation-oxidation formed scales on 310SS.² The effect of most of the variables on the erosion behavior was minor. Formation of the scale at 1650° and 1800°F had no effect on erosion. Because of the brittle nature of the scale, the erosion at an impingement angle of 90° was greater than that at 30°. There was no effect on erosion of variations in the P_{O_2} of the corroding gas between 15 and 17.4. The preparation of the surface prior to corrosion by burnishing to enhance adherence had no measureable affect on the erosion. While no discernible sulfides or spinels were found in the oxide scale, simultaneous sulfidation-oxidation did approximately double the amount of erosion that occurred. However, the combined gas scale lost its impingement angle sensitivity with both the 90° and 30° tests eroding approximately the same amount of material.

10μ

Type 310 S.S. Oxidized: $\text{Log } P_{O_2} = -15$, 1650°F

5μ

Fe 18Cr 5Al 1Hf Oxidized: $\text{Log } P_{O_2} = -15$, 1800°F

XBB797-8943

Fig. 15. Cross sections of scales formed on 310SS and FeCrAlHf alloy

Table 2.

Surface	°F	Log P _{O₂}	Log P _{S₂}	Scale thick- ness (m)	90° weight loss after		30° weight loss after	
					0.3 g SiC	1.2 g SiC	0.3 g SiC	1.2 g SiC
<u>Low P_{O₂} Oxidation of 310 S.S.</u>								
As	1650	-15	---	4	0.4	0.7	0.1	0.3
received	1650	-15	---		0.3	0.6		
	1650	-16.5	---		0.4	0.7		
	1800	-16.5	---		0.4	0.7		
<u>Simultaneous Sulfidation-Oxidation of 310 S.S.</u>								
As	1800	-15	-6	5	0.8	1.3	0.8	1.2
received	1800	-15.2	-8.5		0.7	1.35		
	1800	-16.9	-6		1.1	1.8		
	1800	-17.4	-7		1.0	1.6		
Burnished	1800	-15	-6	2.5	0.9	1.7	0.7	1.2
	1800	-15.2	-8.5		0.7	1.2		
	1800	-16.9	-6		0.9	1.6		
	1800	-17.4	-7		0.9	1.6		

The differences in the morphology of the internal SiO_2 formed in the 310SS is shown in Fig. 16. The increased quantity and orientation parallel to the oxide-metal interface is

10μ

Type 310 S.S. Oxidized: $\text{Log } P_{\text{O}_2} = -15$, 1650°F

10μ

Type 310 S.S.

Oxidized - Sulfidized: $\text{Log } P_{\text{O}_2} = -15$, $\text{Log } P_{\text{S}_2} = -6$, 1800°F

XBB 797-8951

Fig. 16. Differences in morphology of internally oxidized SiO_2

clearly seen. The more the SiO_2 paralleled the oxide-metal interface, the greater the erosion became. The amount of erosion doubled as shown in Table 2 between the specimen shown in the upper picture and that shown in the lower picture. This effect alone could account for the difference in erosion between the low P_{O_2} oxidized and the sulfidized-oxidized specimens. However, other more subtle effects may also account for part of the difference.

The FeCrAlHf alloy also exhibited more erosion in the oxidized-sulfidized state. Figure 17 shows the curves for the low P_{O_2} oxidized and sulfidized-oxidized specimens. The total amount of erosion on this alloy is low because the thickness of the Al_2O_3 scale is low. However, a distinct difference in erosion was measured. Figure 18 shows the cross section of the two scales. The development of the pegs in the sulfidized-oxidized specimen was not as great as in the straight oxidized specimen. The low initial amount of erosion in both specimens indicates that the Al_2O_3 is basically more erosion resistant than the Cr_2O_3 and that pegging has a lesser effect on the erosion behavior than might be expected.

XBL 801-95

Fig. 17. Erosion of FeCrAlHf alloy scales formed in low P_{O_2} and sulfidizing-oxidizing gases

5μ

Fe 18Cr 5Al 1Hf Oxidized: $\text{Log } P_{\text{O}_2} = -15$, 1800°F

5μ

Fe 18Cr 5Al Hf

Oxidized - Sulfidized: $\text{Log } P_{\text{O}_2} = -15$, $\text{Log } P_{\text{S}_2} = -8.5$, 1800°F

XBB797-8946

Fig. 18. Al_2O_3 scale cross-sections formed in low P_{O_2} and sulfidizing-oxidizing gases

Conclusions

Work on determining the erosion behavior of thin oxide scales on metals is continuing. The work to date indicates that significant differences in erosion behavior do occur as the result of different elevated temperature corrosion conditions forming different scale and interface compositions and morphologies. Generally the differences were relatively minor. However, under conditions of combined erosion-corrosion, these differences could become important. Even though the scales were thin and on ductile matrices, their erosion behavior variation with impingement angle was that of a brittle rather than a ductile material. The use of erosion tests to study the adherence of various scales has some merit. However, interpretation of some of the smaller differences could be difficult.

REFERENCES

- (1) S. Bhattacharyya, D. Yates, and V. Hill, "Erosion-Corrosion of Materials in Coal Gasifier Environments," Proceedings of 5th International Conference on Erosion by Liquid and Solid Impact, Cambridge, England, 3-6 September, 1979.
- (2) I. G. Wright, "Solid Particle Erosion of Oxidation-Resistant Alloys in High Temperature, Low Velocity Gas Streams," Proceedings of 5th International Conference on Erosion by Liquid and Solid Impact, Cambridge, England, 3-6 September, 1979.
- (3) H. Leavenworth, J. Kelley, and J. Hansen, "Development of Wear Resistant Valve Materials," Contract Ex-76-A-2219, Albany Metallurgy Research Center, Bureau of Mines, Albany, Oregon.
- (4) L. K. Ives and A. W. Ruff, "Transmission and Scanning Electron Microscopy Studies of Deformation at Erosion Impact Sites," Wear 46, 1978, pp. 149-162.
- (5) N. Gane and M. J. Murray, "The Transition from Ploughing to Cutting in Erosive Wear," Proceedings 5th International Conference on Erosion by Solid and Liquid Impact, 3-6 September, 1979, Cambridge, England.

- (6) G. A. Sargent, M. K. Keshavan, and H. Conrad, "The Effect of Microstructure on the Erosion by Solid Particles," Metallurgical Engineering and Materials Science Department, University of Kentucky, Lexington, Kentucky 40506, 1979.
- (7) I. Finnie, "Erosion of Surfaces by Solid Particles," Wear 3, 1960, pp. 87-103.
- (8) I. Finnie and D. H. McFadden, "On the Velocity Dependence of the Erosion of Ductile Metals by Solid Particles at Low Angles of Incidence," Wear 48-1, May 1978, pp. 181-190.
- (9) J. G. A. Bitter, "A Study of Erosion Phenomena," Parts 1 and 2, Wear 5, 1963.
- (10) J. H. Neilson and A. Gilchrist, "Erosion by a Stream of Solid Particles," Wear 11, 1968, p. 111.
- (11) G. L. Sheldon and A. Kanhere, "An Investigation of Impingement Erosion Using Single Particles," Wear 21, 1972, pp. 195-209.
- (12) T. Christman and P. G. Shewmon, "Erosion of a Strong Aluminum Alloy," Wear 52, 1979, pp. 57-70.
- (13) R. E. Winter and I. M. Hutchings, "Solid Particle Erosion Studies Using Single Angular Particles," Wear 29, 1974, pp. 181-190.
- (14) A. V. Levy, "The Role of Plasticity in Erosion," Proceedings 5th International Conference on Erosion by Solids and Liquid Impact, 3-6 September, 1979, Cambridge, England.
- (15) R. A. Mayville, "The Mechanisms of Material Removal in the Solid Particle Erosion of Ductile Metals." M.S. Thesis, Lawrence Berkeley Laboratory report, LBL-7333, Berkeley, California, 1978.

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